Effects of Deformation Induced Phase Transformation and Twinning on the Mechanical Properties of Austenitic Fe–Mn–Al Alloys

Kazunori SATO,1) Michiyuki ICHINOSE,2) Yoshihiko HIROTsu3) and Yasunobu INOUE1)

1) Analysis Center, Nagaoka University of Technology, Kamitomiokamachi, Nagaoka, Niigata-ken, 940-21 Japan. 2) Graduate Student of Mechanical Engineering, Nagaoka University of Technology. Now at Production Engineering Center, Hitachi Kasei Corp., Higashi-cho, Hitachi, Ibaraki-ken, 317 Japan. 3) Division of Mechanical Engineering, Nagaoka University of Technology, Kamitomiokamachi, Nagaoka, Niigata-ken, 940-21 Japan.

(Received on January 20, 1989; accepted in the final form on May 19, 1989)

Structure and mechanical properties of austenitic Fe–(20 and 30)Mn–(0 to 7) Al alloys in the temperature range between 77 and 295 K have been studied in relation to the occurrences of phase transformation and deformation twinning. Additions of aluminum to the 20 wt% Mn alloys significantly decreased the \( \gamma \rightarrow \varepsilon \) transformation temperature. The yield stress of these alloys increased with increasing aluminum content, whereas the strain hardening of them decreased. This tendency is prominent at low temperatures. In the 30 wt% Mn alloys the yield stress and strain hardening were almost identical regardless of aluminum contents. Additions of aluminum strongly suppress the \( \gamma \rightarrow \varepsilon \) transformation and give birth to the occurrence of deformation twinning. Calculated stacking fault energy based on a regular solution approach shows that the austenitic Fe–Mn–Al alloys which have the stacking fault energy approximately larger than 20 erg/cm\(^2\) favor the deformation twinning leading to the increase in low temperature ductility.

KEY WORDS: phase transformation; mechanical properties; austenitic Fe–Mn–Al alloys; martensite; deformation twins; low temperatures; stacking fault energy.

1. Introduction

Binary austenitic (\( \gamma \)) Fe–Mn alloys transform partially to hcp (\( \varepsilon \)) and bcc (\( \alpha \)) structures within a restricted composition range during quenching.\(^\text{11}\) Plastic deformation also gives birth to the formation of \( \gamma \) twins, \( \varepsilon \) and \( \alpha' \) phases in the \( \gamma \) matrix. Formation of these deformation-induced phases being dependent on alloy composition and deformation temperature significantly affects mechanical properties of these alloys.\(^\text{2,3}\)

Additions of alloying elements to the Fe–Mn alloys affect their phase stability. As for the \( \gamma \rightarrow \varepsilon \) transformation, small additions of Ni, V, Mo, Al, Cu, C, Ti and Nb to Fe–17Mn\(^\text{*}\) alloy significantly decrease the \( \gamma \rightarrow \varepsilon \) martensitic transformation start temperature, \( M_{\gamma\rightarrow \varepsilon} \), whereas additions of Co and Si to Fe–(20 to 28)Mn alloys do not change the \( M_{\gamma\rightarrow \varepsilon} \) temperature but strongly reduce the Neel temperature, \( T_N \).\(^\text{4,5}\) Additions of these elements give birth to various properties in Fe–Mn alloys such as strain hardening associated with deformation twinning in famous Hadfield manganese steel (Fe–13Mn–1.2C)\(^\text{6}\) and shape memory effect associated with the \( \gamma \rightarrow \varepsilon \) transformation in Fe–30Mn–1Si alloy.\(^\text{7}\)

Alloys based on the Fe–Mn–Al system is not only a possible substitute for the Ni–Cr austenitic stainless steels\(^\text{8–12}\) but a promising cryogenic material which possesses excellent cryogenic properties.\(^\text{13,14}\) Recently, Kim et al. have shown that deformation twinning takes place in Fe–30Mn–5Al–0.3C–0.1Nb(or –0.1V) alloys at low temperatures and that increasing amount of aluminum in the same compositional alloy contributes to an increase in uniform elongation with decreasing temperature.\(^\text{15}\) However, the mechanism of low-temperature deformation mode which is considered to be dependent on alloy composition remains uncertain.

The present study was undertaken to examine compositional dependency of the plastic deformation behavior in the austenitic Fe–Mn–Al alloys at low temperatures. In particular, the role of aluminum was investigated in detail for the Fe–(20 and 30)Mn alloys in order to establish the relation between mechanical properties and formations of deformation twinning and/or stress-induced martenitic transformations. Further, thermodynamic analysis was made in order to characterize phase stability of austenitic Fe–Mn–Al alloys. The transformation products were reviewed in terms of the stacking fault energy.

