Deformation Twinning Behavior of Twinning-induced Plasticity Steels with Different Carbon Concentrations – Part 1: Atomic Force Microscopy and Electron Backscatter Diffraction Measurements

Motomichi KOYAMA,1,2)* Takahiro SAWAGUCHI1) and Kaneaki TSUZAKI1,2)

1) National Institute for Materials Science, 1-2-1 Sengen, Tsukuba, Ibaraki, 305-0047 Japan.
2) Department of Mechanical Engineering, Kyushu University, 744 Motoooka, Nishi-ku, Fukuoka, 819-0395 Japan.

Received on February 9, 2015; accepted on March 31, 2015; originally published in Tetsu-to-Hagané, Vol. 100, 2014, No. 10, pp. 1246–1252

The deformation twinning behavior in Fe–17Mn–0.6C, Fe–17Mn–0.8C, and Fe–18Mn–1.2C (wt.%) twinning-induced plasticity (TWIP) steels was investigated by atomic force microscopy (AFM) and electron backscatter diffraction pattern (EBSD) analyses. AFM-based surface relief analysis combined with EBSD measurements was employed to determine the active twinning direction and deformation twin fraction for specific crystallographic orientations. The addition of carbon is known to increase the stacking fault energy. The <111> tensile orientation grains revealed suppression of deformation twinning with increasing carbon concentration; however, the deformation twin fraction in the <144> tensile orientation did not change as a function of the carbon concentration. These results imply that another factor in addition to stacking-fault-energy-based criteria is required to interpret the deformation twinning behavior of carbon-added TWIP steels.

KEY WORDS: surface relief analysis; deformation twinning; austenitic steel; atomic force microscopy (AFM); electron backscatter diffraction (EBSD).

1. Introduction

The work hardening capacity is a major factor in the determination of uniform elongation as long as premature fracture does not occur because an increase in the work hardening rate suppresses local necking. Because the improvement of uniform elongation plays important roles in the enhancement of the shock absorption capacity and workability, clarifying the factors affecting the work hardening behavior are crucial for structural materials such as automobile parts. From the microstructural viewpoint, the primary factor is the increase in the dislocation density and its associated dislocation substructure evolution. As characteristic work hardening factors in austenitic steels, martensitic transformation and deformation twinning also markedly enhance the work hardening capacity via the transformation-induced plasticity (TRIP) and twinning-induced plasticity (TWIP) effects, respectively. To develop advanced high-strength steels with an exceptional ductility-strength balance, the TRIP and TWIP effects must be controlled. The TWIP effect has recently drawn particular attention for industrial application. Therefore, detailed analyses on the deformation twinning in TWIP steels are required to clarify the underlying TWIP mechanism.

Deformation twins are formed in face-centered-cubic (FCC) metals with low stacking fault energy through the motion of a/6<112> Heidenreich-Shockley leading partial dislocations on every {111} plane. FCC deformation twinning has been reported in Cu,3,4) Cu–Al alloys,5) Ag–Au alloys,6) Co alloys,7) and Fe–Mn austenitic steels.8–10) The important factors for FCC twinning are the critical stress and critical strain for twinning. A critical strain for twinning is required to achieve a critical stress by work hardening and to provide the nucleation sites11) and is usually greater than a 10% tensile strain. For example, the critical strain in pure copper is 22.8% at 78 K in the <111> tensile direction, which is the most suitable crystallographic orientation for deformation twinning in common FCC metals.4) However, twinning deformation in Fe–Mn–C austenitic steels such as TWIP and Hadfield steels occurs at a relatively small strain, e.g., 1–3% tensile strain.12–14) The small critical strain for the onset of deformation twinning is an important characteristic of carbon-added high Mn austenitic steels.

