Effect of Chromium Content on Relationship Between \( r \)-value and \{111\} Recrystallization Texture in Ferritic Steel

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The deep drawability of ferritic steels can be improved by increasing the \( r \)-value, which is closely related to the \{111\} recrystallization texture. Control of the \{111\} recrystallization texture in ferritic steel has been extensively studied as the means of improving the deep drawability of the steel sheet. However, no published research has compared the relationship between the \( r \)-value and \{111\} recrystallization texture in mild steels and ferritic stainless steels.

The aim of the present research is to clarify the effect of chromium on the relationship between the \{111\} recrystallization texture and \( r \)-value in order to improve the deep drawability of ferritic stainless steels. The following results were obtained through the study using body-centered cubic steels with various chromium content: (1) Ferritic stainless steels show higher \( r \)-values than mild steels with the same level of \{111\} recrystallization texture. (2) Chromium addition is advantageous for realizing higher \( r \)-values in ferritic steel.

The authors propose that some slip systems in ferritic steels are constrained by chromium atoms, which results in higher \( r \)-values in ferritic steels.

KEY WORDS: ferritic stainless steel; mild steel; texture; \( r \)-value; deep drawability; formability; slip system; forming; chromium; iron.

1. Introduction

Increasing the \( r \)-value (Lankford value) is effective for improving the deep drawability of mild steels (IF steels). The \{111\} orientation texture is known to show the highest \( r \)-value in body-centered cubic (BCC) structure.\(^3\) Therefore, control of the \{111\} recrystallization texture in mild steel has been investigated with the aim of increasing the \( r \)-value.\(^2,3\) Since ferritic stainless steels have the same BCC structure as mild steel, it is also thought possible to realize high \( r \)-values by controlling the \{111\} recrystallization texture in a similar manner. Representative controlling methods of \{111\} recrystallization texture include 1) high purification (mainly reduction of the solid solution carbon), 2) refinement of the microstructure of hot rolled and annealed sheets, 3) enhancement of the \{111\} recrystallization component in hot rolled and annealed sheets, 4) grain size control in cold rolled and annealed sheets, and 5) optimization of the cold rolling reduction ratio and others.\(^2,3\) Based on these \{111\} recrystallization texture control technology, high drawability mild steel sheets with \( r \)-values as high as 3.0 have been developed,\(^3\) and the authors also found that high \( r \)-value of 2.6 can be realized in ferritic stainless steel.\(^7\)

The \( r \)-value is defined as the ratio of the logarithmic strain in the sheet transverse direction to the logarithmic strain in the sheet thickness direction. Materials with high \( r \)-values are characterized by low thickness reduction ratio during forming, which is advantageous for deep drawing applications.

As described above, control of the \{111\} recrystallization texture in ferritic steel is important for increasing the \( r \)-value. However, no published research has compared the relationship between the \( r \)-value and \{111\} recrystallization intensity in mild steels and ferritic stainless steels at the same time. Therefore, the authors investigated the effect of chromium content on the relationship between \( r \)-value and \{111\} recrystallization texture in order to acquire a better understanding for the improvement of the deep drawability of ferritic steels.

The present research attempts to explain why it is possible to obtain high \( r \)-values in ferritic stainless steel in spite of the remarkable difference in the \{111\} recrystallization intensity compared with mild steels.

2. Experimental Procedure

The three alloy compositions shown in Table 1 were used as test materials. Steel A and steel B are titanium bear-

| Table 1. Chemical composition of specimens (mass%). |
| Steel | C   | Cr  | Mo | N  | Ti |
|-------|-----|-----|----|----|----|
| A     | 0.004 | 18.0 | 1.2 | 0.007 | 0.22 |
| B     | 0.005 | 11.0 | ---- | 0.007 | 0.20 |
| C     | 0.002 | ---- | ---- | 0.003 | 0.07 |

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ing high purity ferritic stainless steel. Steel C is titanium bearing mild steel (IF steel). Slabs of these compositions were reheated, hot rolled, annealed as hot band, cold rolled, annealed, for recrystallization and finally provided for experiments in the form sheets with a thickness of 0.8 mm. Average $r$-values of these specimens ranged from 1.8 to 2.7.

Tensile tests (JIS Z 2241) were performed at a cross-head speed of 10 mm/min using JIS 13B test pieces. The $r$-value was measured at an elongation of 15%. The average $r$-value was calculated using the following equation

$$\text{Average } r\text{-value} = \frac{r_0 + 2r_{45} + r_90}{4}$$

(The angle refers to the direction at which the test piece was taken relative to the rolling direction.)

Specimens for the recrystallization texture measurements were prepared by #600 wet polishing, and surface residual stress was relieved by chemical etching. Pole figures were measured by Schultz type X-ray diffraction at the 1/2 thickness with 5° steps. The 3-dimensional crystallographic orientation distribution function (ODF) was obtained from the three measured incomplete pole figures ($\{110\}$, $\{200\}$ and $\{211\}$). For the observations of the slip lines and internal microstructure, JIS-5 test pieces taken parallel to the rolling direction were prepared. The slip lines in test pieces given 25% tensile strain at room temperature were observed using an optical microscope. The dislocation morphology of test pieces given 10% tensile strain at room temperature were observed using a transmission electron microscope (TEM: Hitachi HF-2000).

