

DEVELOPMENT OF A {111} RECRYSTALLIZATION TEXTURE ASSOCIATED WITH DYNAMIC STRAIN AGING DURING HOT ROLLING IN THE FERRITE REGION

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Hot rolling in the ferrite region is applied in an extra-low C sheet steel without cold rolling by employing dynamic strain aging. When the amount of solute C is about 10 ppm before rolling, the r -value and {222} intensity ratio of sheet steel annealed after rolling are maxima at the rolling temperature of 773K, during which dynamic strain aging occurs. The {222} residual strain in the specimen rolled at 773K is higher than that in specimens rolled at other temperatures. It is proposed that dynamic strain aging would provide high stored energy in the {111} component of an as-rolled specimen, with the result that the region of high stored energy would recover and nucleate rapidly so that a strong {111} recrystallization texture develops.

KEY WORDS: Hot rolling, dynamic strain aging, recrystallization, texture, solute carbon, extra-low carbon sheet steel.

1. INTRODUCTION

Deep-drawable sheet steels have mainly been produced by conventional cold rolling and subsequent annealing processes to form the {111} recrystallization texture. In such cases, hot rolling is performed only in the austenite region, leading to a random orientation. A few studies have been made on the recrystallization texture of sheet steels hot-rolled in the ferrite region in order to obtain a high Lankford value (r -value) without cold rolling (e.g. Hayashi and Okamoto, 1978; Senuma *et al.*, 1987; Hashimoto and Yakushiji, 1988; Matsuoka *et al.*, 1988). Senuma *et al.* and Hashimoto and Yakushiji reported that a strong {111} recrystallization texture was developed by hot rolling in the ferrite region when little solute C and/or N existed in the parent steels before hot rolling. These authors suggested that the recrystallization texture of sheet steels hot-rolled in the ferrite region depended on the rolling strain rate, and that a strong {111} texture would be obtained by introducing high strain-rate rolling.

It is well known that the occurrence of the blue brittleness at around 573K results in an abrupt decrease of ductility. This phenomenon is caused by dynamic strain aging, which is due to the interaction between dislocations and interstitial atoms of C and/or N. Since this promotes work-hardening in the deformed material and leads to a refined

substructure (Li and Leslie, 1978), it can be expected that the recrystallization texture after hot rolling in the ferrite region would change significantly with dynamic strain aging according to the oriented nucleation theory (Burgers and Louwse, 1931) or high-energy block theory (Cahn, 1950) for recrystallization. Ushioda and Abe (1984) reported that a $\{110\} \langle 001 \rangle$ recrystallization texture was preferentially formed when Fe-0.01N (mass %) alloys were rolled at 573K, at which dynamic strain aging occurred, and were then annealed. However, the effect of solute C content on the formation of the $\{111\}$ recrystallization texture that is associated with dynamic strain aging has not been reported.

In the present work, the recrystallization texture and associated r -value were investigated as functions of the solute C content and rolling temperature in the ferrite region for materials with three different solute C contents. As a result, it was demonstrated that a strong $\{111\}$ recrystallization texture and high r -value could be obtained without cold rolling, especially by applying dynamic strain aging.

2. EXPERIMENTAL PROCEDURES

The steels used in this study were three kinds of commercially produced hot bands with different solute C contents, the chemical compositions being shown in Table 1. It was confirmed by a chemical analysis that almost all N could be stabilized as TiN in the steel A and as AlN in steels B and C. The solute C content just prior to rolling in the ferrite region was estimated by means of the aging index (AI), which is related to the solute C content (Konishi *et al.*, 1979). As shown in Figure 1, aging indices were measured for specimens that had been soaked at a certain temperature for 10 minutes and then water-quenched without rolling, leading to an associated solute C content. For steel C, the amount of solute C exceeded 30 ppm as shown in Figure 1 was estimated, although the result in the temperature range of 673K to 973K is larger than a previous result of 22 ppm (Stephenson, 1962). Further microstructural changes cannot be observed through the soaking treatment.

