$\alpha + \gamma$ and γ Phases Annealing in Ultra Low-carbon Sheet Steels

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$\alpha + \gamma$ and γ phases annealing in Ti and Nb-added ultra low-carbon sheet steels with and without Mn, P, and/or Cr has been investigated.

In high-strength steels with Mn, P and/or Cr, BH increases as the annealing temperature increases, whereas YP-EI after aging at 100°C for 1 hr clearly decreases. It is speculated that the coexistence of BH and nonaging properties stems from the high dislocation density introduced by $\gamma \rightarrow \alpha$ transformation. In contrast, YP-EI in mild steel without Mn, P, and/or Cr increases as BH increases because dislocation density is not high enough to provide non-aging property.

In mild steel, r -value distinctly decreases and texture is randomized by γ phase annealing, while r -value increases and $\{111\}$ component develops by intercritical annealing. On the contrary, in high-strength steel, recrystallization texture in α is assumed to be inherited even after $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation takes place. Apparently, variant is distinctly selected when at least γ transforms into α in high-strength sheet steels. The variant selection in $\gamma \rightarrow \alpha$ transformation is speculated to be caused by residual stress introduced by $\alpha \rightarrow \gamma$ transformation.

KEY WORDS: ultra low-carbon steel; cold rolled steel sheet; dual-phase annealing; austenite phase annealing; bake-hardenability; transformation texture; aging property; manganese addition.

1. Introduction

It is important to reduce the thickness of sheet steels used in automobiles because global environmental problems have attracted interest in recent years. To reduce the thickness, the sheet steels must first be strengthened. However, the usual methods lead not only to high ultimate strength but also to high yield strength, which causes surface deflection after press forming. The bake-hardenable (BH) sheet steels have a comparative low yield strength before press forming, and the yield strength of them increases 30 to 50 MPa during the paint-baking process after press forming. Therefore, BH steels could have both antisurface deflection and dent-resistant properties. But higher bake-hardenability (BH) causes stretcher strains. On the contrary, it is well known that low-carbon dual-phase (DP) steels consisting of ferrite and martensite¹⁾ have both prominent BH and nonaging characteristics. DP steels, however, have a problem in terms of their deep-drawing property. In contrast to DP steels, it is reported that intercritical annealing in low- or ultra low-carbon steels increases the Lankford-value (r) compared with steels annealed in

α -region.^{3–6)} In this study, emphases were placed on investigating 1) the relationship between BH property and nonaging property and 2) texture formation in ultra low-carbon sheet steels annealed in $\alpha + \gamma$ and γ phases.

2. Experimental Procedures

Vacuum-melted ultra low-C steels were used and their chemical compositions are listed in **Table 1**. NT is Nb and Ti-added ultra low-C steel. In this base steel, Mn and P were added in the specimen MP and Mn, P, and Cr were added in the specimen MPC. The ingots were hot-rolled, heat treated for coiling simulation, cold-rolled, annealed and temper-rolled. The process conditions are listed in **Table 2**, where SRT: slab reheating temperature, FT: finishing temperature, CT: coiling temperature, CR: cold-rolling reduction. To investigate the influence of the cooling rate after annealing, the two heating patterns described in **Fig. 1** were employed. In **Fig. 1**, ST represents the annealing temperature.

A tensile test was performed using the JIS13B specimen. BH was tested by measuring the difference between the flow stress at 2% prestraining and the lower yield

Table 1. Chemical compositions of steels used. (mass%)

	C	Si	Mn	P	S	Cr	Al	Ti	Nb	N
NT	0.0021	0.04	0.10	0.007	0.0002	—	0.040	0.009	0.010	0.0013
MP	0.0022	0.03	1.54	0.062	0.0062	—	0.040	0.011	0.008	0.0013
MPC	0.0032	0.03	1.54	0.063	0.0061	1.06	0.034	0.011	0.008	0.0014

Table 2. Experimental procedures

Steel	SRT	FT	Thickness	CT	C.R.	Annealing	Temper rolling
NT	1100°C, 1 hr	922°C	4 mm	730°C, 1 hr	80 %	800–960°C	0.5 %
MP	1100°C, 1 hr	903°C	4 mm	730°C, 1 hr	80 %	830–930°C	0.5 %
MPC	1100°C, 1 hr	928°C	4 mm	730°C, 1 hr	80 %	830–930°C	0.5 %

Fig. 1. Schematic illustrations of annealing process.

