Advances in Materials Science and Engineering

Cumulation of Failure and Crack Growth in Materials

Lead Guest Editor: Dariusz Rozumek
Guest Editors: Alexander Balitskii, Grzegorz Lesiuk, and Stanislav Seitl

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Editorial

Cumulation of Failure and Crack Growth in Materials

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In order to ensure the safety and reliability of materials and structures, for example, in power and energy systems, structural integrity and lifetime prediction are hot topics of research in these systems. For many countries, currently facing a potential future mismatch between energy production and transformation, increasing interest is being paid to new techniques to both discover and understand the failure mechanisms, as well as provide lifetime prediction for engineering materials and structures.

Working structures and their elements are subjected to the influence of various loads. These can be static, cyclic, or dynamic loads. In order to ensure an adequate level of safety and optimal durability of structural elements, experimental tests and simulations are required to determine the effect of various factors. Such factors include the effects of notches, voids, and environment.

Studies and research outcomes carried out in this field are necessary to guide the development of new and advanced standards for a better selection of materials that meet the requirements of the designers.

This issue would aim to provide the data, models, and tools necessary to perform structural integrity and lifetime prediction based on the multiaxial stress state and finally mixed-mode fatigue crack growth resulting in the use of advanced mathematical, numerical, and experimental techniques.

This special issue aims to gather the most recent research achievements on the initiation and growth of cracks in structural components made of different metals and composites. The analysis of mixed modes is of particular interest, but studies related to the effect of simple load (pure mode conditions) are also well fitted with the aim and scope of the special issue.

Over the last year, we have received and analyzed 28 submitted research works by outstanding international scientists. The papers submitted by the authors have been subjected to the rigorous journal peer-review process. During the intensive process of reviews and in-depth discussions, 17 articles were accepted for the publication, while other documents were rejected (or were withdrawn by the authors) mainly due to the discrepancy to the topic of the special issue. Accepted papers, after passing the positive review process as well as the editorial processes, will be published in this special issue of *Advances in Materials Science and Engineering* entirely devoted to the cumulation of failure and crack growth in materials.

The topics considered in the accepted papers include several aspects influencing the fatigue and fracture of structural materials such as the effects of factors affecting the fatigue life and crack initiation in welded joints, the influence of geometric effects such as notches and fillets, whose investigation has been performed by adopting theoretical, numerical, and experimental approaches. Among the areas emphasized in the special issue are case histories, sample computations of practical design problems, material characterization procedures, failure mechanisms assessment of critical components, energy approach, initiation and propagation mechanisms, effect of microstructure and defects on fatigue behavior, as well as prediction of durability and remaining useful life estimation.

Conflicts of Interest

The editors declare that they have no conflicts of interest regarding the publication of this Special Issue.

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Dariusz Rozumek
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Research Article

Inductive Experimental Study of Corrosion Products of Medium Carbon Steel CK45 Hardened by Magnetized Water

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This paper concerns the inductive experimental study of corrosion products of the medium carbon steel CK45DIN using magnetized water as hardening media, where the water has been exposed to a magnetic field of 1000 Gauss and 2000 Gauss, respectively, for time durations of 1 to 5 hours, with a volumetric flow rate of 4 gal/min. Medium carbon steel CK45DIN samples were exposed to air as corrosive environment for reasonable time interval, and the corrosion product was weighed in order to determine the loss of weight. The results showed that the corrosion rate of all samples decreased as the experiment progresses over time. Corrosion resistance of the medium carbon steel CK45DIN increased when we raised its temperature to 870°C for 40 minutes and hardened by magnetized water applied in a magnetic field with a strength of 1000 Gauss for 3 hours as compared with the sample hardened by normal water.

1. Introduction

Corrosion is a natural process, in which an exposed surface of a refined metal is deteriorated to a more chemically stable form when it comes into contact with a gas or liquid, as a result of chemical reactions typically in the form of oxidation between the metal and its surrounding environment. Corrosion is dangerous, and in many cases it represents an enormous economic loss. Because of corrosion, buildings and bridges can fall, oil pipelines can fracture, and sewerages can flood. Corroded electrical contacts can produce blazes and other accidents; corrosion of medical implants in the human body can cause infection.

As a low-cost and attainable metal, the CK45 steel has applications in industries, which are abundant but still restricted because of its poor wear and corrosion resistance [1]. Hence, a large number of studies have been conducted in order to enhance the corrosion resistance. Heat treatment is a process of heating and cooling metals to obtain good physical and mechanical properties by modifying the crystalline structure. The heating temperature, length of time, and cooling rate are some of the important variables that can affect the properties of the heat-treated metals [2, 3].

The effect of heat treatment on mechanical properties under different cooling media (water and palm oil) was studied, which showed the best properties in strength and hardness would be in water compared to palm oil [4].

The aim of this paper is to conduct an experimental study on corrosion products of the CK45 carbon steel using magnetized water as hardening media and compare the results of that hardened by normal water. In [5], it is explained that the cavitational wear resistance of electroplated nickel composite layers was inspected by ASTM G32. Particles of diverse hardness (titania and silicon carbide) and diverse sizes from microscale to nanoscale were merged up to 30 vol.% into a nickel matrix. Martens hardness is developed by grain refinement via a particle merger. Magnetized water is simply known as the water that passes through a magnetic field or the water inside it or near a magnet for a certain period of time. The exposure of water to the magnetic field causes changing in many physical properties of water because water is a polarized liquid [6, 7].

In [8], this is compared between the effects of N-alloying through diverse processing and manufacturing routes on mechanical, metallurgical, wear, and corrosion properties of high N-containing stainless steels. Furthermore, the

investigation of relationship between the stress intensity factor and corrosion crack growth rate depended on static corrosion [9]. Damage of metals due to the influence of hydrogen and stress corrosion cracking is quite frequent and focuses on risky failures as well as on decreasing property and large compensational payments by insurance companies [10].

The properties of the water solution change when exposed to a magnetic field according to the following variables:

(1) The flow rate of water through the magnetic field.

(2) The presence of water in the magnetic field.

(3) magnetic field strength.

2. Experimental Procedures

(1) Eleven cylindrical samples of the medium carbon steel CK45DIN with a diameter 20 mm and a height of 30 mm were used in this experiment. The chemical composition of the samples was determined by using a chemical analytical instrument (BrukerS-1 TURBO SD) as shown in Table 1.

(2) Dipole-magnetized water was prepared using the device shown in Figure 1 by applying a magnetic field with a strength of 1000 Gauss on distilled water at a flow rate of 4 gallons/min and for periods of time extended from 1 hour to 5 hours. Then the magnetic field strength increased to become 2000 Gauss at the same periods of time and flow rate. The number of samples hardened by magnetized water becomes ten samples, in additon to the sample hardened by normal water. Thus, the number of samples becomes 11.

(3) The eleven steel samples were placed in the oven, and their temperature was raised to a hardening temperature of 870°C for 40 minutes.

(4) The samples were numbered from (1) to (11), so that the sample hardened by normal water took number (11) while the samples hardened by magnetized water were numbered from (1) to (10). The sample hardened by magnetized water applied in a magnetic field with a strength of 1000 Gauss for 1 hour took number (1), the sample hardened by magnetized water applied in a magnetic field with a strength of 1000 Gauss for 2 hour took number (2), and thus the rest of the samples took till number (5). The sample hardened by magnetized water applied in a magnetic field with a strength of 2000 Gauss for 1 hour took number (6), the sample hardened by magnetized water applied in a magnetic field with a strength of 2000 Gauss for 2 hour took number (7), and thus the rest of the samples took till number (10).

(5) After the completion of the hardening process, the samples were prepared as follows:

(i) Grinding of the samples was done with smooth satin paper with softness from 180 to 1200 Grit number.

(ii) The samples were polished using aluminum oxide solution (Al_2O_3) in water by 15% and by using a universal rotary polishing machine. After that, we used a 1-micron fine linen sheet.

Table 1: Percentage of sample components under study.

Symbol	Fe	Mn	Si	C	Cr	Sn	Cu	P
Percentage	98.02	0.72	0.63	0.53	0.04	0.03	0.02	0.01

Figure 1: Water magnetization device.

3. Results and Discussion

3.1. Weight Loss. All medium carbon steel CK45DIN samples were exposed to air as a corrosive environment for reasonable time interval and completely polished with polish paper to remove all corrosion products and then the pieces of each sample were weighed to determine the loss of weight. The weight losses in a gram of sample hardened by normal water and samples hardened by magnetized water applied in a magnetic field for time intervals (1, 2, 3, 4, and 5 hours) with a strength of 1000 Gauss and 2000 Gauss, for a certain exposure time, are presented in Figures 2 and 3.

From Figure 2, it can be observed that the sample applied in a magnetic field with a strength of 1000 Gauss for 3 hours has the highest corrosion resistance since the lowest weight was lost during the exposure period. This is followed successively by the samples applied in a magnetic field for 3 hours, the samples applied in a magnetic for 2 hours, the sample hardened by normal water, the sample applied in a magnetic field for 4 hours, and lastly the sample applied in a magnetic field for 5 hours.

From Figure 3, it can be observed that the sample applied in a magnetic field with a strength of 2000 Gauss for 5 hours has the highest corrosion resistance. This is followed successively by the sample hardened by normal water and lastly the sample applied in a magnetic field for 3 hours has the lowest corrosion resistance.

The general observation on the results shown in Figures 2 and 3 is an evident increase of weight loss with exposure time and a similar progression pathway of cumulative weight losses for all the samples with increase in exposure time. The reason for the constant difference in weight loss from the start until the end could be due to the difference in composition and structure generated by heat treatment.

3.2. Corrosion Rate. The corrosion rate was calculated from decrease in weight observed in samples in weight loss tests using the following formula [11]:

Figure 2: Weight losses versus exposure time for the samples hardened by magnetized water applied in magnetic field with a strength of 1000 Gauss.

Figure 3: Weight losses versus exposure time for the samples hardened by magnetized water applied in magnetic field with a strength of 2000 Gauss.

$$\text{corrosion rate (mpy)} = \frac{534\,W}{D * A * T}, \tag{1}$$

where mpy is mils penetration per year (1 mil·10^{-3} inch.); W is the weight loss (mg); D is the density of the sample (g/cm^3); A is the area of the sample (inch^2); and T is the exposure time (hr)

The corrosion rates are calculated for the sample hardened by normal water, samples hardened by magnetized water applied in a magnetic field for time intervals (1, 2, 3, 4, and 5 hours) with a strength of 1000 Gauss and a strength of 2000 Gauss, for a certain exposure time, and the data obtained are plotted in Figures 4 and 5.

Figure 4: Variation of corrosion rates with exposure time for the samples hardened by magnetized water applied in magnetic field with a strength of 1000 Gauss.

Figure 5: Variation of corrosion rates with exposure time for the samples hardened by magnetized water applied in magnetic field with a strength of 2000 Gauss.

As in Figures 4 and 5, the corrosion rate of all samples decreased as the experiment progresses over time. The corrosion rate was very low with the sample hardened by magnetized water applied in a magnetic field with a strength of 1000 Gauss for 3 hours as compared with the sample hardened by the normal water since it has the highest corrosion resistance. However, the corrosion rate was very high with the sample hardened by magnetized water applied in a magnetic field with a strength of 2000 Gauss for 3 hours as compared with the sample hardened by the normal water since the corrosion attack increased.

4. Conclusion

From the results of the investigation, the following inferences can be deduced:

(1) The corrosion rate of all samples decreased as the experiment progressed over time.

(2) Corrosion resistance of the medium carbon steel CK45DIN increased when we raised its temperature to 870°C for 40 minutes and hardened by magnetized water applied in a magnetic field with a strength of 1000 Gauss for 3 hours as compared with the steel hardened by normal water.

(3) The corrosion rate decreased by 4% if magnetized water was applied in magnetic field with a strength of 2000 Gauss.

Data Availability

No data were used to support this study.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Experimental Study on Mechanical Properties of Cemented Paste Backfill under Temperature-Chemical Coupling Conditions

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To investigate the effect of temperature-chemical coupling on the mechanical properties of cemented paste backfill, three temperatures (20°C, 35°C, and 50°C) and sodium sulfate solution mass concentrations (3%, 5%, and 7%) are applied to simulate the complex environment in a mine. Uniaxial compressive strength and the CPB stress-strain relationship are investigated by applying stress, and the deterioration mechanism was analyzed theoretically according to physical and chemical reactions. At the same time, a structural model of the CPB deterioration mechanism under TC coupling is constructed. Combined with analysis through X-ray diffraction and scanning electron microscopy, it is shown that ettringite and gypsum are the main erosive substances that destroy the structure of CPB and that increased temperatures accelerate the chemical reaction. The concentration change consumes calcium hydroxide, changing the relationship between ettringite and gypsum. Sodium sulphate crystallization is the main form of physical deterioration. The continuous load accelerates the inelastic deformation time of CPB, resulting in a large yield deformation process.

1. Introduction

With the background of vigorously promoting ecologically friendly construction in China, coal mine filling and mining technology has been promoted as a green mining technology in coal mining and goaf management [1, 2].However, the high temperature caused by oxidative heat release, blasting operations, and deep-well mining for underground coal [3–5] along with the stress of overlying strata and deterioration by chemical substances such as acids, alkalis, and salts contained in groundwater cause damage to the structure of cemented paste backfill (CPB). The effect is to reduce or eliminate CPB effectiveness, which in turn affects the stability of the goaf. Therefore, the experimental study on the mechanical properties of CPB under temperature-chemical (TC) coupling has important theoretical significance and practical engineering value for predicting stability in filling engineering.

The effect of filling and mining is closely related to the structural characteristics of CPB. The structure type of CPB filled into goaves is of the plain concrete type; there are already microcracks in the internal structure of CPB when it enters a goaf before reaching the coagulation and finalization stage. Under complex geological conditions and environmental factors, CPB should bear the effects of load and geothermal heat, as well as chemical action of the external environment. Owing to the triple damage posed by the TC combination, microdamage to CPB gradually evolves and expands, finally forming a penetrating crack that leads to structural failure. This results in a sharp drop in the bearing capacity and durability of CPB. Fall et al. and Aldhafeeri et al. [6, 7] used an experimental approach to study the influence of curing temperature and combined effects of temperature and CPB components on the main mechanical properties of CPB (strength, modulus of elasticity, and stress-strain behavior); results showed that the reaction activity of CPB is related to temperature and that reaction activity increases with atmospheric temperature. However, the effect of temperature depends on the curing time. The effect of temperature on mechanical properties of CPB depends on the type of binder, the ratio of water to cement, and the type of tailings. Han and Liu [8] studied the effect of temperature on the microstructure of a hardened paste of cement-slag composite cementitious material and the effect of the mortar's subsequent strength. The results showed that temperature has little effect on the degree of hydration of pure cement but that high temperature (60°C) reduced the post-compressive strength of the pure cement paste.

In addition, most mine water in China is rich in sulfates, and the oxidation of sulfur minerals and the use of retarders added into cement slurries raise the sulfate content of CPB. Sulfate causes severe deterioration of CPB through physical and chemical action [9–11]. Feng et al. [12] showed through experiments that sulfate reacted to form an expansion product, ettringite, and damage induced by expansion modifies both the local effective transport properties and linear elastic properties of each microstructure at different depths, thereby potentially altering the rates of sulfate ingress and expansion. On the other hand, physical crystallization of sulfate is also one of the main forms of sulfate attack. Chen et al. and Zhang et al. [13, 14] studied microstructure changes of cement paste using semiconductors, XRD, MIP, and X-CT. The results showed that the crystalline sodium sulfate formed by sulfate intrusion creates a cement slurry. The internal structure of the body produces crystal stress, which causes its volume to increase, causing severe deterioration. Xu et al. [15] concluded that external sulfate attack is a progressive degradation process that may cause expansion, cracking, loss of binder cohesion, and increased permeability in cementitious materials. Crystallization pressure theory has often been referred to as the most likely mechanism. Chen et al. [16] concluded that the main factors influencing concrete compressive strength are exposure time, water-cement ratio, and sulfate ions. Among these, sulfate attack had the greatest effect. Abedalrazzaq [17] conducted an empirical study on the effects of crack growth in cement mortar exposed to two sulfate solutions. The testing technique was conducted to measure compressive strength, and ultrasonic pulse velocity (UPV) tests simultaneously determined the stress/strength ratio at which cracks propagated. The results showed that crack propagation takes place in specimens exposed to sulfate solutions at stress/strength ratios higher than those kept in tap water. Yu et al. [18] immersed the cement slurry in different concentrations of sulphate solutions; the test showed that sulphate deterioration was staged. In the initial stage of deterioration, the presence of sulfate had a positive effect on the cement slurry strength, and subsequently these properties changed.

According to previous research, many factors such as sulfate presence, geothermal factors, and rock pressure in a mine water context have significant effects on the properties of CPB. Most studies only considered the mechanical properties of CPB with respect to a single factor such as temperature or ion concentration. However, CPB is used in a complex environment of different temperatures, different ion concentrations, and various pressures for each rock layer at the same time. The mechanical properties of CPB are different at different ages [19]; at present, there is a lack of research on the mechanical properties of CPB under coupled conditions of temperature, pressure, and ion concentration. Long-term filling of CPB in this complex environment causes damage to the internal structure of CPB and affects the stability of goaves in which it is used. In order to explore the deterioration mechanism of CPB under these coupled conditions and improve its durability, it is necessary to study the influence of TC coupling on the mechanical properties of CPB. Therefore, based on existing research, we select concentrations of 3%, 5%, and 7% sodium sulfate solution to simulate the chemical environment of CPB [20] and apply three temperatures of 20°C, 35°C, and 50°C to simulate the temperature of underground CPB [21]. The uniaxial compressive strength (UCS) of CPB is tested at 1, 3, 7, and 28 days, respectively. The variation of mechanical properties of CPB under temperature-stress-chemical coupling conditions is investigated, and the deterioration mechanism of CPB under TC coupling conditions is constructed. The structural model is examined through X-ray diffraction (XRD) and scanning electron microscopy (SEM), two microtest methods, to explain and verify the macroscopic phenomena, providing a theoretical basis for studying the durability of CPB in mines.

2. Experimental Materials and Methods

2.1. Experimental Materials

2.1.1. Cement. Cement (no. 32.5) is sourced from Shandong Shanshui Cement Group Co., Ltd., and its main components are shown in Table 1.

2.1.2. Fly Ash. The fly ash is from Huangdao Power Plant, Shandong Province, China. The apparent density is 2040 kg/m^3, and the bulk density is 750 kg/m^3. The main chemical composition is shown in Table 2. The particle size distribution and SEM test are shown in Figure 1.

2.2.3. Coal Gangue. The coal gangue is from the Xinzhuang Coal Mine in Shandong Province, China. The main chemical composition is shown in Table 3. The particle size ratio is shown in Table 4.

2.2. Preparation of CPB. The composition of the CPB was determined based on the principle of reducing the amount of cement needed and utilizing as much solid waste (fly ash and coal gangue) as possible. With these goals in mind, the ratio of the cement, fly ash, and coal gangue was selected to be 1 : 4 : 6. Subsequently, water was added to the CPB until the solid content in the final mixture satisfied the requirements for use in mines. In general, a solid concentration of 70–80% is suitable for gravity- or pumping-driven transport through the pipelines [22]. Cube specimens 70.9 mm × 70.9 mm × 70.9 mm were produced with a mass concentration of 72%. The measured slump was 20 cm, the initial setting time of the specimens was 4 h, and the final setting time was 8 h.

2.3. Experimental Methods. For more realistic simulation of the mine environment, the prepared test pieces were placed in sodium sulfate solutions with mass concentrations of 3%, 5%, and 7% in advance, and they were then placed in a constant-temperature curing box at temperatures of 20°C, 35°C, and 50°C, respectively. The curing time was 1, 3, 7, or 28 days. Another clear-water fluid was used to create a control group. Hence, a total of 48 treatment conditions were studied with three test pieces for each condition for a total of 144 test pieces. After the corresponding age was reached, the filling test piece was taken out of the incubator

Table 1: The chemical component and mineral composition of ordinary silicate cement.

Category	Chemical components (%)					Mineral composition (%)			
	SiO_2	Al_2O_3	Fe_2O_3	CaO	SO_3	C_3S	C_2S	C_3A	C_4AF
Content	21.38	4.23	3.58	66.49	0.1	59.95	12.02	5.94	13.53

Table 2: Chemical composition of fly ash (%).

Source	SiO_2	Al_2O_3	Fe_2O_3	CaO	MgO
Content	53.94	30.91	2.38	6.53	0.92

Table 3: The main chemical composition of coal gangue (%).

Composition	SiO_2	Al_2O_3	Fe_2O_3	CaO	MgO	Na_2O	K_2O
Content (%)	59.1	18.9	4.3	2.36	1.41	0.43	1.89

(a)

(b)

Figure 1: Particle size distribution of fly ash (a) and SEM image (b).

and immediately loaded in a DY-2008DX automatic pressure-testing machine (load capacity: 50 kN and loading rate: 0.01 mm/s), and only the average value was considered. Owing to the short loading time, the scenario can be approximated as a TC coupled-state loading.

2.3.1. UCS Measurements. The UCS of the CPB was measured using an automatic pressure testing machine (DY-2008DX,223 Xianxian Jingqiang Construction Equipment Factory, China; loading capacity = 50 kN and loading rate = 0.01 mm/s). For each curing time, three samples were taken from the curing chamber, and their respective compression forces were measured using the pressure testing machine. The UCS values of the 226 CPB samples were then calculated using

Table 4: Coal gangue particle size grading.

Size (mm)	+10	10–6	6–3	3-1.5	1.5–0
Content (%)	5.18	13.68	22.55	14.97	43.62

$$P = \frac{F}{S}, \tag{1}$$

where P is the UCS (MPa), F is the compression force (N), and S is the cross-sectional area of the CPB test sample (mm^2).

2.3.2. XRD Analysis. Several small blocks were taken from within the CPB samples after breaking them into small pieces. These small blocks were immersed in absolute ethanol to terminate the hydration reaction. Next, these blocks were dried in vacuum at 55°C and ground into powders. These powders were characterized using an X-ray diffractometer (D/MAX-2400, Rigaku Corporation, Japan; scanning range = 5–60°; interval between data points = 0.02°; scanning speed = 4°/min).

2.3.3. SEM Analysis. Several small blocks were taken from within the CPB samples after breaking them into small pieces. These small blocks were immersed in absolute ethanol to terminate the hydration reaction. Next, these samples were dried in vacuum at 55°C. Cross sections of the block samples were taken and imaged using an SEM system (JSM-6510LV JEOL, Japan; magnification = 40× to 400,000×; accelerating voltage = 50 V to 30 kV, continuously adjustable; tilting angle = −10° to 70°).

3. Results and Discussion

3.1. UCS Test Results. According to the experimental data, the influence curve for TC coupling on the uniaxial compressive strength of CPB is plotted, as shown in Figure 2.

According to Figure 2, the TC coupling condition has a significant effect on the UCS of CPB. There are similar variations at each temperature, roughly divided into two phases: the intensity-increasing phase (before 7 days) and the intensity-declining phase (after 7 days). At the ages of 1 day and 3 days, the UCS increased at any temperature and concentration, and the water control group and the test samples at 20°C, 35°C, and 50°C indicated UCS increases of 18%, 25%, 29%, and 34%, 42%, 44%, respectively. The increase of temperature promotes the hydration reaction of cement of lower age, while Na_2SO_4 contributes to the

Figure 2: Variation of UCS of CPB under TC coupling conditions. (a) 20°C, (b) 35°C, and (c) 50°C.

strength of CPB in the early stages [23], resulting in a slight increase in CPB strength over a short time. However, compared with the results at 7 days, this increase was small.

3.2. Analysis of the Influence of TC Coupling on the UCS of CPB. According to the experimental results, the UCS of CPB was in the increasing stage before 7 days, but the increase at 1 day and 3 days was small. This study mainly discusses changes of UCS after 7 days and 28 days for CPB. Following the results shown in Figure 2, the compressive strength after 7 days and 28 days under TC coupling is plotted in Figure 3.

It can be seen from Figure 3(a) that at 20°C and 35°C, temperature and sodium sulfate concentration contribute to the UCS of CPB. The effect of temperature is more significant as it increases the UCS by 70% and 81% compared with the control group. On the one hand, this can be explained by the hydration reaction of dicalcium silicate and tricalcium silicate in cement [24], and a calcium silicate hydrate gel and calcium hydroxide are formed. In terms of filling materials, CPB does not have a dense structure. In the coagulation and finalization stage, the internal structure of CPB already has pores, and the increase of temperature accelerates the speed of the hydration reaction [25, 26]. CSH is insoluble in water; it precipitates and gradually aggregates into a gel, which fills the pores between the filler particles and provides early strength to CPB. On the other hand, sodium sulfate reacted with CH to form gypsum, which subsequently reacts with C_3A in the cement to form ettringite [27]. Consumption of CH by pozzolanic hydration of fly ash also occurs, and increasing temperature promotes the progress of these

Figure 3: Change in UCS of CPB. (a) 7 d and (b) 28 d.

reactions, thereby producing more hydration products. At 7 days, regardless of what kind of product is formed, it fills the pores of CPB. The amount of the product increases with the concentration of sodium sulfate solution, thereby improving the compactness of CPB. The macroscopic performance shows an increase in UCS.

At 50°C, the UCS of CPB changes; its strength increases rapidly when the concentration of sodium sulfate solution is low. At the sulfate level of 5%, the UCS increased by 50%; when the solution concentration was 7%, the UCS no longer increased but suddenly dropped by 32% compared with the results at 5%. The reason for this may be that the temperature was higher, completing the hydration reaction more quickly; when the temperature exceeds 45°C, ettringite begins to gradually decompose, resulting in increased porosity of the cementitious material, which ultimately reduces the strength of the CPB. Higher temperatures also result in increased adsorption on the sulphate ions in CSH, which can reduce the strength of the CPB. This process was verified through SEM analysis.

It can be seen from Figure 3(b) that, at 28 days, the role of sulfate is more obvious, and the UCS of CPB showed a downward trend in all working conditions. Compared with the control group, the UCS decreased by 41%, 50%, and 54% at 20°C, 35°C, and 50°C, respectively. The underlying mechanism of the change in strength was the coupling process between the physical and chemical deterioration, as well as the auxiliary effect of temperature on the deterioration. The reaction between sodium sulfate and CH was subsequently accelerated and the acicular expansion products ettringite and gypsum were formed. Increased concentration of sulfate results in the formation of more products and generation of greater expansion stress inside the CPB. The stress gradually increases with time, causing the CPB to crack, thereby decreasing its strength [30, 31]. On the other hand, the volume expansion is caused by the crystal precipitation of the sodium sulfate solution. The resulting crystalline product can expand the volume by a factor of four [32–34], and the resulting crystalline stress causes the generation of cracks, which ultimately leads to a decrease in the strength of CPB. As the temperature and solution concentration increase, the deterioration of CPB is accelerated, and the process can be further verified by XRD diffraction analysis.

3.3. The Influence of Stress and Strain on CPB under TC Coupling Conditions. Through the comparison of the full stress strain curves obtained by uniaxial compression testing of CPB, it was found that the shapes of the response curves were basically the same and followed the same rules. Therefore, a simplified diagram of the stress-strain curve for single axial compression of CPB is shown in Figure 4.

According to the stress-strain curve for the rock [35, 36], the stress-strain process of CPB is roughly divided into the following five stages: the first stage is the initial deformation stage (OA section). The simplified diagram shows the tendency of the curve to bend upwards, indicating that the slope of the curve is gradually increasing and that the stress of CPB increases with increasing deformation. This is because, under the action of the load, the cracks and pores perpendicular to the stress direction existing inside the CPB are closed owing to the pressure. In this process, through the subsidence of the roof and the deformation of the surrounding rock, the CPB forms a "passive support" for the roof and surrounding rock. Therefore, in the actual filling and mining process, it is necessary to ensure that CPB is fully connected in order to reduce the deformation of the roof and surrounding rock. The second stage is the elastic deformation stage (AB section) in which the curve is approximated as a straight line; that is, the slope of the curve is

Figure 4: Stress-strain curve of CPB under TC coupling condition.

constant in this region, and the strain increases with increasing stress. In this stage, the CPB has a great supporting effect on the surrounding rock, which can better control the deformation of the roof and surrounding rock. The third stage is the inelastic deformation stage (BC section) in which the curve is bent downward. This stage is the crack extension stage; that is, the slope of the curve is gradually reduced. As the external force increases, the stress field value of the end of the microdefects inside the CPB reaches its limit value, that is, its peak intensity. During this stage, the support of the CPB reaches its maximum under the condition of no damage. The fourth stage is the destruction stage (CD section) in which the curve is lowered; this stage is the CPB failure stage in which the slope of the curve becomes negative. During this period, the UCS of CPB gradually decreases as the strain continues to increase, and its residual strength is reached after reaching point D. Although the inside of the CPB has been damaged to a certain extent at this stage, it still retains a certain residual strength and can withstand a certain load, which is very advantageous for the support of the roof. After the fifth stage, the *D* point is reached, at which the curve tends to be gentle. This stage is the residual deformation stage of CPB in which the slope approaches zero. During this period, the UCS of CPB is maintained at a low level as the strain continues to increase, but still reaches 1/3 of the highest strength, exerting a bearing effect to some extent.

According to the test results, the model of the deterioration mechanism of CPB under TC coupling conditions was constructed, as shown in Figure 5. It can be seen from Figure 5 (1) and (2) that, in the absence of sodium sulfate, the hydration of CPB proceeds normally and that there are many pores present. Under a continuous load, the initial deformation stage (OA) in the stress-strain curve lasts for a long period of time, and the hydration products provide early strength. When the concentration of sodium sulfate solution increases, as shown in Figure 5 (3) and (4), ettringite, gypsum, and hydration products gradually increase in quantity, filling the internal pores of CPB, so that the CPB structure is more compact. The elastic deformation stage (AB) appears for a longer time in the stress-strain curve, and the corresponding UCS is large. When both the temperature and the concentration of sodium sulfate solution increase, as shown in Figures 3–5 (5), physical and chemical interactions occur simultaneously, and the growth of the crystal product is limited owing to excessive presence of the crystal product. This results in expansion stress, increased internal structure cracks in the packed body, and increased pore diameter. The elastic deformation stage (AB) under the load is small, lasting from initial deformation (OA) to the inelastic deformation stage (BC) and finally to the failure stage (CD).

3.4. Analysis of XRD Diffraction Results. In order to explore the deterioration products of CPB under TC coupling conditions, the UCS of CPB was sampled, and samples were subjected to XRD diffraction testing. The test results are shown in Figure 6.

After 7 days of deterioration, as can be seen from Figure 6(a), when the temperature is low, ettringite and gypsum diffraction peaks appeared at $2\theta = 11.05^\circ$, 20.91°, 31.12°, and 39.45° [37]. This is explained simply by the fact that there are two products of ettringite and gypsum at high sulfate concentrations. The higher the temperature, the more the product fills the inner pores of CPB, improving the UCS. When the temperature reaches 50°C, the ettringite and gypsum diffraction peaks were greatly enhanced while the CH diffraction peak was smaller (as shown in the enlarged part) because the higher concentration of sodium sulfate consumed CH. Hence, the formation of the two expansion products of ettringite and gypsum was rapidly increased, thereby generating expansion stress and resulting in a decrease in the UCS of CPB. From Figure 6(b), when the age of deterioration was 28 days, the diffraction peaks of ettringite and gypsum gradually increased with increasing temperature, and there was no CH diffraction peak at this time, indicating that it was completely consumed by sodium sulfate. In addition, the $Na_2SO_4{\cdot}10H_2O$ [38] diffraction peak appeared at $2\theta = 22.87^\circ$ and 41.48°, and the intensity of the diffraction peak increased with increasing temperature. This indicates that at, 28 days of age, the strength of the filler was further reduced owing to the crystal stress generated by the crystallization of sodium sulfate and the deterioration rate was positively correlated with the temperature and the solution concentration. The quartz and calcium carbonate (calcite) appearing in the figure include quartz contained in the fly ash itself, and may also contain calcium carbonate components brought by sand and stones when grinding the cement slurry [6, 39].

3.5. SEM Results Analysis. Since there are similarities in the SEM images at various temperatures, only partial SEM at 35°C and 50°C are shown.

It can be seen from Figures 7(a)~7(c) that when the solution does not contain sodium sulfate, there are some pores inside the CPB, the product is less, the particles are larger, and the hydration is incomplete. As the concentration of sodium sulfate solution increases, substances such as

Figure 5: Model of deterioration mechanism of CPB under TC coupling condition.

Figure 6: XRD diffraction analysis of CPB at a solution concentration of 7%: (a) 7 d and (b) 28 d.

Figure 7: SEM results. (a) 35°C, water, 7 d; (b) 35°C, 3%, 7 d; (c) 35°C, 7%, 7 d; (d) 50°C, 7%, 7 d; (e) 50°C, 5%,28 d; (f) 50°C, 7%, 28 d.

ettringite gradually increase and fill in the pores of the CPB, which improves the compactness of the CPB. It can be seen from Figure 7(d) that when the temperature and sodium sulfate concentration are higher at the same time, the hydration reaction rate is increased, and sodium sulfate reacts rapidly with $Ca(OH)_2$, resulting in excessive formation of ettringite and gypsum, both of which are expansive. The excessive amount of product will cause expansion stress in the pores inside the CPB and will sporadically grow from the inside to the outside, destroying the internal structure. Due to the short age of deterioration (7 d), the degree of damage is small. When the age reaches 28 d, as shown in Figures 7(e) and 7(f), on the one hand, the expansive product gradually increases in all directions in the internal structure of the CPB, and the expansion stress increases, resulting in an increase in cracks. On the other hand, due to the large pores at the surface of the aggregate and mortar, it is beneficial to the transport of sulfate ions. $Ca(OH)_2$ is consumed in a large amount, and the pH value inside the CPB is lowered, and the hydrated calcium silicate cannot be stably existed [40], and the final CPB strength decreased.

4. Conclusion

This study investigated the influence of TC coupling on the mechanical properties of CPB. Before 7 days at 20°C and 35°C, thermochemical coupling conditions had no negative influence on the UCS of CPB, and the UCS of CPB increased with the increase of temperature and sodium sulfate concentration, compared with the control group; the compressive strength increased by 70% and 81% at 7 d, and the growth rate at 35°C was obviously greater than 20°C. When the temperature was 50°C and the concentration of sodium sulfate was 3% or 5%, the rate of increase of the UCS of CPB was reduced, but was always advantageous compared with the control case of exposure to tap water. When the concentration of sodium sulfate reached 7%, the UCS decreased significantly, and TC coupling conditions began to have a negative impact on the CPB. When the age reached 28 days, the UCS of the CPB showed a decreasing curve for every temperature and concentration, and the rate of decrease accelerated with increasing temperature and concentration, compared with the control group, 7 d decreased by 41% at

20°C, decreased by 50% at 35°C, and decreased by 55% at 50°C. By analyzing the stress-strain relationship for CPB, the whole process of eroding CPB under TC coupling was analyzed. There is a certain compression deformation space inside the CPB, and the vertical deformation of the CPB is larger when it is subjected to the action of the roof plate. The TC coupling conditions were further validated for the proposed model of the CPB deterioration mechanism. The erosive mechanism of the deterioration products on the UCS of CPB was explained from the physical and chemical reactions. Moreover, the relationships between ettringite, gypsum, calcium hydroxide, and $Na_2SO_4{\cdot}10H_2O$ were clarified by XRD diffraction analysis and SEM analysis of TC coupling conditions. Among these, ettringite and gypsum were involved in the major material chemistry, leading to CPB structure damage. Additionally, $Na_2SO_4{\cdot}10H_2O$ resulted mainly from the physical effects of the CPB internal crystallization stress. According to the tests' conclusions, the variation of mechanical properties of CPB under TC coupling conditions is roughly understood, providing a theoretical basis to improve CPB durability for use in filling engineering.

Data Availability

All data are available within the article or from the corresponding author upon request.

Conflicts of Interest

The authors declare no conflicts of interest. The funders will not affect the submission.

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Research Article

Fatigue Life Behavior of Laser Shock Peened Duplex Stainless Steel with Different Samples Geometry

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Two different stress raiser geometries (fillets and notched) were treated by laser shock peening (LSP) in order to analyze the effect of sample geometry on fatigue behavior of 2205 duplex stainless steel (DSS). The LSP treatment was carried through Nd : YAG pulsed laser with 1064 nm wavelength, 10 Hz frequency, and 0.85 J/pulse. Experimental and MEF simulation results of residual stress distribution after LSP were assessed by hole drilling method and ABAQUS/EXPLICIT software, respectively. The fatigue tests (tensile-tensile axial stress) were realized with stress ratio of $R = 0.1$ and 20 Hz. A good comparison of residual stress simulation and experimental data was observed. The results reveal that the fatigue life is increased by LSP treatment in the notched samples, while it decreases in the fillet samples. This is related to the residual stress distribution after LSP that is generated in each geometry type. In addition, the fatigue crack growth direction is changed according to geometry type. Both the propagation direction of fatigue crack and the anisotropy of this steel results detrimental in fillet samples, decreasing the number of cycles to the fatigue crack initiation. It is demonstrated that the LSP effect on fatigue performance is influenced by the specimen geometry.

1. Introduction

Through surfaces engineering, it is possible to improve the surface properties of materials without changing the bulk material. Different surface treatments, such as shot peening, cavitation peening, laser shock peening, ultrasonic peening, deep rolling, among others, have proven to be effective techniques to increase the service time of mechanical elements. In recent years, the LSP treatment has been focused on fatigue behavior in different metal alloys due to its effectiveness in improving their mechanical properties [1]. During the first stage of the fatigue process, when applying an external load, the yield stress in localized regions (crystalline defects) is exceeded. The plastic deformation continues accumulating as more load cycles take place until one or more cracks are incubated. By inducing residual compression stresses through the LSP treatment and taking into account the superposition principle, in order to exceed the yield stress in these localized regions, it is necessary that the external load first counteract the compression state induced in the material and, subsequently, begins to deform in a tension state. In this way, the LSP treatment decreases the effect of the applied external load and delays the incubation time of the cracks.

Several investigations have been carried out around different LSP parameters in order to optimize the LSP process on fatigue response of different metal alloys. Rubiogonzalez et al. founded that as the density of pulses increases, the fatigue crack growth rate decreases in compact tension samples of Al-6061-T6 [2] and 2205 DSS [3]. The fatigue life behavior as function of laser swept direction was optimized for bending fatigue in Ti-6Al-2Sn-4Zr-2Mo [4]

Figure 1: Chemical composition of 2205 cold rolled plate.

Figure 2: Specimens dimensions for fatigue tests (mm): (a) fillet and (b) notched.

Figure 3: Experimental arrangement of LSP.

and tensile-tensile fatigue test to 2024-T351 [5], 2205 DSS [6], and 316L [7] alloys. Different fatigue behaviors were observed varying the swept direction for each case, demonstrating a high influence of this LSP parameter on fatigue life behaviors of these alloys. Sheng et al. [8] and Huang et al. [9] analyzed the morphology of fracture surface and the fatigue crack growth rate of 6061-T6. It is founded that as the laser pulse energy increases, both the fatigue striation spacing and fatigue crack growth rate decrease. The fatigue response as a function of coverage area paths was analyzed by Huang et al. in Ti-6Al-4V [10] and 6061-T6 [11] alloys. It is demonstrated the fatigue properties are influenced directly by this LSP parameter decreasing the fatigue crack growth rate in both materials.

On the other hand, for other test conditions such as specimens submitted to high temperature during fatigue test [12, 13] and specimens with previous damage (pre-fatigued) [14, 15], the LSP also has been an effective technique to improve the fatigue properties. Further, the effects on chemical composition after LSP and geometric

Figure 4: LSP Pulse sequence.

Figure 5: Finite element model of (a) notched and (b) fillet samples.

Figure 6: Plasma pressure profile.

Table 1: Johnson-Cook parameters for DSS 2205.

A (MPa)	B (MPa)	C	n	m	$\dot{\varepsilon}_0$
520	840.50	0.0124	0.1904	0.965	1

Figure 7: Experimental and simulation comparison of residual stress distribution.

Figure 8: Results of MEF simulation: (a) Residual stress distribution after LSP in notched sample and (b) Residual stress distribution after LSP in filled sample.

Figure 9: MEF simulation results of residual stress distribution in depth at midpoint of width.

factors, such as thickness and stress raisers, also have great influence on fatigue behavior. Rubio-gonzalez et al. [16] demonstrated that the LSP treatment in addition to the other treatment (Cold expansion) improves the fatigue life in open-hole samples of 6061-T6 Al alloy. However, individually, the LSP treatment resulted to have a detrimental effect on the fatigue crack initiation stage and a beneficial effect on the fatigue crack growth stage. Spadaro et al. [17] analyzed the effect of pulse density on low cycle fatigue in super-ferritic stainless steel and it is observed that the LSP causes thermal effects on the surface microstructure, generating intergranular corrosion, which decreases the fatigue life of this alloy. Bergant et al. [18] founded that the LSP increases fatigue crack growth rate in 6082-T651 Al alloy due the equilibrium tensile stresses in the center of specimen and notches edge as explained by Correa et al. [5, 7] through MEF simulation. Granados-Alejo et al. [19] analyzed the effect of samples thickness in 2205 DSS, in which it is observed that in thinner samples, higher fatigue life is obtained by LSP. This behavior is explained through the equivalent plastic strain distribution variations associated with the change in thickness of LSP treated samples. Similar trend about the sample thickness effects in 2024 Al alloy was observed by Achintha et al. [20]. For thick samples (~15 mm), the LSP had no effect on the fatigue life of this alloy. While for thin samples (~5 mm), the fatigue life was improved. The operation sequence of LSP without coating [21] and with coating [22] was investigated in thin open-hole samples for 6082-T6 and LY12CZ Al alloys, respectively. Similar observations are realized in both works, and higher fatigue life was observed in samples with hole drilled after LSP than samples with hole drilled before LSP. This phenomenon is related to residual stress distribution that was assessed by MEF simulation of LSP process.

It is known that other factors unrelated to LSP treatment parameters such as sample geometry, operation sequence and chemical composition, among others, are extremely sensitive to fatigue behavior of different metal alloys. In additions, no previous work has been reported the effect of LSP on fatigue behavior of fillet samples. In this work, two different geometries are compared (notches and fillets) taking into account the same operation sequence and LSP conditions in order to analyze the influence of sample geometry on fatigue life behavior of 2205 DSS. Additionally, the finite element simulation of the LSP process on notched and fillets samples are presented. The residual stress distribution was performed using the commercial code ABAQUS/Explicit and hole drilling method. In addition, the fracture surface is analyzed by SEM microscopy.

2. Material and Methods

The chemical composition results (in weight percentage) and the spectrum obtained through the EDS analysis are shown in the Figure 1. The mechanical properties were investigated by conducting the tensile test according to ASTM E24-08. The ultimate tensile strength, yield strength, elastic modulus, and elongation are 710 MPa, 520 MPa, 190 GPa, and 50%, respectively. Cold rolled plate of 9.5 mm thickness was machined to 4 mm (samples thickness). Special care was taken to cut the samples with a high-pressure water jet to reduce thermal damage on stress raiser geometries. The sample dimensions used in this work are displayed in Figure 2. The specimens were machined with the rolling axis parallel to the longitudinal axis. As reference, both 5 mm and 2 mm radius were selected with a similar stress concentration factor (Kt = 2) for the notched and fillet samples, respectively.

The laser technology system implemented to perform the treatments is shown in Figure 3 and consists of a Nd : YAG pulsed laser source with a wavelength of 1064 nm and a frequency of 10 Hz. During LSP, the laser beam (0.85 J, 6 ns) is directed and focused to 1 mm (spot diameter) on the sample surface through an optical system. The sample is moved with a position control device to generate the pulse sequence (zig-zag pattern) as shown in Figure 4. The pulses density and treated area were 2500 pulses/cm^2 and

Figure 10: Fatigue life for different sample geometry in function of samples geometry for treated and untreated samples.

(a) (b)

Figure 11: Fatigue crack growth direction of (a) notches; (b) fillets.

(a) (b)

Figure 12: Fatigue crack growth evolution in notched samples: (a) AR; (b) LSP.

Figure 13: Fatigue crack growth evolution in fillet samples: (a) AR; (b) LSP.

$25 \times 25\,\text{mm}^2$, respectively, in both sides of the samples. A water jet device was used to form thin layer water on samples surface during the LSP treatments. The samples without LSP treatment will be hereinafter called as as-received (AR).

Tensile-tensile fatigue test was carried out with MTS-810 servo-hydraulic machine with a load ratio of $R = 0.1$ and a frequency of 20 Hz. The experiments were realized at room temperature in the air. The loading applied on all the specimens was 24 kN, corresponding to maximum applied stress in reduced area of 300 MPa. The fatigue crack growth was monitored by a high-speed CCD camera using a magnification of 10x and the residual stress measurements were realized according to ASTM E28-04 by hole drilling method.

2.1. MEF Simulation. The commercial code ABAQUS/Explicit was used to estimate the residual stress distribution. The transient response of the laser pulse and the subsequent relaxation of the stress state constitute the numerical analysis. The FE model is shown in Figure 5. 190,512 C3D8R linear hexahedron elements were used for fillet sample and 224,826 for notched sample.

The pressure pulse was applied on 25×25 mm square. The time evolution is presented in Figure 6. Peak pressure of 5.21 GPa was obtained according to [23]. The LSP simulation process was simplified to one big laser shot according to [19].

The material model used to simulate the behavior of the material under the shock conditions (equation (1)) was the Johnson-Cook model,

$$\sigma = \left(A + B\varepsilon_{\text{eq}}^{n}\right)\left[1 + C\ln\left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right)\right]\left[1 - \left(\frac{T - T_0}{T_m - T_0}\right)^{m}\right], \quad (1)$$

with material constants are given in Table 1 [19].

3. Results and Discussion

3.1. Residual Stress Distribution. The comparison of the experimental and simulation results of the residual stresses distribution is shown in Figure 7. Compressive residual stresses are induced after LSP. The MEF simulation results show a similar trend to the experimental data. Figure 8 shows MEF simulation of residual stress distribution in the cross-section for notched (Figure 8(a)) and fillet (Figure 8(b)) samples.

