Effect of Hot Band Microstructural Factors on Recrystallization Texture in Ti-added Ultra-low Carbon Cold-rolled Sheet Steels

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An examination has been made of the effect of hot rolling conditions, considered as thermal history, on the deep drawability and recrystallization texture formation in Ti-added ultra-low carbon cold-rolled sheet steels.

The findings show that the average r -value of IF steel is significantly improved by refining the hot band grains, by decreasing the finish-rolling temperature in large thermal drop, to temperature even below the A_{r3} transformation temperature, during hot rolling process. However, the average r -values of sheets hot-rolled almost isothermally at a finish-rolling temperature below the A_{r3} transformation temperature deviate detrimentally, when compared with hot band of the same grain size. In this study, an attention has paid to the formation of texture throughout the hot rolling, cold rolling and annealing processes, in relation to deep drawability.

KEY WORDS: Ti-added ultra-low carbon cold-rolled sheet steel; recrystallization texture; r -value; hot rolling pattern; hot band grain size; hot band texture; cold rolling texture.

1. Introduction

Recently, interstitial-atom-free (IF) sheet steels, which are ultra-low carbon steels with the strong carbonitride formers such as Ti and/or Nb in quantities in excess of equivalent to the carbon, nitrogen and sulfur content, are extensively applied in such areas as automotive panels because of their excellent deep drawability and non-aging property. Numerous efforts have made to clarify the mechanism of the improved plastic strain ratio, the average r -value, in IF sheet steels.^{1–3)} Several explanations have been proposed for the effect of titanium addition:

- (1) the scavenging effect, that is, removing carbon and nitrogen from the solution by fixing them as stable precipitates,
- (2) the effect of fine dispersion of TiC precipitates, and
- (3) the effect of a substitutional element in solid solution.

Further improvement of the deep drawability of IF sheet steels has since been achieved through the optimization of chemical composition and processing conditions, such as by decreasing the carbon content, reducing the slab reheating temperature, refining the hot band grain size, or increasing the annealing temperature.^{4–6)}

Concerning the effect of thermal history on the hot rolling process, some positive effects of an α -region hot rolling on the deep drawability of IF steels have been reported.^{7–10)} The most of the papers^{7–9)} were concerned with the deep drawability of hot bands, but a few reports^{10–12)} dealt with the influence of an α -region hot rolling on the deep drawability of cold-rolled IF steels, since the texture of hot band has a substantial

influence on the textural development during subsequent cold rolling and annealing. Senuma *et al.*¹²⁾ has indicated that a cold-rolled IF sheet steel which was previously warm-rolled showed a more favorable recrystallization texture on its midplane for deep drawability than conventionally manufactured sheets. Lubrication during warm rolling is required, however, in order to reduce the friction between roll and sheet, and avoid the formation of the surface texture unfavorable for deep drawability.

Other important factors influential on the deep drawability such as the finish-rolling temperature through its effects on the grain size of hot band have also been investigated. It is well-known that the refinement of hot band grain is one of the most important factors in improving the deep drawability of cold-rolled sheet steels, and many studies have consequently been made on the control of hot bands grains.^{13,14)} However, few studies have been made on the effect of thermal history during hot rolling process.¹⁵⁾ In the existing studies, attention has mainly been focused on the effects of finish-rolling temperature and cooling conditions after hot-rolling. Furthermore, the result has not been discussed in detail from the viewpoint of texture formation.

The present study has been undertaken to examine the effect of hot-rolling conditions, taking thermal history account, on the deep drawability and recrystallization texture of Ti-added ultra-low carbon cold-rolled sheet steels.

2. Experimental Procedure

2.1. Specimens

The chemical compositions of steels used in this experiment are shown in **Table 1**. Ti-bearing ultra-low carbon steels with a carbon content of 0.0014–0.0056 %

Table 1. The chemical compositions of steels used in this experiment. (mass%)

Steel	C	Si	Mn	P	S	Al	Ti	N	Ti*/C
A	0.0014	0.01	0.15	0.002	0.001	0.025	0.038	0.0025	4.98
B	0.0056	0.01	0.15	0.002	0.001	0.024	0.024	0.0022	0.67

Ti*/C: Atomic ratio, $Ti^* = Ti - (48/32)S - (48/14)N$

Fig. 1. Thermal history during hot rolling process.

