1. Introduction

Dynamic transformation in steels is one of the promising means to reduce ferrite grain size down to 1 μm.1–9) The dynamic transformation is known to occur when a specimen is deformed just above the ferrite transformation start temperature (Ar3). To understand the underlying mechanism of grain-refinement through dynamic transformation, hot-deformed microstructures of austenite have been intensively examined by several researchers,10–13) because it is closely related to final ferrite microstructures.

The hot-deformed microstructure of austenite has been studied using a model alloy such as Ni–30mass%Fe 10–13) to avoid the difficulty in observing austenite at room temperature due to inevitable martensitic transformation during cooling. Stacking fault energy of the Ni–Fe alloy is around 75 mJ/m²,14) which is close to that of austenite in low carbon steels, so that deformation characteristics including dislocation substructures of the alloy are expected to be similar to that of low carbon steels. It was found11) that microbands were introduced in the place of dislocation cells by lowering deformation temperature. Misorientation of microbands is slightly large relative to that of dislocation cells. In addition, according to Beynon et al.,12) largely misoriented region was formed at the intersection of microbands. By comparing the process window of microbands and 1 μm ferrite, Hodgson et al.,10) Adachi et al.,11) and Beynon et al.12) have suggested that microbands with characteristics above-mentioned operate as nucleation sites for ferrite and are at least partly responsible for the grain refinement through dynamic transformation.

On the other hand, Suh et al.13) showed that grain boundaries in fcc-matrix phase in the Ni–Fe alloy were serrated at a relatively early stage of hot deformation, and subsequently region near grain boundaries was dynamically recrystallised into fine equiaxed grains beyond a critical strain. This dynamic recrystallisation is considered to occur through continuous dynamic recrystallisation instead of discontinuous dynamic recrystallisation. The serrated or fine grain-bearing grain boundaries provide a chance for the formation of multiple intergranular ferrite variants, which results in retardation of ferrite grain growth through coalescence between neighboring grains with similar orientation. Actually, Torizuka et al.15) observed that at serrated austenite grain boundaries, various variants of intergranular ferrite were formed. Based on above findings, the grain-refinement through dynamic transformation is considered to occur due to the following three reasons. (i) Microbands are densely introduced into austenite by deformation at relatively low temperature and operate as effective nucleation sites for ferrite. As a result combined with large supercooling, nucleation rate of ferrite is increased. (ii) Ferrite grains transformed partially from austenite are soon impinged by others, and further grain growth of ferrite ceases due to this impingement. (iii) The grain boundary serration or continuously recrystallised fine grains at grain boundaries might partly be responsible for ferrite grain refinement by preventing neighboring intergranular ferrite grains from coalescence.

However, most of the previous studies concentrated on ‘static’ transformation at microbands or grain boundaries, leaving behind the physical meaning of ‘dynamic’ transformation. One of the features of dynamic transformation is a transformation during deformation (Fig. 1). Deformation is introduced into austenite both before and after the onset of transformation (Fig. 1(b)). To understand the meaning of
‘dynamic’ transformation, effects of the pre-/post-deformations on transformed microstructures of ferrite should be clarified. It has been reported\textsuperscript{16} that orientation of dynamically transformed intergranular ferrite is distributed more widely and then the orientation distribution prevents neighboring grains with the same orientation from coalescence. Therefore, the present study examines effects of pre-/post-deformations on the orientation distribution of intergranular ferrite to deepen our insight into the mechanism of dynamic transformation. The pre-deformation here is less than 20% in reduction that does not cause grain boundary serration. A Ni–43mass%Cr alloy is used as a model alloy to enable simultaneous observation of fcc-matrix and bcc-product phase at room temperature. An intergranular bcc-phase and fcc-matrix phase in the Ni–Cr alloy are characterized using orientation information obtained by electron backscattered diffraction (EBSD). In particular, a change of the deviation angle from the Kurdjumov–Sachs orientation relationship (KS O.R.)\textsuperscript{11} of intergranular bcc-phase is investigated with pre-/post-deformations.

