Research Article

Effect of PWHT on the Carbon Migration and Mechanical Properties of 2205DSS-Q235 LBW Joint

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Received 2 June 2018; Revised 7 September 2018; Accepted 6 November 2018; Published 2 December 2018

Guest Editor: Marek Smaga

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The effect of postweld heat treatment (PWHT) on the carbon migration and mechanical properties of the 2205DSS-Q235 laser beam welding (LBW) joint was investigated. The carbon-rich zone (CRZ) and carbon-depleted zone (CDZ) generated at the welding seam/Q235 (WS-Q235) interface as the carbon migration occurred after heat-treated at 600°C, 700°C, and 800°C for 1 h. The softening was found in the CDZ. Only the CRZ in joints heat-treated at 800°C was hardened because of the retaining of high-carbon untempered martensite. The thick CDZ in joints heat-treated at 700°C and 800°C contributed to the tensile fracture and the low elongation. The strength of the joint was roughly determined by the hardness of the fracture zone.

1. Introduction

Dissimilar steel welding is common in pressure vessel manufacturing. Dissimilar joints, such as martensitic steel/martensitic steel [1], martensitic steel/austenitic steel [2–4], and Ni-based superalloy/austenitic steel [5], have been widely investigated. Among these dissimilar joints, carbon migration from carbon steel to high-Cr weld metal (WM) was usually observed. The carbon migration was in uphill diffusion, as the carbon atom continuously diffusing from the low-carbon steel to the CRZ. Many investigations revealed that the uphill diffusion depended on the higher Cr concentration in the WM [6–8]. Cr atoms combine with the carbon atoms from carbon steel and generate $M_{23}C_6$ and M_7C_3 (M stands for Cr, Fe, and Mn) in the WM near the WM/carbon steel interface [6]. Therefore, the CRZ forms in the WM near the WM/carbon steel interface. As the loss of carbon atom near the interface occurs, the CDZ generates in the carbon steel near the WM/carbon steel interface. Mas et al. [6] indicated that the diffusion of carbon atom to high-Cr region was driven by the high chemical potential gradient across the interface.

The generation of CDZ in the dissimilar joint always contributed to the change in mechanical properties. Ming

et al. [9] employed Inconel 52M as the transition layer to join the 316L and SA508 steel. They found that the hardest area was in the 1st layer of 52Mb just adjacent to the SA508-52Mb interface due to the CRZ. The CDZ just adjacent to the fusion boundary had the lowest hardness. Sarikka et al. [10] found that the PWHT improved the carbon migration, thus resulting in the wider and softer CDZ compared to the as-welded state in the SA508-Alloy 52 interface. Wu et al. [11] observed that the CDZ in the 9%Cr fusion zone induced the fracture during the high-cycle fatigue test at 470°C. It can be concluded that the generation of CDZ contributed to the poor performance of the dissimilar joint.

Laser beam welding (LBW) provides outstanding characteristics of high energy density and high welding speed [12]. It contributes to the rapid joining of aluminum alloys [13, 14], magnesium alloys [15, 16], and steel [17, 18]. In our previous study, 2205DSS and Q235 steel with a thickness of 6.5 mm was joined using LBW [19]. In order to improve the toughness of the LBW joint, the PWHT process was conducted. After the PWHT process, CDZ and CRZ were generated as the carbon migration [20]. In this investigation, the mechanical properties of the joints with

different PWHT processes were investigated. Based on these results, the effect of PWHT on the carbon migration and mechanical properties were derived.