and a titanium content of 0.024–0.038% were vacuum-melted in the laboratory. Steel A is an interstitial-atom-free steel and is the main steel used here. Steel B is non-interstitial-atom-free steel used for comparison. The amounts of effective titanium, Ti^* , the atomic ratio of titanium to carbon, Ti^*/C , are defined by the following equations.

$$Ti^* (\%) = Ti (\%) - [(48/32) \cdot S (\%) + (48/14) \cdot N (\%)]$$

$$Ti^*/C = (Ti^*/48)/(C/12)$$

2.2. Experimental Procedure

The 30 mm thick slabs were reheated at 1200°C for 30 min and were hot-rolled to 3.6 mm in three passes at the finish-rolling temperature of 800 to 950°C, following two different thermal histories during the hot rolling process, as shown in **Fig. 1**. In the first process, the steels were hot-rolled with large thermal decrease (This process is hereinafter referred to as Pattern I.). In the second process, the steels were hot-rolled almost isothermally (This process will hereinafter be referred to as Pattern II.). The amounts of thermal drop during hot rolling were controlled by rolling speed, which was 25 mpm for Pattern I and was 100 mpm for Pattern II.

After hot rolling, the sheets were rapidly cooled by water-spray, and held at 700°C for 1 h, followed by furnace-cooling to room temperature to simulate the thermal history of coiling. The sheets were then cold-rolled by 80% reduction and mainly annealed at 850°C for 90 s in salt-bath followed by air-cooling, simulating continuous annealing. The plastic anisotropy ratio, r -value, was calculated by measuring the widths at three points on tensile specimens taken in the longitudinal (r_L), transverse (r_T), and diagonal (r_D) directions, of which the width was 12.5 mm and the length 120 mm, after straining by 15%. The average r -value was calculated using the following standard formula;

$$r = (r_L + 2r_D + r_T)/4$$

The process of recrystallization in the specimens annealed at each temperature for 90 s was examined by means of Vickers hardness and X-ray measurements.

The pole density measurements were made with 222,

Fig. 2. Relationship between finish-rolling temperature and average r -value in annealed sheet steels A and B hot-rolled by Patterns I and II.

211, 200 and 110 X-ray diffraction lines by Mo-K α radiation. The crystallite orientation distribution function, ODF, analysis was made using Roe's method.¹⁶⁾

3. Results

3.1. Effect of Finish-rolling Temperature

Because it was reported that the average r -value depends on finish-rolling temperature, and the finish-rolling temperature which shows the highest average r -value is nearly coincident with the Ar_3 transformation temperature, the relationship between finish-rolling temperature and average r -value was investigated in annealed sheet steels A and B, hot-rolled by Patterns I and II, as shown in **Fig. 2**. The average r -value of steel A hot-rolled in Pattern II, in which the thermal drop during the hot rolling process is small, is not significantly affected by finish-rolling temperature, since it shows a weak peak at about 900°C. On the other hand, the average r -value of steel A hot-rolled in Pattern I, in which the thermal drop during hot rolling process is large, is dependent on the finish-rolling temperature. The average r -value increases as the finish-rolling temperature decreases, and reaches a peak at approximately 850°C, which is estimated to be just below the Ar_3 transformation temperature. As a result, the average r -value for Pattern I specimen hot-rolled at a finish-rolling temperature below about 880°C is higher than that of Pattern II specimen. In steel B, which is the non-interstitial-atom-free steel, the average r -value deteriorates, and the dependence on finish-rolling temperature is not so remarkable, even with Pattern I hot rolling.

Figure 3 shows the effect of finish-rolling temperature on the pole densities of the (222), (211), (200) and (110) planes parallel to the sheet plane, and the ratio of (222) to (200) pole density, which is considered to be close to the average r -value, at the midplane of annealed sheet steel A hot-rolled by Patterns I and II. In the textural change for Pattern I specimen, there is the tendency to show an increase in the (222) pole density and a decrease

Fig. 3. Effect of finish-rolling temperature on pole densities of (222), (211), (200) and (110) planes at midplane of annealed sheet steel A.