2. Experimental

2.1. Alloy Preparation

Eight 50 kg ingots having various compositions were prepared by air induction melting of high purity iron, electrolytic manganese and high purity aluminum. The charges were cast into a magnesia mold. The ingots with approximately 120 mm in square and 370 mm in height were soaked at 1473 K for 30 min

* All in weight percent unless otherwise specified.
and subsequently forged to a thickness of about 25 mm. The slabs were cut into several pieces and some of them were hot rolled at 1243 K to about 3 mm thick plates. Finally, both the slabs and plates were heated at 1273 K for 60 min under a protective atmosphere of argon and then quenched into water. The commercial type 304 steel was also prepared for comparison. Chemical analyses of the alloys are given in Table 1.

2.2. Mechanical Testing

The heat-treated plates were machined into specimens for tensile tests with a gage section 2.6 mm in thickness, 12.5 mm in width and 50 mm in length. Tensile tests were conducted at a constant cross-head speed of 1.4 x 10⁻⁴ m s⁻¹ with a 25 Ton strain-type testing machine. True stress–strain curves were obtained from the measured loads and cross-head displacements. To obtain a better accuracy, room temperature stress–strain curves were verified by elongations measured by an extensometer. No significant difference was observed. Room temperature tests were conducted at 295±5 K and low temperature ones were made at 77 K by immersing specimens in liquid nitrogen during the tests. The percent elongation reported is the total elongation measured after fracture. Each test was carried out at least twice under the same condition. The obtained stress–strain curves were sufficiently reproducible in each case.

2.3. Structural Observations

Volume fractions of the phases present were measured with an X-ray diffractometer using Cu-Kα radiation (40 kV and 30 mA). Determination of the volume fraction was made by evaluating integrated intensities of the ε (10-1), γ (200) and α (200) peaks. Integrated X-ray intensities were corrected for the dependence of the atomic scattering factor, and Lorentz-polarization factor on scattering angle. The fraction of the α phase was also compared with the magnetic saturation measurement with a ferrite meter. Surface preparation of specimens for both the X-ray diffraction and magnetic measurements was carried out by a chemical polishing in a hydrogen peroxide and hydrofluoric acid (3 %) solution in order to avoid mechanical damages during the preparation. However, the obtained volume fractions may be semiquantitative for deformed specimens due to the presence of a texture structure. The γ → ε transformation temperature was also determined on continuous cooling from 1743 K to room temperature at a rate of 3 deg min⁻¹ by dilatometry.

Microstructures of the specimen before and after plastic deformation were examined by optical and transmission electron microscopy. Tensile fracture surfaces were observed with a scanning electron microscope (SEM). Optical photomicrography was performed for specimens etched with a 3 % nitric. The structural observation was conducted by a transmission electron microscope (TEM) at an accelerating voltage of 200 kV. Thin foils for electron microscopy were prepared by an electropolishing in a solution of chromic (80 g CrO₃) and phosphoric (400 ml H₃PO₄) acid solution at 40 to 70°C. The electropolishing was conducted at 10 V and 0.5 to 1.0 A.

3. Results

3.1. As-quenched Structures

Volume fractions of the γ, ε and α phases present in the alloys after the quench are summarized in Table 1. No α' martensite was formed for these alloys in the as-quenched condition. Among the alloys investigated only alloy A9 exhibited a mixed ε+γ structure, in which ε phase occupied 88 % in volume fraction, as shown in Fig. 1(a). This structure is consistent with that obtained by slow cooling in Fe-20Mn alloy. On the other hand, alloys A2 and A4 gave birth to a fully austenitic structure as represented by Fig. 1(b). Further addition of aluminum to the 20 % Mn alloys produced a two-phase structure of α+γ in which small ferritic regions exist. Thus, the amount of aluminum addition was limited for the 20 % Mn alloys in order to obtain the fully austenitic phase. Maximum content of aluminum to the 30 % Mn alloys with keeping a single-phase austenitic structure was higher than that in the 20 % Mn alloys as shown in Table 1.

The occurrence of the γ → ε transformation was examined by dilatometry for alloys A0, A2, A4, B0, B2

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Composition (wt%)</th>
<th>Phases*</th>
</tr>
</thead>
<tbody>
<tr>
<td>A0</td>
<td>Mn 20.60, Al &lt;0.05, C 0.070, Si 0.070, P 0.018, S 0.005</td>
<td>ε+γ(88 % ε)</td>
</tr>
<tr>
<td>A2</td>
<td>19.10, 1.81, 0.071, 0.140, 0.020, 0.007</td>
<td>γ</td>
</tr>
<tr>
<td>A4</td>
<td>19.40, 4.40, 0.033, 0.030, 0.011, 0.005</td>
<td>γ</td>
</tr>
<tr>
<td>A6</td>
<td>20.60, 5.70, 0.080, 0.047, 0.014, 0.002</td>
<td>α+γ(25 % α)</td>
</tr>
<tr>
<td>B0</td>
<td>31.30, &lt;0.05, 0.070, 0.000, 0.015, 0.008</td>
<td>γ</td>
</tr>
<tr>
<td>B2</td>
<td>29.14, 2.05, 0.070, 0.110, 0.020, 0.007</td>
<td>γ</td>
</tr>
<tr>
<td>B4</td>
<td>29.18, 5.81, 0.060, 0.110, 0.011, 0.007</td>
<td>γ</td>
</tr>
<tr>
<td>B6</td>
<td>20.08, 7.15, 0.069, 0.050, 0.017, 0.002</td>
<td>γ</td>
</tr>
<tr>
<td>SUS 304</td>
<td>8.63, 18.23, 0.07, 0.48, 0.029, 0.006</td>
<td>γ</td>
</tr>
</tbody>
</table>