2. Materials and Experiments

The 2205DSS and Q235 plates with a thickness of 6.5 mm were joined using LBW without filler metal. The power of the laser beam was 3.7 kW. The defocusing distance was 0 mm. The welding speed was 1.2 m/min. The shielded gas was Ar gas with a flow rate of 25 L/min. The joints were heat-treated at 500°C, 600°C, 700°C, and 800°C for 1 h. The cross-sectional specimen of the weld joint perpendicular to the WD was prepared using an electrical discharge machine (EDM). The specimens were mechanically polished using waterproof SiC emery papers of up to 7000 grit and mirror polished using a colloidal Al_2O_3 (100 nm) suspension. The mirror-polished specimens were then etched using a solution consisting of HNO_3 (40 vol%) + $\text{C}_2\text{H}_5\text{OH}$ for ~5 s and subjected to the optical microscopy (OM, Zeiss Axio Scope A1) observation, scanning electron microscopy (SEM, MIRA3 LMH) observation, and electron probe microanalysis (EPMA, EPMA-1600). The tests of SEM and EPMA were under the secondary electron imaging (SEI) mode. For the electron backscatter diffraction (EBSD, MIRA3 LMH + Oxford) observation, the specimens were mechanically polished in a similar manner and then electropolished in a solution consisting of 10 vol.% perchloric acid and 90 vol.% ethanol at 20 V for ~10 s at room temperature. The step size of 1 μm was set for EBSD observation. The results from the EBSD were analyzed using Channel 5 software. The films with a diameter of 3 mm were prepared using a twin-jet electropolishing device and observed using a transmission electron microscope (TEM, JEM-2100).

The microhardness was measured using a microhardness tester with a load of 200 g and a dwell time of 5 s. The tensile test was conducted using a tensile machine with the help of fixture. The tensile speed was 100 $\mu\text{m}/\text{min}$.

The schematic of the joint is shown in Figure 1(a). The cross section of the joint is shown in Figure 1(b). A narrow WS (approximately 1 mm) was observed from Figure 1(b). The chemical compositions of the WS, Q235 BM, and 2205DSS BM are listed in Table 1. The interface of the WS/Q235 is shown in Figure 1(c). The structure of the tensile sample is shown in Figure 1(d). The tensile sample and fixture are shown in Figures 1(d) and 1(e). The fractured samples are shown in Figure 1(f).

3. Results and Discussion

3.1. Microstructure after PWHT

3.1.1. WS Evolution. The microstructure of WS is shown in Figure 2 [20]. The microstructure of the as-welded WS is shown in Figures 2(a) and 2(g). High dislocation density can be observed in Figure 2(g). Moreover, a small amount of retained austenite is observed in Figure 2(g). The microstructure of as-welded WS was mainly untempered martensite [19, 20].

The phase diagram of the WS (Figure 2(f)) was calculated using JMatPro software. According to the phase fraction of the WS, the Ac_1 temperature (621°C) and Ac_3 temperature (772°C) were derived. In this investigation, the PWHT temperatures of 500°C and 600°C were lower than the Ac_1 temperature. The untempered martensite would become tempered martensite. The microstructures of the WS heat-treated at 500°C and 600°C are shown in Figures 2(b) and 2(c). The martensitic lath was clearer. The microstructure of the WS heat-treated at 600°C is revealed in Figure 2(h). It could be observed that the $(\text{Cr,Fe})_{23}\text{C}_6$ particles precipitated in the martensitic lath boundary. The high dislocation density disappeared. The subgrain was observed. Moreover, the dislocation density declined.

When the PWHT temperature was increased to 700°C, part of martensite transformed into the γ phase. After cooled to the room temperature, the γ phase became the untempered martensite again. The martensite, which did not transformed into the γ phase, became the deeply tempered martensite during the PWHT. Therefore, the microstructure of the WS heat-treated at 700°C was consisted of untempered martensite and tempered martensite.

When the PWHT temperature was increased to 800°C, the martensite totally transformed into the γ phase which can be observed from Figure 2(f). After cooled to the room temperature, the γ phase became the untempered martensite again. From the microstructure in Figure 2(e), the martensitic lath could not be clearly observed. It meant that carbide did not precipitate. The TEM image of the WS heat-treated at 800°C is revealed in Figure 2(i). The $(\text{Cr,Fe})_{23}\text{C}_6$ particles on the grain boundaries disappeared. The martensitic lath with a high dislocation density was observed again.

From the above results and discussion, the untempered martensite transformed into the tempered martensite after heat-treated at 500°C and 600°C. The WS heat-treated at 700°C consisted of tempered martensite and part of untempered martensite. The WS heat-treated at 800°C was total untempered martensite.