The critical strain for deformation twinning has been reported to increase with increasing deformation temperature.8,15–17) The stacking fault energy increases with increasing deformation temperature,18) and the critical stress and
strain for twinning increase with increasing stacking fault energy.\textsuperscript{11,15,18} For example, Remy et al. reported\textsuperscript{8} that deformation twinning occurs at a stacking fault energy of 15–42 mJ/m\textsuperscript{2} in high Mn steels and that the deformation twin fraction decreases with increasing stacking fault energy. Allain et al. reported\textsuperscript{19} that deformation twinning occurs with stacking fault energies of 12–35 mJ/m\textsuperscript{2}. In addition to the deformation temperature dependence, the chemical composition dependence of the twinning behavior can also be determined by the stacking fault energy in austenitic steels.\textsuperscript{20,21} Thus, the deformation twinning behavior, including the critical strain and temperature composition dependence, can be determined by the stacking fault energy.

Exceptionally, the addition of carbon to a high Mn steel was observed to extend the deformation temperature and stacking fault energy regions for deformation twinning.\textsuperscript{8} Despite the stacking fault energy increasing with increasing carbon content in high carbon austenitic steels,\textsuperscript{20–25} For example, deformation twinning occurred in the temperature range of 323–423 K in a Fe–29Mn binary alloy. In contrast, deformation twinning occurred in the range of 323 K to over 600 K in a Fe–26Mn–0.2C steel.\textsuperscript{26} Fe–28Mn–7.5Al–0.75C exhibited deformation twinning in an abnormal deformation temperature range of 298–923 K.\textsuperscript{20} Other carbon-added Fe–Mn–C austenitic steels were also reported to exhibit deformation twinning up to a stacking fault energy of approximately 60 mJ/m\textsuperscript{2}.\textsuperscript{27,28} which is significantly higher than the upper limit of the stacking fault energy for deformation twinning reported by Remy\textsuperscript{8} and Allain.\textsuperscript{19} Bouaziz also claimed\textsuperscript{29,30} that carbon plays a positive role in deformation twinning because some Fe–Mn–C austenitic steels exhibited a considerable number of deformation twins, whereas Fe–Mn binary alloys with comparable stacking fault energy did not.\textsuperscript{30} Because carbon is a basic element in TWIP steels,\textsuperscript{31–33} Specifically, AFM analysis can determine the twinning direction as well as the twinning plane when the surface crystallographic orientations are known. The volume fraction of deformation twins is also determined by the AFM-based surface analysis proposed in this study. Namely, the combination of EBSD and AFM analyses can provide detailed information on deformation twinning in the grains with specific crystallographic orientations of polycrystalline specimens with different chemical compositions. In this paper, we demonstrate the effectiveness of the EBSD/AFM analyses on twinning phenomena in Fe–Mn–C TWIP steels and report new insights with respect to the carbon concentration dependence of the deformation twinning behavior.

2. Experimental Procedure

2.1. Materials

Fe–17Mn–0.6C, Fe–17Mn–0.8C, and Fe–18Mn–1.2C steels were prepared by vacuum induction melting. The ingots were forged and hot rolled at 1 273 K; they were then solution treated at 1 273 K for 3.6 ks under an argon gas atmosphere and subsequently water quenched to suppress the formation of carbides. The average grain size of the three steels was approximately 35 μm, including annealing twin boundaries. The chemical compositions of the steels are shown in Table 1. Hereafter, the compositions will be described in wt.%. These steels are TWIP steels, which exhibit an excellent combination of elongation and ultimate tensile strength (UTS), mainly because of deformation twinning.\textsuperscript{16,17,34} The solution-treated bars were cut by spark machining to obtain the specimen geometries required for the following experiments.

2.2. Microstructure Observations

X-ray diffraction (XRD), EBSD, and AFM analyses were performed to investigate the deformation-induced microstructures. The specimens for the XRD analyses were chemically polished with a solution (H\textsubscript{2}O\textsubscript{2} : HF = 10:1) at ambient temperature after mechanical polishing to remove the layers affected by the cutting and heat treatment. The XRD was operated at 35 kV, 300 mA and a scan rate of 0.02° s\textsuperscript{–1} with a Cu target. The specimens for the EBSD analyses were prepared by mechanical polishing for 3.6 ks under a low-stress condition. The EBSD analyses were conducted at 20 kV with a beam step size of 300 nm. The specimens for the AFM were electropolished at 277 K after mechanical polishing. To reduce the effect of grain boundaries, the region for the AFM observations was selected in a central part of the grain interiors approximately 10 μm from the grain boundaries.