- a) Pattern-1 is process for rapidly cooling from soaking temperature,
 b) Pattern-2 is process for slowly cooling from soaking temperature to 600°C.

 Fig. 2. Influence of heating temperature and time on volume fraction of γ in steel NT.

stress after subsequent heating at 170°C for 20 min. YP-EI after artificial aging, i.e. heating at 100°C for 1 hr, was measured to examine the aging characteristics. The cold-rolled specimens mentioned above were subjected to dilatation measurement to investigate the transformation behavior of each steel. Electron microscopic observation was carried out for some specimens. The dislocation densities of the steels were compared by observing the grains whose normal direction (ND) of sheet plane was commonly parallel to $\langle 111 \rangle$ and whose thickness was considered to be almost constant. The textures of annealed sheets were determined by the conventional X-ray diffraction method to the quarter-plane of the specimens.

3. Experimental Results

3.1. $\alpha \rightarrow \gamma$ Transformation Behavior

The relationships between holding time and volume fraction of γ transformed when the specimens were heated at the rate of 10°C/s and held at the various temperatures were shown in Figs. 2 to 4. The volume fraction

 Fig. 3. Influence of heating temperature and time on volume fraction of γ in steel MP.

 Fig. 4. Influence of heating temperature and time on volume fraction of γ in steel MPC.

when time equals 0 does not always equal 0 because transformation already started during heating.

It is evident from Fig. 2 that steel NT has a narrow temperature range of $\alpha + \gamma$ and its transformation finishes in a brief period. It is considered that the narrow $\alpha + \gamma$ two-phase region in such Ti, Nb-added ultra low-C mild steels is caused by the solute C attributable to the redissolution of TiC and NbC. Steels MP and MPC that contain Mn, P, and Cr have lower $\alpha \rightarrow \gamma$ transformation temperatures and comparatively wide two-phase regions as represented in Figs. 3 and 4. Furthermore, it is obvious that in these specimens it takes a relatively long period to complete the $\alpha \rightarrow \gamma$ transformation. For instance, transformation does not complete yet after heating at 860°C for 120 sec in steel MP.

3.2. Optical Microstructure

The optical microstructures of steels NT and MPC when the specimens were heated by pattern-1 (Fig. 1) at various temperatures are shown in Fig. 5. In steel NT, the grains are clearly coarsened by $\alpha + \gamma$ or γ phase annealing. In contrast, bainitic ferrite is obtained by $\alpha + \gamma$ or γ phase annealing in steel MPC. Almost the same

microstructures were obtained when the specimens were heated by pattern-2 (Fig. 1).

3.3. Relationship between BH Property and YP-EI after Artificial Aging Treatment

The influence of annealing temperature on BH and YP-EI after artificial aging is represented in Figs. 6 to

Fig. 5. Changes in optical microstructure with annealing temperature in steel NT and MPC (annealed by pattern-1). NT annealed in a) α , b) $\alpha + \gamma$, c) γ -region, MPC annealed in d) α , e) $\alpha + \gamma$, f) γ -region.

Fig. 6. Influence of annealing temperature on BH and YP-EI in steel NT. Number on marks are volume fraction of γ during annealing. a) Annealed by pattern-1 in Fig. 1, b) Annealed by pattern-2 in Fig. 1.

Fig. 7. Influence of annealing temperature on BH and YP-EI in steel MP. Numbers on marks are volume fraction of γ during annealing. a) Annealed by pattern-1 in Fig. 1, b) Annealed by pattern-2 in Fig. 1.

Fig. 8. Influence of annealing temperature on BH and YP-EI in steel MPC. Numbers on marks are volume fraction of γ during annealing.
a) Annealed by pattern-1 in Fig. 1,
b) Annealed by pattern-2 in Fig. 1.

Fig. 9. Influence of annealing temperature on r -value (annealed by pattern-1). Numbers on marks are volume fraction of γ during annealing.

8. The numbers on each symbol indicate the volume fraction of γ transformed during annealing as shown in Figs. 2 to 4.

A prominent BH can be obtained by increasing the annealing temperature in steel NT as shown in Fig. 6. YP-EI also increases with the increase in BH. There are the tendencies that BH and YP-EI are slightly decreased when heating pattern-2 (Fig. 1) is employed.

