Effect of grain boundary on martensite transformation behaviour in Fe–32 at.%Ni bicrystals

Masato Ueda, Hiroyuki Yasuda, Yukichi Umakoshi*

Department of Materials Science and Engineering, Graduate School of Engineering and Frontier Research Centre, Osaka University, 2-1 Yamada-oka, Suita, Osaka 565-0871, Japan

Received 19 November 2001; revised 3 December 2001; accepted 3 December 2001

Abstract

Two types of Fe–32 at.%Ni bicrystals containing a 90°(211) tilt or a 90°[211] twist grain boundary were prepared to investigate the effect of grain boundary character on the martensitic transformation behaviour. The martensite-start temperature (Ms) of bicrystals with the tilt boundary was significantly higher than that of single crystals, while Ms of bicrystals with the twist boundary showed no significant difference from that of single crystals. Near the tilt boundary, the coarse lenticular martensites were symmetrically formed in the neighbouring grains. In contrast, the tiny martensites were homogeneously distributed in bicrystals with the twist boundary, similar to those in single crystals. In the vicinity of the tilt and the twist boundaries, some variants with the habit plane almost parallel to the boundaries were preferentially selected among 24 variants; moreover, the equivalent variants in neighbouring grains were adjoined at the tilt boundary. As a result, the compatibility of shape strains across the boundary was maintained in the case of the tilt boundary, resulting in increasing the Ms. Such characteristic nucleation of martensites can be regarded as an example of self-accommodation across the boundary, which is called cooperative nucleation (C–N). From a crystallographic viewpoint, C–N can occur only at the symmetric tilt boundary. Effects of pre-deformation and applied stress on the heterogeneous nucleation at the boundary were also examined; C–N was always confirmed to occur at the tilt boundary, and the advantage of the boundary for nucleation of martensites did not change even under pre-deformation or applied stress. Furthermore, the martensite-start stress (σM) and the morphology of martensites in the stress-assisted transformation were strongly influenced by C–N.

Keywords: Bicrystal; Phase transformation; Martensite; Grain boundary character; Strain compatibility

1. Introduction

Martensitic transformation has an important role in material properties like the shape memory effect and the transformation induced plasticity (TRIP) phenomenon. In general, a nucleation of martensite is known to occur heterogeneously depending on lattice defects of dislocations and grain boundaries. Many researchers pointed out that grain boundaries act as a favourable nucleation site in the martensitic transformation [1,2]. Kajiwara [2] suggested that grain boundaries do not necessarily act as a favourable nucleation site; only boundaries with a special character do so, although the exact nature of such boundaries could not be identified. Although Magee [1] reported an annealing twin boundary acts as the effective nucleation site for martensitic transformation, Tsuzuki et al. [3] found no significant effect of the twin boundary on the martensitic transformation by comparing the martensite-start temperature (Ms) of bicrystals containing the annealing twin boundary with that of single crystals. Thus, the martensitic transformation behaviour in the vicinity of grain boundaries is not yet clear.

It is known that this behaviour is strongly influenced by external factors such as applied stress and magnetic field [4–10]. The effects of applied stress and pre-strain on the martensitic transformation were reported [6–10]. In particular, the effect of applied stress on several phenomena of an increase in the Ms and a selection of specific variants has been theoretically examined [11–14]. However, how the external factors affect the heterogeneously nucleated martensites at the grain boundary is still unknown.

Exact knowledge of the effect of grain boundary and external factors on the heterogeneous nucleation contributes to the development of more favourable thermo-mechanical processing through martensitic transformation, such as ausforming or to the improvement of the shape memory effect in polycrystalline materials. Control of the grain boundary character is also important for the improvement of material properties.

To understand details of the martensitic transformation in
the vicinity of grain boundaries, many controlling factors in the transformation behaviour must be simplified. Bicrystals are useful to easily control the grain boundary character and the direction of external factors to examine the effect of the boundary. The purpose of this work is to investigate the effect of grain boundary character on the martensitic transformation behaviour using Fe–Ni bicrystals focusing on the Ms, the morphology of martensites and the variant selection. In addition to the thermal transformation behaviour, the effects of pre-deformation and applied stress on the heterogeneous nucleation near grain boundaries were also investigated.

