Effect of initial grain size on inhomogeneous plastic deformation and twinning behavior in high manganese austenitic steel with a polycrystalline microstructure

R Ueji1,*, N Tsuchida2, K Harada3, K Takaki3 and H Fujii1
1 Joining and Welding Research Institute, Osaka University, 11-1 Mihogaoka, Ibaraki, Osaka 567-0047, Japan
2 Faculty of Engineering, University of Hyogo, 2167 Shosya, Himeji, Hyogo 671-2280, Japan
3 Faculty of Engineering, Kagawa University, 2217-20 Hayashi-cho, Takamatsu, Kagawa 567-0047, Japan

Email: *ueji@jwri.osaka-u.ac.jp

Abstract. The grain size effect on the deformation twinning in a high manganese austenitic steel which is so-called TWIP (twinning induced plastic deformation) steel was studied in order to understand how to control deformation twinning. The 31wt%Mn-3%Al-3%Si steel was cold rolled and annealed at various temperatures to obtain fully recrystallized structures with different mean grain sizes. These annealed sheets were examined by room temperature tensile tests at a strain rate of 10^{-4}/s. The coarse grained sample (grain size: 49.6 μm) showed many deformation twins and the deformation twinning was preferentially found in the grains in which the tensile axis is parallel near to [111]. On the other hand, the sample with finer grains (1.8 μm) had few grains with twinning even after the tensile deformation. The electron back scattering diffraction (EBSD) measurements clarified the relationship between the anisotropy of deformation twinning and that of inhomogeneous plastic deformation. Based on the EBSD analysis, the mechanism of the suppression of deformation twinning by grain refinement was discussed with the concept of the slip pattern competition between the slip system governed by a grain boundary and that activated by the macroscopic load.

1. Introduction
Austenite is a useful phase to manage both strength and elongation in steel [1, 2]. Among several kinds of typical austenitic steels, high manganese austenitic steel gathers much attention from materials scientists during the last decade [3-6]. Especially, recent research works focus on the high manganese steel with alloying of aluminum in order to prevent the dynamic phase transformation from austenite to ε phase within HCP structure. This alloy usually has a relatively low stacking fault energy of around 40 mJ/m² at ambient temperature and it is reported that deformation twinning occurs occasionally so that this steel was widely called a Twinning Induced Plasticity (TWIP) steel [3]. Previous papers reported that the deformation twinning preferentially occurs when the load direction is near parallel to [111] direction at the tensile deformation [6]. The deformation twinning provides new boundaries for the object against the dislocation slip, which provides a higher work hardening rate followed by a larger uniform elongation. It was also clarified that the adequate ductility of the TWIP steel keeps even at a
high strain rate deformation up to $10^2$/s [7, 8]. The excellent properties of TWIP steel seems to be attractive also for industrial applications [3].

By the way, grain refinement is one of the strengthening methods without any additional alloying. It is well-known that when the mean grain size becomes as small as a few micrometers in the BCC steels or single phase aluminum, the strength increases however, the elongation decreases [9]. On the other hand, for the 31%Mn-3%Al-3%Si TWIP steel, the grain refinement provides strengthening while keeping adequate elongation, although the deformation twinning is suppressed by the grain refinement. Ueji et al. [6] clarified the crystallographic tendency that [111] parallel to the tensile direction is preferable for the twinning even in a fine grained sample. However, the mechanism of the suppression of the twinning has not been clarified yet.

The relationship between initial grain size and deformation twinning in polycrystalline metal has been studied for example by Barnett et al.[10]. They studied the grain size effect in a Mg alloy and clarified the mechanism of the suppression by the difference between the Hall-Petch slope for the dislocation slip and that for the twinning. It is reasonable because deformation twinning in the magnesium alloy occurs at plastic yielding. However, the twinning in the TWIP steel generally appears after the leading plastic deformation by dislocation slip [11] as is found in other FCC metals [12, 13]. This fact implies the importance of the investigation on the change of dislocation structure by the grain refinement. Consequently, this study was conducted to examine the deformation induced substructure in the fully recrystallized TWIP steels with different grain sizes and to discuss the mechanism of the twinning suppression by the grain refinement.

