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The role of deformation twinning in the formation of a fine-grained structure in cold-rolled 310 steels

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The role of deformation twinning in the formation of a fine-grained structure in cold-rolled 310S steels has been investigated by using transmission electron microscopy. When 310S steels were rolled at room temperature, a deformation twin appeared in the early stage of the rolling process. A fine lamellar structure consisting of the deformation twin and matrix (T-M) developed with increased rolling strain. At moderate strain, micro shear bands appeared in the area of the T-M lamellar structure, and the shear bands grew and multiplied during further cold-rolling. Such multiplication of shear bands destroyed the T-M lamellar structure, which caused the development of a fine-grained structure. In shear bands, a highly misoriented structure with a submicron grain size was generated. In this paper, the mechanism of the transmutation process from the T-M lamellae into the fine-grained structure is discussed. Particular emphasis is laid on the role of deformation twinning in the formation of the fine-grained structure.

1. Introduction

There is great advantage in grain refining polycrystalline materials to obtain high-performance mechanical properties. Recently, in order to attain ultra fine-grained structures in bulk materials, much attention has been paid not only to the techniques using traditional heat treatments but also to several new methods using severe plastic deformation such as ECAP (equal-channel angular pressing) [1] or ARB (accumulative roll-bonding) [2], where the materials to which it is applied mainly have a relatively high stacking fault energy.

In the case of ordinary cold-rolling, microstructures evolve by inhomogeneous plastic deformation, which develops deformation bands or boundaries that give rise to the subdivision of crystal grains. As is well known, the stacking fault energy (SFE) is one of the most important parameters influencing the formation of such deformation microstructures in fcc materials. In fcc pure metals with a high SFE such as aluminum, the prominent structures are dense dislocation walls (DDWs) and micro bands (MBs) which are geometrically necessary boundaries (GNBs) bearing the misorientations between adjacent subdivided areas in a grain [3]. On the other hand, in fcc alloys with a low SFE such as α-brass, deformation twinning has an important role: when they are cold-rolled, a dense twin-matrix (T-M) lamellar structure develops which is then destroyed by the development of shear bands (SBs) to form a fine-grained structure.

We have been studying the microstructural evolution due to cold rolling in austenitic stainless steels 310S(Fe-25%Cr-20%Ni) whose SFE is thought to be as low as that of α-brass. In this work [4], it was found that the key process for the grain refinement is the shear banding in the T-M lamellar structure. However, little attention has been paid to the influence of deformation twinning on the process of grain refinement in low SFE metals.

In the present paper, we will report the development of fine-grained structure, where special emphasis is laid on the role of deformation twinning on grain refinement in low SFE metals.

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TABLE I Chemical composition of SUS310S (mass%)

C	Si	Mn	P	S	Ni	Cr
0.05	0.84	1.20	0.016	0.001	19.2	24.8

2. Experimental

We employed an austenitic stainless steel 310S whose chemical composition is shown in Table I. Polycrystalline plates (initial thickness: 12 mm, average grain size: 100 μm) of this steel were rolled by reductions of 20–90% in thickness at room temperature. Deformation structures in the rolled sheets were observed by TEM. For preparing TEM samples, cylindrical rods (3 mmϕ) were trepanned the along the transverse direction (TD) in rolled sheets by spark cutting, and subsequently sliced into disks 1 mm thick. These disks were mechanically polished and electro-polished into thin 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. In every case the foil surface was perpendicular to the transverse direction. Local orientations were measured by selected area diffraction (SAD).