2. Experimental

2.1. Material and Thermomechanical Treatment

A 57Ni–43Cr (mass%) alloy was used as a model alloy in the present study. The material was processed by hot rolling, followed by annealing. The hot rolling results in equiaxed matrix grains in 5 mm sheets. The alloy was heated up to 1200°C at heating rate of 10 K/s using an infrared heating furnace and then isothermally solution-treated at 1200°C for 1 h, followed by water quenching to obtain a Cr-supersaturated fcc-matrix. Then it was aged at 1000°C for 2 h to stimulate bcc-Cr precipitation (hereafter, undeformed specimen). To examine the effect of pre-deformation (Fig. 2(b)), the solution-treated specimen was lightly cold rolled to thickness of 4 mm from 5 mm (20% in reduction), which corresponds to the pre-deformation. Subsequently the specimen was isothermally aged at 1000°C for 2 h to stimulate bcc-Cr precipitation (hereafter, pre-deformed specimen). To investigate the effect of post-deformation (Fig. 2(c)), firstly the solution-treated specimens were aged at 1000°C for 2 h to stimulate bcc-Cr precipitation. Then specimens were deformed up to 20% by tension (or partly cold rolling) at room temperature (hereafter, post-deformed specimen).

2.2. Specimen Preparation and Observation

Microstructural observation was made by optical microscopy (OM), transmission electron microscopy (TEM) and scanning electron microscopy (SEM) in combination with EBSD measurement. The EBSD measurement was conducted using a field emission gun-equipped SEM (FESEM) with TexSEM Laboratories OIM attachment. Interval of EBSD measurement points was set at 50 nm. For the pre-deformed specimens, EBSD measurement was conducted on the transverse sections of rolled specimens, whereas for the post-deformed specimens, on the normal plane of tensile specimens. The orientation of precipitates and matrix phase was obtained from the center region of precipitates and the region about 200 nm distant from interphase boundaries, respectively. Specimens for EBSD were mechnochemically polished with colloidal silica as an abrasive compound to realize the specimen surface as smooth as possible. The specimens were polished before post-deformation, and never polished again after post-deformation. Specimens for OM and SEM observation were electrochemically polished and etched with a solution of 6vol% perchloric acid, 35vol% butanol and 59vol% methanol at room temperature at 30 V DC. Thin foils for TEM were prepared by the conventional twinjet method using the aforementioned solution at 243 K at 80 V DC.

There are two possible orientation relationships between bcc and fcc phases; the KS O.R.\textsuperscript{17} (\{111\}_fcc/\{110\}_bcc and \[110\]_fcc/\[111\]_bcc) and the Nishiyama–Wasserman (NW) O.R.\textsuperscript{18,19} (\{111\}_fcc/\{110\}_bcc and \[\bar{1}2\bar{1}\]_fcc/\[\bar{1}10\]_bcc). Deviation angles from the KS O.R. and the NW O.R. of precipitates and matrices were calculated using orientation information by EBSD. The detailed procedure of the calculation was mentioned elsewhere.\textsuperscript{21,22} According to the accompanying experiments,\textsuperscript{21,22} the deviations of intergranular bcc-precipitates from the exact \{111\}_fcc/\{110\}_bcc and \[110\]_fcc/\[111\]_bcc were found less than 6 degrees, while from the exact \[\bar{1}2\bar{1}\]_fcc/\[\bar{1}10\]_bcc, often more than 6 degrees. This result suggests that intergranular precipitates examined tend to be related with respect to the adjacent matrix.
grains in the KS O.R. rather than the NW O.R., but the plane parallel orientation relationship and the direction parallel orientation relationship deviated from the exact \( \{111\}_{\text{fcc}}//\{110\}_{\text{bcc}} \) and \( [110]_{\text{fcc}}//[111]_{\text{bcc}} \) to a certain extent. Based on this result, the orientation relationship was defined as the KS O.R. when both the plane/direction parallel orientation relationships were satisfied within 6 degree deviation. The measurement error in orientation by EBSD system used in the present study was less than 0.819 degree according to the pre-examination.\(^{20}\) Sixty, seventeen and eleven precipitates in undeformed, pre-deformed and post-deformed specimens, respectively, were subjected to the orientation examination. Those were formed at thirty five, nine and five grain boundaries, respectively.