3.1.2. Interface Evolution. The WS-Q235 interfaces after heat-treated are shown in Figure 3 [20]. From Figure 3(a), the carbon migration was not observed in the WS-Q235 interface heat-treated at 500°C. When the PWHT temperature was increased to 600°C, CRZ and CDZ were observed. When the PWHT temperature was raised to 700°C, CRZ and thick CDZ were observed (Figures 3(c) and 3(f)). However, when the PWHT temperature was increased to 800°C, only the CDZ was observed (Figures 3(d) and 3(e)). Moreover, the thickness of the CDZ heat-treated at 800°C was smaller than that heat-treated at 700°C.

In order to observe the CRZ in the WS-Q235 interface heat-treated at 800°C, the cross section was etched by $\text{FeCl}_3 + \text{HCl}$ solution again. The optical image of the WS-Q235 interface is shown in Figure 3(g). The CRZ was distinguished.

The formation of the CDZ and CRZ was attributed to the Cr in the WS [6–8, 21]. During the PWHT process of

FIGURE 1: (a) Schematic of the joint; (b) cross section of the joint; (c) microstructure of the WS-Q235 interface; (d) structure of the tensile sample; (e) fixture of the tensile sample; (f) fractured tensile samples.

TABLE 1: Chemical compositions of WS, Q235 BM, and 2205DSS BM (wt.%).

Materials	C	Si	Mn	Ni	Cr	Mo	N	S	P	Fe
WS	0.107	0.350	0.823	2.067	8.800	1.186	0.055	0.013	0.03	Bal.
Q235	0.16	0.21	0.55	—	—	—	—	0.020	0.030	Bal.
2205DSS	0.025	0.570	1.250	5.300	22.570	3.040	0.14	0.002	0.030	Bal.

2205DSS-Q235 joint, the carbon atom continuously migrated from the Q235 BM to the WS. It combined with the Cr and Fe atoms and contributed to the generation of the $(Cr,Fe)_{23}C_6$ carbide in WS. Therefore, CDZ and CRZ generated in Q235 BM and WS, respectively. The thickness of CDZ and CRZ as a function of PWHT temperature is revealed in Figure 3(h). The CDZ thickened as the increase of temperature when the PWHT temperature was below 700°C. However, when the PWHT temperature was raised to 800°C, the thickness of the CDZ declined to approximately 130 μm . The rapid thickening of the CRZ is also marked in Figure 3(h).

3.1.3. Carbon-Depleted Zone Evolution. The A_{c1} temperature of Q235 BM was approximately 710°C as shown in Figure 4. Therefore, when the PWHT temperature was 500°C, 600°C, and 700°C, the phase in the CDZ was mainly the α -Fe phase. The generation of CDZ mainly relied on the migration of carbon atom from Q235 to WS. Therefore, the thickness of CDZ depended on the diffusion rate of carbon atom. When the temperature increased (below 710°C), the carbon migration gradually accelerated. Therefore, the thickness of the CDZ increased as the increase of PWHT temperature.

When the temperature was 800°C, the CDZ consisted of α -Fe and γ -Fe as shown in Figure 4. At the early stage of the PWHT at 800°C, the carbon concentration of Q235 near the WS/Q235 interface was 0.16 wt.%. The volume fraction of

γ -Fe was 54%. However, when the PWHT time extended to 1 h, the carbon concentration of CDZ was only approximately 0.087 wt.% (from EPMA result). According to Figure 4, the volume fraction of γ -Fe was 31% after heat-treated for 1 h. Huang et al. [21] indicated that the γ -Fe in the CDZ slowed the migration of carbon atom. Therefore, the CDZ in the joint heat-treated at 800°C was thinner than that heat-treated at 700°C.

