Effects of $\alpha+\gamma$ Intercritical and $\gamma$ Single-Phase Annealing on Martensitic Texture Evolution in Dual-Phase Steel Sheets

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Dual-phase (DP) steel sheets composed of soft ferrite and hard martensite phases are typical advanced high strength steel sheets applicable to a variety of automobile parts. The crystallite texture of steel sheets is an important factor which influences the press formability. However, the texture of the martensite itself in DP steel sheets has not been discussed, since the texture was generally measured by the X-ray diffraction method, which does not distinguish the texture of martensite from that of ferrite. The objective of this study is to investigate the texture evolution behavior of each compromising phase; especially the martensite phase, in DP steel sheets by a newly-developed analysis method using Electron Back-Scatter Diffraction (EBSD). The chemical composition of the steel used was 0.088%C-1.23%Si-2.29%Mn-0.093%Ti (mass%). The two sequent annealing was conducted, changing the second annealing temperature, both in the intercritical region and in the $\gamma$ single-phase region. The obtained DP microstructures were controlled to have the same volume fraction of martensite of approximately 40%. The overall texture including martensite after the intercritical annealing was similar to the texture before 2nd-annealing, while the texture after the $\gamma$ single-phase annealing became weak. The new analysis technique using OIM clearly revealed that the discriminate textures from only martensite were close to, but slightly weaker than those of ferrite under the two annealing conditions.

KEY WORDS: crystallographic texture; dual-phase steel; cold-rolled steel sheets; martensite; transformation; intercritical annealing; $\gamma$ single-phase annealing; ODF; EBSD.

1. Introduction

Advanced high strength steel sheets (AHSS) have been extensively applied to various automobile parts during the past few decades in consideration of passenger safety and fuel efficiency. The application of AHSS makes it possible to lighten the weight of the automobile parts by reducing sheet thickness,1) and is expected to accelerate to accommodate increasingly strict automotive fuel efficiency requirements. Among many kinds of developed AHSS, dual-phase (DP) steel sheets having microstructures composed of soft ferrite and hard martensite phases are typical AHSS. DP steel sheets are still attractive due to their superior balance of strength and ductility.2,3) In this connection, the press formability of steel sheets is classified into four categories,4) among which deep drawability is clearly known to depend on the crystallite texture of the steel sheet. Quite a few studies, however, have examined texture control in DP steel sheets with the aim of improving press formability. It has been reported that the texture of DP steel sheets was randomized as the volume fraction of martensite increased, and the hard martensite phase reduced the influence of the texture of the ferrite matrix phase on the planar anisotropy of mechanical properties.5,6) Mondal et al.7) investigated the change in the texture of DP steel sheets after cold rolling and recrystallization annealing in DP steel sheets with a 30–54% volume fraction of martensite before cold rolling. They concluded that the cold-rolled texture was similar to the texture after recrystallization annealing, and the intensity of the $\{111\}$ texture of DP steel sheets decreased with an increasing volume fraction of martensite. However, the texture of the martensite in DP steel sheets has not yet been discussed, since the texture was generally measured by the X-ray diffraction method in prior studies, and it is difficult to distinguish the texture of martensite from that of ferrite by this conventional method.

In this paper, the texture evolution behavior of each component phase, and especially the martensite phase, in DP steel sheets was investigated in Ti-bearing DP steel sheets with the same volume fraction of martensite under different annealing conditions of annealed in the intercritical (two phases, namely, ferrite and austenite) region and annealing in the $\gamma$ single-phase region. The data measured by Electron Back-Scatter Diffraction (EBSD) were analyzed by a newly-developed analysis method to discriminate the crystal texture of the martensite itself.

