Effect of Prior Deformation of Austenite on the $\gamma \rightarrow \epsilon$ Martensitic Transformation in Fe–Mn Alloys

Kaneaki Tsuzaki*, Shin-ichi Fukasaku**, Yo Tomota***, and Tadashi Maki*

The influence of prior deformation of austenite ($\gamma$) on the formation of hcp ($\epsilon$) martensite in Fe–Mn binary alloys during subsequent cooling was investigated. The formation of bcc ($\alpha'$) lath martensite in deformed austenite was also examined for comparison. In both of the martensitic transformations to $\alpha'$ in the Fe–9%Mn alloy and to $\epsilon$ in the Fe–16%Mn and Fe–24%Mn alloys, the transformation start temperature ($M_s$) monotonously decreases with increasing the reduction of austenite at 773 K. It is demonstrated that the decrease in $M_s$ temperature for the $\gamma \rightarrow \epsilon$ transformation is much larger than that for the $\gamma \rightarrow \alpha'$ transformation. The $M_s$ temperature of the Fe–24%Mn alloy is lowered over 80 K by 40% reduction and the structure becomes fully austenitic at room temperature, whereas the decrease in $M_s$ temperature of the Fe–9%Mn alloy is 16 K. The decrease in $M_s$ temperature by the prior deformation of austenite is quantitatively accounted for by the strengthening of austenite and the extra energy required for the transformation dislocations to advance through forest dislocations in austenite. The marked stabilization of austenite against the $\gamma \rightarrow \epsilon$ martensitic transformation is mainly associated with the smaller increase in chemical driving force for the $\gamma \rightarrow \epsilon$ transformation with decreasing temperature than that for the $\gamma \rightarrow \alpha'$ transformation.

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1. Introduction

The influence of prior deformation of austenite ($\gamma$) on the martensitic transformation during subsequent cooling has been well studied for the case of bcc ($\alpha'$) lenticular martensite in ferrous alloys. It has been reported that the martensitic transformation start temperature ($M_s$) increases by the austenite predeformation$^{1(6),7}(7)}$. It is generally believed that dislocations in austenite act as favorable nucleation sites and thus $M_s$ rises by the prior deformation of austenite. Recently, Kajiwara$^{7}$ has proposed an alternative explanation for the effect of austenite predeformation. He$^{7}$ considered that the plastic accommodation in austenite is a vital step in the martensitic nucleation and the dislocations introduced by the prior deformation serve the plastic accommodation of the shape strain of martensite. However, the role of dislocations introduced into austenite in the martensitic nucleation has not been clarified.

It has been well established that fcc ($\gamma$) → hcp ($\epsilon$) martensitic transformation proceeds by one $a/6\langle 112 \rangle$ Shockley partial dislocation advancing on every second (111) austenite plane$^{8(9)$. The $\gamma \rightarrow \epsilon$ martensitic transformation is simple at least compared with $\gamma \rightarrow \alpha'$ martensitic transformation. Thus, the $\gamma \rightarrow \epsilon$ martensitic transformation would be a good example to examine the effect of prior deformation of austenite on subsequent transformation and to clarify the role of dislocations in the martensitic nucleation. There have been, however, a few investigations$^{9(10)}$ for the $\gamma \rightarrow \epsilon$ martensitic transformation. Tamura et al.$^{10}$ first reported that, in contrast to the case of $\alpha'$ lenticular martensite, the $M_s$ temperature for the $\gamma \rightarrow \epsilon$ transformation decreased monotonously with increasing the prior deformation of austenite in an Fe–15%Cr–13%Ni alloy. Tomota et al.$^{10}$ reported that the prior deformation of austenite suppressed the $\gamma \rightarrow \epsilon$ martensitic transformation in Fe–high Mn alloys. However, the reason for the decrease in $M_s$ has not been clarified. Tomota et al.$^{10}$ also reported that the mechanically stabilized austenite showed the good tensile property. The mechanical stabilization of austenite prevented the intergranular fracture at 77 K in the Fe–17Mn alloy and enhanced the uniform elongation in the Fe–25Mn alloy$^{10}$. Thus, the prior deformation of austenite is also important, from a technological viewpoint, to improve the mechanical properties of high-Mn steels.

