Microscopic Studies on Stress-induced Martensite Transformation and Its Reversion in an Fe-Mn-Si-Cr-Ni Shape Memory Alloy

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Martensitic transformation induced by extension at room temperature and its reverse transformation on heating in an Fe-14Mn-6Si-9Cr-5Ni (mass%) shape memory alloy were studied in detail by optical microscopy, electron microscopy and electron diffraction. The stress-induced transformation is initiated by random formation of extremely thin martensite plates with 1-2 nm width and then these plates are clustered and some of them coalesce to form thicker martensite plates with increasing deformation. The clustered regions are 400-600 nm wide and considered to correspond to the thinnest martensite plates observable with optical microscope. It is presumed that these clustered regions may also correspond to the lamella structures of a mixture of f.c.c. and h.c.p. phases reported in a previous work. In optical microscopic scale, the thin martensite plates with the smallest width are formed rather uniformly in an austenite grain, and with further increasing deformation, they are clustered and coalesce into thicker plates with 3-8 μm width. On heating, the reverse transformation is accomplished exactly by the reverse process of the martensite formation mentioned above. Stacking disorders in martensite revealed by electron diffraction analysis are such that the stacking sequence at the fault represented by parameter α is equal to that of the parent phase, in other words, extremely thin layers of f.c.c. phase (parent phase) with the minimum width of 3 layers are contained in the martensite with a high density.

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I. Introduction

Although there are many shape memory alloys so far reported, a very few of them are practically used. Among them, the NiTi alloy and its ternary alloys are most widely employed in the various fields of application. However, these alloys are very expensive, so, various attempts have been made to find cheaper shape memory alloys; especially, in the past decade, Fe-based shape memory alloys have drawn much attention because of their low costs. These are Fe-Mn-Si⁽¹⁾⁻⁽¹²⁾, Fe-Mn-Si-Cr-Ni⁽¹³⁾⁻⁽¹⁹⁾, Fe-Ni-Co-Ti⁽²⁰⁾⁻⁽²³⁾ and Fe-Ni-C alloys⁽²⁴⁾⁽²⁵⁾. The martensitic transformations involved in these shape memory alloys are the f.c.c. $\rightarrow$ h.c.p. transformation for the first two cases and the f.c.c. $\rightarrow$ b.c.t. transformation for the latter two cases.

Among these alloys, the most promising alloy will be the Fe-Mn-Si-Cr-Ni alloy which is a modification of Fe-Mn-Si alloy and has a property of corrosion-resistance comparable to the stainless steels. However, in order to obtain a good shape memory effect (more than 80% shape recovery) for this alloy, some kinds of thermomechanical treatments are necessary. For the as-solution treated specimen, only about 50% shape recovery is obtained. One of the thermomechanical treatments is so-called the "training" in which deformation by 2.5% at room temperature and annealing at 870 K for 10 min are repeated several times⁽¹⁰⁾⁽¹⁴⁾. The other is such that a specimen which has been solution-treated at 1320 K for 30

min is first cold-rolled by 10% at room temperature and then heated at 970 K for 10 min.

Recently, in order to know the origin of the improved shape recovery in the thermomechanically treated specimen of this alloy, Ogawa and Kajiwarara⁽²⁶⁾ studied in detail the deformation structures produced by 4% extension, using a high resolution electron microscope. In their experiment, specimens were subjected to the thermomechanical treatment of the latter type mentioned above. They found that the minimum unit of deformation band with 0.2 μm generated in the thermomechanically treated specimen consists of lamella structures having a mixture of h.c.p. and f.c.c. phase with 1-10 nm widths, but in the case of non-thermomechanically treated specimen, the deformation band mostly consists of h.c.p. phase only. They have attributed the observed good shape memory effect to the easiness of the reverse transformation (h.c.p. $\rightarrow$ f.c.c.) on heating, because there is no need for the nucleation of the f.c.c. phase in the h.c.p. martensite, that is, extremely thin plates of the f.c.c. phase in the lamella structure serve as starting places for the reverse transformation by thickening. It was presumed that stacking faults in austenite introduced by the thermomechanical treatment provide nucleation sites for the f.c.c. to h.c.p. transformation and, since the stacking faults are rather uniformly distributed, the formation of the fine lamella structure consisting of the f.c.c. and h.c.p. phases will be easily realized.