3. Results and discussion

Fig. 1 shows bright-field (BF) TEM images demonstrating morphological characteristics developed with increasing rolling strain. In this observation, particular attention was paid to the deformation twinning observed in longitudinal sections. At a 20% thickness reduction (Fig. 1a), relatively sharp and straight bands along the $\langle 211\rangle$ direction are visible against a background of a kind of dislocation cell structure. The diffraction pattern obtained from the area including the straight bands clearly exhibits the twin-matrix double spot pattern, indicating that deformation twinning has already occurred in spite of the rather early stage of the cold-rolling. It is to be noted here that those twin bands are rather thin and the thickness of those bands is less than 0.1 μm at the minimum. This morphological characteristic observed here is a general feature of deformation twinning in fcc materials with very low SFE [5]. At 50% reduction (Fig. 1b), deformation twinning is plentiful, and dense structures due to thin twinned layers are seen approximately horizontal, although they are somewhat curved. The direction of the twinned layers tends to be parallel to the rolling direction (RD) when it is compared with that of the twin bands observed in Fig. 1a. Selected area diffraction patterns showed that these boundaries are parallel to the trace of the $\{111\}$ plane, i.e., the twinning plane in fcc crystals. The density of such twinned layers increases with increasing rolling strain. In Fig. 1c (70% thickness reduction), the whole area observed is covered by a dense lamellar structure which is almost straight and parallel to the RD. Since this lamellar structure is considered to consist of twin and matrix thin layers, we call this structure a "T-M (twin and matrix) lamellar structure". Fig. 2 shows the change of the inclination of the T-M lamellae to the rolling direction with increasing rolling strain. Here, horizontal and vertical axes indicate the thickness reduction (the rolling strain) and the angle between the RD and the direction of lamellae, respectively. Since the deformation is inhomogeneous and the lamella angle changes from place to place, a variation in angle was recorded. In Fig. 2, the scatter is shown by a bar graph. As is seen in this figure, the average angle decreases with increasing reduction, indicating that the inclination of T-M lamellae to the rolling direction decreases with the rolling strain. In addition,

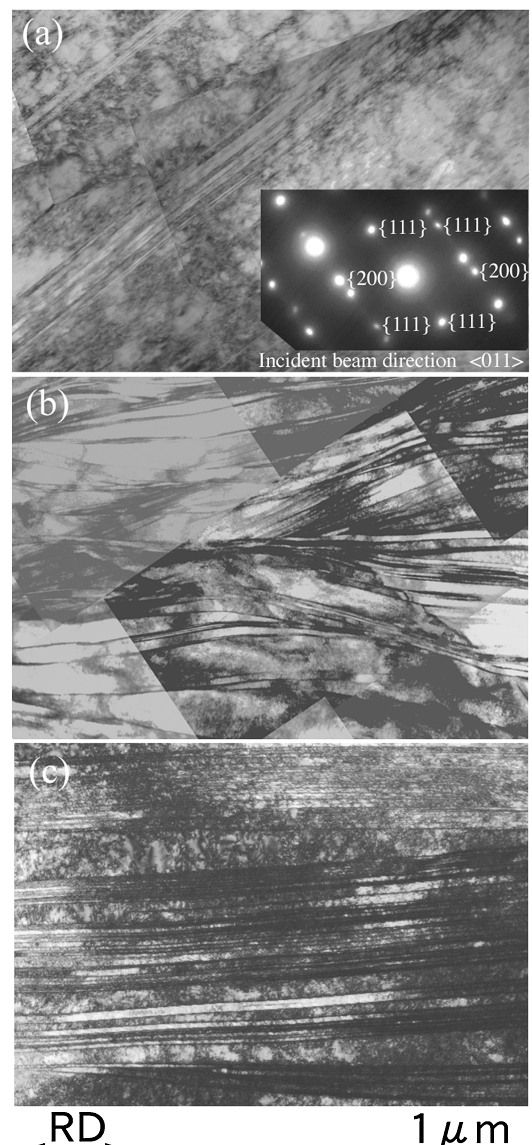

Figure 1 Evolution of the twin-matrix (T-M) lamellar structure observed in longitudinal sections. Bright field images at (a) 20% (b) 50% (c) 70% thickness reduction. Selected area diffraction pattern in (a) indicates the occurrence of deformation twins.