Fig. 4. Effect of finish-rolling temperature on grain sizes of hot bands and annealed sheet steels A and B hot-rolled by Patterns I and II. (Arrows indicate data concerned with microstructure which consist of partly un-recrystallized grains)

in the (200) and (110) pole densities with decrease in the finish-rolling temperature. Then, the former reaches a weak peak and the latter reaches a bottom at approximately 850°C. No apparent discrepancy between Pattern I and Pattern II specimens was observed in the behaviors of texture as a function of finish-rolling temperature above 880°C. With decrease in the finish-rolling temperature below 880°C in Pattern II specimen, the (222) pole density decreases and the (200) pole density increases abruptly. As a result, the trend in ratio of (222) to (200) pole densities corresponds with that of the average r -value shown in Fig. 2.

Figure 4 shows the effect of finish-rolling temperature on the grain sizes of hot bands and annealed sheet steels A and B hot-rolled in Patterns I and II. The grain sizes of hot bands decrease as the finish-rolling temperature decreases, and the minimum hot bands grain sizes occur at about 850°C in specimens of both hot rolling patterns.

Fig. 5. Relationship between average r -value and grain sizes of hot bands of steels A and B hot-rolled by Patterns I and II.

The grain sizes of steel A hot bands are larger than those of steel B hot bands in the finish-rolling temperature range of 800 to 950°C. Furthermore, the minimum grain size of hot band for the Pattern II specimen at about 850°C is larger than that of the Pattern I specimen. The grain size of annealed sheet steel A is coarser than that of steel B. There is no apparent difference in the change in grain size of annealed sheets as a function of either the finish-rolling temperature or the hot rolling patterns.

It is well-known that the grain size prior to cold rolling is an important factor controlling the average r -value, by controlling the preferred orientation in the primary recrystallization texture. Therefore, the average r -value of steels A and B hot-rolled in Patterns I and II have been plotted against the grain sizes of hot bands, as shown in Fig. 5. The increase in average r -value of steel A is closely related to the decrease in the grain size of hot band. It can therefore be said that the peak average r -value is obtained at a finish-rolling temperature of approximately 860°C due to the finer grain size of the hot band. Some of the data concerning Pattern II specimen deviate detrimentally from this relationship, when compared with the same grain size of hot band. The dependency of grain size of hot band on the average r -value in steel B specimens is less pronounced than for steel A specimens.

3.2. Comparison of Hot Rolling Pattern

As mentioned above, in Pattern I hot rolling, in which the thermal drop during hot rolling process is large, the average r -value of steel A is improved by refining the hot band grain. This is caused by the decrease in the finish-rolling temperature to temperatures as low as below the A_{r3} transformation temperature. However, the average r -value deteriorates sharply when the finish-rolling temperature drops below the A_{r3} transformation temperature with Pattern II hot rolling, in which the thermal drop during the hot rolling process is small, although the grain size of hot band is as fine as that of the Pattern I specimen. Therefore, in order to clarify the reasons why the above described behaviors take place, the behavior of recrystallization texture formation was investigated for the following specimens;

(1) The effect of the finish-rolling temperature, that is, hot band grain size: steel A hot-rolled at the finish-rolling temperatures of 940 and 860°C in Pattern I (these specimens will be referred to hereinafter as Patterns Ia and Ib specimens, respectively)

(2) The effect of hot rolling pattern: steel A hot-rolled at a finish-rolling temperature of 860°C in Patterns I and II (these specimens will be referred to hereinafter as Pattern Ib and IIB specimens, respectively)

Figure 6 shows the change in pole densities of (222), (211), (200) and (110) planes parallel to the sheet plane of hot band of steel A as a function of distance from the surface. The textures in all specimens are almost random at the surface, and show general tendencies toward increases in the (222), (211) and (200) pole densities and decrease in the (110) pole density with shifts away from the surface toward the center. No discrepancy between the Pattern Ia and Pattern Ib specimens is apparent in the behaviors of pole density as a function of distance from the surface. However, the texture in Pattern IIB specimen is non-homogeneous through the thickness. The (110) pole density takes the maximum value at a certain depth from the surface, and the (200) and (211) pole densities increase with changing from the surface to the center and show the maximum values at the midplane. This phenomenon shows redundant shear strain which brings about between roll surface and strip.⁸⁾ The (110)

Fig. 6. Change in pole densities of (222), (211), (200) and (110) planes of hot band of steel A as a function of distance from surface.

pole density is higher for a certain depth from surface, a position 1/18 of the hot band thickness below the surface, than for the surface layer. This result corresponds the general observation that the maximum shear strain does not emerge at the surface but at a certain distance below the surface.^{17,18)} **Figure 7** shows that the main orientation of 1/18 thickness layer is dominated by (110)[uvw] while that of the midplane is (100)[01 $\bar{1}$] ~ (211)[01 $\bar{1}$].

