In-situ Observation of Butterfly-type Martensite in Fe–30mass%Ni Alloy during Tensile Test Using High-resolution EBSD

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The microstructural change of the thermal butterfly-type martensite (α’) in Fe–30mass%Ni alloy during a tensile test was investigated by in-situ observation using high-resolution electron backscatter diffraction (EBSD). When the specimen is plastically deformed, the α’ plate in a thermal butterfly-type α’ starts to grow toward its width direction from the α’–γ interface with (2 2 5)γ habit plane. As the α’ plate grows, the crystal orientation relationship (OR) between α’ and γ changes from Greninger–Troiano to Kurdjumov–Sachs. Moreover, an orientation gradient is formed in the α’ plate as its growth proceeds. The latter is due to the inheritance of the piled-up transformation dislocations into the α’ plate. On the other hand, the orientation gradient in the γ matrix around the α’–γ interface decreases. These results indicate that the thermal butterfly-type α’ changes its character to lath-type during growth. This is caused by the change of the accommodation process from single slip for butterfly-type martensite to multiple dislocation slip due to the tensile deformation at R.T. It is concluded that the growth behaviour of the butterfly-type α’ during tensile deformation depends on the change of the accommodation process.

KEY WORDS: electron backscatter diffraction (EBSD); phase transformation; martensite; iron alloy; crystal orientation relationship (OR).

1. Introduction

Martensite phases (α’) are separated into several types depending on its morphologies.1–6) The morphologies of the α’ are determined by the accommodation mechanism for the transformation strain. Because the accommodation process strongly depends on the temperature, the types of the α’ are mainly determined by the formation temperature (Martensite start temperature: Ms). In addition, the crystal orientation relationship (OR) between the α’ crystals and the γ matrix crystals is also affected by Ms.3–6)

Table 1 shows features of the different α’ types in Fe–Ni alloys.1–13) When α’ is formed at very low temperature such as liquid nitrogen temperature, it develops into a thin plate-type due to the accommodation process by twinning.3–11) The thin plate-type α’ has the habit plane {3 1 0 1 5}γ and satisfies the Greninger–Troiano (G-T) ( {1 1 1}α’ || {0 1 1}γ, {1 1 1}α’ || 2.5° from {0 1 1}γ ) OR with the γ matrix. At slightly higher temperatures lenticular-type α’ is often formed.4,6,9) Because the transformation strain for the lenticular-type α’ is accommodated by twinning and single-slip dislocation, this α’ has {3 1 0 1 5}γ or {2 5 9}γ habit planes and its OR with the γ matrix is of G-T, Nishiyama–Wassermann (N-W) ( {1 1 1}γ || {0 1 1}α’, {1 1 1}γ || 2.5° from {0 1 1}α’ ) or Kurdjumov–Sachs (K-S) ( {1 1 1}γ || {0 1 1}α’, {1 0 1}γ || {1 1 1}α’ ) types. At intermediate temperatures, both, dislocation slip on limited slip systems and twinning act as accommodation process for the transformation strain. As a result, the morphology of the α’ becomes butterfly-like.1,2,5) The habit plane of the butterfly-type α’ is {2 2 5}γ, and its OR has been reported as G-T, N-W or K-S ORs. As described above, for the lenticular-type α’ and the butterfly-type α’, several ORs have been reported.4–6) This is due to the fact that both types have heterogeneous defect structures created during the accommodation process.4–6) At high temperatures, lath-type α’ is formed by the accommodation due to multiple dislocation slip.3,14,15) The lath-type α’ often exhibits K-S OR with the γ matrix, and its habit plane is {5 5 7}γ or {1 1 1}γ.