One-pass rolling was conducted at different temperatures in the range of room temperature to 973K in the ferrite region, the rolling strain rate being widely varied from 45s^{-1} to 1000s^{-1} with a reduction of 55%. The rolled specimens were water-quenched within 0.2s after rolling. A lubricant was adopted in this process to obtain a homogeneous microstructure in the thickness direction (Matsuoka *et al.*, 1986). The friction coefficient between a roll and the material deformed under the foregoing conditions was calculated to be as low as 0.1 with the rolling load used, the radii of the rolls being 125mm and 355mm. The average rolling strain rate was then calculated from the following equation (Shida, 1966):

Table 1 Chemical composition of the sheet steels (mass %)

	<i>C</i>	<i>Si</i>	<i>Mn</i>	<i>P</i>	<i>S</i>	<i>Al</i>	<i>N</i>	
A	0.0020	0.012	0.07	0.013	0.008	0.069	0.0014	Ti0.036
B	0.0034	0.020	0.15	0.013	0.006	0.040	0.0019	Nb0.015
C	0.043	0.015	0.30	0.014	0.008	0.041	0.0040	

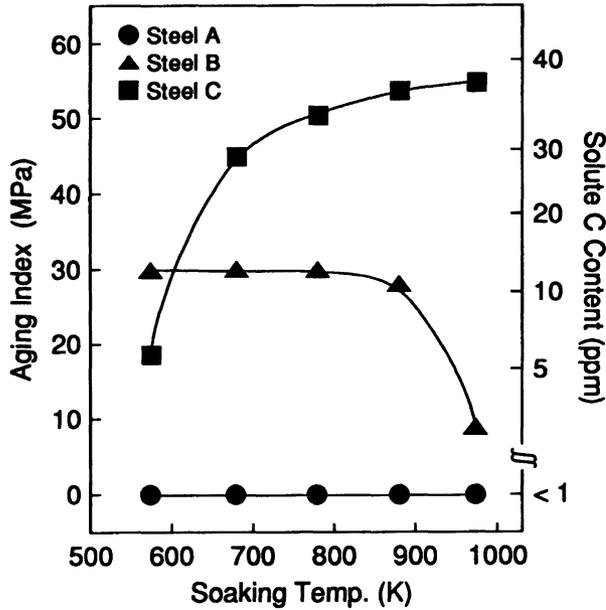


Figure 1 Effect of soaking temperature on the aging index (AI) and solute C content of specimens water-quenched from soaking temperature.

$$\dot{\epsilon} = \frac{2\pi n}{60\sqrt{r}} \sqrt{\frac{R}{H}} \ln \left[\frac{1}{1-r} \right]$$

- where n = roll speed (rpm)
- H_0 = thickness before rolling (mm)
- r = rolling reduction (%/100)
- R = roll radius (mm)

Recrystallization annealing was performed on the rolled materials with an infrared heating device, the specimens being held at 1103K for 40s with a heating rate of 10K/s and then air-cooled. Hardness was measured by a Vickers hardness tester with a 5kg load on the quarter plane of the specimen.

X-ray measurements, using the Mo characteristic ray, were performed on the quarter plane of the specimens. The X-ray diffraction intensity ratio and its line broadening were measured for the main components of {222}, {211}, {200} and {110}. Line broadening was obtained by separating the $K_{\alpha 1}$ peak from the diffraction line profile.

The Lankford value (r-value) was calculated by measuring the width change of the specimen after stretching by 15%. In this study, the mean r-value defined by the following equation was adopted:

$$r = (r_L + 2r_D + r_T) / 4$$

where r_L , r_D and r_T denote the values measured for specimens cut in the longitudinal, diagonal and transverse directions, respectively.

3. EXPERIMENTAL RESULTS

3.1 Effects of Chemical Composition and Rolling Temperature on the r-value

The effects of chemical composition and rolling temperature on the r-value of annealed specimens is shown in Figure 2. The r-values for steel A, in which the amount of solute C was less than 1 ppm, progressively increase with decreasing rolling temperature. For steel B, the r-values also increase with decreasing rolling temperature, except for the data at 773K. The r-value shows its significant peak at 773K, corresponding to about a 10ppm solute C content. The r-values for steel C, in which the amount of solute C exceeded 30 ppm at about 773K, increase with decreasing rolling temperature, except for the rapid decrease at a rolling temperature of 773K.