Figure 8 shows the changes in hardness and in the ratio of recrystallized grain size to hot band grain size as a function of annealing temperature for 90 s, for Patterns Ia, Ib and IIB specimens. The difference in overall recrystallization behavior that occurs with the changes in hot rolling pattern and finish-rolling temperature can not be discerned within the scope of the present investigation. The recrystallization commences at about 620°C and is complete at approximately 700°C. With Pattern I hot rolling, the size ratio of grains obtained by recrystallization to hot band grains is larger for the Pattern Ib specimen than for the Pattern Ia specimen. Additionally, the increase in size ratio that occurs with elevation of the annealing temperature, that is the degree of grain-growth, is larger for Pattern Ib specimen than for the Pattern Ia specimen. There is no distinct difference between the Pattern Ib and the Pattern IIB specimens in the recrystallized grain size ratio and the degree of grain-growth except for the higher annealing temperature region. This means that the deterioration in the average r -value for the Pattern IIB specimen compared with the Pattern Ib specimen is due to the change in recrystallization texture.

Therefore, the textural changes that occur during isochronal annealing were examined for the specimens of all three groups, Ia, Ib and IIB specimens. **Figure 9** shows the change in pole densities of (222), (211), (200) and (110) planes parallel to the sheet plane at the mid-plane of steel A, as a function of annealing temperature. These pole densities remain almost constant before the commencement of recrystallization. The (222) pole density in Patterns Ia and Ib specimens begins to increase with the start of recrystallization, and continues to increase with the progress of recrystallization and grain-growth, due to the elevation of the annealing temperature. The increment is more remarkable in the Pattern Ib specimen than in the Pattern Ia specimen. On the

Fig. 7. (200) pole figures at a) 1/18 thickness layer and b) midplane of hot band of Pattern IIB specimen.

Fig. 8. Changes in hardness and ratio of recrystallized grain size to hot band grain sizes of Patterns Ia, Ib and II specimens as a function of annealing temperature for 90 s.

Fig. 9. Changes in pole densities of (222), (211), (200) and (110) planes at the midplane of steel A as a function of annealing temperature for 90 s.

other hand, the (222) pole density in the Pattern IIb specimen does not change during the progress of recrystallization, and increases only slightly with the grain-growth, as a result of the elevation of annealing temperature. There are a general tendency towards an inverse correspondence between the (222) pole density and the (211) and (200) pole densities: an increase in (222) pole density is always accompanied by the decreases in (211) and (200) pole densities. The textural behaviors concerning the (211) and (200) pole densities are fundamentally similar among these specimens, except that the (211) pole density remains a high value in the Pattern Ib specimen, even after recrystallization. The (110) pole density is not so much influenced by elevation of the annealing temperature, especially in the Pattern Ib specimen, since there is a tendency to show a slight increase in pole density due to the progress of recrystallization, followed by a monotonic decrease due to grain-growth. The increment in (110) pole density is distinctive in the Pattern IIb specimen. Moreover, the

Fig. 10. Example of orientations distribution at stage of recrystallization of about 50% fraction of steel A determined by means of ECP technique. (d_H indicates grain size of hot bands.)

(110) pole density remains a high value after grain-growth in the Pattern IIb specimen, in comparison with the Pattern Ib specimen. The behavior of (110) pole density in Pattern IIb specimen is assumed to be closely related to that of (222) pole density.

Figure 10 shows the grain orientations distribution at about the 50% fraction stage of recrystallization, determined by means of the Electron Channeling Pattern (ECP) technique at the midplane of steel A annealed at 660°C. At this stage, the recrystallized grains over $10 \mu m$, which is the analyzing limits size in the ECP technique, are orientated in vicinity of (111), (100) and (110). The figure clearly shows that the orientations in the vicinity of (110) is not recrystallized in the Pattern Ib specimen, which has a fine grain in the hot band. In addition, the orientations in the vicinity of (100) is much more recrystallized in the Pattern IIb specimen which has a {100} texture in the hot band in spite of its fine grain size, than in the other specimens.