As shown in Table 1, the features of the α’ are mainly determined by Ms. However, it has been reported that the features of the lenticular and the butterfly-type α’ depend on the transformation temperature during its growth. Shibata et al. investigated the microstructure of lenticular-type α’ in Fe–33mass%Ni alloy using electron backscatter diffraction (EBSD).41) According to their study, the midrib part containing twins has G-T OR with the γ matrix, similarly to the case of thin plate-type α’. On the other hand, the untwined part near the α’–γ interface contains a high dislocation density and satisfies K-S OR with the γ matrix, similarly to the lath-type α’. They suggested from these observations that the lenticular-type α’ is at first formed with the character close to the thin plate-type α’ under very low temperature, and then grows with the features of lath-type α’ because of the heat generation caused by martensitic transformation.41) On the other hand, the present authors...
have investigated about the spatial orientation distribution and OR between a butterfly-type $\alpha'$ plate and its $\gamma$ matrix in Fe–30mass%Ni alloy.\(^5\) In the butterfly-type $\alpha'$, one $\alpha'$–γ interface of the $\alpha'$ plate mainly exhibits a G-T OR while the other side shows both K-S and G-T ORs. In addition, the $\alpha'$–γ interface which exhibits exclusively G-T OR has a habit plane of $\{2\,5\,9\}_\gamma$, while the other one exhibits a habit plane of $\{2\,2\,5\}_\gamma$ and $\{5\,5\,7\}_\gamma$. Based on these results, they proposed that the butterfly-type $\alpha'$ is formed with the lath-type $\alpha'$ character at $M_S$ and then, as the butterfly-type $\alpha'$ grows, its character changes to the lenticular-type $\alpha'$ because of the drop of transformation temperature.\(^5\) From these growth mechanisms, it is seen that the character of the $\alpha'$ is determined not only by $M_S$ but also by the transformation temperature during its growth.

The growth processes of the $\alpha'$ phase described above have been inferred from microstructural observations of already grown $\alpha'$. Therefore, the growth process of the $\alpha'$ itself is still unclear. In order to clarify these formation processes, it is necessary to carry out in-situ observation. In the case of thermally induced martensite formation, however, an in-situ observation is very difficult because of its very fast growth. Another way to induce the growth of the $\alpha'$ is the strain-induced martensitic transformation. Zhang et al. have investigated the morphological change of thermal butterfly-type $\alpha'$ in Fe–30mass%Ni alloy during a tensile test by comparing the microstructure of a non-deformed specimen with that of the deformed one.\(^12,13\) In their study they observed strain-induced growth of the original thermal butterfly-type $\alpha'$ and showed that the martensite character changed from butterfly-type to lath-type. In addition, Maki et al. reported that the microstructure of thermal $\alpha'$ is essentially the same as that of strain-induced $\alpha'$.\(^10\) It can therefore be assumed that in-situ observations of strain-induced martensite may be used to conclude on the temperature dependent growth of thermal martensite.

In the present work we report on an in-situ observation of the microstructural changes of a butterfly-type $\alpha'$ in Fe–30mass%Ni alloy during a tensile test using EBSD-based orientation microscopy. Because EBSD can analyze the crystal orientation of materials under bulk, its distribution can be obtained with small strain release. From the results of the microstructural change of the $\alpha'$ during its growth, growth process of the butterfly-type $\alpha'$ are discussed. As shown in details later, we found that the butterfly-type $\alpha'$ grows by tensile test, and that morphology of the $\alpha'$ and the ORs between the $\alpha'$ and the $\gamma$ matrix are continuously changed during its growth.

## 2. Experimental Procedures

### 2.1. Preparation of Specimens for in-situ Observation of $\alpha'$ during Tensile Test

An Fe–30mass%Ni alloy ingot was prepared by induction melting in vacuum, and then homogenized at 1200°C for 20 h. After the heat treatment, the ingot was hot forged at 1100°C. Specimens with 40 mm×15 mm surface area and 1.5 mm thickness were cut from the hot forged sample. The specimens were austenitized at 1100°C for 30 min, and then water quenched. Temperature of the water used for the quench was 15°C. In previous studies, it had been reported that the $M_S$ temperatures of Fe–30mass%Ni alloy and Fe–29mass%Ni alloy are around R.T. and 73°C, respectively.\(^1,10\) Hence, in the present specimen, $\alpha'$ was formed during water quenching. Tensile specimens for in-situ observation were spark-cut from the quenched specimens and then mechanically and electrolytically polished. The electro-polishing was carried out using Struers AC2 electrolyte at 20°C and 32 V for 20 s in a Struers Electropol device. Finally, tensile specimens with 4 mm gage length and 1 mm×0.3 mm cross section were obtained.