It is observed that the LSP generates different residual stress distribution in each geometry type. The LSP treatment induces compressive residual stresses near surface. However, the balance of stress state generate tensile residual stresses at middle section of the thickness. The fillet sample present high

Figure 14: Fracture surfaces of the specimens with filet: (a, d) crack initiation, (b, e) fatigue crack growth, and (c, f) final rupture.

levels of tensile residual stresses near middle section of thickness in comparison of the notched sample. The residual stress profile in depth at midpoint of width is shown in Figure 9. In this point, the fillet sample presents a maximum value of tensile residual stress of 95.0 MPa at 1.6 mm depth. While the notched sample presents 5.3 MPa at 1.54 mm depth, a significant difference of 89.7 MPa between notched and fillet samples was observed. The constant area reduction of fillet sample could modify the residual stress distribution. The notched geometry has a symmetric axis in cross-direction that distributes the residual stress field more evenly in comparison with asymmetric axis corresponding to fillet sample. Several works have demonstrated that both the pulse sequence [24, 25] and the coverage area [26] modify the distribution of residual stress in sample edge and the middle of thickness. Further, other geometrical parameters such as the sample thickness also have a significant effect on the distribution of residual stresses as reported in [19, 27]. In this work, the same treatment condition was used in both sample geometries. It is evident that the geometry type is also an important factor which influences the residual stress distribution.

3.2. Fatigue Test. Fatigue test results are shown in Figure 10. It is observed that in notched samples, the LSP increases the number of cycles to failure of 521,500 for AR samples and to 779,450 after LSP. While in fillet samples, the LSP decreases the number of cycles to failure of 528,900 for AR samples and to 309,500 after LSP. It is observed that the LSP has different effects on fatigue behavior according to the sample geometry type tested, improving only the fatigue properties in notched samples. It is correlated with residual stress results in notched samples that showed higher compressive residual stress near surface and less tensile residual stress at middle of the sample thickness in comparison with fillet samples. Figure 11 shows the fatigue crack growth direction for each geometry type. The notched samples present a symmetric horizontal axis (perpendicular to loading axis) and the fatigue crack growth direction is given in this direction as shown in Figure 11(a). From Figure 11(b), it is observed that the fillet samples present an asymmetric horizontal axis and it has changed the fatigue short crack growth direction to an inclined plane and then return to the horizontal direction perpendicular to axis loading. It is

Figure 15: Fracture surfaces of the 4 mm thick specimen: (a, d) crack initiation, (b, e) fatigue striation, and (c, f) final rupture. Figure 15 is reproduced from Granados-Alejo et al. [19].

evident that the geometry type changes the preferential direction of fatigue crack growth. According to anisotropic effects reported in [28, 29], this direction change could influence fatigue results in fillet samples.

The fatigue crack evolution in treated and untreated notched samples is shown in Figure 12. For untreated samples (Figure 12(a)), the fatigue crack appears at surface at 487,000cycles and then 9,500 cycles elapse, accumulating 496,500 cycles to failure. While in treated samples (Figure 12(b)), the fatigue crack appears at surface at 877,500 cycles and then 12,400 cycles elapse, accumulating 889,900 to failure. It is observed that the LSP improves both the fatigue crack initiation and growth in notched samples.

The fatigue crack evolution in treated and untreated fillet samples is shown in Figure 13. For untreated samples (Figure 13(a)), the fatigue crack appears at the surface at 479,000 cycles and then 5,400 cycles elapse, accumulating 484,400 cycles to failure. While in treated samples (Figure 13(b)), the fatigue crack appears at surface at 313,000 cycles and then 6,500 cycles elapse, accumulating 319,500 to failure. It is observed that the LSP improves the fatigue crack growth stage while it is detrimental for the fatigue crack initiation stage in fillet samples. It is well known that the fatigue crack initiation depends on microstructural parameters such as crystallographic orientation, grain size, phase distribution, and the interaction of elastic and plastic properties between the austenite and ferrite phases [28, 30, 31]. In this stage, the short crack is propagating in the plane of the maximal shear stress. In the fatigue crack growth stage, the microstructural effects are inhibited, and the fatigue crack grows perpendicular to loading axis. In this stage, the compressive residual stress induced by LSP decreases stress ratio R according to superposition principle [32] and consequently the fatigue crack growth rate (da/dN) decreased.

3.3. Fracture Surface. The fracture surface for notched and fillet samples is shown in Figures 14 and 15, respectively. The LSP effect changes the fatigue crack initiation zone in

both samples' geometries from the border in AR samples (Figures 14(a) and 15(a)) at mid-thickness in treated samples (Figures 14(d) and 15(d)). The LSP induces high level of compressive residual stresses and hardening layer [33] that improves the surface properties causing this shift. Details of stable crack growth zone are shown in Figures 14(b) and 15(b) for untreated samples and Figures 14(e) and 15(e) for treated samples. The average value of fatigue striation spacing in AR samples is 0.748 and 1.337 μm for notched and fillet samples, respectively. While in treated samples, the average value of fatigue striation spacing is 0.346 and 0.403 μm for notched and fillet samples, respectively. In both cases, a lower value of fatigue striation spacing after LSP was observed, indicating a decrease in fatigue crack growth rate according to experimental results of fatigue tests. The final fracture is shown in Figures 14(c) and 15(c) for untreated samples and Figures 14(f) and 15(f) for treated samples. In both cases, similar mechanism of ductile failure is observed.

4. Conclusions

The influence of LSP on fatigue life of 2205 DSS with different sample geometry has been investigated. The LSP improves the surface properties of 2205 DSS increasing the fatigue life on notched samples. However, the fillet samples modify the residual stress distribution and change the short fatigue crack growth inhibiting the LSP effect.

The numerical simulation of residual stress distribution was used as powerful tools to predict and correlate this result with the fatigue behavior of the 2205 DSS. A good approximation between experimental and simulation results was obtained.

Both the reduction of fatigue striation spacing and the monitoring of fatigue crack growth corroborate that the LSP decreases the fatigue crack growth rate in both sample geometries. However, the propagation direction of fatigue crack growth in fillet samples decreases the crack incubation time. This material behavior is related to the residual stress distribution and the anisotropy of this steel. It is evident that the sample geometry is an important factor to the residual stress distribution and the fatigue life of 2205 DSS.

Data Availability

The experimental data used to support the findings of this study are included within the article.

Conflicts of Interest

The authors declare that there are no conflicts of interest regarding the publication of this paper.

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Research Article

Influence of Flaw Inclination Angle on Cracking Behavior of Rock-Like Materials under Uniaxial Compression

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The inclination angle of the flaw has an important effect on the cracking behavior of fractured rock. The influence of the inclination angle of closed flaw on the strength, cracking behavior, failure modes, and AE behaviors of specimens has not been investigated thoroughly. Uniaxial compression tests were conducted on gypsum samples containing a single closed flaw with different inclination angles. The flaw orientation influence on strength, cracking mechanism, failure modes, and AE behaviors was analyzed in detail. With the increase of the flaw inclination angle from 30° to 75°, the mechanical parameters such as elastic modulus, wing crack initiation strength, and uniaxial compression strength decrease first and then increase and reach the minimum value at 45°. The wing crack initiation angle decreases from 65° to 35° as the flaw inclination angle increases from 30° to 75°. The wing crack is more difficult to initiate from the tips of closed flaws compared with that of open flaws. The secondary cracks are initially shear cracks and propagate in a stable manner. However,when the axial stress reaches the peak strength, the secondary crack propagates unstably, which results in the macroscopic failure of the sample. The AE behaviors of samples are also analyzed during the uniaxial compression test. The experimental results are expected to provide helpful guidance for safe construction under fractured rock mass condition.

1. Introduction

Discontinuity of rock mass is a common phenomenon in slope engineering and underground engineering. A large number of engineering practices have proved that the instability and failure of rock mass engineering are closely related to the internal joints, flaws, and weak surfaces. In rock engineering, new cracks initiate from the tips of preexisting flaws and unstable failure of surrounding rocks will happen along with the coalescence of these cracks. In order to better understand the strength characteristics and cracking mechanisms of fractured rock, extensive research has been done under uniaxial compression [1–13].

The influence of the flaw inclination angle on cracking behavior has been extensively studied by many researchers. Bobet and Einstein [14] conducted compression tests on samples of gypsum to investigate the orientation influence on failure mechanism and crack coalescence modes. By using uniaxial compression tests on specimens, Park and Bobet [15] observed that initiation stress of wing cracks increased with flaw inclination angle. Yang and Jing [16] conducted several groups of uniaxial compression tests to investigate the influence of the flaw angle on the strength and cracking behaviors of sandstone samples. Zhang and Wong [17] used a bonded particle model to simulate the influence of the flaw inclination angle on cracking process in rock-like materials containing a single flaw. They concluded that the flaw inclination angle had a strong influence on the crack initiation and propagation patterns. Manouchehrian et al. [18] used PFC2D (particle flow code in 2 dimensions) to study the influence of the flaw orientation on the crack propagation mechanism in brittle materials under various compressive loads and detected two types of secondary crack propagation mechanisms. Zhou et al. [19] conducted uniaxial compression tests for rock-like materials containing multiple flaws to further research the influences of the layout of preexisting flaws on mechanical properties and crack coalescence types.

The closed flaw is one of the most common structural surfaces in rock mass, and the orientation of the flaw has an important influence on the cracking behavior of fractured rock. Although fruitful research findings were achieved in the past several decades, most of the experimental and numerical research studies were conducted on specimens containing open flaws. The influence of the inclination angle of closed flaw on the strength, cracking behavior, failure modes, and AE behaviors of specimens has not been investigated thoroughly. To address these problems, uniaxial compression tests were carried out on specimens containing a single flaw with four different flaw inclination angles. The experimental results are expected to provide helpful guidance for safe construction under fractured rock mass condition.

2. Specimen Preparation and Testing System

2.1. Specimen Preparation. In this experiment, the specimens were made from a mixture of water, high-strength gypsum, and quartz sand at a mass ratio of 1 : 3 : 0.5. Rectangular prismatic specimens, with dimensions of $120 \times 60 \times 40$ mm (height × width × thickness), shown in Figure 1, were prepared. The closed flaw in the specimen was made by inserting a sheet of resin flake, with thickness of 0.2 mm and length of 20 mm, during the solidification period of the specimen. The resin flake was pulled out before the sample reached the final solidification strength, and the closed flaw naturally formed during the solidification process as a result of the characteristic of volume expansion of gypsum. To investigate the influences of the inclination angle of the preexisting flaw on the mechanical characteristics of the specimen under uniaxial compression, four different inclination angles were designed. The inclination angle α (the angle of the flaw to the horizontal direction) of the flaw is 30°, 45°, 60°, and 75°, as shown in Figure 1. The physical and mechanical parameters of the specimen are listed in Table 1.

2.2. Testing System. The testing system as shown in Figure 2 includes loading equipment, AE monitoring equipment, and high-speed camera. The RMT-150C rock mechanics servo-controlled testing system, developed by the Institute of Rock and Soil Mechanics, Chinese Academy of Sciences, was chosen as the loading equipment. All tests were carried out under displacement control conditions with a loading rate of 0.002 mm/s. The AE signals were recorded by a DISP full-information AE measuring system made by American Physical Acoustics Corporation. In order to effectively remove the noise and to get the real AE signals generated by the specimens, the threshold value of AE signals was set to 40 db. The frequency of AE system was fixed to 1 MHz.

3. Analysis of Experimental Results

3.1. Characteristics of Strength and Deformation. Typical axial stress-strain curves for specimens containing single flaw under uniaxial compression are shown in Figure 3. As can be seen from Figure 3, the curve shape is obviously different for samples containing different flaw inclination angles which indicate that the flaw inclination angle has an important influence on the mechanical behavior of samples under uniaxial compression.

It can be seen from Figure 3 that after the peak strength, the load value drops rapidly and gradually stabilizes to a certain constant value, which is the residual strength of the sample. The residual strength of the sample indicates that the bearing capacity is not decreasing and no new cracks are produced in the sample. In the residual strength phase, continued loading is meaningless for the purposes of this study. Therefore, after the macroscopic damage of the sample, that is, the stress-strain curve shows obvious characteristics of residual strength, the loading stops.

The influence of the flaw inclination angle is investigated on the strength and deformation parameters of specimens under uniaxial compression, which are listed in Table 2. Three samples were tested under each flaw inclination condition. It should be pointed out that the test data with large deviation are not listed in Table 2. The definition for the mechanical parameters in Table 2 is described as follows. E denotes the elastic modulus, σ_i is the wing crack initiation stress, σ_p is the uniaxial compressive stress, and σ_r represents the residual stress.

Before analyzing the flaw orientation influence on strength and deformation parameters of specimens, it should be pointed out that a comprehensive method based on the stress-strain and AE hit curve of the specimen was used to determine the wing crack initiation strength σ_i. Taking the sample 30-2, which contains a flaw with an inclination angle of 30°, as an example, the comprehensive method to determine σ_i is analyzed as follows. The curve of stress-strain and AE hit of 30-2 is shown in Figure 4. As can be seen from Figure 4, the AE hit rate stayed at a low level during the stage of fissure closure and elastic deformation. However, when the axial stress increased to about 26.49 MPa, the wing crack emanated from the tip of flaw. The corresponding axial stress dropped rapidly from 26.49 MPa to 26.20 MPa; meanwhile, the AE hit rate increased abruptly to about 120 times per second. Therefore, the wing crack initiation strength σ_i of sample 30-2 is identified as 26.49 MPa.

The influence of the flaw inclination angle on σ_i, σ_p, σ_r, and E of specimens containing a single flaw is demonstrated in Figure 5. Combined with the experimental curves drawn in Figure 5 and experimental results listed in Table 2, the influence of the flaw inclination angle on σ_i, σ_p, σ_r, and E of specimens containing a single flaw is analyzed as follows.

The influence law of α on the elastic modulus of the specimen is shown in Figure 5(a). E first decreased and then increased and reached the minimum value at 45°. As α increased from 30° to 45°, the average value of E decreased from 13.30 Gpa to 12.83 GPa. However, the average value of E showed a steady increasing trend from 12.83 GPa to 13.53 GPa with the increase of α from 45° to 75°.

Figure 5(b) shows the influence of α on the uniaxial compression strength of specimens, which is similar to the variation tendency on the elastic modulus. The average value

Figure 1: Geometry of the specimens.

Table 1: Physical and mechanical parameters of the specimen.

Uniaxial compressive strength (MPa)	Elastic modulus (GPa)	Tensile strength (MPa)	Dry density (g·cm^{-3})	Poisson's ratio
39.27	13.46	1.48	1.83	0.23

Figure 2: Test loading, acoustic emission monitoring, and camera system.

Figure 3: Axial stress-strain curves of specimens containing a single flaw under uniaxial compression.

of σ_p at $\alpha = 30^\circ$ is 33.54 MPa, but the average value of σ_p at $\alpha = 45^\circ$ decreases to 28.80 MPa. With the increase of α from 45° to 75°, the average σ_p shows a steady increasing trend from 28.80Mpa to 32.85MPa.

Figures 5(c)–5(e) reveal the influence of the flaw inclination angle on the wing crack initiation strength σ_i and the ratio of σ_i to σ_p, respectively. After comparing the two curves plotted in Figures 5(c)–5(e), it is very clear that both σ_i and σ_i/σ_p have a similar trend with the increase of the flaw inclination angle from 30° to 75°. The two parameters decreased in the range from 30° to 45° and then increased from 45° to 75°. For instance, when α increased from 30° to 45°, the average wing crack initiation strength decreased from 25.69 MPa to 17.49 MPa. However, when α increased from 45° to 75°, the average wing crack initiation strength increased from 17.49 MPa to 31.85 MPa.

It needs to be mentioned that the ratio of σ_i/σ_p can be used to describe the degree of difficulty of the initiation of wing crack. According to the experimental results concluded by Zhao et al. [20], the ratio of σ_i/σ_p of gypsum samples containing open flaws with different inclination angles is plotted in Figure 6. From Figure 6, we can conclude that the

Table 2: Mechanical parameters of specimens containing a single flaw under uniaxial compression.

α (°)	Sample	E (GPa)	σ_i (MPa)	σ_p (MPa)	σ_i/σ_p (%)	σ_r (MPa)
30	30-1	13.49	24.88	33.21	74.92	0
	30-2	13.11	26.49	33.86	78.23	3.64
45	45-1	12.42	17.38	28.01	62.07	4.71
	45-2	13.24	17.60	29.59	59.48	3.98
60	60-1	12.73	29.41	31.58	93.13	5.98
	60-2	13.39	28.44	30.98	91.80	7.89
75	75-1	14.07	31.64	32.08	98.63	7.51
	75-2	12.99	32.06	33.62	95.36	6.87

Figure 4: Stress-strain and AE hit curve of the specimen 30-2.

wing crack is more difficult to initiate from the tips of closed flaws compared with that of open flaws.

Figure 5(d) is a plot of the residual strength σ_r versus the flaw inclination angle α. The change rule is that σ_r kept increasing with the flaw inclination angle α. The minimum average value of σ_r ($\alpha = 30^\circ$) was 1.82 MPa, and the maximum average value of σ_r was 7.19 MPa ($\alpha = 75^\circ$).

3.2. Cracking Behaviors. When a crack is stably expanded, the slowly expanding path can be clearly observed from the captured video. Moreover, the stable expansion of the crack can be promoted only when the axial load continues to increase. In the form of the stress-strain curve, the crack is stably expanded and the load is slowly increased. At the same time, the acoustic emission hit rate and energy of the sample will also show a slowly increasing trend. When the crack is unstably expanding, its expansion speed is very fast, and it is difficult to capture its expansion path from the captured video. At the same time, even if the axial load does not continue to increase, the crack will spontaneously expand. At the same time, the acoustic emission hit rate and energy of the sample will show a sharply increasing trend.

Taking the sample 30-2, which contains a flaw with an inclination angle of 30°, as an example, the cracking characteristics and failure process are described as follows. Tensile wing crack initiated when axial stress reached 26.49 MPa (see Figure 7(a), T stands for tensile wing crack). The generation of the wing crack resulted in a small drop of axial stress. With the further increase of axial stress, the wing crack lengthened and propagated to the edge of the sample (as can be seen from Figure 7(a)). When the sample was loaded to about 32.5 MPa, the secondary crack emanated from the upper tip of flaw. The secondary crack was originally shear crack and propagated in the plane of the flaw in a stable manner (see Figure 7(b), S stands for shear crack). The propagation of the secondary crack resulted in surface spalling phenomenon around the area of propagation route (see Figure 7(c), the grey area represents surface spalling). When the axial stress reached the peak strength, the secondary crack propagated in an unstable manner toward the upper boundary of the sample which resulted in an obvious shear movement along the plane of flaw.

As can be seen from Figures 8(a)–8(c), for the specimens containing a single flaw with the inclination angle of 45°, its cracking process is very similar to that of specimens containing a single flaw with the inclination angle of 30°. In fact, for the specimens containing a single flaw with the inclination angle of 60° and 75°, their cracking processes are all similar to those of specimens containing a single flaw with the inclination angle of 30°. However, for the specimens containing a single flaw with the inclination angle of 0° and 90°, no cracks were observed to initiate from the tips of the flaw during the whole loading process which is consistent with the numerical simulation results conducted by Liu and Yang [21]. This phenomenon is very different from that conducted on samples containing open flaws [6, 7, 17, 18, 22]. The reason for this phenomenon may be due to the large coefficient of friction between the preexisting flaw surfaces.

Furthermore, the influence of the flaw inclination angle α on the wing crack initiation angle θ (the angle between the tangential direction of the wing crack and the long axis of the preexisting flaw as shown in Figure 7(a)) is also analyzed as demonstrated in Figure 9. The results show that wing crack initiation angle θ decreased from 65° to 35° as the flaw inclination angle α increased from 30° to 75°. The change trend is consistent with that conducted on samples containing open flaws [14].

There is one point that should be pointed out. Yang and Jing [16] categorized nine different crack types based on their geometry and crack propagation mechanism of sandstone samples containing a single flaw under uniaxial compression. However, in this experimental investigation, there are only two main types of cracks, e.g., the tensile wing crack and the shear secondary crack initiated from the tips of the flaw. Three main reasons can be concluded to explain this discrepancy. The fist and the most important point is the difference in the type of the flaw. In this research, we adopted a closed flaw, but in Yang's research, he used an open one. The second one is the difference of experimental material. The sample used by Yang is sandstone, but high-strength gypsum is used in this experiment. The third point is the

Figure 5: Influence of the flaw inclination angle on the strength and deformation parameters of specimens under uniaxial compression: (a) elastic modulus; (b) uniaxial compressive strength; (c) wing crack initiation strength; (d) residual strength; (e) ratio of σ_i/σ_p.

difference in the length of the flaws. Further research will be conducted to reveal the inherent reasons.

3.3. Failure Modes. Typical failure modes of specimens containing a single flaw with different inclinations are described in Figure 10. For specimens containing a single flaw with the inclination angle of 30°(Figure 10(a)), 45°(Figure 10(b)), 60°(Figure 10(c)), and 75°(Figure 10(d)), they exhibit similar failure modes. The initiation and propagation of wing crack destroyed the integrity of the sample. However, the initiation and propagation of secondary crack resulted in the ultimate macrofracture of samples. After the experiment, as shown in Figure 11, the

Figure 6: Influence of the flaw inclination angle on the ratio of σ_i/σ_p.

Figure 7: Crack initiation and propagation process of sample 30-2 under uniaxial compression: (a) wing crack initiation and propagation; (b) secondary crack initiation and propagation; (c) ultimate failure.

fractured sample can be easily forced apart with hands along the irregular plane of propagation tracks of secondary crack.

4. Acoustic Emission Characteristics

As a nondestructive testing method, the acoustic emission (AE) monitoring technique has been widely used to better understand the failure process of rock in laboratory [23, 24]. Figure 12 shows the AE hit rate and AE cumulative energy of some typical samples in the process of uniaxial compression test. From Figure 12, one can find that for all the samples containing a single flaw with different inclinations, both the AE hit rate and AE cumulative energy remained at a very low level in the stage of fissure closure and elastic deformation.

For the samples containing a single flaw with the inclination angle of 30°(Figure 12(a)), 45°(Figure 12(b)), 60°(Figure 12(c)), and 75°(Figure 12(d)), the AE hit rate increased sharply due to the initiation of wing crack. Meanwhile, the AE cumulative energy increased abruptly to a high level. As the axial stress reached the peak strength, both the AE hit rate and cumulative energy increased abruptly due to the unstable propagation of secondary crack which resulted in the ultimate failure of the sample.

The AE events were very few at the stage of residual strength for the samples containing a single flaw with the inclination of 30°, 45° and 75°. However, for the samples containing a single flaw with the inclination of 60°, as shown in Figure 12(c), the AE events still maintained a certain level of activity which is characterized by several slight increases of the AE cumulative energy. The reason for this phenomenon is that some certain parts of the samples broke off due to the constraint influence of the loading plate.

5. Conclusions

Uniaxial compressive tests were carried out to investigate the strength, cracking process, failure modes, and AE behaviors of rock-like materials containing a single closed flaw with four different inclination angles from 30° to 75°. The following conclusions can be summarized based on the analysis of experimental results.

Figure 8: Crack propagation process of sample 45-1 under uniaxial compression: (a) wing crack initiation and propagation; (b) secondary crack initiation and propagation; (c) ultimate failure.

Figure 9: Influence of the flaw inclination angle on the wing crack initiation angle.

Figure 10: Continued.

(c) (d)

Figure 10: Failure modes of specimens containing a single flaw with different inclination angles under uniaxial compression: (a) $\alpha = 30^{\circ}$; (b) $\alpha = 45^{\circ}$; (c) $\alpha = 60^{\circ}$; (d) $\alpha = 75^{\circ}$.

Figure 11: Internal rupture of sample.

Figure 12: Continued.

Figure 12: Relation between axial stress, AE hit rate, AE cumulative energy, and time of samples containing a single flaw with different inclinations: (a) $\alpha = 30^\circ$; (b) $\alpha = 45^\circ$; (c) $\alpha = 60^\circ$; (d) $\alpha = 75^\circ$.

With the increase of the flaw inclination angle from 30° to 75°, the corresponding mechanical parameters, such as elastic modulus E, wing crack initiation strength σ_i, and uniaxial compressive strength σ_p, decrease first and then increase and reach the minimum value at 45°. However, the residual strength σ_r exhibits an increasing tendency with flaw inclination angle. The wing crack is more difficult to initiate from the tips of closed flaws compared with that of open flaws.

Wing crack and secondary crack were detected in the specimens containing flaws with the inclination angle of 30°, 45°, 60°, and 75°. Wing crack initiation angle θ decreases from 65° to 35° as the flaw inclination angle α increases from 30° to 75°. The secondary cracks are originally shear cracks and propagate in a stable manner in a plane coplanar with the flaw. However, when the axial stress reaches the peak strength, the secondary crack propagates in an unstable manner which results in the ultimate failure of the specimens.

For the samples containing a single flaw with the inclination angle of 30°, 45°, 60°, and 75°, the AE hit rate increases sharply due to the initiation of wing crack. When the axial stress reaches the peak strength, both the AE hit rate and cumulative energy increase abruptly due to the unstable propagation of secondary crack which results in the ultimate failure of the sample.

Data Availability

Any reader or researcher who wishes to obtain the research data of this article can contact the author by e-mail.

Conflicts of Interest

The author declares no conflicts of interest.

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Research Article

Mode I Fracture Toughness Test and Fractal Character of Fracture Trajectory of Red Sandstone under Real-Time High Temperature

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To evaluate the stability and compactness of high-temperature underground construction, it is necessary to test the fracture toughness of surrounding rock (red sandstone) under real-time high temperature. In this paper, SCB specimens recommended by the International Society for Rock Mechanics are used to measure the mode I fracture toughness of red sandstone at real-time high temperatures. Also, to reveal its fracture characteristics and fracture mechanism, the fracture morphology observation (SEM experiment), XRD experiment, mercury intrusion porosimetry testing, and fractal measurement of fracture trajectory are carried out on the red sandstone specimens at various temperatures. The results show that (1) temperature may have a significant impact on the fracture toughness and fracture characteristics of red sandstone. On the whole, the fracture toughness values decrease with the increase in temperature, while the fractal dimensions of fractal trajectories increase with the increase in temperature. (2) Temperature has a significant influence on the fracture mode of red sandstone. At relatively low temperatures (20°C–400°C), the main fracture mode is transgranular failure. At relatively high temperatures (400°C–700°C), the fracture mode is mainly intergranular failure. (3) The weakening mechanism of red sandstone is mainly due to the effect of thermal dehydration when the temperature is between 100°C and 400°C. When the temperature is between 400°C and 700°C, the weakening mechanism is mainly due to thermal cracking and the α-β phase transition of quartz.

1. Introduction

In the deep geological disposal of high-level radioactive waste such as nuclear waste, the design requirements of surrounding rock and artificial barrier system are very strict. It is necessary to consider the original strength of surrounding rock as well as the influence of thermal stress on its microstructures. According to the study of Zuo et al. [1–4], the highest temperature in the surrounding rock will reach about 100°C–1000°C during the geological disposal of nuclear waste, which will make the surrounding rock undergo drastic high-temperature physical and chemical changes. To prevent the leakage of nuclear wastes, it is essential to study the evolutional processes of mechanical strength and internal microstructures of surrounding rock at high temperature in real time. Besides, in the fields of geothermal resource development, oil and gas and coalbed methane development, oil and gas geological storage, deep underground space utilization, and marine deep well drilling [5–8], it is also necessary to study the variational trend of physical and mechanical parameters of rock at high temperature and its weakening mechanism.

In geotechnical engineering science, fracture toughness is a property which describes the ability of a rock to resist fracture and is one of the most important features of any solid material for many underground tunnel design applications [9–12]. The linear-elastic fracture toughness of rock is determined from the stress intensity factor (K) at which a thin crack in the material begins to grow and is a measurement of the energy required to produce a tiny crack. Proverbially, the fracture behavior of rock strata will be significantly affected by geostatic stress and thermal stress. And, it has been widely proved that many initial

cracks and microcracks induced by external stress disturbance (i.e., excavation-induced cracks and fissures) are widely and randomly distributed in surrounding rock. Thus, under the action of in situ stress and thermal stress, when the stress intensity factor at the crack tip is higher than the fracture toughness of surrounding rock, microcracks in surrounding rock will further expand, propagate, and coalesce and eventually form macroscopic cracks, which will ultimately affect the stability and safety of underground engineering buildings. Therefore, to accurately evaluate the stability and safety of underground engineering, it is necessary to conduct experimental research on the fracture behavior of rock and the corresponding crack growth law.

It should be noted that, for some complex underground and tunneling projects, surrounding rock is often greatly affected by extremely high temperature. For examples, when the radioactive nuclear waste is deeply buried, a large amount of heat will be generated due to the decay of radioactive elements (137Cs, 60Co, 90Sr, etc.) [13]. Usually, these heats will cause the temperature of the surrounding rock to be as high as 1000°C. However, the existing studies show that local thermal stress will be generated between mineral particles due to the differences in the thermal expansion coefficient between rock mineral particles at high temperature. In this way, the internal structure of the particles will change substantially, and thermal cracking may occur in the rock, which eventually affects the physical and mechanical properties of the rock [14–19]. Besides, it is also well known that the fracture trajectories and fracture roughness and surface morphology are strictly related to the seepage behavior of geofluids. Therefore, the fracture toughness and fracture characteristics of surrounding rock at real-time high temperature need to be studied to evaluate the migration paths of geological storage of nuclear wastes.

As a typical sedimentary rock, sandstones are widely distributed in nature. Due to the influence of the geochemical environment, there are substantial differences in mineral composition, particle shape, and cement type. After treatment with high temperature, the minerals in the sandstone undergo a polycrystalline transformation, and their mineral composition and physical and mechanical parameters will correspondingly change. Previously, several scholars have concentrated on the study of the damage mechanism, failure criterion, microstructure, and texture evolution of rocks after treatment with high temperature and have obtained some useful research results. Zuo et al. [1–4] systematically studied the thermal cracking, failure mechanism, and fracture morphology of sandstone at various temperatures in real time and determined the influence of temperature on the fracture behavior of sandstone; that is, with the increase in temperature, the fracture mode of rock changes from local brittle fracture to local brittle-ductile coupling fracture. And they considered that 300°C is the threshold temperature for the brittle-ductile transition. In the transition zone, the fracture morphologies of rocks are more diverse and complex. Mahanta et al. [9] have carried out the model I fracture toughness tests of Manoharpur sandstone of India after high-temperature treatment. The results show that the fracture toughness of sandstone reaches its maximum when the temperature reaches 100°C. Then, the value decreases sharply with the increase in temperature. Feng et al. [20, 21] studied the effect of temperature on the mode I fracture toughness and fracture characteristics of Chinese sandstone. The results show that the change rate of fracture toughness of sandstone is closely related to temperature. Tian et al. [22] studied the variational trend of physical properties of sandstone subjected to high temperature. It is found that the permeability increased gradually with the increase in temperature. The bulk density of sandstone decreased significantly at a temperature above 500°C, and the porosity increased substantially at a temperature above 300°C. Johnson et al. studied the change characteristics of wave velocity of granite, limestone, diabase, and sandstone after treatment with high temperature. It is found that the wave velocity will attenuate at high temperature, and the phenomenon is closely related to grain boundary cracking.

In summary, under the action of high temperature, the minerals in sandstone will undergo physical and chemical changes such as dehydration, melting, decomposition, and phase transformation, which will lead to great changes in the crystal structure and composition of minerals. However, the above scholars have not revealed the thermal damage mechanism of rocks from the mineralogical direction, and the correlation between micro- and macromechanical parameters has not been discussed. Although the relevant scholars have qualitatively analyzed the changes of chemical and mineral composition using XRD and FT-Raman spectroscopy, a great number of research works still need to be done on the quantitative analysis of compositional variation in minerals, the relationship between the evolution process of mineral components and physical and variational trend of mechanical parameters, microfracture morphology, porosity, and other micro-macroparameters. Given the complicated temperature-related engineering problems, the fracture toughness and fracture characteristics of typical red sandstone at real-time high temperature are carried out in this paper. Besides, microscopic analysis such as quantitative XRD analysis (mineral composition), SEM fractographic analysis (fracture topography), mercury intrusion porosimetry measurement, and fractal measurement (fracture propagation trajectory) reveal the thermal fracture mechanism of sandstone and the correlations of micro-macromechanical parameters.

2. Materials and Methods

2.1. SCB Specimens and Calculation of Mode I Fracture Toughness. In fracture mechanics, there are three types of fracture: mode I (tensile fracture), mode II (shear fracture), and mode III (out-of-plane tearing fracture) [20–29]. Among them, mode I fracture is the most frequently encountered in underground geotechnical engineering [20, 30, 31]. Several tests have been proposed to estimate the mode I fracture toughness of rocks, namely, the chevron bend (CB) test

[32, 33], the edge cracked circular specimen subjected to wedge loading [34], the center-cracked ring specimen subjected to diametral compression [35], the single-edge-cracked round bar in the bending test [36], noncircular shape specimens such as the edge-cracked rectangular beam specimen subjected to three- or four-point bending [37, 38], the short rod (SR) test [38, 39], the central straight through the Brazilian disk test [33], the cracked chevron-notched Brazilian disk (CCNBD) test [40, 41], the SCB test, and the straight-notched disk bending test [42], and edge crack triangular (ECT) specimen [43]. The International Society for Rock Mechanics (ISRM) [42–44] suggested CB, SR, and CCNBD tests for the determination of the static fracture toughness of rocks. The SCB test was added to the suggested methods in 2014 [45]. In this paper, the SCB test was chosen to estimate the fracture toughness of the rocks for its simplicity of specimen preparation, equipment, and testing procedure [46] (Figures 1 and 2).

The data obtained from the experiment are calculated and processed according to the ISRM recommended method. Fracture toughness K_{IC} of SCB specimens is calculated from the following formula [45]:

$$K_{\mathrm{IC}} = \frac{P_{\max}\sqrt{\pi\alpha}}{2\mathrm{RB}} Y^{*}, \tag{1}$$

$$\begin{aligned} Y^{*} = {} & -1.297 + 9.516\left[\frac{S}{(2R)}\right] \\ & - \left\{0.47 + 16.457 \cdot \left[\frac{S}{(2R)}\right]\right\}\beta \\ & + \left\{1.071 + 34.401\left[\frac{S}{(2R)}\right]\right\}\beta^{2}, \end{aligned} \tag{2}$$

$$\beta = \frac{a}{R}, \tag{3}$$

where $P_{\max}$ is the peak load and Y^{*} is the dimensionless stress intensity factor.

The geometric dimensioning of SCB specimens used in this paper is as follows: $B = 20$ mm; $a = 12.5$ mm; $R = 25$ mm; $S/(2R) = 0.61$; $a/R = 0.5$; $Y^{*} = 4.7672775$ [45].

2.2. Sample Preparation. Red sandstones are taken from the Taiyuan area, Shanxi Province. The main components of sandstone are albite and quartz. The specimens are light red and cemented with shale. The particle size is 0.1–0.7 mm, and the average particle size is about 0.4 mm (see Table 1 for detailed composition and content). The sandstone has a density of 1.93 g/cm^3, a porosity of 7.16%, a uniaxial compressive strength of 37.6 MPa, a uniaxial tensile strength of 3.4 MPa, a P wave velocity of 1982 m/s, and a S wave velocity of 1269 m/s.

The sample preparation process can be divided into three parts:

(1) Place the sandstone on a flat platform. When the fixing is completed, the cylindrical cores are drilled at a speed of 10 mm/s using a drill with a diameter of 50 mm.

Figure 1: SCB specimen geometry. R, sample radius; B, sample thickness; a, precrack length; S, support span; P, load.

Figure 2: The SCB specimens.

(2) The cores obtained are cut into the semidisc samples with a thickness of 25 mm by a cutter. Then, the ends of the tested samples are smoothed using 30-grit fine sandpaper.

(3) Clamp the semicircular specimen with pliers. After the fixing is completed, the crack prefabrication is performed on the tested samples by using the crack prefabrication equipment. The thickness of the saw blade in the crack prefabrication equipment is 0.4 mm.

3. Experimental Results

3.1. Load-Displacement Curves. During the thermomechanical loading, the experimental load-displacement curves were recorded automatically. By analyzing the experimental load-displacement curves, the fracture in compression is a complex process consisting of 4 stages of fracture development (as shown in Figure 3). (1) Compaction stage (OA):

Table 1: Compositions and contents of the tested specimens (mean value).

Chemical component	Quartz (SiO_2)	Albite (Ca-rich ((Na,Ca)Al (Si,Al)$_3$O$_8$))	Albite ($NaAlSi_3O_8$)	Anorthite ($CaAl_2Si_2O_8$)
Content	85.4%	8.7%	4.8%	1.1%

Figure 3: Typical load-displacement curve of sandstone.

Figure 4: Typical load-displacement curve of sandstone at various temperatures.

the curve bends slightly upward, which is caused by the closure of origin cracks. (2) Elastic stage (AB): at this stage, the shape of the curve is close to a straight line. This is because the axial force is linear with the displacement. (3) Plastic deformation stage (BC), sandstone: at this stage, the force-displacement curves bend downward; that is, the processes exhibit stronger nonlinear trends in the relationship between force and displacement. Besides, the slope of the force-displacement curve gradually decreases to zero as the force increases. (4) Complete failure stage (CD): the force drops sharply. At this point, the load reaches its peak, which means that the rock has been completely broken into small pieces.

Since the force-displacement curves of sandstone specimens at the identical temperature have similar variational trend, the typical force-displacement curves of each condition are listed in this paper, as shown in Figure 4. From the force-displacement curves of sandstone at various temperatures, it can be concluded that (1) the variational trend of complete force-displacement curves of sandstone is all the same, which has been divided into compaction stage, elastic-plastic deformation stage, and failure stage; (2) at room temperature (20°C), sandstone suddenly breaks after the plastic stage, indicating a typical brittle failure; (3) when the temperature rises to 100°C–400°C, the curve slope at the elastic stage (AB) changes obviously; that is, the slope decreases gradually with the increase in temperature. In addition, the phenomenon of "load drop" appears on the curve. The main reason for this change is that sandstone produces a series of complex microcracks under the action of high temperature. When the microcracks expand and evolve through each other during loading, the load will drop suddenly; (4) when the temperature rises to 400°C–600°C, the prepeak AC and CD sections of the load-displacement curve tend to be gentle gradually, and the maximum displacement will also show an increasing trend with the increase in temperature. This is mainly due to the decrease in brittleness index and the enhancement of ductility of the specimens.

At real-time high temperature, the relationships between the mean of fracture toughness of sandstone specimens and temperature are shown in Figure 5 and Table 2. On the whole, the fracture toughness of sandstone at different temperatures has a certain degree of discreteness, which is mainly due to the structural differences of sandstone. The variational trend of fracture toughness of sandstone with temperature has the following characteristics: (1) with the increase in temperature, the value of fracture toughness of sandstone first increases then decreases; the value of fracture toughness reaches the maximum at 100°C, which is 4.27 MPa m$^{0.5}$, and the value is 0.18 MPa m$^{0.5}$ when the temperature rises to 700°C, which is 95.7% lower than that at 100°C. This fully demonstrates that the fracture behaviors of sandstone are greatly affected by high temperature. (2) According to the relationships between fracture toughness and temperature (i.e., curve slope), the variational trend of the curve can be divided into three stages: (1) the slow increase stage (20°C–100°C). The curve slope at this stage is 0.0055. (2) The slow descent stage (100°C–400°C). The curve slope at this stage is 0.0048. (3) The rapid descending stage (400°C–700°C), in which the slope of the curve is 0.0087. Based on these, it seems that the brittle-ductile transition of sandstone occurs at 400°C.

3.2. Fracture Trajectory. It is well known that the shape and trajectory of rock fracture are determined by its

Figure 5: The relationship between temperature and the mean of fracture toughness.

Table 2: Fracture toughness values of sandstone at various temperatures.

Temperature (°C)	#1	#2	#3	#4	#5	#6	Mean	Standard deviation
20	4.05	3.42	4.08	3.62	3.91	3.29	3.73	0.34
100	4.19	4.46	4.69	4.17	4.08	4.06	4.27	0.25
200	4.05	4.09	3.46	3.97	3.69	3.80	3.84	0.24
300	3.40	3.30	3.90	3.28	3.87	3.72	3.58	0.28
400	2.78	3.08	2.49	3.21	2.20	2.57	2.72	0.38
500	1.57	1.29	1.35	1.41	1.15	1.19	1.33	0.16
600	0.81	1.25	0.89	1.02	1.18	0.80	0.99	0.19
700	0.15	0.12	0.17	0.26	0.14	0.25	0.18	0.06

microfracture mechanism, which can directly reflect the macrofracture characteristics of the rock. Since the initiation cracks generally occur at the surface, subsurface, stress concentration zones, and internal defects of the specimen, the fracture propagation path of the specimen is tortuous and continuously changing. According to the previous studies, it is widely recognized that fracture modes can be classified into 3 categories: transgranular fracture, intergranular fracture, and mixed fracture, which can be presented on the surface of the specimen. According to the research results of Zuo et al. [1–4], it can be concluded that the roughness of the fracture surface after fracture is a comprehensive reflection of its microstructure, a magnitude of load and temperature [47, 48]. Therefore, the fracture trajectory is closely related to the roughness of the rock mass structural surface [49–53]. According to the research results of Liu et al. [53], the roughness of the structural plane can directly affect the shear strength and permeability of surrounding rock, whereas the shear strength and permeability are closely related to the stability and compactness of nuclear waste storage. Therefore, in this section, to quantitatively evaluate the tortuosity and complexity of fracture trajectory, the fractal theory and box counting method are used to describe the change law of fracture trajectories and quantitatively obtain the relationship between the tortuosity of fracture propagation path and temperature of sandstone (Figure 6).

The average of fracture deviations from the center line originating at the notch tip of sandstone specimens at the temperature ranging from room temperature (20°C) to

Figure 6: Fracture trajectory of red sandstone at 700°C.

700°C is shown in Figure 7 and Table 3. It can be clearly obtained that the value gradually increases with the increase in temperature. And, it is smaller at relatively low temperatures (20°C–400°C) but larger at relatively high temperatures (400°C–700°C). This is because sandstone belongs to a kind of brittle material at a relatively low temperature. According to the theory of energy dissipation, when the brittleness coefficient of the specimen is higher, it is easier to be entirely destroyed, which is mainly manifested by dynamic instability behavior such as the violent spattering of broken rock blocks. Regarding this kind of fracture mode, transgranular failure is more likely to occur so that the fracture deviation is smaller. When the temperature increases, the fracture mode of the specimen gradually transforms from brittleness to ductility. Hence, the postpeak deformation of the specimen becomes more substantial, and the elastic release energy of the specimen decreases. Under this condition, the intergranular failure is more prone to occur to the specimen. In addition, due to thermal cracking,

Figure 7: The average of fracture deviation of sandstone at different temperatures.

Table 3: The average of fracture deviation from center line of sandstone at various temperatures.

Temperature (°C)	Fracture deviation (left side) (mm)	Fracture deviation (right side) (mm)
20	2.6	1.9
100	2.4	1.8
200	3.5	2.8
300	3.8	4.5
400	4.3	5.2
500	5.7	6.4
600	6.8	7.5
700	7.9	9.6

a serious of complex microcracks, fissures, and voids are randomly generated in the specimen, which will make the crack propagation path be extremely irregular. Therefore, under high-temperature conditions, due to the change of fracture mode and thermal cracking, a large fracture deviation is formed.

According to the previous studies, the fracture trajectories of the specimens obviously have strong statistical self-affinity. Hence, the fracture propagation path on the sandstone surface can be described by fractal dimension [18–20]. By performing binarization on the crack images, black-and-white bitmaps of simple cracks are obtained. Then, the black-and-white bitmaps of the crack are covered by the boxes. Since the total number of boxes varies with the size W, the cracks with different shapes and sizes will correspond to the different data sets (W, N). When logarithms of $1/W$ and N are taken, respectively, a new set of data ($\ln(1/W)$, $\ln N$) is obtained. Then, these data are plotted into a logarithmic graph, and a set of equation is obtained after linear regression analysis, as shown in equation (3):

$$\ln N = a + D\ln\left(\frac{1}{W}\right), \tag{4}$$

where D is the slope of the regression line, and it is also the fractal dimension of the fracture propagation path.

Note that we register exactly the same trajectory of a specimen for three times and we calculate the fractal dimensions immediately. Note that six SCB specimens were used for exploring the fracture characteristics under each set of temperature. Therefore, 18 tests (the calculation of fractal dimension of fracture trajectories) were performed in total under each set of temperature. The average fractal dimensions and total errors of fracture trajectories at each temperature are shown in Table 4. From the experimental results (see Figure 8 and Table 4), we can obtain that the fracture trajectory of specimen has strong irregularity and fractality. With the increase in temperature, the average fractal dimension value increases gradually. For example, when the experimental temperature is 20°C, the average fractal dimension is 1.19 ± 0.03, and when the temperature is raised to 700°C, the average fractal dimension is increased to 1.84 ± 0.06, which is 52.1% higher than the room temperature. This fully demonstrates that the fracture trajectory of the sandstone samples becomes more and more complicated and tortuous.

4. Discussion

4.1. Correlation of Micro-Macromechanical Parameters. The main mineral components of sandstone are quartz and albite. After treatment with high temperature, the pore structure, surface morphology, crystal structure, and mineral composition of the specimens have undergone a series of complex changes. This will strongly influence the fracture behavior of sandstone. As shown in Table 5, the diffraction peaks of quartz and feldspar change obviously after treatment with a high temperature of 200°C. Compared with room temperature and 100°C, the content of quartz decreases, while feldspar increases. When the temperature is raised to 300°C, the quartz content in the sample is further reduced, and the feldspar content is also significantly reduced. Among them, the content of calcium-rich albite and calciferous feldspar increases compared with that at 200°C. In addition, it is interesting to find that a new mineral component, laumontite, appears in the samples. This is because the sodium feldspar in the sample undergoes a dehydration reaction at a high temperature, and the decomposition product chemically reacts with quartz to form the new mineral. When the temperature is between 400°C and 700°C, the laumontite disappears, and new minerals (i.e., calcium feldspar rich in sodium) appear in the sample. And the quartz content is gradually increased at this stage as the temperature increases.

It is worthy to note that chemical reactions between mineral particles (crystalline water removal, the formation of new materials, decomposition reaction, and lattice transformation) will lead to a series of physical changes, such as the increase in pore structure, the generation of new pore and fracture, and volume expansion. According to the experimental results, we can conclude that the higher the temperature is, the more complex the chemical reactions occur in the crystals of sandstone. For example, when the temperature of sandstone rises from room temperature to 700°C, quartz decomposition and synthesis, the transformation between

Table 4: Average fractal dimension values of crack development paths under different temperatures.

Temperature (°C)	Average fractal dimension (D)	Total error
20	1.19	0.03
100	1.12	0.07
200	1.26	0.04
300	1.32	0.04
400	1.39	0.02
500	1.43	0.09
600	1.68	0.11
700	1.84	0.06

Figure 8: The relationship between average fractal dimension and temperature.

quartz crystals, dehydration of feldspar, and synthesis and decomposition of feldspar will occur. Because the heating rate in this paper is relatively low, it means that the chemical reactions experienced by mineral crystals at 700°C include not only all chemical reactions from room temperature to 600°C, but also those occurring at 700°C. But for 200°C, only a small part of quartz and feldspar is decomposed or synthesized. Therefore, the macroparameter (i.e., fracture toughness) gradually decreases with the increase in temperature.