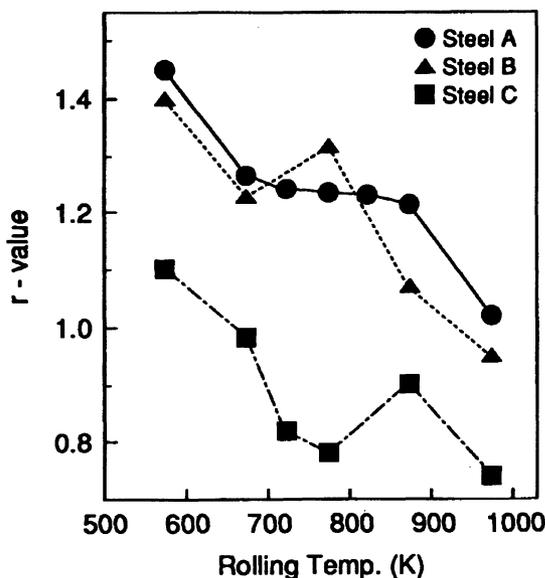


Figure 2 Effects of chemical composition and rolling temperature on the r-values of specimens annealed at 1103K for 40s after rolling with a strain rate of 1000s^{-1} .

The r-values and diffraction intensity ratios for annealed specimens rolled at 773K, at which the r-value changed in this characteristic way, are shown in Figure 3. The r-value and {222} intensity ratio slightly increase with increasing solute C content from less than 1 ppm to 10 ppm. However, a solute C content exceeding 30 ppm leads to an abrupt decrease in r-value and {222} intensity ratio. The {110} intensity ratio rapidly increases with increasing solute C content from 10 ppm to over 30 ppm, while the {211} and {200} intensity ratios are independent of the solute C content.

As already mentioned, the r-value and recrystallization texture were significantly affected by the solute C content existing prior to rolling. This was especially true when rolling was performed at about 773K, about 10 ppm of solute C (steel B) leading to a maximum r-value, and solute C above 30 ppm (steel C) leading to a minimum

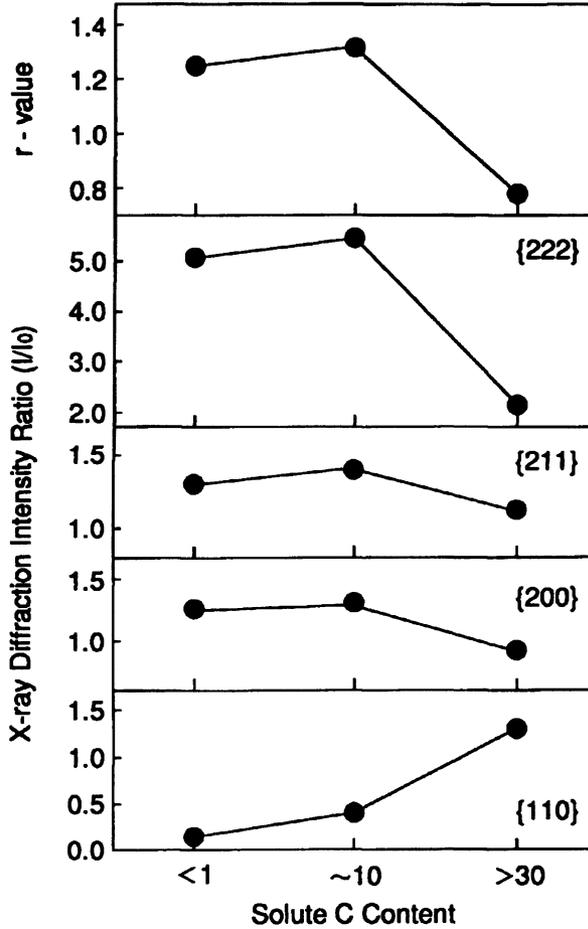


Figure 3 Effect of solute C content on the r-value and diffraction intensity ratio of specimens annealed at 1103K for 40s after rolling at 773K with a strain rate of 1000s⁻¹. Solute C contents of less than 1 ppm, about 10 ppm and over 30 ppm correspond to steels A, B and C, respectively.

r-value. In order to clarify the effect of solute C content on the r-value and recrystallization texture, the texture formation of steels B and C, which showed the characteristic phenomenon at a rolling temperature of about 773K, were investigated in detail.

3.2 Recrystallization Texture of the Sheet Steel C with a Solute C content over 30 ppm

Figure 4 shows optical micrographs of as-rolled specimens in which the amount of solute C exceeded 30 ppm at about 773K. A uniformly deformed microstructure can be seen in the specimens rolled at 973K and 573K. On the other hand, deformation bands at an angle of 45° to the rolling direction at regular intervals can be seen in