2.3. Surface Relief Analysis

The AFM observations were performed to measure the surface-relief angles produced by deformation twinning. The measured relief angles were used to identify the deformation twins and their shear directions and to determine the deformation twin fraction quantitatively. The theoretical relief angles for the deformation twins θ were calculated using the following equation:\textsuperscript{35}

\[ \tan \theta = (\sin \gamma \sin^2 \beta / (\sqrt{2} + \sin \gamma \cos \beta \sin \beta)), \quad (1) \]

where β and γ are the angles between the surface and \{111\} plane and between the Burgers vector and surface trace, respectively. A schematic illustration of the calculations is presented in Fig. 1. The normal orientations of the surfaces required for the calculations were determined from the EBSD analyses. Twelve twinning directions are possible on four \{111\} slip planes. Because the twinning plane can be determined by the surface trace analyses, the number of candidates for the twinning direction is reduced to three. A comparison between the measured and calculated surface relief angles identifies the shear direction of the observed deformation twins.

In this study, the deformation twin fraction is also quanti-
titatively evaluated. Note that surface relief information overestimates the deformation twin fraction compared with measurements on a smooth surface. As shown in Fig. 1, the twin widths obtained from the surface relief and smooth surface images are \( w \) and \( z \), respectively. Therefore, the width \( w \) in the surface relief image must be converted to \( z \). Assuming that an area fraction of deformation twins is equal to the volume fraction, the volume fraction is obtained using the following equation:

\[
V = \sum_{i=2}^{n} \frac{F(T) \cdot w_i}{L} \quad \text{(2)}
\]

\[
F = z / w = 1 - (\tan \theta / \tan \beta) \quad \text{............ (3)}
\]

where \( L \) is the length of the measured line and \( F \) is the correction value to convert the area of twins on a surface relief image (\( w \)) to the area of twins on a smooth surface image (\( z \)), as indicated in Fig. 1.

2.4. Calculation of Stacking Fault Energy

Many researchers have measured\(^{18,22,23}\) and calculated\(^{10,19,20,36,37}\) the stacking fault energies in Fe–Mn–C austenitic steels. In the high carbon concentration region of more than 0.3 wt.%, the measured stacking fault energy increases linearly with increasing carbon concentration.\(^{22,23}\) Lee’s report\(^{21}\) also demonstrated that the stacking fault energy increased linearly with increasing total concentration of interstitial atoms (N + C). Because Nakano’s calculation\(^{37}\) can describe the linear relationship and absolute values, we used this approach to determine the stacking fault energies of the present steels. The calculation is based on the following equation:\(^{38}\)

\[
\gamma = 2 \rho (\Delta G^a + \Delta G^\text{strain}) + 2 \sigma \gamma / \epsilon \quad \text{............ (4)}
\]

where \( \Delta G^a \) is the free energy change by \( \epsilon \)-martensitic transformation from FCC (\( \gamma \)) to HCP (\( \epsilon \)), \( \rho \) is the molar surface density along \{111\}, and \( \sigma \) is the interfacial energy of \( \gamma / \epsilon \). The free energy change and molar surface density were extracted from the other references\(^{19,39-45}\) in Nakano’s paper.\(^{37}\) The interfacial energy was assumed to be 16 mJ/m\(^2\).\(^{37}\)