As for the other type of the thermomechanical treatment (i.e., the "training") mentioned above, Otsuka *et*

al.⁽¹⁰⁾⁽¹⁴⁾ concluded that the improvement of the shape recovery is attributed to (1) the increase in yield stress of austenite and (2) the decrease in the critical stress for the martensite formation. However, it is also considered that the "training" brings about the increase of uniformly distributed stacking faults in austenite, which will produce the above mentioned lamella structure. Therefore it is very important to clarify the formation process and its reversion of these lamella structures of the f.c.c. and h.c.p. phases in this alloy. In the present work, microscopic observations on martensite transformation and its reversion in an Fe-Mn-Si-Cr-Ni alloy have been made, using a conventional optical microscope, a high resolution differential interference optical microscope and an electron microscope. The main purpose of this work is to inspect the formation mechanism of the above mentioned lamella structures with nanometric scales and to know its reversion process on heating. Stacking disorders in the h.c.p. martensite were also examined by electron diffraction.

II. Experimental Method

The alloy used is an Fe-14Mn-6Si-9Cr-5Ni (mass%) alloy. The result of chemical analysis is shown in Table 1. The specimens were subjected to one of the thermo-mechanical treatments mentioned above, namely, a specimen which had been solution-treated at 1320 K for 30 min was first cold rolled by 10% at room temperature and then heated to 970 K for 10 min. By this treatment, the shape recovery for 4% extension is increased to 80% from 50% of the non-thermomechanically treated specimen. The specimen thickness was 0.4 mm in the cold-rolled state. The M_s temperature of this thermo-mechanically treated specimen is 250 K and the martensitic transformation is induced by deformation at room temperature. For optical microscope observation, the specimen was electrolytically polished before extension and the change in surface relief due to the stress-induced martensite transformation was followed continuously at room temperature. The reverse transformation, i.e., reversion, of the martensite was followed by both of stepwise heating and continuous heating. (The A_f temperature is 480 K.) For the continuous heating, a heating stage by the infrared image furnace was employed. With this attachment, the change in the martensite surface relief on heating was observed continuously with the heating rate of 50 K/min. For electron microscopic observation, specimens were thinned by the twin-jet polishing technique after extension of 0.2, 0.8, 1.6, 3.0 and 4.0% at room temperature, respectively. The accelerating voltage of the electron microscope used was 200 kV and the specimen was able to be tilted within the range of $\pm 25^\circ$.

Table 1 Chemical composition (mass%) of the alloy used.

Mn	Si	Cr	Ni	C	P	S	Fe
14.09	6.17	8.97	5.29	0.004	0.003	0.0004	bal.

Fig. 1 Series of optical micrographs showing the formation of ϵ martensite consecutively.

III. Experimental Results

1. Optical microscopic observation

Figure 1 shows a series of optical micrographs showing formation of h.c.p. (ϵ) martensite consecutively. The specimen was extended stepwise by a hand-made extension device. The observation was made by a CCD camera and the photographs were printed out by a video printer. It seems in this figure that most of the initially formed martensite plates in (a) have thickened in the later stages of extension in (b) and (c). However, a high resolution differential optical micrograph of ϵ martensite, such as shown in Fig. 2, can reveal that a martensite plate which seems to be a single plate by a conventional optical microscope consists of much finer plates with the width of about $0.2 \mu\text{m}$. Figure 3 shows a series of micrographs taken consecutively by stepwise extensions. These micrographs were taken with a high resolution differential

Fig. 2 High resolution differential optical micrograph, revealing very fine martensite plates.

optical microscope and indicate how a seemingly thick martensite plate has formed. Note the martensite plates indicated by arrows. A thin martensite plate is formed very close to the initially formed plate and coalesces with the latter. This process seems to be repeated and a thick martensite plate which looks like a single plate by a conventional microscope is realized. Of course, each thin martensite plate has thickened to some extent as evidently seen in the other martensite plates. On heating, the reverse process occurs as shown in Fig. 4, where a series of micrographs on the identical area to Fig. 3 were taken stepwise by heating to the temperatures 290, 410, 440 and 470 K, and shown in (a), (b), (c) and (d), respectively. A seemingly one plate becomes split and these split plates have disappeared in the reverse manner as they had formed.

