Effect of Ti on the Development of Rolling Textures in High Purity Iron

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Dedicated to the memory of Professor Günter Wassermann

With the method of the crystallite orientation distribution function analysis, the development of the rolling texture was studied in detail on a Fe-0.004%C-0.20%Ti alloy cold rolled up to 90% reduction in thickness. It was found that the effect of Ti was widely different among various rolling texture components. The addition of Ti enhanced the development of the (110)RD fiber rolling texture component by promoting the rotation about the (110)RD axis. This resulted in the development of the strong {112}(110) texture component. As to the (110)TD fiber texture component, it was found that, although the addition of Ti did not affect the rotation about the (110)TD axis in the range {110}(001) to {554}(225), it strongly suppressed the rotation of {554}(225) into {111}(112). As a result, strong {554}(225) rolling texture component was developed below 70% rolling reduction. This component seems to provide origins of the {554}(225) recrystallization texture commonly observed in the Ti-stabilized steel. Such changes in the rolling texture would be expected, since slip modes are strongly affected by scavenging of C atom by Ti, or by the presence of very fine TiC precipitate particles.

KEY WORDS: Rolling texture, ODF-analysis, iron, titanium-alloyed, crystal rotation.

1 INTRODUCTION

It is well known that the addition of Ti greatly enhances the development of the {111} recrystallization texture in low carbon steels. As possible explanations for this, it has been thought that
solute Ti atoms (Goodenow and Held, 1970), precipitation of the TiC particles (Matsuo and Takahashi, 1971, Fukuda and Shimizu, 1972, 1975) and scavenging of C atoms by Ti (Kokubo et al., 1973) play important roles in the formation of the recrystallization texture.

On the other hand, rather little attention had been directed to the effect of Ti on the development of the cold rolling texture (Willis et al., 1975, Matsuo et al., 1972 and V. Schlippenbach et al., 1986). Since this effect might not be so pronounced as to be readily detected by the conventional pole figure method, more reliable and quantitative methods of the texture analysis should be adopted. By using the method of the three dimensional crystallite orientation distribution function analysis (Bunge, 1969), it has been found that the addition of Ti enhances the development of the \{112\}\{110\} component in the cold rolling texture (Willis et al., 1975, Matsuo et al., 1972, V. Schlippenbach et al., 1986). In most cases, this has been attributed to the presence of rather strong \{112\}\{110\} transformation textures in hot bands (Willis et al., 1975 and V. Schlippenbach et al., 1986), and contributions of other factors have not been clarified as yet.

Also rolling reductions adopted in the previous investigations have been limited between 70 (Willis et al., 1975 and Matsuo et al., 1972) and 80\% (V. Schlippenbach et al., 1986), and the effect of rolling reductions has not been studied systematically as yet. For this reason, it is not clear at present how the crystal rotation which occurs during cold rolling is affected by the addition of Ti.

In the previous paper (Inagaki, 1987), it has been found in a 0.05\%C rimmed steel that the path of this crystal rotation consists of following two branches, i.e., (i) \{110\}\{001\}→\{554\}\{225\}→\{111\}\{112\}→\{111\}\{110\}→\{332\}\{011\} and (ii) \{100\}\{001\}→\{100\}\{011\}→\{211\}\{011\}→\{322\}\{011\}, both leading to the common \{322\}\{011\} stable end orientation.

In this investigation, it was tried to clarify how the rate of rotation along these two pathes was affected by the addition of Ti. The influence of the transformation texture was eliminated by using hot bands having weak transformation textures. The effect of the transformation texture on the development of the cold rolling texture was independently evaluated by using hot bands in which strong transformation textures were developed by controlled rolling.
2 EXPERIMENTS

The starting materials used in the present investigation were 150 kg ingots of high purity Fe-0.004%C-0.20%Ti and Fe-0.023%C-0.13%Ti alloys, both melted in a vacuum induction furnace. Contents of other elements were Mn, 0.35%, P, 0.002%, S, 0.001% and Al, 0.032%, respectively. Fe-0.004%C-0.20%Ti alloy was used to investigate the effect of Ti on the rate of the crystal rotation which occurs during cold rolling. After slabing, the ingot of this alloy was hot rolled to the thickness of 3 mm. In order to avoid the development of the strong transformation texture, hot rolling was finished at 930°C. After pickling, the hot bands were cold rolled to 50, 70 and 90% reduction in thickness. Specimens cut from these cold rolled sheets were chemically thinned down to 0.05 mm thickness. From the {110}, {200} and {211} pole figures determined on these specimens, crystallite orientation distribution functions were calculated with the method of Roe up to \( l = 22 \). By comparing orientation distribution functions observed at each rolling reductions with those of the 0.05%C rimmed steel (Inagaki, 1987), the effect of Ti addition on the crystal rotation was quantitatively evaluated.