* Identified after heat-treatment at 1273 K for 60 min followed by water quenching.
Only alloy A0 showed the $\gamma \rightarrow \varepsilon$ transformation at 380 K which was almost consistent with the transformation temperatures previously reported for the same compositional alloy.\textsuperscript{13-51} The other alloys did not reveal the $\gamma \rightarrow \varepsilon$ transformation within the temperature range examined. Alloys A2 and A4 were further cooled to liquid nitrogen temperature and examined by optical microscopy and X-ray diffraction method at room temperature. However, the occurrence of the $\gamma \rightarrow \varepsilon$ transformation was not observed. This implies that aluminum addition to the austenitic Fe–Mn alloys considerably suppresses the $\gamma \rightarrow \varepsilon$ transformation.

### 3.2. Yielding and Stress–Strain Behavior

True stress–strain curves for alloys A0, A2, A4, B0, B2 and B4 deformed at 77 K are shown in Fig. 2. Tensile behavior of the 20 % Mn alloys was significantly affected by the aluminum content. Alloy A0 showed the highest yield stress and the highest work hardening rate. Comparison of the tensile behavior between alloys A2 and A4 shows that the yield stress of alloy A2 was lower than that of alloy A4 but the work hardening rate of alloy A2 was higher than that of alloy A4. This suggests that yielding and the following plastic deformation of these alloys take place through different mechanisms. In contrast to this, no distinct effect of aluminum addition was observed for the stress–strain curves of the 30 % Mn alloys. Temperature dependence of the stress–strain curves were also obtained in the temperature range between 77 and 295 K for alloys A2, A4, B0 and B4. In each alloy the yield stress and the work-hardening rate increased almost monotonically with decreasing temperature. A typical example is shown in Fig. 3 for alloy A2.

The yield stress at zero plastic strain is plotted as a function of aluminum concentration for the 20 and 30 % Mn alloys strained at 77 and 295 K as shown in Fig. 4. The phase boundary of $\gamma | (\alpha + \gamma)$ for the 20 % Mn alloys were estimated and marked in the Fig. 4(a). Within the $\gamma$ phase region, the yield stress of the 20 % Mn alloys increased with increasing aluminum content, which is more prominent at 77 K than at 295 K. In the 30 % Mn alloys, a very slight increase in the yield stress with increasing aluminum content was observed at each temperature. This indicates that solid solution hardening by aluminum addition can be neglected when one considers the compositional dependence of the yield stress in the austenitic 20 % Mn alloys. The values of the work-hardening rate, $\dot{\sigma}/\dot{\varepsilon}$, at a plastic strain of 0.03 for
these alloys as well as the SUS 304 steel at 295 and 77 K were determined from the stress–strain curves and are listed in Table 2. Thus, the results of the tension tests at 77 K for the austenitic 20 % Mn alloys indicate that the distinct increase in the yield stress with the decrease in the work hardening rate on increasing aluminum content is ascribed to a structural change under tensile loading.

The values of the yield strength, tensile strength and ultimate tensile elongation for the alloys tested and those for type 304 stainless steel are summarized in Table 3.

### 3.3. Deformation Mode and Fine Structure Associated with Deformation

Microstructural changes with increasing tensile plastic strain were observed for alloys A2, A4, B0 and B4 at 77 and 295 K. Figs. 5 through 7 show the optical micrographs of specimens sectioned transversely for alloys A2, A4 and B4, respectively, after deformation at 77 K.

Alloy A2 showed a distinct structural change from the early stage of deformation as shown in Fig. 5.

The as-quenched structure was completely austenitic, whereas sharp striations appeared almost over the whole grains at a plastic strain of 0.02. Formation of ε martensite was identified for the specimen strained to 0.02 by X-ray diffraction. The amount of the ε phase increased with increasing plastic strain. In addition, α' martensite was also identified at a plastic strain higher than 0.09. Formation of ε martensite followed by the α' martensite production with increasing plastic strain was also identified at 295 K. Fig. 8 shows changes in microstructure determined by X-ray diffraction method with increasing plastic strain for alloy A2 at 295 and 77 K.