### 2.2. In-situ Observation of $\alpha'$ with High-resolution EBSD during Tensile Test

In-situ observation of $\alpha'$ during tensile test was made by EBSD-based orientation microscopy using a field emission gun scanning electron microscope (FE-SEM, JEOL6500F) with EBSD attachment (OIM of EDAX/TSL). All observations were made at 15 kV acceleration voltage and a working distance of 15 mm. Tensile test was performed in a tensile-test machine of Kammrath & Weiss installed inside the FE-SEM.

Tensile test was carried out at a strain rate of $1.1 \times 10^{-3}$ 1/s at R.T. At total nominal strains of 2.0%, 4.6%, 10.4% and 14.5%, tensile test was paused and microstructural observations were performed using EBSD. Orientation mapping was performed with a step size of 0.3 µm using pattern analysis parameters leading to an angular resolution of 0.5°. From the obtained orientation maps, the ORs between $\alpha'$ and $\gamma$ were determined using the axis-angle representation of misorientations between the $\alpha'$ and $\gamma$ crystals. Table 2 shows the axis-angle values calculated for the exact K-S,
Table 2. Axis-angle values of orientation relationship between \(
\alpha'\) and \(\gamma\). \(^{5,17}\)

| ORs between \(\alpha'\) and \(\gamma\) | Axis-angle |
|-----------------------------------|------------|
| K-S                               | 42.85° < 17.8 17.8 96.8 > |
| G-T                               | 44.26° < 12.2 18.4 97.5 > |
| N-W                               | 45.99° < 8.3 20.1 97.6 > |
| Bain                              | 50.00° < 0   0   1 > |

N-W, G-T and Bain ORs \(^{5,17}\). The actually occurring ORs were determined by comparing the experimentally obtained axis-angle pairs with the calculated ones. Moreover, in order to determine the habit planes of \(\alpha'\), the traces of well-known habit planes were drawn on the orientation maps and then that one which was most parallel to the actual interface trace was selected to be the correct one.

3. Results and Discussion

3.1. Microstructural Change of Butterfly-type \(\alpha'\) during Tensile Test

Figure 1 is a stress–strain curve of Fe–30mass%Ni alloy obtained from an in-situ tensile test. Microstructural observations using EBSD were performed by pausing the tensile test at nominal strains of \(\varepsilon=2.0\%\), 4.6\%, 10.4\% and 14.5\%. Observation point of \(\varepsilon=2.0\%\) is in the region of elastic deformation, while all other points are the plastic regime. The tensile test was stopped when the stress exceeded the tensile strength, because the surface of the tensile specimen became rough as will be shown later.