The grain orientation (in the X direction) of WS/Q235 interface heat-treated at 700°C and 800°C is shown in Figure 5. The columnar α -Fe grain with a straight grain boundary was observed in the CDZ heat-treated at 700°C. However, the part of grain boundaries of the α -Fe grain in the CDZ heat-treated at 800°C was not straight anymore as indicated in Figure 5(b). It should be attributed to the transformation from γ -Fe to α -Fe when the joint cooled from 800°C to room temperature. The grain boundary of γ -Fe and α -Fe at 800°C should also be the straight. When the temperature declined, the α -Fe grains would nucleate in the grain boundary. The prior austenitic grain was partitioned by the new α -Fe grains. Therefore, the zigzag grain boundaries generated after heat-treated at 800°C.

The difference in the Fe_3C particle after heat-treated at 700°C and 800°C should also be discussed in this investigation. The Fe_3C particle is indicated in Figures 5(c) and 5(d). It can be observed from Figures 5(c) and 5(d) that the Fe_3C particle in the CDZ heat-treated at 700°C was less than that heat-treated at 800°C. The α -Fe phase in the CDZ heat-treated at 700°C exhibited a smaller solubility for carbon

FIGURE 2: Microstructure and phase diagram of WS: (a, g) as-welded; (b) heat-treated at 500°C, (c, h) heat-treated at 600°C, (d) heat-treated at 700°C, (e, i) heat-treated at 800°C, (f) phase diagram.

atoms when compared with that heat-treated at 800°C. Moreover, the γ -Fe phase in the CDZ at 800°C exhibited a larger solubility for carbon atoms when compared with the α -Fe phase at 700°C. Therefore, more carbon atoms were dissolved in the CDZ at 800°C. After cooled to room temperature from 800°C, more Fe_3C particles precipitated from the CDZ.

3.1.4. Carbon-Rich Zone Evolution. The Ac_1 and Ac_3 temperatures of the WS as a function of carbon concentration are shown in Figure 6(a). From Figure 6(a), the microstructure of the WS at 700°C consisted of α -Fe and γ -Fe. The volume fraction of the γ -Fe in the as-welded WS was 21%. When heat-treated at 700°C for 1 h, the carbon concentration of the CRZ was approximately 0.47 wt.% (from EPMA). The volume fraction of the γ -Fe was 17% from Figure 6(a). The matrix of WS became pure γ -Fe at 800°C as shown in Figure 6(a).

$(\text{Cr,Fe})_{23}\text{C}_6$ carbide would generate at 700°C and 800°C. Its volume fraction is shown in Figure 6(b). The carbon

concentration of the CRZ in the joint heat-treated at 800°C was 0.22 wt.% (from EPMA). The volume fraction of the $(\text{Cr,Fe})_{23}\text{C}_6$ carbide at 800°C was only 3.3%. However, the volume fraction of $(\text{Cr,Fe})_{23}\text{C}_6$ carbide at 700°C reached 8.9%.

From the carbon concentration of the CRZ in the joint heat-treated at 700°C and 800°C, the wider CRZ with a lower carbon concentration was found in the joint heat-treated at 800°C. The formation of the wider CRZ relied on the long-distance diffusion of carbon atom. At 800°C, the grain boundary contained less $(\text{Cr,Fe})_{23}\text{C}_6$ carbide. The diffusion of the carbon atom along the grain boundary at 800°C was easier than that at 700°C as the weaker hindering effect of $(\text{Cr,Fe})_{23}\text{C}_6$. Therefore, the wider CRZ was generated at 800°C.

The CRZ in the joint heat-treated at 600°C consisted of tempered martensite and $(\text{Cr,Fe})_{23}\text{C}_6$. $(\text{Cr,Fe})_{23}\text{C}_6$ came from two sources. The first was the carbon from the WS. The second was the carbon from the Q235 BM. The CRZ in the joint heat-treated at 700°C consisted of tempered martensite, untempered martensite, and $(\text{Cr,Fe})_{23}\text{C}_6$. The

FIGURE 3: Microstructure of WS-Q235 interface and thickness of CDZ and CRZ: (a) heat-treated at 500°C; (b) heat-treated at 600°C; (c, f) heat-treated at 700°C; (d, e, g) heat-treated at 800°C; (h) thickness of CDZ and CRZ as a function of PWHT temperature.