Figure 11 illustrates the grain orientations distribution at the midplane of steel A annealed at 850°C, determined by the ECP technique. The grain orientations of these specimens are almost the same as those of the approximately 50% recrystallization stage, except that the orientation in the vicinity of (111) in the Pattern Ib specimen is further intensified. The relatively developed (111) orientation after the completion of recrystallization and grain-growth suggest the possibility that the nuclei of the (111) orientation, which were determined later to have the component of (554)[225], grow more rapidly than other nuclei.

Figure 12 shows the $\phi = 45^\circ$ section of the ODF of steel A annealed at 850°C. The recrystallization textures of the Patterns Ia and Ib specimens have an intensified

Fig. 11. Example of orientations distribution of steel A annealed at 850°C determined by means of ECP technique.

Fig. 12. $\phi = 45^\circ$ section of ODF at midplane of steel A annealed at 850°C.

γ -fiber and are characterized through $(554)[22\bar{5}]$ as a main orientation. It is known that this type of texture is formed in IF steel. The $(554)[22\bar{5}]$ component in the Pattern Ib specimen is relatively stronger than that in the Pattern Ia specimen. In contrast, the recrystallization texture of the Pattern IIb specimen mainly consists of $(111)[01\bar{1}]$. Besides having this orientation, this specimen also has the other component developed.

4. Discussion

4.1. Effect of Finish-rolling Temperature in Pattern I

In Pattern I hot rolling, the average r -value of steel A is significantly improved because the hot band grain is refined by the drop in the finish-rolling temperature to

levels even below the A_{r3} transformation temperature. It is well-known that the improvement of formability in IF steel sheets has been achieved by refining the hot band grain size.^{13,14} It has been noted that this is thought to be controlling the preferred orientation in primary recrystallization texture, but the results concerning IF sheet steels were not discussed in detail from the viewpoint of their implications for texture formation. A comparison of the annealed sheet textures in the Patterns Ia and Ib specimens indicates that the $(554)[22\bar{5}]$ component is relatively developed in the Pattern Ib specimen.

In general, recrystallization nuclei are known to be formed in the regions where have local non-homogeneities of deformation, such as grain boundaries. Since it is thought that high energies are stored and the trend is for crystal rotations in these regions to be varied as a consequence of the difference in deformation mode, in comparison with regions where deformation proceeds homogeneously, some extent of development of local non-homogeneity is expected in cold-rolled states and in textural change on annealing.¹⁹ Recrystallization texture is also known to have a certain crystallographic relationship with deformation texture.^{20,21} Accordingly, an experimental study concerning the cold rolling stage was undertaken, to interpret the beneficial development of the $(554)[22\bar{5}]$ recrystallization texture in the Pattern Ib specimen.

Figures 13 and 14 show the $\phi = 45^\circ$ section of the ODF and α - and γ -fiber textures at the midplane of specimens for Patterns Ia and Ib specimens, indicating the change in cold rolling texture at 80% reduction with finish-rolling temperature for each case. The cold rolling textures are similar for the two types of specimens. They both show a strong γ -fiber texture in addition to a strong α -fiber texture which is primarily composed of $(211)[01\bar{1}]$. It should be noted that the crystals rotated through two routes, the $(100)[01\bar{1}] \sim (211)[01\bar{1}]$ route about the RD// $\langle 011 \rangle$ axis, and the $(211)[01\bar{1}] \sim (111)[11\bar{2}]$ skeleton route. The occurrence of crystal rotation through two routes is more remarkable in the Pattern Ib specimen than in the Pattern Ia specimen. This is considered to have resulted from the refinement of the hot band grain. The $(111)[11\bar{2}]$ orientation rotates toward the end stable orientation, $(211)[01\bar{1}]$, which coincides with the orientation predicted by a crystal rotation model,²² by operating multiple-slip if hot band grain is refined. The grain-growth of near $(111)[11\bar{2}]$, possibly $(554)[22\bar{5}]$, is considered to occur mainly in the deformed matrix of $(211)[01\bar{1}]$ by a rotation of approximately 35° around the $\langle 011 \rangle$ fiber axis at -60° from the rolling direction. The phenomenon is more selective in the Pattern Ib specimen than in the Pattern Ia specimen, which is considered to be the result of the high mobility of the interface between recrystallized and unrecrystallized regions.