The effect of the transformation texture on the development of the cold rolling texture was investigated on Fe-0.023%C-0.13%Ti alloy. By controlled rolling (Inagaki, 1984) this alloy with the finishing temperature of 850°C, 3 mm thick hot bands having strong transformation textures could be prepared. This alloy was cold rolled 70% reduction in thickness. By comparing orientation distribution functions of these specimens with those of Fe-0.004%C-0.20%Ti alloy, the effect of the transformation texture was estimated.

3 RESULTS

3.1 Effect of Ti on the crystal rotation

Figure 1 illustrates \( \phi = 45^\circ \) sections of the crystallite orientation distribution functions observed in Fe-0.004%C-0.20%Ti alloy at each rolling reductions. (Positions of ideal orientations on \( \phi = 45^\circ \) sections are illustrated in detail elsewhere, Inagaki, 1987.) The hot
Figure 1  $\phi = 45^\circ$ sections of the crystallite orientation distribution functions observed in Fe-0.004%C-0.20%Ti alloy after 0, 50, 70 and 90% cold rolling.
band had a rather strong \(\{100\}<011>\) texture. However, components of transformation textures derived from the austenite rolling texture, such as \(\{311\}<011>\) and \(\{332\><113>\) (Inagaki, 1977) were considerably weak. Orientation distributions observed at each rolling reductions were qualitatively the same as those observed in 0.05\%C rimmed steel (Inagaki, 1987). However, intensities of textures were generally much stronger in Fe-0.004\%C-0.20\%Ti alloy. These features are more clearly illustrated in Figures 2, 3, 4 and 5, in which orientation distributions of Fe-0.004\%C-0.20\%Ti alloy are directly compared with those of the 0.05\%C rimmed steel.

Figure 2 shows orientation distributions along \(\psi = 90^\circ\) lines on \(\phi = 45^\circ\) sections given in Figure 1. Orientations having the \(\langle110\rangle\) axes parallel to the rolling direction (i.e. members of the \(\langle110\>\parallel RD^\dagger\) fiber texture) are all located on this line. In the as hot rolled condition, Figure 2(a), this fiber texture was somewhat stronger in Fe-0.004\%C-0.20\%Ti alloy. However, the difference was small. After 50\% rolling reduction, Figure 2(b), orientations in the range \(\{100\}<011>\) to \(\{111\><110>\) showed much more remarkable development in Fe-0.004\%C-0.20\%Ti alloy. Between these two specimens, the difference in these components became largest at 70\% rolling reduction, Figure 2(c).

At 90\% rolling reduction, Figure 2(d), the difference became smaller again. However, throughout the orientation range between \((001)[1\bar{1}0]\) and \((\bar{1}10)[\bar{1}1\bar{0}]\), all orientations were clearly much stronger in Fe-0.004\%C-0.20\%Ti alloy than in 0.05\%C rimmed steel. Also \((223)[\bar{1}10]\) stable end orientation (Inagaki, 1987) was much more well-developed in Fe-0.004\%C-0.20\%Ti alloy.

From these results, it is clear that, even in the absence of the strong \(\{112\><110>\) component in hot bands, strong \(\langle110\>\parallel RD\) fiber rolling texture, above all \(\{211\><011>\) rolling texture component can be developed by the addition of Ti.

Figure 3 shows orientation distributions along \(\psi = 0^\circ\) lines on \(\phi = 45^\circ\) sections given in Figure 1. All orientations having the \(\langle110\rangle\) axes parallel to the transverse direction and lying in the range \((001)[1\bar{1}0]\) to \((\bar{1}10)[\bar{1}01]\) (i.e., members of the \(\langle110\>\parallel TD\) fiber textures) are located on the line.

\(^\dagger\) In this paper, RD, TD and ND denote rolling, transverse and normal directions, respectively.
Figure 2 Orientation distributions along $\psi = 90^\circ$ line on $\phi = 45^\circ$ sections given in Figure 1. Rolling reductions; a) 0% (as hot rolled), b) 50%, c) 70% and d) 90%, respectively.

In the as hot rolled condition, Figure 3(a), 0.05%C rimmed steel had a uniform orientation distribution between (111)[112] and (110)[001], whereas Fe-0.004%C-0.20%Ti alloy showed in this range a broad peak whose center was located at about (554)[225] orientation. Since orientations near (554)[225] would be derived from the Cu type rolling texture in the austenite phase via Kurdjmov-Sachs relationship (Inagaki, 1977), these results suggest that some of the austenite in Fe-0.004%C-0.20%Ti were not recrystallized. In this hot band, however, strong (001)[110] and (110)[110] components were also present. Since these components would be derived from the austenite recrystallization texture via
Kurdjmov-Sachs relationship (Inagaki 1977), it might be concluded that most of the austenite were recrystallized in this hot band.