2. Experimental Procedures

TWIP steel with a chemical composition of 31%Mn-3%Si-3%Al-Bal.Fe was used. The ingot was hot rolled at 1200 °C to obtain a plate with a thickness of 12mm. The hot rolled sheet was cold rolled to a reduction in thickness of 12 mm followed by annealing at 900 °C for 9.2 ks for homogenization. The heat treated sample was cold rolled to a reduction of 83% and subsequently annealed at 700 °C, 800 °C or 1000 °C for 1.8ks. These annealed samples have a fully recrystallized microstructures whose mean grain sizes were 1.8 μm, 7.2 μm or 49.6 μm, respectively. This thermomechanical treatment is just the same as performed in the previous work [6]. The tensile specimens with a gage part of 25mm length x 5mm width x 1.2mm thickness were cut from these annealed samples and tensile test were conducted at an ambient temperature at a strain rate of $10^{-4}$/s. The samples were strained to various strains and these were measured by the Electron Back Scattering Diffraction (EBSD) technique in a Scanning Electron Microscope (SEM). The EBSD data were analyzed mainly focused on local misorientation angles. The EBSD scan step size was 0.05 μm at all scans except for the measurements shown in Fig.4(c) and Fig.5(b) where the coarse grained sample were measured with a step size of 0.25 μm.

3. Results and discussion

Figure 1 shows the stress - strain curves of the annealed TWIP steels. The end points of the curves equal to the point where the load reached the maximum which corresponded to the starting point of the necking. All of the curves show a somehow liner morphology, which is typically found in some of the FCC metals with a low stacking fault energy. With decreasing mean grain size, the yield stress and the flow stress increase because of the grain refinement strengthening.

Figure 2 shows the gray scale maps for the distribution of Kernel Average
Misorientation (KAM) angles in the samples with initial grain sizes of 1.8μm (a, b, c) or 49.6μm (d, e, f) tensile tested to a strain of 10%, 20% or 30%. The KAM angle equals to the mean value of all the misorientation angles between the measurement position and the EBSD scan points which have a constant distance (KAM step) from the measurement position [14]. The main purpose for the KAM evaluation is the examination of substructure, so that the vicinity point belonging to the different grain from that of the measurement position is eliminated to estimate the KAM angle. In addition, the EBSD data from the position adjacent of the initial grain boundaries are not evaluated. In the gray scale map, the larger KAM angle are indicated by the whiter color.

In the case of the 1.8 μm sample, 10% tensile deformation provides a local increase of the KAM angle especially at the locations near the grain boundaries or the triple junctions of grain boundaries. It should be noted that no deformation twinning was found and that these substructures were due to the slip deformation of the dislocations. With increasing strain, the area with higher KAM angle extends more widely. On the other hand, a different morphology was observed in the coarse grained sample: as shown in the KAM maps of the 49.6μm sample, many bands having a larger KAM angle are additionally found also in the interior of the grains. These bands extend along different orientations and the intersection of these band produces a kind of cell structure. These cell structures show a different morphology in the different grains. When the coarse grained sample was tensile deformed to a strain of 30%, a lot of deformation twinning occurred in some of the grains although the twinning cannot be found in the fine grained samples. This result confirmed the suppression effect of the grain refinement on the deformation twinning which was reported already [6].