In steels MP and MPC, BH increases as the annealing temperature increases as represented in Figs. 7 and 8, respectively. On the contrary, YP-EI clearly decreases as the annealing temperature increases, that is, increases in volume fraction of γ . For instance, in the case of heating by pattern-1, no YP-EI was observed when the volume fraction of γ became more than 80%, though BH increases to approximately 50 MPa.

Although the heating pattern does not significantly influence BH, it affects YP-EI. YP-EI tends to be larger in the case of heat-treatment pattern-2 compared with pattern-1. Although steel MP annealed at 880°C by pattern-1 has 50 MPa of BH and 0% of YP-EI, steel MP heated by pattern-2 exhibits 0.12% of YP-EI and 25 MPa of BH. In any event, a prominent BH and nonaging property are obtained by $\alpha + \gamma$ two-phase or γ phase annealing independently of the heating pattern in

this experiment.

3.4. r -value and Texture of Annealed Sheet

The influence of annealing temperature on the r -value in each steel is shown in Fig. 9. It is evident from Fig. 9 that although the r -value increases to approximately 2.8 as the temperature increases to 900°C in steel NT, the r -value is abruptly reduced to approximately 1.2 by γ phase annealing. In contrast to steel NT, though the r -value slightly decreases as the volume fraction of γ increases in steels MP and MPC, the degradation is not so large even when annealed in γ region. Furthermore, $\{100\}$ pole figures of the annealed sheets of NT and MPC at various temperatures are shown in Fig. 10. In steel NT, the texture is randomized by γ phase annealing, while ND// $\langle 111 \rangle$ is developed by the increasing temperature, especially by $\alpha + \gamma$ two-phase annealing. In contrast to steel NT, in steel MPC, even if the specimen is annealed in γ phase region, the texture is extremely similar to that of the sheet annealed at α phase temperature, in other words, the recrystallization texture is inherited even after $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation. These results support the variation of r -value with annealing temperature.

The changes in r -value and texture when the specimens were annealed by pattern-2 in Fig. 1 are shown in Figs. 11 and 12, respectively. When the specimens cooled slowly, the tendencies obtained are somewhat the same as the results from when the steels cooled rapidly. Namely, r -value in steel NT is abruptly reduced by annealing in γ -region whereas r -value in steels MP and MPC decreases as the volume fraction of γ increases, but the degradation is not so large even when annealed in γ region. Nevertheless, the degradation of r -value when the specimens are cooled slowly is smaller than when the specimens are cooled rapidly. The texture formed after annealing by pattern-2 is almost same as that formed by pattern-1. Moreover, almost the same textures were obtained in steel MP.

4. Discussions

4.1. Coexistence between BH and Nonaging Properties

The reasons that BH increased as the annealing temperature increased are thought to be the result of C

Fig. 10. Changes in $\{100\}$ pole figures of steels NT and MPC with annealing temperature (annealed by pattern-1).

Fig. 11. Influence of annealing temperature on r -value (annealed by pattern-2). Numbers on marks are volume fraction of γ during annealing.

in solid solution generated by redissolution of TiC and NbC. It is considered to be reasonable that TiC and NbC redissolve at high annealing temperature, because BH is quite small when steel is annealed at 800 or 830°C. It is thought that $\alpha + \gamma$ or γ phase annealing accelerated the redissolution of TiC and NbC because the solubilities of TiC and NbC in γ are much larger than those in α . Electron micrographs of steel MP annealed at α and γ regions followed by rapid cooling (pattern-1 in Fig. 1) are shown in Figs. 13 and 14, respectively. The specimen with a chemical composition of 0.5% Cr added to steel MP was employed to observe the microstructure annealed at $\alpha + \gamma$ dual-phase temperature because it was

relatively easy to distinguish the transformed grains from the untransformed ones. The typical polygonal ferrite structure without dislocations is observed in steel MP annealed at α region. In contrast, numerous dislocations are observed in steel MP annealed in γ region. In steel MP annealed intercritically, dislocations in transformed grains such as B in Fig. 15 are observed equivalent to those in steel MP annealed at fully γ temperature (Fig. 14). Dislocations in untransformed grains, such as A in Fig. 15, are observed less than that of transformed ones, especially in sites adjacent to transformed grains.

Electron micrographs of steel NT annealed in γ region is shown in Fig. 16. Dislocations are observed in grains. However, it is clear that dislocation density is lower in steel NT than in steel MP even if annealed in γ region.

