Effect of Transformation Cycling on the ε Martensitic Transformation in Fe–Mn Alloys

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The effect of $\gamma \rightleftarrows \varepsilon$ cyclic transformation on the subsequent forward and reverse transformations was studied in low carbon Fe–17%Mn and Fe–25%Mn binary alloys. A remarkable decrease in Ms temperature by the cyclic transformation was observed in the Fe–25%Mn alloy. After 7 thermal cycles between 305 and 573 K, Ms decreased by over 60 K and the structure became fully austenitic at room temperature. On the other hand, As and Af temperatures increased with an increase in the number of thermal cycles. As the upper peak temperature of thermal cycling increased, the changes in Ms, As and Af became smaller. The decrease in Ms by the cyclic transformation was smaller in the Fe–17%Mn alloy than in the Fe–25%Mn alloy. A large number of dislocations were observed in the austenite of the Fe–25%Mn alloy after the thermal cycling between 273 and 573 K, although dislocations were scarcely observed before the thermal cycling. The hardness of austenite at 573 K was markedly increased by the thermal cycling between 273 and 573 K. A microstructure memory, which is a phenomenon that martensite forms in the same position where the martensite formed in the last forward transformation, was not observed.

KEY WORDS: cyclic transformation; thermal cycling; stabilization of austenite; ε martensite; martensitic transformation; reverse transformaiton; Fe–Mn alloy.

1. Introduction

Recently, much attention has been paid to shape memory effect associated with fcc (austenite; γ) $\rightleftarrows$ hcp (ε martensite) transformation.[1-3] Sato et al.[4] have first presented that a single crystal of an Fe–30%Mn–1%Si alloy exhibits a complete shape memory effect when it is deformed in a specific direction, and clearly shown that the origin of the shape memory effect is the reverse transformation of stress-induced ε martensite of a single variant. Sato et al.[5,6] have also reported that the shape memory effect of Fe–Mn–Si single crystals depends on orientation and composition as well as physical properties such as magnetic susceptibility and lattice parameter. Murakami et al.[7] have succeeded to develop the polycrystalline Fe–Mn–Si shape memory alloys, and found that Fe–Mn–Si alloys with 28–33 % Mn and 4–6 % Si exhibit a nearly complete shape memory effect.

More recently, Otsuka et al.[8] have presented that the shape memory effect of an Fe–32%Mn–6%Si alloy is remarkably increased by the repetition of slight deformation (2.5 % in tension) at room temperature and annealing at 773–973 K. After 5 times of this repetition a complete shape memory effect was achieved.[8] This repetition treatment was called as " training ".[8] Moriya et al.[9] have reported that the training is also effective for stainless type Fe–Cr–Ni–Mn–Si shape memory alloys associated with $\gamma \rightleftarrows \varepsilon$ transformation. These alloys have been developed separately by Moriya et al.[9] and Otsuka et al.[10] The mechanism of the training effect is, however, not well clarified. Before achieving a complete shape memory effect, the slight deformation in the training accompanies not only stress-induced ε martensitic transformation but also slip deformation of austenite. It is thus apparent that the training process consists of two metallurgical phenomena, i.e., formation of stress-induced ε martensite and its reversion to austenite, and slip deformation of austenite and its recovery at elevated temperatures. Therefore, the influence of the $\gamma \rightleftarrows \varepsilon$ cyclic transformation per se on the subsequent forward and reverse transformations must be made clear, in order to understand the mechanism of the training in the Fe–high Mn shape memory alloys. However, the effect of $\gamma \rightleftarrows \varepsilon$ cyclic transformation on the subsequent transformations is still far from being fully understood even in Fe–Mn binary alloys which are typical ones showing $\gamma \rightleftarrows \varepsilon$ transformation. There have been several studies[11-15] on the cyclic transformation in Fe–Mn–C alloys containing about 0.05 % C or more. However, the study in carbon free Fe–Mn binary alloys has been limited.[15]

The purpose of this paper is to make clear the effect of the $\gamma \rightleftarrows \varepsilon$ cyclic transformation by simple cooling and heating on the subsequent transformations in carbon free Fe–Mn binary alloys.

2. Experimental Procedure

Binary Fe–17%Mn and Fe–25%Mn alloys used in the present study were prepared by induction melting in an argon gas atmosphere. The chemical compositions of the alloys are shown in Table 1. The alloys were homogenized at 1 423 K for 86.4 ks followed by hot rolling at 1 473 K to plates with a thickness of 10 mm. The plates were solution-treated at 1 273 K for 3.6 ks and then cut or machined to specimens for dilatometric measurement, microstructure observation and hardness measurement.