However, to accurately determine the relationship between the macro- and microparameters of the sample, the XRD experiment is obviously not enough. Because the analysis of mineral composition or XRD experimental study only shows the variation of parameters concerning temperature from a qualitative point of view, which does not quantitatively reflect the correlation between macroparameters and microparameters. To quantitatively explain the variation of fracture toughness of sandstone concerning temperature, mercury injection experiments are also needed for sandstone.

The variation of average pore size and porosity of sandstone with respect to temperature is shown in Table 6 and Figure 9. According to the experimental results, it is implied that the pore structures gradually transit from dense to porous and fluffy. For example, when the temperature is 100°C, the average pore size and porosity of sandstone slightly decrease compared with that at 20°C. When the temperature is between 100°C and 700°C, these parameters of sandstone gradually increase with the increase in temperature, which indicates that the micropores in sandstone gradually change to mesopores and macropores. On the whole, the variation of these parameters with respect to temperature can be divided into three stages: slow increase stage (20°C–100°C), slow decline stage (100°C–400°C), and sharp decline stage (400°C–700°C). Besides, it should be pointed out that the variational trend of microparameters is basically consistent with that of macroparameters (i.e., fracture toughness), which give strong support to the correlation of macroscopic and microscopic parameters.

4.2. Correlation of Micro-Macrofracture Mode. Based on the load-displacement curves, it can be obtained that when the temperature is between 20°C and 400°C, the fracture mode of the sandstone is mainly characterized by brittle fracture. When the temperature is between 400°C and 700°C, the fracture mode of the sandstone begins to show some ductility. The macroscopic fracture mode and fracture trajectory of the rock are fundamentally determined by the microscopic fracture mechanism. Therefore, the SEM observation of the fracture surface is necessary.

On the whole, the fracture mode of the specimen at a relatively low temperature (20°C–400°C) is mainly characterized by transgranular fracture. With the increase in temperature, the mineral particles gradually become loose, and the opening degree of the interfaces between the crystals gradually becomes more considerable. Moreover, due to the influence of thermal stress and expansion force, a series of complicated cracks are randomly distributed in the samples. As a result, the fracture mode of specimens become more complex and diversified. The fracture mode of the specimen at this temperature stage (400°C–700°C) is mainly characterized by intergranular fracture. This is also consistent with previous scholars' research results. It is worth noting that cracks tend to propagate and evolve along with the interface of crystals under these conditions. Besides, the fractures will also coalesce with the thermal cracks under loading. Hence, the roughness of the fracture surface at high temperature is more significant than that at low temperature. Therefore, the fracture trajectory of the sample becomes more tortuous, and the fractal dimension gradually increases.

Interestingly, however, there are some differences in the fracture morphology of the specimens at real-time high temperatures, which have not been found in the specimens after treatment with high temperatures (the results of Feng et al. [17]); that is, some cracks appear on the fracture surface at 400°C and the density, number, and aperture of the fractures increase with the increasing temperature, as shown in Figure 10. This is because some of the crystals in the specimens are molten under real-time high-temperature conditions, and the fracture behaviors are similar to those of soft rock, so the sandstone tends to deform plastically. Plastic deformation can cause tearing cracks in the fracture surfaces of a specimen.

Table 5: The mineral composition and mass content of sandstone at various temperatures.

Temperature (°C)	Mineral composition	Chemical formula	Mass content (%)
25	Quartz	SiO_2	67
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	9
	Albite	$NaAlSi_3O_8$	21
	Anorthite	$CaAl_2Si_2O_8$	3
100	Quartz	SiO_2	67
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	9
	Albite	$NaAlSi_3O_8$	21
	Anorthite	$CaAl_2Si_2O_8$	3
200	Quartz	SiO_2	52
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	14
	Albite	$NaAlSi_3O_8$	29
	Anorthite	$CaAl_2Si_2O_8$	5
300	Quartz	SiO_2	47
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	27
	Anorthite	$CaAl_2Si_2O_8$	19
	Laumontite	$Ca\ [AlSi_2O_6]_2*4H_2O$	7
400	Quartz	SiO_2	57
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	23
	Albite	$NaAlSi_3O_8$	15
	Anorthite	$CaAl_2Si_2O_8$	3
	Anorthite, Na-rich	$(Ca,Na)\ (Si,Al)_4O_8$	2
500	Quartz	SiO_2	53
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	11
	Albite	$NaAlSi_3O_8$	26
	Anorthite	$CaAl_2Si_2O_8$	6
	Anorthite, Na-rich	$(Ca,Na)\ (Si,Al)_4O_8$	4
600	Quartz	SiO_2	67
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	9
	Albite	$NaAlSi_3O_8$	18
	Anorthite	$CaAl_2Si_2O_8$	4
	Anorthite, Na-rich	$(Ca,Na)\ (Si,Al)_4O_8$	2
700	Quartz	SiO_2	75
	Albite, Ca-rich	$(Na,Ca)Al\ (Si,Al)_3O_8$	6
	Albite	$NaAlSi_3O_8$	14
	Anorthite	$CaAl_2Si_2O_8$	3
	Anorthite, Na-rich	$(Ca,Na)\ (Si,Al)_4O_8$	2

Table 6: Pore structure parameters of sandstone at various temperatures.

Temperature (°C)	Average pore size (nm)	Porosity (%)
20	38	3.7
100	26	3.5
200	48	4.2
300	81	5.3
400	176	8.6
500	284	13.5
600	372	18.2
700	449	24.8

4.3. Weakening Mechanism of Sandstone at Real-Time High Temperature. According to the variation of fracture toughness, fractal dimension, and surface morphology of sandstone with temperature, the weakening mechanism of sandstone at a real-time high temperature can be divided into 3 stages.

Figure 9: The relationship between average pore size and porosity of sandstone and temperature.

Figure 10: Fracture morphology of sandstone at different temperatures (SEM images): (a) 20°C; (b) 100°C; (c) 200°C; (d) 300°C; (e) 400°C; (f) 500°C; (g) 600°C.

Stage I (25°C–100°C): at this stage, the fracture toughness of sandstone increases slightly with the increase in temperature. This is because the evaporation of free water and the closure of original pores and cracks occur at this temperature stage. According to the principle of effective stress, the evaporation of water leads to a decrease in pore pressure and an increase in effective stress, which causes the rock mass to be squeezed, and finally, the rock strength is increased. In addition, heating causes thermal stress inside the rock. When the temperature is relatively low, the thermal

stress can only cause recoverable elastic deformation, which will not cause thermal cracking in the samples. Moreover, thermal expansion at relatively low temperatures will result in the closure of existing microcracks and inhibit the coalescence and interaction of existing cracks, which will lead to the enhancement of the overall structural strength of sandstone. Undoubtedly, the increase in rock strength will lead to an increase in resistance to intergranular and transgranular stress corrosion cracking and ultimately lead to an increase in fracture toughness. Since the transgranular fracture occurs mainly at this temperature stage, the deviation of transgranular fracture from the center line tends to be smaller compared with intergranular fracture, thus producing a relatively straight fracture trajectory. In summary, due to the volatilization of water and the closure of the primary pores, the fracture toughness of the specimens increased. However, since the thermal stress is small, the growth rate of the fracture toughness is relatively tiny.

Stage II (100°C–400°C): at this stage, the fracture toughness slowly decreases with the increase in temperature. On the contrary, as the temperature increases, the evaporation of water in the rock becomes more and more severe. Not only the bound water is fully evaporated but also the strong combined water and structural water are gradually lost. Dehydration has a direct effect on the structure of the rock. For example, a type I tensile crack occurs during the dehydration of serpentine (Jung et al. [54]). According to the results of MIP, the microstructures of sandstone change significantly at this stage compared with that at the first stage. Therefore, fracture toughness gradually decreases with the increase in temperature.

Stage III (400°C–700°C): there are three reasons for the weakening mechanism of specimens at this stage. The first reason is the decomposition of some minerals, the recrystallization of some minerals, and the formation of new minerals. In this way, the decomposition of minerals will produce a range of cracks with a broad distribution of dimensions (length and width), orientations, and shapes, which will result in a higher porosity and fractal dimension. Besides, according to the study of Yu et al. [55] and Liu et al. [56], crystal recrystallization facilitates crack propagation in the grain-boundary regions by enabling intergranular fracture and decohesion between mineral grains; secondly, rock is a heterogeneous body composed of different minerals, and the thermal expansion coefficients of each mineral are different at high temperature [57–60]. The inconsistency of expansion and deformation of mineral grains under relatively high temperature will cause high thermal stress in the mineral structure. The maximum thermal stress often occurs at the junction or boundary of mineral particles. If the stress reaches or exceeds the critical strength of rock, grain-boundary microcracking may occur. According to the results of Feng et al. [17], thermal cracking will occur at 400°C; thirdly, quartz in sandstone will transit from the α phase to β phase (Grimm and Dorner [61]) at 573°C. In the quartz lattice structure, two Si-O tetrahedra are connected at an angle of 150°. When the quartz sand is heated to 573°C, the two Si-O tetrahedra in the crystal structure are connected at 180°. Due to the α-β phase transformation of quartz, the volume of quartz changes drastically, resulting in an increase in internal microcracks, which ultimately leads to changes in the microstructure and macroscopic physical and mechanical properties of the rock.

5. Conclusion

In this paper, the macroscopic and microscopic fracture behaviors of red sandstone under real-time high temperature are studied experimentally. The main conclusions are as follows:

(1) With the increase in temperature, the value of fracture toughness of sandstone first increases and then decreases; the value of fracture toughness reaches the maximum at 100°C. According to the relationships between fracture toughness and temperature (i.e., curve slope), the variational trend of the curve can be divided into 3 stages: (1) the slow increase stage (20°C–100°C). The curve slope at this stage is 0.0055. (2) The slow descent stage (100°C–400°C). The curve slope at this stage is 0.0048. (3) The rapid descending stage (400°C–700°C), in which the slope of the curve is 0.0087.

(2) The fractal dimension and the fracture deviation from the hypothetical fracture originating at the notch tip of sandstone specimens gradually increases with the increase in temperature. And, the values are smaller at relatively low temperatures (20°C–400°C) but larger at relatively high temperatures (400°C–700°C).

(3) The higher the temperature is, the more complex the chemical reactions occur in the crystals of sandstone. When the temperature of sandstone rises from room temperature to 700°C, quartz decomposition and synthesis, the transition between quartz crystals, dehydration of feldspar, and synthesis and decomposition of feldspar will occur. And, the average pore size, total volume, and porosity of sandstone first decrease and then increase with the increase in temperature. When the temperature is between 100°C and 700°C, these parameters of sandstone gradually increase with the increase in temperature, which indicates that the micropores gradually change to mesopores and macropores. On the whole, the variational trend of mesoparameters is basically consistent with that of fracture toughness.

(4) There are some differences in the fracture morphology of the specimens at real-time high temperatures, which have not been found in the specimens after treatment with high temperatures. Some cracks appear on the fracture surface at 400°C and the density, number, and aperture of the fractures increase with the increasing temperature. Some of the crystals in sandstone are molten under real-time high-temperature conditions, and the mechanical properties are similar to those of soft rock, so the sandstone tends to deform plastically. Plastic deformation can cause tearing cracks in the fracture surfaces of the specimen.

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Study on Cumulative Damage Law of Stainless Steel-Reinforced Concrete Columns under Step Impact Loading

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In this study, an ultrahigh drop hammer impact test system was adopted for multiple horizontal impact tests on stainless steel-reinforced concrete columns and ordinary-reinforced concrete columns with the same longitudinal reinforcement diameter. The damage performance after impact was studied, and the finite element model was established. The test measured the impact force, displacement, cracking of the specimen during the impact, and the concrete damage near the bottom of the specimen. The test results showed that the failure mode of the stainless steel-reinforced concrete specimen under multiple impacts was the same as that of the ordinary reinforced concrete specimen. Under the same impact conditions, the maximum impact force, the maximum displacement, and the damage degree of stainless steel-reinforced concrete column specimen were lower than those of the ordinary reinforced concrete specimen.

1. Introduction

Compared with ordinary steel bars, stainless steel bars have better advantages in strength, ductility, corrosion resistance, and maintenance cost. A large number of stainless steel bars were used in the Hong Kong-Zhuhai-Macao Bridge that has been opened to traffic. It is a development trend to replace ordinary reinforced concrete structures with stainless steel bars. Reinforced concrete columns are one of the most important components in bridges and building structures. Damage or damage under impact loads may cause partial or total collapse of the structure, causing catastrophic consequences.

Tian and Zhu [1] established a damage assessment method for members based on the vertical residual load capacity of reinforced concrete columns after impact, which was used to determine the damage degree of reinforced concrete columns under impact loads and analysis of the protective measures for column protection. Cheng et al. [2] conducted a numerical study on the dynamic response of reinforced concrete columns under impact and analyzed the effects of collision velocity, mass, collision shape, reinforcement ratio, and concrete strength grade on deformation and impact force. Loedolff [3] conducted a horizontal impact test on the impact resistance of reinforced concrete cantilever columns. Liu et al. [4] studied the failure mode and transformation mechanism of axially compressed reinforced concrete columns under lateral impact loads using the finite element method. Demartino et al. [5] and Cai et al. [6] carried out an experimental study on the impact resistance of scaled reinforced concrete bridge piers. Zhou et al. [7] established a reliability analysis method for shear performance of reinforced concrete piers under falling rock impact load and discussed the influence of different parameters on the shear performance of columns. Wang et al. [8] studied the dynamic response of simply supported RC columns with axial compression under lateral impact loads and analyzed the effect of loading rate, longitudinal

reinforcement ratio, and stirrup reinforcement ratio on the failure mode and carrying capacity. Zhao and Qian [9] used numerical simulation to study the performance of reinforced concrete piers under lateral impact loads, to analyze the dynamic response and impact resistance of the piers under lateral impact loads, and to discuss the effects of different collision parameters on the failure mode and the damage mechanism of reinforced concrete piers. Zhang et al. conducted an experimental study on the impact resistance of stainless steel-reinforced concrete [10, 11] and used ultrasonic waves to simple analyze the damage of the pier specimens after impact [12].

In the present study, based on the application of stainless steel-reinforced concrete piers of the Hong Kong-Zhuhai-Macao Bridge, an ultrahigh multifunctional drop hammer impact test system was applied for the experimental research of the cumulative damage of stainless steel-reinforced concrete columns under the horizontal impact loads. Also, the ordinary reinforced concrete specimens with the same longitudinal reinforcement diameter were compared and analyzed. The results have reference significance for the impact analysis and design of stainless steel-reinforced concrete columns.

2. Experimental Programme

2.1. Design of the Specimen. In this experimental study, two cylindrical column specimens were designed, one of which was a stainless steel-reinforced concrete column numbered R1 and the other which was an ordinary reinforced concrete column numbered R2. The specimens were made of C40 concrete, with column section diameters of 340 mm and a column height of 2,200 mm. The experimental stainless steel bars were an S2304 duplex stainless steel reinforcement, and the longitudinal reinforcement used in the ordinary RC bridge pier specimen was HRB335 reinforcement. The design parameters of the test specimens are shown in Table 1, the properties of the steel materials are shown in Table 2, and the detail drawing of reinforced concrete is shown in Figure 1.

2.2. Test Device. This study's experiment was conducted using an ultrahigh drop hammer impact test system. By adjusting the vertical height of the drop hammer, the free fall of the drop hammer could be used as the traction of a car with different motion rate levels. The mass of the drop hammer was 196 kg, the impacting car body was 1.2 t, the bottom of the column test specimen was fixed with high-strength bolts, and the vertical shaft pressure of 250 kN was applied on the top of the test specimen, as shown in Figure 2.

During the test, the impact force, displacement, and crack development of the test specimens were collected and recorded. Four displacement meters were arranged at the back of the specimen from top to bottom at 100 mm, 800 mm, 1,500 mm, and 2,100 mm, as shown in Figure 3, and numbered as #1, #2, #3, and #4. In order to obtain the damage degrees of the specimens, a ZBL-U520 nonmetal ultrasonic detector was used to test the specimens and the main detection parts were the bottoms of the specimens.

2.3. Model Test. The drop hammer was lifted to a certain height and then released, in order to track the car's impact on the specimen. The impact velocity was obtained using a laser velocity measuring system set at the end of the track, and the impact force was measured using a pressure sensor set at the impacting car's head. In order to record the crack development during the impact of the specimen, the crack width of the specimen was measured by a crack width measuring instrument. The data acquisition system uses the high performance dynamic signal analysis system DH8302 of Jiangsu Donghua Test Technology Co., Ltd.

3. Results and Discussion

3.1. Time-History Curve for the Impact Force. Figure 4 shows the time-history curve of the impact force of the stainless steel-reinforced concrete specimens with the drop hammers lifted to 2 m, 4 m, 6 m, and 8 m, respectively. As the impact energy continues to increase, the maximum impact force also increases. The maximum impact force under the impact of 2 m, 4 m, 6 m, and 8 m was 130264 N, 196819 N, 236707 N, and 269763 N, respectively. It can be seen from Figure 5 that under the same impact conditions, the maximum impact force of ordinary reinforced concrete specimens was 142136 N, 215166 N, 285176 N, and 317289 N, respectively. By comparing the stainless steel-reinforced concrete specimens with ordinary reinforced concrete specimens under the same impact conditions, it was found that the impact force peak of stainless steel-reinforced concrete specimens under the same impact condition was significantly lower than that of ordinary reinforced concrete column specimens, and the shock during the impact of the ordinary reinforced concrete specimen was more complicated than the stainless steel-reinforced concrete specimen. This was mainly due to some characteristics of the stainless steel reinforcement itself.

3.2. Research on Damage Degree Based on the Damage Factor. In order to obtain an accurate description of the damage degree of the specimens, the damage degree of the specimens are quantitatively described based on the ultrasonic wave velocity, and the damage factor D is introduced.

The damage factor is defined as follows:

$$D = 1 - \frac{v^2}{v_0^2}, \tag{1}$$

where v is the wave velocity after damage and v_0 is the initial wave velocity.

From the change of the ultrasonic wave velocity before and after the impact, the damage factor of each specimen can be obtained according to the aforementioned formula. The change of the damage factor after each impact is shown in Table 3.

It can be seen from the table that after each impact, the damage degree of the column specimens will increase and the ultrasonic wave velocity of the specimens will reduce after the damage. With each impact, the damage factor was changing; that is, after each impact, the degree of damage to the specimens was increasing. In order to more intuitively

Table 1: Specimen design parameters.

Specimen no.	Longitudinal reinforcement	Stirrup	Concrete strength grade
R1	10S16	Φ8@50	C40
R2	10Φ16	Φ8@50	C40

S represents the stainless steel reinforcement.

Table 2: Steel reinforcement material performance.

Reinforcement category	Yield strength (MPa)	Tensile strength (MPa)	Elongation (%)	Elastic modulus (MPa)
Reinforcement HRB335	368	552	16	2.0×10^5
Stainless steel reinforcement S2304	748	866	17	1.93×10^5

Figure 1: Detail drawing of reinforced concrete (unit: mm).

obtain the relationship between the damage degree of the specimens and the damage factor D after each impact, the relationship between the damage factor and the impact after the cumulative damage condition is shown in Figure 6.

It can be seen from Figure 6 that the specimens were regarded as nondamaged before impact. After each impact, the damage factor was increased. Under the same impact condition, the damage factor of the stainless steel-reinforced concrete specimen was smaller than the ordinary reinforced concrete specimen, indicating that the damage degree of the stainless steel-reinforced concrete specimen under the same impact condition was lower than that of the ordinary reinforced concrete specimen.

3.3. Crack Description and Analysis. Each of the specimen's crack initially developed from the bottom of the positively impacted specimen. With the increase in the impact energy, the crack propagated from the bottom of the positive impact to the center. Meanwhile, the crack correspondingly occurred to the impact back and point. As the impact energy continued to increase, the original crack at the midlower section of the front continued to develop toward the center. At the same time, the crack located on the reverse side continued to develop upwards. When the impact energy reached a certain level, a larger crack occurred at the bottom of the impact front, and the concrete on the impact back began to partially crush. The crack development areas were quite consistent with the main force areas of the specimens. The following table shows the width development of a major crack at the lower section of the impact front for each of the specimens with the changes in the drop hammer's height.

It can be seen from the aforementioned Table 4 that with the increase in the impact energy the width of this major

Figure 2: Ultrahigh drop hammer impact test system diagram.

(a) (b)

Figure 3: Arrangement diagram for displacement of specimen (unit: mm).

crack also increased. However, under the same impact condition, the development width of this major crack of the stainless steel-reinforced concrete specimen was lower than the crack width of the ordinary reinforced concrete specimen.

3.4. Displacement Time-History Curve. Figures 7 and 8 show the time-history curves of the four measuring points of the drop hammer when it was lifted to 8m. The displacements of the four measuring points were synchronously changed. The four displacement values of the R1 specimen were 17.3 mm, 11.3 mm, 4.9 mm, and 1.2 mm, respectively, and the four displacement values of the R2 specimen were 19.1 mm, 12.4 mm, 5.2 mm, and 1.7 mm, respectively. As it can be seen from the aforementioned data, the displacements of stainless steel-reinforced concrete specimen are lower than those of the ordinary reinforced concrete specimen under the same condition.

Figure 4: Impact force history of the R1.

Figure 5: Impact force history of the R2.

Table 3: Damage factor changes under continuous impact.

Specimen	Numbers of impacts	Acoustic velocity average (km/s)		Damage factor	Impact height
		Before impact	After impact		
R1	1	4.167	4.156	0.0053	2 m
	2	4.156	4.127	0.0139	4 m
	3	4.127	4.051	0.0365	6 m
	4	4.051	4.018	0.0162	8 m
R2	1	4.258	4.243	0.0070	2 m
	2	4.243	4.202	0.0192	4 m
	3	4.202	4.065	0.0641	6 m
	4	4.065	4.024	0.0201	8 m

3.5. Stress Nephogram. The finite element model of the specimen being impacted was established by using LS-DYNA. Figures 9 and 10 show the numerical simulation stress diagrams of the two specimens under the impact energy of the drop hammer to 8 m. The figures show the impact of the impact body on the left side of the specimen.

Figure 6: Damage factor change curve after each impact.

Table 4: Crack development under different impact energies.

Specimen	Crack widths at each drop hammer height (mm)			
	2 m	4 m	6 m	8 m
R1	0	0.230	0.294	0.318
R2	0	0.298	0.422	0.554

Figure 7: Displacement history of R1-8 m.

Figure 8: Displacement history of R2-8 m.

It can be seen from the stress diagram that before the impact, since the vertical load has been applied on the top of the specimens, the stress value at the top of the specimen was large. When the impact body has just come into contact with the specimens, the stress is greater near the position of the impact point and at the bottom of the reverse side. As the impact continues, the stressed part gradually increases mainly at the bottom of the back, and then the area of the back bottom and the area of the impact point develop toward both sides. From the whole impact process, the main stress change areas of the two specimens were concentrated near the bottom of the impact back and the impact point, and the stress changes at the same position were about the same. When the specimen impacted the bottom stress to the maximum, the specimen displacement also reached the maximum. However, it can be seen from the figure that at the same time, the maximum stress of the stainless steel-reinforced concrete specimen was smaller than that of the ordinary reinforced concrete specimen.

(a) (b) (c) (d)

Figure 9: Stress nephogram of R1-8 m (Unit: Pa): (a) R1-8 m ($T = 0.030$ s), (b) R1-8 m ($T = 0.035$ s), (c) R1-8 m ($T = 0.062$ s), and (d) R1-8 m ($T = 0.089$ s).

(a) (b) (c) (d)

Figure 10: Stress nephogram of R2-8 m (Unit: Pa): (a) R2-8 m ($T = 0.030$ s), (b) R2-8 m ($T = 0.035$ s), (c) R2-8 m ($T = 0.062$ s), and (d) R2-8 m ($T = 0.089$ s).

4. Conclusions

In this paper, multiple horizontal impact tests were carried out on the stainless steel-reinforced concrete column specimen and the ordinary reinforced concrete column specimen with the same longitudinal reinforcement diameter and established the finite element model. The following conclusions were obtained by analyzing the impact time history, displacement time history, concrete damage variation, and crack recording results of reinforced concrete column specimens under step impact loading.

(1) With the increase of impact energy, the impact peak gradually increased. By comparing with the ordinary reinforced concrete specimen, the maximum impact force and displacement peak of the stainless steel-reinforced concrete specimen are smaller under the same impact conditions.

(2) Based on the ultrasonic wave velocity and introducing the damage factor, the damage degree of the specimen was quantitatively described and compared with the ordinary reinforced concrete specimen with the same longitudinal reinforcement diameter, the damage degree of the stainless steel-reinforced concrete specimen was lower.

(3) The crack of the specimens start from the bottom area of the front surface of the impact. As the impact energy increases, it gradually extended toward the middle direction. At the same time, the crack occurred at the opposite side of the impact surface and the impact point and gradually developed toward the upper and middle parts of the back surface. When the impact energy reached a certain level, a large crack appeared at the bottom of the impact front, and the concrete at the bottom of the impact was crushed,

which was consistent with the finite element simulation results.

Data Availability

The data sets used and analyzed during the current study are available from the corresponding author on reasonable request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Hydrogen Effect on Cumulation of Failure, Mechanical Properties, and Fracture Toughness of Ni-Cr Alloys

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Influence of hydrogen pressure and internal hydrogen contents on short-term strength, plasticity, and plane-stress fracture toughness of 05Cr19Ni55 alloys at pressure up to 30 MPa was investigated. It was established that the crack resistance parameters K_c of alloys decrease with displacement rates decreasing similar to elongation (δ) and reduction of area (ψ) of smooth specimens. The maximum hydrogen influence is achieved at strain rate speeds less than 0.1 mm/min and hydrogen pressures above 15 MPa when δ and K_c of prehydrogenated samples are reduced by 3 times. The plane-strain conditions required for the evaluation of K_{Ic} were fulfilled on compact tension 05Cr19Ni55 alloy specimens with thickness above 20 mm under hydrogen pressure 30 MPa and preliminary dissolved hydrogen concentration above 20 wppm. Regardless of the test conditions, the value of the characteristics of plasticity (δ, ψ) and fracture toughness (K_c) of alloy (TO1) specimens oriented in the transverse direction (orientation TV) is significantly lower than that of specimens oriented in the longitudinal direction (LT). Alloy cleaning by vacuum arc remelting and optimization of heat treatment regime increase their hydrogen resistance.

1. Introduction

Heat-resistant nickel alloys are widely used in energy and aerospace engineering in contact with high-pressure hydrogen-containing gas mixtures [1–5]. Therefore, one of the most important requirements for such alloys is their resistance to hydrogen degradation. However, it is known that precipitation-hardening alloys are sensitive to hydrogen embrittlement [1–5]. So, for correct estimation of their workability in hydrogen environment, it is important to determine the effect of the structure and parameters of the load on the mechanical properties.

In order to ensure the safety and reliability of materials for a steam and gas turbine, structural integrity and lifetime prediction are of great importance. Working structures and their elements (blades and discs) are subjected to the influence of various loads and hydrogen-containing gas. In order to ensure an adequate level of safety and optimal durability of such structural elements, experimental tests in gaseous hydrogen are required to determine the effect of various factors. By the fracture mechanics approaches, evaluation of static load durability of structures and critical crack (defect) size in technological or operational circumstances have been calculated as the values of critical stress intensity factor K_{Ic} (K_c) [1, 3–6]. Use of this option to evaluate workability details in gaseous hydrogen is limited to a number of factors caused by the methodological aspects of the determination in laboratory conditions and characteristics of initiation and growth of cracks in the presence of hydrogen [3–9]. The materials employed in hydrogen energy should have little sensitivity to cracks and cuts that have a high ability to relax tension in the areas of highest concentration (high plasticity), which does not archive the plane-strain state for research specimens and determines KIc by standard methods in air [5–8]. Therefore, most studies of the impact of hydrogen on mechanical behavior of structural steels and alloys were carried out with tests on short-term static tensile and low-cycle durability [1–3, 10, 11]. The

influence of load conditions (speed for static tension, frequency and amplitude by low-cycle fatigue, temperature, pressure, and dissolved hydrogen content in advance) on strength, ductility, and durability of materials of different structural classes was established [1–4, 11]. However, literature data on regularities of influence of hydrogen on the static crack resistance are limited and controversial [1–5]. The issue of the effect of sample orientation on the sensitivity of materials to the action of hydrogen remains poorly understood. The influence of texture components on hydrogen-induced cracking of pipeline steel has been studied in [12, 13], and similar data on heat-resistant materials are limited and ambiguous [14, 15].

In what follows, we study the influence of high-pressure gaseous hydrogen on short-term strength, plasticity, and static crack resistance of a nickel-based alloy 05Cr19Ni55 with different modes of heat treatment and chemical composition variations.

2. Materials and Experimental Procedure

The alloy 05Cr19Ni55 from which parts used in energy, petrochemical engineering, and aerospace engineering products [16, 17] are manufactured for use in the hydrogen environment has been investigated. Chemical composition of investigated alloys is shown in Table 1. Alloying with niobium, vanadium, titanium, aluminum, and boron leads to the formation of carbides $(Ni, Fe, Cr)_{23}C_6$, borides Me_3B_2, and intermetallics $Ni_3(Al, Mo, Nb)$ in the amounts up to 10%, which substantially increases their high-temperature strength and significantly influences the sensitivity to the action of hydrogen [1–3, 9, 11]. Presented in Table 2 and in Figures 1 and 2 are the results obtained in the first version of the chemical composition (CC).

The orientation of specimens, method of melting, heat treatment regimes, grain sizes, thickness of the specimens for fracture toughness testing, the mechanical properties of alloy in the air and in hydrogen under the pressure of 30 MPa after preliminary hydrogenation (623 K, 30 MPa H_2, 10 h), and coefficients of influence of hydrogen on reduction of area and K_{Ic} are shown in Table 2.

Static tensile tests were carried out on standard fivefold cylindrical specimens with a diameter of working part 5 mm using the strain-rate range $V = 0.01$–100 mm/min. Stress intensity factor under static loading K_c was computed either for the maximum force F_C in the "F-V" linear diagram or for the force F_Q determined by using 5% secant for nonlinear diagrams. Rectangular compact specimens were tested for eccentric tension under the pressures up to 30 MPa at a strain rate of 0.01–100 mm/min. The values of K_c are calculated from the Srawley–Gross formula [18]. Plastic properties of the alloy in the air are so high (see Table 2) that plane-strain conditions on samples with a thickness of 20–25 mm are not implemented. Therefore, plane-stress fracture toughness was determined by the method of J-integral [19]. Despite some reservations, this method is often used to evaluate static fracture toughness of plastic materials and for small thicknesses of parts of structures [1, 6, 7]. For tests, we have used the 50x48x20 mm specimens with previously groved at 293 K fatigue cracks.

Specimens were hydrogenated in a working chamber at 623 K and hydrogen pressure 30 MPa during 1–10 h. The maximum hydrogen content in the specimens varied from 21 wppm to 27 wppm depending on the structural state of the alloy (see Table 2). Hydrogen concentration (C_H) was determined with a LECO TCH 600 instrument [20] with precise 0.1 wppm (3 specimens for point). To determine the indicated mechanical characteristics in hydrogen, the working chambers were preliminarily evacuated, blown-out with hydrogen, again evacuated, and filled up with hydrogen to a given pressure. All tests were performed at room temperature.

3. Results and Discussion

3.1. Influence of Loading Rate on the Cumulation of Failure and Properties in Hydrogen. The behavior of the rate dependences of hydrogen degradation is determined by the kinetics of penetration of hydrogen into the material and changes in the mechanisms of deformation. As the rate of short-term tension decreases, the influence of hydrogen first increases, and then the properties become stable (see Figure 3). Under a constant hydrogen pressure and content, the rate interval and the degree of maximum embrittlement are determined by the chemical composition and the structural state of the material. Substantial reductions in ductility characteristics of the alloy with CC I (see Table 1) after HT III (see Table 2) were found in the range of rates 0.1–10 mm/min and with CC II after HT II in the range 0.1–1 mm/min (see Figure 3, curves 1, 4 and 3, 6).

The maximum embrittlement of hydrogenated material in hydrogen under a pressure of 30 MPa is realized at a rate less than 0.1 mm/min ($0.67 \cdot 10^{-4} \cdot s^{-1}$), which is three orders of magnitude smaller than the regulated rate of determination of the hydrogen resistance of steels with a body-centered cubic structure equal to 10 mm/min [4–7, 11, 12]. This ratio of the rates correlates with the ratio of the diffusion coefficients of hydrogen in alloys with face-centered cubic and body-centered cubic lattices given in [1, 4–7, 11, 21, 22], which testifies to the penetration of hydrogen into the metal in the process of static tension.

Similarly, the level of K_c decreases with decreasing dK/dt to the minimum value, defined of material structural condition (see Figure 1). The widest rate range and greatest sensitivity to the action of hydrogen typical for longitudinal samples after open melting HT1 with a coarse-grained structure (see Figure 1, curve 1). Maximum embrittlement of these specimens and specimens with grain size 36 microns are achieved by loading rate 0.18 $MPa \cdot m^{1/2}/s$ (see Figure 1, curves 1 and 8), a grain size of 21 microns, and the largest fracture toughness in the air -0.054 $MPa \cdot m^{1/2}/s$ (see Figure 1). At the same time, if in air with increasing thickness of the sample, parameter K_c slightly increases, then in hydrogen, it decreases (see Table 2, positions 5, 6, 7, and 8); i.e., the embrittlement effect of hydrogen on the fracture toughness of the alloy increases.

Table 1: Chemical composition of 05Cr19Ni55 alloy (mass %).

CC	C	Fe	Cr	Mo	Nb	Al	N	∑Cr, Mo, Nb, Fe	∑ N + C	(Cr, Ni, Fe)$_{23}$(C, N)$_6$
I	0.05	12.0	19.0	8.87	1.73	1.49	0.07	41.60	0.12	3.1
II	0.04	11.0	17.5	8.97	1.84	1.54	0.03	39.38	0.07	2.0

Table 2: Characteristics of the 05Cr19Ni55 alloy modifications at 293 K and displacement rate 0.1 mm/min.

No. of order	Orientation	Heat treatment	d (μm)	t (mm)	K_{1c} (J_c) (MPa·m$^{1/2}$)	C_H (wppm)	σ_u MPa	σ_{ys} MPa	δ %	ψ %	K_c (MPa·m$^{1/2}$)	β_ψ βK_c
1	TV	OM. Quenching (Q): 1323 K, 1 h Aging (A): 1023 K, 15 h, +923 K, 10 h (HT I)	85	20	86	25	920/ 930	570/ 570	10/ 2.5	12/ 4.5	69/28	0.38/ 0.32
2	LT	OM, HT I	85	20	107	25	1020/ 970	590/ 600	14/ 4.5	18/ 8	88/38.5	0.44/ 0.36
3	LT	VAR, HT I	85	20	128	22	1050/ 1010	620/ 610	20/ 6	32/ 15	98/51	0.47/ 0.40
4	LT	OM. Q: 1253 K, 1 h. A: 1023 K, 15H, +923, 10 h (HT II)	21	20	119	27	1040/ 990	610/ 610	17/ 5	20/ 9	94/45	0.45/ 0.38
5	LT	VAR, HT II	21	20	162	24	1080/ 970	650/ 660	35/ 6	38/ 19	116/68	0.50/ 0.42
6	LT	VAR, HT II	21	25	162	24	1080/ 970	650/ 660	35/ 6	38/ 19	126/64	0.50/ 0.40
7	LT	VAR, after HT II simulation of soldering 1283 K, 15 min, A: 973 K, 10 h (HT III)	36	10	138	21	1180/ 960	710/ 690	30/ 5	34/ 16	103/62	0.47/ 0.45
8	LT	VAR, HT III	36	20	138	21	1180/ 960	710/ 690	30/ 5	34/ 15	108/52	0.44/ 0.38
9	LT	VAR, after HT II simulation of soldering 1473 K, 15 min + 1273 K, 1 h (HT IV)	130	20	—	23	750/ 700	380/ 390	41/ 16	31/ 18	—	0.58/ —

TV: transverse orientation; LT: longitudinal orientation; OM: open melting; VAR: vacuum arc remelting; d: grain size; t: thickness of the specimen. $\beta_\psi = \psi_{hydr}/\psi_{air}$; $\beta_{KC} = K_{c\ hydr.}/K_{Ic(Jc)}$ air.

Figure 1: The effect of rising K rate dK/dt on fracture toughness K_c in hydrogen under pressure 30 MPa of hydrogenated specimens of 20 mm thickness. The numbers near to curves correspond to ordinal numbers in Table 2.

Figure 2: Dependences of fracture toughness K_c of prehydrogenated (623 K, 30 MPa H_2, 10 h) specimens of 20 mm thickness on the hydrogen pressure P at a strain rate of 0.1 mm/min. The numbers near curves correspond to ordinal numbers in Table 2.

3.2. Influence of Hydrogen Pressure on Fracture Toughness. The dependences of the fracture toughness of prehydrogenated specimens on the hydrogen pressure consist of two regions. In the first region (low pressures), the pressure abruptly drops, and in the second, the negative action of hydrogen becomes stable (see Figure 2). This means that there exists a pressure under which degradation of the material with hydrogen reaches its maximum. In our case, this effect is found for pressures more than 15 MPa. Note that the properties of nonhydrogenated specimens of the

Figure 3: Dependences of plasticity characteristics δ (1–3) and ψ (4–6) of prehydrogenated (623 K, 30 MPa H_2, 10 h) specimens in hydrogen under pressure 30 MPa on the strain rate: 1, 4: CC (I) VAR, HT III; 2, 5: CC (I) VAR, HT II; 3, 6: CC II, VAR, HT II.

Figure 4: Dependences of fracture toughness K_c at hydrogen pressure of 30 MPa of specimens of 10 (7), 20 (5, 8) and 25 (6) mm thickness on the hydrogen content C_H at a strain rate of 0.1 mm/min. The numbers near curves correspond to ordinal numbers in Table 2.

alloy 05Cr19Ni55 do not reach the minimum values in the pressure range 0–35 MPa [11].

In hydrogen under the pressure 30 MPa, K_c of the alloy (CC I after HT II and HT III) of the thickness 20 mm decreases to 68 MPa·m$^{1/2}$ and 57 MPa·m$^{1/2}$ at the hydrogen concentration of 19 and 18 wppm, respectively, and does not change with further hydrogenation (see Figure 4, curves 5 and 8). Under these conditions, the diagrams "load–displacement" are linear with a sharp maximum of the load and fracture surface of specimens covered cleavage facet, typical for brittle fracture. K_c value reaches the critical value K_{1c} that satisfies the condition l, $b \geq 2,5\ (K_c^H/\sigma_{ys}^H)^2$, where l is the crack length and b is the specimen thickness [18, 19]. Thus, hydrogen initiates the destruction of the mechanism of normal separation across the whole crack front, causing formation of the plane-strain state in the specimen. Conditions of the plane-strain state on all modifications of the alloy are performed by the specimen thickness 20 mm with the exception of positions 5 and 7 in Table 2. K_{IcH} alloy with CC I after VAR, HT II (position 5) is equal to 64 MPa·m$^{1/2}$ and achieved on specimens of thickness 25 mm (see Figure 4, curve 6). For a specimen thickness of 10 mm, plane-strain conditions for the alloy (CC I after HT III) did not perform even after the hydrogenation to the $C_H = 21$ wppm (see Figure 4, curve 7). Thus, the maximum effect of hydrogen on fracture toughness of specimens with lower thickness is achieved by more intensive modes of action (hydrogen pressure or content) or slower loading rates (see Figure 1), when conditions are provided to allow sufficient hydrogen to enter the crack tip. At the same time, if in air with increasing thickness of the sample, parameter K_c slightly increases, then in hydrogen, it decreases (see Table 2, positions 5, 6 and 7, 8); i.e., the embrittlement effect of hydrogen on the fracture toughness of the alloy increases.

3.3. Influence of Metallurgical Factors on the Alloy Properties. Strength, plasticity, and fracture toughness of the alloy 05Cr19Ni55, melted in vacuum and open electric furnaces, were compared. For the same heat treatment, the alloy after vacuum arc remelting has much higher plasticity than after open melting (position 2, 3 and 4, and 5 in Table 2). Significantly higher are all crack resistance parameters: $K_{1c}(J_c)$ and K_c on air and K_{1c} prehydrogenated specimens in hydrogen (see Figure 5). The resulting open smelting fine-grained structure under the same conditions absorbs the largest number of hydrogen (position 4 in Table 2). To establish the reasons for increasing the properties of the vacuum alloy was determined gases content (by vacuum melting) and impurities (spectral method). Also conducted chemical and X-ray analysis of phases and studied the structure. It was found that open and vacuum melting alloy containing an identical amount of nitrogen have the same lattice parameters of γ- and γ'-phases.

Chemical composition and quantity of intermetallics and carbide phases did not differ. Grain sizes after HT were practically the same (see Table 2, positions 2, 3 and 4 and 5). Difference between metal open and vacuum melting was found in the analysis of phosphorus (decreases by 2 times), sulfur (decreases by 1.5 times), and gases. According to the gas analysis, specimens after vacuum arc remelting contained less (2 times) amount of gases than after open melting, mainly due to hydrogen (3–8 times) and oxygen (2 times) content. Thus, the high content of sulfur, phosphorus, hydrogen, and oxygen affects the characteristics of the nickel alloy 05Cr19Ni55 in air and hydrogen and increases the hydrogen embrittlement.

The relationship between mechanical properties of the alloy and texture that appeared as a result of rolling was investigated. In air, the value of the characteristics of plasticity (δ, ψ) of cylindrical specimens of the alloy (TO1) oriented in the transverse direction (orientation TV) is significantly lower than that of the specimens oriented in the longitudinal direction (LT) (see Table 2, positions 1 and 2). Transversely oriented specimens are more sensitive to hydrogen than longitudinal: δ decreases in 4 and 3.1 times, ψ decreasesin 2.7 and 2.25 times.

Figure 5: Fracture toughness of specimens of the 05Cr19Ni55 alloy of 20 mm thickness after open and vacuum arc melting in air (light bars) and in hydrogen under pressure 30 MPa after hydrogenation (shaded bars). The numbers near curves correspond to ordinal numbers in Table 2.

Resistance to fracture of the alloy also depends on the orientation of the specimens.

The value of static fracture toughness of compact specimens of 20 mm thickness with cracks oriented in the transverse orientation in air was $K_c = 69\,\text{MPa}\cdot\text{m}^{1/2}$ and $K_{Ic}(J) = 86\,\text{MPa}\cdot\text{m}^{1/2}$, respectively, which is approximately 20% smaller than that of specimens with cracks oriented in the longitudinal orientation (see Table 2, positions 1 and 2).

In hydrogen, characteristics of crack resistance transversely and longitudinal oriented specimens are 28 and 38.5 MPa·m$^{1/2}$. Similar results were obtained in the study of hydrogen embrittlement of steels 4130 and 4310 and other materials [22–24].

The most likely reason for anisotropy of mechanical properties can be anisotropic structural boundaries in the alloy, i.e., the dependence of fate crack length, falling on the structural boundary, and on the orientation of the applied load. The accumulation of hydrogen at the grain boundaries makes it easy to crack, while at transverse-loading large areas such boundaries are under the influence of normal stresses [21].

The difference in the degree of hydrogen embrittlement of the TV and LT samples is also due, probably, to the formed rolling process by the dislocation texture, which is decisive in hydrogen cracking [12, 13]. In addition, it is known that the effective rate of hydrogen diffusion depends substantially on the orientation of the samples [23].

3.4. Influence of Heat Treatment Regimes and Chemical Composition on the Alloy Properties. Effect of dispersion structure of the alloy on strength, plasticity, and fracture toughness in air and hydrogen was studied. Quenching temperatures were 1253 K, 1283 K, 1323 K, and 1473 K, grain sizes is equal to 21, 36, 85, and 130 microns, respectively, and the average thickness of grain boundaries is 1, 2, 3, and 4 microns (see Figure 6).

After quench from 1473 K alloy has a large grain (Figure 4(c)), small amount of hardening phases, high relative elongation and low strength (Table 2, position 9).

The method of J-integral does not recommend to evaluate the materials static crack resistance in plane-strain conditions. In the absence of large local stresses during the short-term strength tests, arise the minimal embrittlement effect of hydrogen.

For the same aging regimes with decreasing grain size, parameters of plasticity and static crack resistance increased in air and hydrogen (see Table 2, Figure 2, positions 3, 5, 8). At the same time, strength characteristics differ insignificantly. The best combination of mechanical properties in air and greatest resistance to hydrogen embrittlement was observed on longitudinal samples after VAR and HT II (see Table 2, position 4, 5). In this case, the fine-grained structure with equally distributed over the body of grain particles of γ'-phase is obtained. Additional reduction of carbon, nitrogen (from 0.12 to 0.07 mass %), and total content of chromium, molybdenum, niobium, and iron (see Table 1, CC II) leads to the increasing of alloy hydrogen resistance and fracture toughness in air and hydrogen (Table 3).

Probably, this effect is due to a reduction from 3.1 to 2.0 mass % quantity of carbides and carbonitrides (see Table 3, Figure 7), which promote brittle fracture and hydrogen embrittlement of nickel alloys [1–3, 17–19]. It has been found that at low deformation rates (~1 mm/min) and large hydrogen pressures (~3 MPa), intergranular destruction and destruction along the boundaries of the carbide matrix occur with cracking of carbides (Figures 8(a) and 8(b)). The change in the nature of the fracture from transgranular to intergranular under the action of hydrogen is a characteristic feature of dispersion-hard austenitic steels and alloys due to the concentration at the grain boundaries of carbides, intermetallics, and hydrogen [1, 2, 4, 5, 24–27].

Figure 6: Microstructure of the alloy: after VAR and HT II (a), HT I (b), and HT IV (c).

Table 3: Mechanical properties of alloys with different chemical composition at 293 K and displacement rate 0.1 mm/min.

CC	σ_u (MPa)	σ_{ys} (MPa)	δ %	ψ %	β_ψ	K_c MPa·m$^{1/2}$	$K_{1c}(J_c)$ MPa·m$^{1/2}$	β_{KIc}	(Cr, Ni, Fe)$_{23}$(C, N)$_6$, mass %
I	1080/970	650/660	35/6	38/19	0.50	126/64	162	0.40	3.1
II	1075/1010	630/650	42/12	44/29	0.66	139/89	171	0.52	2.0

Figure 7: X-ray patterns of the alloys with different content of carbides (I, II, see Table 1): a, Ni-base solid solution; b, Ni_3(Al, Mo, Nb); c, (Cr, Ni, Fe)$_{23}$(C, N)$_6$.