2. Preparation of Fe–Ni bicrystals

Two types of bicrystals with a nominal composition of Fe–32 at.%Ni containing a 90°(211) tilt or a 90°[211] twist boundary were prepared by a diffusion bonding method as shown in Fig. 1(a) and (b). It is also noted that the bicrystal system (x, y, z) is set up where y is normal to the grain boundary plane and z is parallel to [941]. Then, three types of tensile specimens with selected angles (ϕ = 0, 45, 90°) between the loading axis and the grain boundary plane were cut from the bicrystals as shown in Fig. 1(c). Single crystal specimens with the angle of ϕ = 0° were also prepared. The gauge dimensions of a tensile specimen are 2 mm × 0.5 mm × 3 mm. Several rods of single crystal were used in this study. These rods were carefully distinguished from each other since the Ms depends strongly on small deviations in their chemical composition.

3. Thermal transformation

3.1. The martensitic transformation behaviour in Fe–Ni bicrystals

The Ms values of Fe–32 at.%Ni single crystals and bicrystals with the tilt or the twist boundary were determined by the change in electrical resistivity during cooling. The specimens were cooled slowly at a constant rate of 1 K min⁻¹. During this process, the electrical resistivity gradually decreases and then suddenly drops at the Ms due to the burst transformation to martensite as shown in Fig. 2. The average Ms values for single crystals, bicrystals with the twist boundary and bicrystals with the tilt boundary are 115, 135 and 180 K, respectively. There is no big difference in the Ms between single crystals and bicrystals with the twist boundary, although in the Ms of bicrystals with the twist boundary there is a large scattering of Ms values. In contrast, the Ms of bicrystals with the tilt boundary is significantly higher than that of single crystals by more than 60 K. The difference in the Ms of these crystals does not come from the effect of their chemical composition since

Fig. 1. Schematic illustration of Fe–Ni bicrystals with (a) 90°(211) tilt or (b) 90°[211] twist boundaries, and (c) relation between tensile axis and grain boundary plane in tensile specimens.
these specimens were prepared from the same single crystal ingot.

Optical micrographs of thermally transformed martensites are shown in Fig. 3. The lenticular martensites which are popular in Fe-30 at.%Ni alloy were observed in all specimens tested, however, their size and distribution depend on the type of crystal. In single crystal, tiny martensites are homogeneously formed and the retained γ phase is small, as shown in Fig. 3(a). In contrast, coarse lenticular martensites are symmetrically formed in the vicinity of the tilt boundary as shown in Fig. 3(b). Moreover, there exists a great deal of retained γ phase apart from the boundary. On the other hand, the tiny martensites are homogeneously distributed in bicrystal with the twist boundary, similar to that in single crystals as shown in Fig. 3(c). This suggests that the twist boundary does not act as an effective nucleation site for the martensitic transformation although the tilt boundary strongly does. Thus, the martensitic transformation behaviour in Fe–Ni bicrystals depends strongly on the grain boundary character.

In pursuit of the dominant controlling factors for this behaviour depending on the type of boundary, the crystallographic analysis of martensites was carried out using a SEM–EBSP method. The variant selection of martensites provided useful information on the transformation behaviour near grain boundaries. All variants of the observed martensites with the bcc structure satisfied the Nishiyama–Wassermann (N–W) relationships with the γ parent phase with fcc structure; (111)₁//(110)₁ₘ and (112)₁//(110)₁ₘ. There exist 12 possible variants and 24 habit plane variants in the relationship. These variants are represented as V₁–V₁₂ and V₁ʹ–V₁₂ʹ as defined earlier [15].

In single crystals, all N–W variants are confirmed in the phase image and the (100) pole figure obtained by the EBSP analysis as shown in Fig. 4(a) and (b). In contrast, some variants are preferentially selected among 24 habit plane variants and the characteristic arrangements of variants depending on the type of boundary are observed near the grain boundary in bicrystals as shown in Fig. 4(c) and (d). Specific variants of V11 and V2' are formed in the neighbouring grains near the tilt boundary and the equivalent variants are symmetrically adjoined at the tilt boundary (Fig. 4(e)). On the other hand, the variants of V9', V8' and V7 are chosen near the twist boundary although the symmetric nucleation about the boundary is not observed near this boundary (Fig. 4(d)). All observed variants in the vicinity of both grain boundaries have habit planes almost parallel to each boundary, though the arrangement of variants was quite different between the two types of bicrystals. Thus, martensitic transformation behaviour is strongly influenced by the type of grain boundary.