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2. Experimental Procedure

The chemical composition of the model steel used is shown in Table 1. The C, Si and Mn contents were adjusted so as to obtain substantially the same volume fraction of martensite in both \(\alpha + \gamma\) intercritical annealing and \(\gamma\) single-phase annealing, which are explained below. Ti was added for the purpose of developing the \([111]\) crystallographic texture of ferrite after recrystallization annealing.8)

In specimen preparation, a 30 kg vacuum melted ingot was forged to a thickness of 20 mm. The slabs were reheated to 1 523 K for 3 600 s and hot-rolled to about 4.2 mm in 4 passes with a finishing temperature of about 1 193 K. The hot-rolled sheets were subsequently air-cooled to 873 K and then held at 873 K for 3 600 s to simulate coiling, followed by furnace cooling. After pickling, the sheet thickness was reduced to 3.5 mm by grinding both surfaces to remove scale and surface defects, and the ground sheets were cold rolled to a thickness of 1.0 mm with a 71% cold-rolling reduction. Consecutive two-step annealing cycles were then carried out in an alumina fluidized-bed furnace. The heat cycles are depicted schematically in Fig. 1. The 1st-annealing temperature was 948 K, which was just below the \(\text{Ac}_1\) transformation temperature. This temperature also corresponds to the higher side of the ferrite recrystallization temperature region. Recrystallization was assumed to be completed after 1st-annealing, and the work strain induced by cold rolling was almost completely removed so as to focus on the texture evolution through the \(\alpha-\gamma-\alpha\) transformation in the second process.

In the second heat cycle, the steel sheets were annealed at 1 123 K in the intercritical region and at 1 223 K in the \(\gamma\) single-phase region. When the ferrite single-phase steel at room temperature was heated at 1 123 K in the intercritical region, the austenite phase nucleated and grew at the ferrite grain boundaries above the \(\text{Ac}_1\) temperature (985 K), forming a two-phase macrostructure of ferrite and austenite at 1 123 K. The two-phase macrostructures of this type were retained during gas-cooling to 773 K. The remaining austenite phase at 773 K was subsequently transformed to martensite. On the other hand, when heated at 1 223 K in the \(\gamma\) single-phase region, the steel formed an austenite single phase at 1 223 K since the \(\text{Ac}_1\) temperature was 1 149 K. This austenite single phase changed to a two-phase ferrite and austenite region at 773 K, where the ferrite phase nucleated and grew at the grain boundaries of the austenite phase during gas cooling. The remaining austenite phase at 773 K was subsequently transformed to martensite, precisely as under the intercritical annealing condition. Although the phase transformation processes were different in each case, adjustment to a DP microstructure with the same amount of martensite volume fraction of approximately 40% was possible by controlling the chemical composition, annealing temperature and cooling-stop temperature. The steel sheets were water-quenched at various temperatures during heating and gas cooling in the 2nd-annealing process.

The 2nd-annealed steel sheets were tempered at 573 K for 150 s to improve the reliability of Kikuchi pattern recognition of the martensite phase, which has a high dislocation density. Martensite having a body centered tetragonal (BCT) structure decomposed into ferrite (a body centered cubic (BCC) structure) and carbide (cementite) during the tempering treatment. In this heat treatment, lattice strain is relaxed and the dislocation density decreases due to some form of recovery. The tempering temperature was set to 573 K to minimize the crystal orientation change during the temper-

![Fig. 1. Schematic diagram of heat treatment.](image-url)
ing process. As a result, the reliability of the crystal orientation information of martensite improved, and it was possible to facilitate pattern acquisition from the martensite phases without changing the crystal orientations by the SEM/EBSD (Electron Backscatter Diffraction pattern) method.

The microstructures on the ND (normal direction) × RD (rolling direction) section were observed with a scanning electron microscope equipped with a field emission gun (FE-SEM, S-4100, Hitachi) after etching with 1% nital. The mechanical properties of the steel sheets after both 1st-annealing and 2nd-annealing were investigated by Vickers hardness testing under a load of 9.8 N. The average value of the five tested data was taken as the measured value of the respective specimens.