**Figure 2** Grey scale maps for the distribution of KAM angles in the 31%Mn-3%Al-3%Si TWIP steels with initial grain size, d of 1.8 μm (a, b, c) or 49.6 μm (d, e, f) tensile tested to a true strain, ε of 10% (a, d), 20% (b, e) or 30% (c, f). The EBSD measurements were conducted on the cross-sectional plane parallel to the normal direction (ND) and the rolling direction (RD).
The mean value of all the KAM angles in each scan of the samples is shown in Fig. 3. Although various kinds of the KAM steps, 0.05 μm (a), 0.15 μm (b) and 0.25 μm (c) were adopted, the basic tendency is not changed by the change of KAM step. When compared with the same sample with an identical initial grain size, the mean KAM value increases with increasing of the tensile strain. This result indicates that the magnitude of the KAM angle is correlated with the magnitude of the accumulation of the dislocations. In addition, with decreasing mean grain size, the increasing rate of the mean KAM angle becomes larger, indicating a rapid increase of dislocations which contributes the local misorientation. Concerning with the effect of the KAM step, when comparing the data with the same initial grain size and the same strain, the mean KAM angle becomes larger with increasing KAM step.

As mentioned in the introduction of this paper, deformation twinning show a strong orientation dependence. That is, the [111] is a more preferable tensile direction than [001] and this tendency is not changed even in the fine grained TWIP steel. Consequently, the orientation dependence of the KAM angle change is required in the study to link the dislocation structure to twinning if it exists. For the first trial, the EBSD scan data were separated between grains and the mean KAM angle of each grain were calculated to evaluate the substructure evolution related to the tensile orientations. Figure 4 shows inverse pole figures showing the crystallographic orientations parallel to the tensile direction of each grain. The grayscale of a data point indicates the mean KAM value of the corresponding grains. Concerning the distribution of the tensile directions, the sample with smaller grains has randomly distributed orientations; whereas in the coarse grained sample the tensile direction preferentially distributes along the oblique line indicating the orientation from [001] to [111]. This change of the texture evolution by grain refinement can be confirmed by the texture plots as inverse pole figures shown in Fig. 5. As is found in a previous paper [6], the initial textures showed no strong orientation concentration and the maximum intensity decreased with the progress of grain growth. This tendency is opposite to what was found in this work. This result clearly indicates the difference of the lattice rotation between the coarse and fine grained samples.

Concerning the mean KAM angles, it is difficult to find the orientation dependence although these mean KAM angles were evaluated with the larger KAM step (0.25 μm) which seems to be better to show the change of substructure evolution as shown in Fig. 3. More careful observation clarified that in the 7.2 μm sample (b), there are some concentrations of the points with relatively larger mean KAM angles at the orientations from [111] to [112]. This tendency seems to be related with the previous study reported especially by the research group in Denmark [15-18]. They studied the substructure evolution...
During the tensile deformation of FCC metals with medium to high stacking fault energies and they found that the dislocation density related to local misorientation preferentially increases when the tensile direction is closely parallel to [111]. However, a more clear orientation dependence of the mean KAM angle cannot be found in the other samples. This result indicates that conditions for the evolution of the mean KAM angle are needed to clarify the orientation dependence.

As shown in Fig. 3, with increasing of the tensile strain and/or the KAM step, the mean KAM angle tends to increase. The larger value of the KAM angle should be preferable to detect the change of the orientation dependence. Figure 6 shows the mean KAM angles evaluated with a relatively large KAM step which does not exceed both the grain size and the cell size in the samples tensile strained to various strains. The samples with a mean grain size of 1.8μm (a) or 49.6μm (b) were examined with a KAM step of 0.25μm and illustrated by the grey scale.

**Figure 4** Inverse pole figures showing the mean crystallographic orientations parallel to the tensile axis of the grains in the 31%Mn-3%Al-3%Si TWIP steels with various mean grain sizes, d tensile strained to a strain, ε of 10%. The mean KAM angle was calculated with KAM step of 0.25μm and illustrated by the grey scale.

**Figure 5** Inverse pole figures showing the tensile axis orientations in the 31%Mn-3%Al-3%Si TWIP steels with various mean grain sizes, d tensile strained to a strain, ε of 10%.

during the tensile deformation of FCC metals with medium to high stacking fault energies and they found that the dislocation density related to local misorientation preferentially increases when the tensile direction is closely parallel to [111]. However, a more clear orientation dependence of the mean KAM angle cannot be found in the other samples. This result indicates that conditions for the evolution of the mean KAM angle are needed to clarify the orientation dependence.