In the present study, the effect of prior deformation of austenite on the subsequent $\gamma \rightarrow \epsilon$ martensitic transformation in Fe–Mn binary alloys was examined in order to clarify the role of dislocations introduced into austenite in the martensitic nucleation and to give a quantitative analysis to the decrease in $M_s$. In the Fe–Mn binary alloys $\alpha'$ lath martensite forms when the Mn content is less than about 10 mass%. In order to make clear the effect of type of martensite, the formation of $\alpha'$ lath martensite in deformed austenite was also examined.

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II. Experimental Procedure

The chemical compositions of three Fe-Mn binary alloys used in the present study are shown in Table 1. ε martensite forms in the Fe–16Mn and Fe–25Mn alloys, while α' lath martensite forms in the Fe–9Mn alloy. An Fe–23Ni binary alloy showing the formation of α' lath martensite was also used. All of the alloys were prepared by induction melting. After hot swaging and hot rolling, the alloys were homogenized at 1473 K for 36 ks. The alloys were then cut and machined into cylindrical specimens 8 mm in diameter and 12 mm in length.

Heat treatment and compressive deformation (prior deformation of austenite) of specimens were performed by means of a hot deformation simulator (Thermomaster-Z) with a laser equipment measuring the diameter of a specimen. A schematic diagram of the heat treatment is shown in Fig. 1. The heat treatment was performed in an argon gas atmosphere for the Fe–Mn alloys and in vacuum for the Fe–Ni alloy. The specimens were first austenitized at 1373 K for 0.6 ks and cooled to 773 K. After holding at 773 K for 300 s, the specimens were compressed (ausformed) at a strain rate of 1 × 10^-3/s. The amount of reduction varied from 0 to 40%. Subsequently, the specimens were rapidly cooled with He gas to a temperature which was about 100 K higher than the Ms temperature of the non-deformed specimen (Ms°), in order to avoid the recovery of deformed austenite. The specimens were then cooled at a rate of 0.2 K/s to room temperature.

The transformation behavior was measured by detecting the diameter change of a specimen with the laser equipment. The amount of ε martensite was also determined by dilatometry according to the fact that the γ → ε transformation in Fe–high Mn alloys accompanies 0.70% contraction in specimen length. The Ms temperatures of the non-deformed specimens (Ms°) are shown in Table 1. The Ms° temperatures of all the four alloys are higher than room temperature.

Optical microstructure observation was carried out on a center plane parallel to the compressive axis of specimens of the Fe–16Mn alloy. After electrolytic polishing, the specimens were etched with a 100 ml saturated Na_2S_2O_3 water solution + 10 g K_2S_2O_5 to reveal ε martensite structure.

III. Results

Figure 2 shows the change in M_s temperature with increasing the reduction of austenite at 773 K. The difference in M_s temperature between the non-deformed and deformed specimens is plotted against the prior reduction of austenite in Fig. 3. In all of the alloys, M_s temperature decreases monotonously with increasing the reduction of austenite. It should be emphasized that the decrease in M_s temperature is much larger for the γ → ε transformation (the Fe–16Mn and the Fe–24Mn alloys) in comparison with that for the γ → α' transformation (the Fe–9Mn and Fe–23Ni alloys). In the Fe–16Mn alloy, the M_s temperature of the 39% deformed specimen is 40 K lower than that of non-deformed specimen. The M_s temperature of the Fe–24Mn alloy is lowered over 80 K.

Table 1 Chemical compositions (mass%), type of martensite, and M_s temperatures of non-deformed specimens (M_s°) of the alloys.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Ni</th>
<th>Type of martensite</th>
<th>M_s°</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe–24Mn</td>
<td>0.002</td>
<td>0.02</td>
<td>24.4</td>
<td>0.005</td>
<td>0.008</td>
<td>—</td>
<td>ε (hcp)</td>
<td>390 K</td>
</tr>
<tr>
<td>Fe–16Mn</td>
<td>0.006</td>
<td>0.01</td>
<td>15.7</td>
<td>0.005</td>
<td>0.005</td>
<td>—</td>
<td>ε (hcp)</td>
<td>447 K</td>
</tr>
<tr>
<td>Fe–9Mn</td>
<td>0.0023</td>
<td>0.0005</td>
<td>9.28</td>
<td>0.0008</td>
<td>0.003</td>
<td>—</td>
<td>α' (bcc)</td>
<td>602 K</td>
</tr>
<tr>
<td>Fe–23Ni</td>
<td>0.0015</td>
<td>0.0054</td>
<td>0.054</td>
<td>0.0011</td>
<td>0.0007</td>
<td>23.4</td>
<td>α' (bcc)</td>
<td>481 K</td>
</tr>
</tbody>
</table>