The crystal textures of the steel sheets were measured by the X-ray diffraction and EBSD methods. The change in the overall average texture was evaluated by the inverse intensity ratio of the {222} and {200} planes parallel to the ND plane using the X-ray diffraction method (Schulz’s reflection method). The three-dimensional crystal orientation distribution function (ODF) in the Bunge notation was calculated from the three imperfect pole figures ( {110} plane, {200} plane, and {211} plane). The steel sheet surface corresponding to the quarter-thickness position was used for the X-ray texture measurements. The crystal orientations of both the ferrite and martensite phases in the DP steel sheets were analyzed by the FE-SEM/EBSD method. The surface preparation of the steel sheets was performed by wet-polishing and buff-polishing with a colloidal silica, followed finally by chemical etching with 0.1% nital. As a result, the irregularities on the sample surface was minimized and the affected layer caused by polishing was completely removed.

The EBSD measurement area and step size were carefully selected to be 50 × 40 μm and 50 nm, respectively, in order to measure the texture of martensite itself, considering the size of martensite and its substructures.

3. Experimental Results

3.1. Relationship between Annealing Temperature and Recrystallization Behavior

In order to investigate only the texture evolution during the phase transformation in 2nd-annealing, it is necessary to complete the recrystallization of the ferrite phases after 1st-annealing and remove the work strain induced by cold rolling. Figure 2 shows the result of an investigation of the effect of the 1st-annealing temperature on the hardness of the steel sheets to determine the ferrite single-phase region and the completion temperature of recrystallization. The cold-rolled sheets were annealed at holding temperatures of 823 K to 973 K for 3 600 s and then water-quenched. The hardness decreased as the 1st-annealing temperature increased; an abrupt decrease was observed at temperatures over 898 K, but hardness then started to increase at 948 K.

Figure 3 shows SEM images of the cross-sectional microstructures of the specimens annealed at 823 K to 973 K for 3 600 s. The rolling direction is parallel to the horizontal direction in the figure. The bright regions with a size of less than 0.5 μm in Figs. 3(a)–3(c) correspond to carbides (cementite). Unrecrystallized ferrite grains elongated in the rolling direction were observed in the microstructure of the sheets annealed at 823 K, as in Fig. 3(a). In the steel sheet annealed at 898 K in Fig. 3(b), recrystallized ferrite grains with a size of approximately several micrometers had formed in the grain boundaries and/or intra-grain area of the unrecrystallized ferrite. The microstructure then changed to recrystallized ferrite with dispersed spherical carbides at 948 K (Fig. 3(c)). With further elevation of the annealing temperature to 973 K (Fig. 3(d)), uniformly dispersed martensite phases having a size of about 0.5 μm to 2 μm appeared, and the carbides disappeared. Therefore, it was deduced that the decrease of hardness in the temperature region from 823 K to 948 K in 1st-annealing was attributable to recrystallization of the ferrite phase, while the increase of hardness at temperatures over 973 K was due to the formation of martensite, which was transformed during cooling from the two-phase region.

In this way, the 1st-annealing temperature of 948 K was carefully selected so as to obtain the ferrite single phase with fully completed recrystallization.
3.2. Effect of Intercritical and \( \gamma \)-Single-phase Annealing on Texture Evolution in DP Steel Sheets