3. Results and Discussion

3.1. Effects of Pre-/post-deformations on Microstructure

Optical microstructures of (a) undeformed, (b) pre-deformed and (c) post-deformed specimens are shown in Fig. 3. In all specimens here, precipitates are found to form at grain boundaries of a matrix phase. No difference in the microstructure of intergranular precipitates is recognized from (a) though (c) at least at this magnification. In the pre-deformed specimens (Fig. 3(b)), precipitates are also observed in interior of the matrix phase. According to accompanying study,\(^{23}\) the transgranular precipitates are formed at microbands and annealing twin boundaries.

3.2. Orientation Analysis for Pre-deformed Specimens

Optical micrographs of (a) undeformed, (b) pre-deformed and (c) post-deformed specimens. Average standard deviations of the deviation angles are not remarkably changed by 20% pre-deformation (from 1.1/1.4 to 1.4/1.4 degrees).

3.3. Orientation Analysis for Post-deformed Specimens

3.3.1. Line origin-to-point misorientation profiles across grain boundaries

Line origin-to-point misorientation profiles across grain boundaries (Fig. 7(a)) clearly indicate that the maximum local misorientation at the vicinity of grain boundaries increases with increasing post-deformation degree (arrowed in Fig. 7(a)). The local misorientation is found to reach up to 15 degrees and to be present at the vicinity of both the coherent (in matrix A in Fig. 7(a)) and incoherent interphase boundaries (in matrix B) of intergranular precipitates. The orientation relationship between intergranular precipitates and adjacent matrix grains is deviated from the KS O.R. by the post-deformation, because the adjacent matrix lattice is locally rotated crystallographically. As a result, deviation angles \( \Delta \theta_{0.10.20} \) of intergranular precipitates from the plane/direction parallel orientation relationships in the KS O.R. increase with increasing the post-deformation degree as shown in Fig. 7 and Fig. 8, where the subscripts of \( \theta \) mean post-deformation degree. The deviation angle from the plane parallel orientation relationship is increased aver-

* The \( \{111\}_{\text{fcc}} \) closest to the underlying grain boundary plane is often selected as a low-energy interface by the major variant of KS-related GB-precipitates. However the closest \( \{111\}_{\text{fcc}} \) is not selected by variants in the following two exceptional cases. The first case is when the closest \( \{111\}_{\text{fcc}} \) is largely tilted from the grain boundary plane. The other case is when GB-precipitates are KS-related with respect to both adjacent matrix grains\(^{20,21}\). (Such orientation relationship is called ‘KS-KS O.R.’ in the present study.)
agely from 1.2 up to 5.9 degrees with increasing post-deformation degree, whereas that from the direction parallel orientation relationship is not considerably changed (from 2.1 up to 3.4 degrees) by post-deformation. Noteworthy is that the deviation angle in the post-deformed specimens is much larger than that in the pre-deformed specimens. The deviation angles in pre-deformed and post-deformed specimens are compared in detail in Sec. 3.4.

One may notice that the deviation angles of the precipitate, P4a in Fig. 7(b) increase with increasing the post-deformation degree, even though matrix lattice rotation along the direction perpendicular to the grain boundary trace is small. To clarify the reason for this contradiction, line misorientation profile along GB2 is also measured (Fig. 9). In contrast to the direction perpendicular to the grain boundary trace, fluctuation in misorientation is observed along GB2 in matrix A and its magnitude reaches

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**Fig. 4.** Orientation image mappings of (a) undeformed and (b) pre-deformed specimens. Misorientation angles (MO) and deviation angles from the plane/direction parallel orientation relationships are presented. An adjacent matrix gain to which intergranular precipitates are KS-related is indicated with a short line. Transverse section of rolled specimens (b).