3.2. Requirement for a favourable nucleation of martensite at grain boundaries

In this work the tilt boundary acted as an effective nucleation site of martensite, though the twist boundary did not. As Kajiwara reported [2], not all grain boundaries always activate the nucleation effectively. In bicrystals with the tilt boundary, coarse and equivalent martensites were symmetrically formed near the boundary. Such morphology and variant selection must effectively enhance the martensitic transformation near the tilt boundary.

Generally, a lattice deformation accompanying the martensitic transformation can be expressed by a shape deformation $\mathbf{P}$ in the cubic $(\bar{x}', \bar{y}', \bar{z}')$ system. The converted shape strain matrix is given by

$$\mathbf{\tilde{P}} = \mathbf{R}^T \mathbf{P} \mathbf{R}^T$$

where $\mathbf{R}$ is the coordinate transformation matrix and $\mathbf{R}^T$ is the transpose of $\mathbf{R}$. The shape deformation in the bicrystal
system can be described in terms of six strain components of $\varepsilon_{xx}$, $\varepsilon_{yy}$, $\varepsilon_{zz}$, $\gamma_{xz}$, $\gamma_{yz}$ and $\gamma_{zx}$. If the component crystals of A and B maintain the compatibility of shape strain at the grain boundary, the strain components must satisfy the following equations [17]

$$\varepsilon^A_{xx} = \varepsilon^B_{xx}, \quad \varepsilon^A_{zz} = \varepsilon^B_{zz}, \quad \gamma^A_{xz} = \gamma^B_{xz},$$

where superscripts A and B represent the component crystals of A and B, respectively. The shape strain matrices $\mathbf{P}$ of actually observed variants were calculated for each component crystal as follows

$$\mathbf{P}_{V11}^{A,\text{tilt}} = \begin{pmatrix}
0.9454 & -0.2044 & -0.0022 \\
0.0195 & 1.0729 & -0.0196 \\
-0.0010 & -0.0577 & 0.9996
\end{pmatrix},$$

$$\mathbf{P}_{V7}^{B,\text{twist}} = \begin{pmatrix}
1.0190 & -0.0846 & -0.0407 \\
-0.0099 & 1.0440 & 0.0106 \\
0.0229 & -0.1972 & 0.9537
\end{pmatrix},$$

In bicrystal with the tilt boundary, the observed variant of V11 has the equivalent matrix components of $(\varepsilon_{xx}, \varepsilon_{zz}, \gamma_{xz})$ in each component crystal. Thus, the compatibility of shape strains at the tilt boundary is maintained when the same variants in neighbouring grains are adjoined at the boundary. In contrast, the combinations of variants that can satisfy the compatibility requirement at the boundary never exist in the twist boundary. This indicates that even if any variant is adjoined at the twist boundary, the strain compatibility at the boundary cannot be maintained. If the compatibility of shape strains at the boundary is perfectly maintained, the grain boundary does not disturb the transformation since the nucleated variants in the neighbouring grain do not constrain each other at the boundary. In contrast, when the shape strains at the boundary are incompatible, the parent phase deforms plastically in order to maintain the continuity of strain at the boundary. As a result, the Ms of bicrystal with the tilt boundary is higher than that with the twist boundary.

In thermoelastic martensitic transformation, several variants constitute a group to reduce the strain energy by themselves. This is called self-accommodation, which has
been experimentally and theoretically established by Saburi and Wayman [18]. Then, this approach is applied to the martensitic transformation behaviour near the grain boundary. When the compatibility of shape strains at the tilt boundary is maintained, the average shape strain matrix of martensites formed near the boundary is close to the identity matrix compared with the original one as follows:

\[
\frac{1}{2} (P_{V11}^{A, \text{init}} + P_{V11}^{B, \text{init}}) = \begin{pmatrix}
0.9454 & 0 & -0.0022 \\
0 & 1.0729 & 0 \\
-0.0100 & 0 & 0.9996 \\
\end{pmatrix}
\]

(5)

The reduction of shape strain can be regarded as a kind of self-accommodation across the boundary. We call such characteristic nucleation as 'cooperative nucleation (C–N)’. This C–N must enhance the heterogeneous nucleation of martensites in the vicinity of the tilt boundary, and results in an increase in the Ms of bicrystals with the tilt boundary.