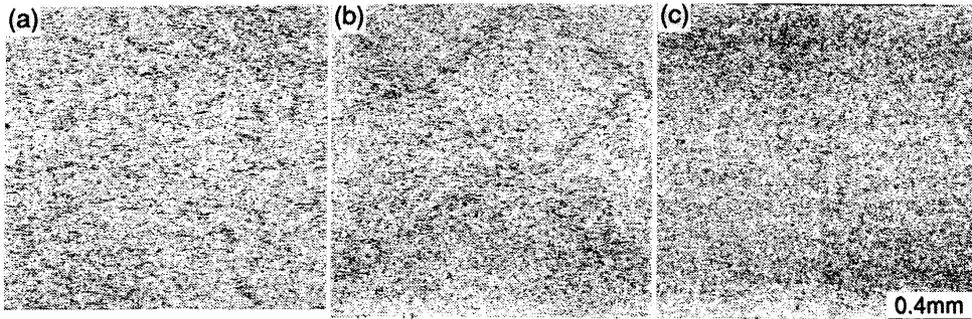


Figure 4 Optical micrographs showing the effect of rolling temperature on the cross-sectional microstructure in the longitudinal direction of specimens rolled at (a) 573K, (b) 773K and (c) 973K with a strain rate of 1000s^{-1} , in which the solute C content exceeded 30 ppm (Steel C).

the specimen rolled at 773K. It would be induced from the non-uniform deformation caused by dynamic strain aging.

The effect of rolling temperature on the diffraction intensity ratio of as-rolled specimens containing solute C of over 30 ppm at around 773K is shown in Figure 5. The $\{222\}$ and $\{200\}$ intensity ratios abruptly decrease with decreasing rolling

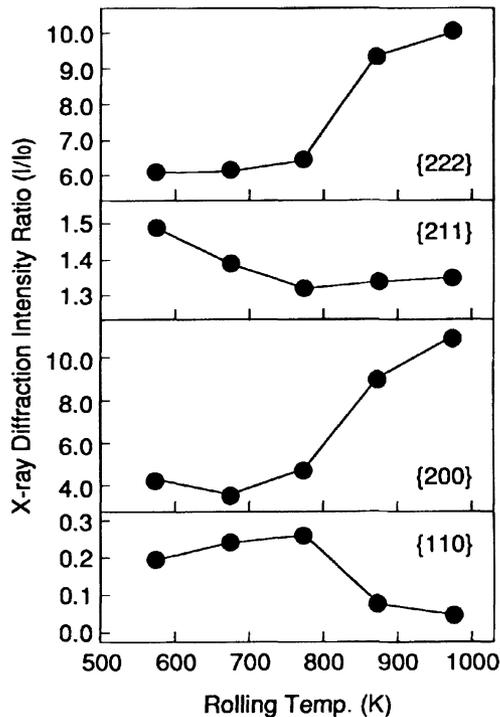


Figure 5 Effect of rolling temperature on the diffraction intensity ratio of specimens rolled with a strain rate of 1000s^{-1} , in which the solute C content exceeded 30 ppm (Steel C).

temperature from 873K to 773K, and are almost constant below 773K. On the other hand, the {110} intensity ratio increases with decreasing rolling temperature from 873K to 773K and remains high below 773K.

Figure 6 shows the effect of rolling temperature on the diffraction intensity ratio of the specimens annealed at 1103K for 40s after rolling. The {222} intensity ratio increases with decreasing rolling temperature, except for the minimum value at a rolling temperature of 773K. On the other hand, the {110} intensity ratio shows a maximum value at the rolling temperature of 773K.