Figure 2 shows an overview on the microstructure of the tensile sample obtained by orientation microscopy at relatively low magnification. Figure 2(a) is an image quality map of the initial microstructure of Fe–30mass%Ni alloy together the results of a trace analysis for \(\alpha'–\gamma\) interface. In this micrograph, solid and broken lines indicate the traces of \(\{5\ 5\ 7\}\_\gamma\) and \(\{2\ 2\ 5\}\_\gamma\), respectively. After water quenching, the Fe–30mass%Ni alloy develops several \(\alpha'\) plates. In these \(\alpha'\) plates, no midrib is observed. From the results of trace analysis for \(\alpha'–\gamma\) interface, one interface in the \(\alpha'\) plate has a \(\{5\ 5\ 7\}\_\gamma\) habit plane, while the other side exhibits a \(\{2\ 2\ 5\}\_\gamma\) plane. As shown in Table 1, \(\{5\ 5\ 7\}\_\gamma\) and \(\{2\ 2\ 5\}\_\gamma\) are the ideal habit planes of lath-type \(\alpha'\) and butterfly-type \(\alpha'\), respectively.\(^{1–13}\) In addition, the present authors reported earlier that a butterfly-type \(\alpha'\) in Fe–30mass%Ni alloy consists of an interface with a \(\{2\ 2\ 5\}\_\gamma\) habit plane and one with \(\{2\ 5\ 9\}\_\gamma\).\(^{3}\) In case of the present \(\alpha'\) plate, although the boundary character in terms of habit plane shows both, lath-type and butterfly-type characters, the microstructure of the present \(\alpha'\) plate is closer to that of butterfly-type than that of lath-type. Therefore, the \(\alpha'\) observed in the present Fe–30mass%Ni alloy can be referred to as butterfly-type \(\alpha'\).

Figures 2(b), (c) and (d) show the microstructures of the Fe–30mass%Ni alloy at \(\varepsilon=2.0\%\), 4.6\% and 14.5\%, respectively. As can be seen in Fig. 2(b), no microstructural change is observed in the specimen during elastic deformation. Exceeding the nominal strain of \(\varepsilon=4.6\%\), the size of the \(\alpha'\) plates gets larger and slip lines become visible in the \(\gamma\) matrix. Especially, the large \(\alpha'\) plates thicken and the small \(\alpha'\) plates lengthen as tensile strain increases. When the nominal strain is \(\varepsilon=14.5\%\), slip lines are clearly visible inside both, the \(\alpha'\) plates and the \(\gamma\) matrix, and the surface of the specimen becomes rough as shown in Fig. 2(d). Moreover, no microstructural changes of the \(\alpha'\) plates were observed when the tensile load was removed. Therefore, the present \(\alpha'\) would not be thermoelastic \(\alpha'\).

It is well known that the surface roughness of specimen affects on the result obtained by EBSD. Although the surface of present specimen becomes rough with increasing tensile strain, influence of its surface roughness on the results in this study would be small. This is because the original part in the \(\alpha'\) plate (thermally transformed part) keeps almost same orientation distribution. In this study, growth process of the \(\alpha'\) during tensile test is discussed as the influence of the surface roughness in deformed specimen is
Figure 3 displays an orientation map of the initial \(\alpha'\) plate marked by a white rectangle in Fig. 2(a). Figure 3(a) shows ORs of the \(\alpha'\) plate together with results of trace analysis for the interface. We denominate the left-hand interface of the \(\alpha'\) plate as the “interface A” and right-hand one as the “interface B”. Red, yellow and green lines show \(\alpha'\)–\(\gamma\) interface with K-S, G-T and N-W ORs, respectively. It can be seen that both \(\alpha'\)–\(\gamma\) interfaces of the initial \(\alpha'\) plate satisfy G-T OR. Habit planes for \(\alpha'\)–\(\gamma\) interface of the \(\alpha'\) plate are \{5 5 7\}_\(\gamma\) at interface A and \{2 2 5\}_\(\gamma\) at interface B. It is therefore again concluded that this \(\alpha'\) plate is close to butterfly-type.

Figures 4(a) through 4(d) are image quality maps showing microstructures of the \(\alpha'\) plate shown in Fig. 3 at nominal strains of 2.0%, 4.6%, 10.4% and 14.5%, respectively. At the nominal strain of \(\varepsilon=2.0\%\), microstructure of the \(\alpha'\) plate is the same as the initial one in Fig. 3(a). When the nominal strain exceeds \(\varepsilon=4.6\%\), the interface B moves toward the right direction and the width of the \(\alpha'\) plate increases as the nominal strain increases. On the other hand, no movement of interface A is observed. After the movement of the interface B, a microstructure with a similar appearance as a midrib is formed at the original position of the interface A. In this paper, we will therefore refer to this as midrib. The habit plane of the midrib keeps \{2 2 5\}_\(\alpha'\) during the tensile test. With respect to the surrounding \(\alpha'\) it shows a \{1 1 2\}_\(\alpha'\) trace. For nominal strains higher than \(\varepsilon=4.6\%\) slip lines are observed in the \(\gamma\) matrix. At \(\varepsilon=14.5\%\) one set of slip lines in the \(\alpha'\) is visible as well.