FIGURE 4: Ac₁ temperature; Ac₃ temperature; γ-Fe phase fraction of Q235 as a function of the carbon concentration.

FIGURE 5: Microstructure of the CDZ; (a) grain orientation of the CDZ heat-treated at 700°C; (b) grain orientation of the CDZ heat-treated at 800°C; (c) CDZ heat-treated at 700°C; (d) CDZ heat-treated at 800°C.

FIGURE 6: (a) Ac_1 , Ac_3 , and γ -Fe fraction as a function of carbon concentration; (b) $M_{23}C_6$ concentration as a function of carbon concentration.

formation of $(\text{Cr,Fe})_{23}\text{C}_6$ was similar to that at 600°C . The untempered martensite came from the $\gamma\text{-Fe}$ at 800°C . When the joints heat-treated at 600°C and 700°C were etched by $\text{HNO}_3(4\text{ wt.}\%) + \text{ethanol}$ solution, the $(\text{Cr,Fe})_{23}\text{C}_6$ carbide was etched easily. Therefore, the CRZ could be observed clearly in Figures 3(b)–3(f).

The CRZ in the joint heat-treated at 800°C was of untempered martensite and small amount of $(\text{Cr,Fe})_{23}\text{C}_6$ carbide. Therefore, it was hard to reveal the microstructure using $\text{HNO}_3(4\text{ wt.}\%) + \text{ethanol}$ as shown in Figures 3(e) and 3(f).

3.2. Mechanical Properties after PWHT

3.2.1. Hardness. The hardness distribution of the joints is shown in Figure 7(a). The hardness change of the 2205DSS and Q235 was small. In this investigation, the hardness evolution of 2205DSS BM and Q235 BM was not discussed as their small influence on the fracture behavior. The hardness evolution of WS, CRZ, and CDZ was discussed in detail. From Figures 7(a) and 7(b), the hardness of the as-welded WS was approximately 550 HV. The hardness of the WS declined slightly after heat-treated at 500°C . When the WS was heat-treated at 600°C and 700°C , the hardness was approximately 350 HV. When the PWHT temperature was increased to 800°C , the hardness increased to approximately 450 HV. Moreover, the hardness of the WS near the WS/Q235 interface significantly increased (approximately 570 HV) as shown in Figure 7.

In this investigation, the hardness decline of WS at 500°C , 600°C , and 700°C should be attributed to the formation of tempered martensite. The hardness increase at 800°C should be attributed to the regain of untempered martensite.

The hardness distribution near the WS/Q235 interface is shown in Figure 7(b). The high-hardness CRZ in the joint heat-treated at 800°C should be attributed to the high-carbon martensite which had a high hardness as discussed in Section 3.1.4. However, the hardness increment of the CRZ in the joints heat-treated at 500°C , 600°C , and 700°C was not observed in Figure 7(b). Although $(\text{Cr, Fe})_{23}\text{C}_6$ carbide or small amount of untempered martensite generated in the CRZ in the joints heat-treated at 500°C , 600°C , and 700°C , its strengthening effect could not balance the hardness reduction of the tempered martensite at all. Therefore, the hardness of the CRZ did not increase after heat-treated at 500°C , 600°C , and 700°C .

The hardness of the HAZ heat-treated at 500°C declined slightly compared with that of the HAZ of the as-welded joint as shown in Figure 7(b). When the PWHT temperature was increased to 600°C , the hardness of the CDZ was approximately 150 HV. The lowest hardness of the CDZ was only approximately 110 HV after heat-treated at 700°C . When heat-treated at 800°C , the lowest hardness of the CDZ was improved to approximately 135 HV.