### Table 2. Work-hardening rate of alloys at a plastic strain of 0.03.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Work-hardening rate (×10^3 KPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>295 K</td>
</tr>
<tr>
<td>A0</td>
<td>2.6</td>
</tr>
<tr>
<td>A2</td>
<td>1.6</td>
</tr>
<tr>
<td>A4</td>
<td>1.2</td>
</tr>
<tr>
<td>B0</td>
<td>1.1</td>
</tr>
<tr>
<td>B2</td>
<td>1.3</td>
</tr>
<tr>
<td>B4</td>
<td>1.3</td>
</tr>
<tr>
<td>SUS 304</td>
<td>1.5</td>
</tr>
</tbody>
</table>

### Table 3. Mechanical properties of Fe-Mn-Al alloys.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Test temperature (K)</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A0</td>
<td>295</td>
<td>365</td>
<td>750</td>
<td>17</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>510</td>
<td>980</td>
<td>17</td>
</tr>
<tr>
<td>A2</td>
<td>295</td>
<td>160</td>
<td>600</td>
<td>66</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>195</td>
<td>1020</td>
<td>68</td>
</tr>
<tr>
<td>A4</td>
<td>295</td>
<td>170</td>
<td>460</td>
<td>67</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>390</td>
<td>885</td>
<td>90</td>
</tr>
<tr>
<td>A6</td>
<td>295</td>
<td>255</td>
<td>515</td>
<td>37</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>390</td>
<td>795</td>
<td>38</td>
</tr>
<tr>
<td>B0</td>
<td>295</td>
<td>200</td>
<td>445</td>
<td>56</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>240</td>
<td>705</td>
<td>60</td>
</tr>
<tr>
<td>B2</td>
<td>295</td>
<td>170</td>
<td>460</td>
<td>47</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>303</td>
<td>755</td>
<td>87</td>
</tr>
<tr>
<td>B4</td>
<td>295</td>
<td>190</td>
<td>460</td>
<td>46</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>340</td>
<td>740</td>
<td>85</td>
</tr>
<tr>
<td>B7</td>
<td>295</td>
<td>215</td>
<td>530</td>
<td>37</td>
</tr>
<tr>
<td></td>
<td>138</td>
<td>360</td>
<td>755</td>
<td>47</td>
</tr>
<tr>
<td>SUS 304</td>
<td>295</td>
<td>235</td>
<td>655</td>
<td>80</td>
</tr>
<tr>
<td></td>
<td>77</td>
<td>630</td>
<td>1 565</td>
<td>41</td>
</tr>
</tbody>
</table>

* 0.2 % offset

Fig. 5.
Optical micrographs of alloy A2 (a) in the as-quenched condition, and deformed in tension at 77 K to strains of (b) 0.02, (c) 0.04 and (d) 0.09.
Alloy A4 also showed a microstructural change at 77 K from the early stage of deformation as shown in Fig. 6. X-ray diffraction analysis of these specimens revealed that neither \( \alpha' \) nor \( \varepsilon \) martensites was formed at 77 K at strains smaller than 0.1. However, the volume fraction of the \( \alpha' \) phase for the specimen strained to 0.4 at 77 K attained to be about 30%. In contrast to this, at 295 K no newly formed phases were identified even for the specimen after fracture.

Fig. 7 shows the change of microstructure for alloy B4 deformed at 77 K to plastic strains of 0.04 and 0.10. A slight amount of the striation was seen to increase with plastic strain. In the 30% Mn alloys, optical micrographic observations of alloys B0, B2 and B4 appeared to show the same microstructural change at deformation temperatures of 77 and 295 K. The structures of the alloys at a plastic strain of 0.1 at 77 K were identified to be fully austenitic by X-ray diffraction. However, after fracture only alloy B0 exhibited a slight amount of \( \varepsilon \) phase (<3%) at both temperatures. The other alloys did not give birth to newly formed phases.

Electron micrographs and corresponding electron diffraction patterns of alloy A2 deformed at 77 K to a plastic strain of 0.02 are shown in Fig. 9. Presence of stacking faults was easily observed for the deformed specimen as shown in Fig. 9(a), whereas the same alloy in the as-quenched condition did not exhibit the stacking faults under the present TEM observations. The deformed specimen also includes the \( \varepsilon \) phase as shown in Fig. 9(b) which was identified from the diffraction patterns shown in Fig. 9(c). These diffraction patterns showed the coexistence of the \( \gamma \) and \( \varepsilon \) phases with the orientation relationship,

\[
(0001), // (111),
[11\overline{2}0], // [110].
\]

The microstructure of alloy A4 deformed at 77 K is demonstrated by the electron micrographs and corresponding electron diffraction patterns of Fig. 10. After a relatively small amount of deformation (0.03), several parallel bands were observed in the austenite matrix with a small reflection spot close to each fcc (200) reflection spot and streaking along the [111] direction as shown in Fig. 10(a). However, microscopic observations for this alloy did not exhibit the stacking faults as found in alloy A2. A plastic strain of about 0.1 in alloy A4 gave birth to a number of parallel bands with densely populated dislocations as shown in Fig. 10(b). The electron diffraction pat-
Fig. 9. TEM micrographs of alloy A2 deformed in tension at 77 K to a strain of 0.02 showing bright-field images of (a) stacking faults (S.F.) and (b) ε plates, and (c) electron diffraction pattern consisting of the [111] and [001] zone axes taken from the image of (b).