4. Thermal transformation after pre-deformation

4.1. Change in the Ms by pre-straining

The effect of prior deformation on the martensitic transformation has been investigated using Fe–Ni and Fe–Ni–Al alloys [19,20]. However, no exact information about Ms is reported in polycrystalline materials. To obtain more clear information about the effect of pre-straining on the martensitic transformation, various amounts of pre-strain were introduced in single crystal and bicrystals with the tilt and the twist boundary at room temperature by tensile deformation at \( \phi = 0^\circ \). After releasing the load, the Ms was determined by electrical resistivity measurement in the same way as the thermal transformation. Even slight deformation rapidly decreased the Ms in all crystals. This retardation of the martensitic transformation by pre-straining is known as the mechanical stabilisation of austenite. With further deformation, the Ms of single crystals gradually increased with increasing pre-strain while neither bicrystals showed any remarkable change in the Ms. However, it should be noted that the Ms of bicrystals with the tilt boundary is always higher than that of other crystals in the measured pre-strain range up to 35%. Hence, the advantage of the tilt boundary for the nucleation of martensites does not change even after pre-straining.

4.2. Characteristic morphology of martensites in pre-deformed bicrystal with the tilt boundary

The most interesting morphology of martensites in pre-deformed crystals was seen in bicrystal with the tilt boundary, which showed higher Ms than the other crystals. Fig. 5 shows the optical micrograph and the corresponding orientation image by the EBSP analysis in the vicinity of the tilt boundary. Large amounts of tiny lenticular martensites align nearly parallel to the boundary, forming a lamellar-like structure, as shown in Fig. 5(a). The occurrence of C–N of V11 or V2′ is confirmed at the tilt boundary (Fig. 5(b)); this variant selection is consistent with that in thermal transformation without pre-straining. Furthermore, the observed lamellar structure of martensites is composed of alternately arranged variants such as V11–V12′–V11 or V2′–V3–V2′. The formation process of such characteristic morphology can be predicted as schematically shown in Fig. 6. First, the C–N of V11 or V2′ occurs at the tilt boundary (Fig. 6(a)). Subsequently, alternate formation of specific variants is activated by the shape deformation of the previously nucleated variant (Fig. 6(b)). In this way, a limited set of variants is continuously nucleated depending on the neighbouring variants and the lamellar structure develops in the intergranular region (Fig. 6(c)). The tilt boundary obviously acts as an effective nucleation site of martensites since the lamellar structure composed of the limited set of variants is never obtained without preferential nucleation at the boundary. To prove the validity of this assumption, the transmission of shape deformation at the interface of variants was estimated using the stress transmission factor, \( N_{ij} \) proposed by Livingston and Chalmers [17], which is given as follows

\[
N_{ij} = (e_A \cdot e_B)(g_A \cdot g_B) + (e_A \cdot g_B)(e_B \cdot g_A)
\]

(6)

where \( e \) and \( g \) are a slip plane normal and a slip direction, respectively. The subscripts denote the component crystals A and B. In general, the \( N_{ij} \) value is used to evaluate the transmission of plastic deformation across a grain boundary. In the present study, a slip plane and a slip direction are regarded as a habit plane and a shear direction, respectively,
in order to apply the $N_{ij}$ value for the transmission of shape deformation at the interface of variants. As the $N_{ij}$ approaches $-1$, the shape deformation is easily transmitted to the neighbouring variant from the viewpoint of self-accommodation. In addition, the neighbouring variants must have close habit planes in order to effectively transmit the shear strain. The $N_{ij}$ values were calculated for all variants against V11 and V2', which were selected in the vicinity of the tilt boundary, and are listed in Table 1. Actually, the observed variant pairs V11–V12' and V2’–V3 have an $N_{ij}$ value close to $-1$ and a close habit plane relation. Thus, the formation process of lamellar structure can be quantitatively explained by calculating $N_{ij}$ values.

