Deformation Microstructure and Texture in a Cold-Rolled Austenitic Steel with Low Stacking-Fault Energy

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Deformation microstructure and texture in a cold-rolled austenitic steel 310S with low stacking fault energy (SFE) have been investigated by SEM and TEM. When 310S plates were rolled by 95% in thickness reduction, a remarkable development of the brass-type preferred orientation was confirmed by X-ray diffraction. SEM observations of the specimens with such texture exhibited a layered structure developed parallel to the rolling plane. TEM images demonstrated the inhomogeneity of highly deformed polycrystalline low SFE fcc metals. Dominant microstructure in 95% cold-rolled specimens is fine-grained structure caused by deformation twin and shear band formation (area (I)), but there exists small amount of areas without twin (area (II)). Dominant texture is the \{110\}(112) brass-type in both areas (I) and (II), although \{111\}(112) (twin-matrix lamellae), \{110\}(001) (Goss), \{112\}(111) (copper-type) are also observed in the area (I).

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1. Introduction

310S stainless steel has a very low stacking-fault energy, and it is known that a typical rolling texture of brass-type (ND // \{111\}, RD // \{112\}) is developed in this alloy.\(^1\) Such texture development must be closely related to the microstructural evolution during the cold rolling.\(^2-3\) However, few studies have been made on the relation between microstructures observed by TEM and textures developed during cold-rolling. In the present study, deformation microstructure developed in cold-rolled 310S steel has been analyzed by a TEM dark-field method. Particular attention has been paid on the fine-grained structure and micro texture developed through deformation twinning and shear band multiplication.

2. Experimental Procedure

We employed an austenitic stainless steel 310S whose chemical composition is shown in Table 1. The stability of \(\gamma\)-phase in this alloy is very high even when it is subject to heavy deformation due to rolling. Polycrystalline plate specimens (initial thickness: 12 mm, average grain size: 100 \(\mu\)m) were rolled up to thickness reductions of 50, 70, 90 and 95% (equivalent true strain 0.8–3.4) at room temperature. Deformation structures of the rolled specimens were observed by both SEM and TEM. For TEM samples, the cylindrical rod (3 mm\(\phi\)) was trepanned along transverse direction (TD) from rolled sheet by spark cutting, and it was sliced into disks with 1 mm thickness. Those disks were mechanically polished and electro-polished into foils by a twin-jet technique. TEM observations were carried out with JEM-200CX operated at 200 kV in the HVEM laboratory at Kyushu University. The specimen foil surface is parallel to the longitudinal plane of rolled sheets, so that the microstructures shown in the present paper are those observed from TD. The development of preferred orientations in the planes normal to the ND was measured at the mid-thickness section of specimen plates by using the X-ray back reflection method. From three incomplete pole figures (\{111\}, \{200\} and \{220\}) measured, the complete pole figures were calculated using the iterative series expansion method, which has been originally proposed by Dahms and Bunge\(^4\) and modified by Inoue and Inakazu.\(^5\)

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
<th>Cr</th>
<th>Fe</th>
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<tr>
<td>0.05</td>
<td>0.84</td>
<td>1.20</td>
<td>0.016</td>
<td>0.001</td>
<td>19.2</td>
<td>24.8</td>
<td>Bal.</td>
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3. Results and Discussion

Figures 1(a), (b) and (c) show the X-ray \{111\} pole figures of the 310S steel cold-rolled by 50%, 70% and 90% in thickness reduction, respectively. A texture is developed with the rolling strain, where it is the typical brass-type texture \{110\}(112) being commonly observed in cold-rolled fcc metals with very low stacking fault energies (SFE) less than 10 mJ m\(^{-2}\).\(^6\)

Figures 2(a), (b), (c) and (d) show SEM micrographs of deformation structures observed by surface etching at thickness reductions of 50, 70, 90 and 95%, respectively (longitudinal plane). Here, the rolling direction (RD) is parallel to the horizontal direction. At 50% in thickness reduction, as shown in Fig. 2(a), we can see grain boundaries and banded structure inclined about 45 degrees to RD. At 70%, grains are elongated along RD, and in addition, coarse bands inclined 35–40 degrees to RD are observed in the elongated grains. They correspond to shear bands where intense shear deformation occurs. The detail of shear bands in 310S was reported in the previous paper by present authors.\(^7\)