Fig. 10(a). Bright-field image and electron diffraction pattern with [110] zone axis of alloy A4 deformed in tension to a strain of 0.03 at 77 K.
(b). Bright-field image of alloy A4 at a strain of about 0.1 at 77 K.
(c). Electron diffraction pattern with [110] zone axis taken from (b).

tern of the same area (Fig. 10(c)) distinctly revealed these bands to be deformation twins. Deformation twins were also observed for alloy B4 deformed in tension to a plastic strain of about 0.1 at 77 K as shown in Fig. 11. These TEM observations indicate that the amount of deformation twins produced during plastic deformation in alloy B4 is smaller than that in alloy A4 under the same tensile condition.
4. Discussion

4.1. Formation of hcp and bcc Phases in Austenitic Fe–Mn–Al Alloys

The \(\gamma \rightarrow \varepsilon\) transformation temperatures, \(M_{\gamma \rightarrow \varepsilon}\) and \(A_{\gamma \rightarrow \varepsilon}\), determined by the dilatometric method in binary Fe–(15 to 28)Mn alloys decrease with increasing manganese content and the alloys exceeding the manganese content than 28 wt\% do not undergo the \(\gamma \rightarrow \varepsilon\) transformation on cooling to room temperature.\(^{1,3,4}\) Breedis and Kaufman\(^{10}\) calculated the phase diagram of the binary Fe–Mn system using the regular solution approach. Ishida and Nishizawa\(^{5}\) experimentally obtained the \(T_{\gamma \rightarrow \varepsilon} = (M_{\gamma \rightarrow \varepsilon} + A_{\gamma \rightarrow \varepsilon})/2\) line, which corresponds to the \(\gamma \varepsilon\) allotropic phase boundary in the binary Fe–Mn system. Based on these data they modified Breedis and Kaufman’s diagram using the same approach. The calculated phase diagram can well explain phase stability in binary Fe–Mn alloys. Thus, accurate evaluation of the difference in free energies between the \(\varepsilon\), \(\gamma\) phases in ternary iron alloys provides detailed information on the phase stability of Fe–Mn–Al alloys if thermodynamic properties for each component are available. Adopting a regular solution model, \(\Delta G^{\gamma \rightarrow \varepsilon}(=\Delta G^{\varepsilon \rightarrow \gamma}(\varepsilon)\) for Fe–Mn–Al alloys will be expressed in a first approximation as:

\[
\Delta G^{\gamma \rightarrow \varepsilon} = X_{Fe}\Delta G_{Fe}^{\gamma \rightarrow \varepsilon} + X_{Mn}\Delta G_{Mn}^{\gamma \rightarrow \varepsilon} + X_{Al}\Delta G_{Al}^{\gamma \rightarrow \varepsilon} + X_{Fe}X_{Mn}(X_{Fe} + X_{Mn})^{-1}\Delta G_{FeMn}^{\gamma \rightarrow \varepsilon} + X_{Fe}X_{Al}(X_{Fe} + X_{Al})^{-1}\Delta G_{FeAl}^{\gamma \rightarrow \varepsilon} + X_{Mn}X_{Al}(X_{Mn} + X_{Al})^{-1}\Delta G_{MnAl}^{\gamma \rightarrow \varepsilon} \quad \text{(1)}
\]

where, \(X_i\): mole fraction of the component, \(i\)

\(\Delta G^{\varepsilon \rightarrow \gamma}\): fcc–hcp free energy difference of the pure component, \(i\),

\(\Delta G_{ij}^{\varepsilon \rightarrow \gamma}\): interaction energy parameter between the components \(i\) and \(j\).

\(\Delta G_{Fe}^{\varepsilon \rightarrow \gamma}\) was referred to Dew-Hughes and Kaufman,\(^{19}\) \(\Delta G_{Mn}^{\varepsilon \rightarrow \gamma}\) to Breedis and Kaufman,\(^{10}\) \(\Delta G_{Al}^{\varepsilon \rightarrow \gamma}\) to Kaufman and Breedis,\(^{20}\) and \(\Delta G_{MnAl}^{\varepsilon \rightarrow \gamma}\) to Ishida and Nishizawa.\(^{5}\) The value of \(\Delta G_{MnMn}^{\varepsilon \rightarrow \gamma}\) is not available at present. However, the experimentally determined \(T_{\gamma \rightarrow \varepsilon}\) temperatures in the Fe–17Mn alloy system including 0.3 to 1.3 wt\% Al\(^{30}\) revealed that contribution of the binary Mn–Al parameter term, \(X_{MnX_{Al}}(X_{Mn} + X_{Al})^{-1}\Delta G_{MnAl}^{\varepsilon \rightarrow \gamma}\), in Eq. (1) to the evaluation of \(\Delta G^{\varepsilon \rightarrow \gamma}\) is very small compared to the other terms in the same equation. Calculated errors of \(\Delta G^{\varepsilon \rightarrow \gamma}\) and \(T_{\gamma \rightarrow \varepsilon}\) by neglecting the Mn–Al terms were found to be about ±25 J/mol and ±20 deg, respectively. Thus, \(\Delta G^{\varepsilon \rightarrow \gamma}\) for Fe-rich Fe–Mn–Al alloys can be expressed as:

\[
\Delta G^{\varepsilon \rightarrow \gamma} = X_{Fe}\Delta G_{Fe}^{\varepsilon \rightarrow \gamma} + X_{Mn}\Delta G_{Mn}^{\varepsilon \rightarrow \gamma} + X_{Al}\Delta G_{Al}^{\varepsilon \rightarrow \gamma} + X_{Fe}X_{Mn}(X_{Fe} + X_{Mn})^{-1}\Delta G_{FeMn}^{\varepsilon \rightarrow \gamma} + X_{Fe}X_{Al}(X_{Fe} + X_{Al})^{-1}\Delta G_{FeAl}^{\varepsilon \rightarrow \gamma} + X_{Mn}X_{Al}(X_{Mn} + X_{Al})^{-1}\Delta G_{MnAl}^{\varepsilon \rightarrow \gamma} \quad \text{(2)}
\]

The same treatment was also applied to the calculation of \(\Delta G^{\gamma \rightarrow \varepsilon}\). Thermodynamic parameters used in the present calculations for the \(\gamma \varepsilon\) equilibrium are summarized in Table 4. The parameters necessary for the calculation of the \(\alpha /\gamma\) equilibrium are reported previously.\(^{19}\) In the present calculation the value of \(\Delta G_{Fe}^{\gamma \rightarrow \varepsilon}\) including magnetic contribution in the temperature range between 500 and 1 043 K was employed. Determination of the \(T_{\gamma \rightarrow \varepsilon}\) and \(T_{\varepsilon \rightarrow \gamma}\) lines were made by setting \(\Delta G^{\gamma \rightarrow \varepsilon}\) and \(\Delta G^{\varepsilon \rightarrow \gamma}\) to zero respectively for the alloy system containing constant aluminum content at each temperature. Hence, the \(T_{\alpha \rightarrow \gamma}\) and \(T_{\gamma \rightarrow \alpha}\) lines in the Fe–Mn–Al system, respectively. These figures revealed that the Fe–Mn–2Al alloy system has a much narrower region of the \(\varepsilon\) phase, i.e., at 0 K an addition of 2 wt\% Al to the Fe–Mn alloy system shifts the \(\gamma /\varepsilon\) boundary from approximately 32 to 24 wt\% Mn. Thus, the addition of 2 wt\% Al to Fe–Mn alloys gives birth to a much restricted region of \(\varepsilon\) phase stability. Further, this calculation shows that on increasing aluminum content more than about 2 wt\% there exists

---

Table 4. Thermodynamic properties for the calculation of \(\gamma /\varepsilon\) equilibrium.

<table>
<thead>
<tr>
<th>(\Delta G^{\gamma \rightarrow \varepsilon}) (J/mol)</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>(\Delta G_{Fe}^{\gamma \rightarrow \varepsilon})</td>
<td>19</td>
</tr>
<tr>
<td>(\Delta G_{Mn}^{\gamma \rightarrow \varepsilon})</td>
<td>18</td>
</tr>
<tr>
<td>(\Delta G_{Al}^{\gamma \rightarrow \varepsilon})</td>
<td>20</td>
</tr>
<tr>
<td>(\Delta G_{FeMn}^{\gamma \rightarrow \varepsilon})</td>
<td>5</td>
</tr>
<tr>
<td>(\Delta G_{FeAl}^{\gamma \rightarrow \varepsilon})</td>
<td>5</td>
</tr>
<tr>
<td>(\Delta G_{MnAl}^{\gamma \rightarrow \varepsilon})</td>
<td>5</td>
</tr>
</tbody>
</table>

---
no region for stable ε phase in the Fe–Mn–Al alloys. The calculated (T_s−X) curve shows that the occurrence of the γ→ε transformation is strongly dependent on manganese content particularly below room temperature.

On cooling to room temperature alloy A2 exhibited a fully austenitic structure whereas alloy A0 showed the γ→ε transformation at 380 K. This can be expected from the calculated phase diagram shown in Fig. 12. Thus, additions of aluminum to the Fe–Mn alloys undergoing the γ→ε transformation significantly decrease the Ms−Mf temperature. In addition, the calculated diagram of Fe–Mn–Al alloys shows that γ phase stability against the γ→ε transformation below room temperature increases with increasing both manganese and aluminum contents. However, increasing amount of aluminum enlarges the α phase region particularly above room temperature. Therefore, stable austenitic alloys are obtained with increasing amount of manganese and limited amount of aluminum.