3. Experimental Results

Cold rolled and annealed sheets with a thickness of 0.8 mm were prepared under various manufacturing conditions to provide various samples with different $\{111\}$ intensity. Figure 1 shows the relationship between the $\{111\}$ intensity and measured average $r$-value of final annealing sheets with average $r$-values ranging from 1.8 to 2.7. The ordinate is the average $r$-value, and the abscissa is the $\{111\}$ intensity ratio to random. Here, (1,0,1) was used as the origin of the straight and broken lines, since randomly oriented sample has to show no anisotropy.

Ferritic stainless steels and mild steel show a positive correlation between the $\{111\}$ intensity and average $r$-value, but ferritic stainless steel shows higher $r$-values than mild steel with the same level of $\{111\}$ intensity. Figure 2 shows effect of the chromium content on slope A (average $r$-value/\{111\} intensity) of final annealing sheets.

This fact indicates that high $r$-values cannot be determined by the $\{111\}$ intensity alone. Therefore, ODFs of steels A and C with average $r$-values of 2.6 were obtained from pole figures, and the textures were compared to each other. Figure 3 shows the ODFs (Euler angular notation as proposed by Bunge) at the $\phi_2$=45° cross section. The numerical values in the figure are the intensity ratio to random. The solid lines show $\gamma$-fibers, which are known advantageous to increase the $r$-value. The $\gamma$-fiber is a fibrous texture in which the $\{111\}$ direction is parallel to the normal direction and is positioned at $\phi=54.7^\circ$. It is clear that steel A has higher $r$-value in spite of comparatively low achievement of the $\{111\}$ recrystallization texture in comparison with steel C. These figures suggest that $\{111\}$ recrystallization texture does not fully explain $r$-values, and chromium atoms play an important role to determine high $r$-values in ferritic steels.

Figure 4 shows the ODFs ($\phi_2=45^\circ$ cross section) for the three specimens of steel A with different $r$-values (average $r$-value=1.9, 2.3, 2.6). This comparison is intended to reconfirm that the $r$-value of ferritic steel with the same composition is dependent on $\{111\}$ recrystallization texture.

The $\{100\}$ recrystallization texture ($\alpha$-fiber) is known to be detrimental to $r$-values. Figure 5 shows the relationship between the measured $\{111\}$ intensity and the $\{100\}$ intensity of final annealing sheets with average $r$-values of 2.6. In comparison with steel C, steel A has higher $\{100\}$ intensity and lower $\{111\}$ intensity. When the relationship between the $r$-value and $\{111\}/\{100\}$ (or $\ln\{111\}/\{100\}$) is compared with Fig. 1, the difference in the slopes shows further increase.

In order to clarify the relationship between the recrystallization texture and $r$-value, the $r$-value was calculated for mild steel and ferritic stainless steel ($r=2.6$) based on the obtained ODF. Assuming full constraint $\{111\}$ pencil glide
The average $r$-values of mild steel (steel C) and ferritic stainless steel (steel A) were calculated as 2.82 and 2.56 respectively. The calculated average $r$-value of mild steel C is about 0.26 higher than that of the ferritic steel A.

These $r$-values calculated from ODFs again show that the measured $r$-value is not explained by the recrystallization texture.

In summary, two points were revealed by the experiments described above:

1) Ferritic stainless steels show higher $r$-values than mild steel with the same level of $\{111\}$ intensity. Chromium addition is advantageous for realizing higher $r$-values in ferritic steels.

2) $r$-values of the ferritic steels are not determined by texture alone. It is considered that chromium atoms play an important role in determining $r$-values in ferritic steels.
4. Discussion

Because the slope of the average \( r \)-value against \{111\} recrystallization texture relation is steeper in ferritic stainless steel than in mild steel. It is clear that chromium addition is advantageous for realizing higher \( r \)-values in ferritic steel.

Plastic deformation of steels occurs due to “slip” along designated crystallographic planes, which are called slip planes. The slip plane is normally the plane that has the largest planar spacing (i.e., crystallographic plane with the highest atomic density), and therefore has the lowest resistance to slip. The slip direction is generally the direction where the atomic spacing is the smallest. In comparison with austenitic steels, which have face-centered cubic (FCC) structure, no plane in BCC steel has a conspicuously higher atomic density, which enables multiple slip systems. Generally, the slip planes in BCC structure are the \{110\}, \{112\}, and \{123\} planes, and the slip direction is \( \langle 111 \rangle \) in all cases. Even considering only \{110\} and \{112\} slip systems, which are near the densest planes, twenty-four slip systems exist.

Considering the results described in the previous chapter, the authors propose the following mechanism by which chromium contributes to higher \( r \)-values.