The difference of dislocation density between in steel NT and MP is explained by Fig. 17, which indicates results of dilatation measurement when phase transformation takes place. Because Ar temperature in steels with Mn and Cr is quite low in comparison with that in mild steels, volume expansion caused by $\gamma \rightarrow \alpha$ transformation makes dislocation density higher. Moreover, the recovery rate of dislocation is assumed to be slower because of lower Ar temperature.

Consequently, it is speculated that the coexistence of BH and nonaging properties in steels MP and MPC stems from the high dislocation density introduced by $\gamma \rightarrow \alpha$ transformation. The increase of immobile dislocation density after paint-baking process brings about high BH because C in solid solution adheres to a part of the dislocations by BH treatment. Nevertheless, it is assumed that residual mobile dislocations still exist after artificial

Fig. 12. Changes in $\{100\}$ pole figures of steels NT and MPC with annealing temperature (annealed by pattern-2).

Fig. 13. Transmission electron micrograph of steel MP annealed by pattern-1 at 830°C (α) for 60 sec.

aging at 100°C for 1 hr because the density is considered sufficiently high, which suppresses recovery of YP-El.

Furthermore, it was observed that YP-El becomes larger when steels were cooled slowly after soaking (pattern-2 in Fig. 2). The assumed reasons are as follows: 1) Ar temperature rose, and 2) the recovery rate of dislocations accelerated.

In contrast to steels MP and MPC, in steel NT, it is assumed that dislocations are adhered by C because dislocation density is not high enough to provide non-aging property.

4.2. Mechanism of Texture Inheritance after $\alpha \rightarrow \gamma \rightarrow \alpha$ Transformation

4.2.1. Previous Investigations on Transformation Texture in Low- and Ultra Low-carbon Steels

Although there are detailed reviews^{7,8)} on the transformation texture of steels, the transformation texture in low- and ultra low-carbon cold-rolled sheet steels is scarcely mentioned. Therefore, previous investigations on the transformation texture formed by annealing in these sheet steels are discussed here.

Horta *et al.*³⁾ suggested that ND// $\langle 111 \rangle$ texture is developed by slow cooling after intercritical annealing. They also reported that the optimum volume fraction of γ for the development of $\{111\}$ texture was approximately 60%. Untransformed α presumably contributes to the increase in ND// $\langle 111 \rangle$ texture.

Hashimoto *et al.*⁶⁾ similarly reported that $\{111\}$ texture is developed by $\alpha + \gamma$ intercritical annealing. The mechanism is explained as follows. Grains of $\{100\}$ and $\{110\}$ preferentially transform to γ during $\alpha \rightarrow \alpha + \gamma$

Fig. 14. Transmission electron micrograph of steel MP annealed by pattern-1 at 930°C (γ) for 60 sec.

Fig. 15. Transmission electron micrograph of 1.5%Mn-0.06%P-0.5%Cr steel annealed by pattern-1 at 860°C ($\alpha + \gamma$) for 60 sec.

Fig. 16. Transmission electron micrograph of steel NT annealed by pattern-1 at 940°C (γ) for 60 sec.

Fig. 17. Relationship between dilatation and temperature of steels NT and MP.

transformation, whereas grains of $\{111\}$ do not transform. Residual ferrite grains of $\{111\}$ grow into transformed grains during $\alpha + \gamma \rightarrow \gamma$ transformation. Hashimoto *et al.*^{4,5)} also discussed the mechanism of the formation of $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation texture with respect to the variant section.

4.2.2. New Mechanism of Texture Inheritance after $\alpha \rightarrow \gamma \rightarrow \alpha$ Transformation

Textures obtained in steels NT, MP, and MPC when annealing was conducted in $\alpha + \gamma$ and γ phase are summarized as follows. Here α before and after $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation is designated as "initial α " and "final α ," respectively.

- (1) Initial α in steels NT, MP, and MPC has a strong $\{111\}$ texture.
- (2) $\{111\}$ texture is sharpened by intercritical annealing in steel NT.
- (3) The texture in steel NT is randomized by annealing in γ phase.
- (4) Intercritical annealing does not change the texture in steels MP and MPC, which is similar to initial α .
- (5) The texture in steels MP and MPC is maintained even when annealing is performed in γ region.