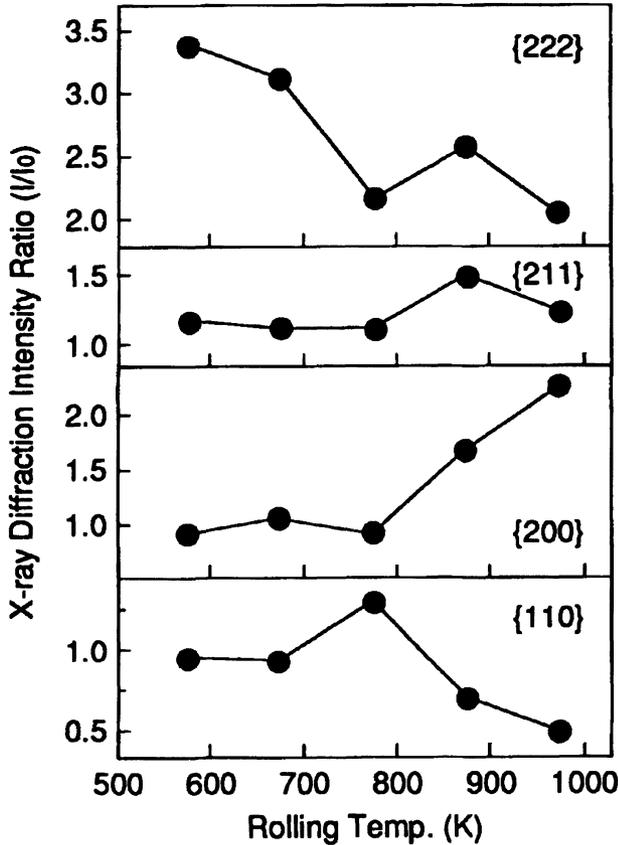


Figure 6 Effect of rolling temperature on the diffraction intensity ratio of specimens annealed at 1103K for 40s after rolling with a strain rate of $1000s^{-1}$, in which the solute C content exceeded 30 ppm (Steel C).

As previously mentioned, the steel containing solute C of over 30 ppm before rolling exhibits an abrupt decrease of r-value and the evolution of an abnormal recrystallization texture is caused by dynamic strain aging, these results being in good agreement with those of Ushioda and Abe (1984).

3.3 Recrystallization Texture of the Sheet Steel B with a Solute C Content about 10 ppm

Figure 7 shows the effect of rolling temperature on the diffraction intensity ratio of as-rolled specimens B containing 10 ppm of solute C at about 773K. The {222} and {200} intensity ratios decrease progressively with almost constant below 773K. These diffraction intensity changes are similar to those for steel C shown in Figure 5. On the other hand, the {211} and {110} intensity ratios are independent of the rolling temperature.

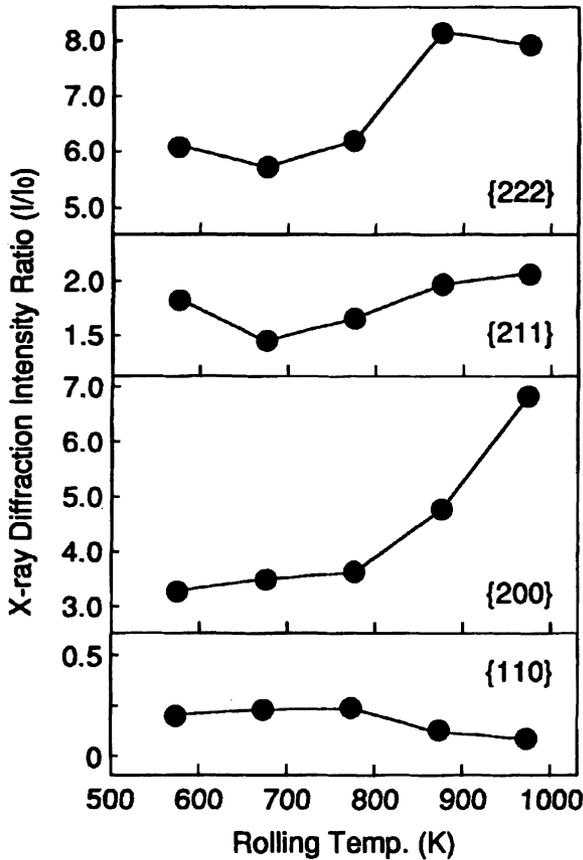


Figure 7 Effect of rolling temperature on the diffraction intensity ratio of specimens rolled with a strain rate of 1000s^{-1} , in which the solute C content was about 10 ppm (Steel B).

Figure 8 shows the effect of rolling temperature on the diffraction intensity ratios of the specimens annealed at 1030K for 40s after rolling. the {222} intensity ratio shows a maximum at the rolling temperature of 773K. While the {110} intensity ratio also shows a maximum at the rolling temperature of 773K, this is lower than that of the specimen containing over 30 ppm of solute C shown in Figure 6. This result is in good agreement with the r -value change for the annealed specimens shown in Figure 2.