Figure 8: Fracture surface investigation of Ni alloy specimen with cumulation of failure due to static loading at 293 K in hydrogen under the pressure of 30 MPa: (a) intergranular fracture in the area of surface cracks and (b) destruction along the boundary of the carbide matrix with carbide cracking.

4. Conclusions

(1) With the decrease of strain rate from 100 to 0.1 mm/min, plasticity characteristics of the Ni alloy specimens in hydrogen are reduced in the interval rates 0.1–0.01 mm/min. Similarly, the level of K_c decreases with decreasing dK/dt to the minimum value, defined of material structural condition and cumulation of failure.

(2) The values of hydrogen pressure and hydrogen content by which has achieved the maximizing effect on investigated alloys fracture toughness has been established. Compact specimens thickness increasing leads to the hydrogen embrittlement increases. With compact specimen thickness increasing, the hydrogen embrittlement increases.

(3) After cumulation of failure, hydrogen has initiated the fracture by the mechanism of normal separation across the whole crack front, causing formation of the plane-strain state in the specimens.

(4) The alloy properties depend on the orientation of the samples. Strength, plasticity, and fracture toughness in air and hydrogen in transversely oriented specimens are much lower than those in longitudinally oriented specimens.

(5) Vacuum arc remelting, formation of fine structures of the thin grain boundaries, and a minimum number of carbides and carbonitrides have increased fracture toughness and hydrogen resistance of the alloy.

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Analysis on Fracture Toughness of the L360QS/N08825 Bimetallic Composite Pipe Welded Joint

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The fracture toughness of the weld and heat-affected zone (HAZ) of the L360QS/N08825 composite pipe welded joint was evaluated by a crack tip opening displacement (CTOD) test. The fracture morphology, microstructure, and inclusion near fracture zones were observed by means of scanning electron microscopy (SEM), transmission electron microscopy (TEM), and energy dispersive spectroscopy (EDS). The grain size and grain orientation of the crack propagation zone in the weld were investigated by electron back-scattered diffraction (EBSD). The results revealed that the average CTOD values of the weld and HAZ samples were relatively high, and a greater dispersion of CTOD values of the HAZ samples is due to the pop-in phenomenon in the P–V curve. The fracture surfaces of the weld and HAZ samples showed the characteristics of ductile fracture to a certain extent, whereas the fracture of the CTOD sample with the pop-in phenomenon exhibited a quasicleavage feature. High-density dislocation and a large number of inclusions were observed in the near fracture zone of the weld and HAZ samples. The stress concentration, caused by hindering the dislocation slip, was the main reason for microcrack formation. The existence of large-size grains and large-scale small-angle grain boundary in the weld implies that the cracks propagate toward the weld.

1. Introduction

Bimetallic composite pipes have been increasingly applied in the chemical industry, for environmental protection, for oil and gas transportation, and in the nuclear industry [1–3] because of their several benefits such as a reduction in material costs and a combination of good strength and excellent corrosion resistance of the two dissimilar materials [4, 5]. Because of great differences in the chemical composition, linear expansion coefficient, and thermal conductivity between the two dissimilar metals, it is easy for the bimetallic composite pipe welded joint to have defects and residual stress [6–8], which results in cracking [9–11]. The crack propagates until fracture occurs, thereby increasing the risk of oil and gas transportation and reducing the service life of the pipeline [12–14]. It is very important to investigate the fracture toughness of the bimetallic composite pipe welded joint to ensure the safety of construction projects.

At present, crack tip opening displacement (CTOD) is thought to be the essential fracture parameter to assess the fracture behavior of steel and welded joints based on nonlinear elastic fracture mechanics [15]. As the performance index and fracture criterion for crack initiation and propagation analysis of steel and welded joints, CTOD can not only effectively evaluate the fracture toughness of steel and welded joints, but it can also provide an experimental basis for the reliability and security assessment of construction [16, 17] through CTOD tests.

A few studies of fracture toughness of welding materials have been done by means of CTOD tests. Wang et al. [18] studied the CTOD fracture toughness of the weld and heat-affected zone (HAZ) of X80 steel at 0°C according to the BS7448 standard and found that the CTOD test could be used to evaluate effectively the fracture toughness of welded joints. Miao et al. [19] conducted a CTOD test on a submerged arc-welded joint of extrathick ocean structural steel S355G10+N. It was shown that the fracture toughness of the S355G10+N welded joint is excellent, and the welding process can be directly used to construct a marine steel structure. In addition, it was suggested that high stress should be avoided in the weld zone owing to the lowest

toughness of the weld. Coronado and Cerón [20] studied CTOD for welded joints of three welding procedures and found that the fracture parameter CTOD was correlated with the fracture surface and microstructures. Leng et al. [21] studied the relationship between the fracture toughness and microstructure of S335G10+N welded joints by CTOD, a Charpy impact test, and scanning electron microscopy (SEM), and they found that the change trend of the CTOD value at the weld and fusion line with a notch was similar to that of the impact value. In addition, there is a big difference in the average grain sizes of S3557, S3556, and S3559 near-fracture microstructures. It was found that the larger the average grain size, the smaller the CTOD value. Wang et al. [22] studied the fracture properties and microstructure of A508/316L welded joints. The results showed that the fracture mechanism of A508, 316L base materials, and the HAZ of 316L is typical ductile fracture of the nucleation, growth, and coalescence of micropores. The fracture mode of the HAZ of A508 and the interface area of A508/52Mb composed of martensite is ductile-brittle mixed fracture, and the presence of martensite leads to low crack growth resistance. At the same time, it was found that the orientation of columnar austenite crystal can affect significantly the fracture mechanism and crack growth resistance in the weld zone. Guo et al. [23] also studied the fracture toughness and microstructure characteristics of various regions of 9Cr/CrMoV welded joints. The research results show that the fracture toughness of the CrMoV side is better than that of the 9Cr side because the microstructure of the former is fine austenite and bainite and the latter is coarse martensite. Li et al. [9] studied the crack and fracture properties of Fe_3Al/Cr18 Ni8 welded joints. The test shows that the crack initiation is in the fusion zone of the Fe_3Al side, in which there is much dislocation; most of the cracks propagate along the fusion zone, but a few cracks propagate horizontally to the HAZ and end at the weld. This is because of $\gamma + \delta$ organization in this area. Ju et al. [24] investigated the variations between ductile-to-brittle transition temperature (DBTT) and CTOD within the HAZ of API X65 pipeline steel. Both values varied dramatically with the distance from the fusion zone, and the region near the fusion zone exhibited the lowest CTOD and highest DBTT, which is possibly caused by the increasing portion of coarse grain in the HAZ. Because most mechanical components are subjected to complex loading conditions with varying magnitude and direction, Mokhtarishirazabad et al. [25] successfully evaluated the overload effect of biaxial fatigue cracking in terms of crack growth rate, crack opening displacement, and stress intensity factor, providing a new method to study the effect of applying overload cycle on the behavior of a crack under cyclic biaxial loading. The CTOD test can not only be used as the basis for selecting the toughness of piping and offshore materials, but also to provide the test basis for evaluating the safety and reliability of structures.

A few research efforts have been made to examine the fracture toughness for a bimetallic composite pipe via CTOD tests because the application of bimetallic composite pipes is becoming more and more extensive. To ensure the safety of bimetallic composite pipe welded joints during operation, in the present study, an experimental investigation was carried out using CTOD tests to evaluate the fracture toughness of the weld and the HAZ in an L360QS/N08825 bimetallic composite pipe welded joint according to BS7448 [26–28], ISO12135 (2002) [29], and ISO15653 (2010) [30] fracture toughness test standards at room temperature. The fracture morphology, microstructure, and inclusion near fracture zones were observed by means of SEM, TEM, and EDS, and the differences in grain size and grain orientation of the crack propagation area were studied by the EBSD technology. The primary aim was to study the fracture toughness characteristics of each part of the welded joint to provide a theoretical reference for engineering applications.

2. Materials and Methods

The cladding and base materials were made of N08825 and L360QS, respectively. The chemical compositions of the L360QS/N08825 bimetallic composite pipe formed by centrifugal casting are given in Table 1. The L360QS/N08825 bimetallic composite pipe circumferential weld was welded using tungsten inert gas (TIG) method through a D/T-HW350 automatic welder with a U-shaped groove shown in Figure 1. The filler metal was an ERNiCrMo-3 welding wire, whose chemical composition is given in Table 2. The diameter of the L360QS/N08825 bimetallic composite pipe welded joint was 610 mm, and the wall thickness was (20 + 3) mm.

2.1. Sample Preparation. The fracture toughness test was conducted according to the BS7448 standard to obtain the CTOD value of the welded joint. Three-point bend specimens of the SE(B) with fatigue precrack in the weld and the HAZ were prepared, and the form of the SE(B) is shown in Figure 2. Thickness (B), width (W), and length (L) are the dimensions of the specimen, M is the machining notch, m is the length of the precrack generated by fatigue, and N is the notch thickness.

For estimating each parameter, the ISO12135 (2002) and ISO15653 (2010) standards were used to intercept samples of the HAZ and weld, respectively. The size of the weld sample was $B \times 2B$, and the notch direction was NP. The size of the HAZ sample was $B \times B$, and the notch direction was NQ. The schematic diagram of the sample interception is shown in Figure 3, where N is the vertical weld direction, P is the direction of the parallel weld, and Q is the direction of weld thickness. NP implies that the direction of the sample length is a vertical weld and the direction of crack propagation is a parallel weld. NQ represents that the direction of sample length is a vertical weld and the direction of crack propagation is in the direction of the weld thickness. There were three samples to be tested for the weld and for the HAZ: W1, W2, and W3 and H1, H2, and H3, respectively.

2.2. Machining of Mechanical Notch. The cutting line of the specimens was marked in accordance with the BS7448 standard. For the HAZ specimens, the line should be drawn on the fusion line to ensure that the crack-tip position is not

Table 1: Chemical composition of L360QS and N08825 (mass fraction/%).

Material	C	Si	Mn	P	S	Ni	Cr	Cu	Mo	V	Ti
N08825	0.022	0.14	0.47	0.015	0.007	41.14	22.07	1.47	3.13	—	—
L360QS	0.07	0.30	1.06	0.012	0.001	0.090	0.090	0.060	0.040	0.040	0.002

Figure 1: Bimetallic composite pipe welded joint.

Table 2: Chemical composition of ERNiCrMo-3 (mass fraction/%).

Material	Ni	C	Cr	Mo	Nb	Fe	Ti	Si
ERNiCrMo-3	64.43	0.011	22.2	9.13	3.53	0.19	0.23	0.05

Figure 2: Three-point bending specimen SE(B).

Figure 3: Schematic diagram of CTOD sample interception. (a) HAZ; (b) weld.

more than 0.5 mm from the fusion line. The line should coincide with the center line for the weld specimens. In addition, the vertical angle between the plane of the cutting line and the sample cutting surface should be $90^\circ \pm 5^\circ$ when the notch is being processed. An electrical-discharge-machining (EDM) wire-cutting process with molybdenum wire of 0.08 mm diameter was used to machine the notch on the side of the N08825 alloy along the thickness direction of the

specimen. The machining depth of the notch was approximately 45% of the specimen thickness. The angle of the notch root was less than 60°, and the radius of the notch root was 0.12 mm.

2.3. Experimental Procedure. The fatigue precracking was carried out on the side of the N08825 alloy using a PLG-200 high-frequency fatigue-testing instrument with a vibration frequency between 115 and 117 Hz at room temperature (20°C). The load ratio was set at 0.1. For the weld and HAZ samples, the loading span was set as 160 and 80 mm, respectively. Meanwhile, maximum fatigue precracking loads of 22.00 and 35.00 kN were applied for the weld and HAZ samples, respectively. The time of fatigue precracking for each sample was approximately 0.4 h. According to the BS7448 standard, crack growth should not be too fast when fabricating the fatigue precracking. In the last 1.3 mm, the fatigue load ratio can be appropriately increased to avoid the growth of the surface crack being larger than that of the internal crack. Loading force needs to be strictly controlled to prevent the plastic deformation of the crack tip during the process of fatigue precracking and ensure that the initial crack length a_0 of all samples should be within the range of 0.10–0.45W (HAZ) or 0.45–0.70W (weld), where W is sample width (mm) [30].

CTOD experiments were performed with a DDL 300 universal testing machine at room temperature; at the same time, the P–V curve was automatically recorded (P is applied load, and V is the crack tip opening displacement). A typical P–V curve is shown in Figure 4. The parameters of the experiment were: speed of application of the load, 0.1 mm/min; span, 160 mm for the weld and 80 mm for the HAZ; and sampling interval for equipment data acquisition, 0.1 s. The initial crack length (a_0) [28] and CTOD value (δ) [26] are calculated according to the following equation:

$$a_0 = \frac{1}{8}\left(\frac{a_{01} + a_{09}}{2} + \sum_{i=2}^{8} a_{0i}\right),$$

$$\delta = \left[\frac{FS}{BW^{1.5}} \times f\left(\frac{a_0}{W}\right)\right]^2 \frac{(1-\mu^2)}{2\sigma_S E} + \frac{0.4(W-a_0)V_p}{0.4W + 0.6a_0 + Z}, \quad (1)$$

where a_{0i} is the initial crack length of the i-th test point ($i = 2, 3, 4, \ldots 8$) (the measure of a_{0i} is shown in Figure 5), F is the applied load, S is span of SE(B), μ is Poisson's ratio, E is elastic modulus, f is function for (a_0/W), V_p stands for the plastic component of CTOD, σ_S is yield strength, and Z is blade thickness.

After the CTOD test, the validity of the CTOD samples was determined according to the BS7448, ISO12135 (2002), and ISO15653 (2010) standards, and specific steps were performed based on standard procedures from the literature [21]. The analysis showed that the CTOD samples of the welded joints were effective.

The fracture surface of the samples was analyzed by SEM using secondary electrons, while the composition of fracture inclusions was analyzed by EDS. The SEM sample was etched by a solution of $CH_3OH + HNO_3$ with the proportion of 3 : 1. TEM was used to observe and analyze the microstructure morphology and the distribution of dislocations in the near fracture of the sample, and the acceleration voltage was 200 kV. The piece of TEM observational sample was cut from the middle thickness region of the specimens along the direction perpendicular to the machining notch, as shown in Figure 6. After mechanical thinning to 50 μm, the TEM

Figure 4: Typical P–V curve.

Figure 5: Measure of a_{0i} ($i = 2, 3, 4, \ldots, 8$).

Figure 6: TEM observation of CTOD near-fracture sample.

specimen was exposed to a double electrolytic jet in an electrolytic double spray device, and the electrolyte was 75% $CH_3OH + 25\%$ HNO_3 (volume fraction). To understand further the direction of crack propagation from the perspective of crystallography, the crystal structure, grain size, and grain boundary misorientation of the weld zone were observed and analyzed using EBSD technology.

3. Results and Discussion

3.1. Discussion of CTOD Value. Table 3 summarizes the data concerning the fracture toughness tests. The average CTOD value (δ_m) of the weld samples is 1.3232 mm, and the minimum value is 1.2022 mm. Meanwhile, the average CTOD value of the HAZ samples is 1.5119 mm, and the minimum value is 0.9787 mm.

As Table 3 shows, the δ_m value of the HAZ is highly discrete. The maximum δ_m value of 2.4457 is more than twice the minimum δ_m value of 0.9787, which is thought to be caused by the heterogeneity structure and performance in the HAZ. The lowest CTOD value of H2 may be caused by the pop-in phenomenon in the P–V curve of sample H2 shown in Figure 7. The pop-in phenomenon shows that the load suddenly decreases and the displacement increases in the P–V curve, and then the load and displacement continue to increase until cracking. Meanwhile, when the pop-in phenomenon occurs, a relatively weak crackling sound can be heard during the loading process, and a brittle fracture zone can also be observed on the fracture surface of the sample. In fact, this pop-in phenomenon was derived from the small brittle crack initiated at a local brittle zone. This rapidly propagating brittle crack was subsequently arrested when it propagated into the higher-toughness region that surrounded the local brittle zone [31]. The macroscopic fracture of sample H2 shown in Figure 8 reveals the signs of crack tip instability in the local brittle zone, and the crack propagation path is extremely unstable. Cleavage steps can also be observed in the crack growth zone, which conforms to the pop-in phenomenon characteristics. In the loading process of the CTOD sample, the reduction in resistance to crack propagation of sample H2 resulted from unstable crack propagation, which resulted in a lower δ_m value. The allowable value of CTOD was set at 0.30 mm by referring to the design value of similar steel structures. The CTOD values of all zones of the above automatic TIG welded joints exceeded the allowable value (0.30 mm), so this process can be used for welding L360QS/N08825 composite pipe.

3.2. SEM Fractography Analysis of CTOD Sample. SEM images of the fracture surfaces of the weld and HAZ samples are shown in Figures 9 and 10, respectively.

Figures 9(a), 9(d), and 9(g) show that there are many shallow dimples and some flat brittle appearance on the fracture surfaces of the crack initiation area of the weld samples, which is believed to be a result of separating the weak grain boundaries with low fracture energy [32]. Figures 9(b), 9(e), and 9(h) are the fracture images of the crack propagation region of the weld sample. In Figures 9(b) and 9(h), the fracture is ductile fracture with large and spherical dimples distributed alternately, and small dimples have side-by-side connections. Figure 9(e) shows the fracture surface that is characterized by parabolic dimples, and inclusions are distributed at the bottom of the dimples. As Figures 9(c), 9(f), and 9(i) show, the fracture surfaces of the final fracture zone consist of equiaxial dimples distributed uniformly, which are typical of microvoid coalescence fracture [33].

Observations of the fracture surfaces of the crack initiation area of the HAZ samples are shown in Figures 10(a), 10(d), and 10(g). Figure 10(a) shows that the shape of the dimples is parabolic, and the size is nonuniform. The number of dimples in Figure 10(g) is less than that in Figure 10(a), which indicates that the toughness of sample H3 is weaker than that of sample H1 in accordance with the CTOD value (δ_m) listed in Table 3. The fracture morphology of the crack initiation area of H2 in Figure 10(d) is a quasicleavage fracture, and macroscopic cracking occurs between some cleavage surfaces. The occurrence of macroscopic cracks may be caused by cracking along grain boundaries or brittle phases. The SEM fracture analysis of the crack propagation region of sample H1 revealed a ductile fracture with large and deep dimples in a parabolic shape, as shown in Figure 10(b). The extensive and deep dimples in the fracture morphology of H1 indicate that a great deal of energy has been consumed during the fracture process [34], contributing to a high CTOD value (δ_m). Some shallow dimples can be seen in the crack propagation region of sample H2 (Figure 10(e)), which is characteristic of local ductile fracture. Figure 10(h) shows that the fracture surfaces of H3 had a ductile appearance characterized by equiaxial dimples that are typical of microvoid coalescence fracture. Figures 10(c), 10(f), and 10(i) show high-resolution fractographs of the final fracture zone of the HAZ samples. Figure 10(c) shows that many parabolic dimples were distributed on the fracture surface of H1, and the fracture had a ductile fracture morphology. The fracture surfaces of the final fracture zone of H2 consisted of brittle fracture with quasicleavage characteristics, including a fluvial pattern, as shown in Figure 10(f). This fracture was accompanied by less plastic deformation and energy consumption, and it corresponds to lower crack growth resistance. In the final fracture zone of H3, some small dimples and brittle facets can be seen, as shown in Figure 10(i), and this is characteristic of mixed ductile–brittle fracture.

Table 3: CTOD values and parameters of specimens of L360QS/N08825 welded joint (20°C).

Samples	Crack location	Mechanical properties	Thickness B (mm)	Width W (mm)	Crack length a_0 (mm)	a_0/W	CTOD value δ_m (mm)	Average value $-\delta_m$ (mm)
W1		$\sigma_s = 370$ MPa	19.82	39.88	21.55	0.54	1.2739	
W2	Weld	$E = 2.1 \times 10^5$ MPa	19.98	39.90	20.59	0.52	1.4935	1.3232
W3		$\mu = 0.3$	19.96	39.98	21.02	0.53	1.2022	
H1		$\sigma_s = 370$ MPa	19.94	19.91	4.21	0.21	2.4457	
H2	HAZ	$E = 2.1 \times 10^5$ MPa	19.96	19.94	3.36	0.17	0.9787	1.5119
H3		$\mu = 0.3$	19.92	19.85	3.77	0.19	1.1113	

Figure 7: Pop-in phenomenon in P–V of sample H2.

Figure 8: Macroscopic fracture morphology of sample H2.

As shown in Figure 10, there is a clear difference in the fracture surface characteristics between sample H2 with the pop-in phenomenon and samples H1 and H3 without the pop-in phenomenon. The fracture morphology of the crack initiation zone and the final fracture zone of samples without the pop-in phenomenon is the ductile dimple. However, the fracture morphology of the same area of the samples with the pop-in phenomenon is quasicleavage. This further confirms that samples containing the pop-in phenomenon have lower CTOD values compared with samples without the pop-in phenomenon in line with data shown in Table 3.

The morphology and composition of inclusions in the fracture surface of the CTOD sample of the HAZ are shown in Figure 11. The spherical inclusions are located at the bottom of the dimple and are identified as oxides, including Al_2O_3 and CaO, by EDS analysis. The inclusion is a favorable position for the formation of dimples, and the stress concentration around the inclusion generates a microstress field, which leads to the formation and propagation of cracks until ultimate fracture.

3.3. TEM Analysis of CTOD Sample near Fracture Zone. The microstructure of the near fracture zone of CTOD fracture samples of the L360QS/N08825 composite pipe welded joint is shown in Figure 12. Figures 12(a) and 12(b) show that the microstructure of the near fracture zone of the HAZ samples is mainly dendritic austenite with a clear dendritic boundary, and the branch width is between 181.61 and 620.69 nm. The appearance of the dendritic austenite microstructure of the near fracture zone of the HAZ samples is because the location of the CTOD fracture crack propagation zone is in the weld. At the same time, there is a high dislocation density inside the austenite grain and along the grain boundary, and dislocation entanglement was also observed in the grain. As reported in the literature [35], dislocation slips forward under the action of external forces. When the dislocation moves to the grain boundary, it is obstructed and causes local stress concentration. The stress concentration at the grain boundary becomes the crack source that eventually develops into a microcrack with the increase of the deformation degree, which eventually leads to fracture after crack growth, propagation, and unstable propagation. In addition, a large number of precipitated phases with a relatively uniform size of approximately 15 nm on the dislocation line were also observed inside the dendritic austenite grain.

Figures 12(c) and 12(d) show that the microstructure morphology of the near fracture zone of the weld samples is similar to that of the HAZ samples. The major microstructure is dendritic austenite with a clear dendritic boundary, and the branch width is in the range of 228.99 to 1660.00 nm. Furthermore, high-density dislocation and precipitated phases with different sizes ranging from 5.88 to 17.65 nm were also observed in the austenite grain and grain boundary.

The above analysis shows that the high-density dislocation was distributed in the grain and grain boundary, providing conditions for the initiation of microcracks that go through the growth and propagation process until fracturing eventually. The relevant literature [36] indicates that the precipitations on the dislocation line could hinder

Figure 9: CTOD microscopic fracture photo of weld of L360QS/N08825 bimetallic composite pipe welded joint: (a), (b), (c) W1; (d), (e), (f) W2; (g), (h), and (i) W3.

dislocation motion effectively and delay the initiation and propagation of cracks to some extent, which is beneficial to the toughness of materials. The appearance of a mass of precipitations in the HAZ and weld might be one of the reasons why the average δ_m value of the CTOD samples was higher than the standard value (0.30 mm).

In general, the inclusions in the matrix of the metal material welded joint play an important role in the fracture process as the nucleus center of microvoid formation and aggregation. Figure 13 shows the distribution of inclusions and EDS analysis of the near fracture zone of the HAZ samples. Figure 13 shows that there are approximately 40 spherical inclusions with different sizes between 3 and 8 μm in diameter in the area of $400 \times 400\,\mu$m. These inclusions were distributed in the grain and the grain boundary, and they were composed of sulfide and oxide, mainly including TiO_2 and CaS.

The distribution of inclusions and EDS analysis of the near fracture zone of the weld samples are shown in Figure 14. Many irregularly shaped inclusions with nonuniform size were distributed unevenly on the weld substrate. The EDS results reveal that the composition of inclusions is mainly SiO_2 oxide. The occurrence of stress concentration happens easily and is common around the inclusions under the action of external force, which increases the opportunity for microcracks to form in the material. Meanwhile, high-density distributed inclusions make the connection of microcracks easier, which eventually leads to the fracture of components after the process of growth and propagation of cracks [35].

3.4. EBSD Analysis of Weld Zone. Metallography observation on the surface perpendicular to the fracture surface showed that fatigue precracks of HAZ and weld samples propagated into the weld zone during the CTOD test. To reveal the relationship between the direction of crack propagation and microstructure and the grain size and grain boundary

Figure 10: CTOD microfracture photo of the HAZ of L360QS/N08825 bimetallic composite pipe welded joint: (a), (b), (c) H1; (d), (e), (f) H2; (g), (h), and (i) H3.

Figure 11: SEM of morphology and composition of inclusions in the fracture zone of the HAZ.

Figure 12: TEM images of near fracture zones of CTOD samples: (a, b) HAZ and (c, d) weld.

misorientation from the perspective of crystallography, EBSD technology was used to analyze the grain size and grain boundary misorientation in the weld zone.

The strength and toughness of materials are closely related to the sizes of grains. Generally, the strength and toughness of materials are improved as the size of the metal grain decreases [37, 38]. Therefore, it is necessary to study the grain size of the crack propagation zone in the weld. Figure 15(a) shows the grain size distribution in the weld zone, and the color of each grain is selected randomly. As the histogram of the weld grain size distribution in Figure 15(b) shows, there are 1172 grains, and the average grain size is 95.2 μm. Based on the above data analysis from Figure 15, the average grain size in the weld zone is relatively large, which leads to a decrease in toughness and promotes crack propagation in this area.

To reveal the crystallographic parameters of the weld microstructure further, the grain orientation distribution of the weld (the same color represents the same grain orientation) and the grain boundary misorientation distribution diagram of the weld are shown in Figures 16 and 17, respectively. Mathematical statistics show that there is an approximately 81% small-angle grain boundary (2°–15°) in the weld zone, and the remaining 19% is a high-angle grain boundary distributed within 49°–62° randomly. This result indicates that the grain boundary misorientation of the weld zone is small, and the appearance of high-angle grain

(a) (b)

Figure 13: SEM images of inclusions and compositions in the near fracture zone of the HAZ.

(a) (b)

Figure 14: SEM images of inclusions and compositions in the near fracture zone of weld.

(a) (b)

Figure 15: Grain size of weld: (a) grain size and (b) histogram.

boundaries (>15°) results from the accumulation of small-angle grain boundaries [39]. According to the relevant literature [40], the existence of small-angle grain boundaries leads to the decline of material toughness, which is conducive to the propagation of cracks. Therefore, a large number of small-angle grain boundaries in the weld zone reduce the resistance

Figure 16: Grain orientation distribution of weld.

Figure 17: Grain boundary misorientation distribution histogram of weld.

to crack propagation, which explains why cracks propagated into the weld zone during the CTOD test.

4. Conclusions

In the present work, the fracture toughness characteristics of bimetallic composite pipe welded joints were studied. The following results were obtained:

(1) The CTOD values of the HAZ and weld samples were high; nevertheless, the greater dispersion of CTOD values of the HAZ samples resulted from the pop-in phenomenon in the P–V curve of sample H2.

(2) The fracture surfaces of the L360QS/N08825 composite pipe welded joint were characterized by ductile fracture to a certain extent, while the fracture

surface of the CTOD sample in the HAZ region containing the pop-in phenomenon had a quasi-cleavage morphology.

(3) The microstructure of the near fracture zone of the HAZ and weld samples was mainly dendritic austenite, and there were high-density dislocation and dislocation entanglement in the grains and grain boundary. A large number of spherical or irregular inclusions were observed, which is the main reason for the appearance of microcracks. Meanwhile, the high-density inclusions distributed on the matrix made the connection of microcracks easier, which eventually leads to the fracture of components after the process of growth and propagation of cracks.

(4) The average grain size of the weld structure was 95.2 μm, and small-angle grain boundaries accounted for approximately 81%. A larger grain size and a larger proportion of small-angle grain boundary can reduce the resistance to crack propagation and promote crack propagation.

Data Availability

The data used to support the findings of this study are included within the article.

Conflicts of Interest

There are no conflicts of interest regarding the publication of this paper.

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Research Article

Elliptical Method for Damage Identification in Carbon Fibre Reinforced Polymer Laminates

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Oftentimes, researchers in the area of vibration-based structural health monitoring (SHM) and damage detection focused their attention on the global properties of structures, which are modal frequencies and modal damping factors. However, the effect on the local properties for SHM, that is, modal constants, has not been extremely explored. In this paper, the elliptical plane modal identification method is proposed to be used as a damage identification method itself. It is observed that when the receptance is plotted in the elliptical plane, the area of the ellipse formed close to the resonant frequencies (which depends on the modal constants) can be used to detect damage, namely, composite carbon fibre reinforced polymer (CFRP) rectangular plates. Although a mathematical correlation has not been established yet, results show that the method is sensitive to the presence of damage in the test plates, as the area of the ellipse changes with damage.

1. Introduction

The ever-developing world as continuously demanded an advancement in every field of endeavour to meet up with the increasing desire for a better society yet does not have a damaging effect on the environment. In the field of engineering, a composite carbon fibre reinforced polymer (CFRP) is one of such materials that has shaped and challenged several industries such as marine, aerospace, civil infrastructural, and automotive and sports equipment, owing to their rare mechanical properties, namely, strength and stiffness to weight ratios [1–5]. Despite the unique mechanical properties of CFRP, low-impact damages have been its shortcomings. Such damages may be introduced, for instance, by a hailstone, bird strike, debris, stones, or tools drop during manufacturing/maintenance [6]. Medium-to-low energy impacts (i.e., 1–10 m/s and 11–30 m/s, respectively) on the surface may result in BVID (barely visible impact damage), i.e., in a mark other than a small indentation that is difficult to identify through visual inspection. However, the impact may have resulted in damage that propagates under different mechanisms through the thickness of the laminate down to the opposite side which is usually hidden. This could compromise the integrity of the structure, reduce the remaining useful life (RUL) [7], and raise safety issues. Hence, it is pertinent to implement systems that can promptly identify and locate damage in composites in order to avert an unexpected breakdown of structures.

Several nondestructive techniques (NDTs), such as acoustic emission, ultrasound, visual inspection, X-ray, or eddy currents, among others [8], have been developed over the years in an attempt to get a more efficient, simple, and economical solution for monitoring and detection of damage in composite structures. However, no single technique has proven appropriate for all circumstances. Each

technique has its uniqueness, effectiveness, and range of applications. It would be beneficial to develop a method that could detect damage in a structure based on its global properties through vibration testing. In that regard, researchers have engaged in continuous studies in the area of the analysis of the vibration characteristics of structures to identify damage without prior knowledge of the location of the damage.

The method is hinged upon knowing the state of the healthy characteristics of the structure and using it as a baseline to compare with the vibration characteristics of the structure at a planned period or at its damaged state. The comparison would highlight any noticeable deviation in the case of the presence of damage and might even reveal the damage location and its severity.

An extensive review on SHM and techniques for damage detection has been presented in [9–11]. Montalvão et al. [12] presented a review of vibration-based SHM with special emphasis on composite materials. Among other damage identification methods, modal analysis is the mostly applied technique [13]. This method utilises the deviation in the modal parameters (modal frequencies, modal damping, and mode shape) of a structure. The dynamic behaviour of the structure is analysed based on the modal parameters extracted from the raw data collected. The modal properties can be used to monitor vibration and damages in a CFRP.

It is difficult to analyse the interactions between all the features of mechanical systems. However, the dynamic properties of the mechanical system can be represented if the basic properties are assumed to be a single-degree-of-freedom (SDOF) system and considered separately [14]. The dynamic characteristics of structures can be described with spatial, modal, or frequency response model as stated by Maia and Silva [14]. It is interesting to note that these models can be linked with each other [14, 15]. The spatial distribution of mass, stiffness, and damping properties are illustrated in terms of matrixes of mass $[M]$, stiffness $[K]$, and damping $[D]$ (for the hysterically damped model) or $[C]$ (for the viscously damped model) [14, 15].

Over the years, researchers have always been focusing on the global properties of structures for SHM, with little interest in the local properties—the modal constants. This work explores the plausibility of damage identification with the modal constants. This study focuses on understanding the possible relationships between the deviation in the ellipse area and damage. The modal constants of the CFRP material determine the area and shape of the ellipse.

2. Materials and Methods

2.1. Materials Preparation. In this investigation, laminates with dimensions as shown in Table 1 were manufactured and used to conduct the experimental modal analysis. The composite consists of plain weave carbon fibres as the reinforcement and epoxy as the matrix.

It is a unidirectional (UD) prepreg FIBREDUX 6268-HTA (12K) carbon/epoxy material. HexPly® 6268 provides good adhesion to honeycomb core and suitable for aircraft structures. All the test samples were manufactured using the hand layup, and a bagging film (Nylon 66) that can withstand high temperatures and pressures was used during the curing process. The bagging process can be seen in Figure 1.

To avoid sticking of the prepreg laminates to the aluminium plate (forming tool), the prepreg laminates were sandwiched between two release films. The vacuum bag was sealed with an inner yellow sealant tape; the pressure gauge and vacuum pipes were connected to the vacuum bagging through valves. For the autoclave curing cycle, start-up heat was 20°C and heat up rate was 1°C/min until it reached 121°C. The laminates were cured at 106 kPa for 2 hours and then naturally cooled.

2.2. Vibration Testing. The composite plates were suspended vertically under a free-free simulated configuration with 2 nylon strings as shown in Figure 2. A force transducer is attached at the corner of the specimen and connected to an electromagnetic shaker through a pushrod (stinger).

The response of the samples due to the generation of a multisine [16] excitation signal for a frequency range of 0 to 800 Hz with a frequency resolution of 0.25 Hz was measured. The responses were measured at a specific location using three lightweight PCB teardrop accelerometers, type 352A24, that weighs 0.8 g each, at the corner of the specimen to acquire the Frequency Response Functions (FRFs). The experiments were performed for both healthy and damage-induced samples of the same configuration.

A number of experiments were conducted by assembling/disassembling, and results were generally consistent although a statistical analysis was not formally conducted.

2.3. Static Testing. In order to introduce damage in the specimen, static testing (ST) was performed using a 25 kN Tinius-Olsen universal testing machine. This technique has been followed by other authors in the absence of impact testing machines, such as [3, 17]. The experimental setup for the ST is illustrated in Figure 3.

An indenter made from carbon steel with a hemispheric tip of 24.5 mm in diameter was used. The setup consists of a 350 mm^2 rectangular fixture base plate with a cutout of 250 × 150 mm. A total of six vertical toggle clamps hold the specimen to the fixture base, three spaced equally on each side of the length of the fixture base. The specimen is clamped to the base plate to avoid movement during loading. The ST was conducted at a loading speed of 10 mm/min, with the application of varying force. After each round of loading, the energy dissipated on the test plates was within the range of 0.41 kJ to 18 kJ. The study considered five different stacking sequences, and they are labelled in alphanumeric style, that is, plates A1, A2, B1, B2, C1, C2, D1, and D2, as shown from Figures 4–6.

Since the area and the shape of the ellipse depend on the real and imaginary modal constants, a possible deviation in the area of the ellipse due to damage would suggest a correlation. The area of the ellipse from the test specimens was identified using the elliptical method [18–20]. The ellipse

Table 1: Types of specimens and designations.

Quantity	Designation	Material	Stacking sequence	Laminate type	Dimensions (mm)	Aspect ratio (a/b)
2	Plate A1–A3	FibreDUX 6268C-HTA 12K	$[90/\pm45/0]_s$	Quasi-isotropic	$310 \times 240 \times 2$	1.29
2	Plate B1-B2	FibreDUX 6268C-HTA 12K	$[90/0/\pm45]_s$	Quasi-isotropic	$310 \times 240 \times 2$	1.29
2	Plate C1–C3	FibreDUX 6268C-HTA 12K	$[90/0]_{2s}$	Cross-ply	$310 \times 240 \times 2$	1.29
2	Plate D1	FibreDUX 6268C-HTA 12K	$[90_2/0_3]_s$	Cross-ply	$310 \times 240 \times 2$	1.29
1	Plate E1	FibreDUX 6268C-HTA 12K	$[90_3/0_3]_s$	Cross-ply	$300 \times 241 \times 3$	1.24

Figure 1: Vacuum bagging process.

Figure 2: Experimental setup for vibration testing [6].

Figure 3: Static testing.

modal identification software can also be used to determine the global properties of a structure, that is, the modal frequency and modal damping.

3. Results and Discussion

The theoretical development of the elliptical method and its properties are presented in [18]. The elliptical method depends on the modal constants—the real and imaginary parts of the modal constant. The real part of the modal constant can be determined by using the following equation:

$$A_{\mathrm{R}} = \sqrt{H^2_{\omega=\omega_{\mathrm{r}}} \eta_{\mathrm{r}}^2 \omega_{\mathrm{r}}^4 - A_I^2}, \tag{1}$$

while the imaginary part of the modal constant is determined by using the following equation:

$$A_I = \sqrt{\frac{H^2_{\omega=\omega_{\mathrm{r}}} \eta_{\mathrm{r}}^2 \omega_{\mathrm{r}}^4}{\left[\tan\left[\sin^{-1}\left(\theta_{\omega \ll \gg \omega_{\mathrm{r}}}\right)\right]\right]^{-2} + 1}}, \tag{2}$$

where A_{R} is the real part of the modal constant, A_I is the imaginary part of the constant, H is the amplitude of the receptance, ω is the natural frequency, ω_{r} is the angular frequency for mode r, η_{r} is the hysterical damping factor for mode r, and θ is the phase angle between the force and the displacement response. From the results, it was observed that the presence of damage in the CFRP reduces the area of the ellipse. Some representative elliptical shapes are shown in Figures 7 and 8.

The amplitudes of the healthy and damaged ellipse for plate A1 shown in Figure 7 are multiplied by 5 and 10,000 scale multipliers, respectively. Despite a large amount of multiplier, the area of the ellipse for the damaged plate is relatively smaller than that for the healthy plate. This

Figure 4: Variation in the area of the ellipse for plates (a) A1, (b) A2, (c) A3, and (d) B1.

Figure 5: Continued.

Figure 5: Variation in the area of the ellipse for plates (a) B2, (b) C1, (c) C2, and (d) C3.

Figure 6: Variation in the area of the ellipse for plates (a) D1 and (b) E1.

Figure 7: Continued.

(b)

Figure 7: Ellipse area for plate A1: (a) healthy; (b) damaged.

(a)

Figure 8: Continued.

(b)

Figure 8: Ellipse area for plate B1: (a) healthy; (b) damaged.

suggests the area of the ellipse changes with damage, and there might be a possible correlation.

Again, the ellipse area for the healthy plate B1 is larger than that for the damaged plate as shown in Figure 8. Although the area of the ellipse for the damaged plate appears to be larger pictorially, in reality, it is opposite. The larger appearance was due to being multiplied by a scale multiplier of 10,000 and that of the healthy ellipse being multiplied by just 1 scale multiplier.

Obviously, from Figures 4–6, there is a variation in the area of the ellipse after the damage was introduced in the test plates. In plate A1, the area of ellipse decreased at all the 10 modes considered.

It is important to note that the amount of reduction in the ellipse areas as shown from Figures 4–6 is within the same range. Apart from specimen C3, the reduction rate across all the modes in other specimens is over 97%. The results indicate that the elliptical method is sensitive to the presence of damage in the composite CFRP rectangular plate.

4. Conclusion

A novel method for damage identification from FRF, based on the representation of the receptance on the elliptical plane, was presented. It was shown that the area of the ellipse, which is related to the modal constants (local modal properties), is sensitive to damage in a consistent manner. Hence, this paper offers new possibilities for other researchers who are concerned with damage diagnosis in lightly damped structures since the elliptical plane modal identification method provides promising results for damage identification in CFRP rectangular plates. However, more studies are still required in terms of experimental work to find what the mathematical correlation is (if any) between damage and the shape of the ellipse, for example, by taking into account that damage in CFRPs has complex morphologies that may affect the plates and how the modal constants are affected by the presence of damage.

Data Availability

The Frequency Response Functions (FRFs) data used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that there are no conflicts of interest.

Authors' Contributions

Daerefa-a Mitsheal Amafabia and Diogo Montalvão conceived and designed the experiments. Daerefa-a Mitsheal Amafabia, Diogo Montalvão, and Opukuro David-West conducted the experiments. Daerefa-a Mitsheal Amafabia conducted the formal analysis. Opukuro David-West, Diogo Montalvão, and George Haritos supervised the work.

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Research Article

Fracture of Parallel Strand Bamboo Composite under Mode I Loading: DCB Test Investigation

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This paper describes the experimental studies on Mode I fracture of parallel strand bamboo (PSB) by the double cantilever beam (DCB) test. R-curves based on the elementary beam theory and specimen compliance are proposed in order to overcome the difficulties to monitor the crack propagation during experiments. The results demonstrate that the energy release rate (ERR) is influenced by the specimen geometry, i.e., the specimen width and initial crack length. The ERR at the plateau level is similar for the range of the analyzed widths ($B = 20$, 40, and 60 mm), while it decreases with width increasing up to 80 mm and 100 mm. The energy release rate for PSB specimens would verge to a stable value with the width increasing up to a specific value, while the value of the energy release rate will be influenced by the initial crack length. Consequently, the DCB tests also show that the obtained R-curve in this study is not a material property.

1. Introduction

Parallel strand bamboo (PSB) is manufactured by parallelly gluing bamboo strands together under controlled temperature and pressure. Because bamboo strips are parallelly glued along the longitudinal direction and uniformly distributed over the transverse direction, PSB can be considered as a unidirectional and orthotropic fibrous composite, as shown in Figure 1 [1]. More recently, PSB wins growing interests as an alternative of wood composites for construction engineering in China due to its fast-growing feature and excellent structural performance. Aimed at the structural use, the failure modes of combined compression and bending PSB members were studied by Huang et al. [2–4]. It was found that the fracture along the fiber interfaces was one of the major failure modes of PSB composites. Once a PSB member subjected to an increasing external load, the initial deflects, such as microvoids and microcracks, could be advancing or growing into macrocracks and consequently result in catastrophic failure of structures. Therefore, the fracture of PSB composite is one of the major concerns in PSB structural design.

The objective of the present study is to investigate the Mode I fracture properties and to measure the fracture toughness of the PSB composite through DCB experiments, which is one of the necessities for establishing design allowable values used in damage tolerance analysis of PSB structures.

The double cantilever beam (DCB) test [5–7], a standard test procedure prescribed in ASTM D5528-13 [8], was employed as the test procedure in this study. The test uses a rectangular specimen with different widths, different initial crack lengths, and constant thickness. It contains a pre-implanted, nonadhesive insert or crack as an initial delamination. The opening load is applied perpendicular to the initial crack surface to induce the Mode I crack. Such a test analysis is based on the beam theory, and the fracture toughness can be measured through the energy release rate (ERR) for orthotropic fibrous composites. The ERR can be obtained by various methods, including direct area integration of loading-unloading curves or by means of

Figure 1: (a) Photograph of PSB. (b) Principal directions of PSB.

compliance measurements [9]. Since the data reduction of DCB test is easier than the other test approaches [10, 11], it is now the most popular method used to determine Mode I fracture toughness.

Recommended by ASTM and ISO standards [9, 12, 13], the energy release rate can be estimated through the change of compliance. Theoretically, it can be obtained by differentiating the specimen compliance, C, with respect to the crack length a, i.e., $G = (P^2/2B)(\partial C/\partial a)$ [14–18], where P is the applied load and B the width of the specimen. In the frame of linear fracture mechanics, the root condition is supposed to be fully built-in; hence, the compliance equation can be given by simple beam theory, i.e., $C = 8a^3/BEh^3$ [9]. Nevertheless, because of the fuzzy boundary conditions at the end of opening arms, the compliance equation $C(a)$ cannot be determined without controversy. On the other hand, the crack length cannot be measured either with desired precision. Therefore, various data reduction methods were developed to overcome this disadvantage, such as the area method, in which the compliance equation was built through data fitting for tests [19, 20]. Hashemi et al. [21] compared the different data reduction methods for obtaining the energy release rate of fibrous composites. They proposed that the correction of crack length was necessary in the beam theory-based method because the end of opening arms was not perfectly clamped. They developed the corrected beam theory (CBT), providing a correction in the crack length based on a compliance calibration. However, the fracture mechanism of fibrous composite is much complicated than that of brittle materials. The major aspect is that the microvoids coalescing, fine cracks extending, and fiber bridging lead to a large fracture process zone (FPZ) in front of the crack tip, which makes the clear location of the crack tip impracticable [22–25]. Furthermore, the development of the large FPZ in the crack-tip front delays the fracture of specimen, which consequently makes the built-in assumption no longer valid. It has been well recognized that the use of elementary beam theory with the built-in assumption at the crack tip has a significant error for calculating the energy release rate of fibrous composites [26]. To date, there is no method available to exactly identify the crack tip for the DCB test of fibrous composites. For this reason, extensive researches have been carried out to avoid locating the crack tip [27–32]. Among these researches, the concept of equivalent linear elastic fracture mechanics (LEFM) was widely accepted to deal with the fracture with a large FPZ [31]. According to the concept of equivalent LEFM, the increasing of compliance, owing to the development of FPZ or main crack propagation with FPZ, is attributed to the elastic crack which equals to the actual crack with its FPZ [33]. Therefore, the crack length can be estimated according to the associated compliance obtained by experiments. Furthermore, the R-curve can also be measured using only a monotonic load-displacement record.

In the present study, the data reduction method was based on the theory of equivalent LEFM. Compliance calibration was introduced to determine the crack length. The direct integral approach was adopted to calculate the energy release rate.