Both mild steel and ferritic stainless steel have the same BCC structure. In ferritic stainless steel, iron contains substitution chromium atoms at a rate of about 10–30%. Chromium atoms might affect some of the slip planes in BCC metals, and as a result, certain slip plane may acquire priority. If some of these slip planes were constrained by chromium atoms, anisotropy of plastic deformation should increase and thus increase \( r \)-value.

To clarify the effect of chromium on constraint of slips, observation of slip bands and dislocation morphology were carried out. Figure 6 shows the results of tensile test performed at room temperature with 25% tensile strain in the rolling direction. Observation of the slip bands revealed many straight slip bands in steel A, and numerous wavy-shaped slip bands in steel C. It is considered that the latter are attributed to multiple slip.

The internal microstructure of specimens given 10% tensile strain in the rolling direction was observed by TEM. Figure 7 shows the dislocation morphology observed from the \( \langle 111 \rangle \) direction. Intertwinement of dislocations is slight, and the dislocations show a linear arrangement in steel A. On the other hand a number of dislocations exist, and the dislocations tend to form interwined cells in steel C. These observations suggest that certain slip systems are constrained in ferritic stainless steel, but fewer or none are affected in mild steel.

In order to understand the mechanism that designated
slip systems are constrained, it may be mentioned that addition of chromium to iron promotes the formation of twin crystals. It has been shown that the formation of twins in ferritic steel suppresses slip on the \{112\} plane. The mechanism responsible for this phenomenon is considered as follows: Sleeswyk\(^6\) proposed that \(a/2\{111\}\) dislocation on the \{112\} plane in BCC metals expands to \(a/6\{111\}\) partial dislocation, and the three partial dislocation are then connected by stacking faults (\(a\) = lattice constant). Based on the fact that chromium addition to iron promotes the formation of twins, the authors assumed that chromium addition reduces stacking fault energy. The expansion of dislocations to partial dislocations is difficult by this reduction in stacking fault energy. This implies that the \{112\}\{111\} slip systems are suppressed by the chromium atoms.

Another factor which may suppress designated slip systems in Fe–Cr alloys, is the fact that chromium atoms tend to occupy designated atomic positions of BCC unit cell. It can be thought that solute atoms (chromium) concentrate in designated sites because modulus of elasticity is anisotropic.\(^8\) The order of the modulus of elasticity in iron is \{100\} < \{110\} < \{111\}. Decomposition of the chromium occurs most easily along the \{100\} plane, which has the smallest modulus of elasticity.

It is well known that spinodal decomposition is characterized by respective phase separation of iron atoms and chromium atoms. This phase separation is one of the cases where chromium atoms concentrate in a designated atomic position of BCC unit cell in Fe–Cr alloy. Mima and Yamaguchi\(^10\) have reported that a plate-like chromium-rich zone forms on the \{100\} plane when Fr–Cr alloys are aged, and as a result, slip deformation is suppressed and twin deformation occurs. Considering these points, it is reasonable to assume that decomposition and precipitation of chromium atoms to designated sites in BCC unit cell may be a factor which constrains slip systems.

Inoue\(^11,12\) calculated \(r\)-values for cases where both \{110\}\{111\} and \{112\}\{111\} slip systems are active, and where only \{110\}\{111\} slip system is active under various constraint conditions, from the same ODF data. The calculated results showed that \(r\)-value is higher when the slip systems are limited only to \{110\}\{111\} under various constraint conditions. From this result, it is demonstrated that limiting designated slip systems increases plastic deformation anisotropy, which is advantageous for realizing higher \(r\)-values. Thus, if chromium limits some slip systems in BCC metals, it is possible to explain why high \(r\)-values are easily achieved in chromium-bearing ferritic steels.

5. Conclusions

To obtain an index for improving the drawability of ferritic stainless steels, the effect of chromium on the relationship between \{111\} recrystallization texture and \(r\)-value was clarified using ferritic stainless steel and mild steel.

(1) Ferritic stainless steels and mild steel show a positive correlation between the \{111\} intensity and average \(r\)-value. But, ferritic stainless steel shows higher \(r\)-values than mild steel with the same level of \{111\} intensity. The slope of the average \(r\)-value against \{111\} intensity shows dependency on chromium, and the slope becomes steeper as the chromium content increases. It is clarified that chromium addition is advantageous for realizing higher \(r\)-values in ferritic steel.

(2) Based on the above mentioned results, the authors thought that chromium atoms suppress some slip systems in ferritic steel. Two hypotheses are proposed. a) From the fact that chromium addition to iron promotes twin formation, chromium addition may reduce stacking fault energy. The expansion of dislocations to partial dislocations is difficult by the reduction in stacking fault energy, which implies that \{112\}\{111\} slips are suppressed by the chromium atoms. b) Chromium atoms tend to occupy the designate site in BCC unit cell. This fact also suggests that chromium atoms constrains slip systems.

Thus, it is considered that limiting designated slip systems increases plastic deformation anisotropy, which is advantageous for realizing higher \(r\)-values. If chromium limits some slip systems in BCC metals, it is possible to explain why high \(r\)-values are easily achieved in chromium-bearing ferritic steels.

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