**Fig. 5.** Orientation image mappings of (a) undeformed, (b) 10% post-deformed and (c) 20% post-deformed specimens. All images here are obtained from the same area on the normal plane of the tensile specimen. Misorientation angles (MO) and deviation angles from the plane/direction parallel orientation relationships are presented. An adjacent matrix gain to which intergranular precipitates are KS-related is indicated with a short line.

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**Fig. 6.** Schematic representation of local lattice rotation in a matrix due to geometrically necessary dislocations.
up to 7 degrees. The fluctuation corresponds to the lattice rotation near GB2 in Fig. 6 and is probably responsible for the deviation of P4a from the KS O.R. To interpret the dependence of line misorientation analysis direction, the slip system is investigated (Fig. 10(a)). Traces of {111}fcc and {110}fcc in the observed plane are exhibited using reference lines in Fig. 10(a), and their tilt angles from the observed plane are indicated with the length of the reference lines. One of the slip systems in matrix A (marked with *) is found almost parallel to that in matrix C, while others in matrix A do not correspond to any slip systems in matrix C. Dislocations probably transfer easily in the slip system marked with * from matrix A to matrix C as schematically shown in Fig. 10(b). In this case, dislocations may not be piled-up at grain boundary and dislocation distribution in the slip system is considered homogeneous. Consequently, geometrically necessary dislocations causing lattice rota-
tion are hardly generated. On the other hand, dislocations are likely to be piled-up at grain boundaries, generating lattice rotation in the case when there is no continuity of slip system availability between the adjacent matrix grains. The above analysis suggests that the local misorientation near grain boundaries depends on line analysis direction because continuity of slip systems between the adjacent matrix grains depends on the direction.

As revealed in Fig. 11, the deviation angle from the KS O.R. is increased by post-deformation (deformation degree here is 20%), but it is varied depending on intergranular precipitates formed at the same grain boundary, GB2. The standard deviation ($\sigma_{0.20}$) in the deviation angles are increased from 0.43/098 (at GB2, but average $\sigma_r$ for all intergranular precipitates examined in undeformed specimens is 1.1/1.4 degrees) to 3.7/1.0 degrees by 20% post-deformation. As shown in Fig. 12, the orientation normal to observed plane of matrix along GB2 is distributed more widely by post-deformation. In addition, Fig. 12 suggests that the orientation of intergranular precipitates along GB2 is also scattered more widely by post-deformation. These results indicate that heterogeneous local lattice rotation of matrix is a main reason of the variation of the deviation angles from the KS O.R., but heterogeneous local lattice rotation of intergranular precipitates is one of the additional reasons.

3.4. Comparison of Deviation Angles from the KS O.R.

Comparison of deviation angles of intergranular precipitates from the KS O.R. is made among undeformed, pre-deformed and post-deformed specimens (Fig. 13). In Fig. 13, average deviations ($\Delta \theta_{KS}$), and standard deviations ($\sigma$) of the deviation angles observed in the present study are also exhibited. The average (maximum) deviation angles from the plane/direction parallel orientation relationships are 1.2/2.1 (5.4/5.1), 1.7/2.8 (5.8/5.3) and 5.9/3.4 (12.2/4.7) degrees in undeformed, 20% pre-deformed and 20% post-deformed specimens, respectively. Corresponding standard deviation of the deviation angles are 1.1/1.4, 1.4/1.4 and 3.7/1.0 degrees. It is found that the deviation angle from the plane parallel orientation relationship becomes largest in the post-deformed specimens, but the deviation angle from the direction parallel orientation relationship is not noticeably different among specimens. Figure 13 indicates that intergranular precipitates rationaly related tend to loose the coherency with respect to the adjacent matrix grains, most noticeably by post-deformation. The degree of coherent-to-incoherent transition of intergranular precipitates is different depending precipitates even formed at the same grain boundaries as shown in Fig. 11. The pre-deformation (Fig. 13(b)) is also found to be effective in increasing the deviation angle compared to the undeformed case (a), though it is less effective than the post-deformation. The reason of the pre-deformation-induced deviation is still not apparent, but geometrically necessary dislocations introduced by the pre-deformation are supposed to affect the balance of related boundary energies and then the equilibrium shape of in-