The changes in the starting temperature of $\gamma \rightarrow \varepsilon$ forward transformation (Ms) and the starting and finishing temperatures of $\varepsilon \rightarrow \gamma$ reverse transformation (As and Af) by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation were measured by means of a dilatometer. The amount of martensite was also determined by dilatometry according to the fact that $\gamma \rightarrow \varepsilon$ forward transformation accompanies 0.70 % contraction in specimen length.[16] Specimens with 3 mm in diameter and 10 mm in length were first austenitized at 1 273 K for 0.6 ks and cooled to 305 K (just above room temperature). Thermal cycling was then performed between 305 K (below Ms) and a temperature above Af. The latter temperature (hereafter " upper peak temperature ") was altered from 573 to 1 173 K. The heating and cooling rates were set at 2.5 K/s. The specimens were not held at either 305 K or the upper peak temperature of thermal cycling. All heat treatments were performed in an argon gas atmosphere. Average values of Ms, As and Af temperatures in the first thermal cycle after the austenitization at 1 273 K are shown in Table 1.

For specimens for microstructure observation and hardness measurement, the thermal cycling was conducted after austenitization at 1 273 K for 0.6 ks, using iced water (273 K) and a salt bath set at 573 K. Specimens for optical metallography were etched with 100 ml saturated $Na_2S_2O_3$ water solution+10 g $K_2S_2O_5$ after electrolytic polishing. Foils for transmission electron microscopy were thinned in a 100 g CrO_3+500 ml CH_3COOH+20 ml H_2O solution using a twin-jet polisher. The foils were examined with a JEM 200CX operated at 200 kV. The hardness of austenite was measured at 573 K (above Af temperature) by a Vickers hardness tester with a hot stage using a load of 49 N.

Throughout this paper the treatment, which is composed of the cooling to 305 or 273 K (below Ms) after the austenitization at 1 273 K and the subsequent heating up above Af, will be referred to as the first cycle of thermal cycling. Accordingly, specimens cooled to 273 or 305 K after the austenitization at 1 273 K will be described to be in the condition

after the first $\gamma \rightarrow \varepsilon$ forward transformation.

3. Results

3.1. Effects of Upper Peak Temperature and Mn Content on Stabilization of Austenite by Thermal Cycling

Fig. 1 shows the changes in Ms, As and Af temperatures and the amount of ε martensite at 305 K by the thermal cycling in the Fe–25%Mn alloy. The upper peak temperatures of thermal cycling were 573, 873 and 1 173 K for Figs. 1(a) to 1(c), respectively. In the case of the upper peak temperature of 573 K (Fig. 1(a)), Ms temperature and the amount of ε martensite at 305 K decrease monotonously with an increase in the number of thermal cycles. After 7 thermal cycles Ms is lowerd by over 60 K and the structure becomes fully austenitic at 305 K. On the other hand, As temperature increases with an increase in the number of thermal cycles. As in the 7th cycle is higher by 15 K than that in the 1st cycle. Af temperature is also somewhat increased by the thermal cycling. In the case of the upper peak temperature of 873 K (Fig. 1(b)), the changes in Ms, As, Af and the amount of martensite at 305 K by the thermal cycling are similar to those in the case of the upper peak temperature of 573 K (Fig. 1(a)), although more thermal cycles (10 cycles) are required for Ms to be below 305 K. When the thermal cycling is conducted with the upper peak temperature of 1 173 K (Fig. 1(c)), the marked stabilization of austenite is not observed. Ms after 4 cycles is constant, although the decrease of 15 K in Ms is observed during first 4 cycles. Moreover, the amount of ε martensite at 305 K is constant unlike in the thermal cycling with the upper peak temperatures of 573 and 873 K (Figs. 1(a) and 1(b)). This indicates that the degree of stabilization of austenite against ε martensitic transformation by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation depends on the upper peak temperature of thermal cycling. The effect of the upper peak temperature in the Fe–25%Mn alloy is demonstrated in Fig. 2. In this figure, Ms temperature in the 7th thermal cycle is plotted against the upper peak temperature. Since different pieces were used for each upper peak temperature, Ms temperature of the 1st forward transformation of each pieces is also shown. Ms of the 1st transformation scatters in the range of $366\pm$ 4 K. Ms in the 7th thermal cycle increases considerably with an increase in the upper peak temperature when it is higher than 773 K.
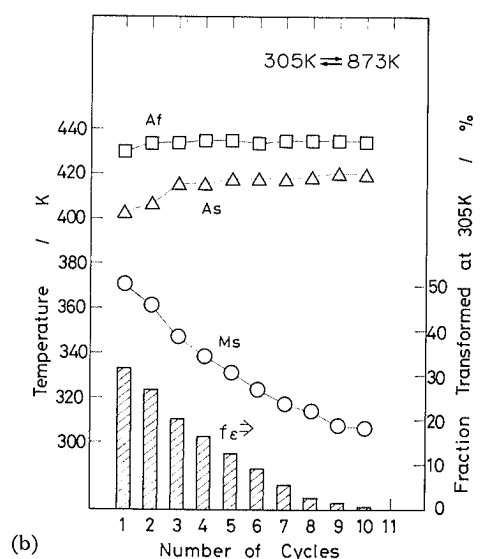
Fig. 3 shows the changes in Ms, As and Af temperatures and the amount of ε martensite at 305 K by the thermal cycling in the Fe–17%Mn alloy. The upper peak temperatures of thermal cycling were