3. Results and Discussion

3.1. Carbon Concentration Dependence of Total Deformation Twin Fraction

Martensitic transformation and deformation twinning may occur in steels with low stacking fault energy. The existence of martensite and deformation twins was confirmed by XRD and EBSD. Figure 2 presents the XRD patterns, where the martensitic phase does not exist in the three steels after 10% tensile deformations at ambient temperature with an initial strain rate of \( 1.7 \times 10^{-4} \text{ s}^{-1} \). Figs. 3(a)–3(c) are the inverse pole figure maps of the rolling direction (RD-IPF) corresponding to the tensile direction. The EBSD images were taken after 10% tensile deformations at ambient temperature and with an initial strain rate of \( 1.7 \times 10^{-4} \text{ s}^{-1} \). Hereafter, the horizontal direction of the micrographs corresponds to the tensile direction. The EBSD analyses detected the deformation twins indicated by the black arrows in Figs. 3(a)–3(c). Figures. 3(d)–3(f) are the image quality (IQ) maps, which enable more detailed analyses. The IQ value deteriorates when two Kikuchi patterns of the parent and twin crystals are overlapped.\(^{13,46}\) The reduction in IQ values indicates deformation twins. Therefore, the IQ maps show the deformation twins more clearly than the RD-IPF maps when other plate-like products, such as martensitic phases, do not form.\(^{13,46}\) Because no second phases were detected in the XRD patterns in Fig. 2, the plate-like contrasts in the IQ maps of Figs. 3(d)–3(f) were identified as deformation twins. The observations indicate that the total number of deformation twins decreased with increasing carbon concentration.

3.2. Deformation Twins in <144> Tensile Orientations

Because the dominant deformation products were identified to be deformation twins, the surface reliefs can be analyzed precisely using the geometry of the deformation twins, as indicated in Fig. 1 and Eq. (1). Figures. 4(a)–4(c) present AFM differential height images showing the surface reliefs induced by 15% tensile deformations at ambient temperature with an initial strain rate of \( 1.7 \times 10^{-4} \text{ s}^{-1} \) in the three steels. The contrasts in color correspond to the difference in the surface tilt angle. The initial tensile orientations of the grains shown in Figs. 4(a)–4(c) are close to the <144> directions and have high Schmid factors of approximately 0.50 for primary deformation twinning. The initial crystallographic orientations, calculated stacking fault energies, Schmid factors, and information required for the calculation of the surface relief angles are summarized in Table 2. The

![Fig. 1. Schematic of the geometry of the surface relief arising from twinning deformation.](image)

![Fig. 2. XRD patterns in the three Fe–Mn–C steels after 10% tensile deformation at ambient temperature with an initial strain rate of 1.7×10^{-4} \text{ s}^{-1}.](image)
Fig. 3. RD-IPF maps after 10% tensile deformation in the (a) Fe–17Mn–0.6C, (b) Fe–17Mn–0.8C, (c) Fe–18Mn–1.2C steels. (d), (e), and (f) corresponding IQ maps for (a), (b), and (c), respectively. The deformations were performed at ambient temperature at an initial strain rate of $1.7 \times 10^{-4} \text{s}^{-1}$. (Online version in color.)

Fig. 4. AFM images observed in a grain with a high Schmid factor in the (a) Fe–17Mn–0.6C, (b) Fe–17Mn–0.8C, and (c) Fe–18Mn–1.2C steels. The tensile direction is approximately parallel to $<144>$. The 15% tensile deformations were provided at ambient temperature at an initial strain rate of $1.7 \times 10^{-4} \text{s}^{-1}$. (Online version in color.)

Table 2. Results of surface relief analyses in Figs. 4(a)–4(c). The twinning plane and direction are for the primary deformation twins. The correction values and deformation twin fraction were estimated using Eqs. (1) and (2). ND: Normal direction.

| Steel (wt.%) | Fe–17Mn–0.6C | Fe–17Mn–0.8C | Fe–18Mn–1.2C |
|-------------|--------------|--------------|--------------|
| Stacking fault energy | 31 mJ/m$^2$ | 35 mJ/m$^2$ | 43 mJ/m$^2$ |
| ND orientation | [8 7 9] | [−3 − 4 10] | [−7 5 13] |
| Tensile direction | [1 4 − 4] | [10 −14 −3] | [−4 −3 −1] |
| Twinning plane | (1−11) | (−11−1) | (11−1) |
| Twinning direction | [21−1] | [1−1−2] | [−1−1−2] |
| Trailing partial | [121] | [−1−2−1] | [−21−1] |
| $\beta$ | 65.5° | −55.3° | −56.0° |
| $\gamma$ | 26.8° | −122.5° | −30.9° |
| Relief angle for twin | 13.3° | −17.5° | −16.5° |
| Schmid factor of leading partials | 0.500 | 0.484 | 0.490 |
| Schmid factor of trailing partials | 0.250 | 0.341 | 0.326 |
| Correction value | 0.89 | 0.78 | 0.80 |
| Deformation twin fraction | 25.5% | 22.2% | 24.5% |
calculated stacking fault energies appear reasonable, as the stacking fault energies of similar steels, Fe–12Mn–0.7C, Fe–12Mn–0.8C, and Fe–13Mn–1.4C, were measured to be 35, 39, and 52 mJ/m² at ambient temperature, respectively,22) and because the additional Mn (approximately 5 wt.% in the present steels reduces the stacking fault energy by 5–10 mJ/m².47) The measured angles of the surface reliefs denoted as A–H are listed in Table 3. All the surface reliefs in Figs. 4(a)–4(c) were identified as primary deformation twins by comparing the theoretical and measured relief angles. The results indicate that the 15% plastic strains in the <144> tensile orientation of all of the steels occurred mainly by the primary deformation twinning. Additionally, as observed in Table 2, the deformation twin fraction did not change significantly upon increasing the carbon concentration in the <144> tensile orientations despite the increase in the stacking fault energy with increasing carbon concentration.