![Fig. 11. An effect of post-deformation on the deviation angles from the plane/direction parallel orientation relationships of intergranular precipitates formed at the same grain boundary.](image)

![Fig. 12. Variation in orientation normal to observed plane of intergranular precipitates along grain boundary 2 (GB 2) and adjacent matrix with post-deformation. (a) Undeformed, (b) 10% post-deformed, (c) 20% post-deformed.](image)

![Fig. 13. Comparison in the distribution of the deviation angles from the plane/direction parallel orientation relationships.](image)
tergranular precipitates. Furuhara et al. showed that the orientation of intergranular ε-precipitates (hcp) in a slightly pre-deformed β-matrix (bcc) in Ti–15V–3Cr–3Sn–3Al alloy was different from that in an undeformed matrix. They considered that the stress field of dislocations in the slip bands intersecting with a grain boundary preferred the specific variants of ε-precipitates.

3.5. Role of Pre-/post-deformations on the Orientation Distribution of Intergranular Product Phase Dynamically Transformed

Above-mentioned results reveal that both pre- and post-deformations lead to an increase of deviation angles. In addition, particularly post-deformation increases standard deviation angles of intergranular precipitates from the KS O.R. Figs. 14(a), 14(b) and 14(c) represent a schematic illustration of microstructural evolution in undeformed, pre-deformed and post-deformed specimens, respectively. One may suppose that if the pre-deformation is introduced heterogeneously along a matrix grain boundary (Fig. 14(b-2)), an orientation of intergranular product phase that is formed at a strain-accumulated region becomes slightly different from that at other regions (Fig. 14(b-2)). The pre-deformation is considered to be fairly effective in distributing the orientation of intergranular product phase, and then in preventing the neighboring intergranular product phase with similar orientation from coalescence. This is attributed mainly to heterogeneous local matrix lattice rotation due to heterogeneous deformation along grain boundaries. Additionally, heterogeneous local lattice rotation of intergranular precipitates also affects the distribution. The post-deformation is, therefore, considered to be more effective in increasing the deviation angle of an intergranular product phase from the KS O.R. (Fig. 13) than the pre-deformation. Hence particularly post-deformation is assumed to play an important roll in distributing the orientation of intergranular ferrite in ‘dynamic’ austenite-to-ferrite transformation. Lee and co-workers have found that external stress during austenite-to-ferrite transformation also has the potential to distribute the orientation of intergranular ferrite. The combined effect of pre-/post-deformations and external stress on the orientation of intergranular ferrite may be a subject of the intense examination on dynamic transformation in a future. The effects of volume fraction of transformed ferrite, carbon content in retained austenite and deformation temperature on the strain distribution to ferrite and austenite are also subjects that should be elucidated.

4. Summary

To understand the features of ‘dynamic’ austenite-to-ferrite transformation in steels, effects of light deformations before and after precipitation (pre-deformation and post-deformation) on the orientation distribution of intergranular bcc-precipitates was examined in a Ni–43mass%Cr alloy by EBSD. The deviation angles of intergranular precipitates from the Kurdjumov–Sachs orientation relationship (KS O.R.) were compared among undeformed, pre-deformed and post-deformed specimens. Obtained results could be summarized as follows.

(1) Post-deformation was found to increase not only the deviation angles but also the standard deviation angles of the deviations from the plane/direction parallel orientation relationships in the KS O.R.

(2) Pre-deformation was also found to increase the deviation angles, though it was less effective than post-deformation.

(3) Based on above results, post-deformation is considered to be more effective in distributing the orientation of
intergranular ferrite than pre-deformation in dynamic austenite-to-ferrite transformation in steels.

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