Table 1. Chemical compositions (mass%) and Ms, As and Af temperatures (K) in the first thermal cycle of the alloys.
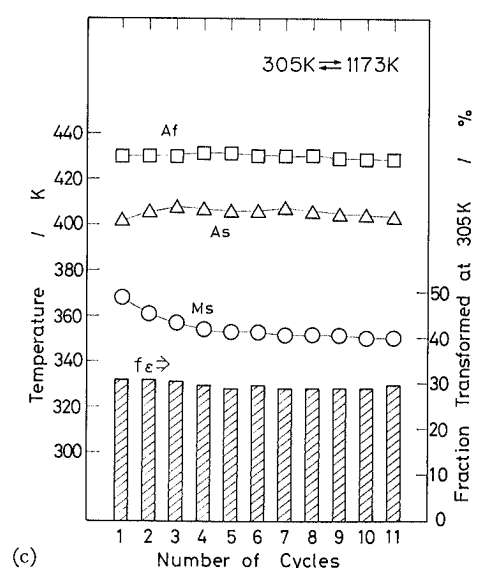
Alloy	C	Si	Mn	P	S	Fe	Ms	As	Af
Fe–17%Mn	0.0010	<0.005	16.6	0.0014	0.0004	bal.	433	453	486
Fe–25%Mn	0.002	0.02	24.4	<0.005	0.005	bal.	366	407	436
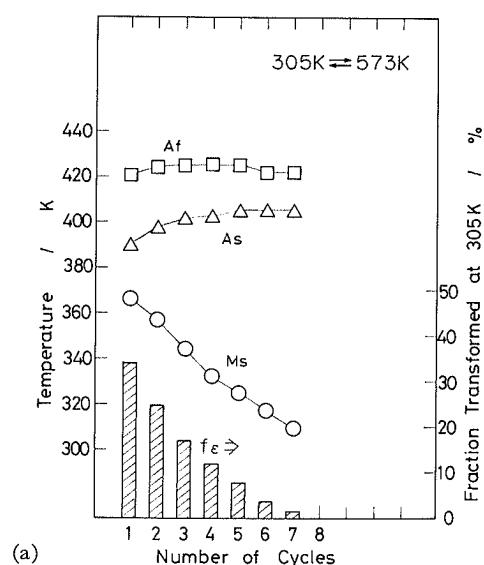
(a)

(b)

(c)

Fig. 1. Changes in Ms, As and Af temperatures and the amount of ε martensite at 305 K by thermal cycling in the Fe–25%Mn alloy.

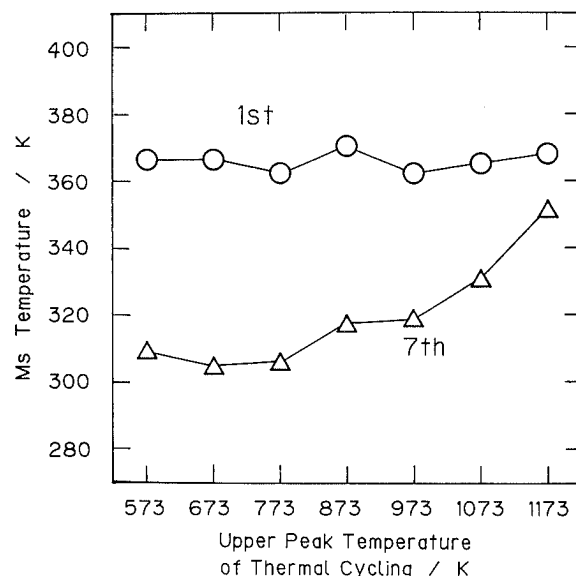The upper temperature of thermal cycling test was set at 573 K (a), 873 K (b) and 1 173 K (c).

Fig. 2. Effect of the upper peak temperature of thermal cycling on the stabilization of austenite in the Fe–25%Mn alloy.

Ms temperatures of the 1st and 7th forward transformation were plotted against the upper peak temperature.

573, 873 and 1 173 K for Figs. 3(a) to 3(c), respectively. A significant decrease of about 20 K in Ms is seen in the cases of the upper peak temperatures of 573 and 873 K (Figs. 3(a) and 3(b)), although the decrease in Ms is smaller than that observed in the Fe–25%Mn alloy (Figs. 1(a) and 1(b)). As and Af temperatures are somewhat increased by the thermal cycling. In the case of the upper peak temperature of 1 173 K (Fig. 3(c)), Ms, As and Af temperatures are almost constant. In all of 3 thermal cycling tests, the amount of ε martensite at 305 K hardly changes.