Figure 4 shows SEM images of the cross-sectional microstructures after 1st-annealing and 2nd-annealing. After 1st-annealing, the microstructure consisted of recrystallized ferrite grains with a size of about 10 \( \mu m \) and carbides with a size of 1 \( \mu m \) or less, the latter being shown by the bright contrast in Fig. 3(a). Figures 4(b) and 4(c) show the microstructures of the specimens annealed in the intercritical region and the \( \gamma \)-single-phase region after 1st-annealing, respectively. The ferrite grain size was about 5 \( \mu m \). The area fractions of the martensite shown in Figs. 4(b) and 4(c) were almost the same, being 37% and 35%, respectively, and the martensite phase was mainly observed adjacent to the ferrite grain boundaries. However, the sizes of the martensite were different under the two conditions; that is, the size was larger with annealing in the \( \gamma \)-single-phase region than with annealed in the intercritical region. The Vickers hardnesses of the steel sheets after 1st-annealing and 2nd-annealing are shown in Table 2. The hardness after 2nd-annealing was higher than that after 1st-annealing due to martensite formation, as shown in Fig. 4. The hardnesses were almost the same, at approximately 280 points after 2nd-annealing in the intercritical region and in the \( \gamma \)-single-phase region. Thus, these results confirmed that the martensite fraction and the hardness values of the steel sheets after 2nd-annealing were approximately the same with annealing in the intercritical region and in the \( \gamma \)-single-phase region.

Figure 5 shows the ODFs in the \( \phi_2=45^\circ \) section calculated from the X-ray pole figures for the steel sheets after 1st-annealing and 2nd-annealing, as shown in Fig. 4. The main component of the orientation texture after 1st-annealing was \( \{223\}<110> \). The orientation intensities of the partial \( \alpha \)-fiber (low \( \phi \) angle side) and \( \gamma \)-fiber were quite high. This main component, \( \{223\}<110> \), was still maintained with a high orientation intensity after 2nd-annealing in the intercritical region. On the other hand, in the case of annealing in the \( \gamma \)-single-phase region, the texture was similar to that after 1st-annealing, but it became significantly weaker or randomized after 2nd-annealing.

From the comparison of the texture evolution during 2nd-annealing in the intercritical and the \( \gamma \)-single-phase regions, although the fractions of ferrite and martensite were almost the same, the overall textures including martensite were quite different under the two condition; the initial texture was maintained in the case of intercritical annealing, while the texture in \( \gamma \)-single-phase annealing was randomized after 2nd-annealing. Therefore, it was found that the textures of the DP steel sheets depended on the annealing temperature in 2nd-annealing.

In order to investigate the origin of the difference in the texture of the DP steel sheets, the change of the texture during annealing in the intercritical region and in the \( \gamma \)-single-phase region was studied. Figure 6 shows the change in the volume fraction of ferrite and martensite in specimens which were water-quenched in the middle of the process in 2nd-annealing. In the intercritical annealing in Fig. 6(a), the volume fraction of ferrite decreased to 53% as the 2nd-annealing temperature increased during heating, while that of martensite increased to 47%. In the subsequent cooling process, the volume fraction of ferrite increased as the 2nd-annealing temperature decreased, and that of martensite decreased.

![Fig. 4. SEM images of the cross-sectional microstructures: (a) 1st-annealing at 948 K for 3 600 s, 2nd-annealing at (b) 1 123 K and (c) 1 223 K for 60 s.](image)

![Fig. 5. ODFs in the \( \phi_2=45^\circ \) section calculated from the X-ray pole figures for the steel sheets 1st-annealed at (a) 948 K for 3 600 s, and 2nd-annealed at (b) 1 123 K and (c) 1 223 K for 60 s.](image)

Table 2. Vickers hardness (HV) of the steel sheets after 1st-annealing at 948 K for 3 600 s, and after 2nd-annealing both in intercritical region at 1 123 K for 60 s and in \( \gamma \) single-phase region at 1 223 K for 60 s.

| After 1st-annealing | After 2nd-annealing |
|---------------------|---------------------|
| \( \alpha+\gamma \) Phase | \( \gamma \) Phase |
| 255                 | 274                 |

\( \alpha \): Ferrite, \( \gamma \): Austenite
decreased to 37% at the cooling stop temperature of 773 K. On the other hand, in the $\gamma$ single-phase annealing in Fig. 6(b), the volume fraction of ferrite decreased to 53% as the 2nd-annealing temperature increased, and the ferrite phase disappeared at 1 223 K. In the cooling process, the ferrite phase was again generated or nucleated, and the volume fraction of ferrite increased as the 2nd-annealing temperature decreased; in this case, the volume fraction of martensite decreased and reached 35% at 773 K.