4.2. Relation between Stacking Fault Energy and Deformation Mode

In low stacking fault energy alloys such as austenitic Fe–Ni–Cr, Fe–Mn and Fe–Mn–C, the yielding and following strain hardening behaviors are closely related to the formation of the γ twins, α and α' phases. The present results show that the differences in the yield stress and work hardening rate of austenitic Fe–Mn–Al alloys are significantly affected by the occurrences of the γ→ε transformation and deformation twinning. This is particularly prominent for alloys A2 and A4 at low temperatures. When plastic deformation in tension was limited to about 2% at which internal stress produced by plastic deformation can be roughly neglected, alloys A2 and A4 produced the ε phase and the γ twins respectively as major deformation products at 77 K. The yield stress at zero plastic strain of alloy A2 was lower than that of alloy A4. The work hardening rate of alloy A2 was larger than that of alloy A4. A similar case was observed for austenitic Fe–Mn–C alloys. An Fe–20Mn–0.35C having the lower stacking fault energy than an Fe–13Mn–1C alloy favors the γ→ε transformation more than deformation twinning which predominantly takes place in the latter alloy. Further, the former alloy gave birth to lower yield stress with higher work hardening rate than the latter alloy. Thus, a correlation between deformation mode and predicted stacking fault energy in austenitic Fe–Mn–Al alloys provides further evidence for the austenitic phase stability against the γ→ε transformation and deformation twinning.

According to a thermodynamic model, the fault energy of fcc alloys, \( \gamma_{\text{eff}} \), is expressed as \(^{(2)}\):

\[
\gamma_{\text{eff}} = 2p \Delta G^{\gamma \rightarrow \alpha} + 2a^{\alpha^\prime} \]

where \( p \) is the density of atoms in a close packed plane in moles per unit area, which was determined to be 2.9×10\(^{-4}\) mol/m\(^2\) for the fcc (111) plane using the lattice constant of the austenitic Fe–(20 to 30)%Mn–(2 to 4)%Al alloys, \( a^{\alpha^\prime} \) and \( \Delta G^{\gamma \rightarrow \alpha} \) is the γ→α interfacial energy. The interfacial energy is very weakly temperature dependent\(^{(20)}\) so that it can be safely regarded as a constant. Since the value of \( a^{\alpha^\prime} \) is not available for Fe–Mn–Al alloys, \( a^{\alpha^\prime} = 10 \, \text{mJ/m}^2 \) for austenitic Fe–(16 to 18)%Cr–(12 to 14)%Mn was employed. The values of \( \Delta G^{\gamma \rightarrow \alpha} \) are obtained by using Eq. (2). The calculated error of \( \gamma_{\text{eff}} \) by neglecting the interaction parameter, \( \Delta E^{\text{inter}} \), was estimated to be about ±1 erg/cm\(^2\) (=1 mJ/m\(^2\)). Thus, calculated values of \( \gamma_{\text{eff}} \) were plotted as a function of aluminum content for the 20 and 30% Mn alloys in Fig. 13. It shows that additions of aluminum significantly increase the stacking fault energy. When the austenite stacking fault energy approaches zero, ε martensite, which is essentially composed of stacking faults on every alternate fcc (111) plane, is easily formed under tensile loading. TEM observations showed that extended dislocations and ε bands existed in alloy A2 (Fe–20Mn–2Al) deformed to a plastic strain of 0.02 at 77 K whereas no extended dislocations were observed for alloy A4 (Fe–20Mn–4Al) deformed under the same condition. In the as-quenched Fe–30Mn alloy small needles of the ε phase were observed under a TEM observation.\(^{(4)}\) Also in the present experiment X-ray diffraction analysis for alloy B0 (Fe–30Mn) after tensile fracture showed the presence of a slight amount of ε phase which was not found for the other austenitic alloys such as alloys A4, B2, B4 and B7. These results.

Fig. 12. Calculated (T_s−X) curves of γ→ε and γ→α for (a) the Fe–Mn system referring to Ishida and Nishi-zawa (γ→α)\(^{(1)}\) and Breedis and Kasumann (γ→α)\(^{(1)}\) and (b) the Fe–Mn–2Al system.

Fig. 13. Calculated stacking fault energy vs. aluminum content for the austenitic 20 and 30 % Mn alloys at (a) 295 K and (b) 77 K.

875
imply that in the Fe–Mn–Al alloys the stress-induced \( \gamma \rightarrow \varepsilon \) transformation predominantly occurs in the Fe–Mn–Al alloys for those having the stacking fault energy approximately smaller than 20 erg/cm\(^2\). Adler et al. showed that austenitic Fe–Mn–C alloys of which the stacking fault energy exceeds about 20 erg/cm\(^2\) undergo deformation twinning,\(^{23} \) which is also consistent with the present results.