In order to examine the reverse transformation more dynamically but in a more macroscopic scale, a continu-

ous observation was made by using heating stage in which a long working distance lense had to be used. Figure 5 shows a series of optical micrographs taken in such an observation with a heating rate of 50 K/min. It is seen that the widths of most of martensite plates have decreased on heating, although a part of plate A was split. The widths of typical martensite plates in this figure are plotted against time in Fig. 6. It is interesting to note in this figure that all of the martensite plates have started to reduce their widths at the same time and the behavior of the γ/α interface movement on heating seem to be about the same. The meaning of an apparent reduction in martensite width in a macroscopic scale on heating will be discussed later.

2. Electron microscope observations

In order to examine further details of stress-induced martensite formation, electron microscopic observations were made on the specimens extended at room temperature by 0.2, 0.8, 1.6, 3.0 and 4.0%, respectively. Figure 7(a) shows an electron micrograph of martensite formed by extension of 0.2%. The martensite plates are oriented in such a way that the habit plane is parallel to the incident beam. In this paper, only electron micrographs with such orientation are presented; otherwise, we can not measure the martensite plate widths and their distribution. It is well known that, in the f.c.c. to h.c.p. transformation, the habit plane is $(111)_{\text{fcc}}$ and parallel to (00.1) plane of the h.c.p. martensite. From this micrograph, it is measured that the thickness of the martensite plates is 1-2 nm for thin plates and about 30 nm even for thick plates. The other characteristic feature is that the martensite plates are rather uniformly formed at this stage of deformation. It was confirmed by taking a darkfield image with an h.c.p. spot that the dark plates

Fig. 3 Series of differential optical micrographs taken consecutively by stepwise extension.

Fig. 4 Series of optical micrographs taken by stepwise heating to 290 K in (a), 410 K in (b), 440 K in (c) and 470 K in (d). The same area to Fig. 3 is shown.

Fig. 5 Series of optical micrographs taken on heating from room temperature, showing the reverse transformation. (a) 333 K, (b) 406 K, (c) 425 K, (d) 473 K.

in Fig. 7(a) are the h.c.p. martensite phase. A part of the dark field image is shown in Fig. 7(b).

Figure 8 shows martensite plates which were formed by extension of 0.8%. Besides very thin martensite plates with about 2 nm, thick martensite plates with the thickness of 20–55 nm have appeared. It seems that these thick

martensite plates have been produced by coalescence of the thinner plates as seen from split patterns at the places indicated by arrows. It is noted that observed martensite plates are clustered within the region of about 400 nm (along the diagonal direction of the micrograph). Many extended dislocations on the other {111} slip planes

Fig. 6 Changes in widths of martensite plates measured at points A, B, C and D in Fig. 5, on heating with the rate of 50 K/min.

which are inclined to the incident beam are also seen in this figure. Most of martensite plates have grown straight across these extended dislocations. This fact may indicate that these dislocations are not so serious obstacles to the passage of the ε martensite plates.

Martensite plates formed by a further extension (1.6%) are shown in Fig. 9. Here thick martensite plates with the width of 60–130 nm in dark contrast have appeared, but it is evident from the left side area of the micrograph that these thick martensite plates consist of much thinner plates with the width of 3–10 nm. It should be noted again in this figure that most of martensite plates are clustered within a certain width (about 600 nm in this case).

An example of a seemingly one thick martensite plate formed by an extension of 3% is shown in Fig. 10(a). Figure 10(b) is a further enlarged micrograph of the framed area in (a), but the diffraction condition of the micrograph has been slightly changed in (b) to reveal distinct striation contrast in the martensite. These striations are considered to be stacking fault fringes. Since the stacking fault plane, $(00.1)_{\text{hcp}}$, is nearly parallel to the incident beam in the present case, only one or two dark striations will correspond to one fault plane. There is also a possibility that these striations are interference fringes of the γ/ε interface of a thin γ phase, however, we cannot distinguish which is the former and which one is the latter because it was found by a previous HREM study⁽²⁶⁾ that the martensite plate in the thermomechanically treated case of this alloy contains very thin layers of γ phase (f.c.c., the parent phase) with the width of 1–10 nm in high density. Short dark contrasts with a rather irregular shape (indicated by arrows) are presumably partial dislocations, for the fringes are distorted at such places.