Table 1
The stress transmission factors, $N_{ij}$ for the combinations of variants against V11 and V2' formed near the tilt boundary

|       | V11 | V1' | V2' | V2' | V11 | V2' |
|-------|-----|-----|-----|-----|-----|-----|
| V1    | 0.479 | 0.625 | -0.489 | -0.371 |
| V2    | 0.330 | 0.434 | 0.852 | -   |
| V3    | -0.874 | -0.816 | -0.423 | -0.371 |
| V4    | 0.486 | 0.596 | -0.490 | -0.429 |
| V5    | 0.237 | 0.330 | 0.797 | 0.924 |
| V6    | -0.923 | -0.918 | -0.510 | -0.493 |
| V7    | -0.417 | -0.485 | 0.601 | 0.478 |
| V8    | 0.929 | 0.795 | 0.331 | 0.246 |
| V9    | -0.490 | -0.490 | 0.511 | -0.916 |
| V10   | -0.383 | -0.500 | 0.645 | 0.483 |
| V11   | -   | 0.852 | 0.437 | 0.337 |
| V12   | -0.387 | -0.426 | -0.828 | -0.876 |

 Apparently, unless the nucleation of martensite first occurs in the vicinity of the tilt boundary, the lamellar structure cannot be formed. Observation of the lamellar structure suggests that the tilt boundary acts as a favourable nucleation site of martensites.

5. Stress-assisted transformation

5.1. Effect of applied stress on the transformation behaviour in the vicinity of grain boundary

Martensitic transformation was induced by an applied stress at $\phi = 0^\circ$ in single crystal and both bicrystals. A tensile stress assisted the martensitic transformation above the Ms in all crystals. Fig. 7 shows the morphology of stress-induced martensites in Fe–Ni single crystal and bicrystals. In bicrystals with the tilt boundary (Fig. 7(b)), the coarse lenticular martensites are symmetrically formed about the boundary accompanied by a large quantity of the retained $\gamma$ phase apart from the boundary. In single crystals (Fig. 7(a)) and bicrystals with the tilt boundary (Fig. 7(b)), each martensite becomes coarse because of applied stress, though the morphology of martensites is similar to that in thermal

Fig. 7. Optical micrographs of stress-induced martensites in Fe–Ni single crystal (a) and bicrystals with the tilt (b) or the twist (c) boundary.
transformation. On the other hand, the stress-induced martensites in bicrystals with the twist boundary (Fig. 7(c)) tend to nucleate near the boundary though the twist boundary is not necessarily favourable for the nucleation intrinsically in the thermal transformation. Thus, the applied stress enhances the heterogeneous nucleation at both tilt and twist boundaries.

An applied stress is known to assist the nucleation of specific variants. Patel and Cohen [6] proposed the following equation to examine the contribution of applied stress to the formation of martensite

\[ U/\sigma = \gamma \cos \theta \cos \lambda + \varepsilon \cos^2 \theta \]  

(7)

where \( U \) is work done, \( \sigma \) applied stress, \( \gamma \) transformation shear strain in the habit plane, \( \varepsilon \) transformation strain normal to the habit plane, \( \theta \) angle between the habit plane normal and the loading axis and \( \lambda \) angle between the shear direction and the loading axis, respectively. The \( U/\sigma \) values at \( \phi = 0^\circ \) were calculated for 24 habit plane variants and are listed in Table 2. Large positive \( U/\sigma \) values are believed to be favourable for the transformation in tension.

The observed variants in bicrystals with the twist bound-

\[ \sigma_M (at \phi = 0^\circ) \gg \sigma_M (at \phi = 45^\circ) > \sigma_M (at \phi = 90^\circ) \]  

(8)

On the other hand, the grain boundary character influenced the variant selection near the tilt boundary even under applied stress. Specific variants of V11 or V2', which have the habit plane nearly parallel to the boundary, were always selected in the vicinity of the tilt boundary in all loading axes. The onset of martensitic transformation must be controlled by the formation of the variants. Therefore, the \( U/\sigma \) values of V11 or V2' at each loading axis were calculated to examine the contribution of the applied stress to the formation of the variants. The \( U/\sigma \) value of V11 or V2' varies depending on the angle \( \phi \), and the following relation is obtained

\[ U/\sigma (at \phi = 0^\circ) \ll U/\sigma (at \phi = 45^\circ) < U/\sigma (at \phi = 90^\circ) \]  

(9)

Generally, the influence of external stress on the Ms has been explained using the equation of Clausius–Clapeyron or its modification [21,22]. The required driving force for the martensitic transformation can be represented by two terms of chemical driving force and mechanical work done. At a constant temperature, the \( \sigma_M \) is only influenced by work done since the chemical driving force is constant. In single crystal, a specific variant with the largest \( U/\sigma \) value may act as a trigger for burst transformation under applied stress. This hypothesis could explain the previous results obtained by Patel and Cohen [6] or Hosoi et al. [23]. In bicrystals with the tilt boundary, the variants formed at the onset of transformation were limited by the effect of the grain boundary character. As a result, the \( \sigma_M \) changes.