4.3. Effect of Strain-induced Transformation and Deformation Twinning on Mechanical Properties

In stable fully austenitic Fe–Mn alloys solid solution hardening by manganese does not contribute very greatly to the overall strength.\(^{25}\) In addition, the fact that the yield stress at zero plastic strain did not vary considerably with aluminum content for the 30 \% Mn alloys confirms negligible solid-solution hardening effects on the analysis of the true stress–strain curves in the austenitic Fe–Mn–Al alloys. Thus, the influence of the phase transformation or deformation twinning on work-hardening rate can be better explained by comparing the alloys at the same temperature.

The strain hardening increment, \( \sigma - \sigma_y \) vs. plastic tensile strain, \( \varepsilon_p \) at 77 K is compared for alloys A4 and B4, and SUS 304 in Fig. 14. At low strains \( (\varepsilon_p \leq 0.1) \) alloy A4 deforms primarily by twinning whereas alloy B4 deforms by slip. However, the strain hardening behaviors are almost identical. This seems to reflect that the small amount of twins induced does not affect the work-hardening rate. At higher strains \( (\varepsilon_p > 0.1) \) alloy B4 showed downward curvature in the strain hardening curve while alloy A4 showed an almost linear increase in the flow stress mainly at high strains. At this stage only alloy A4 exhibited the \( \gamma \rightarrow \alpha' \) strain-induced martensitic transformation with concurrent formation of twins, and the volume fraction of \( \alpha' \) martensite amounts to roughly 50 \% after the fracture, while alloy B4 exhibited only deformation twinning. Thus, formation of \( \alpha' \) martensite during deformation can give rise to a higher work hardening rate than twinning. SUS 304 steel which produces a great amount of \( \alpha' \) martensite during plastic deformation at 77 K showed a considerably high work-hardening rate.

The elongation of alloy A4 at 77 K (90 \%) was larger than that at 295 K (67 \%). The increase in tensile elongation of alloy A4 with the formation of \( \alpha' \) martensite is analogous to that observed in ausformed TRIP steels.\(^{36–29} \) On the contrary, temperature dependence of tensile elongation in Fe–20Mn–4Cr–0.5C alloy showed that formation of the strain-induced \( \varepsilon \) martensite which can give rise to a brittle behavior reduces the homogenous elongation with an increase in its volume fraction.\(^{20} \) Alloy A2 exhibits the \( \gamma \rightarrow \varepsilon \) transformation from the early stage of deformation, however, the occurrences of the \( \gamma \rightarrow \alpha' \) as well as \( \gamma \rightarrow \varepsilon \rightarrow \alpha' \) transformations, which are predominant at low temperatures, are considered to suppress a decrease in elongation probably due to the stress-relaxation effect.\(^{20} \)

5. Conclusions

Structure and mechanical properties of austenitic Fe–(20 and 30)Mn–(0 to 7)Al alloys have been investigated in the temperature range between 77 and 295 K. An addition of 1.7 \%Al to an Fe–20Mn alloy significantly decreased the \( \gamma \rightarrow \varepsilon \) martensitic transformation temperature from 380 K to below room temperature. In the 20 \% Mn alloys increasing amount of aluminum suppresses the \( \gamma \rightarrow \varepsilon \) transformation under tensile loading and gives rise to the formation of \( \gamma \) twins leading to a lower strain hardening and increases in ductility and the yield stress which are particularly prominent at low temperatures. In the 30 \% Mn alloys additions of aluminum do not affect either the strain hardening rate nor yield stress but increased ductility at 77 K. X-ray diffraction analysis revealed that the formation of a slight amount of \( \varepsilon \) martensite which was identified for specimens deformed to fracture at 77 and 295 K is also suppressed by the addition of aluminum. The observation of an Fe–30Mn–4Al alloy deformed to fracture at 77 K with TEM showed the occurrence of deformation twinning. Comparison of the experimental results with a computed dependence of the stacking fault energy on aluminum content in the 20 and 30 \% Mn alloys shows that the stress-induced \( \varepsilon \) martensite is no longer produced for the alloys having the stacking fault energy larger than about 20 erg/cm\(^2\). These alloys produce the deformation twinning in stead of the \( \gamma \rightarrow \varepsilon \) transformation.

Acknowledgments

The authors would like to acknowledge the experimental support on TEM observations by Mr. K. Tagawa. They also acknowledge the supply of the material by R & D Laboratories-II, Nippon Steel Corporation.

This work was supported by the Grant-in-Aid for Scientific Research from the Ministry of Education, Science and Culture, Japan.

REFERENCES

19) D. Dow-Hughes and L. Kaufman: CALPHAD, 3 (1979), 175.