As shown in Fig. 3, with increasing of the tensile strain and/or the KAM step, the mean KAM angle tends to increase. The larger value of the KAM angle should be preferable to detect the change of the orientation dependence. Figure 6 shows the mean KAM angles evaluated with a relatively large KAM step which does not exceed both the grain size and the cell size in the samples tensile strained to various strains. The samples with a mean grain size of 1.8μm (a) or 49.6μm (b) were examined with a KAM step of 0.5μm or 2.5μm, respectively. In order to clarify the orientation dependence, the EBSD scan data were separated in tensile orientation was close to the [111] direction or not. The threshold angle is set as 30° and the orientation whose deviation angle less than 30° is almost the same as the orientation with whom the grains have twinning preferentially when the initial grain size is coarse. When the tensile strain is limited to 10%, it is difficult to find the orientation dependence of the mean KAM value in both the samples as is similar to be found in Fig. 4. However, when the samples were deformed more than 20%, the coarse grained sample shows the orientation dependence, that is, the mean KAM value becomes smaller when the orientation is near to [111]. On the other hand, the fine grained sample still has no clear orientation dependence. Although this orientation dependence is not in agreement with the previous finding, the result at least supports the disappearance of the orientation dependence of the substructure evolution by the grain refinement to a grain size of about 1 μm.
By means of the experimental works in this study, three significant characteristics of the grain size effect on the substructure were found:

(1) When the sample has a coarse grained structure, the KAM angle becomes preferentially large both near the grain boundary and at the interior of the grain; whereas, in the fine grained sample, only the grain boundary preferentially accelerates the increase of the KAM angle. (Fig.2)

(2) The rate of increase of the mean KAM angle becomes larger with decreasing of the initial grain size. (Fig.3)

(3) The orientation dependence of the mean KAM angle can be found only in the coarse grained sample. (Fig.6)

These behaviors can be explained by the difference between the slip pattern at the area near the grain boundaries and that at the grain interior. The term, slip pattern means the combination of the active slip systems and this is related to the substructure development [15-18]. When the grain size is large enough, the slip pattern is determined by the macroscopic load axis and as a result the development of the mean KAM angle shows an orientation dependence. At the area near the grain boundaries, the constraint effect on the plastic deformation [19] is supposed to be accelerated and provide the different slip pattern which includes activation of additional slip systems. In this case, the KAM angle can become larger. When the initial grain size becomes smaller, the slip pattern governed by the grain boundaries becomes dominant and weakens the orientation dependence of the KAM angle. In other words, a slip pattern competition changes with the grain refinement.

The concept of the slip pattern competition between the slip system governed by grain boundary and that activated by the macroscopic load can explain the suppression of twinning by grain refinement. As mentioned in the introduction of this paper, deformation twinning in FCC metals should need some additional force from the dislocation structure which evolves by the slip deformation before the twinning. Actually, Narita et al. [12, 13] proposed a quantitative model to estimate the additional stress and Karaman et al. [11] modified their theory to explain the deformation twinning behavior in single crystal high manganese steel. In these studies, the active slip systems can be predicted with the orientation of deformation axis. That is to say, the slip pattern activated by the macroscopic load is needed for the deformation twinning. The grain refinement extends the area where the slip system are governed by grain boundaries and suppresses the slip systems activated by the macroscopic load. Consequently, deformation twinning should be prohibited by the grain refinement.

For a further study in order to clarify the effect of slip pattern competition quantitatively, the numerical information of the development of the mean KAM value as shown in Fig.6 should be useful. However, the reason why the mean KAM angle increases rapidly in the grains whose [111] orientation is near to the tensile axis has not been clarified yet, although the previous work [15-18] supported the
opposite tendency. This seems to be due to the difference of the materials with different stacking fault energy or the difference of the mythology of microstructural characterization, which is EBSD/SEM or transmission electron microscopy (TEM). However, the FCC metals with a low stacking fault energy hardly have the conventional dislocation cell structure, so that it is difficult to consider the critical method to clarify the substructure development in TWIP steel and further study should be needed.