Table 2

| Variant notation | \( U/\sigma \) | Variant notation | \( U/\sigma \) |
|------------------|---------------|-----------------|---------------|
| V1               | -0.0977       | V1'             | -0.0637       |
| V2               | 0.0787        | V2'             | -0.0397       |
| V3               | 0.0159        | V3'             | 0.100         |
| V4               | -0.103        | V4'             | -0.0745       |
| V5               | 0.0729        | V5'             | -0.0430       |
| V6               | 0.00374       | V6'             | 0.0913        |
| V7               | -0.0468       | V7'             | -0.0577       |
| V8               | -0.00113      | V8'             | 0.0922        |
| V9               | 0.101         | V9'             | 0.0190        |
| V10              | -0.0599       | V10'            | -0.0616       |
| V11              | -0.0004       | V11'            | 0.0954        |
| V12              | 0.101         | V12'            | 0.0108        |

Fig. 8. Change in martensite-start stress (\( \sigma_M \)) in Fe–Ni bicrystals with the tilt boundary depending on the loading axis at \( \phi = 0^\circ \) (●), 45° (○) and 90° (△). Solid marks indicate yield stress (\( \sigma_{0.2} \)).
depending on the $U/\sigma$ values for the variants nucleated at the boundary. As the $U/\sigma$ value increases, the $\sigma_M$ decreases. In fact, the experimental result coincides with the expected one based on the $U/\sigma$ values. Moreover, the $\sigma_M-\Gamma$ relation at the angle $\phi$ in bicrystals with the tilt boundary could be quantitatively estimated [24], although the detail is not shown here. Thus, the onset of martensitic transformation under applied stress is controlled by the interaction between the first-nucleated martensite and the contribution of applied stress for the variant.

6. Summary and conclusions

Effect of grain boundary character on the martensitic transformation in Fe–Ni bicrystals was examined, and the strong dependence of this transformation on the character was confirmed. The following conclusions were reached based on the results obtained.

1. The Ms temperature, the morphology and the variant selection in Fe–Ni bicrystals depend strongly on the type of grain boundary. A 90°(211) tilt boundary acts as a favourable nucleation site for the martensitic transformation in Fe–Ni bicrystals and increases the Ms, while a 90°(211) twist boundary does not effectively activate the martensitic transformation.

2. Some variants whose habit plane is nearly parallel to the grain boundary are preferentially formed near the boundary among 24 N–W variants. In addition, equivalent variants are adjointed at the 90°(211) tilt grain boundary to maintain the compatibility of strains accompanied by the martensitic transformation at the boundary. The cooperative nucleation occurs.

3. Pre-straining suppresses the martensitic transformation and decreases the Ms in Fe–Ni single crystals and bicrystals. The Ms of bicrystals with the tilt boundary is always higher than that of the other crystals tested in the entire pre-strain range up to about 35%.

4. The lamellar morphology composed of a limited set of variants was obtained in pre-strained bicrystals with the tilt boundary. The morphology is closely related to the favourable nucleation at the tilt boundary and the transmission of shape strain from one variant to another.

5. An applied stress enhances the potency of nucleation at both the tilt and the twist boundary, and assists the nucleation of limited variants with large positive $U/\sigma$ values. However, since the effect of grain boundary character at the tilt boundary is so strong even under applied stress, the variant selection in the vicinity of the tilt boundary is not sensitive to the $U/\sigma$ values.

6. The $\sigma_M$ of bicrystals with the tilt boundary varies depending on the loading direction about the boundary. Moreover, the $\sigma_M$ can be estimated by the interaction between the first-nucleated martensites at the boundary and the contributions of applied stress to the formation of the variants.

Acknowledgements

This work was supported by a Grant-in-Aid for Scientific Research on the Priority Area B Harmonic Material Design of Multi-Functional Composites from the Japanese Ministry of Education, Culture, Sports, Science and Technology.