2. Data Reduction Scheme

Figure 2 illustrates the principle of evaluating energy release rate based on the theory of equivalent LEFM. In the schematic load-displacement curve (F, δ), two consecutive points $M_1(\delta(a_1), F(a_1))$ and $M_2(\delta(a_2), F(a_2))$ corresponding to two different crack lengths a_1 and a_2, respectively, are selected. Suppose that the crack extends from length a_1 to a_2 with a small amount of propagation increment Δa and that the load-displacement trajectory goes from M_1 to M_2; thus, the shadow area circulated by the dashed lines OM_1, and OM_2 and the segment of load-displacement M_1M_2 must be equal to the energy released as the crack growth of the extension Δa [19]. Therefore, the energy release rate can be calculated by

$$G_R = \frac{1}{b\Delta a}\left[\frac{1}{2}P(a_1)\delta(a_2) - \frac{1}{2}P(a_2)\delta(a_1)\right], \quad (1)$$

where P and δ stand for the applied load and associated displacement at loading position, respectively. The crack length $a_i\,(i = 1, 2)$ can be determined from the test compliance $C(a_i) = (\delta(a_i))/(P(a_i))$. According to the beam theory, the compliance can be calculated by [34]

$$C(a) = \frac{a^3}{E_L I} + \frac{6a}{5G_{LT}A}, \quad (2)$$

where E_L is the longitudinal elastic modulus, G_{LT} is the shear modulus in LT plane, i.e., the 12 or 13 plane as shown in Figure 1(b), I is the inertia moment of the cross section, and

Figure 2: Schematic diagram of the evaluation of the resistance to crack growth (G_R).

A is the area of cross section. Once the compliance $C(a_i)$is obtained by experiment and the crack length can be determined by resolving equation (2) or by numerical iteration in terms of the following equation:

$$a_i = \left[\frac{3E_L I\lambda C(a_i)}{2\left(1+\left(3E_L h^2/10G_{LT}a_{i-1}^2\right)\right)}\right]^{(1/3)}, \quad (3)$$

where $\lambda = \overline{C}_0/C_0$ is the calibration parameter to eliminate the system error of the experiment, in which C_0 is the test compliance corresponding to the initial crack length a_0 and $\overline{C}_0$ stands for the theoretical compliance corresponding to the initial crack length calculated by equation (2).

3. Experimental Investigation

3.1. Materials. The test PSB was provided by Feiyu Bamboo Products, Jianghxi, China. The material density was 1.26 g/cm^3, and the moisture content was 11%. Mechanical properties in parallel-to-grain direction, which are involved in test analysis, were pretested following the method recommended in ASTM D143-14 [35], and the results are collected in Table 1, where the subscripts L and T are the parallel-to-grain direction and perpendicular-to-grain direction, respectively, and v represents Poisson's ratio of PSB.

3.2. Specimen Preparation. The geometry of DCB specimen is shown in Figure 3. The dimensions of DCB specimen were determined referring to the standard test procedure addressed in ASTM D5528-13 [8]. In this method, the DCB dimension was designed to ensure the damage zone or nonlinear deformation developed along the delamination front and the stable crack growth can be achieved.

Totally four groups of specimens with different initial crack lengths and different widths were prepared, as illustrated in Table 2, in which each group consists of the same initial crack length and 5 types of specimens with different widths. A label system was designed to identify the specimen. In this system, the latter A, B, C, and D correspond to the initial crack length of 34 mm, 67 mm, 100 mm, and 134 mm, respectively. The number prior to the letter of the group name stands for the width of the specimen, as shown in Table 2. The initial crack was firstly introduced by a 1.5 mm thick saw kerf; afterwards, precrack with length about 1 mm was extended by using a cutting blade [10]. At the end of the DCB test, two bolt holes of 8 mm in diameter were drilled for the sake of joining DCB specimen to the actuator of the test machine, as illustrated in Figure 3.

Table 1: Mechanical properties of PSB composites in parallel-to-grain direction.

Items	E_L (MPa)	G_{LT} (MPa)	v_{LT}	v_{TL}
Mean	15363	4890	0.32	0.05
CV (%)	9.575	12.31	12.3	25.6

3.3. Test Procedure. The fracture tests were performed on a servo-hydraulic universal test machine of 20 kN capacity in room ambient circumstance. The test setup is illustrated in Figure 4(a). The specimen was joined to the load actuator through two steel rods of 7.5 mm diameter, as shown in Figure 4(b). Loading was controlled by the displacement of a moveable actuator at the speed of 1.0 mm/min. A microscope digital camera was mounted in front of the specimen to monitor the crack propagation and take the images of the crack tip. The applied load and the displacement at loading position were simultaneously recorded at a frequency of 10 Hz. The opening displacement at the initial crack tip was measured by using a clip-on gauge (COD gauge) symmetrically fixed at the two sides of the crack tip through embedded aluminum flakes (Figure 4(b)). The crack length during propagation can be observed by using the microscopic camera, as shown in Figure 4(b).

4. Test Results

4.1. Load-Displacement Curve. Figure 5 illustrates a typical load-displacement curve. Roughly three stages can be observed. The first one (stage I) is no-damage stage from origin to the proportional limit point where the curve deviates from its original direction. The load-displacement curve exhibits perfect linearity in this stage. Once the load exceeds the point of proportional limit (stage II), damage onset extends and coalesces to form the fracture process zone (FPZ) and consequently leads to the compliance augment and the curve continuously deviates from its original direction as the load increases. Hence, this stage can be understood as the stage of FPZ development. When the FPZ fully developed, the load-displacement curve would reach its critical point, where the load reached its maximum value and the crack propagation began. Hence, the third stage (stage III) is the crack propagation stage.

Figure 6 presents the load-displacement profiles of the specimens. For the specimens of Group A, the load-displacement curve sharply declines once the load reaches the maximum value, which indicates that the crack

Figure 3: The configurations of DCB specimens.

Table 2: Dimensions of DCB specimens.

Group	Label	a_0 (mm)	L (mm)	B (mm)	H (mm)	a_0 (L)	Number
	20-A			20			
	40-A			40			
A	60-A	34	350	60	50	0.097	3
	80-A			80			
	100-A			100			
	20-B			20			
	40-B			40			
B	60-B	67	350	60	50	0.191	3
	80-B			80			
	100-B			100			
	20-C			20			
	40-C			40			
C	60-C	100	350	60	50	0.286	3
	80-C			80			
	100-C			100			
	20-D			20			
	40-D			40			
D	60-D	134	350	60	50	0.383	3
	80-D			80			
	100-D			100			

Figure 4: (a) Experimental setup of the DCB test. (b) Schematic diagram of the DCB test.

propagation is unstable. For the specimens in Groups B, C, and D with longer initial crack length, the load-displacement profile slowly declines in post-summit segments. Furthermore, the load-displacement profile declined more slowly with longer initial crack length, which indicated that the stability of crack propagation is sensitive to the initial crack length. Therefore, enough initial crack length is needed for obtaining stable crack propagation. In the present study, the ratios of initial crack length to ligament length of Groups B, C, and D are 0.191, 0.286, and 0.383, respectively, and their crack propagates in a stable and ductile manner, which means specimens with the ratio of Groups B, C, and D can be used to determine the fracture properties.

The second factor is the width scaling which has a significant influence on fiber bridging in the wake of the crack, consequently affecting the delamination behavior. As shown in Figure 7, the load-displacement curves for partial specimens are presented with the same initial crack length but different widths. It can be concluded that the crack grows in a "thumbnail" shape [36] under constant or decreasing load, while the overall displacement is increased. The "thumbnail" theory means that the crack extends mainly from the center of the thickness of specimens, while the edge regions are plastically deformed. Therefore, with the width increasing, the energy release rate

Figure 5: A typical load-displacement curve.

Figure 6: Load-displacement curves of all tested DCB specimens with different initial crack lengths in the same width: (a) width $B = 20$ mm; (b) width $B = 40$ mm; (c) width $B = 80$ mm; and (d) width $B = 100$ mm.

Figure 7: Load-displacement curves of DCB specimens with different widths in the same initial crack length: (a) initial crack length $a_0 = 100$ mm; (b) initial crack length $a_0 = 134$ mm.

of DCB specimens for PSB composite would converge to a stable value.

It is worth mentioning that only partial specimens exhibit their load-displacement curves or R-curves in later analysis due to the similar tendency of these curves. Moreover, because of the manufacturing defects for some PSB specimens, the data cannot be used to analyze the energy release rate in the DCB test. Therefore, some specimens present two lines in partial load-displacement curves or R-curves.

4.2. Fracture Process Zone. Damage begins when the external load exceeds its proportional limit, which is characterized by local microcracks at the crack tip front zone. This damaged zone is comprised of microcracks between the grains or through the grains. The microcracks are consequently extended and coalesced to form macrocracks and local microcracks at the advancing crack tip front as the external load increases. This fracture process and damaged zone can be called fracture process zone (FPZ). Before the load reaches its maximum value, the crack interfaces are bonded by fiber bridges which either rupture or peel off from the crack surfaces. This fracture mechanism leads to a large FPZ which can be observed at the crack tip front before crack propagation, as shown in Figure 8. Due to the restriction of fiber bridges, the strain energy is smoothly consumed through the fibrous rupture or pull-off, which leads to stable crack propagation and a rising R-curve.

Several micrographs were taken from the tested specimens at the fracture surfaces using the scanning electron microscope (SEM), as shown in Figure 9. Figure 9(a) exhibits the typical fracture surface with a limited amount of fiber pull-out in the debonding plies, which is presented with higher magnitude in Figure 9(b). The pull-out process could have created a fibre-bridged zone in the wake of the advancing crack tip. During the whole experiment, the growth and eventual stabilization of this fibre-bridged zone could account for the tendency observed in later R-curves.

4.3. R-Curve Measurement. Energy release rate for each test specimen was measured by the method addressed in Section 2. The R-curves of partial DCB specimens with different widths and different initial crack lengths corresponding to specific thickness are presented in Figures 10 and 11. It was observed that the energy release rate increases up to a steady-state toughness after the initiation of delamination. As illustrated in Section 4.2, the fiber bridging effect during the whole crack propagation led to eventual stabilization of crack growths for all tested specimens. Meanwhile, the results also implied that the R-curve obtained in this study was not a material property, i.e., it depended on specimen geometry. Consequently, when the crack starts to propagate, the bridging zone is created. By developing the bridging zone length, the strain energy release rate increases up to the steady-state fracture toughness. The length between the initial crack length and crack length corresponding to the steady-state fracture toughness is called as the steady-state bridging zone length. From Figure 10, the value of the energy release rate was always affected by the initial crack length. The major reason may be that the steady-state bridging zone of PSB specimens was always influenced by the preimbedded crack.

From Figure 11, it can also be concluded that the energy release rate of DCB specimens for the PSB composite would verge to a stable value with the width increasing up to a specific value, which will be identified in the future DCB tests for PSB specimens. As illustrated in the abovementioned "thumbnail" theory, the width scaling has

Figure 8: The fiber bridging phenomenon of DCB specimens.

(a) (b)

Figure 9: (a) SEM micrograph of the DCB specimens' fracture surface for PSB composite. (b) Fiber pull-out with higher magnitude.

(a) (b)

Figure 10: R-curves of DCB specimens with different initial crack lengths on the same width: (a) width $B = 20$ mm; (b) width $B = 40$ mm.

a significant effect on fiber bridging in the wake of the crack, which make the crack grow in a "thumbnail" shape, i.e., the crack extends mainly from the center of specimens, while the edge regions are plastically deformed. Therefore, the energy release rate of DCB specimens for the PSB composite would converge to a stable value with the increasing width. Table 3 gives the average fracture toughness for partial test samples.

Figure 11: R-curves of DCB specimens with different widths in the same initial crack lengths. (a) Initial crack length $a_0 = 100$ mm. (b) Initial crack length $a_0 = 134$ mm.

Table 3: The average fracture toughness for partial test samples ($H = 50$ mm).

Specimens	a_0 (mm)	B (mm)	L (mm)	P (N)	G_R (J/m^2)
DCB-C	100	20	350	839	1740
		40		1535	1505
		60		2364	1550
		80		2149	630
		100		2952	820
DCB-D	134	20	350	745	2600
		40		1332	2570
		60		2081	1960
		80		2009	950
		100		2233	870

5. Conclusion

This paper has investigated an experimental procedure to obtain the energy release rate (ERR) using DCB tests for parallel strand bamboo (PSB) composites. The effects of specimen width and initial crack length on large-scale fiber bridging in Mode I fracture of unidirectional PSB composites were investigated. The R-curve measurements, for the range of the analyzed widths ($B = 20$, 40, 60, 80, and 100 mm) and initial crack lengths ($a_0 = 34$, 67, 100, and 134 mm), show that the energy release rate is influenced by both the width and initial crack length of specimens, in that the ERR at the plateau level decreased with an increasing width and initial crack length. Consequently, the DCB tests imply that the R-curve obtained in the present study is not a material property because it depends on the specimen geometry. However, the ERR at the plateau level is decreased to similar values as the width increases up to 80 mm and 100 mm. It may be concluded that the energy release rate would verge to a stable value with the width increasing up to more than a specific value, which can be clarified in the future DCB tests for PSB composites.

Nomenclature

B: Specimen width
C: Specimen compliance
a: Crack length
P: Applied load
G: Energy release rate
E: Young's modulus
h: Height of the cantilever portion, $H/2$
H: Specimen thickness
δ: Loading-line displacement
Δa: Crack extension
E_L: Elastic modulus in longitudinal direction
G_{LT}: Shear modulus in LT plane
I: Inertia moment of cross section
A: Area of cross section
λ: Calibration parameter to eliminate errors from experiments
$\overline{C}_0$: Theoretical compliance corresponding to initial crack length
C_0: Test compliance corresponding to the initial crack length
a_0: Initial crack length
υ: Poisson's ratio
L: Specimen length.

Data Availability

The data of the DCB test for PSB composites used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Three-Point Bending Fatigue Test of TiAl6V4 Titanium Alloy at Room Temperature

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A polycrystalline alpha-beta TiAl6V4 alloy in the annealed condition was used for the three-point bending fatigue test at frequency f~100 Hz. The static preload $F_{\text{stat.}} = -15$ kN and variable dynamic force $F_{\text{dyn.}} = -7$ kN to -13.5 kN were set as fatigue test loading parameters. The fatigue life S-N curve presented the stress amplitude σ_a as a function of a number of cycles to fracture N_f. A limiting number of cycles to run out of 2.0×10^7 cycles were chosen for the 3-point fatigue tests of rectangular specimens. In addition, the Smith diagram was used to predict the fatigue life. The alpha lamellae width has a significant influence on fatigue life. It is assumed that the increasing width of alpha lamellae decreases fatigue life. A comparison of fatigue results with given alpha lamellae width in our material to the results of other researchers was performed. The SEM fractography was performed with an accent to reveal the initiation sites of crack at low and high load stresses and mechanism of crack propagation for the fatigue part of fracture.

1. Introduction

The TiAl6V4 alloy has a relatively high resistance in cyclic loading [1–4]. However, the course of fatigue damage, however, depends on the content of the additive elements, microstructure, surface treatment, and size and type of applied stresses. The fatigue strength value for smooth samples is more than 50% of the tensile strength, but it is largely dependent on surface quality. The greater the roughness of the surface, or the surface saturated with oxygen or nitrogen, the worse the fatigue properties of titanium. The presence of a notch causes a fatigue reduction of 30–35%. The fatigue range for titanium alloys is reached at 10^6-10^7 cycles, but it is dependent on the load frequency [5].

Generally, the grain size of materials is a very important characteristic of fatigue life. Leyens and Peters in their work [2] present results that the reduction of grain size from 110 to 6 μm for commercially pure titanium increases the fatigue strength from 160 to 210 MPa. Likewise, the fatigue strength increases after increasing strength via work hardening. In addition, Gao et al. [6] and Moussaoui et al. [7] have performed an intensive study about influence α grain size, degree of age hardening, and oxygen content on fatigue life of TiAl6V4. They show that the fatigue properties of two-phase near $-\alpha$ and $\alpha + \beta$ alloys are strongly influenced by the morphology and arrangements of the two phases α and β (β grain size, colony size of α and β lamellae, and the width of α lamellae in lamellar microstructures). Another research was performed by Peters et al. and Wagner et al. [8, 9] report that reducing off the α lamellae width from 10 to 0.50 μm in lamellar microstructures raises the fatigue strength from 480 to 675 MPa.

They also show that reducing the grain size from 12 to 2 μm in equiaxed microstructures increases fatigue strength from 560 to 720 MPa [8]. For duplex structures, reducing the α lamellae width in a lamellar matrix from 1 to 0.5 μm leads to an increase in fatigue strength from 480 to 575 MPa [9].

For the fatigue process as itself, it is a well-known fact that most of the fatigue cracks initiate at free surface (due to surface roughness, oxide presence, or carbide particles). The fatigue process may also start from so-called "fish-eye," what is a typical structural phenomenon when fatigue crack

initiates close to free surface on inclusions (oxide or carbide particles). This phenomenon occurs only in special cases when a very high-cycle fatigue test (over 10^9 cycles to failure) is performed at low-stress amplitude. Zuo et al. [10] have confirmed this phenomenon on this type of alloy with bimodal and basket-weave microstructure. This kind of initiation was not confirmed at low-cycle or high-cycle fatigue of TiAl6V4 alloy. Wagner and Lütjering discussed three various fatigue crack initiation sites in their work [11]. They have shown that fatigue crack initiation sites are related to microstructure. The first type of the fatigue crack initiation site is for lamellar microstructure where fatigue cracks initiate at slip bands of the α lamellae as itself or at connection areas between α lamellae and primary β grain boundaries. The α lamellae width has significant effect on slowing down dislocation movement which is closely related to fatigue crack initiation, and on the contrary, it has influence on fatigue strength and yield stress too. The second type of the fatigue crack initiation site is for equiaxed structures. The fatigue cracks initiate in concert with slip bands inside α grains at these types of microstructures. Therefore, fatigue strength is affected by the grain size and on grain size is yield stress dependent as well. The third type of the fatigue crack initiation site is for duplex structures. The fatigue cracks may initiate in the lamellar matrix (interface the lamellar matrix/the primary α phase), or in the primary α phase as itself. Kuhlman [12] showed that the fatigue crack initiation site also depends on the cooling rate. Boyer and Puschnik et al. [13, 14] in their works have discussed the effect of the volume fraction and size of the primary α phase on fatigue crack initiation sites.

The influence of microstructure on fatigue properties of Ti6Al4V alloys at high-cycle fatigue is discussed in the works of Wu et al. [15] and Crupi et al. [16]. Their results prove the fact that various microstructures decrease (from bimodal, lamellar to equiaxed) high-cycle fatigue strength. About Ti6Al4V bimodal structures, the high-cycle fatigue strength is strongly affected by the primary α phase volume and grain size. In the initial stage, the high-cycle fatigue strength increases. However, it gradually decreases with increasing amount and size of the primary alpha phase. A similar effect of decreasing high-cycle fatigue strength can also be observed in the case of equiaxed or lamellar microstructures of Ti6Al4V alloy. Essentially, it can be generalized that the fatigue strength of the Ti6Al4V alloy decreases, either due to the increase in the alpha-phase grain or by increasing the width of the alpha-phase lamellae.

Another factor affecting the fatigue strength of the Ti6Al4V alloy is the load frequency. This issue was studied by Furuya and Takeuchi [17]. Based on their work, it is possible to state that the results of fatigue tests at ultrasonic frequency are similar, respectively, comparable to results performed at 100 Hz load frequency. However, this is only true if the fatigue crack has initiated below the alloy surface. In the case of the Ti6Al4V alloy, when the fatigue crack initiated on the surface, it was shown that the fatigue strength was higher in the tests performed at the ultrasonic frequency than in the conventional 100 Hz frequency. The mean stress effect on fatigue strength is commonly evaluated by a modified Goodman's or Smith diagram. For the Ti6Al4V alloy, according to modified Goodman's (or Smith) diagram, the fatigue strength is considered safe when reaching 10^7 at conventional load frequencies.

Morrissey and Nicholas [18] also studied the impact of load frequency on strain rate or temperature increase due to internal damping. They compare data from ultrasonic tests, servohydraulic test systems (~60 Hz), and electromagnetic shaker systems (400 Hz). According to their data comparison, there are no frequency effects. The comparison of S-N results at ultrasonic and conventional frequencies is shown in Figure 1.

The aim of this paper is to provide information about fatigue resistance of the TiAl6V4 alloy in the annealed condition at the three-point bending loading test at frequency f~100.0 Hz [19, 20] with the run-out number of cycles N_f in range 10^6 to 2.0×10^7 and to compare it with results in references [8, 9, 15] to show how the bending load and microstructure (α lamellae length and width or primary α (α_p) grain size) affect the fatigue life. There is an assumption that bending load should decrease fatigue life of alloy due to more complex stress course in the specimen. The fatigue test frequency influence or sample heating due to internal damping at 100 Hz is not expected according to references [17, 18].

2. Materials and Methods

The $\alpha + \beta$ mixed TiAl6V4 (GRADE 5) titanium alloy was used as experimental material. The alloy chemical composition and selected mechanical properties according to the certificate of quality and weight (the alloy was supplied by BIBUS Metals AG, CZ, with heat No. HX-032) are given in Table 1. The alloy was in the annealed condition. Boyer had described in his work [13, 21] four common heat-treatments for $\alpha + \beta$ alloys, namely, Ti6Al4V. According to its work, the most suitable treatment for increasing the fatigue properties and achieving the reasonable fatigue crack growth is mill annealing (MA or A). After this treatment, the strength of about 896 MPa and moderate fracture toughness 66 MPa·m^{-2} are achieved. The microstructure of experimental material is shown in Figure 2. It consists of α-phase lamellae which is considered as the hexagonal close-packed phase (HCP) presented at lower temperatures (up to 886°C) situated in β grains (body centred cubic, BCC, for temperatures from 886°C to melting temperature 1660°C) (Figure 2(a)). The arrangement of α-phase lamellae is presented more in detail in Figure 2(b), and as is obvious, it creates the "envelop" of aluminium solid solution in the base titanium matrix.

Samples for the three-point bending test were supplied by BIBUS METALS AG. Samples were cut to $11 \times 10 \times 50$ mm blocks. The specimen surface was sanded using a LaboPol-25 double-disk grinder, where the sample was ground on the grinding disk with a grain size of 400 (grinding in direction 1) and 600 (grinding in a direction perpendicular to direction 1) at 250 rpm. The sample was next rinsed in warm water and alcohol and dried after each step.

Figure 1: Comparison of S-N results at ultrasonic and conventional frequencies [18].

Two single-disk MTH polishers were used for polishing. In the first step, diamond paste D2 with a diamond grain size of 2 μm was added to the sample using a Mol roll at 300 rpm. The disk was moistened with alcohol during polishing, and the sample was polished against the counterclockwise rotation of the disc. In the second step, the D07 paste with a diamond grain size of 0.7 μm was applied using a Nap spin at 300 rpm. The disk was moistened with alcohol during polishing, and the sample was polished against the counterclockwise rotation of the disc. The sample was rinsed in warm water and alcohol and dried after each step. One side of the specimen was over polished due to the good observation of fatigue crack propagation. The surface roughness was measured by the MITUTOYO-ABSOLUTE-DIGIMATIC-HEIGHTGAGE device to compare how the surface roughness affects the fatigue life.

The fatigue test at the three-point bending was performed on a ZWICK/ROELL Amsler 150HFP 5100 resonance pulsator (Figure 3(a)) on 10 experimental specimens at room temperature 22°C ± 5.0°C. The specimens were numbered from 1 to 10. Vibrophore Amsler 150HFP 5100 is for fatigue testing of materials or components by applying sinusoid loads using the resonance principle (testing frequency range 35–300 Hz) with constant or variable amplitude (maximum force amplitude range is ±75 kN) and mean load (maximum mean load is ±150 kN). It provides fatigue testing of materials and components, e.g., fatigue tests according to DIN 50100 (S-N curve) for tensile stress and compressive and alternating stress ranges. Testing can be performed either force controlled (precision force measurement through calibration according to DIN 51 221 and US MIL Std. 1312 B) or strain controlled (two measurement channels for additional extensometers—force and strain; optional extensible through two measuring inputs). Tests can also be carried out under various environmental conditions, e.g., temperature range from cryogenic temperatures in a liquid nitrogen atmosphere to high-temperature testing up to 1200°C (for pushpull loading only). In addition, torsion and bending tests can also be carried out. The parameters of the test were set as follows: the static preload force $F_{\text{stat.}} = -15.0$ kN; the dynamic load was represented by dynamic force varied $F_{\text{dyn.}} = -7.00$ kN to -13.5 kN; frequency during the test was $f = 82.50$ Hz–108.6 Hz; stress cycle asymmetry $R < 1$; the number of cycles was set on value 2.0×10^7 representing the run-out. For titanium alloys, this value is considered as the fatigue limit (if the specimens withstand the 2.0×10^7 cycles at set stress amplitude without break, then the fatigue limit is reached).

The sample (of the size 10 mm × 11 mm × 50 mm) was positioned as shown in Figure 3(b), and it means that the center of the sample was loaded by the main force. To prevent specimens heating during the fatigue test, the specimens were cooled by an external fan. The stress amplitude σ_a is calculated according to the following equation:

$$\sigma_a = \frac{3 \times F \times L}{2 \times b \times h^2} \text{ (MPa)}. \quad (1)$$

where F is the applied dynamic force [N], L is the distance of supports which is 30 mm, b is the sample width which is 10 mm, h is the sample height which is 11 mm, and σ_a is the maximum amplitude (MPa). The S-N curve was drawn. The Smith diagram discusses the relation between mean stress σ_m and stress amplitude σ_a and provides information about the secure area of loading at various values of mean stress σ_m.

The three-point bending loading is not so common way to obtain fatigue life values in comparison to push-pull loading with a coefficient of asymmetry $R = -1$. Three-point bending fatigue loading provides more complex loading of specimens due to the shifting of mean stress to negative values with a coefficient of asymmetry $R < 1$. It means that the specimen is preloaded by negative static force with a higher value than the stress amplitude which results in more complex loading in the center of the specimen. There is an expectation that fatigue life shifts to lower values due to more complex loading compared to push-pull results.

Metallography specimens were prepared by cutting with MTH Micron 3000 precise saw and then mounting into bakelite mixture in Struers CitoPress 1 and finally ground and polished using Struers TegraSystem (TegraPol-15 and TegraForce-1) and a special program for titanium alloys. Grinding and polishing consist of a few steps: grinding with SiC sandpaper No. 320, followed by fine grinding with Largo grinding disk and emulsion Allegro Largo with 9 μm grains. The first step of polishing is performed with Dac polishing disk with an emulsion of Diap 3 μm grains and fine polishing with OP-S lubricant with 0.25 μm grains. The sample was rinsed in warm water and alcohol and dried after each step. This procedure was applied on specimens for light microscopy (LM) and for scanning electron microscopy (SEM) analysis as well.

After polishing, the sample was etched; the specimen surface was immersed into 10% solution of HF for 7 seconds and then rinsed with warm water and alcohol and dried by hot air. The microstructure was observed on light microscope

Table 1: Chemical composition in wt.% and selected mechanical properties of the TiAl6V4 alloy.

Chemical composition (wt.%)								
	Fe	C	N	H	O	Al	V	Ti
Specification condition	≤0.40	≤0.08	≤0.05	≤0.015	≤0.20	5.50–6.75	3.50–4.50	Reminder
Analyses	0.059	0.012	0.004	0.001	0.097	6.04	4.00	89.8

Mechanical properties				
	Ultimate tensile strength UTS (MPa)	Yield strength 0.2% (MPa)	Elongation (%)	Hardness HV
Specification condition	Min. 895	Min. 828	Min. 10	
Longitudinal	1011	951	15	266
Transverse	1023	971	14	

(a) (b)

Figure 2: The micrographs of TiAl6V4 alloy: (a) light microscopy, 10% HF acid etching, and (b) SEM observation, equiaxed microstructure with β grains, and α-phase lamellae.

(a) (b)

Figure 3: Three-point bending fatigue test: (a) ZWICK/ROELL Amsler 150HFP 5100 resonance pulsator and (b) sample position on pulsator at the three-point bending fatigue test.

Neophot 32. The grain size and α lamellae length and width were measured by using NIS-elements 4.20 metallography software.

The fractography analysis of specimens' surface after the fatigue test was performed using the TESCAN VEGA LMU II scanning electron microscope with the aim to detect the

Figure 4: The microstructure of experimental material—TiAl6V4 alloy: (a) the grain length measurement, (b) the width of grains measurements, and (c, d) α—lamellae measurements, 10% HF acid etching.

fatigue initiation site, fatigue crack propagation, and static fracture area analysis.

3. Results and Discussion

3.1. Microstructure. The microstructure of experimental material is documented in Figures 2 and 4. It consists of elongated grains of α lamellae in the transformed β phase (Figure 3). The average length of the grains is 479.1 μm (Figure 4(a)), and the average width is 159.3 μm (Figure 4(b)). The average length of the α lamellae is 24.42 μm (Figure 4(c)), and the average thickness is 2.530 μm (Figure 4(d)). This microstructure is in good agreement with β heat-treated TiAl6V4 alloy at 1020°C/20 min/FC with Widmanstätten α structure presented on prior β grain boundaries reported in Videhi Arun work [22]. The morphology of the Widmanstätten α phase may change from a colony of similarly aligned α lath to a basket-weave arrangement with an increase in the cooling rate or alloying content. From the present knowledge, the dependence of the fatigue life on the lamellar α phase is known. The larger the lamella dimensions are, the lower the fatigue life of the alloy is. To confirm this phenomenon, it would be advisable to perform tests for a more fine-grained structure and to compare the obtained values.

3.2. Fatigue Test Results. The S-N curve is shown in Figure 5. The experimental data were interpolated and the following coefficients of the Basquin [23, 24] equation were obtained:

$$\sigma_a = 912.92 \times N_f^{-0.044} \text{ (MPa)}. \quad (2)$$

From the S-N curve of the fatigue life for σ_a, it is clear that, after the translation of the resulting values of the number of cycles N_f to the failure and the stress amplitude σ_a by the power regression curve, the scattering of the obtained values is relatively small. Therefore, the maximum stress amplitude σ_a in tested material at a given number of N_f cycles can be predicted fairly accurately from equation (2).

The fatigue stress σ_c (defined as the highest stress at which the test bar = specimen is not broken even after the set number of cycles 2.0×10^7 has been exceeded) was calculated according to equation (1) as the medium value from three run-out stress amplitudes, and its value is 431 MPa. The

Figure 5: S-N curve for TiAl6V4 alloy at three-point bending fatigue loading at room temperature.

Table 2: Fatigue test results with used specimen numbering.

Spec. no.	Number of cycles N_f	Mean stress σ_m (MPa)	Stress amplitude σ_a (MPa)	Maximum bending stress σ_O (MPa)	Notice
1	2.0×10^7	613.0	287.0	901.0	Polished, run-out
2	1.8×10^7		389.0	1003.0	Polished, break
3	1.7×10^6		493.0	1107.0	Polished, break
4	1.6×10^5		552.0	1166.0	Polished, break
5	2.0×10^7		450.0	1064.0	Polished, run-out
6	1.2×10^6		477.0	1090.0	Polished, break
7	1.6×10^6		460.0	1074.0	Polished, break
8	2.0×10^7		454.0	1068.0	Polished, run-out
9	5.7×10^6		459.0	1073.0	Polished, break
10	1.2×10^7		455.0	1069.0	Polished, break
11	5.3×10^4		455.0	1068.0	Non-polished, not included in S-N curve
12	6.4×10^4		459.0	1073.0	Non-polished, not included in S-N curve

results for all specimens used at the fatigue test with specimen numbering are shown in Table 2.

Results obtained after the three-point bending fatigue test with load cycle asymmetry $R < 1$ were compared to results of Peters et al. and Wagner et al. [8, 9], Wu et al. [15], and Morrissey and Nicholas [18]. They have used common push-pull loading at $R = -1$. The results comparison shows a difference about 30.0 MPa–40.0 MPa lower in fatigue life and stress amplitude at run-out. This result shows a more complex character of fatigue specimens loading than simple push-pull loading.

The three-point bending loading includes a compression loading and a tension loading as well, and the specimen is subjected not only to direct stress but to bending moment with increasing value when approaching the specimen centre.

The influence of the mean stress σ_m on the fatigue stress σ_c is expressed by the Smith diagram [17, 25] in Figure 6. The region bounded by red indicates the area of the stress, where there is no break even after the theoretically unlimited number of cycles at the given mean stress σ_m and the amplitude of the stress σ_a cycles. The diagram is designed for the compressive stress area, and therefore, it is advantageous for its construction to use the values of the yield bearing strength and ultimate bearing strength [26] that involve a

Figure 6: Smith fatigue life diagram [25] for the TiAl6V4 alloy.

more complex load. It can be seen from the diagram that the fatigue resistance of the TiAl6V4 alloy increases with the decreasing amplitude of the stress σ_a.

Based on TiAl6V4 titanium alloy analyses with grains with an average length of 479 μm and a width of 159 μm, and

(a) (b)

Figure 7: Overall view on fracture surface, initiation sites are marked by arrows: (a) spec. no. 9, $F_{dyn} = -11$ kN; $\sigma_{Omax} = 1073$ MPa; $\sigma_a = 459$ MPa and (b) spec. no. 3, $F_{dyn} = -12$ kN; $\sigma_{Omax} = 1107$ MPa; $\sigma_a = 493$ MPa, SEM.

α lamellae with an average length of 24.4 μm and a width of 2.53 μm, it can be concluded that the fatigue interval σ_c is strongly dependent on the stress amplitude σ_a. According to the established rules obtained from numerous experiments [23, 24], the fatigue limit σ_c at the load push-pull and $R = -1$ corresponds to approximately 50% of ultimate tensile stress. Based on the constructed Smith diagram, it can be assumed that, for $\sigma_m = 0$ MPa, σ_c~650 MPa and UTS~1300 MPa, which corresponds to our calculated value UTS = 1323 MPa.

3.3. SEM Fractography. Figure 7 shows the fracture surface where the fatigue region, static break area, and a fatigue crack initiation site are marked (black arrow). The diminution of the fatigue region (Figure 7(b)) with the increasing value of the dynamic loading force and the stress amplitude σ_a as well is visible when macro-fractographic images are compared.

In Figure 8, the images of samples No. 9 and 7 of the fatigue crack initiation region are shown (Figures 8(a) and 8(b)). The fatigue crack in both cases initiated at the free surface of the polished samples at the sites of the highest concentration of stress. Major crack propagation proceeded from the free surface by transcrystalline cleavage of TiAl6V4 alloy grains with the cleavage facets observed at the crack initiation site. The cleavage facets of samples 9 and 7 are shown in Figures 8(c) and 8(d).

Micro-fractographic images of cleavage facets created by transcrystalline cleavage of TiAl6V4 alloy grains along the direction of propagation of the magistral fatigue crack are notable in Figure 9. In Figure 9(a), the transcrystalline cleavage facet with river morphology is visible. The river reliefs on facets are created due to plastic deformation that preceded the crack formation and its growth, presence of grain boundaries, or α-phase lamellae. The origin of the rivers also means energy consumption, which slows down the rate of propagation of the fatigue crack tip. In another case, especially at higher stress amplitude, the transcrystalline cleavage failure of β grains with a higher degree of cleavage has occurred (Figure 9(b)).

In the area of the stable fatigue crack propagation, the striations (Figure 10(a)) are visible on the surface, indicating the position of the crack tip at the given moment and creating ridges spreading from the initiation site. These ridges are perpendicular to the direction of magistral fatigue crack propagation. Another characteristic feature of the fatigue process is the secondary crack parallel to the advancing fatigue crack front documented in Figure 10(b).

From the comparison of the individual micrographs, it is clear that the amount of secondary cracks increases with the rising value of the stress amplitude. The change in the direction of the striation propagation due to the change in the direction of the magistral fatigue crack growth is shown in Figure 11(a). The area of static failure documented in Figure 11(b) is characterized by a transcrystalline ductile fracture with dimple morphology. A ductile fracture occurs by coalescence of microcells that nucleate at the grain boundaries, secondary-phase particles, or inclusions. The orientation of the holes varies with the orientation of the applied stress and β grain orientation to the applied load as well.

The fractography analysis revealed that the polished surface of samples had a single initiation site just below the surface of the sample where the characteristic of initiation and propagation of the fatigue crack was the transcrystalline cleavage of TiAl6V4 alloy β grains, which is supported by images of transcrystalline cleavage facets in the fatigue crack initiation region. The smaller the area of fatigue fracture was, the higher the amplitude of the stress σ_a was. Striations and secondary fatigue cracks, which are features of fatigue fracture, have also been observed in this area. The transcrystalline cleavage failure was caused by loss of coherency at the α lamella interface and the transformed β grain matrix.

Figure 8: Initiation sites of fatigue crack on the free surface: (a) spec. no. 9, (b) spec. no. 7. Cleavage facets made by transcrystalline cleavage failure of the TiAl6V4 alloy grain at the fatigue crack initiation sites: (c) spec. no. 9; (d) spec. no. 7, details of the initiation site, SEM.

Figure 9: Details of transcrystalline cleavage failure of TiAl6V4 grains: (a) transcrystalline cleavage facet with river morphology; (b) transcrystalline cleavage failure of grains with a higher degree of cleavage, SEM.

(a) (b)

Figure 10: Striation (a) in a perpendicular direction to fatigue crack propagation and secondary fatigue cracks (b) parallel to the advancing crack front, SEM.

(a) (b)

Figure 11: Change in the direction of striation propagation (a) and static transcrystalline ductile fracture with dimple morphology (b), SEM.

3.4. Influence of Surface Roughness on Fatigue Life and Hardness Measurements. The surface quality has a significant influence on the fatigue life of the α-β TiAl6V4 alloy. In the fatigue test, all specimens (except of two which of course were not included in S-N curve results) were polished. Polished specimens have surface roughness $R_a = 0.252$–$0.265\,\mu$m. Specimen No. 8 at loading $\sigma_{Omax} = 1068$ MPa withstands the 2.0×10^7 cycles (it was run out, specimen without breaking). Specimen No. 9 at $\sigma_{Omax} = 1073$ MPa withstands the 5.75×10^6 cycles (specimen with break). For nonpolished specimens with a surface roughness $R_a = 3.95$–$5.12\,\mu$m, the number of cycles to failure has dramatically decreased. For specimen No. 11, it was 5.32×10^4 cycles at the same stress level as specimen No. 8, and for specimen No. 12, it was 6.39×10^4 cycles at the same stress level as specimen No. 9. The fatigue life or the number of cycles to failure was decreased by about 74% for specimen No. 11 and about 97% for specimen No. 12.

The maximum bending stress $\sigma_{Omax} = 1090$ MPa which is obviously higher then ultimate tensile stress UTS = 1011 MPa (in the longitudinal direction) reported for this material in the material list was applied at the fatigue test. To explain this phenomenon, the hardness measurements of material were performed and the results of measurement are shown in Table 3. The values are the mean values of four hardness measurements.

There is a relation between the hardness and the material toughness, and it is expressed by using equation (3), where k is the coefficient that depends on the material type, and for titanium alloys, $k = 3$-4. The volumes of the k coefficient for

Table 3: TiAl6V4 hardness measurement.

	Material list hardness 266 HV		
Average value	VICKERS 346 HV 10/6	BRINELL 299 HBW 5/750	ROCKWELL 32 HRC

various material types and their effect on final UTS were discussed more in detail in the work of Tabor et al. [27] and Zhang et al. [28]:

$$\text{UTS} = k \times \text{HV (MPa)}. \quad (3)$$

For our experimental material TiAl6V4 alloy, $k = 3.85$ based on calculation from values provided by the supplier. According to this, the calculated ultimate tensile stress UTS = 1323 MPa. This value is over 301 MPa higher than that reported in the material list. This difference in material toughness and hardness is possible due to secondary hardening, which was not reported in the material list.

4. Conclusions

The titanium alloy TiAl6V4 was subjected to three-point bending fatigue loading at a frequency ~100 Hz for a high number of cycles 2.0×10^7 at room temperature 22°C ± 5.0°C. It was shown that fatigue crack initiates from free surface and has single initiation at lower amplitudes of stress σ_a and multiple initiation sites when higher stress amplitude σ_a was used.

The fatigue life was reached after a 2.0×10^7 number of cycles, and it was set on $\sigma_c = 431$ MPa. This value is about 30 MPa–40 MPa lower compared to common push-pull fatigue results on this alloy with similar equiaxed microstructure and α lamellae width as reported in experiments of other authors. However, it is necessary to take into count the different loading modes, specimen shape, and surface finishing of samples used at push-pull loading. The comparison with push-pull loading is performed due to the lack of data about the three-point bending fatigue test for this alloy in English.

The results show that a three-point bend is more complex loading and is more suitable for obtaining a fatigue life for advanced materials compared to the commonly used push-pull loads. The three-point bending loading includes both compression loading and tension loading, and a specimen is not subjected to direct stress only for the whole cross section but to the bending moment with increasing value when approaching the specimen center.

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request. The data used for Figure 1 and equation (3) are available in [18] and [27, 28], respectively.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

Acknowledgments

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Research Article

Quantitative Anisotropic Damage Mechanism in a Forged Aluminum Alloy Studied by Synchrotron Tomography and Finite Element Simulations

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A highly anisotropic toughness behavior has been revealed on a forged AA6061 aluminum alloy by toughness tests with CT specimens. The toughness values with specimens loaded on the longitudinal direction are larger than that loaded on the transverse direction due to the anisotropic shape and distribution of coarse precipitates induced by the morphological anisotropy of grains during forging process. Synchrotron radiation computed tomography analysis on as-received material and arrested cracks revealed different fracture modes for the two loading configurations. The damage mechanism has been validated by finite element simulations based on the Gurson–Tvergaard–Needleman micromechanical damage model with different sets of damage parameters for the two loading configurations obtained from quantitative void volume fraction analysis on SRCT data, in situ SEM experiments, and SRCT microstructural analysis.

1. Introduction

Aluminum alloys are often used in industrial structures for its light weight, its corrosion resistance [1], and its mechanical properties. Toughness is a crucial mechanical property in the design and use of industrial metal components. The toughness is frequently appeared to be anisotropic for forged aluminum alloys. The study on the anisotropic feature helps engineers to predict the lifetime of industrial parts. The origin of the anisotropy of damage is associated with microstructural anisotropy [2–4].

Several models of anisotropic ductile damage, taking into account the microstructural aspects, exist in the literature. These models are based on either the Gurson micromechanical model [5] or its extension developed by Tvergaard and Needleman (named GTN model) [6] or the Lemaitre phenomenological model [7, 8]. The parameters of the GTN model have been identified by 3D synchrotron laminography by Shen et al. [9] in AA6061 aluminum alloy sheets loaded on the transverse direction. The anisotropic behavior was not taken into consideration in this article. Several authors have modified these models by taking into account the coalescence of anisotropic cavities [10–14] based on the work of Thomason [15]. These studies show an increase in the crack propagation resistance in the case where the cavities (or coarse precipitates) are elongated along the initial loading direction. This anisotropy of the microstructure delays the coalescence of the cavities as the intercavity or interprecipitate distance in the path of the crack is larger. However, a limitation exists in these models: only coalescence by internal necking [15] is taken into account in these models, but the coalescence of microcavity-generated ligament on a

second population of the precipitates is frequently observed in this material [16, 17].

In this article, the initial microstructure of the alloy was analyzed by synchrotron radiation computed tomography (SRCT). These observations were carried out at the European Synchrotron Radiation Facility (ESRF) on the ID19 beamline. Samples relating to deformed states correspond to parallelepipeds in CT specimens in the region of ductile tearing. Two loading configurations have been studied: LS and TL. We will discuss later in this article these test configurations. The isotropic GTN damage model was used with two parameter sets to represent each configuration taking into account the initial microstructure of the alloy, the germination of cavities on coarse precipitates, the growth of cavities, and two mechanisms cavity coalescence: coalescence by internal necking and coalescence ligament. Cavity nucleation on a secondary population of precipitates was taken into account leading to coalescence mechanism ligament. The model parameters are determined by experimental analysis. The results were then compared with the experimental curves of fracture toughness tests.

2. Experimental

A forged AA6061 alloy was used in this study whose chemical composition is given in Table 1. The material used was in the T6 temper (solution heat treated at 530°C during 3 h and water quenching and aged at 175°C during 12 h). The heat treatment was defined to obtain the maximum yielding stress. More details on the manufacturing process, the microstructure, and mechanical properties of the material are described in [9, 18].

The forging direction, the long transverse direction, and the short transverse direction are referenced to the letters L, T, and S, respectively.

Tensile tests have been performed on smooth round specimens with an initial length of 50 mm and diameter of 10 mm. These tests have been conducted through L and T loading directions with a loading rate of $10^{-4}\,\mathrm{s}^{-1}$.

Toughness tests have been carried out on CT12.5 specimens (Figure 1) to investigate the fracture properties according to the ISO standard [19]. Two loading configurations have been conducted: LS and TL. The first letter represents the loading direction and the second the crack propagation direction. A 2.5 mm precrack was introduced by crack-length-controlled fatigue cycling [20]. Tests were performed on a servohydraulic machine with a loading rate of 8.3 μm/s. At least three tests have been performed in each configuration with only a small dispersion of results.

After being taken in the crack initiation and propagation region on arrested CT specimens by electrical discharge machining [21] (Figure 1), samples were scanned by synchrotron radiation computed tomography (SRCT) installed at beamline ID19 [22] of the European Synchrotron Radiation Facility (ESRF) in Grenoble, France. We followed the methods of Shen et al. described in [9]. We used a monochromatic beam of 19 keV X-ray energy. Volumes were reconstructed from angularly equidistant 2000 projections with an exposure time of each projection of 100 ms. A voxel size of 0.7 μm was chosen. Details of the experimental method and image reconstruction are given in [23, 24].

Table 1: Chemical composition of AA6061 alloy (wt.%).

AA6061	Si	Mg	Fe	Cr	Cu	Mn	Zn	Ti
Wt.%	0.65	1.01	0.24	0.18	0.30	0.09	0.20	0.02

For image analysis performed on as-received material, the analyzed volume is $700 \times 700 \times 700\ \mu\mathrm{m}^3$ corresponding to the representative elementary volume (REV). Only objects with a size larger than $2.1 \times 2.1 \times 2.1\ \mu\mathrm{m}^3$ are considered. For the analysis of arrested CT samples, after the reconstruction of tridimensional images, cracks were first binarized with the morphological algorithm "connected threshold growing" by using the software *ImageJ* and then analyzed using a "sum along ray algorithm" [2, 25, 26]. A Visualization Toolkit (VTK) software routine was used to render the three-dimensional (3D) datasets and produce the 3D images. The aim was to precisely determine and quantify the local crack characteristics such as opening within the 3D volumes.

3. Experimental Results

3.1. Microstructure. Figure 2 shows the microstructure of the as-received material obtained by 3D SRCT analysis. Two types of precipitates at the micrometer scale are present in this material: coarse Mg_2Si and iron-rich intermetallics [18, 27] with, respectively, their volume fraction of $0.43 \pm 0.11\%$ and $0.59 \pm 0.15\%$. Little porosity can be observed with volume fraction less than 0.05%.