3.3. Deformation Twins in <111> Tensile Orientations

Figures. 5(a)–5(c) show the surface reliefs in near-<111> tensile orientations of the Fe–17Mn–0.6C, Fe–17Mn–0.8C, and Fe–18Mn–1.2C steels after 15% tensile deformations at ambient temperature with an initial strain rate of $1.7 \times 10^{-4}$ s⁻¹. The corresponding theoretical surface relief angles and other relevant information are summarized in Table 4. In contrast to the <144> orientations, multiple shear systems were activated in the <111> tensile crystallographic orientations of the three steels. The measured relief angles are listed in Table 5. Reliefs A, B, C, and D in Fig. 5(c) were identified as deformation twins on the (−111) primary twinning plane or the (1−11) secondary twinning plane, indicating that twinning markedly occurred as a main plastic deformation mode in the <111> orientation of the Fe–17Mn–0.6C steel. However, in the Fe–18Mn–1.2C steel, the relief angles of reliefs C and D in Fig. 5(c) were significantly larger than the theoretical values for twinning on (−11−1), although the relief angles of reliefs A and B in Fig. 5(c) showed good agreement with the theoretical value for the (111)[−1−12] twinning. The large values imply an occurrence of slip deformation on (−11−1) because dislo-

| Relief | Fe–17Mn–0.6C | Fe–17Mn–0.8C | Fe–17Mn–1.2C |
|--------|-------------|-------------|-------------|
| A      | 12.8°       | −16.1°      | −15.2°      |
| B      | 12.2°       |−15.1°       | −16.1°      |
| C      | 13.8°       |−15.9°       | −15.6°      |
| D      | 12.0°       |−15.8°       | −15.2°      |
| E      | 12.8°       |−15.6°       | −15.2°      |
| F      | 13.7°       |−16.0°       | −16.8°      |
| G      | 12.6°       |−15.9°       | −17.1°      |
| H      | 12.8°       |−15.3°       | −16.8°      |

Table 3. Measured angles of the surface reliefs denoted by A–H in Figs. 4(a)–4(c).

| Steel (wt.%) | Fe–17Mn–0.6C | Fe–17Mn–0.8C | Fe–18Mn–1.2C |
|-------------|-------------|-------------|-------------|
| Stacking fault energy | 31 mJ/m² | 35 mJ/m² | 43 mJ/m² |
| ND orientation | [4 3 6] | [−9 −11 −4] | [−3 13 17] |
| Tensile direction | [18 14 −19] | [−5 3 3] | [−8 −11 7] |
| Twinning plane 1 | (−111) | (11−1) | (111) |
| Twinning direction 1 | [12−1] | [−121] | [−1−12] |
| Trailing partial 1 | [−11−2] | [−21−1] | [1−21] |
| $\gamma$ 1 | 68.3° | −85.4° | 43.8° |
| Relief angle for twin 1 | 13.0° | −163.9° | 40.9° |
| Schmid factor of leading partials 1 | 0.400 | 0.384 | 0.399 |
| Schmid factor of trailing partials 1 | 0.209 | 0.274 | 0.254 |
| Twinning plane 2 | (1−11) | (1−11) | (1−11) |
| Twinning direction 2 | [21−1] | [−12] | [−2−11] |
| Trailing partial 2 | [1−1−2] | [−2−11] | [−1−2−1] |
| $\gamma$ 2 | 58.8° | −85.4° | −88.4° |
| Relief angle for twin 2 | 17.8° | −163.9° | 169.1° |
| Schmid factor of leading partials 2 | 8.2° | −10.9° | 7.6° |
| Schmid factor of trailing partials 2 | 0.277 | 0.384 | 0.342 |