The low hardness of the CDZ in the joint heat-treated at 700°C and 800°C should be contributed to the less Fe_3C particle and the coarse ferrite grain. The difference in CDZ hardness between the two joints should also be discussed. From the discussion about the Fe_3C in Section 3.1.3, the

FIGURE 7: Hardness distribution of the as-welded and heat-treated 2205DSS-Q235 joints: (a) the whole joint; (b) the WS-Q235 interface.

fraction of Fe_3C in the CDZ heat-treated at 700°C was smaller than that heat-treated at 800°C . Moreover, the grain sizes of the CDZ in the joints heat-treated at 700°C and 800°C were $\sim 32\text{ }\mu\text{m}$ and $\sim 22\text{ }\mu\text{m}$ from the EBSD results. Therefore, the coarse grain and less Fe_3C of CDZ in the joint heat-treated at 700°C contributed to its lowest hardness.

3.2.2. Strength, Elongation, and Fracture Behavior. The stress-strain curves are shown in Figure 8(a). The

FIGURE 8: (a) Stress-strain curves; (b) strength vs. elongation; (c) strength vs. hardness.

strength as a function of elongation is shown in Figure 8(b). It was found that all of the strength and elongation gradually declined as the increase of PWHT temperature when the PWHT temperature was below 700°C. However, when the PWHT temperature was 800°C, the strength increased.

The microstructure of the cross section of fractured joints is shown in Figure 9. From Figures 9(a) and 9(b), the fracture location of the joints without PWHT and heat-treated at 600°C was in the Q235 BM. The fracture of the joints heat-treated at 700°C and 800°C was located in the CDZ as shown in Figures 9(c)–9(h). In this paper, the relationship between hardness of the fractured zone and strength of the joints was investigated as shown in Figure 8(c). It could be found that the strength of the joint was mainly determined by the hardness of the fracture zone.

From Figures 9(c) and 9(d), the CDZ in the joints heat-treated at 700°C and 800°C induced the fracture during tensile. However, the CDZ in the joint heat-treated at 600°C could not result in the fracture as shown in Figure 9(b). It should be attributed to the small thickness of the CDZ in the joint heat-treated at 600°C. During tensile, the strength of the WS and the Q235 near the CDZ in the joint heat-treated at 600°C was higher than the thin CDZ. The deformation of the CDZ was restricted. Therefore, the necking did not occur in the CDZ. On the contrary, the Q235 BM was stretched. When the true stress of the Q235 BM increased to a critical value, the necking occurred in the Q235 BM. Therefore, the fracture was located in the Q235 BM.

The plastic deformation of the Q235 BM contributed to the higher elongation of the joints without PWHT and heat-treated at 600°C. With regard to the joints heat-treated at 700°C and 800°C, the necking and fracture rapidly

FIGURE 9: (a) Fractured as-welded joint; (b) fractured joint heat-treated at 600°C; (c, e, g) fractured joint heat-treated at 700°C; (d, f, h) fractured joint heat-treated at 800°C.

generated in the CDZ during tensile. The deformation of the Q235 BM was small. Therefore, the elongation of the joints heat-treated at 700°C and 800°C was small as shown in Figures 8(a) and 8(b).

4. Conclusions

- (1) Tempered martensite was obtained in the WS when heat-treated at 500°C and 600°C, as the generation of precipitates and decline of dislocation density. Untempered martensite was retained in the WS when heat-treated at 800°C which was higher than the A_{c3} temperature.

- (2) The CDZ and CRZ were generated on the WS/Q235 interface after heat-treated at 600°C, 700°C, and 800°C, as the carbon atoms diffused from Q235 BM to WS. The CDZ of the joint heat-treated at 800°C was thinner than that heat-treated at 700°C because of the generation of γ -Fe in Q235 BM during heat-treatment.
- (3) The hardness of the WS declined to ~ 350 HV after heat-treated at 600°C and 700°C as the tempered martensite formed. The hardness of the WS heat-treated at 800°C was ~ 450 HV as the untempered martensite was retained.
- (4) The softening was found in the CDZ. Only the CRZ of the joint heat-treated at 800°C was hardened as

high-carbon untempered martensite was generated which had a hardness of ~570 HV.

- (4) The strength of the joint was roughly determined by the hardness of the fracture zone. The CDZ in joints heat-treated at 700°C and 800°C contributed to the tensile fracture of the joints. The joint fractured in the CDZ exhibited the lower elongation.

Data Availability

All the data in the article came from our experiments.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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