Figure 7 shows the change in the X-ray integrated intensities of $<111>/ND$ and $<100>/ND$ with the annealing temperature during 2nd-annealing in the intercritical region and the $\gamma$ single-phase region. In the intercritical annealing shown in Fig. 7(a), both the $\{111\}$ and $\{100\}$ orientations decreased with increasing 2nd-annealing temperature during heating, while the $\{111\}$ orientation in particular increased with a decreasing 2nd-annealing temperature during cooling. In the $\gamma$ single-phase annealing in Fig. 7(b), however, both the $\{111\}$ and $\{100\}$ orientations decreased with increasing 2nd-annealing temperature during heating, and in particular, the $\{111\}$ orientation decreased remarkably. In the subsequent cooling process, the $\{111\}$ orientation slightly increased at 1 073 K or less with decreasing 2nd-annealing temperature, but the $\{100\}$ orientation did not change greatly. From these results, it was found that the orientation intensity which had decreased during heating then increased and returned to the initial orientation intensity before 2nd-annealing in intercritical annealing, whereas in $\gamma$ single-phase annealing, the texture was randomized during heating and did not return to the initial orientation intensity.

It was considered that the decrease in the densities of the $\{111\}$ and $\{100\}$ orientations during heating shown in Fig. 7 resulted from the decrease in the volume fraction of ferrite$^7$ because the volume fraction of austenite during heating, that is, the volume fraction of martensite produced during water quenching, increased as the annealing temperature increased, as shown in Fig. 6. Next, a SEM/EBSD analysis was performed in order to understand the change of the texture during cooling in the intercritical region and in the $\gamma$ single-phase region, as shown in Figs. 5 and 7.

Figure 8 shows the crystal orientation relationship between ferrite and martensite in the DP steel sheets annealed in the intercritical region at 1 123 K for 60 s and water-quenched at 973 K. In the Image Quality (IQ) map shown in Fig. 8(a), the bright contrast corresponds to ferrite and the dark contrast to martensite. In Fig. 8(b), which is a magnified image of the area surrounded by the white line in Fig. 8(a), the ferrite in the center is represented by blue and the martensite blocks by other colors. The crystal orientation relationship between the martensitic blocks and the adjacent ferrite grains was studied. Some martensitic blocks had an orientation relationship of a rotation angle of 10.5° or 11.9° with a rotation axis of [0.00, −0.71, −0.71] with the ferrite grain shown in blue at the center; other martensitic blocks had a relationship of another axis/angle pair, [0.66, −0.66, −0.36] / 23.2° or 23.8°. In Fig. 8(c), the crystal orientations of the martensitic blocks and the colored adjacent ferrite in Fig. 8(b) are represented on the {001} and {011} pole figures. Because the crystal orientation relationship of the martensitic blocks and the adjacent ferrite had a common {011} pole, having a so-called close-packed plane parallel relationship,$^{13}$ it was assumed that they had a common Bain axis (B1).$^{13}$ The deviation of the common Bain axis between
the martensitic blocks and the adjacent ferrite was small, being less than 3°. **Figure 9** shows the crystal orientation relationship between ferrite and martensite in γ single-phase annealing. The samples were annealed at 1 223 K for 60 s and water-quenched at 973 K. There were several crystal orientation relationships between the martensitic blocks and the adjacent ferrite shown in blue; the axis was \([0.00, -0.71, -0.71]\), and the angles were 14.2°, \([0.93, 0.35, 0.07]\), 20.6° and \([0.66, -0.66, -0.36]\), 24.7°. Therefore, it was confirmed that, in γ single-phase annealing, there was an orientation relationship having the same Bain axis (B1) as in intercritical annealing, although the deviation of the common rotation axis between the martensitic blocks and the adjacent ferrite was 4° to 8°, which was larger than that in intercritical annealing.