Table 4. Results of the surface relief analyses in Figs. 5(a)–5(c).

| Relief | Fe–17Mn–0.6C | Fe–17Mn–0.8C | Fe–18Mn–1.2C |
|--------|-------------|-------------|-------------|
| A      | 7.3°(111)  |−11.5°(11−1) | 10.8°(111)  |
| B      | 8.0°(−111) |−12.2°(11−1) | 9.3°(111)   |
| C      | 7.3°(1−11) |−13.1°(1−11) | 13.4°(−11−1)|
| D      | 7.5°(1−11) |−10.8°(1−11) | 17.9°(−11−1)|

Table 5. Measured surface relief angles and corresponding slip planes of reliefs A–D in Figs. 5(a)–5(c).
cation multiplications on a specific plane can produce large surface relief angles. Namely, the <111> tensile orientation showed a suppression of deformation twinning due to the increase in carbon concentration, unlike the <144> tensile orientation.

3.4. Deformation Twinning in a Large Plastic Deformation Stage

Although the initial orientation dependence of deformation twinning was indicated by the aforementioned results, the effect of the initial orientations must decrease with increasing plastic strain because of grain rotation. AFM analyses cannot be utilized for large plastic deformation because slip deformation, which does not have a specific surface relief angle, acts as a dominant plastic deformation mode. Therefore, XRD and EBSD methods were used to analyze the microstructures with a large plastic strain. Figure 6 reveals that a second phase was not detected by XRD even after 60% tensile deformation at ambient temperature with an initial strain rate of $1.7 \times 10^{-4}$ s$^{-1}$ in the Fe–18Mn–1.2C steel. Figure 7 presents RD-IPF maps at 60% plastic strain. Compared with Fig. 3 showing the RD-IPF at 10% strain, the RD-IPF maps at 60% strain indicate that the number of deformation twins and the activated shear systems increased with increasing tensile strain. Most of the grains were oriented in the <111> tensile orientation and partially in the <001> tensile orientation, as demonstrated by the blue and red colors, respectively, in the RD-IPF maps. The grain rotation corresponds to the formation of the typical texture in Fe–Mn–C TWIP steels. As reported in conventional studies, the <111> tensile orientation showed a considerable amount of deformation twins, and the <001> tensile orientation did not. Namely, the 60% tensile strain activated primary deformation twinning as well as secondary deformation twinning in most of the grains close to the <111> tensile orientation. In addition, even in the Fe–18Mn–1.2C steel, which has the highest stacking fault energy, an extremely high amount of deformation twins appears after exceeding 40% plastic strain. Because relatively-near <111> orientations, such as <144> and <110>, rotate to <111> tensile orientations, the initial crystallographic orientation dependence of deformation twinning behavior becomes weak in the late-deformation stage.

3.5. Deformation Twinning and Stacking Fault Energy

The total number of deformation twins decreased with increasing carbon concentration, as demonstrated in Figs. 3(a)–3(f). In addition, the local twin fraction in the near <111> tensile orientation decreased with increasing carbon concentration (Figs. 5(a)–5(c)). These facts can be explained by the increase in the stacking fault energy upon increasing the carbon concentration. Because the twinning...
stress is known to increase with increasing stacking fault energy,\(^4,12,15,53\) the increase in the stacking fault energy suppresses the deformation twinning, and the slip deformation occurred instead as indicated by the surface relief analysis in Fig. 5(c).