![Fig. 1 Schematic diagram of the heat treatment and compressive deformation of specimens.](image1)

![Fig. 2 Change in M_s temperature by the reduction of austenite at 773 K.](image2)
by 39% reduction of austenite and the structure becomes fully austenitic at room temperature. On the other hand, the decrease in $M_t$ temperature for the $\alpha'$ lath martensite is 16 K even at 40% reduction of austenite.

Figure 4 shows the amount of $\varepsilon$ martensite at room temperature against the prior reduction of austenite at 773 K in the Fe–16Mn and Fe–24Mn alloys. In the Fe–24Mn alloy, the amount of $\varepsilon$ martensite decreases markedly with increasing reduction of austenite, whereas that is almost constant in the Fe–16Mn alloy. On the other hand, the structures were fully martensitic at room temperature in the Fe–9Mn and Fe–23Ni alloys, regardless of the reduction of austenite.

The effect of prior deformation of austenite on the $\gamma\rightarrow\varepsilon$ martensitic transformation curve is shown in Fig. 5(a) for the Fe–16Mn alloy and in Fig. 5(b) for the Fe–24Mn alloy. In the non-deformed specimen of the Fe–16Mn alloy, the amount of $\varepsilon$ martensite sharply increases just below the $M_t$ temperature and almost saturates at around 400 K. The increase in the amount of $\varepsilon$ martensite is very small in the temperature range between 400 K and room temperature. The slope of transformation curve just below $M_t$ temperature becomes small by the prior deformation of austenite. In the case of the 39% deformed specimen, the amount of $\varepsilon$ martensite increases gradually from the $M_t$ temperature to room temperature. The decrease in the amount of $\varepsilon$ martensite by the austenite predeformation is apparent when it is compared at a higher temperature, i.e., 400 K, although the amount of $\varepsilon$ martensite is almost constant at room temperature. In the Fe–24Mn alloy (Fig. 5(b)), the amount of $\varepsilon$ martensite increases gradually from the $M_t$ temperature to room temperature even for the non-deformed specimen.

Figure 6 shows the microstructural change by the prior deformation of austenite in the Fe–16Mn alloy. The amount of $\varepsilon$ martensite of 6% deformed specimen (Fig. 6(b)) is somewhat larger than that of the non-deformed specimen (Fig. 6(a)). The large reduction of austenite makes the $\varepsilon$ plates thin and curved, as shown in Figs. 6(c) and 6(d). Since the amount of $\varepsilon$ martensite of the 18% or 39% deformed specimen is almost the same as that of

Fig. 3 Difference in $M_t$ temperature between non-deformed and deformed specimens as a function of the reduction of austenite at 773 K. $M_t^1$ and $M_t^2$ mean $M_t$ temperatures of deformed and non-deformed specimens, respectively.

Fig. 4 Change in the amount of $\varepsilon$ martensite at room temperature by the reduction of austenite at 773 K.

Fig. 5 Effect of the prior deformation of austenite at 773 K on the $\gamma\rightarrow\varepsilon$ martensitic transformation curve during the subsequent continuous cooling: (a) Fe–16Mn and (b) Fe–24Mn.
Effect of Prior Deformation of Austenite on the \( \gamma \rightarrow \varepsilon \) Martensitic Transformation in Fe–Mn Alloys

![Fig. 6 Optical micrographs showing the change in \( \varepsilon \) martensite structure by the prior deformation of austenite at 773 K in the Fe–16Mn alloys; (a) non-deformed, (b) 6\%, (c) 18\%, and (d) 39\% deformed specimens.](image)

non-deformed specimen (Fig. 4), the finer \( \varepsilon \) plates in the deformed specimens apparently indicate the increase in the number of \( \varepsilon \) plates by the prior deformation of austenite.