3.2. Microstructure Change by Thermal Cycling

Fig. 4 shows the change in optical microstructure at room temperature by the thermal cycling between 273 and 573 K in the Fe–25%Mn alloy. Micrographs were taken from the same area of a specimen except for Fig. 4(a). It is clearly seen that the number of ε martensite plates decreases with an increase in the number of thermal cycles, confirming that austenite becomes stable against ε martensitic transformation by the thermal cycling. It should be noted in Fig. 4 that a microstructure memory, which is a phenomenon that martensite forms in the same position where the martensite formed in the last forward transformation, is not observed. For instance, the ε plates indicated by the arrows in Fig. 4(f) did not appear in the last cycle (Fig. 4(e)).

Figs. 5(a) and 5(b) show transmission electron micrographs of the Fe–25%Mn alloy taken from a specimen after the 1st forward $\gamma \rightarrow \varepsilon$ transformation and from a specimen after 9 thermal cycles between 273 and 573 K, respectively. There are a large number of dislocations in the austenite after the thermal cycling, while dislocations are scarcely observed after
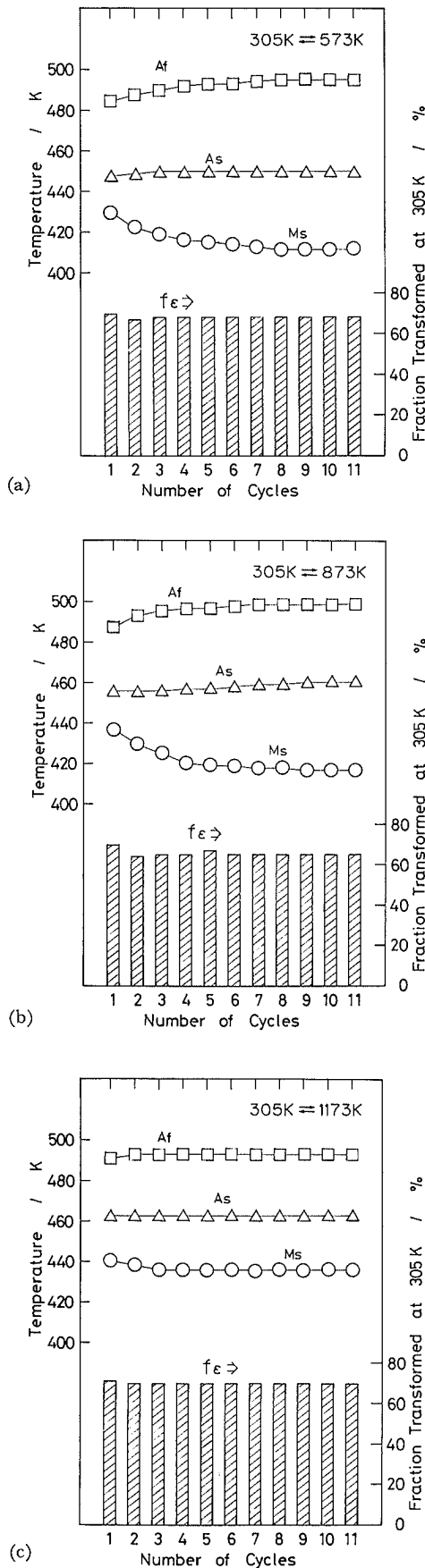
Fig. 3. Changes in Ms, As and Af temperatures and the amount of ε martensite at 305 K by thermal cycling in the Fe–17%Mn alloy.

The upper temperature of thermal cycling test was set at 573 K (a), 873 K (b) and 1 173 K (c).

the 1st forward transformation. This result indicates that the $\gamma \rightleftarrows \varepsilon$ cyclic transformation introduces many dislocations into austenite. It should be noted in Fig. 5(b) that some ε plates are seen, although the specimen was presumed to be fully austenitic after 9 thermal cycles as described in Fig. 1(a). These ε plates were considered to form isothermally during the foil preparation before the TEM observation, as it will be described in Sec. 3.3.

Fig. 6 shows the change in hardness of austenite at 573 K by the thermal cycling between 273 and 573 K. Zero for the number of cycles in the figure means that a specimen was kept at 573 K, over Ms, after the austenitization at 1 273 K. The hardness markedly increases with an increase in the number of thermal cycles. Since many dislocations were observed in the austenite after the thermal cycling (Fig. 5(b)), this increase in hardness of austenite is apparently attributed to dislocations introduced by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation. Moreover, the monotonous increase in hardness of austenite suggests that such an introduction of dislocations occurs in every $\gamma \rightleftarrows \varepsilon$ transformation.