One of the typical TEM micrographs is shown in Fig. 3, where the rolling strain is 70% and the incident beam direction is parallel to TD. In Fig. 3(a), the shear band is seen as a bright band formed in the lamellae which are almost parallel to the rolling direction. Selected area diffraction patterns obtained from the lamellae and shear band are exhibited in Figs. 3(b) and 3(c), respectively. Figure 3(b) shows a diffraction pattern of matrix and twinned area with...
the (110) incidence, which indicates that the lamellae observed are formed by thin layers of deformation twin. Shear bands are formed in such twin and matrix (T-M) lamellar structure, and fine-grained structure appears in the areas of shear bands. Figure 3(c) shows a diffraction pattern obtained from one of shear bands, indicating that orientation scattering is caused by the formation of fine-grained structure.

Figure 4 illustrates the aspect of the microstructural evolution in cold-rolled 310S steel, where the essential process is the development of fine lamellar structure due to deformation twin and the following grain refinement due to shear band formation. The multiplication of shear bands destroys T-M lamellar structure to form fine-grained structure. SEM micrographs in Figs. 2(b) and 2(d) show such grain refinement with increasing rolling strain from 70%
to 95%. Figure 4(b) exhibits the area percentages of T-M lamellae and fine-grains observed at rolling strains from 50 to 95%. The areas of T-M lamellae and fine-grains were determined from TEM bright-field images, where the total image area used for each measurement was around \(9 \times 10^4 \mu m^2\). The bar graph demonstrates a clear tendency of microstructural change from lamellar structure to fine-grained structure. However, in addition to this structural change, note here the white areas in the bar graph of Fig. 4(b) indicates the regions where no T-M lamellae were observed, i.e., neither deformation twin nor shear bands occurred in those regions. The areas without twin still exist even at 95% rolling strain where the brass-type texture has been fully developed. Although such area is not a major part in the overall aspect of microstructures, this suggests that there are two kinds of processes in microstructural development: (I) the process with the occurrence of deformation twin, (II) the process without any relation to twin. Such difference in microstructural developments should have a close relation to the local evolution of texture. In order to examine how texture components relate to such microstructures observed, detailed TEM observations were carried out using the specimen with well-developed rolling texture.

Figure 5(a) shows a bright field (BF) image of the microstructure observed in a specimen rolled by 95%. Here, the rolling direction (RD) is parallel to the horizontal direction. Enlarged BF images of circled areas labeled (I) and (II) in Fig. 5(a) are shown in Figs. 5(c-I) and 5(c-II), respectively. Diffraction patterns obtained from the selected areas shown in (c-I) and (c-II) are also exhibited in Figs. 5(b-I) and (b-II), respectively. The BF image in Fig. 5(a) demonstrates not only the complexity of the microstructure but also the characteristic inhomogeneity of highly deformed polycrystalline materials. The area (I) corresponds to a fine-grained structure formed by deformation twin and shear band multiplication, where large scattering of crystal orientations is observed in the SAD pattern. This kind of microstructure occupies dominant part of specimens, as illustrated by the shaded areas in the bar graph of Fig. 4(b). On the other hand, the area (II) corresponds to those of white areas in the bar graph of Fig. 4(b), where no indication of twin is observed and its diffraction pattern consists of the diffracted spots of the (111) incident beam direction. In addition, the directions of (110) and (211) in the pattern of (b-II) coincide with those of ND and RD, respectively (see the arrows in the bottom right of Fig. 5(a)). This orientation is in good agreement with the preferred orientation of brass-type texture. Thus, cold-rolled microstructure developed without deformation twin does not have large orientation scattering, and it converges to the [110][112] brass-type texture. Next, in order to make more detailed analysis for the microstructure of the area (I) with large orientation scattering, dark field (DF) image analyses were carried out. Figure 6(a) shows the same fine-grained structure as shown in Fig. 5(c-I). Its diffraction pattern is also exhibited again in Fig. 6(f), where the illustrations of diffracted spots due to important textures are superimposed on the diffraction pattern. Here, four kinds of symbols □, ○, △, ▽ represent the diffraction patterns of crystals whose orientations of ND and RD (see the arrows in the bottom right of Fig. 6(f)) are parallel to {110}[112] (brass-type), {111}[112] (twin-matrix lamellae), {110}[001] (Goss), {112}[111] (copper-type), respectively. Figures 6(b)(c)(d) and (e) show the DF images obtained using the diffracted beam indicated by the arrows labeled (b)(c)(d) and (e), respectively. Those diffracted beams correspond to those due to the components of {110}[112], {111}[112], {110}[001] and {112}[111]. The DF images indicate that the fine-grained area observed in Fig. 6(a) includes all the four components, although the most important component is the [110][112] brass-type texture. It is also to be noted that the orientation of twin-matrix lamellar structure still remains as a major component in spite
that the T-M lamellae seems to have been already destroyed by the multiplication of shear bands. Fine distribution of Goss and copper type areas are also observed like fibers.