Figure 11 shows a high magnification micrograph of a part of region in a seemingly thick martensite plate (about 1 μm), which is similar to Fig. 10(a) but contains a less density of the fringes. The specimen was extended by 4%. (The right upper corner with dark contrast is a region of the parent phase which could be easily iden-

Fig. 7 (a) Electron micrograph of very fine martensite plates formed by the extension of 0.2%. (b) Part of the dark field image of (a) taken with an h.c.p. spot.

tified in a smaller magnification.) At the place indicated by double arrow, a pair of fringes are almost merged into one and terminated at that place. It is very likely that each of the fringes corresponds to the interference fringe due to the γ/ε interface, for it becomes split as it goes to the thicker part of the foil (the upper left direction). That is, these fringes are due to the γ/ε interfaces separated about 6 nm as measured in the upper part of the figure, which means that a wedge-shaped plate of γ phase exists at this place. (The width of the thicker part at the upper left corner is 6 nm.) A detailed inspection of this micrograph reveals that the number of fringes is, in general, larger in the upper left corner than in the lower right corner, namely, it increases with increasing thickness of the foil. This fact shows clearly that these fringes are

Fig. 8 Electron micrograph of martensite plates formed by the extension of 0.8%.

Fig. 9 Electron micrograph of martensite plates formed by the extension of 1.6%.

either stacking fault fringes or the interference fringes of the γ/ε interface in which the fault plane or the interface is slightly inclined to the incident beam. There are several places where the fringes terminate, as indicated by single arrows. At these places with strong dark contrast, partial dislocations exist.

In order to know a more general mode of deformation-induced transformation in this alloy, low magnification electron micrographs with about 10,000 direct magnification were taken and distributions of martensite plates were examined along a distance normal to the habit plane. The area examined was at least 10 μm wide. Since the magnification of the micrograph was low, the martensite plates were often seen as merely dark plates even if they contained lamella structures of the f.c.c. phase such as shown in Figs. 8 and 9. The obtained distribution is schematically shown in Fig. 12 for various amounts of deformation, where the widths of rectangles represent the widths of martensite plates and the vertical lines denote extremely thin martensite plates. This figure indicates that thin plates of martensite are generated

(a)

(b)

Fig. 10 (a) Electron micrograph of a seemingly one thick martensite plate formed by the extension of 3%. (b) Enlarged micrograph of the framed area in (a).

Fig. 11 High magnification electron micrograph of a part of region in a seemingly thick martensite plate (about 1 μm) formed by the extension of 4%.

rather uniformly in the early stage of deformation and later they are merged into thicker plates as the amount of

Fig. 12 Schematic pictures showing examples of the distribution of martensite plates for various amounts of extension.

deformation increases. Moreover, the martensite plates are not uniformly distributed in the later stage of deformation.

3. Electron diffraction analysis of stacking disorders in the martensite

As described in the preceding section, many stacking faults exist in the h.c.p. martensite. It was shown nearly 20 years ago by one of the present authors⁽²⁷⁾ that the type of stacking faults and their densities, i.e., stacking disorders, in the h.c.p. structure can be analyzed by electron diffraction. The same method was employed in the present study to examine stacking disorders in the martensite as a function of deformation. In general, two fault parameters, α and α' , are necessary to describe stacking disorders in the h.c.p. crystal⁽²⁸⁾. The meanings of these parameters are schematically shown in Fig. 13. Namely, if the faulting occurs in such a way that three consecutive layers form the sequence ABC (the normal order), the faulting is expressed by α . On the other hand, if the three consecutive layers resulted by the faulting is BAC (the reverse order), the faulting is expressed by α' . $\alpha=1$ means a perfect f.c.c. crystal with the normal stacking order, while $\alpha'=1$ means a perfect f.c.c. crystal with the reverse stacking order (i.e., a twin crystal of the former one). Of course, $\alpha=\alpha'=0$ means a perfect h.c.p. crystal. However, the number of fault parameters can be reduced by considering the observed characteristic features of

Fig. 13 Schematic figure showing the meanings of fault parameters of α and α' .

diffraction spots as in the previous work⁽²⁷⁾.