### IV. Discussion

It was reported in Fe–9\%Cr–1.1\%C and Fe–10\%Cr–1.1\%C alloys that the \( M_s \) temperature of \{225\} plate \( \alpha' \)-martensite also monotonically decreased with increasing prior reduction of austenite\(^7\). Kajiwara\(^9\) has proposed that the plastic accommodation in austenite is an indispensable condition in the martensitic nucleation and any factors impeding the plastic accommodation suppress the transformation. According to the above model, Kajiwara\(^9\) explained the monotonous decrease in \( M_s \) by the prior reduction of austenite as follows. A large number of dislocations introduced in austenite by the prior deformation acts as a barrier to moving dislocations in austenite for the plastic accommodation. In order to overcome such a barrier, a larger chemical driving force (namely, the larger chemical free energy change for the transformation) is required, resulting in the decrease in \( M_s \) by the prior deformation. However, we must point out that the chemical driving force itself does not give the driving shear stress acting on the dislocations in austenite for the plastic accommodation. Such a driving shear stress arises only from the elastic strain due to the formation of martensite plate. Therefore, the driving shear stress is almost constant irrespective of temperature and the undercooling does not have any effects on the plastic accommodation in austenite, as far as the shape strain of martensite is independent of formation temperature. Moreover, it is very difficult to account for how the plastic accommodation in austenite requiring the formation of martensite can control the martensitic nucleation. Furthermore, the plastic accommodation in austenite has not been reported in the case of the \( \gamma \rightarrow \varepsilon \) martensitic transformation. Thus, the alternative model or analysis for the decrease in \( M_s \) by the prior deformation of austenite is required at least in the \( \gamma \rightarrow \varepsilon \) transformation.

It has been well established that the \( \gamma \rightarrow \varepsilon \) martensitic transformation occurs by motion of \( \alpha/6\langle 112 \rangle \) Shockley partial dislocations\(^8\). With one \( \alpha/6\langle 112 \rangle \) Shockley partial dislocation advancing on every second \{111\} fcc plane, a fcc structure transforms into a hcp structure. Dislocations introduced in austenite by the prior deformation must act as obstacles for the movement of Shockley partial transformation dislocations. In order to overcome these obstacles, an extra energy is required. This extra energy can be supplied by chemical driving force. Thus, the \( M_s \) temperature is lowered by the prior deformation of austenite. Consequently, in the present analysis, we consider the interaction of the advancing transformation dislocations themselves with forest dislocations introduced in austenite by the prior deformation. Olson and Cohen\(^1\) have proposed that the initial stage of nucleation in the \( \gamma \rightarrow \alpha' \) martensitic transformation occurs with one Shockley partial dislocations advancing on every three \{111\} austenite plane. Therefore, the decrease in \( M_s \) temperature for \( \alpha' \)-lath martensite may be also discussed from the same point of view as that for \( \varepsilon \) martensite, i.e., the extra energy for Shockley partial transformation dislocations to advance through
forest dislocations.

The critical or resistive shear stress for dislocations to advance in a matrix by cutting forest dislocations is expressed by the Bailey-Hirsch relation\(^{(14)}\) as

\[ \tau = \tau_0 + \alpha G b \sqrt{\rho}, \]  
where \(\tau_0\) is the friction stress, \(\alpha\) a constant, \(G\) the shear modulus, \(b\) the Burgers vector, and \(\rho\) the dislocation density. It is apparent that the work hardening or the increase in dislocation density of austenite by the prior deformation increases the resistive shear stress for the transformation during subsequent cooling. The stress-strain curves of austenite at 773 K are shown in Fig. 7. The work hardening is larger in the Fe-16Mn and Fe-24Mn alloys than in the Fe-9Mn and Fe-23Ni alloys. This may be one of the reasons for the larger decrease in \(M_t\) temperature for the \(\gamma \rightarrow \varepsilon\) transformation (Figs. 2 and 3).