However, the deformation twin fraction in the \(<144>\) tensile orientations after the 15% deformation did not change with increasing carbon concentration, as observed in Figs. 4(a)–4(c) and Table 2, despite the increase in the stacking fault energy. Hence, the present experimental results do not correspond to the reported relationship between the stacking fault energy and twinning behavior. The peculiar tendency of the results at the \(<144>\) tensile orientation regarding the carbon concentration dependence is discussed in detail in Part 2 of this paper.

4. Conclusion

EBSD and AFM analyses were employed to clarify some characteristics of deformation twinning in Fe–Mn–C austenitic steel with high concentrations of carbon. The microstructural examinations were conducted at different carbon concentrations, plastic strains, and crystallographic orientations. The observations revealed the following unconventional findings:

(1) The total number of deformation twins decreases with increasing carbon concentration at ambient temperature. In addition, the deformation twinning was suppressed with increasing carbon concentration in \(<111>\) tensile orientations. The carbon concentration dependence is attributed to an increase in stacking fault energy.

(2) The local deformation twin fraction in the grain with the tensile orientation of \(<144>\) did not change as a function of carbon concentration, despite the increase in stacking fault energy. This fact does not follow the stacking fault energy-based consideration. Hence, the specialty of carbon effects must be considered in understanding the TWIP effect in Fe–Mn–C austenitic steels. The underlying mechanism of the carbon effect on twinning is proposed in Part 2 of this paper.

Acknowledgements

M. K. acknowledges the Research Fellowship of NIMS Junior Researcher (2009–2010) and the Japan Society for the Promotion of Science for Young Scientists (2011) and is grateful to the Materials Manufacturing and Engineering Station and Materials Analysis Station at the National Institute for Materials Science for supporting this work by preparing the samples and performing the analysis of the chemical compositions.

This paper is based on results obtained from a future pioneering program commissioned by the New Energy and Industrial Technology Development Organization (NEDO).