3.3. Isothermal $\gamma \rightarrow \varepsilon$ Transformation after Thermal Cycling

In the Fe–25%Mn alloy, fully austenitic structure was obtained at room temperature by the thermal cycling when the upper peak temperature of thermal cycling was not higher than 873 K. However, this austenite was not stable and ε martensite formation was observed when the specimens were kept at around room temperature. Fig. 7 shows the $\gamma \rightarrow \varepsilon$ transformation behavior during isothermal holding at 313, 333 and 353 K after 9 thermal cycles between 305 and 573 K in the Fe–25%Mn alloy. At 313 K the amount of ε martensite remarkably increases with holding time and reaches 22 % at 40 ks. The transformation rate is fast in the early stage and gradually decreases with time. The increase of only 2 % in the amount of martensite was observed during the isothermal holding up to 72 ks after 40 ks. As the holding temperature increases, the transformation rate becomes slower. At 333 K the amount of martensite is less than 2 %. At last the isothermal martensite formation is not observed at 353 K, although the holding temperature is lower than Ms of the 1st forward transformation (368 K). A similar isothermal ε martensite formation after $\gamma \rightleftarrows \varepsilon$ cyclic transformation was reported by Schumann and Heider[15] in an Fe–16.4%Mn–1.01%Si–0.09%C alloy and by Olson and Cohen[17] in an Fe–16.5%Mn–6.7%Ni alloy. To keep away from the isothermal $\gamma \rightarrow \varepsilon$ transformation, the optical microstructure observation was carried out within 1.6 ks after the heat treatment. However, in the case of TEM observation the specimens were left at room temperature over 20 ks after the heat treatment for the foil preparation. Therefore, it is believed that the ε plates in Fig. 5(b) formed isothermally during the foil preparation before the TEM observation.
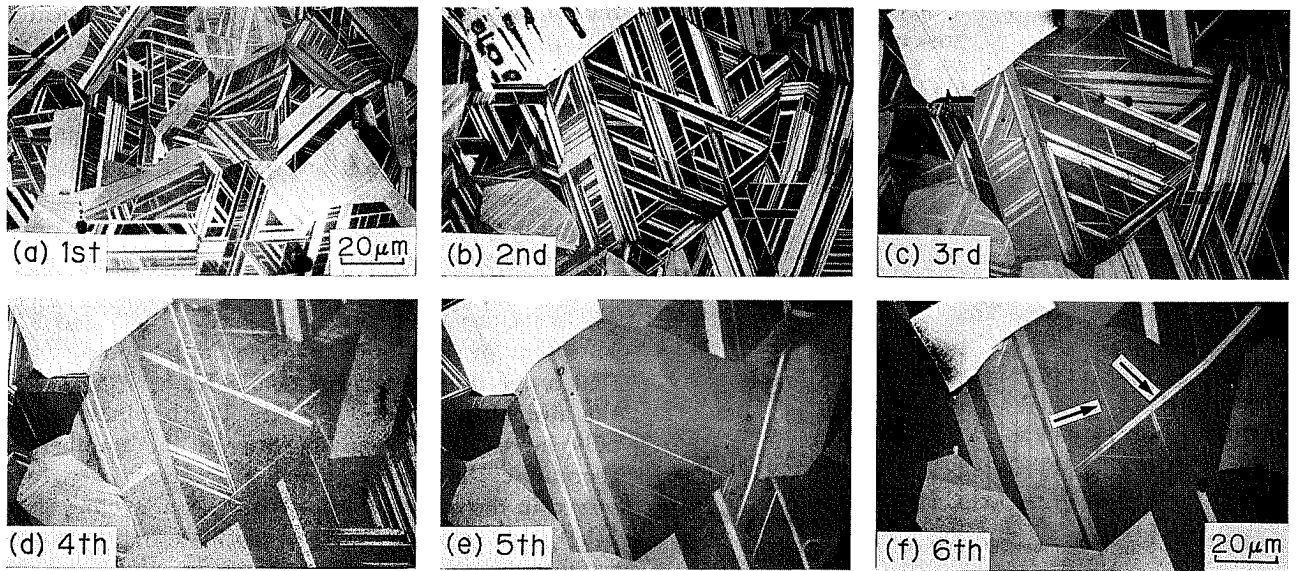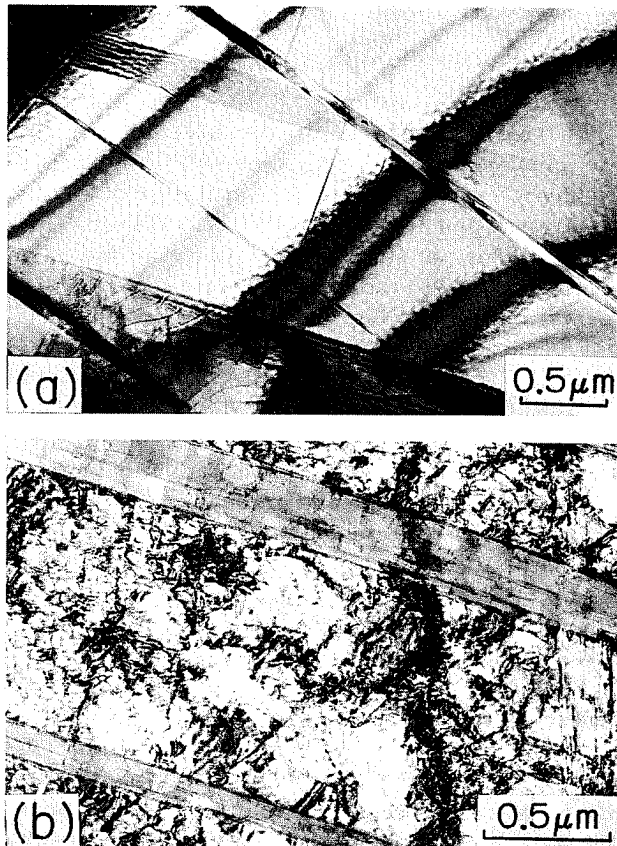
Fig. 4. Optical micrographs showing the change in ε martensite structure by the thermal cycling between 273 and 573 K in the Fe–25%Mn alloys.