4. Summary
The mechanism of the suppression effect of the grain refinement on the deformation twinning was proposed based on the EBSD analysis of the grain size dependence of the substructure formed by dislocation slip in the 31%Mn-35Al-3%Si TWIP steel. The summary can be drawn as following:

(1) When the grain size is large enough, the evolution of an inhomogeneous substructure which can be detected by KAM angle changes whether the observation location is the grain interior or the vicinity of the grain boundaries. Near the grain boundary, the substructure is formed mainly by the slip pattern which can easily produce higher KAM angles. On the other hand, at the interior of the grain, the slip pattern should be governed by the macroscopic load direction, so that the evolution of the KAM value shows an orientation dependence when the grain size is large enough.

(2) The grain refinement provides the wider area where the slip pattern controlled by grain boundaries operates and spontaneously the slip pattern governed by the macroscopic load direction is weakened.

(3) The mechanism of the twinning suppression by the grain refinement can be explained by the competition between the slip pattern governed by the macroscopic load direction and that controlled by the grain boundaries, because the deformation twinning needs an additional stress which is due to substructure produced by the slip pattern related to the macroscopic load direction.

Acknowledgements
The authors would like to thank Prof. N. Tsuji in Kyoto University, Japan for useful discussion. This work was supported by a Grant-in-Aid for Scientific Research on Innovative Area “Bulk Nanostructured Metals” (ID: 22102002) from the Japan Society for the Promotion of Science (JSPS) and the Japan Science and Technology Agency (JST) under Collaborative Research Based on the Industrial Demand, "Heterogeneous Structure Control: Towards Innovative Development of Metallic Structural Materials".

References
[1] Angle J 1954 J. Iron Steel Inst., 177 pp 165–174
[2] Adler P H, Olson G B and Owen W S 1986 Metall. Mater. Trans. A 17A pp 1725–1737
[3] Bouaziz O, Allain S, Scott C P, Cugy P and Barbier D 2011 Curr. Opin. Solid State Mater. Sci. 15 pp 141–168
[4] Park K-T, Jin K G, Han S H, Hwang S W, Choi K and Lee C S 2010 Mater. Sci. Eng. A 527 pp 3651–3661
[5] Rahman K M, Vorontsov V A and Dye D 2015 Acta Mater., 89 pp 247–257
[6] Ueji R, Tsuchida N, Terada D, Tsuji N, Tanaka Y, Takemura A and Kunishige K 2008 Scr. Mater., 59 pp 963–966
[7] Ueji R, Harada K, Tsuchida N and Kunishige K 2007 Mater. Sci. Forum 561-565 pp 107-110
[8] Ueji R, Harada K, Takemura A and Kunishige K 2008 Mater. Sci. Forum 584-586 pp 673-678
[9] Tsuji N, Ito Y, Saito Y and Minamino Y 2002 Scripta Mater., 47 pp 893-899
[10] Barnett M R, Keshavarz Z, Beer A G and Atwell D 2004 Acta Mater. 52 pp 5093-5174
[11] Karaman I, Sehitoglu H, Gall K, Chumlyakov Y I and Maier J 2000 Acta Mater., 48 pp 1345-1359
[12] Narita N and Takamura J 1992 Dislocations in Solids 9 pp 135-189
[13] Miura S, Takamura J and Narita N 1968 Trans. JIM 9 pp 555-561
[14] Choi J Y, Ji J H, Hwang S W and Park K -T 2012 Mater. Sci. Eng. A 535 pp 32-39
[15] Hansen N and Huang X 1998 Acta Mater. 46 pp 1827-1836
[16] Hansen N, Huang X and Hughes D A 2001 Mater. Sci. Eng. A, 317 pp 3-11
[17] Huang X and Winther G 2007 Phil. Mag. 87 pp 5189-5214
[18] Winther G and Huang X. 2007 Phil. Mag. 87 pp 5215-5235.
[19] Higashida K, Takamura J and Narita N 1986 Mater. Sci. Eng. 81 pp 239-258