The driving shear stress acting on Shockley partial transformation dislocations should arise from the chemical driving force and can be derived as follows. As proposed by Olson and Cohen\(^{(12)(13)}\), we here consider the special array of dislocations as the nucleating defects. A schematic illustration of the nucleating defect is shown in Fig. 8. Here, the thickness of the defect is \(nd\), where \(n\) is the number of (111) austenite plane in the defect and \(d\) spacing between (111) austenite planes. Shockley partial dislocations are arrayed on every \(m\) (111) austenite plane. The value of \(m\) must be 2 for the \(\gamma \rightarrow \varepsilon\) transformation, while that is 3 for the \(\gamma \rightarrow \alpha'\) transformation according to Olson-Cohen’s model\(^{(12)(13)}\). When the free energy change per \(1 \times 1 \times nd\) volume by the advancement of Shockley partial transformation dislocations is expressed as \(\Delta W\), the driving shear stress \(\tau_{ch}\) acting on the dislocations is written as

\[ \tau_{ch} = -\frac{\Delta W}{nb/m}, \]  
where \(n/m\) means the number of advancing Shockley partials in the defect. This equation means that the free energy change \(\Delta W\) is consumed by the work necessary for the advancement of transformation dislocations. The free energy change \(\Delta W\) of the defect was given by Olson and Cohen\(^{(12)(13)}\) as

\[ \Delta W = nd[\Delta G_{ch} + \Delta G_{el}] + 2\gamma, \]  
where \(\Delta G_{ch}\) is the chemical free energy change per unit volume for transformation from austenite to martensite, \(\Delta G_{el}\) the elastic strain energy per unit volume, and \(\gamma\) the interfacial energy per unit area for top and bottom planes of the defect. Substituting eq. (3) into eq. (2), the driving shear stress can be expressed using the term of chemical free energy change as

\[ \tau_{ch} = -\left(\frac{md}{b}\right)[\Delta G_{ch} + \Delta G_{el}] - 2\gamma, \]  
At \(M_t\) temperature, the driving shear stress \(\tau_{ch}\) must be equal to the resistive shear stress \(\tau\) given by eq. (1). The resistive shear stress for non-deformed austenite can be assumed to be \(\tau_0\). The elastic strain energy \(\Delta G_{el}\) and the interfacial energy \(\gamma\) of the predeformed specimens may be the same as those of the non-deformed specimens. Thus, the extra energy \(\Delta G_{ex}\) required for the transformation of deformed austenite can be written using eqs. (1) and (4) as

\[ \Delta G_{ex} = \Delta G_{ch}(M_t^\varepsilon - M_t^\alpha) = -\left(\frac{b}{md}\right)\alpha G b \frac{1}{2} \rho_d \]  
where \(M_t^\varepsilon\) and \(M_t^\alpha\) are \(M_t\) temperatures of deformed austenite and non-deformed austenite, respectively, and \(\rho_d\) is the dislocation density of deformed austenite.

Figure 9 shows the chemical free energy change \(\Delta G_{ch}\) for the \(\gamma \rightarrow \varepsilon\) transformation in the Fe-16Mn and Fe-24Mn alloys as a function of temperature\(^\dagger\). \(\Delta G_{ch}\)-tem-
Fig. 9 \( \Delta G_{\text{ex}} \)-temperature curves for the \( \gamma \rightarrow \varepsilon \) transformation in the Fe-16Mn and Fe-24Mn alloys and for the \( \gamma \rightarrow \alpha' \) transformation in the Fe-9Mn and Fe-23Ni alloys.

Fig. 10 Relation between the extra energy (\( \Delta G_{\text{ex}} \)) and the amount of work hardening of austenite at 773 K (\( \sigma - \sigma_{0.2} \)).

The effect of prior deformation on the \( \gamma \rightarrow \varepsilon \) martensitic transformation in Fe-Mn alloys is also shown. The arrows in Fig. 9 indicate the temperature (\( M^* \)) of the non-deformed austenite. It should be noted that the slope of curve at \( M^* \) is different among the alloys. The slope is smaller in the Fe-16Mn and Fe-24Mn alloys than in the Fe-9Mn and Fe-23Ni alloys. For instance, \( \Delta G_{\text{ex}}/\Delta T \) at \( M^* \) is 0.36 MJ m\(^{-3}\) K\(^{-1}\) (0.60 cal mol\(^{-1}\) K\(^{-1}\)) for the Fe-24Mn alloy and is 0.82 MJ m\(^{-3}\) K\(^{-1}\) (1.37 cal mol\(^{-1}\) K\(^{-1}\)) for the Fe-9Mn alloy. This clearly indicates that, in order to obtain a given extra energy, the larger undercooling is required in the Fe-16Mn and Fe-24Mn alloys than in the Fe-9Mn and Fe-23Ni alloys. It can be thus concluded that this small increase in chemical driving force for the \( \gamma \rightarrow \varepsilon \) transformation with decreasing temperature results in the marked mechanical stabilization of austenite against the \( \gamma \rightarrow \varepsilon \) martensitic transformation (Figs. 2 and 3).