The granular structure of the material has been analyzed by SRCT on the gallium-wetted sample. In fact, when liquid gallium is brought in contact with aluminum, gallium can penetrate the grain boundaries. Gallium has a very high X-ray attenuation coefficient compared to aluminum; this renders gallium layers visible by SRCT [28]. Figure 3 shows the superposition of reconstructed tomography images performed on the sample before and after gallium wetting. Coarse precipitates and grain boundaries can clearly be seen owing to the local segregation of gallium. It is observed that both coarse Mg_2Si precipitates and iron-rich intermetallics are mostly distributed on grain boundaries.

The image analysis conducted on tomography data has revealed the Feret diameters [29] of precipitates defined as the diameters derived from the distance of two tangents to the contour of the particle in L, S, or T direction, referred as F_L, F_S, and F_T, respectively. The average values for all precipitates are listed in Table 2. Due to the forging process, these coarse precipitates have a slightly elongated shape (Figure 2).

To analyze the spatial distribution of the particles, the Voronoi diagrams [30] are often used that decompose the threshold image by cells in each of which contains a particle. The particles are assimilated as a point in this method so that the elongated shape of the particles is not taken into account.

Figure 1: Geometry of CT specimens and position of SRCT samples (red parallelepiped).

In this study, the Voronoi cell diagram is computed on the 3D image by the watershed transform [31, 32] to overcome the limit.

Figures 4(a) and 4(b) show, respectively, 2D sections of Voronoi cells computed by the watershed transform associated with coarse Mg_2Si precipitates and iron-rich intermetallics. The Feret diameters of cells were quantified and are shown in Table 2 for both coarse phases through *L*, *S*, and *T* directions. Alignment of precipitates is shown in Figure 2 through the *L* direction. It is expected that the Voronoi cells are "compressed" along this direction as the interprecipitate distance is smaller in this direction. Nevertheless, this conclusion cannot be drawn for coarse Mg_2Si precipitates, which is probably due to the presence of threshold noises (small objects). This conclusion is however observed for cells of iron-rich intermetallics with a flat shape, which means an alignment of these intermetallics along the *L* direction.

3.2. Tensile Tests on Smooth and Notched Specimens. Figure 5 shows the results of tensile tests on smooth specimens loaded through *L* and *T* directions. The yield strength (YS) and ultimate tensile strength (UTS) are approximately identical for the two loading directions. The fracture surfaces for both loading directions are spherical, which indicates an isotropic behavior of plasticity. However, the ductility, i.e., fracture strain is significantly different, with specimen loaded through the *L* direction is 3 times higher than that loaded through the *T* direction.

3.3. Toughness Tests. Figure 6 shows the results of toughness tests for LS and TL configurations in terms of the force as a function of the crack mouth opening displacement (CMOD). The maximum load is higher for the LS configuration (~4.7 kN) as compared to the TL configuration (~3 kN).

3.4. Fractography of CT Specimens. Fracture surfaces have been analyzed by using the scanning electron microscope (SEM) for specimens on LS and LT configurations (Figure 7). It is found that most zones are covered by dimples at 10 to 30 μm with the presence of Mg_2Si and iron-rich intermetallics (Figure 7(a)), interconnected by submicron dimples or shear zones (Figure 7(b)). Whatever the test configuration, dimples are spherical which means that the growth of cavities is isotropic in the fracture plane.

For the specimen loaded in the TL configuration, the crack propagates through the direction of alignment of coarse precipitates. The fracture mode is primarily intergranular due to the intergranular distribution of coarse precipitates. The absence of transgranular coarse precipitates requires the sudden crack bifurcation during its propagation to join another alignment of intergranular precipitates. This

Figure 2: Reconstructed 2D images obtained by SRCT showing the microstructure of the as-received material.

Figure 3: Reconstructed 2D image showing coarse Mg_2Si precipitates, iron-rich intermetallics, and grain boundaries obtained by superposition of tomography images performed on the sample before and after gallium wetting. "Rings" artifact is also visible on the image.

bifurcation is also observed by Achon and Di Russo [33, 34], which occurs along cliffs perpendicular to the propagation plane (Figure 7(a)). Dimples are close to each other and well defined as the void coalescence takes place primarily by internal necking.

In the case of a LS test configuration, crack propagates perpendicular to the plane of alignment of coarse precipitates and therefore gets through grains in their thickness direction. Ductile dimples are smaller connected by smooth areas of a size close to the width of grains (Figure 7(b)). In the smooth area, submicron-sized dimples could be observed with the presence of a second population of precipitates much smaller. The authors showed that this second population of precipitates is probably homogeneously distributed dispersoids of chromium and manganese with a size between 50 nm and 300 nm [35–38]. The fracture mode for the LS configuration is therefore the combination of intergranular and transgranular modes. The primary void coalescence occurs then by shear band with a presence of void sheeting [16].

3.5. SRCT Study of Fracture Mechanism in Interrupted CT Specimens. Tomography scans of interrupted cracks allow observation of the fracture mechanism during the crack initiation and propagation in 3D as well as the subsequent evolution of the fracture process in front of the crack tip. In this study, the CT specimens have been interrupted immediately after the maximum of loads is reached during toughness tests in two configurations (TL and LS) where samples in the crack initiation and propagation region have been taken and observed by SRCT (Figure 1).

Figures 8(a) and 8(b) show the local crack opening map (COD map) of those cracks for TL and LS loading configurations. In both configurations, we distinguish the area of the fatigue precrack region and the ductile crack

Table 2: Feret dimensions of coarse precipitates and Voronoi cells through L, S, and T directions. Deviation is estimated to about 25% owing to the threshold uncertainty.

	Mg_2Si			Iron-rich intermetallics		
	F_L (μm)	F_S (μm)	F_T (μm)	F_L (μm)	F_S (μm)	F_T (μm)
Precipitates	8.5 ± 2.1	6.6 ± 1.6	5.2 ± 1.3	8.7 ± 2.1	8.3 ± 2	6.3 ± 1.5
Voronoi cells	94 ± 23	95 ± 23	104 ± 26	67 ± 16	83 ± 20	104 ± 26

Figure 4: Reconstructed 2D images of the Voronoi cells [30] computed by the watershed transform [31, 32] superposed with threshold precipitates: (a) coarse Mg_2Si precipitates; (b) iron-rich intermetallics.

Figure 5: Tensile tests: experimental and simulation with elastoplastic hardening law (FEM).

Figure 6: Experimental toughness tests and finite element simulations (FEMs) for configuration LS and TL.

propagation. Only damage percolated with the crack is represented, whereas isolated damage in front of the crack is not shown in the COD map.

In the TL configuration (Figure 8(a)), the crack is continuously formed by spherical dimples with a size related to isolated cavity a head of the crack. The interdamage distance is low (about 50 microns) indicating that coalescence by internal necking is dominant. Some narrow and opened ligaments are also recognized which can be referred to cliffs observed in fracture surfaces (Figure 7(a)).

In the LS configuration (Figure 8(b)), the fatigue precrack is more open than in the TL configuration indicating a better resistance to the crack propagation in this configuration. We note an absence of well defined crack tip. The crack propagation zone is not continuous with the presence of bridges between clusters of cavities.

Figures 9(a) and 9(b) show the 2D sections through white lines in Figure 8. These sections confirm the conclusions obtained in fractography analysis (Figure 7). For TL configuration (Figure 9(a)), the crack propagates parallel to the direction of alignment of coarse precipitates. The failure mode is mainly intergranular. For LS configuration (Figure 9(b)), in addition to the intergranular damage mode, the transgranular narrow ligaments are observed between void clusters.

Figures 9(c) and 9(d) show the 2D sections through green lines in Figure 8. For TL configuration (Figure 9(c)), the crack is continuous with small-coalesced cavities. The void coalescence by internal necking is early and rapid with a very limited void growth stage. For LS configuration (Figure 9(d)), continuous crack propagation is not observed but large clusters of cavities at grain boundaries are linked by narrow ligaments. Void coalescence takes place later by a second population of voids initiated on dispersoids of chromium and manganese [35–38].

Analysis presented in Figure 8 in terms of crack opening displacement is calculated in terms of void volume fraction in regions of interests (ROI) equal to $140 \times 140 \times 70\,\mu m^3$. Figure 10 shows a map of average void volume fraction in each ROI in the crack propagation plane, the value selected and shown in the figure corresponds to the maximum value in the loading direction (normal to the propagation plane).

Figure 11 illustrates evolution of void volume fraction averaged through width of specimens in function of distance in the crack propagation direction. The two values indicated by arrows represent the average fraction measured at crack tip of the two configurations (dotted lines in Figure 10). The position of the crack tip is chosen at the maximum crack propagation for which all void volume fraction values are nonzero. These values, 1.5% for TL configuration and 5.6% for LS configuration, will be used as critical void volume fraction at failure in the modeling.

4. Modeling

4.1. Parameter Identification. The material model is described in Appendix A including the Voce hardening law [39] and the Gurson–Tvergaard–Needleman (GTN) damage model [5, 6] where the material damage is associated with a void volume fraction. The main material characteristic parameters included in the models are (i) parameters of the hardening law, (ii) preexisting voids and void nucleation parameters, (iii) void growth parameters, and (iv) void coalescence parameters.

The hardening curve is experimentally measured using tensile tests performed on smooth specimens. Beyond uniform elongation of the specimen, the true stress/strain tensile relationship was determined by using the Bridgman correction [40]. The parameters of the isotropic Voce hardening law have been fitted to the experimental data, which leads to the values presented in Table 3. The Voce law only describes stage III. The adjustment of the law is in good agreement with the experimental data (Figure 5). Stage IV could be neglected.

Figure 7: Fracture surfaces of toughness specimens in the crack propagation zone for (a) TL and (b) LS configurations.

Figure 8: Representation of the local crack tip opening via a "sum along ray method" for (a) TL and (b) LS configurations (with lines indicating locations of the 2D sections).

As concluded obviously by synchrotron laminography in situ analysis [9], the coarse Mg_2Si precipitates start to be damaged in the very early loading stage. This has been confirmed in in situ SEM tensile tests (not shown here) where the coarse Mg_2Si precipitates fracture already in the elastic part of the stress-strain curve. As a consequence, these precipitates are considered as preexisting voids f_0 in the simulations.

Cavities nucleating on iron-rich intermetallics are taken into account in the void nucleation parameters. The volume fraction of iron-rich intermetallics obtained by SRCT is considered as the maximum value of the voids that could be nucleated f_n. The two other parameters of the void nucleation law (ε_{n0} and s_{n0}) are determined by fitting the evolution of density of cavities measured by image analysis during in situ SEM tensile tests and the void nucleation law (not presented here).

The values proposed by Tvergaard and Needleman [6] ($q_1 = 1.5$ and $q_2 = 1$) are used as void growth parameters in the GTN model.

The quantitative analysis described in Section 3.5 provides us the critical void volume fraction at failure for both configurations TL and LS. The crack extends over one element when void volume fraction in ROI has reached this critical value $f_f = 1.5\%$ for TL configuration and $f_f = 5.6\%$ for LS configuration throughout the entire element. It is noted that this value is obtained experimentally with an ROI of $140 \times 140 \times 70\,\mu m^3$ in front of the crack. The same element size must be used in finite element analysis, which is $70 \times 70 \times 70\,\mu m^3$ with the symmetry condition around the center of specimen.

For the present material, two reasons lead us to consider that the material loses its strength once the void coalescence begins. First of all, we cannot observe any coalescence

Figure 9: 2D sections of the SRCT data through white lines in Figure 8 for (a) TL and (b) LS configurations as well as through green lines in Figure 8 for (c) TL and (d) LS configurations.

Figure 10: Void volume fraction map in region of interests (ROI) = $140 \times 140 \times 70\,\mu m^3$ for configurations (a) TL and (b) LS.

Figure 11: Evolution of mean void volume fraction, averaged through the width of the specimen, in function of the distance in the direction of crack propagation. The two values in red and blue represent the average fraction measured at crack tip (dotted lines in Figure 10).

Table 3: Parameters identified for SRCL finite element simulations.

(i)	(ii) Voce hardening law				
	E in GPa	ν	σ_0 in MPa	σ_s in MPa	b
	70	0.33	288	375	12
(iii)	(iv) Preexisting voids and void nucleation				
	f_0	f_n	ε_{n0}	s_{n0}	
	0.43%	0.59%	10%	0.1	
(v)	(vi) Void growth				
	q_1	q_2			
	1.5	1			
(vii)	(viii) Void coalescence				
	f_c	δ	f_f		
TL	1.5%	∞	1.5%		
LS	5.6%	∞	5.6%		
(ix)	(x) Element sizes				
	$(140 \times 140 \times 140)\ \mu m^3$				

between voids without involving the crack, which means that the crack begins to propagate at the same time or earlier than void coalescence at the present stress state. Secondly, it was observed previously by ex situ SRCT tensile tests that the damage evolution is extremely rapid after the coalescence has started. Therefore, the critical void volume fraction at coalescence f_c is assumed to be the same as the critical value at which the crack extends f_f, and the acceleration factor $\delta = ((f_u - f_c)/(f_f - f_c))$ is thus infinite.

To sum up, the parameters used for the finite element simulations are resumed in Table 3.

4.2. Model Predictions. The finite-element simulation technique is described in Appendix B. The experimental study showed a strong anisotropy of toughness for the forged AA6061 aluminum alloy. The toughness is significantly lower in TS configuration than in LS configuration. The two configurations have been simulated by using different sets of damage parameters of the GTN model.

Figure 6 shows load-CMOD curves for both experimental results and finite element simulation results for CT specimens in TL and LS configurations. Only the parameters of critical void volume fraction at coalescence and failure are different for the two configurations, i.e., $f_c = f_f = 1.5\%$ for TS configuration and $f_c = f_f = 5.6\%$ for LS configuration.

In the case of TL configuration (red symbols and curve in Figure 6), the maximum load and the opening displacement are well predicted. It should be noted that no parameter is adjusted to obtain a good agreement between simulations and experimental results.

In the case of LS configuration (green symbols and curve in Figure 6), the simulation has overestimated the maximum load because the critical failure parameter $f_f = 5.6\%$ is too high. This can be improved by a more precise identification of the parameters. As we have shown in Sections 3.4 and 3.5, the intergranular and transgranular failure modes coexist in this configuration. Coalescence between void clusters takes place on narrow ligaments across grains involving the nucleation, growth, and coalescence of a second population of precipitates, i.e. dispersoids of chromium and manganese. This second damage mechanism during coalescence of the primary voids is partly neglected in the traditional GTN model.

5. Conclusions

Toughness tests on CT specimens have been carried out in two loading configurations: TL and LS on a forged AA6061 aluminum alloy in T6 condition. The material is highly anisotropic in terms of toughness values with LS configuration larger than TL configuration. Nevertheless, tensile tests performed on smooth specimens reveal an isotropic plastic behavior, which is used to fit an isotropic Voce hardening law. A previous study by in situ synchrotron laminography [9] and fractography of CT specimens have concluded fracture mechanisms linked to coarse precipitates and shear bands. The anisotropy of toughness is due to the anisotropic shape and distribution of coarse precipitates induced by the morphological anisotropy of grains. Anisotropic initial shape and distribution of coarse precipitates have been identified through SRCT studies of the as-received material. SRCT analysis of arrested cracks has revealed different fracture modes for the configurations TL and LS. Voids nucleate first on coarse Mg_2Si precipitates, followed by growth and coalescence. The iron-rich intermetallics enter in this procedure much later. Void coalescence is easier if the coarse precipitates are close to each other. For TL configuration, the coarse precipitates are intergranular and aligned in the plane of crack propagation. Void coalescence takes place by internal necking and the crack propagates mainly by the intergranular failure mode where the fracture energy is lower. For LS configuration, the coarse precipitates are aligned perpendicular to the plane of the crack propagation. Intergranular and transgranular fracture modes coexist in this case and the fracture energy is therefore higher [3, 41]. The void coalescence takes place by internal necking and a second mechanism on a second population of precipitates, i.e., dispersoids of chromium and manganese [35, 38].

The quantitative void volume fraction analysis has been conducted on SRCT data of the arrested cracks for the two

configurations. The void volume fraction at crack tip was evaluated at 1.5% for TL configuration and 5.6% for LS configuration for a region of interests of $140 \times 140 \times 70\,\mu m^3$.

The anisotropic damage mechanism has been validated by the finite element simulations based on the GTN micromechanical damage model with different sets of damage parameters for different configurations. The void volume fraction at crack tip was used as the critical void volume fraction for failure of elements with the same element size. Other parameters were identified by in situ SEM experiments and SRCT microstructural analysis. In this identification procedure, all parameters were identified experimentally by a local approach and in situ observation without parameter adjustment. Finite element simulation conducted on the CT specimens shows the good agreement for TL configuration, whereas the simulation on LS configuration overestimates the maximum load owing to the neglecting of the role of second population of precipitates.

Appendix

A. Material Models

A1. Voce Hardening Law. The plastic hardening is represented by a Voce-type stress saturation equation and is expressed as follows [39]:

$$\sigma = \sigma_s + (\sigma_0 - \sigma_s)\exp(-b\varepsilon), \tag{A.1}$$

where σ_0 is the yielding stress and σ_s and b are the undamaged material constants.

A2. GTN Damage Model. The Gurson–Tvergaard–Needleman (GTN) micromechanical model [5, 6] is used to introduce damage which is represented by a single scalar variable, the void volume fraction f. The plastic flow potential Φ is written as follows:

$$\varnothing = \frac{\sigma_{eq}^2}{\sigma_y^2} + 2q_1 f^* \cosh\left(\frac{3q_2}{2}\frac{\sigma_m}{\sigma_y}\right) - 1 - (q_1 f^*)^2 = 0. \tag{A.2}$$

The function f^* is the effective porosity and is justified to describe the onset of the void coalescence beyond a critical porosity f_c. The void coalescence is represented by an acceleration of damage rate [6] with

$$f^* = \begin{cases} f, & f < f_c, \\ f_c + \delta(f - f_c), & f \geq f_c, \end{cases} \tag{A.3}$$

where q_1 and q_2 are the void growth parameters involving the yield surface, σ_y is the yielding stress of nondamaged material, and $\delta = ((f_u - f_c)/(f_f - f_c))$ is the void coalescence acceleration factor. The material loses its stress carrying capacity at $f = f_f$. In this case, the crack is assumed to propagate.

The evolution of void volume fraction includes two parts: the growth of the existing voids and the nucleation of new voids. The increase of void volume fraction in the model is written as

Figure 12: 3D mesh of the quarter of the CT specimen with loading roller.

$$\dot{f} = \dot{f}_g + \dot{f}_n. \tag{A.4}$$

Because the matrix material is incompressible, the growth part $\dot{f}_g$ is related to the equivalent plastic strain $\dot{\varepsilon}_{eq}^p$ and is described as

$$\dot{f}_g = (1 - f)\dot{\varepsilon}_{eq}^p. \tag{A.5}$$

Chu and Needleman [42] expressed the void nucleation rate by a distribution function considering the heterogeneous nucleation process. This void nucleation law is dependent exclusively on the equivalent plastic strain:

$$\dot{f}_n = A\dot{\varepsilon}_{eq}^p. \tag{A.6}$$

The coefficient A is selected so that the void nucleation follows a normal distribution function [42] described as

$$A = \frac{f_{n0}}{s_{n0}\sqrt{2\pi}}\exp\left(\frac{1}{2} - \left(\frac{\varepsilon_{eq}^p - \varepsilon_{n0}}{s_{n0}}\right)^2\right), \tag{A.7}$$

where f_{n0} is the maximum value of nucleated void fraction, ε_{n0} is the mean strain for which the nucleation appears, and s_{n0} represents the deviation of the mean strain for which the nucleation appears.

B. Simulation Technique

The GTN model was implemented in the finite element software Cast3m (http://www-cast3m.cea.fr/), developed by the CEA in France. Due to symmetry, a quarter of the three-dimensional SRCL specimen is meshed by using quadratic elements with reduced integration (Figure 12).

When modeling crack propagation using continuum damage mechanics, the crack is a thin volume which height is half the element height in the case of quadratic elements with reduced integration (type CU20 with 20 nodes and 8 Gauss points) [43]. The mesh is refined in the crack region with element size of about $140 \times 140 \times 140\,\mu m^3$, which means $70 \times 70 \times 70\,\mu m^3$ for inter Gauss point size equivalent to the size of ROI in the quantitative experimental analysis in Section 3.5 with respect to the symmetric condition. The loading roller of experimental setup is modeled as a rigid body. The friction is neglected between roller and specimen. Loading is applied via imposed displacement through the Y-axis at line l_char. The same as in the experimental analysis,

$\delta 5$ is calculated at every loading step as twice the displacement of the marker P_$\delta 5$ positioned on the surface of the specimen at 2.5 mm through the Y-axis.

Data Availability

All data used in this article are restricted by the French Atomic Energy and Alternative Energy Commission in order to protect privacy. Requests for data may be considered by the corresponding author for researchers who meet the criteria for access to confidential data under conditions.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Study on Electric-Magnetic-Acoustic Signal Regularity and Its Correlation during Rock Shear Failure

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During mining activities, the deformation and damage of coal rock materials might result in coal rock dynamic disasters, such as rock burst. It leads to serious casualties and property losses. Generally, the occurrence of dynamic failure of coal and rock are caused by shear failure of coal seam. Geophysics signals are generated and related to damage evolution in this loading process. In this paper, sandstone samples were subjected to shear failure laboratory experiments, and the electric-magnetic-acoustic signal regularity was measured and analyzed comparatively. The results indicated magnetic signals were more correlated with stress and acoustic emission (AE) signals, while the amplitude of electric signal fluctuation was larger when main failure occurred. With the increase of sample size and shear strength, the strength of electric-magnetic-acoustic signals increased. The correlation coefficients between the magnetic signal and stress as well as AE energy were superior to those of electric signals. The coupling model between AE and electric signals was established, which shows good statistical correlation. This study lays the foundations for further interpreting the generation mechanism of the electric signal. It provides a new method to indicate the damage evolution of coal rock materials.

1. Introduction

Generally, coal resources play a significant role during the energy consumption, especially for industrial raw materials and electric power production [1–3]. Nevertheless, coal rock dynamic failure and even disasters occur during mining activities, which results in serious casualties and property losses [4–7]. Coal or rock fracture, roof collapse, and fault activation in mines are all accompanied by shear failure, which will further induce mine earthquake, rock burst, coal and gas outburst, and other dynamic disasters [2, 8]. It is of great significance to determine damage sources and their characteristics for predicting dynamic disasters in coal mines [9–11].

During the damage and failure process of the rock, acoustic emission (AE) signals are generated [12]. Therefore, AE technology is used to study the expansion and evolution mechanism of cracks [13, 14]. Khazaei et al. analyzed the variations in b values, energy of acoustic emission for rock specimens during 73 uniaxial compression tests [15]. Liang et al. investigated the mechanical and acoustic emission characteristics of the rock during loading and unloading confining pressures at the postpeak stage [16]. Wang et al. investigated the features of similar material mechanics and concluded that the law of AE behavior and the characteristics of similar material fracture can obtained the regularity of stress and deformation during material loading processes [17]. Kong et al. concentrated on the thermal mechanical properties and AE characteristics during the deformation and fracture of the rock under the action of continuous heating and after high-temperature treatment [18].

In addition to AE signals, electromagnetic radiation (EMR) has been recognized and analyzed in rock failure and earthquake precursor observation in early days [19–21].

Freund and Sornette revealed the mechanism for the low-frequency electromagnetic emissions and electric phenomena and conjectured the intermittent and erratic occurrences of EMR signals are a consequence of the progressive and of their release, which provided a conductive pathway [22]. Gade et al. argued that the electromagnetic signals are generated as a result of the collaborative consequence of charge separation, relaxation, and vibration of charged crack surface during the growth of cracks [23]. Interestingly, Carpinteri et al. reported that growth of both microcracks and macrocracks would lead to EMR signals [24]. Wang et al. reported that coal rock materials could produce ultralow frequency (ULF) EMR signals during the damage and failure processes, and the change trend of the signals was well similar to stress and AE signals [25]. To predict the collapses more accurately using EMR, a loaded coal rock EME coupling model based on statistical damage mechanics was established [26]. Kong et al. studied the time-varying characteristics of EMR during three metamorphic grades of the coal-heating process [27].

Owing to the complexity of rock outburst, the precursor signals of AE and EMR might be not very obvious, but with improvements in monitoring methods, the electric signals are beginning to be analyzed gradually [11, 28–30]. Freund has tried to establish the physically coherent model for resistivity changes, ground potentials, and electromagnetic behavior and explained the mechanism of earthquake-related electrical signals and emission [31]. Li et al. studied the surface potential of coal during different fracture modes, including uniaxial compression, tensile fracture, and three-point bending and concluded that surface potential was produced during the fracture of coal and that it had a good correspondence to the load [32]. To explain the mechanism of electric potential (EP), the change law and distribution characteristics of electric potential were tested with the precracked rock samples, and free charges were found to be produced at the tip of the cracks and the newly formed crack surface [33].

Several studies on AE, EMR, and electric signals have been published. However, studies evaluating these signals in shear failure, especially rock shear failure, are scarce. Besides, the correlation characteristics of electric-magnetic-acoustic signal regularity were not reported. In this paper, under shear failure, the electric-magnetic-acoustic signal regularity of sandstone was studied, and the impact of sample size on experimental results was analyzed. Furthermore, the correlation characteristics among electric signals, EMR, AE, and stress were illustrated. Variation laws of the electric signal were interpreted by establishing an acoustic-electric coupling model. It is promising to further reveal the dynamic disaster process of coal rock materials and perfect the EMR theory as well as the corresponding test equipment.

2. Materials and Experimental Procedure

2.1. Material Preparation. The tested rock samples were collected from roof sandstones of Sanhejian Coal Mine in Xuzhou, China. According to the International Society for Rock Mechanics (ISRM) standard, samples were prepared and processed in Strata Control Experimental Center, State Key Laboratory of Coal Resources, and Safety Mining of China University of Mining and Technology. After sample coring, two types of specimen (50 mm × 50 mm × 100 mm and 50 mm × 50 mm × 50 mm) were prepared. The prepared standard specimen was labeled and kept in a glass container that was sealed up with Vaseline to preserve the original state of the specimen.

2.2. Test System. The experimental system mainly includes hydraulic loading devices, data-acquisition system, various sensors, etc. (Figure 1). Loading system: the control system is an electrohydraulic servo pressure-testing machine (YAW4306), and the maximum pressure is 3000 kN. The loading process can fully realize program control and can simultaneously collect parameters such as stress and displacement during the whole loading process at high speed. The LB-IV electric signal data-acquisition system was used to collect the low-frequency electric signal during deformation and failure process of the rock. Data sampling frequency could be set arbitrarily below 1000 Hz. The frequency of sensors used to collect the low-frequency and acoustic and magnetic signals is higher than that of the low-frequency electric signal, so the CTA-1 acoustic-electric dynamic high-speed data-acquisition system, manufactured by Physical Acoustics (American), was used.

Signal sensors: the copper electrodes (Figure 2(a)) were used to receive low-frequency (<1 kHz) electric signals. AE sensors (Figure 2(b)) with 42.3 kHz resonant frequency were used to monitor acoustic signals, and the ferrite rod antennas (Figure 2(c)) of low frequency (1 kHz and 5 kHz) and common frequency (300 kHz) were used to collect magnetic signals. In the experiments, the cooper electrodes and the acoustic emission sensors were fixed onto the rock sample surface using adhesive tapes. A Vaseline coupling agent was used at the contact surface to ensure the elastic wave is received by the sensors. The ferrite rod antennas were arranged around the specimen, less than 10 cm away from the specimen.

2.3. Test Scheme. Shear strength of the rock is the tangential stress on the shear surface when rocks suffer shear failure and is the ultimate strength of the rock at shear failure. The oblique compression molding method is widely used to test the shear strength of the rock, with which the normal stress (σ) and shear stress (τ) on the shear failure surface can be calculated.

The laboratory rock shear experiment and the layout of sensors are shown in Figure 3. Two electrodes were pasted on the specimen surface, about 10 mm to two sides of the shear surface AB, respectively, and the third electrode was pasted on the upper left corner of the rear surface of the specimen, 10 mm away from the specimen edge. The AE sensor was attached at the bottom left corner of the rear surface of the specimen, 5 mm away from the specimen edge. The prepared rock specimen was placed in between two angle-variable shear clamps, and shear force was applied onto the specimen by controlling the loading rate. During the loading process, the

Figure 1: Experimental system.

electric, acoustic, and magnetic signals were monitored simultaneously. Finally, the sample would experience failure along the preset shear surface of AB.

3. Electric-Acoustic-Magnetic Signals upon Rock Shear Failure

3.1. Sample 1. The stress law and electric-magnetic-acoustic signal regularity of sample 1 are presented in Figures 4(a)–4(h). The size of shear failure surface was about 50 mm × 100 mm. From Figure 4(a), it can be concluded that the maximum shear strength of sample 1 was 46.8 MPa. Clearly, stress, AE, and EMR changed almost similarly throughout the whole loading process. In the initial 30 s, a self-balancing process of the shear mold occurred. Subsequently, shear force increased gradually. At 95 s, the magnetic signals of channel 5 (ch5: Figure 4(f)) and those of channel 8 (ch8: Figure 4(h)) showed an anomalous response. The signals increased rapidly and decreased immediately. Subsequently, the AE signals (Figure 4(e)) presented obvious changes when stress fluctuates slightly after 107 s. Compared with AE signals, low-frequency magnetic signals occurred earlier. After 120 s, stress increased quickly and the magnetic signals increased accordingly. At 160 s, stress reached 43.4 MPa, which is close to the peak intensity. At this moment, the main macrocrack is formed. The stress fluctuates violently, whereas AE and magnetic signals reach the peak values. At 176 s, the shear surface was cut through and stress suddenly dropped to 0. The corresponding channel signals declined quickly.

The electric signals of different channels differed significantly during the whole loading process. The electric signals are slightly weak in Channel 2 (ch2: Figure 4(b)) and Channel 9 (ch9: Figure 4(d)), because the sensors of two channels did not contact with the rock surface directly. Before the main failure, the overall variation of electric signals was only about 2 mV in these channels, whereas a huge amplitude of variations was detected after the development of main fracture. Considering the direct contact between electrode and the sample surface, the electric signals from channel 8 (ch8: Figure 4(c)) were more sensitive and produced violent increase and reduction when stress fluctuated at 107 s, which was consistent with the variation trends of AE and EMR. After 120 s, the EMR signal had the increase trend, which is correlated with the continuous increase of stress. Later, pulse fluctuation became increasingly frequent in the early stage of shear failure, until the occurrence of the main failure and the signal variation amplitude of all channels reached the peak. In general, the electric signal was slightly correlated with stress but was closely related to fluctuation of the AE signal. However, the electric signal fluctuates upon stress fluctuation or generation of the AE signal.

3.2. Sample 2. Sample 2 was also collected from roof sandstone at the same position as Sample 1, but the size of shear surface in the former was 50 mm × 100 mm. Accordingly, the failure load was decreased to 67 kN, but they had the same shear strength. It could be concluded that there is a good correlation between low-frequency magnetic signal of channel 5 (ch5: Figure 5(f)), channel 7 (ch7: Figure 5(g)), and the stress. However, the magnetic signal of channel 8 (ch8: Figure 5(h)) was relatively weaker, showing no obvious variation throughout the loading process. During the loading process, two internal damages were detected at 75 s and 123 s, and the main failure occurred at 140 s. The AE signal had strong responses at these three time points, and it was stronger at 75 s than that at 123 s. Interestingly, as shown in Figure 4(f), the EMR signal met its valley values around 75 s, 123 s, and 140 s. It is most striking around 123 s. Meanwhile, in Figure 4(h), the EMR signal met its significant mutation response around 123 s, while the response is less remarkable around 75 s and 140 s. Compared with Sample 1, the overall electric-magnetic signal strength was significantly weaker. In particular, the high-frequency magnetic signal was not only weak but also poorly corresponding to the stress and AE signal.

The variation law of low-frequency electric signal was similar to that of Sample 1. The signals of channel 2 (ch2: Figure 5(b)) and channel 9 (ch9: Figure 5(d)) changed only slightly, about 1 mV before the occurrence of main fracture. By contrast, they changed significantly after the main failure developed. Because of the higher sensitivity, the electric signal of channel 8 (ch8: Figure 5(c)) fluctuated greatly when stress began to change at 75 s and it presented the second change at 123 s when the overall basic value had reduced. This implied that the internal failure at 75 s brings about some negative charges surrounding the 8# test point. Because of the weak electroconductivity of the rock, this potential remained constant until the abundant new charges formed at the second fracture site (123 s). When the main failure occurred at 140 s, the amplitude of signal fluctuation of channel 8 reached the maximum, but this was still smaller than that of Sample 1. This indicated that sample size could affect intensity of the electric signal, but they had consistent variation tendency.

4. Discussion

4.1. Electric-Magnetic Signal Generation Mechanism. The variable motion of charges is an important way to generate the electric-magnetic wave and includes two mechanisms, such as vibration or transition of electric dipole and relaxation of separated charges. Most mechanisms of EMR

(a) (b) (c)

Figure 2: Sensors. (a) Electrode. (b) AE sensor. (c) Ferrite rod antenna.

1–3: Electrodes
4: AE sensor
5: Ferrite rod antenna

Figure 3: Shearing equipment: 1–3: electrodes, 4: AE sensor; 5: ferrite rod antenna.

generation induced by deformation and failure of coal or rock are related to development of cracks. One is asymmetric charges along the axial direction of cracks, for example, electric dipole model. The other is positive and negative charges generated on cracks wall, such as variable motion of charges and relaxation of separated charges caused by friction electrification or piezoelectric effect. Therefore, it is necessary to study how charges are generated and moved when exploring the mechanism of EMR generation upon stress failure of coal or rock.

In these experiments, the rock suffered shear failure, which is different from tensional failure. Therefore, there is residual stress on crack surfaces even after sufficient sliding of these surfaces. Such residual frictional resistance could be regarded as the main source of stress. Meanwhile, the energy release rate of shear failure was two magnitudes higher than that of tensional failure. In other words, shear failure released an increased concentration of energy at the failure, which was related to different development processes of shear cracks and tensional cracks at the end point.

In the shearing process, charges of different polarities would be generated on the upper and lower shear surfaces, which were attributed to piezoelectric effect, frictional effect, and asymmetric breakages of crystal bonds (Figure 6).

Charge generation induced by the piezoelectric effect is mainly manifested during growth of cracks. With the extension of cracks, the stress of microunit on the upper surface of cracks changes significantly, thus attracting charges to some microunits containing piezoelectric materials. After further extension of the cracks, these microunits are distributed on the upper and lower surfaces of the cracks, making both upper and lower surfaces charged. When the main crack was cut through (i.e., the failure of materials), charges generated by the piezoelectric effect are reduced, which would contribute less to EMR signals.

EMR signal, induced by triboelectric charge, exists throughout the whole shear failure. However, its contributions are different in different stages. During the early loading period, due to the small pressure stress, the relative displacement between upper and lower surfaces of the cracks was very small and the electrification by friction was relatively weak. Nevertheless, the relative displacement between upper and lower surfaces increased significantly after the principal crack was cut through, accompanied by strong friction effect and increased generation of charges, both of which contributed to increased generation of EMR signals. Meanwhile, the contributions of friction effect to the EMR signal under shear failure changed with the shear angle. Obviously, friction played a greater role under a small shear angle and high-pressure stress.

These effects influence the whole shear failure process together, resulting in the continuous accumulation of charges on the upper and lower crack surfaces. Moreover, rock materials are of poor conductive features, which have low rate of charge release. Charges in rock materials will not disappear in a short period but instead move with the surface displacement. Because the total charge on crack surface increases continuously, the electric signal strengthens gradually in the whole failure process when viewed from the macroscopic perspective. Besides, the distance between the testing point and charges changed when upper and lower surfaces had relative displacement, thus causing the pulse fluctuation of the electric signal. This conforms to the variation law of the electric signal

Figure 4: Continued.

Figure 4: Electric-magnetic-acoustic signals of 1# rock sample (1 mm/min). (a) Stress-time. Electric signals of (b) ch2, (c) ch8, and (d) ch9. (e) Acoustic emission signals. Magnetic signals of (f) ch5, (g) ch7, and (h) ch8.

observed in the experiments. Furthermore, the motion of charges generates the EMR signal.

4.2. Correlation of Various Signals. At present, most scholars accept that sample damages and failures could be analyzed by variation of stress and AE signal. If the high correlation between electric-magnetic signals and stress as well as AE could be proved, it would lay foundations for using electric-magnetic signals to analyze sample damage and dynamic disasters of coal or rock. Based on the aforementioned analysis, we found that there is certain correlation between electric-magnetic signals and stress as well as AE. Meanwhile, the correlation between EMR signal of conventional frequency and AE as well as stress has been recognized by most scholars. However, no quantitative evaluation index was proposed for such correlation. In this paper, the correlation coefficient ($|r|$) between electric-magnetic signals and stress as well as AE was calculated using the Origin mathematical software, which was used to evaluate their consistency.

Take the correlation between stress and electric signal (Figure 7(a)) for example. It demonstrated low linear correlation between two signals when the calculated maximum correlation coefficient $|r| < 0.4$, significant correlation when $0.4 \leq |r| < 0.7$, and high linear correlation when $0.7 \leq |r| < 1$. When the delay (τ) corresponding to the maximum correlation coefficient is negative, stress changes earlier than the electric signal; otherwise, the electric signal changes earlier.

The calculated correlation coefficient between electric-magnetic signals of Sample 1 and stress as well as AE is shown in Figure 7. In most cases, τ is close to 0, but in some cases, τ is not 0. Corresponding to the maximum correlation coefficient between stress and electric signal of channel 8 (Figure 7(a)), τ was 10 s, indicating that electric signal changed earlier than stress. However, the electric signal data of channel 9 lagged for 2 s behind the stress (Figure 7(a)). In Figure 7(d), the delay between the magnetic signal of channel 5 and channel 8 and AE signal of channel 2 was 2 s, indicating that the EMR signal changed earlier than the AE signal. Besides, correlation coefficients of electric signal had positive and negative values, which were related to accumulation of positive and negative charges in the electrode region. A similar phenomenon was observed in correlation coefficients of Sample 2 (Table 1). Meanwhile, we calculated another experimental data other than those used in this paper. According to various estimates, the correlation coefficient between electric signal and stress might be either positive or negative at equivalent probability. This was also caused by different charge polarities of different regions upon deformation or failure of the rock. When more positive charges accumulate in one region, the local electric signal increased generally, forming a positive correlation with stress variation. Otherwise, more negative charges accumulate, resulting in the negative correlation between the electric signal and stress.

Table 1 presents peak correlation coefficients of signals from Samples 1 and 2. Based on a comparison of 12 groups, correlation coefficients between electric-magnetic signals and stress were higher than those between electric-magnetic signals and AE in eight groups; correlation coefficients were close to those between electric-magnetic signals and AE in two groups; and correlation coefficients were smaller in two groups. With respect to correlation coefficients between stress and electric-magnetic signals, there were high linear correlation (in 6 groups), significant correlation (in 3 groups), and low linear correlation (in 3 groups). However, concerning the correlation coefficients between AE and electric-magnetic signals, there were no high linear correlation, significant correlation (in 9 groups), and low linear correlation (in 3 groups). Therefore, the correlation between magnetic signal and stress as well as AE was superior to that of the electric signal.

Figure 5: Continued.

Figure 5: Electric-magnetic-acoustic signals of 2# rock sample (1 mm/min). (a) Stress-time. Electric signals of (b) ch2, (c) ch8, and (d) ch9. (e) Acoustic emission signals. Magnetic signals of (f) ch5, (g) ch7, and (h) ch8.

Figure 6: Direction of the electromagnetic field in shear failure.

It has to be pointed out that because correlation coefficient is calculated based on linear correlation calculation, the correlation coefficients between the data that seems correlated are low in reality. Figure 8 presents a comparison between AE energy and the electric signals of ch8. The electric signal enhanced when the AE signal strengthened, indicating the strong correlation between them. We believed that such correlation might be nonlinear, which explains the smaller calculated correlation coefficient.

4.3. Acoustic-Electric Coupling Model and Verification. From Figure 8, it could be seen that AE and electric signals were correlated with each other. As shown in Figure 8, at most moments, the acoustic emissions energy and electric values are relatively stable with some fluctuation. At some moments, when there is a relatively significant damage occurring inside the coal body (it is recorded that there is the pulsed increase of acoustic emission energy, which indicates a relatively significant microrupture inside the sample), the electric signal increases suddenly, which is cooperative with the acoustic emission energy. The response is correlated, especially when the sample ruptures, and both the electric signal and the acoustic emission signal exhibit a rapid increase to the maximum value. The variation mechanism of the electric signal can be viewed as follows: the increase of electric signals results from charge accumulation during the loaded failure, whereas their decrease results from deprivation of charges lost gradually. The speed of increase and decrease is also related to the coupling capacitance between electrode and samples. Based on the aforementioned theory, a circuit model was established for electric signal variation during shear failure of the rock (Figure 9).

In this model, the condenser voltage signal (the observed signal) is correlated with signal source (S), resistance to

Figure 7: Correlation coefficient of signals (sample 1). (a) Stress vs. electric signals. (b) AE energy vs. electric signals. (c) Stress vs. EMR energy. (d) AE energy vs. EMR energy.

Table 1: Statistics on correlation coefficient values between two indexes of electric-magnetic signals as well as stress and AE energy.

Sample number	Channel number	Correlation coefficients between two indexes			
		Electric-stress	Electric-AE	EMR-stress	EMR-AE
1	Ch2	0.345	0.357	—	—
	Ch8	−0.785	−0.535	—	—
	Ch9	0.550	0.570	—	—
	Ch5	—	—	0.711	0.506
	Ch7	—	—	0.301	0.540
	Ch8	—	—	0.679	0.481
2	Ch2	−0.855	−0.406	—	—
	Ch8	−0.601	−0.193	—	—
	Ch9	0.137	0.386	—	—
	Ch5	—	—	0.925	0.452
	Ch7	—	—	0.867	0.421
	Ch8	—	—	0.772	0.411

Figure 8: Comparison between AE energy and the electric signals of ch8. (a) Sample 1. (b) Sample 2. Light blue dotted lines and orange arrows indicate abnormal response of signals.

Figure 9: Acoustic-electric coupling model.

ground (R), and capacitance (C). The following charge balance exists:

$$\int_{t_0}^{t_1}\left(a \times E - \left(\frac{U}{R}\right)\right)d_t = \left(U_{t_1} - U_{t_0}\right) \times C, \tag{1}$$

where E represents the accumulated AE energy, a represents the ratio between electric charge quantity and accumulated AE energy, and U denotes the voltage of the condenser.

Size of S is related to generation and development of cracks, which is proportional to AE energy. Accumulated AE energy in Δt is E. The quantity of electric charge generated by cracks is $q_l = a \times E$, and the quantity of electric charge released to ground is $q_s = ((U_{t_1} + U_{t_0})/2R)$. The overall charge balance can be simplified as

$$\begin{aligned} aE - \frac{U_{t_1} + U_{t_0}}{2R} &= \left(U_{t_1} - U_{t_0}\right) \times C, \\ aE &= U_{t_1}\left(C + \frac{1}{2R}\right) - U_{t_0}\left(C - \frac{1}{2R}\right), \\ \frac{2aR}{C + 2R}E &= U_{t_1} - U_{t_0} \times \frac{C - 2R}{C + 2R}. \end{aligned} \tag{2}$$

Let $b = (2aR/(C + 2R))$ and $c = ((C - 2R)/(C + 2R))$, equation (2) can be expressed as follows:

$$bE = U_{t_1} - cU_{t_0}. \tag{3}$$

Using equation (3), the collected electric signal was converted through adjusting c and its correlation with AE is displayed in Figure 10.

As shown in Figure 10, the evolution characteristics of AE energy and converted electric signal response are similar and cooperative. For example, in the majority of moments, the AE energy and converted electric values are low and stable in Figure 10(a). Around 17 s and 24 s, both of them show growth spurt. Around 100 s and 116 s, the cooperative anomalous characteristics are more significant. At 171 s, both values reach their maximum values cooperatively. Hence, there is a strong correlation exits between AE energy and electric signal with both showing a similar variation tendency, which will lay foundations for further interpreting the generation mechanism of the electric signal by improving its applications.

Figure 10: Correlation between AE energy and converted electric signals.

5. Conclusions

In this paper, the laboratory experiments of sandstone samples under shear loading were carried out; meanwhile, the electric-magnetic-acoustic signal regularity was measured and analyzed comparatively. The results were obtained as followed:

(1) The electric-magnetic-acoustic signals are generated during the shear failure of the rock. The magnetic signal is highly correlated with stress and AE signal, but the electric signal changes randomly. When the main failure occurs, the amplitude of the electric signal fluctuation is larger than that of magnetic and AE signals. The strength of electric-magnetic-acoustic signals is basically proportional to the rock strength, the energy released from failure, the number of cracks, and so on.

(2) The correlation coefficients between electric-magnetic signal and stress as well as AE are calculated and used for quantitative evaluation of the correlation degree between signals. The significant or highly linear proportion between the electric-magnetic signal and stress as well as AE reaches 75%. The correlation degree between the electric-magnetic signal and stress as well as AE is even higher.

(3) A coupling model between AE and electric signal is established, in which the magnitude of electric signal is related to AE energy (E), resistance to ground (R), and other parameters. There is a good correlation between the electric signal and AE energy, which confirms the variation law of the electric signal and verifies rationality of the model.

Data Availability

The data used to support the findings of this study are included within the article.

Conflicts of Interest

The authors declare that there are no conflicts of interest regarding the publication of this paper.

Authors' Contributions

Huilin Jia and Yue Niu contributed equally to this work.

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Research Article

Research on Characteristic Stress and Constitutive Equation of Confined Sandstone during Damage Evolution Based on Energy Evolution Analysis

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This study implements the cyclic loading-unloading triaxial compression tests with confining pressures of 5, 10, 20, and 30 MPa for determining the characteristic stress and constitutive equation of sandstone during damage evolution. The energy evolution characteristics and transformations are analyzed, the energy proportion evolution law is defined and analyzed, and its process is divided into five stages according to the whole stress-strain curve. Then, the dissipation energy proportion method for determining the characteristic stress is proposed and compared with the lateral strain method and the volumetric strain method, and the differences of characteristic stress determined by the three methods are within the acceptable error range, i.e., the proportion of crack initiation stress and crack damage stress to peak stress are all about 40% and 70%, respectively, and least affected by the confining pressures. The new method has strong maneuverability, clearer response to the damage evolution, better coordination with the damage evolution, and less subjective initiative. In addition, the dissipation energy proportion behavior is captured by fitting function and established a relationship with elastic modulus of sandstone, and then the damage evolution constitutive equation of sandstone is established based on energy proportion. Finally, the theoretical curve and experimental curve of damage evolution are compared according to the five stages of dissipation energy proportion evolution, and the result shows that the damage evolution constitutive equation is reasonable.