Although the above observations seem to indicate the growth of the initial \(\alpha'\) plate in Fig. 3, it is difficult to determine it only from Figs. 3 and 4. In order to make it clear, the continuity of crystal orientation across the \(\alpha'\) plate at a nominal strain of \(\varepsilon=14.5\%\) was investigated. The results are displayed in Fig. 5. Figures 5(b) and 5(c) show sets of inverse pole figures of the \(\alpha'\) plate at positions of (1) through (4) in (a). In (b) and (c), right side and left side inverse pole figures are ones of ND and TD direction shown in (a).
continuously from (1) to (8). This means that the movement of the interface B is due to the growth of the initial α′ plate in Fig. 3. Therefore, it is found that the present α′ plate grows toward its width direction from the midrib with {2 2 5}γ habit plane.

Umemoto et al. have suggested a growth process of thermal butterfly-type α′ based on its microstructure.2) According to their study, the butterfly-type α′ nucleates at the α′–γ interface with a {2 2 5}γ habit plane and then grows to one side or both sides from the original α′–γ interface. Although this growth process is valid for thermal butterfly-type α′, we make the same observations for the strain induced growth. This was also reported by Maki et al.10) The growth process reported by Umemoto et al. supports our growth process of the α′ plate induced by tensile test.2)

3.2. Change of Crystal Orientation Relationship between α′ and γ during Tensile Test

In order to investigate the α′–γ OR change during the growth of the α′ plate the axis-angle values of the interface B are displayed in stereographic projections in Figs. 6(a), 6(b) and 6(c), for nominal strains of ε=0%, 4.6% and 14.5%, respectively. In these stereographic projections, the precise K-S, G-T and N-W ORs are marked as reference. At the nominal strain of ε=0%, almost all of points are distributed around the precise G-T OR with a maximum deviation of 3°. Only few points are found around K-S and N-W ORs. When the nominal strain increases up to ε=4.6%, few points start to accumulate around the precise K-S OR. However, the main OR between α′ and γ is still G-T OR. With further increase of the nominal strain, the main α′–γ OR moves towards the K-S OR, as shown in Fig. 6(c). Although G-T OR is still observed at ε=14.5%, the amount of the points around G-T OR is smaller than that of K-S OR. Moreover, there is no point around the N-W OR. As will be explained later, this is because the type of the α′ plate changes from butterfly-type to lath-type during its growth since multiple dislocation slip becomes active as accommodation process. Summarizing the above results, the α′–γ OR of the thermal butterfly-type α′ changes from G-T to K-S during its strain induced growth.