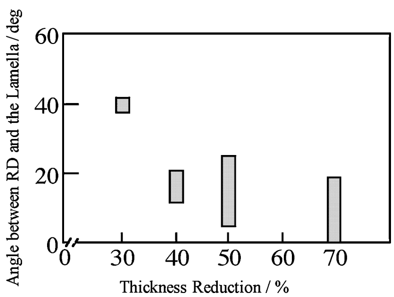

Figure 2 Change of the angle between the rolling direction (RD) and the plane of twin-matrix lamellae during cold-rolling.

it is also seen that the scatter in lamella angle increases with strain which indicates the development of inhomogeneous deformation due to the rolling. Thus, at a reduction of 70%, the angle is $\sim 10^\circ \pm 10^\circ$.

In order to make the detailed structure of T-M lamellae clear, they were imaged in dark-field (DF) mode. Figs 3a and b show DF images obtained from the spots indicated by (a) and (b) in Fig. 3c, respectively. In these DF images, very fine lamellae with thicknesses less than 100 nm are observed, and dark and bright areas in Fig. 3a are reversed in Fig. 3b. The diffraction pattern in Fig. 3c shows the superposition of two $\langle 011 \rangle$ diffraction patterns which are symmetrical to each other with respect to the $\{111\}$ plane, i.e., twin and matrix double spot pattern with the $\langle 011 \rangle$ incidence. The volume fraction of twin layers seems to be roughly equal to that of the matrix, which can be understood from the equivalency of their spot intensities. This indicates that the lamellar structure consists of alternate stacks of twin and matrix (T-M) thin layers.

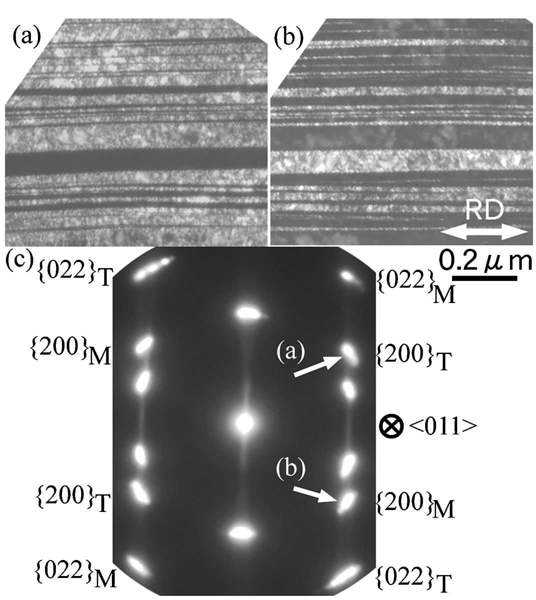

Figure 3 (a) and (b) are DF images of (T-M) lamellae observed in the longitudinal section at 70% reduction, (c) selected area diffraction (SAD) pattern obtained from the T-M lamellae. White arrows indicate spots which were used to obtain the DF images.

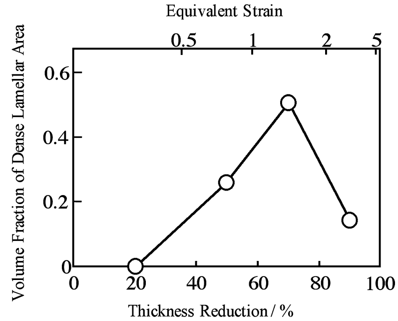

Figure 4 The volume fraction of T-M lamellar structure as a function of strain.

Fig. 4 shows the volume fraction of dense T-M lamellae at 20, 50, 70 and 90% thickness reduction. Although deformation twinning had been initiated at 20% reduction, a prominent lamellar structure was not seen at this rolling strain. In Fig. 4, therefore, the volume fraction of T-M lamellae at 20% reduction was taken to be zero. The volume of dense T-M lamellae increases with increasing strain up to 70%, then markedly decreases again at 90% thickness reduction. This indicates that the T-M structure is being destroyed by another microstructure after about 70% thickness reduction. The new microstructure consists of shear bands. At moderate strain (nearly 50% reduction) small shear bands are initiated, and the number and the width of these bands increase with strain breaking up the T-M lamellar structure to form a fine-grained structure.