Using the relation of \( \sigma = M \tau \), where \( M \) is the Taylor factor\(^{(10)} \), the extra energy \( \Delta G_{\text{ex}} \) is expressed by the amount of work hardening of austenite as

\[
\Delta G_{\text{ex}} = \Delta G_{\text{ch}}(M^*_\gamma) - \Delta G_{\text{ch}}(M^*_\alpha)
\]

\[
= -b/(mdM)(\sigma - \sigma_{0.2})
\]

(6)

where \( \sigma \) is the applied flow stress of austenite and \( \sigma_{0.2} \) is the 0.2% proof stress at 773 K. \( m \) is 2 for the \( \gamma \rightarrow \varepsilon \) transformation and 3 for the \( \gamma \rightarrow \alpha' \) transformation. \( M \) is 3.06 for fcc polycrystalline\(^{(10)} \), \( b \) is \( a/6(112) \) and \( d \) is \( a/\sqrt{3} \) for the present cases (\( a \) is the lattice constant of austenite). Thus, the value of \( b/(mdM) \) becomes 0.12 for the \( \gamma \rightarrow \varepsilon \) transformation and 0.08 for the \( \gamma \rightarrow \alpha' \) transformation.

Figure 10 shows the relation between \( \Delta G_{\text{ex}} \) and \( (\sigma - \sigma_{0.2}) \). Although the extrapolation of data does not go through \( \Delta G_{\text{ex}} = 0 \) at \( (\sigma - \sigma_{0.2}) = 0 \), the value of slope is very close to the calculated one. This suggests that the mechanical stabilization of austenite can be quantitatively accounted for by the strengthening of austenite and the extra energy for the transformation dislocations to advance through forest dislocations. In order to reconfirm the present analysis, the dislocation density of austenite is estimated from the value of extra energy \( \Delta G_{\text{ex}} \) using eq. (5). The results are shown in Table 2. For the calculation, \( \alpha = 0.5 \), \( G = 8 \times 10^{10} \) N m\(^{-2}\) and \( a = 3.6 \times 10^{-10} \) m were used as representative values for austenitic alloys. The calculated value of the dislocation density is reasonable for the deformed materials, as far as the order of magnitude is concerned.

The calculated value of \( b/(mdM) \) is larger for the \( \gamma \rightarrow \varepsilon \) transformation than for the \( \gamma \rightarrow \alpha' \) transformation, as was described above. The value of \( b/(md) \) is the shear strain itself by the advancement of transformation dislocations. It is thus apparent that the dependence of the extra energy \( \Delta G_{\text{ex}} \) on the work hardening of austenite (\( \sigma - \sigma_{0.2} \)) should be larger for the \( \gamma \rightarrow \varepsilon \) transformation with the large shear strain than for the \( \gamma \rightarrow \alpha' \) transformation. However, in order to discuss such a difference in detail from the present results, we need more precise thermodynamics data for \( \gamma \rightarrow \varepsilon \) transformation in Fe-Mn alloys.