Thus, the present TEM observation demonstrates that deformation twinning and the following shear band formation are the main processes in microstructural development in cold-rolled fcc metals with low stacking fault energy. Such microstructural development enhances grain refinement with large scattering of crystal orientation. Origin of such grain refinement is in the morphology of deformation twin in low SFE metals. When the SFE is low enough, twin nucleation occurs very easily so that a high density of thin twins is introduced. The thickness of the twin layer is less than a few hundred nanometers, and the misorientation at the boundary
is large enough to be an effective barrier against the motion of dislocations. Therefore, the progressive development of a fine twinned structure leads to subdivision of a crystal grain. In addition, when such fine lamellae are formed, a high density of dislocations must accumulate at the boundaries during subsequent deformation. This suggests that the T-M lamellar structure should work harden rapidly, which induces the next deformation mode, the formation of shear bands related to a type of recovery or recrystallization. Thus, because of twinning, grain-refinement in low SFE metals is very effective, compared with high SFE metals in which very large strains are required to form boundaries with sufficient misorientation, as was pointed out in Ref. 9).

Finally, it should be noted that there remain areas with the microstructure without any relation to deformation twin, i.e. plastic deformation in such area occurs only by slip deformation. Although such area may be a minor part, the microstructural characteristics such as grain size, orientation scattering and preferred orientations are quite different from those of the area relating to deformation twin, as shown in Fig. 5. The reason why such difference in the occurrence of deformation twin appear has not been clarified yet, but one possibility is the orientation dependence of deformation twin in fcc crystals.\(^8\) The onset shear stress \(\tau_T\) for twinning in tensile tests of fcc crystals is reduced as the specimen axis approaches the \(\{111\}\) pole. On the other hand, \(\tau_T\) is increased as the specimen axis approaches the \(\{100\}\) pole. This orientation dependence of \(\tau_T\) is understood by the complementary relation between cross slip and twinning deformation.\(^9\) Such orientation dependence of the ease with which twinning occurs may cause the inhomogeneity of the occurrence of twinning in polycrystalline fcc metals, which can induce the two types of microstructural development (I) and (II) as observed in Fig. 5(a).

4. Summary

Deformation microstructure and texture in a cold-rolled austenitic steel 310S with low stacking fault energy (SFE) have been investigated. Dominant microstructure in 95% cold-rolled specimens is fine-grained structure caused by deformation twin and shear band formation (area (I)), but there exists small amount of areas without twin (area (II)). Dominant texture is the \(\{110\}\{112\}\) brass-type in both areas (I) and (II), although \(\{111\}\{112\}\) (twin-matrix lamellae), \(\{110\}\{001\}\) (Goss), \(\{112\}\{111\}\) (copper type) are also observed in the area (I).

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REFERENCES