(a) A specimen after the austenitization at 1 273 K and before the thermal cycling

(b) A specimen after 9 thermal cycles between 273 and 573 K.

Fig. 5. Transmission electron micrographs of the Fe–25%Mn alloy.

Fig. 6. Change in hardness of austenite at 573 K by the thermal cycling between 273 and 573 K.

4. Discussion

Yershova and Bogachev[11] first reported that the amount of ε martensite at room temperature is decreased by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation in an Fe–
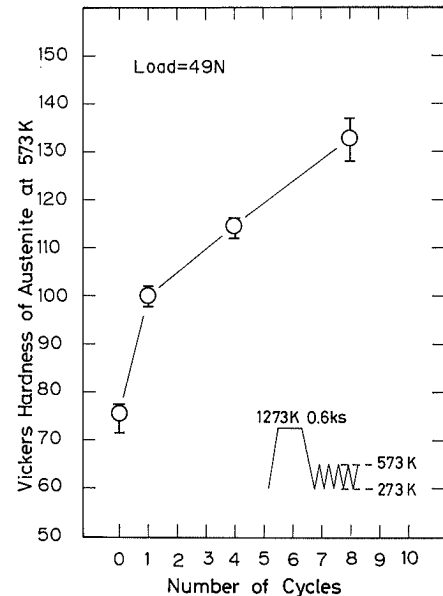
19.7%Mn–0.06%C alloy. Bogachev et al.[12,13] also reported the similar stabilization of austenite against ε martensitic transformation in Fe–Mn based alloys containing about 0.05 % C. In order to make clear the effect of carbon content on the stabilization of austenite by the cyclic transformation, Lysak and Nikolin[14] carried out the thermal cycling between 77 and 673 K and measured the amount of ε martensite at room temperature in Fe–(16–20)%Mn–C alloys with different carbon content of 0.001, 0.10, 0.20 and 0.35 %. They[14] reported that the amount of martensite is hardly changed by the cyclic transformation in the Fe–Mn–0.001%C alloy, while the amount of martensite is markedly decreased in the other alloys containing a larger amount of carbon. They[14] concluded that the stabilization of austenite by the cyclic transformation does not occur in the carbon free Fe–
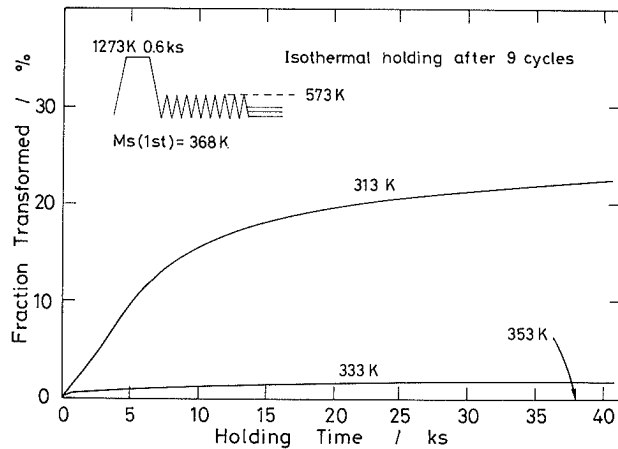
Fig. 7. Isothermal formation of ε martensite at 313, 333 and 353 K after 9 thermal cycles between 305 and 573 K in the Fe–25%Mn alloy.

Mn alloy. They[14] also suggested that the segregation of carbon atoms in stacking faults is responsible for the stabilization of austenite. It has been thus believed[18] that certain content of carbon is necessary for the stabilization of austenite. However, it must be pointed out that the conclusion of Lysak and Nikolin[14] is based only on the measurement of the amount of ε martensite at room temperature. In the present study, Ms temperature was significantly lowered by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation in the Fe–17%Mn alloy with carbon content of 0.001 %, although the amount of martensite at 305 K was constant as reported by Lysak and Nikolin.[14] The carbon content of 0.001 % of the present Fe–17%Mn alloy is the same as their alloy. Furthermore, both of Ms temperature and the amount of ε martensite at 305 K were markedly decreased by the cyclic transformation in the Fe–25%Mn alloy with 0.002 % C. Therefore, it is clear that austenite is stabilized against ε martensitic transformation by the cyclic transformation even in the carbon free Fe–Mn alloys and the existence of carbon atoms is not essential.