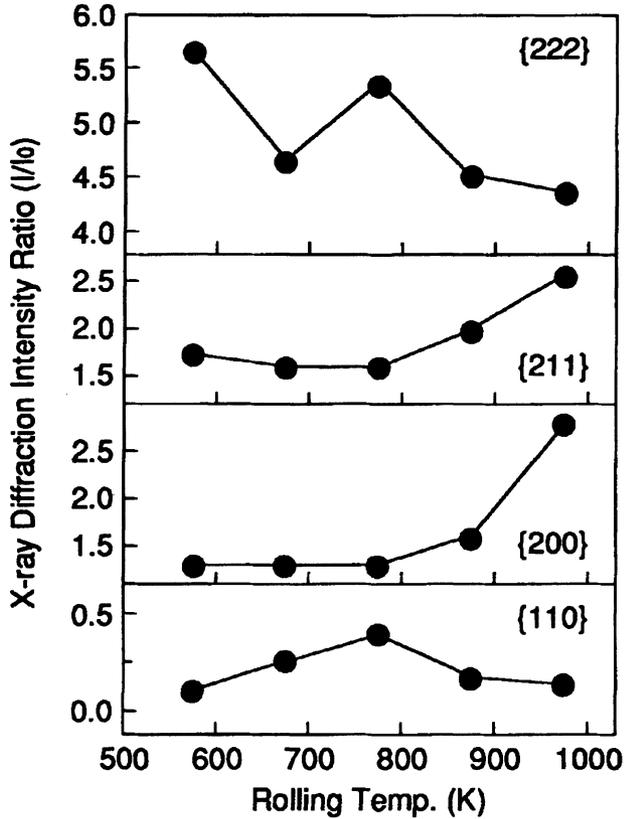


Figure 8 Effect of rolling temperature on the diffraction intensity ratio of specimens annealed at 1103K for 40s after rolling with a strain rate of $1000s^{-1}$, in which the solute C content was about 10 ppm (Steel B).

The effect of rolling temperature on the X-ray line broadening of the as-rolled specimen containing about 10 ppm of solute C before rolling is shown in Figure 9. The X-ray line broadenings of the {222}, {211}, {200} and {110} components tend to increase with decreasing rolling temperature. Particularly, those of the {222} and {200} components show maxima at the rolling temperature of 773K.

In order to confirm whether the specific phenomenon at the rolling temperature of around 773K was caused by dynamic strain aging, the hardness of as-rolled specimens was investigated with various rolling temperatures and rolling strain rates. Figure 10 shows the effects of rolling temperature and rolling strain rate on the hardness of as-rolled steel B. The hardness shows maxima at rolling temperatures from 673K to 773K, depending on the rolling strain rate. The deformation resistance calculated in terms of the rolling load also showed peaks corresponding to the hardness, the peak temperature for maximum hardness increasing with increasing rolling strain rate. As a result, it is concluded that the anomalous phenomenon at the rolling temperature of about 773K was caused by dynamic strain aging when solute C existed.

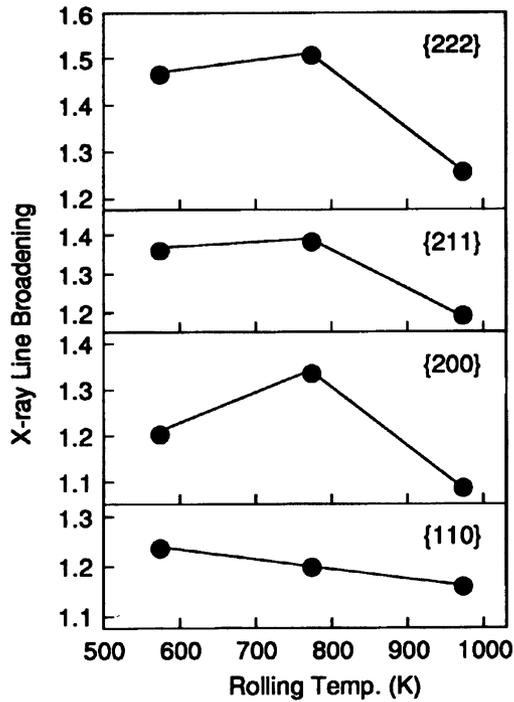


Figure 9 Effect of rolling temperature on the X-ray line broadening of specimens rolled with a strain rate of 1000s^{-1} , in which the solute C content was about 10 ppm (Steel B).