Hence, when austenite was transformed to martensite from the two-phase region, it was found that the martensite and the adjacent ferrite had a specific orientation relationship, having the same Bain axis (B1), regardless of the annealing temperature. However, it was still unclear why the deviation of the common rotation axis between the martensitic blocks and the adjacent ferrite was changed by the 2nd-annealing temperature. Although it is conceivable that the martensitic transformation temperature affected the change in the deviation angle, a detailed investigation will be necessary in the future.
4. Discussion

4.1. Relationship of Textures of Ferrite and Martensite in DP Steel Sheets

Generally, in texture measurements of DP steel sheets by the X-ray diffraction method, it is difficult to distinguish the texture of martensite itself from that of ferrite because it is impossible to obtain accurate positional information corresponding to the microstructure in DP steel sheets by the conventional method. Nevertheless, it is necessary to determine the crystallite texture of martensite itself in order to understand the texture evolution in DP steel sheets. Hence, in this research, the texture of martensite itself was quantitatively analyzed with high accuracy and in an extensive range by separating the crystal orientation of each phase in the DP steel sheets by the FE-SEM/EBSD method.\(^\text{14}\) This method makes it possible to acquire specific orientation information at each position in the microstructure. The details of this analysis method are as follows. Figure 10(a) shows the IQ map of the DP steel sheet annealed in the intercritical region at 1 123 K. The bright and dark regions indicate ferrite and martensite, respectively. The martensite BCT structure can in principle be distinguished from the ferrite BCC structure by the SEM/EBSD method. However, since the DP steel sheet is subjected to tempering treatment at 573 K, the martensite of the DP steel sheet in this analysis was considered to be tempered martensite composed of ferrite and carbide. For this reason, discrimination between ferrite and tempered martensite by the difference in crystal structure was thought to be difficult, and it was necessary to separate the crystal orientation of each phase in order to analyze the crystal orientation of the respective phases independently. Using analysis software (TSL OIM Analysis 6), in this method, first, the region of the martensite and the adjacent ferrite grains which had a similar orientation with a misorientation angle within 15° was selected by the highlighting function (Figs. 11(b) and 11(c); the black regions in the figure indicate unselected ferrite). Second, the measured points of martensite were extracted by selecting the low IQ value indicated by the arrow in Fig. 11(d) using the IQ histogram (Figs. 11(e) and 11(f)). This new method was able to evaluate the texture information of each phase independently. Figures 10(c) and 10(d) show the IPF (Inverse Pole Figure) in which the crystal orientations of ferrite and martensite in the DP steel sheet after 2nd-annealing were separated by using this method. Since the IPF map obtained from martensite itself did not contain patterns with low crystallinity, such as the grain boundaries of ferrite, it was considered that only martensite was accurately extracted.

Figure 12 shows the separated ODFs of the ferrite and martensite phases in the DP steel sheet after 2nd-annealing by using this new analysis method. Regardless of the 2nd-annealing temperature, the sharpness of the martensite texture is lower than that of ferrite, while the tendency of the texture including the main orientation of ferrite and martensite was consistent. Therefore, there was considered to be a strong correlation between the textures of the two phases, i.e., ferrite and martensite. The texture of ferrite had a higher texture accumulation in the partial alpha fiber from \{001\} <110> to {112} <110> after intercritical annealing compared with \(\gamma\) single-phase annealing; the texture of ferrite clearly depended on the 2nd-annealing temperature. From these results, the differences between the textures of martensite in DP steel sheets annealed in the intercritical and \(\gamma\) single-phase regions were closely related to the change in the texture of ferrite depending on the difference in the 2nd-annealing temperature.