1. Introduction

The stability of geotechnical engineering depends on the physical and mechanical environment in which rock materials are located and their mechanical behavior. As a heterogeneous natural geological material with internal microstructure (joints, cracks, inclusions, etc.), the study of the damage evolution process of rock materials has important guiding significance for the design, construction, and operation stability evaluation of geotechnical engineering. At present, laboratory tests of rock materials generally show that the damage evolution process of loaded rock specimens has experienced crack closure, crack initiation, crack penetration, and crack coalescence [1–3]. According to the damage evolution state of rock samples, many scholars divide the whole stress-strain curve into five important stages, including the crack closure stage, linear-elastic deformation stage, stable crack growth stage, unstable crack growth stage, and postpeak damage stage. Corresponding to the five stages of damage evolution, the characteristic stress and crack evolution laws in each stage are given, namely, the crack closure stress (σ_{cc}), the crack initiation stress (σ_{ci}), the crack damage stress (σ_{cd}), the peak stress (σ_p), and the residual stress (σ_c) [4–8]; in particular, the σ_{ci} and σ_{cd} can guide geotechnical engineering design, construction, and operation stability evaluation. Hoek and Brown [9] found that when the far-field maximum stress magnitude exceeded 0.15 of the UCS, spalling was observed around the underground opening of square tunnels in South Africa. Martin and Christiansson [10] and Andersson et al. [11] voted that the σ_{ci} determined by uniaxial compressive tests could be used as an estimate for the in situ spalling

strength, while the ISRM makes no mention of the method for determining characteristic stress in its "Suggested methods for determining the uniaxial compressive strength and deformability of rock materials." Thus, finding a reliable method for determining characteristic stress may be beneficial to geotechnical engineering when estimating the damage evolution state.

To better estimate characteristic stress, various scholars have proposed different methods based on unconfined and confined laboratory tests. Brace et al. [4] utilized the volumetric strain-axial strain curve to determine the σ_{ci}; they noted that the volumetric strain presented linear characteristics before the σ_{ci}, while it presented nonlinear characteristics after the σ_{ci}, so the point deviated from the linear portion was the σ_{ci}. Considering initial cracks of rock materials, Martin and Chandler [5] proposed that using the crack volumetric strain-axial strain curve determined the σ_{ci}; they noted that the curve had a horizontal section with a value of zero, the onset of the horizontal section was the σ_{cc}, and the end was the σ_{ci}. It is well known that the lateral strain is more sensitive than the axial strain, for the lateral dilatancy of rock samples at the stable crack growth stage. Consequently, Lajtai [12] drew a tangent line on the curve of axial stress-lateral strain, and the point where the lateral strain deviated from linearity was the σ_{ci}. Nicksiar and Martin [13] proposed the lateral strain response method; firstly, using the volumetric strain method determined the σ_{cd} and connected the value of lateral strain corresponding the σ_{cd} to the original point as a reference line; then, the difference between lateral strain and reference line was captured and drawn vs axial stress, and the maximum value point on the curve was the σ_{ci}. Latterly, with the application of acoustic emission in monitoring the damage evolution of loaded rock specimens, it is feasible that acoustic emission reveals the phenomenon of crack at different stages of damage deformation. Eberhardt et al. [14], Zhao et al. [15], and Xue et al. [16] drew a tangent line on the AE count-axial stress curve, and the point where the AE count deviated from linearity was the σ_{ci}. In practical applications, some of these methods feebly respond to the crack evolution and more or less have subjectivity, such as determining the linear segment by using the lateral strain method and AE method. In addition, the change of index of these methods is not in coordination with the damage evolution, which cannot reflect all characteristic stress in the whole stress-strain curve, such as the volumetric strain method only determining the σ_{cd}, and the σ_{cd} should be determined firstly by the volumetric strain method when using the lateral strain response method [17–19]. To divide the process of damage evolution more intuitively and determine characteristic stress more accurately, it is urgent to find new parameters and propose new methods.

In nature, a series of physical changes during loading rock specimens, such as crack closure, initiation, growth, and coalescence, are energy transformation, i.e., the damage of loaded rock specimens going through energy absorption, accumulation, dissipation, and release. Xie et al. [20, 21] considered that the deformation and damage of rock were a process of exchanging material and energy with the outside and delineated inner relation between deformation damage and energy dissipation/release. Zhang and Gao [22] determined the elastic strain energy (ESE) and dissipation energy (DE) evolution and distribution through energy tests for red sandstone under uniaxial compression. Deng et al. [23] designed uniaxial cyclic loading tests and explored the parameter variation and interrelation, such as the total absorption energy (TAE), the ESE, and the DE. Meng et al. [24] explored the character of energy accumulation and dissipation during deformation and damage of rock specimens under various loading and unloading schemes and revealed the energy accumulation and dissipation evolution and distribution at the prepeak stage. Under the current scope of work, a unified understanding of energy transformation during loading has been obtained: the TAE transforms into the ESE and the DE, and more absorption energy transforms into the ESE which is stored in rock specimens than the DE at the prepeak stage, while the DE increases and the ESE dissipates rapidly at the postpeak stage. In contrast, few articles care about energy transformation corresponding to the physical change of crack closure, initiation, growth, and coalescence during loading rock specimens. Whether the key points of energy evolution associate with the characteristic stress in damage evolution or not, whether the energy evolution relates to the material parameter evolution of rock specimens or not, and whether the constitutive equation of rock is established from the energy evolution point of view or not have not been explored in depth.

Based on this fact, this paper designed the cyclic loading-unloading triaxial compression tests, introduced the energy proportion, proposed a new method to research crack closure, initiation, growth, and coalescence of loaded rock specimens, and established and verified the constitutive equation from the point of view of energy evolution.

2. Experimental

2.1. Rock Specimen Preparation. The experimental specimens were sandstone, all machined from the same sandstone block, exhibiting fine consistency and integrity with no obvious joints and cracks, and were carefully produced standard specimens with 50 mm in diameter and 100 mm in height and were ground to produce flat parallel surfaces with ±0.02 mm parallelism (see Figure 1) according to the ISRM-suggested method.

2.2. Test Equipment and Procedure. We implemented conventional triaxial compression tests and cyclic loading-unloading triaxial compression tests with the RLJW-2000 triaxial shear-creep testing system (see Figure 2(a)), with 2000 kN maximum axial loading capacity and 50 MPa maximum confining pressure. Axial and lateral displacement data were collected by extensometers (see Figure 2(b)).

Firstly, the conventional triaxial compression tests were implemented with confining pressures of 5, 10, 20, and 30 MPa, and the triaxial compressive strength (TCS) was scored. Then, the cyclic loading-unloading triaxial compression tests were implemented with confining pressures of 5, 10, 20, and 30 MPa: (1) First, the specimens were wrapped

Figure 1: Sandstone specimens.

(a) (b)

Figure 2: Test equipment. (a) Rock triaxial testing system. (b) Extensometers.

Figure 3: Experimental loading-unloading path.

up with plastic film to prevent pressure fluid entry. (2) When testing, specimens were fixed by loading small axial stress and then confining pressure was loaded to a specified value; afterwards, axial stress was loaded to a specified value at a displacement rate of 0.005 mm/s; finally, axial stress was unloaded to zero at a displacement rate of 0.005 mm/s with constant confining pressure; the load and unload process was done repeatedly until the test ended. (3) At the prepeak stage, the unloading points were chosen which normalized to TCS (by the conventional triaxial compression tests) varying from 0.1, 0.2, 0.3, 0.4, 0.5, 0.6, 0.7, 0.8, to 0.9; at the strain-softening stage and residual strength stage, one or more loading-unloading tests were implemented, respectively (see Figure 3).

Figure 4: The axial stress-strain curves from (a) conventional triaxial compression tests and (b) cyclic loading-unloading triaxial compression tests.

Figure 5: Relationship among the TAE, the ESE, and the DE during the i-th loading-unloading.

3. Experimental Results and Analysis

The axial stress-strain curves during conventional triaxial compression tests are shown in Figure 4(a). Two experimental groups were performed at each confining pressure during the cyclic loading-unloading triaxial compression tests. As the limitation of this article, we only showed one experimental group at each confining pressure in Figure 4(b).

We analyzed energy evolution of rock specimens based on test data. Considering the heterogeneity of rock specimens and differences between rock specimens, some simplifications were made when calculating energy for single-rock specimens: (1) neglecting the strain energy stored in rock by hydrostatic pressure working; (2) calculating the elastic modulus of a single cycle by using secant modulus, i.e., connecting the maximum stress point with the minimum stress point on a single cyclic stress-strain curve [25]; and (3) calculating the TAE and the ESE of a single cycle by the integral method (see Figure 5).

3.1. Energy Density Evolution Analysis. Energy is the driving force for internal defect evolution and propagation in rock materials [20, 21]. At each stage of deformation of loaded rock specimens, the energy transformation is in progress [26, 27]. During the triaxial stress state, the loaded rock specimens were taken as a closed-loop system with the supposition that no thermal transmission occurred between the rock specimen and the external environment [28]. According to the first law of thermodynamics, the TAE per unit volume of the rock specimen could be described as [28]

$$W = \int \sigma_1 d\varepsilon_1 + 2\int \sigma_3 d\varepsilon_3, \tag{1}$$

where σ_1 and ε_1 are the axial stress and strain, respectively, and σ_3 and ε_3 are the lateral stress and strain, respectively.

Taking the i-th time loading-unloading stress-strain curve as an example (see Figure 5), the relationship among the TAE, the ESE, and the DE was explained. As shown in Figure 5, the loading curve AB was higher than the unloading curve BC; the total deformation (ε_i) of rock caused during loading contained recoverable deformation (ε_i^e) which was released at the unloading stage and irrecoverable deformation (ε_i^d) used for rock damage and plastic strain. From the energy perspective, the TAE was the area under the loading curve AB and the ESE was the area under the unloading curve BC, and the difference of the above two energies was the DE:

$$\begin{aligned} W_i &= \int_0^{\varepsilon_i} \sigma d\varepsilon, \\ W_i^e &= \int_0^{\varepsilon_i^e} \sigma d\varepsilon, \\ W_i^d &= W_i - W_i^e = \int_0^{\varepsilon_i} \sigma d\varepsilon - \int_0^{\varepsilon_i^e} \sigma d\varepsilon, \end{aligned} \tag{2}$$

where W_i, W_i^e, and W_i^d are the TAE, the ESE, and the DE during the i-th loading-unloading, respectively.

On the axial stress-strain curves during the cyclic loading-unloading triaxial compression tests, the TAE-

Figure 6: Energy evolution curves during the cyclic loading-unloading triaxial compression tests with confining pressures of 5 MPa (a), 10 MPa (b), 20 MPa (c), and 30 MPa (d).

strain curves, the ESE-strain curves, and the DE-strain curves were drawn, as shown in Figure 6:

(1) At the prepeak stage of the same rock specimens, the deformation changed from nonlinearity to linearity and then to nonlinearity with the increase of the number of cycles, corresponding to the crack closing first and then growing stably and finally growing unstably till macroscopic damage. At the postpeak damage stage, due to the confining pressure, the rock specimen still had carrying capacity. With the increase of confining pressure, the degree of nonlinear deformation at the stage of crack closure slowed down gradually, the TCS of rock specimens increased, and the residual strength also increased; that is, the carrying capacity increased when rock specimens were damaged.

(2) At the prepeak stage, the TAE, the ESE, and the DE increased with the axial strain increase; basically, the TAE and the ESE reached a maximum value. After the peak point, i.e., the initial postpeak damage stage, the ESE which was stored in the rock specimens at the prepeak stage dropped rapidly and converted into the DE for the macroscopic crack growth, coalescence, and caving damage of rock specimens, and the DE reached the maximum value. At the residual strength stage, the energy decreased and finally leveled off with no new damage. On the whole, with the confining pressure increase, the TAE, the ESE, and the DE increased during a single cyclic loading-unloading.

(3) Taken as a whole, the TAE, the ESE, and the DE displayed nonlinearity at the crack closure stage, while the degree of nonlinearity gradually slowed down with the confining pressure increase, corresponding to the change of stress; at the linear-elastic deformation stage and the stable crack growth stage, the three kinds of energies represented a continuous linear increase, while displayed nonlinearity at the unstable crack growth stage; the ESE and the DE changed suddenly at the initial postpeak damage stage.

3.2. Material Parameter Evolution Analysis. We calculated the elastic modulus of a single cycle at each confining pressure and drew elastic modulus-strain curves during cyclic loading-unloading, as shown in Figure 7.

Taken as a whole, the elastic modulus rose with the confining pressure increase. Under the same confining pressure, the initial elastic modulus rose, whereas the rate

Figure 7: Elastic modulus-strain curves during cyclic loading-unloading.

of rise gradually decreased and then underwent a period of approximately linear increase and reached the maximum value, corresponding to the crack closure stage, the linear-elastic deformation stage, and the stable crack growth stage, respectively. The stable crack closing and growth prompted the ESE to be stored in the rock specimens; that is, the elasticity of rock specimens was enhanced, and the maximum values of the elastic modulus with confining pressures of 5, 10, 20, and 30 MPa were 16, 22, 24, and 28 GPa, respectively. During the unstable crack growth stage, the elastic modulus gradually decreased with the ESE release. The elastic modulus decreased suddenly at the initial postpeak damage stage and then gradually decreased with no new damage.

4. Proposed Method for Determining Characteristic Stress

As previously mentioned, the loaded rock specimens went through energy absorption, accumulation, dissipation, and release. The ESE was accumulated firstly and then released when the rock specimens damaged, while the DE which existed in the whole loading process was used for original crack closure and new crack initiation, growth, and coalescence. So, the proportion of ESE or DE in TAE reflected the relation of energy transformation during loading. Especially under some stress or strain levels, the proportion of DE in TAE could explain the condition of damage development and plastic deformation of rock specimens which directly affected rock material parameters [29, 30]. Therefore, we defined the parameter of energy proportion, i.e., the ESE proportion (γ) and the DE proportion (η), and analyzed their evolution law; then, a new method for determining the characteristic stress in damage evolution was proposed based on DE proportion, and it was applied and verified finally.

4.1. Definition of Energy Proportion. We defined the parameter of energy proportion, which was taken as the proportion of ESE and DE in TAE. The ESE proportion of the i-th loading-unloading could be described as

$$\gamma_i = \frac{W_i^e}{W_i}, \tag{3}$$

and the DE proportion of the i-th loading-unloading could be described as

$$\eta_i = 1 - \gamma_i = 1 - \frac{W_i^e}{W_i}. \tag{4}$$

On the axial stress-strain curves during the cyclic loading-unloading triaxial compression tests, we drew energy proportion-strain curves (see Figure 8).

4.2. Energy Proportion Evolution Analysis. As shown in Figure 8, under the confining pressures of 5, 10, 20, and 30 MPa, the ESE proportion increased, while the DE proportion deceased at the initial stage, but most of the TAE transformed into the DE for original crack closing, corresponding to the crack closure stage on the axial stress-strain curves; at the later crack closure stage, two kinds of energy proportion were equal. When the loaded rock specimens came into the stage of linear-elastic deformation, the ESE played an important role; contrary to the DE proportion evolution, the ESE proportion increased, while the evolution rate of two kinds of energy proportion gradually decreased. With loading, the evolution rate was gradually steady, and the ESE proportion and the DE proportion approximately linearly developed; that is, the energy transformation developed steadily, and the loaded rock specimens stayed at the stable crack growth stage. When the evolution rate was zero, the ESE proportion reached the maximum value and the DE proportion reached the minimum value in the meanwhile. When the loading stress exceeded the crack damage stress of rock specimens, the rock specimens came into the unstable crack growth stage, and the crack growth consumed energy and caused the DE proportion increase and the ESE proportion decrease, while the evolution rate gradually increased. When the loading stress exceeded the peak stress, the evolution rate suddenly increased, the DE proportion reached the peak value and finally leveled off, and the ESE proportion evolution was opposite.

Taken as a whole, under the confining pressures of 5, 10, 20, and 30 MPa, the energy proportion (for example, the DE proportion) evolution was similar to the whole axial stress-strain curves, which could be divided into five stages (see Figure 8):

(i) Decreased suddenly (stage I)

(ii) Decreased nonlinearly (stage II)

(iii) Decreased approximately linearly (stage III)

(iv) Reached the minimum value and increased gradually (stage IV)

(v) Increased suddenly and leveled off (stage V)

In contrast, the evolution of the ESE proportion was in the opposite direction (see Figure 8).

Figure 8: Energy proportion-strain curves with confining pressures of 5 MPa (a), 10 MPa (b), 20 MPa (c), and 30 MPa (d).

4.3. Proposed New Method. As previously mentioned, corresponding to the stages of the whole axial stress-strain curves during the cyclic loading-unloading triaxial compression tests, the energy proportion evolved obviously at each stage. Based on this, considering geotechnical engineering focused on the rock damage, we proposed a new method, energy proportion method (EPM), for determining characteristic stress during rock damage evolution using the DE proportion curves:

(i) Specimen preparation: rock samples are selected on-site, and standard specimens are produced by the ISRM-suggested method.

(ii) Test procedure: conventional triaxial compression tests are performed to determine the TCS; cyclic loading-unloading triaxial compression tests are performed with the unloading points which normalized to TCS varying from 0.1, 0.2, 0.3, 0.4, 0.5, 0.6, 0.7, 0.8, to 0.9 (or the hierarchical interval is reduced moderately to improve the accuracy).

(iii) Calculation: energy density is calculated (the TAE and the ESE) during a single cyclic loading using the area integral method by MATLAB, and then the DE proportion is calculated.

(iv) Plot: the DE proportion-strain curves are drawn on the axial stress-strain curves during the cyclic loading-unloading triaxial compression tests.

(v) Determination (taking the confining pressures of 5 and 20 MPa as an example (Figure 9)): firstly, the minimum value of the DE proportion is found (the green point), the vertical line is drawn (the purple line) starting with the green point, and the σ_{cd} contact with the axial stress-strain curve is determined; then, the tangent line is drawn starting with the green point (the blue line) and the σ_{ci} contact with the axial stress-strain curve is determined where the DE proportion deviated from linearity (the yellow point); and finally, the tangent is drawn starting with the yellow point (the black line) and the σ_{cc} where the axial stress deviated from linearity is determined. This could be concluded as "two points and three lines."

4.4. EPM: Application and Verification. We determined characteristic stress of sandstone specimens during damage evolution under the confining pressures of 5, 10, 20, and 30 MPa. The determination process is shown in Figure 9

Figure 9: Determination process of characteristic stress by EPM with the confining pressures of 5 MPa (a) and 20 MPa (b).

Table 1: Comparison of characteristic stress determined by the three methods.

Confining stress (MPa)	Peak stress σ_p (MPa)	Characteristic stress (MPa)						Difference (MPa)			Ratio of characteristic stress to σ_{ci} (%)					
		EPM			LSM		VSM				EPM			LSM		VSM
		σ_{cc}	σ_{ci}	σ_{cd}	σ_{cc}	σ_{ci}	σ_{cd}	σ_{cc}	σ_{ci}	σ_{cd}	σ_{cc}/σ_p	σ_{ci}/σ_p	σ_{cd}/σ_p	σ_{cc}/σ_p	σ_{ci}/σ_p	σ_{cd}/σ_p
5	126.00	26.80	52.15	87.80	28.40	52.25	85.80	1.60	0.10	−2.0	21.27	41.39	69.68	22.54	41.47	68.10
10	160.00	20.35	62.70	100.00	22.40	65.30	105.00	2.05	2.60	5.00	12.72	39.19	62.50	14.00	40.81	65.63
20	194.00	18.70	77.00	135.50	21.82	77.50	137.00	3.12	0.50	1.50	9.64	39.69	69.85	11.25	39.95	70.62
30	234.00	12.15	90.00	165.50	15.80	91.00	170.00	3.65	1.00	4.50	5.19	38.46	70.73	6.75	38.89	72.65
Mean	—	—	—	—	—	—	—	2.61	1.05	2.25	—	39.68	68.19	—	40.28	69.25

(only the confining pressures of 5 and 20 MPa are shown), and the results are shown in Table 1.

In Table 1, the characteristic stress was determined by EPM: with the confining pressure increase, the σ_{cc} decreased and the σ_{ci} and the σ_{cd} increased, whereas the ratio of characteristic stress to peak stress was different; the confining pressure had a significant effect on σ_{cc}/σ_p which decreased with the confining pressure increase, while the σ_{ci}/σ_p and the σ_{cd}/σ_p barely changed. Under the confining pressures of 5, 10, 20, and 30 MPa, the σ_{ci}/σ_p was 41.39%, 39.19%, 39.69%, and 38.46%, with an average of 39.68%, and the σ_{cd}/σ_p was 69.68%, 62.50%, 69.85%, and 70.73%, with an average of 68.19%; that is, under different confining pressures, the σ_{cd}/σ_p was all about 40% and the σ_{cd}/σ_p was all about 70%. From an energetic standpoint, the confining pressure amplified the ability of rock specimens to store and dissipate energy, while the distribution proportion of energy almost unchanged.

For the memory function of the rock specimen, we drew the outer envelopes of the axial stress-lateral strain curves and the volumetric strain-axial strain curves which were obtained from the cyclic loading-unloading triaxial compression tests (see Figure 10). The lateral strain method (LSM) and the volumetric strain method (VSM) were chosen to verify the EPM. The determination process is shown in Figure 10 (taking the confining pressures of 5 and 20 MPa for example): the start and the end of the linear region on the outer envelope of the lateral strain corresponded to the σ_{cc} and the σ_{ci}, respectively, and the peak of volumetric strain corresponded to the σ_{cd} (see Table 1). For the changing trend of characteristic stress, the results determined by LSM and VSM were similar to those by EPM. For the difference of characteristic stress, the results determined by LSM and VSM were larger than those by EPM, with an average of 2.61 MPa (σ_{cc}), 1.05 MPa (σ_{ci}), and 2.25 MPa (σ_{cd}). For the ratio of characteristic stress to peak stress, the trend of σ_{cc}/σ_p determined by LSM was similar to that by EPM, i.e., decreasing trend with the confining pressure increase; the σ_{ci}/σ_p was still about 40% determined by LSM, and the σ_{cd}/σ_p was still about 70% determined by VSM.

Comparisons showed that the EPM was feasible. In practical applications, the DE proportion had clearer response to damage evolution, better coordination with damage evolution, and less subjective initiative, and the σ_{cd}, σ_{ci}, and σ_{cc} could be determined in turn by "two points and three lines" using the EPM.

5. Equation of Damage Evolution

The mechanical evolution of loaded rock specimens is a progressive damage. Research on damage evolution from an energetic standpoint contains mechanical processes, such as strength and deformation characteristics, so it can reveal the damage and failure mechanism of rock more essentially. Jin et al. [31] defined and calculated the damage variable from an energetic standpoint and researched the damage propagation law of rock materials. Zhao et al. [32] revised the damage variable, derived the relationship between dissipated

Figure 10: Determination process of characteristic stress by LSM and VSM with the confining pressures of 5 MPa (a) and 20 MPa (b).

energy and damage variable, and then established the damage equation. Liu et al. [33] introduced the damage variable and compaction factor based on energy and revised the equation of damage evolution and verified it using experimental data. Li et al. [34] defined the damage variable and calculated damage modulus from an energetic standpoint and then analyzed damage modulus evolution with confining pressures. Moreover, the above research studies also showed that the energy proportion evolution could explain the deformation of loaded rock specimens. Based on this, we established the relationship between elastic modulus and DE proportion, fitted the mathematical expression of the DE proportion, and then derived the equation of damage evolution of loaded rock specimens.

Theevolution laws of elastic modulus which were illustrated in Section 3.2 were similar to the evolution laws of energy proportion illustrated in Section 4.2, so we defined the relationship between elastic modulus and DE proportion:

$$E = a(1-\eta)E_0, \tag{5}$$

where a is a scaling factor, greater than 1, relating to the confining pressures, and E_0 is the maximum value of the elastic modulus at the linear-elastic deformation stage.

5.1. Equation of Energy Proportion. We fitted the DE proportion under confining pressures of 5, 10, 20, and 30 MPa (see Figure 11), and the fitting function and parameters are shown in Table 2. From the fitting curves, at the initial stage of DE proportion, i.e., the I, II, and III stages, the difference of DE proportion gradually decreased to be aligned at the same strain level, with the confining pressures of 5, 10, 20, and 30 MPa, while the difference of DE proportion at the same strain level gradually increased at the IV stage and the previous V stage and decreased to be stable at the later stage of V.

The meaning of fitting function and parameters is shown in Figure 12, and combined with the fitting curves in Figure 11, we concluded the following:

(i) η_0 is the maximum value of DE proportion at the V stage

(ii) A is the difference of DE proportion between the minimum value at the III stage and the maximum value at the V stage

(iii) ε_c is the axial strain corresponding to the minimum value of the DE proportion at the III stage

(iv) w is the total axial strain corresponding to the III stage and IV stage of the DE proportion

From Table 2, it is seen that the fitting parameters, like η_0 and A, almost did not change with the confining pressure change, as explained in Section 4.4 (the distribution proportion of energy in rock almost unchanged with the confining pressure); ε_c and w increased with the confining pressure increase, which also could be obtained in Figure 8.

Figure 11: Experimental and fitting curves of DE proportion.

Table 2: Fitting function and parameters of DE proportion.

Confining stress (MPa)	Fitting function and parameters $\eta = \eta_0 - Ae^{-((\varepsilon-\varepsilon_c)^2/2w^2)}$			
	η_0	A	ε_c	w
5	0.7	0.42	0.75	0.35
10	0.63	0.43	0.75	0.37
20	0.64	0.43	0.8	0.46
30	0.62	0.42	0.9	0.5

Figure 12: Meaning of the function image and the parameters of the DE proportion.

5.2. Constitutive Equation of Damage Evolution Based on Energy Proportion. In Figure 5, during the i-th cyclic loading, the total axial strain ε_i contained the recoverable strain ε_i^e and the residual strain ε_i^d:

$$\varepsilon_i = \varepsilon_i^e + \varepsilon_i^d. \tag{6}$$

Figure 13: Theoretical stress-strain curves based on energy proportion with the confining pressures of 5 MPa (a), 10 MPa (b), 20 MPa (c), and 30 MPa (d).

By using formulas (2) and (3), the following relations were obtained approximately:

$$\begin{aligned} \varepsilon_i^e &= (1-\eta_i)\varepsilon_i, \\ \varepsilon_i^d &= \eta_i \varepsilon_i. \end{aligned} \tag{7}$$

The stress of loaded rock specimens contains the elastic part and damage part, which can be described by using the elastic part [35]:

$$\sigma = E\varepsilon^e. \tag{8}$$

Taking E and ε^e which contain damage parameter (the DE proportion), the stress is

$$\sigma = a(1-\eta)^2 E_0 \varepsilon = a\left(1-\left(\eta_0 - Ae^{-(\varepsilon-\varepsilon_c)^2/2w^2}\right)\right)^2 E_0 \varepsilon. \tag{9}$$

This equation could describe the axial stress-strain curves, and the values of a were 1.65, 1.48, 1.38, and 1.28. Then, we obtained the theoretical axial stress-strain curves under the confining pressures of 5, 10, 20, and 30 MPa, respectively (see Figure 13).

As shown in Figure 13, fitting function for the theoretical curves described the experimental curves. But some of the shortcomings existed corresponded to the stages of the experimental curves which were divided by the DE proportion (see Figures 8 and 13): At the I stage, the theoretical curves coincided with the experimental curves under the confining pressures of 5 and 10 MPa, while they gradually deviated and depressed with the confining pressure increase. They had an upward departure with low confining pressures at the II stage, while downward deviation with the confining pressure increase, and the deviation degree was minimum with the confining pressure of 10 MPa. They also had an upward departure with low confining pressures at the III stage, but downward deviation with the confining pressure increase, and the deviation degree was minimum with the confining pressure of 20 MPa. They had an upward departure at the IV stage, while the deviation degree gradually decreased with the confining pressure increase. At the IV stage, the deviation

degree was minimum with the confining pressure of 20 MPa.

6. Conclusion

The objective of this paper is to propose a new method for determining characteristic stress and to establish a constitutive equation for damage evolution of sandstone from an energetic standpoint. The cyclic loading-unloading triaxial compression tests which were implemented with confining pressures of 5, 10, 20, and 30 MPa revealed that the energy evolution corresponded to the damage deformation of rock specimens, and the energy transformation could describe crack closure, initiation, growth, and coalescence under the loading process. Furthermore, the parameter of energy proportion was defined. Taking the DE proportion as an example, it was divided into five stages corresponding to the five stages of the whole stress-strain curve. The EPM for determining characteristic stress was proposed and was used to determine the characteristic stress of sandstone during the cyclic loading-unloading triaxial compression tests. The results showed that, with the confining pressure increase, the σ_{cc} decreased, while the σ_{ci} and the σ_{cd} increased; under different confining pressures, the σ_{cc}/σ_p decreased, the σ_{cd}/σ_p was about 40%, and the σ_{cd}/σ_p was about 70% because the distribution proportion of energy in rock specimens almost was unchanged with different confining pressures. Also, the results were verified by LSM and VSM: for the difference of characteristic stress, the results determined by LSM and VSM were larger than those by ERM, while they were in the acceptable error range; for the ratio of characteristic stress to peak stress, the trend of σ_{cc}/σ_p determined by LSM was similar to that by ERM, the σ_{ci}/σ_p was still about 40% determined by LSM, and the σ_{cd}/σ_p was still about 70% determined by VSM.

Based on the analysis of material parameter evolution, the relationship between elastic modulus and DE proportion was defined. The DE proportions fitted under confining pressures, and the fitting function and parameters had certain significance. Furthermore, the equation of damage evolution based on energy proportion was established. According to the five stages of DE proportion evolution, the theoretical curve and experimental curve of the damage evolution equation were compared, and then the rationality of the damage constitutive theory established by DE proportion evolution was verified.

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Study on Dynamic Erosion Behavior of 20# Steel of Natural Gas Gathering Pipeline

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In this paper, a finite element model for the particle-impact target is established for the abrasion erosion failure behavior of 20# steel used in natural gas gathering pipelines. The JC plastic model and failure criterion are used in the model to simulate the failure and damage of the material, and the reliability of the model was verified by comparison with experimental data. The dynamic behavior of particle erosion was studied by using this model. The changes of energy, stress, and morphology at the microscopic level were analyzed. The influence of particle shape, particle size, impact velocity, and impact angle on the erosion rate of 20# steel was obtained. The model is further used to study the influence mechanism of stress interference on erosion damage behavior under multiparticle random drop impact.

1. Introduction

The phenomenon of sand production in the gas field makes it inevitable to carry tiny solid particle impurities in the gas gathering stage, and these particles can cause continuous erosion and wear damage to natural gas gathering pipelines [1, 2]. Liu et al. and Pereira et al. carried out erosion research on a horizontal elbow [3, 4]. Fu and Wang carried out erosion research on a throttle valve [5]. At present, the classical fitting models of experimental parameters are generally used to predict the erosion failure law of pipelines, and most of these models do not consider the influence of material properties on model coefficients. Material properties are the most important factor affecting erosion wear. In addition, the impact parameters of particles, such as erosion velocity, particle shape, and impact angle, are also the primary factors on erosion wear.

With the development of computing technology, it is possible to study the microbehavior of erosion failure of materials with different properties [6–8]. Wang et al. proposed a method of a coupling finite element model and smooth particle dynamics model to study the erosion and failure behavior of particles on ductile materials, and the results were in good agreement with previous work [9]. Takaffoli and Papini used a 2D finite element model to simulate the impact behavior of rhomboid particles on a copper target. Both high distortion element removal and adaptive grid technology were tried to deal with large deformation and distortion elements, and the results showed the effectiveness of the model [10]. Takaffoli and Papini used coupled FEM and smooth particle hydrodynamics (SPH) to investigate the removal mechanism of deformation elements in single particle impacts on ductile materials [11]. Furthermore, three-dimensional models of coupled FEM and SPH were used to study the impact failure behavior of Al6061-T6 materials subjected to rhombohedral particles [12]. Liu et al. used the finite element technology based on Johnson–Cook and failure model to discuss the erosion damage of different shape particles on ductile materials, and the results revealed the rule of the impact of particle shape on the erosion damage [13]. Dong et al. simulated the surface erosion behavior of 6061-T6 aluminium alloy and oxygen-free copper with angular particles by SPH. The J-C model and J-C failure criterion were adopted. This method has better simulation effect than the traditional finite element method on large deformation and removal of material [14].

Kumar and Shukla established the finite element model of single particle impacting titanium alloy by using the abrasive water jet. The Hashish wear model was adopted, and the influence factors of particle shape were considered on the basis of the Finnie cutting model [15]. Wang and Yang divided 100 spherical particles into 10 groups to impact plastic materials and brittle materials, respectively, and obtained reasonable erosion rate results [9]. ElTobgy et al. found that the erosion rate of the same impacted position tended to be stable with the increase of impact times through the finite element simulation of multiple steel ball particles impacting Ti-6Al-4V, and at least three particles were required to reach this stable point value [16].

Based on the understanding of the microstructure and mechanical properties of 20# steel, which is commonly used in gas gathering pipelines in southwest China, the failure rules of 20# steel under particle impact were studied by using the finite element method, providing theoretical support for the prediction of the erosion rule of natural gas gathering pipeline components.

2. Numerical Model

2.1. Material Properties. 20# steel is a common type of pipeline steel, and its chemical composition (mass fraction/%) is 0.16 C, 0.19 Si, 0.36 Mn, 0.01 Ni, 0.01 Cr, and other Fe and belongs to the high-quality low carbon steel, in which plasticity and toughness are good. The microstructure is typical ferrite and pearlite, distributed in sheets, as shown in Figure 1. The mechanical properties of the target material, such as Poisson's ratio, yield strength, and Young's modulus at a temperature of 20°C, are given in Table 1.

2.2. Constitutive Models. In this paper, the Johnson–Cook plastic model is adopted, which can accurately predict the plastic deformation of the impacted target under high strain rate. The equivalent stress depends on the equivalent strain $\bar{\varepsilon}_p$, equivalent strain rate $\dot{\varepsilon}$, and temperature, and the expression is as follows:

$$\sigma = \left(A + B\left(\bar{\varepsilon}_p\right)^n\right)\left[1 + C \cdot \ln\left(\frac{\dot{\varepsilon}_p}{\dot{\varepsilon}_0}\right)\right]\left[1 - \left(\frac{T - T_r}{T_m - T_r}\right)^m\right], \tag{1}$$

where σ is the equivalent stress, A is the yield strength, B is the hardening modulus, $\bar{\varepsilon}_p$ is the equivalent plastic strain, n is the hardening exponent, C is the strain rate coefficient, $\dot{\varepsilon}_p$ is the strain rate, and $\dot{\varepsilon}_0$ is the reference strain rate, usually takes $1.0\,s^{-1}$. T is the target temperature, T_r is the room temperature, T_m is the melting point, and m is the strain rate sensitive index.

The Johnson–Cook failure criterion is derived from the accumulative damage method of equivalent plastic strain based on the element integral point, which takes into account stress triaxiality, strain rate, and temperature effect. The failure parameters are defined as follows:

$$D = \sum \frac{\Delta\bar{\varepsilon}_p}{\varepsilon_f}, \tag{2}$$

Figure 1: Microstructure of 20# steel.

Table 1: Mechanical material property

Density (kg/m^3)	Young's modulus (GPa)	Yield strength (MPa)	Tensile strength (MPa)	Poisson's ratio
7850	206	275	410	0.3

where D is the failure parameter, $\Delta\bar{\varepsilon}_p$ is the load increases caused by equivalent plastic strain increment, and ε_f is the failure strain, which is obtained by using the following equation:

$$\varepsilon_f = \left[d_1 + d_2\exp\left(d_3\frac{\sigma_n}{\sigma}\right)\right]\left[1 + d_4\ln\left(\frac{\dot{\varepsilon}_p}{\dot{\varepsilon}_0}\right)\right] \cdot \left[1 + d_5\left(\frac{T - T_r}{T_m - T_r}\right)\right], \tag{3}$$

where σ_n is the compressive stress. σ is the equivalent stress. $\dot{\varepsilon}_p$ is the equivalent strain rate; $\dot{\varepsilon}_0$ is the reference strain rate, often takes $1.0\,s^{-1}$; T is the target temperature; T_r is the room temperature; T_m is the melting point; and d_1~d_5 are the material failure constants.

The failure strain ε_f and the cumulate of damage are a function of mean temperature, strain rate, and temperature. When the failure parameter $D = 1$, it is the initial point of failure initiation.

For the explicit dynamic finite element method, the material is discretized into many finite elements, and all elements $(\sum \Delta\varepsilon)_i$ are calculated by using the dynamic algorithm. For any element i in any time step, as long as the failure parameter $D_i = 1$, element i will be labeled as failure and enter the damage evolution stage. In this paper, Ti-6Al-4V was used as the target material for simulation, and the results are compared and verified with the published literatures. Then, the target material will be replaced with 20# steel for research. The specific material parameters are shown in Table 2.

The failure model calculates the critical point of material strength limit, and before the element is deleted, the damage evolution of material needs to be defined. In this paper, the displacement change of linear development is adopted to control the damage:

Table 2: Material parameters.

	Ti-6Al-4V	20#
ρ (kg·m^{-3})	4428	7840
E (GPa)	113.8	206
ν	0.31	0.3
A (MPa)	1098	258
B (MPa)	1092	329
n	0.93	0.235
C	0.014	0.0323
m	1.1	1.0
d_1	−0.09	0.05
d_2	0.27	3.44
d_3	0.48	−2.12
d_4	0.014	0.002
d_5	3.87	0.61

$$\dot{d} = \frac{L\dot{\bar{\varepsilon}}^{\mathrm{pl}}}{\bar{u}_{\mathrm{f}}^{\mathrm{pl}}} = \frac{\dot{\bar{u}}^{\mathrm{pl}}}{\bar{u}_{\mathrm{f}}^{\mathrm{pl}}}, \tag{4}$$

where L is the feature length of the element, $\dot{\bar{\varepsilon}}^{\mathrm{pl}}$ is the equivalent plastic strain rate, $\bar{u}_{\mathrm{f}}^{\mathrm{pl}}$ is the failure displacement of element deletion, and $\dot{\bar{u}}^{\mathrm{pl}}$ is the plastic displacement rate. When the value of $\dot{d}$ reaches 1 the element stiffness degradation will automatically be deleted completely.

During the damage evolution process, the element stress is calculated by using the following formula:

$$\sigma = (1 - \dot{d})\bar{\sigma}, \tag{5}$$

where the parameter $\dot{d}$ has been calculated from the above formula and $\bar{\sigma}$ is the equivalent stress calculated by using the plastic equation assuming that the material has not failed.

2.3. Geometric and Mesh Models. The multiparticle repeated impact finite element model established in this paper is shown in Figure 2. In order to consider the influence of particle corner, three different particle shapes: spherical, tetrakaidecahedron, and triangular prism particles, are adopted. For repeated impacts, a six-particle impact model is set. The initial spacing between particles, initial impact angle, and velocity are all equal to ensure continuous impact on the same position of the target to obtain a stable erosion rate. In order to reduce the influence of boundary effect on simulation results, the target was selected as a rectangular solid of 1200 μm × 1200 μm × 360 μm; that is, the size of the target in the horizontal direction and the thickness direction was 10 times and 3 times of the particle diameter, respectively. Considering that erosion mainly occurs on the contact area in the center, the region of 360 μm × 360 μm × 36 μm in the center is selected for mesh refinement.

2.4. Boundary Conditions. It is assumed that the solid particles are less deformed in the impact process, and the reference point is the center of mass. In the initial analysis step, the particles and the target were defined as the surface contact, the particle surface was the main surface, the point set in the contact area was the slave surface, and no contact was set between the particles. Normal contact obeys the kinematics contact condition, tangential contact obeys the coulomb friction law, and the mode of slip in contact is the finite slip. The target in the study is equivalent to a small piece embedded in an infinite plate, so all degrees of freedom at the bottom of the target are constrained.

At present, there are four main sources of solid particles in gathering and transportation pipelines: formation particles carried by natural gas, cement, and fracturing proppant produced by drilling, completion, and stimulation processes; rust particles produced by wear and corrosion of pipelines and ground pipelines; and impurity particles settled during pipeline construction, with a maximum diameter of about 200 μm. According to the working conditions of natural gas gathering pipelines in Sichuan area, the impact velocity selected in simulation is 10 m/s to 20 m/s.

2.5. Model Validation. Yerramareddy and Bahadur carried out erosion experiments on Ti-6Al-4V wafer with the sand blasting machine [17]. The impact particle material was SiC with a diameter of 120 μm and density of 3200 kg/m^3. In order to compare with the experiment, the tetrakaidecahedron particle material in the finite element impact model is SiC with the equivalent diameter of 120 μm, and the target material is Ti-6Al-4V. The repeated impact particle number is six, and the impact angles are 20°, 30°, 45°, 60°, and 90° respectively. The time step of simulation is 1.4×10^{-5} s, which is enough to ensure that all particles finish the impact, and the kinetic energy tends to be stable.

After the simulation of 6 times repeated impacts is completed, the equivalent stress distribution on the surface of the target is shown in Figure 3. As can be seen from the figure, some elements are deleted from all angles, and more elements are deleted when impact angles are 20°, 30°, and 45°. At a low impact angle, there is a raised lip on the right side of the impact pit, which is consistent with the phenomenon pointed out by cutting theory. However, with the increase of the impact angle, the depth of stress effect increases continuously because the component velocity in the z direction increases with the increase of angle.

Figure 4 shows the statistical change trend of erosion rate at different impact angles. It can be seen that the erosion rate increases first and then decreases. When the impact angle is 30°, the erosion rate reaches its maximum value. When the impact angle increases from 30° to 90°, the erosion rate decreases rapidly. Compared with the experimental results, it can be seen that the simulated erosion rate is larger than the experimental value, but the development trend of the curve is basically the same, and there is no difference of orders of magnitude in the numerical value. The reason for the difference between the experiment and simulation is that numerical simulation is an ideal case where the sharp corner of each particle contacts the same position of the target. In the experiment, the placement of the particles is relatively dispersed, and a large number of particles are needed to impact to meet the repeated impact at any position of the material

(a) (b) (c)

Figure 2: Geometric model of the impact target. (a) Spherical model. (b) Tetrakaidecahedron model. (c) Triangular prism model.

surface, so the experimental results will significantly be smaller than the simulation results.

3. Results and Discussion

3.1. Changes of Stress and Energy during Impact. In this paper, the impact failure morphology of material surface, the energy change in the system, and the movement rule change in the single particle are firstly observed. Particle shape is tetrakaidecahedron, equivalent diameter of 120 μm, density of 2600 kg/m^3, and impact speed of 15 m/s and the impact angles are 15°, 30°, 45°, 60°, 75°, and 90°, respectively, in the target material 20# steel (physical parameters and failure parameters shown in Table 2); the whole analysis step time is 2×10^{-6} s.

Figure 5 is a time-dependent curve of the equivalent stress at the center of the erosion zone when particles contact the target at different angles. It can be seen that the equivalent stress of the material is almost over 600 MPa, exceeding the strength limit of the material, and the material is already in the state of damage accumulation or failure, but the change trend is different under different impact angles. When the impact angles are 15° and 75°, respectively, it shows an obvious valley value. When the impact angle is 45°, two peaks and a valley value appear.

This indicates that the plastic stress of the target is constantly changing and related to the failure of the material element. At the beginning of the collision, the sharp part of the particle intrudes and pushes the target material, which causes plastic deformation. In this process, the damage of contact element continuously accumulates until the failure limit value is reached. The element is deleted and the stress is rapidly reduced. If the particles at this time still maintain enough kinetic energy, the subsequent exposed elements will be failed and be deleted, and a peak and a valley value will be formed again, which is like the process of cutting material, indicating that the microcutting phenomenon does exist.

Figure 6 shows the change of kinetic energy of particles with impact angle. It is found that the kinetic energy of particles decreases first and then stabilizes at each impact angle. Some kinetic energy is lost when particles contact the target, which is used to produce plastic deformation on the surface of the target and increase the internal energy of the whole system. When the particle completes the impact and leaves the target, it retains a certain kinetic energy, which corresponds to the stable section of the curve.

The impact time of particles at different impact angles is different. When the impact angle is 15°, the impact time is the longest. When the impact angles are 30° and 45°, the impact time decreases in turn. When the impact angles are 60°, 75°, and 90°, the impact time is close and lower than that of the low angle impact. This is because the particles mainly cut the target at a low impact angle, and the velocity component in the X direction determines the cutting ability. The stronger the cutting action, the longer the particles stay on the target surface. On the contrary, the plastic fluidity of the target in the Z direction is weaker than that in the X direction. The bigger the impact angle, the faster the material enters the hardening stage. Therefore, the impact time at a high impact angle is relative short.

Figure 7 shows the change of kinetic energy consumption of particles under different impact angles. As mentioned above, the amount of kinetic energy consumption can reflect the severity of target erosion. The longitudinal coordinates in the figure represent the ratio of particle kinetic energy consumed to initial kinetic energy. It can intuitively be seen that different impact angles have different effects on erosion. The minimum kinetic energy consumption is 58% when the impact angle is 15°, and the maximum value is 99% when the impact angle is 60°. This indicates that serious plastic deformation occurs when the particles are embedded in the target.

3.2. Effect of Particle Velocity. The effect of velocity on erosion rate was studied by using the finite element model of continuous impact of particles on the target. In the simulation, the particle shape is tetrakaidecahedron, the particle equivalent diameter is 120 μm, and the particle density is 2600 kg/m^3. According to the current velocity range of gas gathering pipeline, five kinds of particle velocities are determined: 10 m/s, 12.5 m/s, 15 m/s, 17.5 m/s, and 20 m/s. The impact angles of particles are 30° and 75° as representatives of high and low impact angles. It can be seen intuitively from Figure 8 that the number of deleted elements in the target increases with the increase of particle velocity, and the

Figure 3: Erosion stress contour of the target material after repeated impacts at different impact angles. (a) 20°, (b) 30°, (c) 45°, (d) 60°, and (e) 90°.

impact damage to the target becomes more and more serious.

Figure 9 shows the increase of the erosion rate under two impact angles in the form of a curve. It can be seen from the figure that the erosion rate increases dramatically with the increase of impact velocity. When the impact angle is 30° and the velocity increases from 10 m/s to 20 m/s, the erosion rate increases from 0.33 mg/g to 1.35 mg/g, which is four times the original value. When the impact angle is 75°, the velocity increases from 10 m/s to 20 m/s, and the erosion rate increases

Figure 4: Change of the erosion rate with the impact angle.