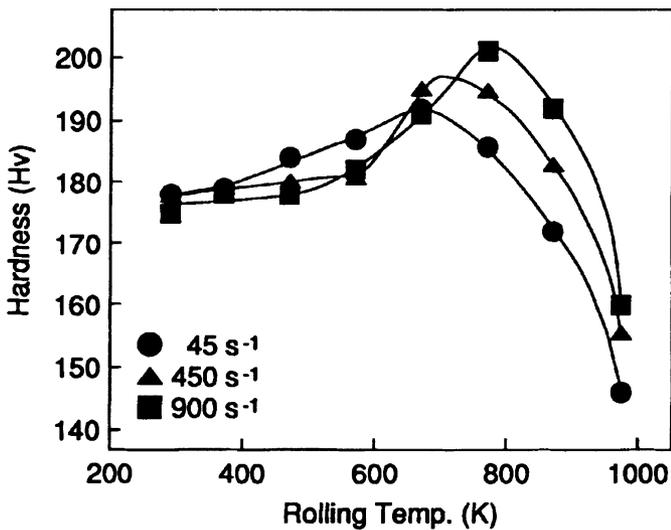


Figure 10 Effects of rolling temperature and rolling strain rate on the hardness in Steel B rolled in a temperature range of 293K to 973K with a rolling strain rate of 45s^{-1} to 900s^{-1} , in which the solute C content was about 10 ppm.

4. DISCUSSION

It is well known that solute C influences the formation of a recrystallization texture in cold-rolled sheet steel, and that solute C before cold rolling and annealing leads to less development of the {111} recrystallization texture. Since there can be a high proportion of solute C during hot rolling in the ferrite region as shown in Figure 1, it is considered that the deformation behavior is very much affected by the amount of solute C.

Several investigations into the effect of solute C on the formation of a recrystallization texture in sheet steel rolled in the ferrite region have been described in the literature. Senuma *et al.* (1987, 1988, 1989) and Hashimoto and Yakushiji (1988) reported that a strong {111} recrystallization texture and resultant high r -value could be obtained by hot rolling in the ferrite region without solute C before rolling. As shown in Figure 2, the r -value for specimen A, in which the amount of solute C was less than 1 ppm, is highest in comparison with the values for steels B and C. This is in good agreement with the results reported so far.

However, it should be noted that the r -values for steels B and C, in which the respective amounts of solute C were about 10 ppm and more than 30 ppm, had converse behavior at the rolling temperature of 773K at which dynamic strain aging occurred. In these cases, the r -value and {222} diffraction intensity ratio for steel B containing about 10 ppm of solute C were maxima, the resultant r -value being higher than that for steel A with less than 1 ppm of solute C, while the r -value and {222} diffraction intensity ratio for steel C with over 30 ppm of solute C were minima.

Ushioda and Abe (1984) reported that relatively wide and coarsely dispersed deformation bands, which had {110}<001> and {100}<011> orientations locally curved with high dislocation density, were formed when an Fe-0.01%N alloy was rolled at temperatures from 573K to 673K at which dynamic strain aging occurred. These recrystallization grains with the same orientation as the deformation bands nucleated preferentially at the bands, resulting in the development of a {110}<001> recrystallization texture. Although an analysis of orientation in the recrystallized grains through the recrystallization process was not carried out in the present study, the results for steel C with a large amount of solute C shown in Figure 6 correspond to those of Ushioda and Abe. These phenomena are considered to have been caused by strong interaction between the dislocations induced by rolling and the interstitial solute atoms, resulting in the formation of the deformation bands already mentioned, when a large amount of interstitial solute C and/or N existed and the sheet steel was rolled at the temperature at which dynamic strain aging occurred.

The rolling temperature at which dynamic strain aging occurred was 773K in this study, which is considerably higher than that reported by Ushioda and Abe. The reason for this is that the rolling strain rate of 1000s^{-1} used in this study is extremely higher than that of 3s^{-1} used in Ushioda's study, although the diffusion coefficient of N in iron is larger than that of C in iron.

On the other hand, the r -value and {222} diffraction intensity ratio for the sheet steel B containing about 10 ppm of solute C showed maxima at the rolling temperature of 773K at which the dynamic strain aging occurred. Moreover, the {222} and {200} line broadenings of the as-rolled specimen also showed maxima at the rolling temperature of 773K. Wide X-ray line broadening in as-rolled specimens indicates the high residual strain in specifically oriented grains, corresponding to the stored energy. As a result, it is considered that a uniform deformation structure formation of deformation bands,

and that the stored energy would be increased by the interaction between solute C and dislocations when the hot rolling is performed at the rolling temperature that introduces dynamic strain aging.

Oriented nucleation theory and high-block energy theory both assume that there is a non-homogeneous distribution of strain energy in an as-rolled sheet steel, and that the region of high strain energy in the deformed grains recovers rapidly and nucleates during annealing, resulting in the development of a recrystallization texture. In this study, the r -value and the $\{222\}$ diffraction intensity ratio of as-annealed specimens showed maxima at the rolling temperature of 773K by dynamic strain aging. This change is considered to have arisen as the result of an increase in the $\{222\}$ residual strain caused by dynamic strain aging. With this phenomenon, the region of high stored energy in the $\{222\}$ component would recover and nucleate faster, and the $\{111\}$ recrystallized grains form more numerous and grow easily to a greater size, resulting in the maximum r -value and $\{222\}$ diffraction intensity ratio at the rolling temperature of 773K.