Figure 14(a) shows a typical example of a diffraction pattern taken from an area of lamella structures such as shown in Figs. 7–9. This pattern consists of f.c.c. and h.c.p. structures. A key diagram of this diffraction pattern is shown in Fig. 14(b), where open circles are h.c.p. spots and crosses represent f.c.c. spots, and triangles denote double reflections. Miller indices at the right side in the figure are those of h.c.p. lattice. Here n is an integer. First of all, we notice a strong streak along the direction parallel to c^* -axis. This fact alone indicates the existence of high density of stacking faults. Beside this, we can reveal by detail measurement that the distance between 10.2 and 10.1 spots (or between 10.0 and 10.1 spots) is smaller than that between 10.1 and 10.0 in this figure. This fact can be recognized in an enlarged part of the diffraction pattern shown in Fig. 15. That is, the h.c.p. spots have shifted towards the nearest f.c.c. spots of the parent phase. The directions of shifts of h.c.p. spots are indicated by arrows in Fig. 14(b). (Spots for $h-k=3n$ are not affected by the existence of stacking faults on the basal plane.) This fact indicates that the faulting occurs in such a way that the resulting stacking sequence approaches that of the f.c.c. structure with the normal order, i.e., that of the parent phase. Therefore, to a first approximation, we have analysed the stacking disorders in the present case by using only one parameter α , namely, $\alpha'=0$ was assumed.

The above mentioned distance between 10.2 and 10.1 spots and its equivalent distances in the observed diffraction patterns were measured for the specimens extended by various amounts of 0.2–4%. The measured distances were in the range of 168° – 179° when one period along c^* -axis of the reciprocal lattice of the h.c.p. structure (i.e., $l=2$) is taken as 360° . The corresponding values of fault parameter α were obtained using the curve in Fig. 6 of the previous paper by Kajiwara⁽²⁷⁾. Figure 16 shows the values of α against amount of deformation, where each circle represents the α value obtained from one diffraction pattern such as shown in Fig. 14(a). Although there is a considerable scatter in the value of α , we can see in this figure a tendency that α increases with increasing amount of deformation.

In order to understand the stacking disorders from the

Table 2 Values of α , $\bar{m}$, $1/(\bar{m}-2)$ for various amounts of extension.

Extension (%)	α (averaged)	$\bar{m}$	$1/(\bar{m}-2)$
0.2	0.043	2.045	22.2
0.8	0.041	2.043	23.3
1.6	0.094	2.10	10.0
3.0	0.126	2.14	7.1
4.0	0.101	2.11	9.1

(a)

(b)

Fig. 14 (a) Electron diffraction pattern taken from such an area as shown in Figs. 7-9. The pattern consists of $[0\bar{1}1]$ zone of f.c.c. structure and $[01.0]$ zone of h.c.p. structure. (b) Key diagram of diffraction pattern in (a).

other view point, we calculate the average layer spacing, $\bar{m}$, between "the faults" which are defined as layer shifts from the normal stacking sequence, ABC, of the f.c.c. structure. These layer shifts are either from C to A, from A to B, or from B to C such as schematically shown in Fig. 17. With this definition of the "faults", it is evident that $\bar{m}=2$ is for an unfaulted h.c.p. crystal. The followings are known from Fig. 17, that is, the probability of the occurrence of the "fault" at the 1st layer is zero because we have assumed $\alpha'=0$, the probabilities of the occurrence of the "faults" are $1-\alpha$ at the 2nd layer, $\alpha(1-\alpha)$ at the 3rd layer, $\alpha^2(1-\alpha)$ at 4th layer, and $\alpha^{m-2}(1-\alpha)$ at the m th layer. Therefore,

$$\bar{m} = \sum_{m=2}^{\infty} m \alpha^{m-2} (1-\alpha) = (2-\alpha)/(1-\alpha).$$

In Table 2 the values of $\bar{m}$ are shown for various amounts of deformation together with the values of $1/(\bar{m}-2)$. The values of α listed in this table are the values averaged for each amount of deformation. The value of $1/(\bar{m}-2)$ means the average layer spacing between the "random" stacking faults in the conventional sense. Since the one

Fig. 15 Enlarged part of diffraction pattern, showing the shifts of diffraction spots of h.c.p. lattice toward those of the f.c.c. lattice.