Tamura et al.[19] examined the mechanical stabilization of austenite against $\gamma \rightarrow \varepsilon$ martensitic transformation in an Fe–15%Cr–13%Ni alloy. They[19] reported that Ms temperature markedly decreases with an increase in the amount of prior deformation of austenite at 573 K. Recently, some of the present authors[20] have reported a similar large decrease in Ms by prior deformation of austenite at 773 K in the same Fe–25%Mn alloy used in the present study. Therefore, it is very possible that the decrease in Ms by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation is caused by the strengthening of austenite due to the introduction of a large number of dislocations. It is not known at present, however, how these dislocations were introduced into austenite during the $\gamma \rightleftarrows \varepsilon$ cyclic transformation process. Nevertheless, it should be noted that the introduction of dislocations may be closely associated with the $\varepsilon \rightarrow \gamma$ reverse transformation, because dislocations were scarcely observed in the specimen after the first forward transformation.

Based on the hypothesis that the mechanical sta-

bilization of austenite is responsible for the decreases in Ms temperature and the amount of ε martensite at 305 K by the cyclic transformation, the loss of stabilization effect at higher upper peak temperature as shown in Fig. 2 can be explained by faster annihilation of dislocations at higher temperature. Therefore, the degree of stabilization of austenite by the cyclic transformation should also depend on holding time at the upper peak temperature. In fact, when the thermal cycling test with holding time for 1.8 ks at the upper peak temperature of 873 K was carried out in the Fe–25%Mn alloy, the formation of ε martensite of 7 % was observed at 305 K even after 15 thermal cycles. Since the recrystallization of austenite was not observed even at the upper peak temperature of 1 173 K, it is clear that the annihilation of dislocations takes place by recovery.

The decrease in Ms temperature by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation means that the chemical driving force necessary for ε martensitic transformation is increased by the cyclic transformation, namely, excess driving force is required for ε martensitic transformation. Fig. 8 shows the difference in the free energy between austenite and ε martensite $(\Delta G^{\gamma \rightarrow \varepsilon})$ as a function of temperature. $\Delta G^{\gamma \rightarrow \varepsilon}$ was calculated, based on a regular solution model, using the method proposed by Ishida and Nishizawa[21] and the well known thermodynamics data reported by Blackburn et al.[22] and by Breedis and Kaufman.[23] The arrows in Fig. 8 indicate the Ms temperature of the 1st forward transformation observed in the Fe–17%Mn and Fe–25%Mn alloys. It is worth to note that the slope of curve at the Ms temperature is smaller in the Fe–25%Mn alloy than in the Fe–17%Mn alloy. This indicates that, in order to get a given excess driving force, the larger temperature decrease from the Ms of the 1st transformation is required in the Fe–25%Mn alloy. This may be associated with the larger decrease in Ms by the $\gamma \rightleftarrows \varepsilon$ cyclic transformation in the Fe–25%Mn alloy than in the Fe–17%Mn alloy.

It has been believed that ε martensitic transformation occurs by motion of $a/6\langle 112 \rangle$ Shockley partial dislocations.[4] Sato et al.[4] reported that the motion of these Shockley partial dislocations is controlled by thermally activated overcoming of short range obstacles. The occurrence of $\gamma \rightarrow \varepsilon$ isothermal transformation after the thermal cycling in the Fe–25%Mn alloy (Fig. 7) suggests that the motion of the Shockley partial dislocations is controlled by thermally activated process even after the thermal cycling. Dislocations introduced into austenite by the cyclic transformation may act as additional short range obstacles as well as long range obstacles. Anyhow, since the transformation temperature range is higher in the Fe–17%Mn alloy than in the Fe–25%Mn alloy, it is apparent that the thermally activated overcoming of short range obstacles occurs more quickly during the cooling in the thermal cycling in the Fe–17%Mn alloy. This may be another reason for the smaller decrease in Ms by the cyclic transformation in the Fe–17%Mn alloy compared with the Fe–25%Mn alloy.
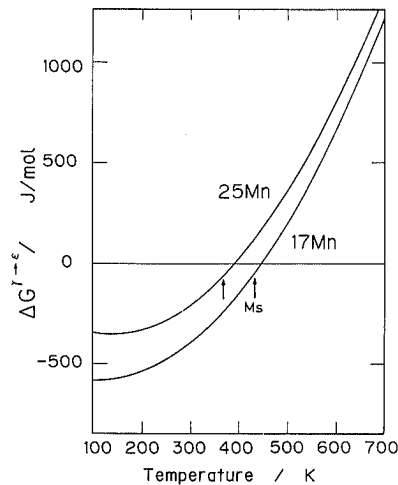
↑ : Ms temperatures in the first forward transfor-
mation of the Fe–17%Mn and Fe–25%Mn
alloys.