4.2. Effect of Hot Rolling Pattern in Finish-rolling Temperature of 860°C

The average r -value for the Pattern IIb specimen deteriorates sharply when the finish-rolling temperature is dropped to below the A_{r3} transformation temperature, although the grain size of the hot band is as fine as that of the Pattern Ib specimen. The main orientation of the recrystallization texture in the Pattern IIb speci-

Fig. 13. $\phi = 45^\circ$ section of ODF at midplane of steel A cold-rolled of 80%.

Fig. 14. α - and γ -fiber texture at midplane of steel A cold-rolled of 80%.

men is $(111)[011]$, which is the same as that of batch-annealed Al-killed steel sheets. According to Fig. 7, the hot rolling texture of the Pattern IIb specimen is characterized by $(100)[011] \sim (211)[011]$ at the midplane and $(110)[uvw]$ in the neighborhood of the surface as a main orientation. Recently, several experiments concerning the hot band texture hot-rolled at below the A_{r3} transformation temperature have been reported.⁷⁻⁹ It is now well-known that the texture in α -region hot-rolled sheets is non-homogeneous throughout the thickness. Saito *et al.*¹⁰ has reported that the main textures at the 1/16 thickness layer of hot band finish-rolled at 650°C are $\{110\}\langle 001 \rangle \sim \{110\}\langle 111 \rangle$ and $\{211\}\langle 111 \rangle$, whereas those at the midplane are α - and γ -fiber. The latter is the same as for a typical cold rolling texture. Moreover, the texture of sheet fully annealed at 700°C is quite similar to that of as-hot-rolled sheet at the 1/16 thickness layer, but differs from the texture of hot-rolled sheet at the midplane, which consist of a sharp $\{111\}$ texture with a maximum at $\{111\}\langle 112 \rangle$, together with other minor components such as $\{100\}\langle 001 \rangle$.

In this experiment, the main texture at the midplane of the Pattern IIb specimen, however, is $(100)[011] \sim (211)[011]$, although that in the vicinity of the surface is $(110)[uvw]$. The difference midplane textures obtained in the present study and in Saito's study¹⁰ are assumed to be due to the difference in finish-rolling temperature,

which influences the extent of stored lattice energies and deformed texture during hot rolling. In Saito's study, the texture of the hot band finish-rolled at 650°C resembles the texture from cold rolling, and the stored lattice energies are hardly released at all during hot rolling, because the most of the total reduction achieved in hot rolling is performed at temperatures far below the A_{r3} transformation temperature, that is, in the α -region. Therefore, the subsequent annealing of hot band brings about a sharp $\{111\}$ texture with its maximum at $\{111\}\langle 112 \rangle$, which is the typical recrystallization texture obtained in the cold-rolled and annealed IF sheet steels.

In contrast, in the present study, the microstructure of sheet hot-rolled at 860°C , which is regard to be just below the A_{r3} transformation temperature, that is, the $\gamma + \alpha$ -region, consists of the equi-axial ferrite grain transformed from the austenite phase work-hardened from coarse grain, and the ferrite grain obtained by the dynamic recovering mainly or recrystallization from the deformed ferrite phase. This is considered to be the reason why the $(100)[011] \sim (211)[011]$ orientation is formed at the midplane in the Pattern IIb specimen.

In addition, the hot bands textures of the Patterns Ia and Ib specimens, which were hot-rolled at a finish-rolling temperature below the A_{r3} transformation temperature, 860°C , are homogeneous almost throughout the thickness. In this case, the austenite microstructure before final-pass rolling is fine grain and random texture because of the deformation from a huge reduction and subsequent recrystallization during the previous passes at the γ -region. The austenite phase deformed by final-pass rolling transforms finely and at random to ferrite phase during the subsequent water-spray cooling. Additionally, a small amount of ferrite phase is also deformed at final-pass rolling enough to recrystallize in an extremely short time. The texture of that transformed and recrystallized ferrite phase is considered to bring about the nearly random microstructure, even though the sheet is hot-rolled at below the A_{r3} transformation temperature.