Thus, this study clarified the fact that the texture of martensite in DP steel sheets was similar to the texture of ferrite regardless of the annealing temperature. As discussed in the previous chapter, this was because the martensite
and the adjacent ferrite in DP steel sheets has a specific orientation relationship, having the same Bain axis (B1). Moreover, the fact that the textures in DP steel sheets annealed in the intercritical and γ single-phase regions were different was also clarified. The recrystallization texture of 1st-annealing was maintained after intercritical annealing, whereas the orientation intensity decreased drastically after γ single-phase annealing, in which the entire structure was completely transformed to austenite. The reason for the texture difference depending on the 2nd-annealing conditions will be discussed in the next section.

4.2. Formation Mechanism of Transformation Texture during Intercritical Annealing and γ Single-phase Annealing

In the previous section, it was noted that the texture after 2nd-annealing in intercritical annealing was similar to that before annealing, whereas in γ single-phase annealing, the texture became significantly weak and randomized after 2nd-annealing. It was deduced that the reason for the dif-
Fig. 13. Schematic diagram of the mechanism of transformation texture formation during 2nd-annealing in the intercritical region and in the γ single-phase region.

The austenite which was transformed at the triple junctions of ferrite during heating was thought to have an orientation relationship close to the Kurdjumov-Sachs (K-S) relationship15 at one or more interfaces with the neighboring ferrite.16 In intercritical annealing, since the initial ferrite grains before annealing remained at the holding temperature, the volume fraction of each phase gradually changed during the annealing process while maintaining the K-S relationship at the interface between the initial ferrite and austenite. The initial ferrite texture could develop because a nucleus having a similar orientation with the initial ferrite grew during gas cooling. In subsequent water quenching, the remaining austenite was transformed to martensite having the variant with the same Bain axis (B1) as the initial ferrite. Therefore, as shown in Fig. 5(b), the texture of DP steel sheets annealed in intercritical annealing was considered to be similar to the texture before annealing. On the other hand, in γ single-phase annealing, a fresh ferrite phase nucleated and grew from the grain boundaries of austenite during gas cooling. It has been reported that the texture formed before γ single-phase annealing was randomized.
The orientation of the initial recrystallized ferrite could not be inherited because a single phase of austenite had formed, and ferrite newly nucleated from this austenite single phase after the austenite grains had grown to some extent during γ single-phase annealing. In subsequent water quenching, austenite was transformed to martensite so as to have the same Bain axis (B1) as the adjacent ferrite. The orientation intensity of the DP steel sheet after γ single-phase annealing was lower than that before annealing because the ferrite which newly nucleated from the austenite single phase tended to have a different orientation from the initial recrystallized ferrite before γ single-phase annealing.

5. Conclusion

The texture evolution of Ti-bearing DP steel sheets annealed in the intercritical and γ single-phase regions was analyzed by the X-ray diffraction and EBSD methods, and the following points were clarified.

1) In the case of intercritical annealing, the annealed texture was similar to that before annealing despite a martensite fraction of approximately 40%. In contrast, in the case of γ single-phase annealing, the annealed texture became significantly weak or randomized.

2) A new quantitative analysis method was proposed for evaluation of the texture of DP steels, and especially martensite itself, by using FE-SEM/EBSD. This method clearly revealed that the texture of martensite itself in DP steel sheets was similar to that of ferrite.

3) In intercritical annealing, where the initial recrystallized ferrite remained, the volume fraction of each phase changed while maintaining the orientation relationship between the initial ferrite and the neighboring austenite. In subsequent water quenching, the remaining austenite was transformed to martensite having variants with the same Bain axis (B1) as the initial ferrite. Therefore, the texture of DP steel sheets annealed in intercritical annealing is similar to the texture before annealing.

4) In γ single-phase annealing, where ferrite newly nucleates from an austenite single phase and grows during gas cooling, the crystal orientation of the ferrite could not be inherited from the initial ferrite before annealing. In subsequent water quenching, the austenite was transformed to martensite so as to have the same Bain axis (B1) as the adjacent ferrite. Therefore, the orientation density of the DP steel sheet after γ single-phase annealing was lower than that before annealing.

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