Fig. 8. Difference in free energy between austenite and ε
martensite ($\Delta G^{\gamma-\varepsilon}$) as a function of temperature.

The $\gamma \rightleftarrows \varepsilon$ cyclic transformation also influenced the
subsequent $\varepsilon \rightarrow \gamma$ reverse transformation. As and Af
temperatures were increased by the thermal cycling
(Figs. 1 and 3). It should be noted again that the
amount of ε martensite at 305 K decreased with an
increase in the number of thermal cycles. Therefore,
the increases in As and Af might be associated with
the decrease in the amount of ε martensite. In order
to examine the effect of the amount of ε martensite
on As and Af temperatures, a following experiment
was carried out in the Fe–25%Mn alloy. After the
austenitization at 1 273 K for 0.6 ks, specimens were
cooled to various temperatures between Ms and 305
K to obtain various amount of martensite. The
specimens were then heated up for the reverse trans-
formation. As and Af temperatures measured in this
experiment are shown in Fig. 9 against the amount of
ε martensite before the heating. As and Af tempera-
tures are almost independent of the amount of ε
martensite. It is thus clear that the increase in As
and Af shown in Figs. 1 and 3 is essential feature
caused by the cyclic transformation, although the rea-
son for the increase in As and Af is not known at
present.

Recently, Sade et al.[24] studied the effect of cyclic
transformation on the subsequent transformations in
Fe–(24–36)%Mn–(2–4)%Si alloys. They[24] reported
that Ms is decreased by only 10 K even after 1 000
thermal cycles between 77 and 598 K. Moreover,
Ghosh et al.[25] reported that the amount of ε marten-
site at room temperature is significantly increased by
the thermal cycling between room temperature and
573 K in Fe–(26–32)%Mn–(3–6)%Si alloys. These
results are different from the present results of Fe–Mn
binary alloys. This discrepancy between Fe–Mn–Si
alloys and Fe–Mn binary alloys suggests that silicon
has an important effect on $\gamma \rightarrow \varepsilon$ forward and $\varepsilon \rightarrow \gamma$
reverse transformations per se in addition to the pre-
viously reported effects,[6,7] i.e., the solution hardening
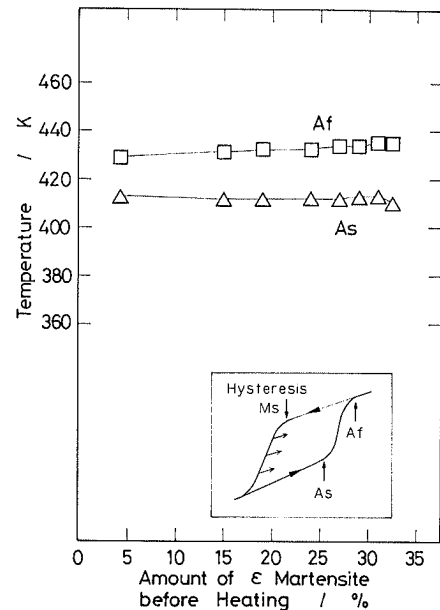of austenite and the decrease in Néel point.

Fig. 9. Effect of the amount of ε martensite before the heat-
ing on As and Af temperatures in the Fe–25%Mn
alloy.

5. Conclusions

The effect of $\gamma \rightleftarrows \varepsilon$ cyclic transformation by cool-
ing and heating on the subsequent forward and
reverse transformations was studied in low carbon
Fe–17%Mn and Fe–25%Mn binary alloys. Main
results obtained are as follows.

(1) A remarkable decrease in Ms temperature by
the cyclic transformation is observed in the Fe–
25%Mn alloy. After 7 thermal cycles between 305
and 573 K, Ms is decreased by over 60 K and the
structure becomes fully austenitic at room tempera-
ture. However, this austenite transforms isother-
mally to ε martensite when the specimen is kept at
around room temperature. As and Af temperatures
are increased by the cyclic transformation. The
changes in Ms, As and Af become smaller, as the
upper peak temperature of thermal cycling increases.

(2) The decrease in Ms by the cyclic transforma-
tion is smaller in the Fe–17%Mn alloy than in the
Fe–25%Mn alloy. In the Fe–17%Mn alloy, the
amount of ε martensite at 305 K is hardly changed
by the cyclic transformation.

(3) A large number of dislocations are observed
in the austenite of the Fe–25%Mn alloy after the
thermal cycling between 273 and 573 K, while dis-
locations are scarcely observed before the thermal
cycling. The hardness of austenite at 573 K is
markedly increased by the thermal cycling between
273 and 573 K. It is concluded that the decrease
in Ms by the cyclic transformation is attributed to the
strengthening of austenite due to the dislocations in-
troduced into austenite.

(4) A microstructure memory, which is a phe-
nomenon that martensite forms in the same position
where the martensite formed in the last forward trans-
formation, is not observed.

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