REFERENCES

1) I. Tamura: Metall. Sci., 16 (1982), 245.
2) O. Grüssel, L. Krüger, G. Frommeyer and L.W. Meyer: Int. J. Plast., 16 (2000), 1391.
3) T. H. Blewitt, R. R. Colman and J. K. Redman: J. Appl. Phys., 28 (1957), 651.
4) D. Weiner: Acta Metall., 20 (1972), 1235.
5) T. Mori and K. Fujita: J. Jpn. Inst. Met., 18 (1977), 17.
6) H. Suzuki and C. S. Barrett: Acta Metall., 6 (1958), 156.
7) L. Remy and A. Pineau: Mater. Sci. Eng., 26 (1976), 123.
8) L. Remy and A. Pineau: Mater. Sci. Eng., 28 (1977), 99.
9) Y. G. Kim, S. Y. Park and J. K. Han: Metall. Trans. A, 16A (1985), 689.
10) P. H. Adler, G. B. Olson and W. S. Owen: Metall. Trans. A, 17A (1986), 1725.
11) L. Remy: Acta Metall., 26 (1978), 443.
12) I. Karaman, H. Sehitoglu, K. Gall, Y. I. Chumlyakov and H. J. Maier: Acta Mater., 48 (2000), 1345.
13) D. Barbier, N. Gey, N. Bozzolo, A. Allain and M. Humbert: J. Microsc., 235 (2009), 67.
14) E. G. Astafurova and Y. I. Chumlyakov: Phys. Metal. Metallogr., 180 (2009), 541.
15) D. R. Steimmetz, T. Jäpel, B. Wietbrock, P. Eisenlohr, I. Gutierrez-Urrutia, A. Saeed-Akbari, T. Hickel, F. Roters and D. Raabe: Acta Mater., 61 (2013), 494.
16) M. Koyama, T. Sawaguchi and K. Tsuzaki: Tetsu-to-Hagané, 98 (2012), 229.
17) M. Koyama, T. Sawaguchi and K. Tsuzaki: ISIJ Int., 53 (2013), 323.
18) L. Remy: Acta Metall., 25 (1977), 173.
19) S. Allain, J. P. Chateau, O. Bouaziz, S. Migot and N. Guetlon: Mater. Sci. Eng. A, 387–389 (2004), 158.
20) A. Saeed-Akbari, J. Imlau, U. Prahl and W. Bleck: Metall. Mater. Trans. A, 40 (2009), 3076.
21) T. H. Lee, E. Shin, C. S. Oh, H. Y. Ha and S. J. Kim: Acta Mater., 58 (2010), 317.
22) P. Y. Volotovsich, V. N. Grindnev and Y. N. Petrov: Fiz. Metall. Metalldoved., 40 (1975), 554.
23) Y. N. Petrov: Z. Metallkd., 94 (2003), 1012.
24) W. Charnock and J. Nutting: Met. Sci., 1 (1967), 123.
25) P. J. Broman and G. S. Ansell: Metall. Mater. Trans. A, 9A (1978), 879.
26) J. S. Chou and C. G. Hao: Scr. Mater., 26 (1992), 261.
27) A. S. Hamada: Doctoral Thesis, University of Oulu, Linnanmaa, Finland, (2007).
28) Y. S. Han and S. H. Hong: Mater. Sci. Eng. A, 222 (1977), 76.
29) O. Bouaziz, S. Allain, C. P. Scott, P. Cugy and D. Barbier: Curr. Opin. Solid State Mater. Sci., 15 (2011), 141.
30) S. Allain, O. Bouaziz and J. P. Chateau: Proc. Int. Conf. SHSS, Associazione Italiana Di Metalurgia, Milano, (2010).
31) M. Koyama, M. Murakami, K. Ogawa, T. Kikuchi and T. Sawaguchi: J. Jpn. Inst. Met., 71 (2007), 672.
32) M. Koyama, M. Murakami, K. Ogawa, T. Kikuchi and T. Sawaguchi: Mater. Trans., 49 (2008), 812.
33) M. Koyama, T. Sawaguchi, K. Ogawa, T. Kikuchi and M. Murakami: Mater. Sci. Eng. A, 497 (2008), 353.
34) M. Koyama, T. Sawaguchi, T.-K. Lee, C-S. Lee and K. Tsuzaki: Mater. Sci. Eng. A, 483–484 (2008), 184.
35) J. Nakano and P. J. Jacques: Calphad, 34 (2010), 167.
36) G. B. Olson and M. Cohen: Metall. Trans. A, 7A (1976), 1897.
37) Y. K. Lee and C. S. Choi: Metall. Mater. Trans. A, 31 (2000), 355.
38) A. T. Dinsdale: Calphad, 15 (1991), 317.
39) W. Huang: Calphad, 13 (1989), 243.
40) P. Gustafson: Scand. J. Metall., 14 (1985), 259.
41) W. Huang: Scand. J. Metall., 19 (1990), 26.
42) W. Huang: Metall. Trans. A, 21A (1990), 2115.
43) S. M. Cotes, A. F. Guillermot and M. Sade: Metall. Mater. Trans. A, 35A (2004), 83.
44) J. E. Jin and Y. K. Lee: Mater. Sci. Eng. A, 527 (2009), 157.
45) P. Y. Volotovsich, V. N. Grindnev and Y. N. Petrov: Fiz. Metall. Metalldoved., 42 (1976), 372.
46) M. Koyama, T. Sawaguchi, K. Ogawa, T. Kikuchi and M. Murakami: J. Jpn. Inst. Met., 73 (2009), 174.
47) M. Koyama, T. Sawaguchi, K. Ogawa, T. Kikuchi and M. Murakami: Mater. Sci. Eng. A, 520 (2011), 7310.
48) M. Koyama, T. Sawaguchi, K. Ogawa, T. Kikuchi and M. Murakami: Mater. Sci. Eng. A, 51 (2010), 1194.
49) D. Barbier, N. Gey, S. Allain, N. Bozzolo and M. Humbert: Mater. Sci. Eng. A, 500 (2009), 196.
50) J. A. Jiménez and G. Frommeyer: Mater. Charact., 61 (2010), 221.
51) I. Gutierrez-Urrutia, S. Zaefferer and D. Raabe: Mater. Sci. Eng. A, 527 (2010), 3552.
52) M. A. Meyers, O. Vohring and V. A. Lubarda: Acta Mater., 49 (2001), 4025.