Concerning (1) above, an explanation is omitted because the texture is a typical recrystallization one. The interpretation for (2) and (4) may be possible using the model proposed by Hashimoto.⁴⁻⁶⁾ Therefore, (3) and (5) mentioned above are discussed here.

In steels MP and MPC, final α could memorize the texture of initial α despite the $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation, whereas the texture is randomized by the transformation in steel NT. The assumed reasons are as followings:

- (a) The variant in transformation was selected based on the model of Hashimoto *et al.*
- (b) Since austenitization was not completed in steels MP and MPC, the residual α grew in $\gamma \rightarrow \alpha$ transformation.
- (c) The difference in grain growth rates in γ region among steels MP, MPC, and steel NT brought about the

difference in final texture.

(d) γ preferentially transformed into α from precipitates that memorized the texture of initial α .

(e) α preferentially transformed into γ from segregation bands of Mn in MP and MPC. The transformed γ and retransformed α memorize the information of initial α .

(f) Residual stress generated by $\alpha \rightarrow \gamma$ transformation contributes to the variant selection in $\gamma \rightarrow \alpha$ transformation.

(g) Mn and Cr led the shear transformation that affects the variant selection.

Hereafter the probability of each item is discussed.

First, (a) is considered impossible. Although Hashimoto reported that 1) variant cannot be selected when heating and cooling rates are fast, and 2) the texture when variant selection occurs becomes strong in $\{110\}$ component or the orientation shifts 20–30° from $\{100\}$ parallel to ND, 1) the textures of steels MP and MPC were not randomized even after faster heating and cooling rates were employed in annealing, and 2) the texture obtained was $\{111\}$ parallel to ND. The probability of (b) may also be quite small because Ac_3 temperatures in MP and MPC are below 900°C, while the annealing was conducted at 930°C which is high enough for full austenitization. (c) does not explain the difference in transformation texture between steel NT and MP, MPC under the assumption that the same variant among steels NT, MP, and MPC might be selected in $\gamma \rightarrow \alpha$ transformation. The possibility of (d) is small because precipitates in α phase are presumably TiC and FeTiP in steels MP and MPC and would be dissolved in γ . Although MnS also exists in these steels, it is impossible for MnS to memorize the initial texture of α because Mn depleted zone would be formed around MnS, which retard $\alpha \rightarrow \gamma$ transformation. (e) is also not very possible because there seems to be no reason why $\{111\}$ texture is inherited even if the texture of γ strongly developed owing to Mn segregation.

Simulation of the orientation generated by transformations was attempted to examine (f). Furubayashi^{9,10)} positively introduced Bain's relationship for the analysis of transformation texture. The reasons are as follows^{9,10)}:

(1) Among Kurdjumov–Sachs (K–S), Nishiyama (N) and Bain's relationship (B), which are well known as orientation relation of transformation, it is easy to handle B because the number of variants is 3. The number of variants is 24 for K–S and 12 for N.

(2) Furthermore, there does not appear to be substantial differences between K–S, N and B, the differences of orientation are within the range of experimental accuracy of X-ray diffraction. Therefore, B will be employed here as the orientation relationship for transformation. Calculation was made using Furubayashi's¹¹⁾ program modified by one of the authors.¹²⁾

The orientations of transformed γ and final α calculated by Bain's relationship when $\{111\}\langle 211 \rangle$ is introduced as initial α are represented in Fig. 18. As mentioned above, there exist 3 variants in B that bring about 9 orientation relationships after $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation. It is natural that the orientation of final α always includes that of

Fig. 18.

Calculated $\{100\}$ pole figures based on Bain's relationship.

● represents $\{100\}$ poles when the expansion axis in $\alpha \rightarrow \gamma$ transformation agrees with the compression axis in $\gamma \rightarrow \alpha$ transformation.

Fig. 19. Schematic illustrations of Bain's relationship.

X, Y, Z and X', Y', Z' are axes for γ and α , respectively.

initial α . The orientation of final α could correspond with initial α where the expansion axis in $\alpha \rightarrow \gamma$ transformation agrees with the compression axis in $\gamma \rightarrow \alpha$.