Figure 5: Equivalent stress changes at the erosion center with time.

Figure 6: Change of kinetic energy of particles with time.

from 0.28 mg/g to 1.49 mg/g, which is 5.3 times of the original value. It can be seen that, regardless of the impact angle, the impact velocity is generally low, the erosion rate is very sensitive to the change of impact velocity, and the sensitivity can be seen from the velocity index. Through power function regression of the two angle curves, it is found that the velocity exponent is 2.04 when the impact angle is 30° and the velocity exponent is 2.41 when the impact angle is 75°. The velocity

Figure 7: Ratio of kinetic energy consumption varies with the impact angle.

index predicted by using the finite element method coincides with the conclusions of the previous studies.

The reason for the velocity index is different in the case of high and low impact angles may be that the contact position of particles changes constantly in the process of continuous impact. At the low impact angle, when the velocity is large enough, the particle tip removes the target surface material by cutting or ploughing, and the impact point of subsequent particles always remains at the end of the cutting groove, which is conducive to lip extrusion but not conducive to lip removal, so the contribution of the increase of kinetic energy to the erosion rate decreases. Deformation and wear occur on material surface under high angle impact as long as the speed is sufficient to cause the plastic deformation. Particle kinetic energy increases rapidly with the increase of impact velocity, and the impact pit deepens and the plastic deformation of the extrusion lip increases. Under the repeated impacting, the material is easier to harden and exfoliate, which makes the erosion rate raise and occur faster, and the speed index is larger.

3.3. Effect of Impact Angle. Similarly, the finite element model of continuous impact of particles on the target material was used to study the impact angle on the erosion characteristics. Other conditions were the same as in 3.2, and the particle velocity was 15 m/s. The stress of target material surface continuously impacted by particles under different impact angles was calculated. It can be found that serious deformation and failure deletion of elements occur on the target surface, and the stress contour changes obviously with different impact angles. The results show that the peak erosion rate appears at 45°, and the peak erosion rate of ordinary plastic materials occurs at 30~35°. It is shown that the angle of the peak value of the erosion rate depends on the target material characteristics, and the development trend of the simulated erosion rate curve is to increase first and then decrease.

The simulation results compared with the erosion prediction model established by Finnie [15] is shown in Figure 10. It can be seen that the calculated value of the Finnie

Figure 8: Continued.

Figure 8: Stress contour of the target at different impact velocities: (a) 10 m/s, (b) 12.5 m/s, (c) 15 m/s, (d) 17.5 m/s, and (e) 20 m/s (the left figures are 30° and the right figures are 75°).

Figure 9: Effect of impact velocity on the erosion rate.

Figure 10: Comparison of simulation results with the Finnie model.

model is larger when the impact angle is lower than 30°. The biggest difference between the two models is that the predicted value of the Finnie model decreases rapidly after the impact angle is higher than 30°, while the erosion rate decreases slowly when the impact angle is larger. Many scholars believe that the Finnie model has good prediction ability at the low impact angle, while the prediction error of the erosion rate at the high impact angle is larger [1, 4, 10]. The variation trend of the simulation curve in the figure is close to the Finnie model at the low impact angle, and the numerical difference is very small. This proves again the validity of finite element simulation, and the development trend of the simulation curve at the high impact angle is closer to reality.

3.4. Effect of Particle Diameter. Erosion-related parameters, such as particles equivalent diameter of 50 μm, 120 μm, 160 μm, and 200 μm; particle shape of tetrakaidecahedron; density of 2600 kg/m^3; impact speed of 15 m/s; and impact angle of 45°, are selected in this article. Figure 11 shows the stress contour of the section after the impact by four kinds of particles with different equivalent diameters. It can be seen from the figure that with the increase of equivalent diameters of particles, more and more elements are deleted.

Figure 12 shows the trend of erosion rate with the increase of particle diameter. It can be seen that when the particle diameter is 50 μm, the minimum erosion rate is 0.38 mg/g, and when the particle diameter is 200 μm, the maximum erosion rate is 1.2 mg/g. When the diameter of particles increases from 50 μm to 160 μm, erosion rate increases rapidly. When the diameter of particles increases from 160 μm to 200 μm, the erosion rate changes gently. The rule that the erosion rate increases rapidly first and then tends to be stable with the increase of particle diameter has been mentioned by other scholars [16, 17].

The reason for the erosion rate tending to be a constant value with the increase in particle diameter is that the particle mass increases with the diameter, and its kinetic energy and particle surface area also increase synchronously. However, the growth rate of the deleted mass on the surface of the target material could not keep up with the growth rate of the particle mass, so the proportion tended to be stable. On the contrary, the effect of particle size may affect the impact response of the target material, and the change of the hedging corrosion rate also has an effect.

3.5. Effect of Particle Shape. In order to study the effect of particle shape on the erosion characteristics of target, spherical, tetrahedral, and triangular prism shapes were set from blunt to sharp, and the effects of the impact angle, impact velocity, and particle equivalent diameter on the erosion rate were simulated separately.

The results for tetrahedral particles are shown in Section 3.3 in this article. The erosion results of a sharp triangular prism and spherical particles at different impact angles are shown in Figure 13. The equivalent diameter of particles is 120 μm, and the impact speed is 15 m/s. It can intuitively be seen that the failure and deletion of elements occur at all angles due to the erosion of the triangular pyramid, while the failure and deletion of elements occur only at 45° due to the spherical particles.

Figure 14 shows the variation of the erosion rate with the impact angle of three different shaped particles with an equivalent diameter of 120 μm at a velocity of 15 m/s. It can be seen from the figures that the change trend of the erosion rate curves of the three shape particles is about the same. The spherical particles are special, when the impact angles are 15°, 60°, and 75° and erosion rate is 0. When the impact angle is 45°, the corresponding erosion rate of each shape particle reaches the maximum value, which indicates that the impact damage of the particles on the 20# steel target is the most serious when the impact angle is 45°. Comparing the influence of particle shape, it is found that the severity of impact damage on the target is triangular prism, tetrakaidecahedron, and sphere in descending order, which indicates that the sharper the particles are, the greater the impact damage is, which accords with the erosion characteristics of plastic materials

Figure 15 shows the statistical erosion rate of the target impacted, respectively, by the triangular prism particles and spherical particles with an equivalent diameter of 120 μm, impact angle 30°, and velocity of 10 m/s to 20 m/s. It can be seen that the erosion rates of triangular prism particles and tetrakaidecahedron particles increase rapidly with increasing velocity. The velocity indices are 2.31 and 2.04, respectively. The erosion rate is 0 of spherical particles when the impact velocity is less than 15 m/s, and the erosion rate increases rapidly when the velocity is greater than 15 m/s.

The variation of the erosion rate of different shape particles with different equivalent diameters is shown in Figure 16. It can be seen from the figure that the curve corresponding to the triangular prism and the tetrakaidecahedron has the same trend, and the erosion rate is first increased rapidly and then tends to be stable, while the sphere shows a tendency to increase first and then decrease, and the erosion rate is much smaller than that of sharp particles.

3.6. Influence of Stress Interference on Erosion Behavior

3.6.1. Particle Random Impact Model. Natural gas gathering pipelines have small particle content, and few particles continuously impact the same position. In order to study the influence of stress interference on the impact results, it is necessary to consider the randomness of the falling points. In this paper, the random impact of particles on the surface of the material is used to study the stress interference.

The random erosion model uses spherical particles with a particle diameter of 120 μm, a density of 2600 kg/m^3, an impact velocity of 15 m/s, and an impact angle of 45°. In the simulation, 10 layers of particles are generated, and the height difference between layers is 121 μm. Given five particles in each layer, the particle position is determined by using the random function to ensure that the spacing between particles is not less than particle diameter. As in the previous six-particle model, there is no contact between particles, and the finite element mesh of the stochastic impact model is obtained as shown in Figure 17.

3.6.2. Stress Interference Analysis. Figure 18 shows the stress contour distribution of the surface after 50 spherical particles randomly impacting the target. It can be seen from the figure that the particles with similar falling points show obvious stress interference, and the maximum stress is 551.2 MPa. A large number of elements have exceeded the strength limit of the target and are in a failed state, while a few elements that have experienced multiple similar point impacts are deleted. Comparing the impact angle of 45° and the speed of 15 m/s, the simulation results of the six-particle repeated impact at the same position show that the element is deleted when the same position is subjected to the third impact of the spherical particles, indicating cumulative

Figure 11: Impact stress contour impacted by tetrakaidecahedron particles with different equivalent diameters: (a) 50 μm, (b) 120 μm, (c) 160 μm, and (d) 200 μm.

Figure 12: Change of the erosion rate with particle diameter.

impact damage and drop point location coincidence degree have a direct relationship. Further combined with the drop point diagram, it is found that the larger equivalent stress value appears in the places where several drop points are very close to each other, which proves that the closer the drop point position is, the more obvious the cumulative failure effect is. Therefore, it can be predicted that if the number of particles and the number of impacts continue to increase, the first place where the element is deleted must be the place where these equivalent stress values are the largest.

It can be seen from the figure that although a few point position is relatively far apart, there is a clear stress communication between the falling points in the contour image, and these connected areas are indirectly affected by the particles. The force is sufficient to cause plastic deformation of the element therein, which confirms the existence of stress interference. PJ Woytowitz believes that the stress interference damage increases first and then decreases with the increase of the distance of the falling point. When the ratio of the distance between the falling point and the particle radius is around 0.125, the stress interference is the strongest [18].

Figures 19(a)–19(c) correspond to the ZX plane, cross section of $Y = -110\,\mu$m, $Y = -60\,\mu$m, and $Y = 0\,\mu$m,

Figure 13: Continued.

Figure 13: Impact stress contour impacted by multiparticles at different impact angles: (a) 15°, (b) 30°, (c) 45°, (d) 60°, (e) 75°, and (f) 90°(the left figures are of triangular prism shape, and the right figures are of spherical shape).

Figure 14: Effect of the impact angle on the erosion rate.

Figure 15: Effect of impact velocity on the erosion rate.

respectively (the specific position can be referred to the ordinates of Figure 17(a)). As shown in Figure 19, the four falling points on the left side in (a) are relatively close, with the spacing less than 10 μm, the individual elements have been deleted, and the maximum downward influence depth is 15 μm, with strong stress interference. In (b), the distance between the three falling points on the right side is about 20 μm, and the maximum influence depth is 12 μm, and the stress interference is obvious. In (c), the spacing between the falling points exceeds 60 μm, and the maximum influence depth is not more than 9 μm, and it is basically free from stress interference. Therefore, for particles with an equivalent diameter of 120 μm, it can be considered that when the ratio between the distance of falling point and particle radius exceeds 0.16, the stress interference will obviously aggravate the erosion damage of particles to the target.

Figure 16: Effect of particle diameters on the erosion rate.

Figure 17: Random particle model and location of the drop. (a) Drop location map. (b) Random impact model grid.

Figure 18: Stress contour impacted randomly by 50 spherical particles.

Figure 19: Stress contour result after filtering (>258 MPa) on ZX plane cross section: (a) $Y = -110\,\mu$m, (b) $Y = -60\,\mu$m, and (c) $Y = 0\,\mu$m.

4. Conclusion

(1) According to the working conditions of natural gas gathering pipelines, the 20# steel erosion dynamics model based on the finite element method was established. The Johnson–Cook plasticity and failure model were used to characterize the stress, strain, and failure behavior. The simulation results are in agreement with the experimental results.

(2) The variation of kinetic energy of impact particle at different angles was obtained. It was found that the main effect of particle on the target is microcutting at the low impact angle. The kinetic energy loss tends to increase first and then decrease with the increase of the impact angle, and the contact time between particle and target decreases with the increase of the impact angle.

(3) The effects of the impact angle, impact velocity, particle size, and particle shape on the impact stress distribution and erosion failure rate of the material were obtained by simulation. Compared with the empirical model and experimental results, the simulation results show a high degree of rationality.

(4) The stress interference research model under random drop impact conditions is established to obtain the surface stress distribution law under the impact of adjacent particles. When the ratio of the drop distance to the particle radius exceeds 0.16, stress interference will significantly aggravate the erosion damage of the target.

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

Dynamic Crack Propagation and Fracture Behavior of Pre-cracked Specimens under Impact Loading by Split Hopkinson Pressure Bar

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Deformation and fracture of brittle materials, especially crack propagation, have drawn wide attention in recent years. But dynamic crack propagation under impact loading was not well understood. In this paper, we experimentally tested Brazilian disk (BD) fine sandstone specimens containing pre-cracks under cyclic impact loading by the Φ 74 mm diameter split Hopkinson pressure bar (SHPB) test device. The pre-cracked specimens were named central straight through crack flattened Brazilian disk (CSCFBD). By using the low air-pressure loading conditions (0.1 MPa, equal to the impact velocity of 3.76 m/s), a series of dynamic impact tests were detected successfully, and the effects of pre-cracks on dynamic properties were analyzed. Experimental results show that the multiple cracks mostly initiate at/or near the pre-crack tips and then propagate in different paths and directions varying by inclination angles, leading to the ultimate failure. Compared to static or quasi-static loading, dynamic crack propagation and fracture behavior are obviously different. Furthermore, we characterized the crack propagation paths, directions, and fracture patterns and discussed the influences of the pre-cracks during the breakage process. We concluded that the results obtained are significant in investigating the failure mechanism and mechanical properties of brittle materials under impact loading.

1. Introduction

Dynamic deformation and fracture of brittle materials are complex processes. Mining, tunnel excavation, and natural disasters such as landslides and earthquakes are all involved in the problems of dynamic damage. The failure of brittle materials is usually associated with crack propagation initiated from natural or artificial pre-existing defects. Moreover, the mechanical properties of brittle materials are closely related to external loading conditions, such as loading rate and load magnitude [1–3]. Rock is one of the most complex brittle materials containing different scale voids, cracks, and other defects, as shown in Figure 1, which are the main mechanical factors that affect the rock deformation and failure [4]. Crack growth and catastrophic failures initiated from pre-existing defects subjected to multiaxial loads are the main concerns for geotechnical engineers and designers of underground structures. The defects in rock can promote the initiation of new defects, which in turn may propagate and coalesce with other defects, and then can further decrease the strength of the rock mass [5–8]. The presence of pre-cracks may obviously reduce the fracture toughness, dynamic uniaxial compressive strength, and dynamic tensile strength and lead to fragmentation and multiple crack interactions, branching, and coalescence [9].

Due to the difficulties of in-situ tests, the laboratory experiment is an important and effective research method to investigate rock failure modes and fracture mechanisms. Over the past few decades, many experiments have been devoted to use semi-circular core in the three-point bending (SCB) specimen [10], Brazilian disk (BD) specimen with chevron flaws or other pre-existing flaws [11], radial cracked

Figure 1: Schematic view of pre-existing defects in the natural rock mass [4].

ring (RCR) specimen [12], and modified ring (MR) specimen [13] to investigate fracture toughness and crack propagation. Note that, in previous studies, the fracture behavior and crack propagation of brittle materials were mainly investigated under static/quasi-static loading or used in intact specimens. Irwin et al. [14–16] divided the simple cracks into three types in the basic failure process: Mode I (the tension/opening mode) crack, Mode II (the sliding mode), and Mode III (the tearing mode) crack. In engineering, Mode I cracks are of prime importance. More complicated cracks can form from these simple cracks, called mixed-mode cracks. Many experimental studies have been conducted to explain the crack initiation, propagation, and coalescence in pre-cracked brittle materials under static or quasi-static loading [17–21]. The BD test is an effective way to study crack propagation and fracture behavior of brittle materials [22, 23]. Al-Shayea [24] studied crack propagation paths in pre-cracked limestone central straight though crack Brazilian disk (CSCBD) specimens loaded with diametrical compression, and they investigated the possibility of using outcrop specimens to estimate the fracture toughness behavior of the reservoir rock at in-situ conditions of temperature and confining pressure. Haeri et al. [25–27] used Portland Pozzolana Cement (PPC) experimentally and studied crack propagation and crack coalescence of BD pre-cracked and pre-holed specimens under quasi-static compressive and tensile loading. Aliha and Bahmani [28] investigated fracture toughness under mixed-mode loading using different cylindrical and disc shapes for the brittle material. However, in deep strata, the damage and destruction of rock mass are often caused by dynamic even cyclic dynamic loading. None of these investigations listed above has ever captured the dynamic multiple crack propagation. In such cases, it is of great significance and essential to investigate the dynamic crack propagation and fracture patterns of rock in deep strata under impact/cyclic impact loading.

Various experiments devices have been used to explore a wide range of strain rates. Split Hopkinson pressure bar (SHPB) technique, which decouples cleverly the inertia effect in structures and strain rate effect in materials, has been widely used to characterize the dynamic performance of various engineering materials at high strain rate, such as rock [29–33], concrete [34–38], and ceramics [39, 40] at high strain rates (10^2~10^4 s^{-1}). The strain rate sensitive behavior of brittle materials has been under investigation for several decades. The strain rate sensitivities are mainly measured by strength or the strains at the maximum stress [41, 42]. In recent years, a variety of researchers have investigated and demonstrated the dynamic properties of natural or artificial brittle materials [43–47], and they found that the dynamic strength (including dynamic compressive strength and dynamic tensile strength) and impact toughness increases with strain rate and the strain rate sensitivity of brittle materials.

In this paper, we defined some central straight through crack flattened Brazilian disk (CSCFBD) specimens to investigate dynamic crack propagation and fracture patterns under impact loading by the SHPB device, and the term "pre-crack(s)" is used to describe the artificially created crack. The motivation for this work will focus on the following two points: (1) investigating the dynamic fracture patterns, multiple crack propagation paths, and directions in pre-cracked specimens; (2) characterizing and analyzing the crack types initiated from pre-cracks under cyclic impact loading. This paper is organized as follows: in Section 2, the preparation of tested CSCFBD specimens and experimental procedure are discussed. The SHPB experimental scheme is also introduced. In Section 3, we present the experimental results. We look at both the multiple crack propagation paths and directions. In Section 4, we discuss the experimental results and characterize the dynamic fracture patterns and failure modes. The conclusions concluded upon the foregoing results are given in Section 5.

2. Preparation of Disk Specimens and Experimental Procedure

Due to the crystalline and blocky structures, fine sandstone is widely used to investigate the fracture behavior and crack propagation of defected brittle materials. In this paper, we used fine sandstone to study the dynamic crack propagation under impact loading. The fine sandstone samples were excavated by geologic drilling from about 900-meter depth underground strata in Juye coalfield, whose Cenozoic formation is very thick. The average thickness of strata in the fourth system is 158.43 m, and the average thickness of the upper tertiary strata is 497.01 m. And the thickness of the new boundary layer is 530~720 m, mainly composed of clay, sandy clay, sand, fine sand, and gravel. Main coal seam roof and floor sandstone thickness is 4.80~75.65 m, mainly fine sandstone, local sandstone, and siltstone [4]. The pre-cracked specimens' preparation and experimental procedures will be explained as follows.

2.1. Preparation of CSCFBD Specimens. For manufacturing CSCFBD specimens, the whole tests used six samples with each one cored into 30~50 cm long, diameter $D = 62$ mm cylindrical columns in the construction site. In order to avoid external environmental influence, the surface of the samples was wrapped in the multilayer food preservation film after removing from the formation. The specimens were cut into BD shapes with the size of Φ 62×30 mm. Three

specimens with same pre-crack geometry were prepared to guarantee the reproducibility of these experimental tests, and a total number of 21 CSCFBD specimens were manufactured in this work. The physical and mechanical properties of the tested fine sandstone are listed in Table 1.

Various Brazilian tests were conducted on CSCFBD specimens containing a single crack with different inclination angles. These pre-cracks were created by the high-speed water jet cutting machine [4], as shown in Figure 2. Figure 3 shows pre-cracked specimens with different inclination angles and the crack inclination angle β (the angle between the normal line of the pre-crack surface and the vertical direction of the impact load), $\beta = 0^\circ$, 15°, 30°, 45°, 60°, 75°, and 90°. Figure 4 shows a schematic view of CSCFBD specimens. The pre-cracks length, $2b$ (b is the half of pre-crack length), is equal to 10 mm, and the crack width is 1 mm. The radius and thickness of the CSCFBD specimen are $R = 31$ mm and $H = 30$ mm. Crack length ratio is an important parameter for the pattern, trajectory, and the number of fractures; in this work, the ratio b/R is 0.16.

2.2. SHPB Experimental Procedure. In this work, the dynamic tests were taken in the structure laboratory of Hohai University, adopting Φ 74 mm diameter straight taper variable cross sections SHPB device. The Φ 74 mm diameter SHPB device mainly composes of the power system (which propelled by a gas gun), elastic bars (an incident bar, a transmitter bar, and an absorbing bar), a damper, energy absorbing setup, high dynamic strain indicator and data processing systems. The power system consists of an air compressor and pressure vessel. The impact velocity is measured by the light electric tachometer. Schematic of the SHPB test device is shown in Figure 5.

The SHPB device is based on the one-dimensional theory of elastic wave and uses Lagrangian coordinates to describe all the physical parameters. The control equations of waves in the bars are on the following assumptions. One-dimensional assumption: speed v and the strain ε are only the function of point X and time t and the assumption that the strain rate is independent. Assuming axial wave propagation and homogeneous stress distribution in the specimen, the resulting stress $\sigma_s(t)$, strain $\varepsilon_s(t)$, and strain rate $\dot{\varepsilon}(t)$ of the specimen are obtained by the following equation [48]:

$$\begin{aligned} \sigma_s(t) &= \frac{S_B E}{2S_S}\left[\varepsilon_t(t) + \varepsilon_r(t) + \varepsilon_i(t)\right], \\ \dot{\varepsilon}_s(t) &= \frac{C_0}{L_s}\left[\varepsilon_t(t) + \varepsilon_r(t) - \varepsilon_i(t)\right], \\ \varepsilon_t(t) &= \frac{C_0}{L_S}\int_0^t \left[\varepsilon_t(t) + \varepsilon_r(t) - \varepsilon_i(t)\right] d\tau, \end{aligned} \tag{1}$$

where S_B, E, and C_0 are the cross-sectional area (mm^2), Young's modulus (GPa), and the wave velocity (km/s) of the bar material and L_S and S_s are the length (mm) and cross-sectional area (mm^2) of the specimen. $\varepsilon_i(t)$, $\varepsilon_r(t)$, and $\varepsilon_t(t)$ are the strain singles in specimen. The diameter, Young's modulus, and density of the elastic bars are 74 mm, 210 GPa, and 7850 kg/m^3, respectively.

Table 1: Physical and mechanical properties of the tested fine sandstone.

Properties	Value
Young's modulus, E (GPa)	11.1
Compressive strength, σ_c (MPa)	84.1
Brazilian tensile strength, σ_t (MPa)	3.0
Longitudinal wave velocity, V_P (m/s)	3526.4
Porosity (%)	8.3
Poisson's ratio, ν	0.33
Density, ρ (kg/m^3)	2692

Figure 2: Pre-crack preparation in CSCFBD specimens using a water jet cutting machine [4].

Figure 3: Schematic view of CSCFBD specimens with different inclination angles of pre-cracks.

In fact, during the experimental tests by BD specimens, the cracks of samples are first produced by the center and then expanded along the radial direction. Wang et al. [49] presented that if the samples between two planes parallel to the plane of the degree of smoothness and not less than 0.05 mm, then they can ensure that the fracture initiated from the specimen center. The loading areas corresponding

Figure 4: Schematic view of the fine sandstone CSCFBD specimen.

Figure 5: Schematic of the split Hopkinson pressure bar (SHPB) device.

to the center angle 2α to meet $20^\circ \leq 2\alpha \leq 30^\circ$. Figure 6 shows the flattened BD specimen under radial loading. Before the installation of the specimens, a layer of Vaseline is evenly applied to the contact between the end of the specimen and the end of the compression bar, and the specimens need to be tightened between the incident bar and the transmitter bar. During SHPB tests, the brittle materials may fail before stress uniformly is achieved within the specimens. Modification of the incident pulse to closely match the elastic response is required. The pulse shaping technique has been widely applied in SHPB testing of engineering materials, and it is especially used for investigating the dynamic response of brittle materials. In this paper, we used a pulse shaping technique, a thin copper disk (12 mm-diameter and 1 mm-thickness) which was placed on the impact side of the incident bar. The pulse shape can attenuate high-frequency oscillations of the incident stress wave to improve the stress wave shape. Pulse shaping technique allows for controlling the damage of brittle materials [29, 30].

Figure 6: Flattened BD specimen under radial impact loading.

3. Experimental Results

3.1. Impact Time Analysis of CSCFBD Specimens. During testing, the air pressure in the gas gun chamber was kept constant 0.1 MPa, equal to the impact velocity 3.76 m/s, and the bullet must be brought back to its original position before the next loading. The times of impacts were recorded for each specimen until final failure. As shown in Figure 7, to reach the final crack propagation paths forms, 6 times impact loading are needed when $\beta = 0^\circ$ and the impact times are 3, 4, 6, 4, 2, and 2 when $\beta = 15^\circ$, 30°, 45°, 60°, 75°, and 90° respectively. It is obvious that the impact times in final failure are related to the pre-existing inclination angles. Because pre-cracks can make the specimen strength to reduce, so when $\beta = 15^\circ$, 75°, and 90°, the impact times are usually less than the cases of $\beta = 0^\circ$, 30°, 45°, and 90°.

3.2. Dynamic Crack Propagation Paths and Directions of CSCFBD Specimens. In this paper, we investigated the dynamic crack propagation paths and directions in CSCFBD specimens. In Figure 8, the crack propagation paths in

Figure 7: Impact times in final crack propagation path forms when $\beta = 0^\circ$, 15°, 30°, 45°, 60°, 75°, and 90°.

CSCFBD specimens with different inclination angles are indicated, and for all figures, the upper plane edges of the specimens are contacted with the incident bar. As shown in Figures 8(b)–8(g), cracks initiated from the tips of pre-cracks and approximately propagated towards the direction of the maximum stress. It should be noted that, in Figure 8(a), there is a crack initiated from the middle portion of the pre-crack when $\beta = 0^\circ$.

Figure 8(a) shows that five crack propagation paths appeared: two cracks started from the pre-existing crack's left-end tip and propagated to the specimen's lower plane edge; two cracks started from the pre-existing crack's right-end tip; the rest one crack propagated to the upper and lower plane edges. However, because of the high values of crack orientation with respect to the loading direction, there is one crack which did not propagate from the tip of the pre-crack. In Figures 8(c) and 8(d), $\beta = 30^\circ$, 45° respectively, five crack propagation paths appeared. Figure 8(c) shows that there are two cracks that started from the pre-crack upper tip and propagated to the upper plane edge, two cracks started from the pre-crack lower tip and propagated to the lower plane edge, and the rest one crack started from the pre-crack upper tip and propagated to the lower plane edge. While in Figure 8(d), there is one crack that started from the pre-crack lower tip and propagated to the upper plane edge. In Figures 8(b), 8(e), and 8(f), $\beta = 15^\circ$, 60°, and 75°, respectively, four main crack propagation paths appeared, and two of them started from the pre-crack upper tip and propagated to the upper plane edge. The other two started from the pre-crack lower tip and propagated to the lower plane edge. In Figure 8(b), an intermittent crack appeared, starting from the pre-crack lower tip, and has a downward trend to the lower plane edge. Figure 8(e) shows that the scatter of the four cracks presents regular symmetrical characteristic. As shown in Figure 8(g), $\beta = 90^\circ$, and the upper pre-crack tip propagated two crack propagation paths to the upper plane edge and one to the lower plane edge.

From Figures 8(a)–8(g), it can be clearly seen that the final dynamic crack propagation paths and fracture patterns under impact loading are obviously different compared to those under static or quasi-static loading (e.g., Figure 9 [4]). In many studies, there is only one fracture that propagates from each tip of BD specimens under static or quasi-static loading. But under impact loading in the SHPB test, due to fractures with high strain rate, there are multiple crack propagation paths. The results also show the pre-existing inclination angles affect the multiple crack propagation paths and directions initiated from tips of the pre-cracks under cyclic dynamic loading.

4. Discussion

The crack propagation patterns were obtained in previous investigations of brittle materials with pre-cracks as shown in Figure 10 [1, 2]. From the experimental results above, two types of cracks were observed: wing cracks and secondary cracks. Usually, the wing cracks are tensile cracks and the secondary cracks are shear cracks (oblique shear cracks and coplanar or quasi-coplanar shear). Table 2 summarizes crack types initiated from the pre-cracks in CSCFBD specimens under cyclic impact loading. Most of the tensile cracks and shear cracks initiated from the tips of pre-cracks at an angle and then propagate to the parallel to the compressive direction. Shear cracks' initiation patterns depend on the inclination angles of the pre-cracks. Under impact loading, shear cracks caused the failure of the tested specimens mostly.

Fracture and failure of pre-cracked CSCFBD specimens under cyclic impact loading involve tensile and shear crack types. Because of pre-cracks, the fracture patterns and failure modes are much more complex than those of intact brittle materials. Note that all the CSCFBD specimens present multiply crack propagation paths. But the geometries of pre-cracks appear to play limited effects on the crack type of CSCFBD specimens under cyclic impact loading.

In this paper, we studied the dynamic crack propagation and fracture patterns in pre-cracked CSCFBD specimens from deep underground strata under cyclic impact loading. The dynamic crack propagation paths and directions are obviously different from those under static or quasi-static loading in the previous studies, and the experimental results present some regular symmetrical characteristics. The dynamic crack propagation paths and propagation directions are not only determined by material related but also dependent on the geometries of the pre-cracks. However, the geometries of pre-cracks appear to play limited effects on the final dynamic fracture patterns and failure modes of CSCFBD specimens under cyclic impact loading. Further research may be a focus on the dynamic crack propagation mechanism of brittle materials studied by experimental and numerical simulation comprehensively. Various numerical simulation methods, e.g., finite element method (FEM), extended finite element method (XFEM), finite differential method (FDM), and different criteria, e.g., strain energy density (SED) criterion, cohesive zone model (CZM) [50] for theoretical analysis, have been developed to investigate

Figure 8: Continued.

(g)

Figure 8: Crack propagation paths in fine sandstone CSCFBD specimens with different inclination angles of pre-cracks: (a) $\beta = 0$; (b) $\beta = 15^\circ$; (c) $\beta = 30^\circ$; (d) $\beta = 45^\circ$; (e) $\beta = 60^\circ$; (f) $\beta = 75^\circ$; (g) $\beta = 90^\circ$.

(a) (b)

(c) (d)

Figure 9: Continued.

Figure 9: Experimental results of crack propagation paths in tested specimens with different inclination angles of pre-cracks: (a) $\beta = 0$; (b) $\beta = 15^\circ$; (c) $\beta = 30^\circ$; (d) $\beta = 45^\circ$; (e) $\beta = 60^\circ$; (f) $\beta = 75^\circ$; (g) $\beta = 90^\circ$ [4].

Figure 10: Various crack types initiated from the pre-cracks identified in the present studies (T and S denote tensile and shear cracks).Type I: tensile crack; Type II: tensile crack; Type III: tensile crack; Type IV: mixed tensile-shear crack; Type V: shear crack; Type VI: shear crack (coplanar or quasi-coplanar shear crack); Type VII: shear crack (oblique shear crack) [1, 2].

Table 2: Types of cracks initiated from pre-cracked CSCFBD specimens under cyclic impact loading.

Tested CSCFBD specimen no.	Crack types						
	Type I tensile	Type II tensile	Type III tensile	Type IV tensile-shear	Type V shear	Type VI shear	Type VII shear
$\beta = 0^\circ$	√	√	√	√			
$\beta = 15^\circ$	√	√		√			
$\beta = 30^\circ$	√	√	√	√	√		
$\beta = 45^\circ$	√	√			√		√
$\beta = 60^\circ$	√			√	√		
$\beta = 75^\circ$		√	√	√			
$\beta = 90^\circ$	√		√	√			

crack propagation, fracture patterns, and failure modes in brittle materials but have not shown satisfactory effectiveness in modeling dynamic damage evolution and crack propagation [51–53]. Using mesh-free particle methods to numerically study dynamic crack propagation may be an effective way. Especially in peridynamics, cracks are part of the solution, not part of the problem [54]. Therefore, using peridynamics to simulate dynamic multiple crack propagation and fracture patterns in brittle materials under impact or cyclic impact loading would be very meaningful which we plan for in the future.

5. Conclusions

In this paper, we provide experimental results of dynamic fracture tests carried out on fine sandstone CSCFBD specimens under cyclic impact loading by the Φ 74 mm-diameter SHPB device. Dynamic crack propagation and fracture mechanism are rather complicated processes. From the results presented and analyzed above, the following conclusions could be drawn:

(1) Compared to static or quasi-static loading, the dynamic crack propagation and fracture behavior are much more complex, and it presents multiple crack propagation paths and directions in dynamic fracture.

(2) Both tensile and shear cracks were observed, most of them mainly initiated from tips of the pre-cracks and propagated in a stable manner. According to the different geometries of pre-cracks, the tested CSCFBD specimens experienced tensile or shear crack propagation failure.

(3) The natural or artificial pre-existing defects can change the crack propagation paths, especially directions under impact loading compared to static or quasi-static loading. However, the geometries of pre-cracks appear to play limited effects on cracks type of CSCFBD specimens under cyclic impact loading.

(4) Dynamic crack propagation and fracture mechanism are rather complicated processes. This study revealed multiple dynamic crack propagation behavior that had not been observed previously. Numerical simulation is further needed to be summarized and explored which we plan for in the future.

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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Research Article

An Approach to Estimate SCC Growing Rate in Slow Strain Rate Tensile Test Based on EPFEM

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The slow strain rate tensile test (SSRT) is a common means to investigate stress corrosion cracking (SCC) in key engineering structural materials of light-water reactors, and it is an important task to real-time monitor the crack growing length and rate of the specimen during the test. Because the specimen is placed in an autoclave with high-pressure and high-temperature water environment-simulated light-water reactor, the current potential drop method, which includes current potential drop (DCPD) and alternating current potential drop (ACPD), is the main means to real-time monitor crack growth rate in the SCC test. As a supplementary means to obtain the crack growth rate during the test, the SSRT process of nickel-based Alloy 600 CT specimens is investigated by using the elastic-plastic finite element method (EPFEM) in this paper. Based on the consideration that both the elastic-plastic deformation and crack length of the specimen would affect the relationship between the load and the displacement of the loading point during the SSRT test, the relationship between the loading point displacement caused by crack propagation ΔL_c and plastic deformation ΔL_p is separated by EPFEM. Then, the SCC crack growth rate and the real-time crack length are obtained. This proposed approach could be used to improve the test results in the SSRT test.

1. Introduction

To improve the corrosion resistance of nuclear power equipment, austenitic stainless steel and nickel-based alloys are widely used as the structural materials in the primary circuit of pressurized water reactors (PWRs). Both austenitic stainless steel and nickel-based alloys, however, were found to be susceptible to stress corrosion cracking (SCC) by both laboratory experiment and field experience [1–6]. The SCC is typically associated with the combined effect of residual stress and working stress in a high-temperature water environment; such cracking can affect the reliability, integrity, and economy of PWRs and may become a potential service life-limiting issue [7].

The method for quantitative monitoring of the SCC propagation behaviour of austenitic stainless steel in the high-temperature water environment has been investigated for many years. Hwang et al. have studied the acoustic emission (AE) characteristics of SCC on 304 austenitic stainless steel and conclude that during the SCC development process, specific AE waveforms were generated, which can distinguish the SCC stages by heating, saturation, maximum temperature and pressure, and leakage [8]. Kovac et al. have studied the IGSCC propagation of sensitized type AISI 304 stainless steel by using four measuring techniques such as AE, electrochemical noise (EN), digital imaging, elongation measurements, and crack propagation were observed in all measured signals, although AE activity occurred only when the crack was large enough. The propagation was also successfully monitored with digital imaging [9]. Satoh et al. compared the EAC growth rate in high-temperature water with or without irradiation by using quantitatively predicting the theoretical model of SCC combined with the finite element method. The results show that the finite element simulation of crack growth rate agrees with the theoretical calculation results under irradiation

conditions; however, there are deviations in results under nonirradiation conditions [10]. Shi et al. proposed an artificial neural network algorithm (ANN) for predicting the rate of SCC growth of Alloy 600. The results show that the crack propagation rate prediction results obtained by ANN are consistent with the experimental data [11]. Ritchie et al. used the finite element method and the electrical analogue method to study the relationship between crack length and voltage drop of CT specimens. The results show that the crack length and voltage drop are approximately linear [12]. Zhang et al. used the direct current potential drop (DCPD) to conduct SCC growth tests of cold-worked nuclear-grade 316L austenitic stainless steel containing various dissolved oxygen (DO) contents but no dissolved hydrogen. The results showed that the SCC crack growth rates increase with increasing DO content in the simulated PWR primary water environment [13]. Zhai et al. have investigated the SCC initiation of three mill-annealed Alloy 600 heats in simulated pressurized water reactor primary by using constant load tests equipped with DCPD measurement capabilities. The results showed that the initiation time in the ground samples was slightly delayed, and the SCC initiation times were greatly reduced by a small amount of cold work [14].

The slow strain rate tensile test (SSRT) is an important experimental method to evaluate the stress corrosion sensitivity of materials in the high-temperature water environment. In the SSRT test, the direct current potential drop (DCPD) or the alternating current potential drop (ACPD) method are usually used to the real-time measure the crack growth length [15, 16]. Since DCPD mainly uses time changes in the potential around the crack tip to estimate the crack propagation, the results are highly susceptible to the experimental environment. An estimated method of real-time crack growing rate during the SSRT test is proposed by using the elastic-plastic finite element method (EPFEM) in this paper. Based on the consideration that both the elastic-plastic deformation and crack length of the specimen would affect the relationship between the load and the displacement of the loading point during the SSRT test, the relationship between the loading point displacement caused by crack propagation ΔL_c and plastic deformation ΔL_p are separated by EPFEM. Then, the SCC crack growth rate and the real-time crack length are obtained.

2. SSRT Experiment in High-Temperature Water

2.1. Materials and Specimen. The type Alloy 600 selected from the safety-end welded joint in pressurized water reactors (PWRs) is used in the SSRT experiment. The chemical compositions of the material are shown in Table 1 [17, 18]. Standard 12.7 mm thickness compact tension (CT) specimen with 5% side grooves on each side was prepared, which is shown in Figure 1 [19].

2.2. Experiment Procedures. SCC crack growth tests were measured using a slow strain rate stress corrosion testing machine equipped with an autoclave, a high-precision heating controller, a servo loading system, and a direct current drop (DCPD) crack length measurement devices. The precrack length on the CT specimen is 10 mm, and the CT specimen is loaded by a displacement load. The load-line displacement change rate $V = 0.005$ mm/h, and the test period is 900 hours. The stress corrosion test procedure for Alloy 600 is shown in Figure 2(a). The CT specimen with precrack is fixed by a pin, and a load P was applied to the specimen in the vertical direction as shown in Figure 2(b); the load-line displacement and the crack growth behaviour during the tests was separated monitored by using the sensor on the SSRT machine and the DCPD device.

3. Calculation Model

3.1. Material Model. The true stress-strain curve of Alloy 600 can generally be expressed in Ramberg–Osgood relationship, and the equation is written as

$$\frac{\varepsilon}{\varepsilon_0} = \frac{\sigma}{\sigma_0} + \alpha\left(\frac{\sigma}{\sigma_0}\right)^n, \tag{1}$$

where σ_0 is the yield strength of the material, ε_0 is the yield strain of the material, respectively, α is the yield offset coefficient of the material, and n is the strain hardening exponent of the material [20]. The mechanical properties of Alloy 600 are shown in Table 2 [21].

3.2. Geometric Model. To investigate the effect of the SSRT process of nickel-based Alloy 600, simulated SCC numerical tests with CT specimen were performed in this investigation. The numerical simulation process was guided by the American Society for Testing and Materials (ASTM) standards [19]. Since the shape and loading mode of the sample are both symmetrical, a one-half model of the CT sample is used in this simulation to reduce the amount of calculation, and the pre-crack length of the sample is 10 mm.

3.3. Finite Element Mesh. The loading process and the crack growth process were simulated by commercial FEM code ABAQUS, which is expected to represent the crack tip stress and strain conditions in the entire SSRT experimental process [22]. The mesh of the specimen is shown in Figure 3, and the mesh nearby the crack tip region of specimen is significantly refined to obtain a more detail and accurate data at the crack tip, where the X-axis is the opposite direction of the crack growth, and Y-axis is the normal direction of the crack growth in the coordinate system. The crack length extends from 24 mm to 27 mm by using node releasing technique during the simulating process.

4. Results and Discussions

4.1. SSRT Experiment Results. After 900 hours of slow strain rate stretching, the crack on the CT specimen was expanded from 24 mm to 27 mm by using DCPD, and the crack propagation length was about 3 mm. The relationship between the tensile load P and the load line displacement ΔL is shown in Figure 4.

Table 1: Chemical composition of Alloy 600 (%).

C	Mn	Fe	S	Si	Cu	Cr	P	Nb	Ti	Ni
0.065	0.22	9.50	0.0021	0.12	0.1	15.57	0.007	<0.05	0.28	Bal.

(a) (b)

Figure 1: Compact tension specimen: (a) specimen; (b) geometric dimensioning.

(a) (b)

Figure 2: The SSRT experiment: (a) the test procedure for Alloy 600; (b) the installation position of the CT specimen.

Table 2: Nickel-based Alloy 600 material mechanical properties at 340°C.

σ_0 (MPa)	E	α	n	E (GPa)	ν
436	0.239	3.075	6.495	189.5	0.286

4.2. Calculation Result and Analysis. After the numerical simulation is completed, the data of the load line displacement ΔL and the tensile load P at the load point are extracted, and the relationship between the tensile load P and the load line displacement ΔL is obtained as shown in Figure 5. Because the whole crack growing length during the test process is about 3 mm, the relation of the tensile load P and the load-line displacement ΔL are calculated when the crack extends every 0.375 mm, for which the crack length are 0 mm, 0.375 mm, 0.75 mm, 1.125 mm, . . ., 3 mm, respectively.

The above results are obtained by fixing the deformation path, which fixes the crack propagation length during the slow strain rate test (SSRT), and the relationship between the tensile load P and the load-line displacement ΔL of the fixed crack length is obtained. By incorporating the FEA results with SSRT result, we could get the intersection of the SSRT test curve and the numerical simulation curve under different crack lengths, thus the load-line displacement ΔL_c caused by the change in crack length is separated from the total line displacement obtained from the SSRT test, as shown in Figure 6.

In Figure 6, the black scatter curve and the colour curves are separated and used to represent SSRT result and the

Figure 3: Mesh model: (a) global mesh; (b) mesh refinement on crack tip.

Figure 4: The relationship between tensile load and load line displacement.

Figure 5: The relationship between tensile load and load line displacement.

numerical simulation results. The green lines are used to represent the intersection of numerical simulations and test results. By extracting the intersection of these curves, we can

Figure 6: The integration of the numerical simulation and SSRT.

get the relationship between the crack propagation length Δa and the load line displacement ΔL, as shown in Figure 7.

From the relation of the crack growth length Δa and the load-line displacement ΔL, we could estimate the crack growth rate da/dt by the load-line displacement change rate V and the relation of the crack growth length Δa with the load-line displacement ΔL during the test process, which is very useful for SCC experiments on the new environmental, material, and mechanical conditions. The SSRT time, Δa, and ΔL data are shown in Table 3.

The relationship between the crack growth rate and the crack growth length is shown in Figure 8.

The stress intensity factor K_I and J-integral J along the thickness of specimen could be worked out by using the ABAQUS software, as shown in Figures 9 and 10, which indicate that the K_I and J in the centre area of the specimen are bigger than two sides. At the same time, the difference of J between centre areas with two sides is bigger than K_I, which indicates that the crack growth will be more rapid in the central area than on the two sides.

The change laws of the stress intensity factor K_I and J-integral J with the crack growing in the centre area (plane

Figure 7: The relationship between crack growth length and load-line displacement.

Table 3: The SSRT experimental time, Δa, and ΔL data.

ΔL (mm)	Δa (mm)	T (h)	da/dt (mm/s)
0.36	0	72	—
1.1	0.375	220	$4.735E-07$
1.75	0.75	350	$5.952E-07$
2.32	1.125	464	$6.735E-07$
2.82	1.5	564	$7.388E-07$
3.3	1.875	660	$7.8918E-07$
3.72	2.25	744	$8.400E-07$
4.1	2.625	820	$8.892E-07$
4.485	3	897	$9.290E-07$

Figure 8: The relationship between crack growth rate and crack growth length.

Figure 9: The stress intensity factor K_I along the thickness of specimen.

Figure 10: The J-integral along the thickness of specimen.

strain condition) are shown in Figures 11 and 12, which indicate that both K_I and J will increase as the crack propagation because of both load and crack length increasing during the test process.

Finally, the relational curve of the SCC growth rate da/dt with stress intensity factor K_I and the relational curve of the SCC growth rate da/dt with J-integral J could be obtained during the whole test process, which provide a clear

Figure 11: The relationship between crack growth length and stress intensity factor.

Figure 12: The relationship between crack growth length and J-integral.

Figure 13: The relationship between stress intensity factor and crack growth rate.

Figure 14: The relationship between J-integral and crack growth rate.

understanding about the effect of the mechanical factors on SCC crack growth. It is also very convenient to analyze the effect of environmental and material factors on SCC crack growth rate and removing the mechanical factors. The relationship between K_I, J, and crack growth rate is shown in Figures 13 and 14, respectively.

5. Conclusions

(1) A quantitatively estimated approach of SCC growing rate in the slow strain rate test based on EPFEM is proposed, which can be used to estimate the actual crack growth rate during the SSRT test.

(2) During the crack propagation process, the K_I and J in the centre area of the specimen are bigger than those on the two sides. At the same time, the difference of J between centre areas with two sides is bigger than K_I, which indicates that the crack growth will be more rapid in the central area than on the two sides.

(3) When the crack propagation length is less than 3 mm, the K_I and J at the crack tip position gradually increase as the crack propagation length increases, and the increasing of J is much quicker than the increasing of K_I, which provides a clear understanding of the effect of the mechanical factors on EAC crack growth.

Data Availability

Previously reported Alloy 600 material chemical composition, cited as [18] within the text, and mechanical data, cited as [21] within the text, were used to support this study and are available at https://doi.org/10.1016/j.corsci.2015.05.026 and https://doi.org/10.1115/pvp2011-57317, respectively. The SSRT experiment data and the numerical simulation data of the SSRT experiment used to support the findings of this study are included in the article.

Conflicts of Interest

The authors declare that there are no conflicts of interest regarding the publication of this paper.

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