In contrast to the present results, the increase in \( M \),

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Reduction of ( \gamma ) (MN/m(^2))</th>
<th>( \Delta M_{\alpha'} )</th>
<th>( \Delta G_{\text{ex}} ) (MJ/m(^3))</th>
<th>( \rho_\beta ) (cm/cm(^3))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe-16Mn</td>
<td>6%</td>
<td>148</td>
<td>-14 K - 6.88</td>
<td>-11.5 ( 1.1 \times 10^9 )</td>
</tr>
<tr>
<td>Fe-16Mn</td>
<td>18%</td>
<td>247</td>
<td>-26 K - 12.5</td>
<td>-20.9 ( 3.6 \times 10^9 )</td>
</tr>
<tr>
<td>Fe-16Mn</td>
<td>39%</td>
<td>295</td>
<td>-40 K - 18.9</td>
<td>-31.6 ( 3.6 \times 10^9 )</td>
</tr>
<tr>
<td>Fe-24Mn</td>
<td>7%</td>
<td>143</td>
<td>-32 K - 11.6</td>
<td>-19.4 ( 3.1 \times 10^9 )</td>
</tr>
<tr>
<td>Fe-24Mn</td>
<td>18%</td>
<td>238</td>
<td>-61 K - 20.9</td>
<td>-35.0 ( 1.0 \times 10^9 )</td>
</tr>
</tbody>
</table>
temperature by prior deformation of austenite has been
frequently reported for the case of \(\alpha'\) lenticular martensite in Fe–high Ni and Fe–high Ni–C alloys\(^{(13-17)}\). Hosoi and Kawakami\(^{(1)}\) reported that the \(M_s\) temperature increased from 222 K to 238 K by 30% reduction of austenite in an Fe–31.7%Ni alloy. A similar significant increase in \(M_s\) has been reported in an Fe–26.4%Ni–0.42%C alloy by Gooch and West\(^{(19)}\) and in an Fe–31.2%Ni–0.10%C alloy by Tokizane\(^{(20)}\). It has been believed that dislocations in austenite act as favorable nucleation sites and thus \(M_s\) rises by the prior deformation of austenite. In the present study, however, the \(M_s\) temperature monotonically decreases with increasing prior reduction of austenite (Figs. 2 and 3). It is apparent that the most potent nucleation sites operate at \(M_s\) temperature. The present result thus suggests that the prior deformation of austenite does not introduce more potent nucleation sites compared with pre-existing nucleation sites in the present alloys. However, the reason for that is not known at present. It is worth to note again that the number of \(\varepsilon\) plates increases by the prior deformation of austenite (Fig. 6). This suggests that at least some of dislocations introduced into austenite by the prior deformation act as the nucleation sites, although those potency is not so high. It can be concluded that most of dislocations introduced into austenite by the prior deformation act as obstacles against the advancing transformation dislocations and some of those act as nucleation sites.

V. Conclusions

The effect of prior deformation of austenite on the \(\gamma \rightarrow \varepsilon\) martensitic transformation in Fe–Mn binary alloys during subsequent cooling has been investigated. The formation of \(\alpha'\) lath martensite in deformed austenite has been also examined for comparison. The main findings in the present study would be summarized as follows.

(1) In both of the transformations to \(\varepsilon\) (hcp) in the Fe–16Mn and Fe–24Mn alloys and to \(\alpha'\) (bcc) in the Fe–9Mn and Fe–23Ni alloys, \(M_s\) temperature decreases monotonically with increasing reduction of austenite at 773 K.

(2) The decrease in \(M_s\) temperature for the \(\gamma \rightarrow \varepsilon\) transformation is more remarkable than that of the \(\gamma \rightarrow \alpha'\) transformation. The \(M_s\) temperature of the Fe–24Mn alloy is lowered over 80 K by 40% reduction of austenite and the structure becomes fully austenitic at room temperature. On the other hand, the decrease in \(M_s\) temperature for the \(\alpha'\) lath martensite is 16 K.

(3) In the Fe–16Mn alloy, the \(\varepsilon\) plates become thin and curved by the prior deformation of austenite. The amount of \(\varepsilon\) martensite at room temperature is almost constant irrespective of the reduction of austenite. These results indicate that the number of \(\varepsilon\) plates is increased by the prior deformation of austenite and suggest that some of dislocations introduced into austenite act as nucleation sites.

(4) The decrease in \(M_s\) temperature by the prior deformation can be quantitatively accounted for by the strengthening of austenite and the extra energy required for the transformation dislocations to advance through forest dislocations in austenite. The marked stabilization of austenite against the \(\gamma \rightarrow \varepsilon\) martensitic transformation is mainly associated with the smaller increase in chemical driving force for the \(\gamma \rightarrow \varepsilon\) transformation with decreasing temperature compared with that for the \(\gamma \rightarrow \alpha'\) transformation.

Acknowledgments

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