Here, the possibility of (f) mentioned above is discussed. Expansion along one axis and compression along the other axes occur. When $\alpha \rightarrow \gamma$ transformation takes place, residual stress is thought to work in opposite direction, as described as σ_x , σ_y , and σ_z in Fig. 19. Assuming the residual stress affects the variant selection in $\gamma \rightarrow \alpha$ transformation, it is considered reasonable that the orientation of final α agrees with that of initial α . The differences in texture formation in mild steel (NT) and in high-strength steels (MP and MPC) likely stem from the differences in strength. Because the higher residual stress might be generated by $\alpha \rightarrow \gamma$ transformation in steels with higher strength, variants could be distinctly selected in steels MP and MPC compared with steel NT. However whether residual stress can exist at such a high temperature as used in this experiment or not is the future theme. However, there is an example of successfully simulating the transformation texture by introducing residual stress at high temperature. That is, one of the authors¹²⁾ determined that hot-rolled sheet steels with a chemical composition of 0.008% C–1.5% Mn–0.071% Ti–0.02% Nb–0.002% N transformed from unrecrystallized γ have a high r -value and that transformation texture can be simulated by introducing residual stress generated by hot-rolling, as proposed by Furubayashi.^{9,10)} Therefore, by analogy it seems reasonable to assume the existence of residual stress at high temperature.

Concerning (g), it has not been revealed whether shear

transformation could occur in Mn or Cr-added ultra low-carbon steels. Nevertheless, if shear mechanism is predominant in those steels, it presumably assists the variant selection. Therefore, it might be important to add the elements that retard the $\gamma \rightarrow \alpha$ transformation.

Moreover, the inherited texture in intercritical annealed sheet steels can be understood based on this mechanism besides the theory proposed by Hashimoto *et al.*⁶⁾

The formation mechanism of transformation texture in ultra low-carbon sheet steels is thought not to be understood completely. Therefore, it is necessary to clarify the mechanism in terms of 1) phase transformation mechanism in IF steels, and 2) *in situ* measurements of texture in γ region or residual stress in the future.

5. Conclusions

Annealing in $\alpha + \gamma$ and γ phase of Ti and Nb-added ultra low-carbon mild sheet steel and Mn, P, and/or Cr-added high-strength sheet steels was investigated. Results obtained in this study are summarized as follows;

(1) Mild steel has a narrow range of $\alpha + \gamma$ temperature and $\alpha \rightarrow \gamma$ transformation finishes in a brief period. On the contrary, high-strength steels, which contain Mn, P, and Cr, have lower $\alpha \rightarrow \gamma$ transformation temperature and a comparatively wide two-phase range (30–40°C). Furthermore, it takes a relatively long period to complete $\alpha \rightarrow \gamma$ transformation.

(2) In mild steel, both BH and YP-El increase as the annealing temperature increases. In high-strength steels, BH increases as the annealing temperature increases, whereas YP-El clearly decreases. It is evident that prominent BH and nonaging properties are present in high-strength steels.

(3) It is speculated that the coexistence of BH and nonaging properties in high-strength steels stems from the redissolution of TiC and NbC, and the high dislocation density introduced by $\gamma \rightarrow \alpha$ transformation. In contrast, YP-El in mild steel increases as BH increases. The difference in dislocation density between mild and high-strength steels is considered to stem from the differences in Ar temperature, which causes the difference in volume expansion and recovery rate.

(4) In mild steel, r -value distinctly decreases and texture is randomized by complete γ phase annealing, while r -value increases and $\{111\}$ component develops by intercritical annealing. On the contrary, in high-

strength steel, recrystallization texture in α is inherited after the transformation of $\alpha \rightarrow \gamma \rightarrow \alpha$.

(5) Apparently, variant is distinctly selected when at least γ transforms into α in high-strength sheet steels. It is speculated that residual stress introduced by $\alpha \rightarrow \gamma$ transformation contributes to the variant selection in $\gamma \rightarrow \alpha$ transformation. The texture of initial α is inherited after $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation because the expansion axis in $\alpha \rightarrow \gamma$ transformation agrees with the compression axis in $\gamma \rightarrow \alpha$ transformation. This is inferred to be controlled by residual stress. Because residual stress generated by $\alpha \rightarrow \gamma$ transformation is considered small in mild steel, the variant cannot be strongly selected. Consequently, the texture is randomized by annealing in γ phase.

The formation of transformation texture in ultra low-carbon steels is considered closely related to the transformation mechanism. It is therefore necessary to systematically investigate the transformation mechanism in ultra low-carbon steels in the future.

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