When a shear band forms the intense local shear deformation is accompanied by a characteristic microstructure in the area of the shear band. Fig. 5a shows an enlarged BF image of a shear band. In this figure, the shear band is running from the bottom left to the top right. The lamellar structure can still be seen in the top left and bottom right in the figure. In the inner part of the shear band, substructures like an aggregation of fine grains are observed, and the shape of each fine grain seems to be somewhat elongated along the extending direction of the shear band. Note that the size of these elongated fine grains seems be larger than the lamella spacing. Considering that shear bands appear in the T-M lamellar structure, this suggests that collapse of the lamellar structure and a type of grain growth occurs during the intense shear deformation, the detail of which will be described later.

Fig. 5b shows the typical image of a selected area diffraction (SAD) pattern obtained from a shear band. It is to be noted that arcs of Debye rings are observed, although the pattern basically coincides with the twin-matrix double spot pattern with the $\langle 011 \rangle$ incidence which was shown in Fig. 3c. This indicates that the shear band consists of highly misoriented small regions, and the misorientation occurs around the $\langle 011 \rangle$ axis. Such fine-grained structure is seen not only in the BF image in Fig. 5a, but also in a previous observation using dark-field images [6].

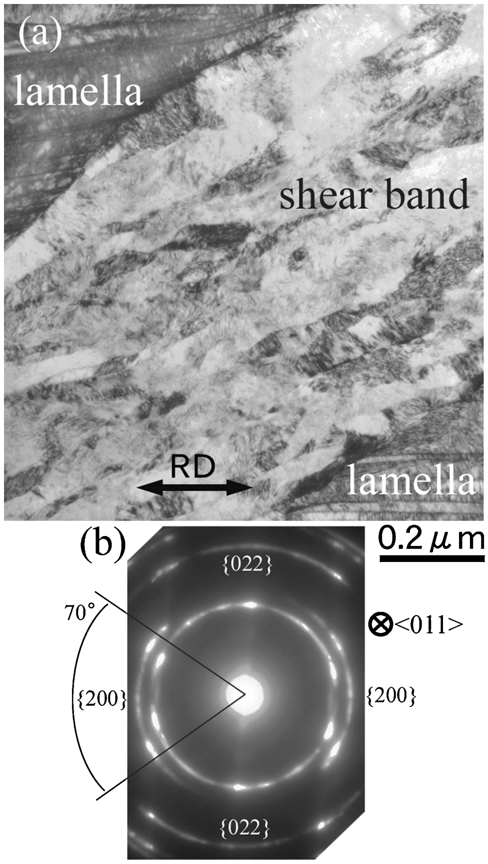

Figure 5 (a) BF image of a shear band formed in the T-M lamellae, (b) SAD pattern obtained from the foil which includes both the shear band and the adjacent T-M lamellae in (a).

In addition, there is another characteristic to be noted in the SAD pattern in Fig. 5b. When comparing the {200} and {220} arcs in Fig. 5b with the corresponding the {200} and {220} spots in Fig. 3c, it is found that the Debye arcs appear mainly in the range of 70° between the matrix spots and the twin spots, e.g., the 70° range between the {200} spot of the matrix ({200}M) and that of the twin({200}T). This indicates that the crystal orientation around the ⟨011⟩ axis in the twinning region rotates clockwise, while in the matrix anti-clockwise. This characteristic found in the SAD pattern is essential to understand the formation mechanism of fine-grained structure in shear bands, which is explained by the model illustrated in Fig. 6.