References

[1] C.L. Magee, The nucleation of martensite, Phase Transformations, ASM, Metals Park, OH, 1969, pp. 115–156.
[2] S. Kajiwara, Roles of dislocations and grain boundaries in martensite nucleation, Metall. Trans. A 17A (1986) 1693–1702.
[3] K. Tsuchiya, Q. Harada, T. Maki, Potency of grain boundaries as martensitic nucleation site, J. Phys. IV 5 (C8) (1995) 167–172.
[4] T. Kakeshita, K. Shimizu, S. Funada, M. Date, Composition dependence of magnetic field-induced martensitic transformations in Fe–Ni alloys, Acta Metall. 33 (1985) 1381–1389.
[5] T. Kakeshita, K. Shimizu, R. Tanaka, S. Nakamichi, S. Endo, F. Ono, Effect of spontaneous volume magnetostriiction on martensitic transformation in an Fe–29.9 at.%Ni invar alloy under hydrostatic pressures, Mater. Trans., JIM 32 (1991) 1115–1119.
[6] J.R. Patel, M. Cohen, Criterion for the action of applied stress in the martensitic transformation, Acta Metall. 1 (1953) 531–538.
[7] H. Onodera, H. Oka, I. Tamura, A role of strain on deformation-induced martensitic transformation in an Fe–Ni–C alloy, J. Jpn. Inst. Metals 42 (1978) 898–905.
[8] R.H. Richman, G.F. Bolling, Stress, deformation, and martensitic transformation, Metall. Trans. 2 (1971) 2451–2462.
[9] D. Goodchild, W.T. Roberts, D.V. Wilson, Plastic deformation and phase transformation in textured austenitic stainless steel, Acta Metall. 18 (1970) 1137–1145.
[10] J.F. Breedis, Influence of dislocation substructure on the martensitic transformation in stainless steel, Acta Metall. 13 (1965) 239–250.
[11] G.H. Olsen, W.A. Jesser, The effect of applied stress on the F.C.C.–B.C.C. transformation in thin iron films, Acta Metall. 19 (1971) 1299–1302.
[12] Y. Higo, F. Lecroyseiy, T. Mori, Relation between applied stress and orientation relationship of $\alpha$ martensite in stainless steel single crystals, Acta Metall. 22 (1974) 313–323.
[13] M. Kato, T. Mori, Stress-induced martensite in single crystals of an Fe–23Ni–5Cr alloy, Acta Metall. 24 (1976) 853–860.
[14] Y. Murakami, K. Otsuka, S. Hanada, S. Watanabe, Self-accommodation and morphology of 14M(7R) martensites in an Ni–37.0 at.%Al alloy, Mater. Sci. Eng. A139 (1994) 191–199.
[15] M. Ueda, H.Y. Yasuda, Y. Umakoshi, Effect of grain boundary character on the martensitic transformation in Fe–32 at.% Ni bicrystals, Acta Mater. 49 (2001) 3421–3432.
[16] M.S. Wechsler, D.S. Lieberman, T.A. Read, On the theory of the formation of martensite, Trans. AIME 197 (1953) 1503–1515.
[17] J.D. Livingston, B. Chalmers, Multiple slip in bicrystal deformation, Acta Metall. 5 (1957) 322–327.
[18] T. Saburi, C.M. Wayman, Crystallographic similarities in shape memory martensites, Acta Metall. 27 (1979) 979–995.
[19] M. Tokizane, The effect of tensile pre-straining on nucleation of martensite in an Fe–Ni–C alloy, Scripta Metall. 10 (1976) 459–462.
[20] N. Jost, E. Hornbogen, Influence of prior plastic deformation on the thermal-martensitic transformation in an Fe–Ni–Al alloy, J. Mater. Sci. Lett. 6 (1987) 491–494.
[21] J. Ortin, A. Planes, Thermodynamics of thermoelastic martensitic transformations, Acta Metall. 37 (1989) 1433–1441.

[22] M. Andersson, R. Stalmans, J. Ågren, Unified thermodynamic analysis of the stress-assisted $\gamma \rightarrow \epsilon$ martensitic transformation in Fe–Mn–Si alloys, Acta Metall. 46 (1998) 3883–3891.

[23] Y. Hosoi, Y. Kawakami, Effect of plastic deformation and stress on martensite transformation in Fe–Ni alloy, Iron and Steel 12 (1963) 1780–1787.

[24] M. Ueda, H.Y. Yasuda, Y. Umakoshi, Stress-induced martensitic transformation in Fe–Ni bicrystals, Acta Mater. 49 (2001) 4251–4258.