Besides boundary character Fig. 3(b) shows the orientation gradients occurring inside the α′ plate and the γ matrix in the specimen before tensile test. The orientation gradients are calculated with respects to the crosses in the image and displayed in form of color gradient. More details of the orientation gradients from white point in the α′ plate and the γ matrix are given by the misorientation profiles measured along the lines indicated in Fig. 7 for the α′ and Fig. 8 for the neighboring γ. All measurements are measured parallel or perpendicular to interface B. The orientations inside the α′ plate in Fig. 7(a) continuously change from interface A to the interface B covering a total misorientation of up to about 3° across a distance of 2 μm. On the other hand, no orientation gradient parallel to the interface B is observed. Also, the orientation gradient perpendicular to the interface B in the γ matrix covers a total misorientation of 4° across a distance of 2.5 μm, while no orientation gradient parallel to the interface B is observed in Fig. 8(a). In our previous study, we reported the same orientation gradient profiles for thermal butterfly-type α′.5) According to
In this study, such orientation gradients in the $\alpha'$ plate and the $\gamma$ matrix have a common rotation axis and are generated by dislocation inheritance during the growth of the $\alpha'$ plate by quenching. The evolution of the orientation gradients with increasing tensile strain is displayed in the Figs. 7(b) and 7(c) for $\alpha'$ and Figs. 8(b) and 8(c) for the $\gamma$ matrix. As can be seen in Figs. 7(b) and 7(c), the total misorientation across the $\alpha'$ plate in the deformed specimen increases towards the midrib and then decreases towards interface B. The maximum misorientation covers $3.5^\circ$ at midrib independent of the tensile strain. Moreover, variation of the misorientation in the $\alpha'$ plate at nominal strain of $\varepsilon=14.5\%$ becomes small exceeding a distance of about $3\,\mu m$. On the other hand, the orientation gradient approximately perpendicular to the interface B in the $\gamma$ matrix decreases as the nominal strain increases, and no orientation gradient is observed at nominal strain of $\varepsilon=14.5\%$. These observations can be explained in terms of dislocation inheritance into $\alpha'$ plate and the change of accommodation process in the $\gamma$ matrix as it will be explained in more detail in the next section.

3.3. Growth Process of the Butterfly-type $\alpha'$ by Tensile Test

From all microstructural observation, the mechanisms of growth of the butterfly-type $\alpha'$ during tensile deformation can be inferred. Before describing the growth process of the $\alpha'$ plate during tensile test, it is necessary to explain the formation process of the thermal-induced butterfly-type $\alpha'$ observed in the initial microstructure. As already mentioned in the introduction, the present authors suggested the following formation process of the thermal butterfly-type $\alpha'$. When the thermal butterfly-type $\alpha'$ plate is formed in the $\gamma$ matrix at relatively high temperature, the accommodation process is multiple dislocation slip in $\gamma$. Consequently, the formation starts with lath-type character. With proceeding quenching, the transformation temperature decreases and the transformation continues. The accommodation process changes to single or double slip and generates
a strong orientation gradient in the γ matrix. Finally, the accommodation dislocations in γ are inherited into α' with increasing thickness of α' thus generating the observed orientation gradient in the α' plates. In the present study, the orientation gradients in the α' plate and the γ matrix are observed in the initial microstructure as shown in Fig. 3(b).

Figure 9 displays the image quality map and selected pole figures of α' and γ in the specimen before tensile test. Figures 9(b) and 9(c) are (1 1 0) pole figure of γ and (1 1 1) pole figure of α' at area marked by a broken rectangle in Fig. 9(a). From these pole figures, it is seen that the common rotation pole is parallel to a (1 1 0) crystal direction in γ and parallel to a (1 1 1) pole in α'. In addition, these common rotation poles of a (1 1 0) direction in γ and a (1 1 1) direction in α' correspond to the rotation axis of orientation gradient in the γ matrix and the α' plate around α'–γ interface, respectively. This indicates that the accommodation dislocations in γ are inherited into the α' plate. Therefore, these observations are in agreement with our earlier observations on the growth of thermal butterfly-type α'.