As mentioned above, the recrystallization texture is known to have a certain crystallographic relationship with deformation. Accordingly, investigations were also carried out concerning the deformation stage. The experimental results for the 80% cold rolling texture of the Pattern IIb specimen are included in Figs. 13 and 14. The results obtained for the Pattern IIb specimen differs from those of the Patterns Ia and Ib specimens. The Pattern IIb specimen has an intensified $(111)[112]$ as its main orientation, but a markedly weakened α -fiber is observed in the Pattern Ib specimen. It resembles the recrystallization texture of IF steels.

It is difficult to conclude now why the cold rolling texture of $(111)[112]$ was formed from the hot rolling texture of $(100)[011] \sim (211)[011]$, even though there is an orientation relationship described by $35 \sim 55^\circ$ rotation about the $\langle 011 \rangle$ axis between $(111)[112]$ and $(100)[011] \sim (211)[011]$. On the other hand, the main orientation of the recrystallization texture of the Pattern IIb specimen is $(111)[011]$. A comparison of the cold rolling texture with the recrystallization texture indicates that the crystal rotation due to recrystallization mainly occurs by 30° about the ND// $\langle 111 \rangle$ axis.

It should be true that the recrystallization texture formed on the midplane of the Pattern IIb specimen affects the deep drawability and deteriorates it. Additionally, the non-homogeneity in the texture formed in the vicinity of the surface of the hot band may not be favorable for the overall average r -value of cold-rolled and annealed sheet, and consequently also for deep drawability. It has been determined that this non-homogeneity in texture stems from the friction between material and roll, as mentioned above.

5. Conclusion

For the purpose of improving the deep drawability of IF sheet steels, an examination has been made of the effect of hot rolling conditions, in particular the influence of thermal history, on the recrystallization texture in Ti-added ultra-low carbon cold-rolled sheet steels. The key findings can be summarized as follows:

(1) The average r -value of IF steel is significantly improved by refining the grain of the hot band, by means of decreasing the finish-rolling temperature to levels even below the A_{r3} transformation temperature, during hot rolling in Pattern I, in which the amount of thermal decrease during hot rolling is large.

(2) In hot rolling Pattern I, the annealed sheet textures with (554)[225] component, and the crystal rotation by cold rolling through two route, the (100)[01 $\bar{1}$] ~ (21 $\bar{1}$)[011] route about the ND//<111> axis and (211)[01 $\bar{1}$] ~ (111)[11 $\bar{2}$] skeleton route, developed more remarkable in the specimen hot-rolled with the finish-rolling of 860°C (Ib) than in the specimen hot-rolled at 940°C (Ia).

(3) From these results, it can be concluded that the beneficial development of (554)[225] component in the Pattern Ib specimen is brought about by selective grain-growth, mainly in the deformed matrix of (211)[01 $\bar{1}$] by a rotation of approximately 35° around the <011> fiber axis at -60° from the rolling direction.

(4) The average r -value deteriorates sharply with decrease in the finish-rolling temperature to levels below the A_{r3} transformation temperature with Pattern II hot rolling, in which the thermal drop during hot rolling process is small, although the grain size of the hot band is as fine as that of the Pattern Ib specimen.

(5) In the Pattern IIb specimen, the hot rolling texture is characterized by (100)[01 $\bar{1}$] ~ (211)[01 $\bar{1}$] at the midplane, and is (110)[uvw] in the vicinity of the surface as its main orientation, while the main orientation at the midplane is (111)[11 $\bar{2}$] for the cold rolling texture and (111)[01 $\bar{1}$] for the annealing texture.

(6) It should be concluded that the average r -value

of annealed sheet is deteriorated by the consequent recrystallization texture at the midplane, and that the texture formed in the vicinity of the surface in hot band may also affect its deep drawability. A comparison of the cold rolling texture with the annealing texture indicates that the crystal rotation due to recrystallization mainly occurs by 30° about the ND//<111> axis.

(7) The formation of (100)[01 $\bar{1}$] orientation at the midplane of the Pattern IIb specimen is believed to be attributed to the ferrite microstructure transformed from austenite phase work-hardened from coarse grain, and the ferrite microstructure obtained by dynamic recovering or by recrystallization from deformed ferrite phase.

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