The reason for the deformation behavior being significantly different between the specimens containing solute C of about 10 ppm and over 30 ppm is proposed as follows: Considering that deformation and rotation of the sheet steel would occur by the slip associated with the movement of dislocations, it is necessary for a uniformly deformed microstructure to move the dislocations smoothly without any strong interaction between the solute C and dislocations. On the other hand, it is effective for storing working strain that the solute C would interact with the dislocations due to dynamic strain aging. As a result, when there is a large amount of solute C (over 30 ppm) in the parent steel before rolling, a significant deformation band would be formed by dynamic strain aging, resulting in the abrupt decrease in r -value and $\{222\}$ diffraction intensity ratio. Conversely, a small amount of solute C (about 10 ppm) in the parent steel before rolling would induce both uniform microstructure and the high $\{222\}$ residual strain by dynamic strain aging, resulting in the increase in r -value and $\{222\}$ diffraction intensity ratio.

The authors have already reported that an increased in hot rolling strain rate in the ferrite region can lead to an increase in the $\{222\}$ residual strain of an as-rolled sheet steel, and that an intense $\{111\}$ recrystallization texture developed (Matsuoka *et al.*, 1988). Considering the results from the previous study, it is obvious that an increase in the residual strain of the $\{222\}$ component in an as-rolled sheet steel would be effective for the development of a $\{111\}$ recrystallization texture. The authors believe that there are two other ways to increase the residual strain of the $\{222\}$ component of as-rolled steel and to develop a strong $\{111\}$ recrystallization texture than by decreasing the rolling temperature. One is to increase the hot rolling strain rate in the ferrite region, and the other is to hot roll at the temperature that induces dynamic strain aging.

As shown in Figure 9, the $\{200\}$ diffraction intensity ratio for the annealed specimen was not affected by dynamic strain aging, even when the $\{200\}$ residual strain also showed a maximum at the rolling temperature of 773K. Since the $\{200\}$ deformed grains would recover and nucleate later than the $\{222\}$ deformed grains, the recrystallization behavior of the $\{200\}$ deformed grains would hardly be changed by the increase in residual strain.

The authors believe that the abnormal phenomenon at the rolling temperature of around 773K is attributable to dynamic strain aging, and the effect of solute C has been investigated only during hot rolling in this study. Since solute C influenced the formation of the recrystallization texture during annealing as well as during hot rolling, it is necessary to investigate, in the future, the effect of solute C during annealing.

5. CONCLUSION

The formation of a {111} recrystallization texture and the associated *r*-value were investigated as functions of the solute C content and rolling temperature in the ferrite region. The results are summarized as follows:

- (1) When the amount of solute C is less than 1 ppm before rolling, the *r*-value of a sheet steel annealed after rolling in the ferrite region progressively increases with decreasing rolling temperature.
- (2) When the amount of solute C exceeds 30 ppm before rolling, deformation bands are formed and the {110} recrystallization texture preferentially develops at the rolling temperature of around 773K, resulting in an abrupt decrease in *r*-value, which is almost the same effect as that reported for the case of solute N.
- (3) The deformation bands at the rolling temperature of around 773K are considered to have been induced by non-uniform deformation, which is caused by strong interaction between the large amount of solute C of over 30 ppm and the dislocations associated with dynamic strain aging.
- (4) When the amount of solute C is about 10 ppm before rolling, the *r*-value and {222} intensity ratio of the sheet steel annealed after rolling increase with decreasing rolling temperature, except for showing maxima at the rolling temperature of around 773K.
- (5) An X-ray line broadening analysis of the as-rolled sheet steel shows that the residual strain in the {222} component of the specimen rolled at 773K is higher than that of the specimen rolled at other temperatures.
- (6) Based on the foregoing results, it is considered that dynamic strain aging provides high stored energy in the {222} component, with the result that an intense {111} recrystallization texture develops.
- (7) It is concluded that the {111} recrystallization texture would develop with dynamic strain aging at the rolling temperature of around 773K by retaining a small amount of solute C (about 10 ppm) before rolling in the ferrite region.

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