The growth process of the α' during tensile test can be also explained in terms of the inheritance process of dislocations and the change of accommodation process. When plastic deformation is imposed onto the specimen, the thermal α' plate starts to grow toward its width direction. Figure 10 displays the image quality map and selected pole figures of α' and γ in the specimen deformed to a nominal strain of ε=14.5%. From Figs. 10(b) and 10(c), it appears that the α' and the γ matrix still have the same common rotation pole as in the initial microstructure. This indicates that piled up dislocations around α'–γ interface in the γ matrix continue to be inherited into the growing α' plate, which creates the orientation gradient inside this phase (Fig. 7(b)). However, the accommodation process of transformation strain in the α' plate and γ matrix may change from single or double dislocation slip to the multiple dislocation slip with the growth of the α' plate by tensile test proceeds. This is because the dislocation slip not only comes from transformation strain, but also from the external stress. The same process with this is proposed for the growth process of the strain-induced α' by Zhang et al.12 Also, the tensile temperature at R.T. may be one of the reasons for the change of the accommodation process because the tensile temperature was a bit higher than water temperature used for quench (15°C). As a result, the orientation gradient in γ decreases with increasing strain (Fig. 8(b)) since the multiple dislocation slip is occurred under several slip systems. Miyamoto et al. have reported about the orientation gradient and the strain distribution in γ matrix around thermally induced lath α'.13 According to their study, orientation gradient in γ matrix around the lath α' is generated by piled-up dislocation due to the multiple slip. However, in case of the present α' plate, the orientation gradient in γ matrix around α'–γ interface becomes small by multiple dislocation slip. Because the present deformed specimen may has a lot of tangled dislocation in a whole γ matrix, it is considered that the orientation gradient in γ matrix around α'–γ interface obtained in the present study is different from that reported by Miyamoto et al. With further increasing tensile strain, all piled-up dislocations around α'–γ interface by dislocation slip on limited slip systems are inherited into α' plate (Fig. 7(c)) while the continuously arising transformation strain is accommodated only by multiple dislocation slip. Finally, the orientation gradient in the γ matrix is exhausted (Fig. 8(c)) and the ORs of the α' plate with the γ matrix change toward K-S OR (Fig. 6). Due to the change of the accommodation process and boundary character, the type of the α' plate changes toward lath-type. It can therefore be concluded that the features of the α' plate are changed during its growth depending on the accommodation process.

In this study, the growth process of the butterfly-type α' is discussed basing on the results of EBSD analysis and the previous literatures. Although EBSD result can show orien-
tation distribution based on dislocation distribution, this result does not give dislocation structure directly. In order to discuss about the growth process of the $\alpha'$ more detail, it is necessary to understand the dislocation structure inside both, the $\alpha'$ plate and the $\gamma$ matrix, by transmission electron microscopy. The investigation of the change of the dislocation structure inside the $\alpha'$ plate and the $\gamma$ matrix during its growth is our future work.

4. Conclusions

The microstructural changes of a thermal butterfly-type $\alpha'$ in Fe–30mass%Ni alloy during a tensile test were investigated by in-situ observations using SEM-EBSD. From the obtained results, the growth process of the butterfly-type $\alpha'$ was discussed. The main results are as follows.

1) When plastic deformation is imposed onto the specimen the $\alpha'$ plates start to grow toward its width direction from the $\alpha'$$-\gamma$ interface with a $\{2\overline{2}5\}_[\gamma]$ habit plane. On the original position of the $\alpha'$$-\gamma$ interface with $\{2\overline{2}5\}_[\gamma]$ habit plane, a microstructure similar to that of a midrib in plate-type martensite is formed with a $\{2\overline{2}5\}_[\gamma]$ habit plane.

2) An orientation gradient is formed in the $\alpha'$ plate as its growth proceeds because the piled-up dislocations around $\alpha'$$-\gamma$ interface are inherited into the growing $\alpha'$ plate. On the other hand, the orientation gradient in the $\gamma$ matrix around $\alpha'$$-\gamma$ interface decreases as tensile strain increases. This is because the accommodation process for transformation strain changes to multiple dislocation slip.

3) As the $\alpha'$ plate grows the OR of the butterfly-type $\alpha'$ with the $\gamma$ matrix changes from G-T to K-S ORs. This arises due to the fact that the thermal butterfly-type $\alpha'$ plate changes toward the lath-type $\alpha'$ because the accommodation process shifts from dislocation slip on limited slip systems to multiple dislocation slip during tensile test at R.T. Therefore, the growth behaviour of the butterfly-type $\alpha'$ during tensile test depends on the accommodation process for transformation strain.

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