Fig. 16 Measured values of fault parameter α as a function of the amount of extension.

Fig. 17 Schematic picture showing the layer shifts from the normal stacking sequence of ABC expressed by fault parameter α .

layer spacing between the close packed planes of (00.1) of this structure is about 0.2 nm, the actual spacings between the "random" faults are in the range of 2-5 nm. In some of high magnification electron micrographs, we could observe the equivalent spacings between the stacking faults as will be reported in another paper⁽³⁰⁾.

IV. Discussion

From the experimental results described in the preceding section, we can construct the formation sequence of stress-induced martensite as follows. In the earliest stage of deformation, very thin martensite plates with 1–2 nm width appear rather uniformly and then these plates coalesce to form a thicker martensite plate. In the next step, martensite plates are clustered in a certain region of 400–600 nm width, where several thick martensites are situated in the center and thinner martensite plates are distributed around them. Here the width of the thicker martensite plates which are formed by the coalescence of thinner plates is in the range of 20–130 nm and their thickness increases with increasing deformation. Since the smallest width of martensite plates which could be observed by optical microscope is 0.2–0.6 μm , the above mentioned clustered region may correspond to the minimum unit of martensite observable with optical microscope. This clustered region may also correspond to the minimum unit of deformation band which consists of lamella structures of f.c.c. and h.c.p. phases reported in the previous work⁽²⁶⁾. In the further stage of deformation, the martensite thickness becomes up to about 1 μm in the view of electron microscopic scale. On the other hand, in the view of optical microscopic scale, martensite plates 0.2–0.6 μm in thickness are generated rather uniformly at first, then clustered and coalesce to form further thick martensite plates with 3–8 μm . It should be noted here that the “thickness increase” does not mean the lateral growth of the martensite plate, instead, very thin plates nucleate close to the existing martensite plate and coalesce with it in both the optical microscopic and electron microscopic scales. It is worth emphasizing that the coalescence is never complete and the f.c.c. phase is left between them as can be known by the previous work⁽²⁶⁾.

As mentioned above, the stress-induced transformation in the present alloy is basically accomplished by formation of groups of very thin martensite plates. This transformation mode will make it possible that the shape strain of martensite is accommodated elastically in the matrix (the parent phase). This nature greatly improves the shape memory effect, because (1) the plastic deformation in the parent phase can be avoided in principle at the time of martensite formation and (2), on heating, the strain field existing at the tip of a martensite plate will make easy the reverse motion of the Shockley partial dislocations which were associated with the formation of the h.c.p. phase.

On heating, an apparent reduction in the martensite thickness is observed in some cases by optical microscope as shown in Figs. 5 and 6. This means that thin martensite plates which constitute a thick martensite and are situated at the outer side of the plate disappear first. In most of the other cases, a seemingly one thick martensite plate becomes split into thin plates on heating and then these plates disappear in the reverse order as they form. These phenomena are the reverse process of the

martensite formation sequence.

The reverse transformation will be easier in the present case because there exist thin f.c.c. layers between martensite plates; there is no need for the nucleation of the f.c.c. phase in martensite, namely, the reverse transformation can be initiated simply by thickening of the existing f.c.c. thin plates. Since the easier the reverse transformation is, the better the shape recovery in the shape memory test is, the existence of thin f.c.c. layers in martensite will be essential to improve the shape memory effect. From the present experimental facts described in Section III.2 and the results of the previous HREM study⁽²⁶⁾, it is evident that the deformation structures produced by extension of the thermomechanically treated present alloy correspond to such case as mentioned above. This is one of the important reasons why the thermomechanically treated specimen of Fe-Mn-Si-Cr-Ni alloy shows a good shape memory effect.