As shown in Fig. 2, with increasing the rolling strain, T-M lamellae become nearly parallel to the RD. Fig. 6a schematically shows this lamella configuration parallel to the RD, and slip systems on the {111} planes crossing both twin and matrix lamellae. Such slip deformation causes crystal rotation as shown in Fig. 6b: in this case the crystal rotation in the twinned region is clockwise, while in the matrix it is anti-clockwise. This crystal rotation causes the characteristic scatter of crystal orientations observed in the SAD pattern shown in Fig. 5b. Also, opposite crystal rotation in the matrix and twinned layers gives rise to the accumulation of geometrically necessary dislocations (GNDs) as an array at the boundaries. Fig. 6b also illustrates dislocation behavior when the T-M lamellar structure is subjected to a compressive stress, where b1 and b2 are dislocations with the Burgers vectors of a/2⟨011⟩ twin and a/2⟨110⟩ matrix, respectively. In this illustration, the dislocation with the Burgers vector b1 + b2 is the boundary dislocation which should be formed as an array by usual slip dislocations in the twinned and matrix region. Among three slip systems on the slip plane (in both twin and matrix), the slip system whose slip direction is normal to the plane of the diagram in Fig. 6b will not be activated since its Schmid factor is equal to zero, which indicates that the remaining two slip systems with the same Schmid factor will be activated. This means that there will be four families of dislocations accumulating at the T-M lamella boundaries, i.e., the dislocation reaction at the boundary may be more complicated than the illustration in Fig. 6b. However, the activation of the two slip systems with the same Schmid factor will be equivalent causing the mutual cancellation of the screw components of the activated dislocations. Thus, the boundary dislocations can be considered to be almost edge dislocations as illustrated in Fig. 6b.

With increasing compressive strain, the density of boundary dislocations will increase to make the tilt angle larger, which transforms a stable twin boundary into a

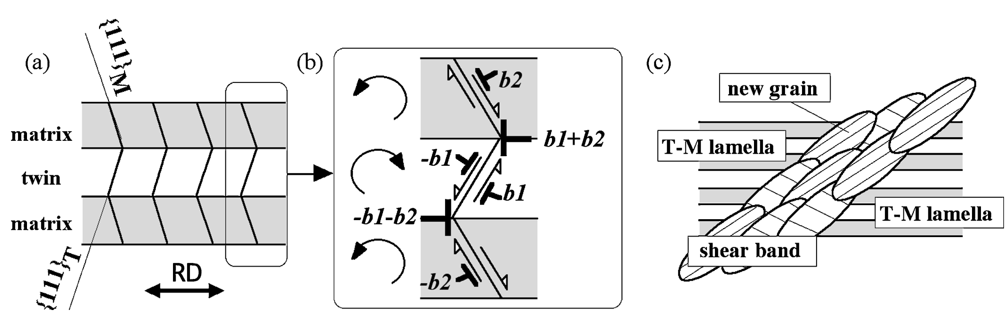

Figure 6 Schematic illustrations of shear band formation in (T-M) lamellae: (a) T-M lamellae almost parallel to rolling direction, (b) dislocation accumulation in the T-M lamellae, (c) new grains formed from the T-M lamellae.

high-energy boundary. In order to release the large accumulated strain energy, mutual annihilation of neighboring boundaries with opposite sense, may occur locally resulting in the formation of new grains as shown in Fig. 6c. In these processes, if the plastic strain due to the slip crossing the lamellae varies locally, the orientations of "new grains" must distribute among those of the twin and matrix in the original lamellae. In fact, such a distribution of orientation in the shear bands is observed in the diffraction patterns of Fig. 5b.

Finally, it should be emphasized that deformation twinning has an essential role in the process of grain refinement in low SFE fcc 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.

4. Conclusions

The evolution of a dense lamellar structure due to deformation twinning and the subsequent formation of a fine-grained structure in cold-rolled low SFE fcc metals was investigated by TEM using 310S austenitic stainless steels. The key process for grain refinement is in the development of a fine twin-matrix (T-M) lamellar structure and the subsequent process of the transmutation of T-M lamellae into a fine-grained structure by shear band formation. The shear bands destroy the T-M structure completely and new grains are formed. The T-M lamellar structure has an essential role for grain refinement, since the spacing of neighboring twin boundaries is very small and the misorientation is large enough to accumulate a high density of dislocations. This contributes to effective formation of new grains without huge strain.

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