Although the Seeger's pole mechanism⁽²⁹⁾ for the f.c.c. to h.c.p. transformation was recently confirmed to occur by the *in situ* electron microscopic observation⁽⁶⁾, it does not seem that this mechanism operates also in the present case because of the following reasons. (1) There are so many stacking faults represented by parameter α . (2) The minimum thickness of h.c.p. martensite observed in the early stage of deformation is 1–2 nm. If the pole mechanism is working, there will be no such large number of stacking faults and the minimum thickness of h.c.p. martensite will be equal to the length of the pole dislocation which is much larger than the observed value of the minimum thickness (1–2 nm).

Stacking disorders in martensite revealed by electron diffraction analysis are such that the stacking sequence at the fault represented by parameter α is equal to that of the parent phase, in other words, extremely thin layers of f.c.c. phase (parent phase) with the minimum width of 3 layers are contained in the martensite with a high density. Therefore, it is considered that these places at the faults provide favorable nucleation sites in the h.c.p. to f.c.c. reverse transformation. However, such situation is not so helpful to promote the reverse transformation, because, as reported in the previous work⁽²⁶⁾, the same type of stacking faults is also observed with an approximately same density in the non-thermomechanically treated specimens which show only 50% shape recovery. (The 50% shape recovery in the latter specimens may be due to the existence of such stacking faults, though.)

Finally we discuss why the stacking disorder in the h.c.p. martensite has a characteristic feature of $\bar{m} > 2$. $\bar{m} > 2$ means that the stacking sequence is likely to approach that of the f.c.c. structure of the parent phase. (It should be noted that the stacking sequence in the h.c.p. martensite is not faulted towards that of the twin structure of the f.c.c. parent phase.) The martensitic transformation from f.c.c. to h.c.p. structure is accomplished by introduction of stacking faults on every second layer of (111)_{fcc}. This lattice deformation is macroscopically equivalent to a simple shear on the habit plane, (111)_{fcc}. The characteristic faulting mentioned above may be as-

sociated with this simple shear, that is, the macroscopic shear of the lattice deformation in the f.c.c. to h.c.p. transformation is opposed by a force due to the constraints imposed by the surrounding matrix crystal, and then the transforming crystal is inclined to remain in the stacking sequence of the matrix crystal. As a result, $\bar{m}$ becomes greater than 2. It is worth noting in Table 2 that $\bar{m}$ increases with increasing amount of deformation. This fact may indicate that the force due to the constraints by the surrounding matrix increases with increasing deformation.

V. Summary

Martensitic transformation induced by extension at room temperature and its reverse transformation on heating in an Fe-14Mn-6Si-9Cr-5Ni alloy were studied in detail by using a conventional optical microscope, a high resolution differential interference optical microscope and an electron microscope. The results are summarized as follows.

(1) The stress-induced transformation is initiated by random formation of extremely thin martensite plates with 1–2 nm width and then these plates are clustered and some of them coalesce to form thicker martensite plates.

(2) In the optical microscopic scale, thin martensite plates with 0.2–0.6 μm thickness which are the minimum observable unit are formed rather uniformly at first, then clustered and coalesce into thick plates with 3–8 μm . The thinnest martensite plate observable with optical microscope corresponds to the lamella structure consisting of a mixture of the f.c.c. and h.c.p. phases reported in the previous work⁽²⁶⁾.

(3) On heating, an apparent reduction in the martensite thickness is observed in some cases by optical microscope, but this is a consequence of removing a thin martensite plate one by one from the outer side of the thick martensite plate. In most of the other cases, a seemingly one thick martensite plate becomes split into thin plates and then these plates disappear in the reverse order as they form. These phenomena are exactly the reverse process to the formation sequence of martensite.

(4) It seems that the existence of thin f.c.c. layers in martensite is essential to improve the shape memory effect as reported in the previous work⁽²⁶⁾.

(5) Another characteristic feature to improve the shape memory effect is the generation of extremely thin martensite plates of which the shape strains can be accommodated elastically in the matrix (the parent phase). In such a situation, the reverse motion of the Shockley partial dislocations associated with the forward transformation will be easy on heating.

(6) Stacking disorders in martensite revealed by electron diffraction analysis are such that the stacking sequence at the fault represented by parameter α is equal to that of the parent phase and the average spacing between such stacking faults is in the range of 2–5 nm.

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