After 50% cold rolling, Figure 3(b), crystal rotation occurred in both specimens about the (110) axis lying parallel to the transverse directions, forming distinct peaks at $\theta = 55$ or 60°. However, this rotation seemed to occur more readily in 0.05%C rimmed steel than
in Fe-0.004%C-0.20%Ti alloy. In the former case, the peak was located at (111)[112], whereas in the latter case, it was located at (554)[225]. The height of the peak was nearly the same in both cases.

After 70% cold rolling, Figure 3(c), both peaks increased their heights appreciably. This increase was much larger in Fe-0.004%C-0.20%Ti alloy. Positions of peaks, however, did not show any change. It should be noted that (554)[225] orientation was quite stable in Fe-0.004%C-0.20%Ti alloy. In this alloy, rotation about the ⟨110⟩∥TD axis beyond (554)[225] orientation seems to be strongly suppressed. Further rotation about the ⟨110⟩∥TD axis toward (111)[112] occurred only after 90% cold rolling, Figure 3(d).

Figure 4 shows orientation distributions along θ = 55° line on φ = 45° sections given in Figure 1. All orientations having ⟨111⟩ axes parallel to the normal direction (i.e., members of the ⟨111⟩∥ND fiber texture) are located on this line. Below 70% rolling reduction, no marked difference was observed between 0.05%C rimmed steel and Fe-0.004%C-0.20%Ti alloy. In both cases, rather uniform orientation distributions were observed after 70% cold rolling, indicating the formation of the uniform ⟨111⟩∥ND fiber texture.
texture. After 90% cold rolling, \(\{111\}\langle011\rangle\) orientations became in both cases the main orientation among the members of the \(\langle111\rangle\parallel\text{ND}\) fiber texture. But it should be noted in the case of Fe-0.004%C-0.20%Ti alloy that, above 70% rolling reduction, all members of the \(\langle111\rangle\parallel\text{ND}\) fiber texture increased their intensities appreciably.

### 3.2 Effect of the transformation texture

In Figure 5, \(\phi = 45^\circ\) sections of the crystallite orientation distribution functions observed in hot bands of Fe-0.004%C-0.20%Ti and Fe-0.023%C-0.13%Ti alloys are compared. It is evident that crystallite orientation distribution functions are quite different between these two alloys.

This is more clearly shown in Figures 6 and 7. Figure 6 shows orientation distributions along \(\psi = 90^\circ\) line on \(\phi = 45^\circ\) sections given in Figure 5. In Fe-0.023%C-0.13%Ti alloy, a distinct peak was present at \((\overline{1}13)[\overline{1}10]\) orientation. \((001)[\overline{1}10]\) and \((110)[\overline{1}10]\) were much weaker than in Fe-0.004%C-0.20%Ti alloy.

![Figure 5](image)

**Figure 5** \(\phi = 45^\circ\) sections of the crystallite orientation distribution functions observed in the hot bands of Fe-0.004%C-0.20%Ti and Fe-0.023%C-0.13%Ti alloys.
Figure 6  Orientation distributions along $\psi = 90^\circ$ line on $\phi = 45^\circ$ sections given in Figure 5.

Figure 7  Orientation distributions along $\psi = 0^\circ$ line on $\phi = 45^\circ$ sections given in Figure 5.
Figure 7 illustrates orientation distributions along $\psi = 0^\circ$ line on $\phi = 45^\circ$ sections given in Figure 5. In Fe-0.023%C-0.13%Ti alloy, a strong peak was present at $\langle 554 \rangle [225]$ orientation.

Thus, it is evident that the hot bands of Fe-0.023%C-0.13%Ti alloy had a relatively strong texture consisting of $\langle 554 \rangle \langle 225 \rangle$ and $\langle 311 \rangle \langle 011 \rangle$ main orientations. These are components of the transformation textures derived from the Cu-type austenite rolling texture via Kurdjumov-Sachs relationship (Inagaki, 1977). In this alloy, recrystallization of the austenite might have been more effectively suppressed than in Fe-0.004%C-0.20%Ti alloy, since volume fraction of TiC particles precipitated was larger because of higher C content. In addition to this, since controlled rolling with the lower finishing temperature was adopted in hot rolling, austenite rolling texture developed in this alloy might have been much stronger. As a result, transformation textures inherited from these austenite rolling textures were much stronger in this alloy. Figure 8 shows $\phi = 45^\circ$ sections of the crystallite orientation distribution functions observed after 70% cold rolling. At first glance it is clear that the rolling texture is weaker in Fe-0.023%C-0.13%Ti alloy than

![Figure 8](image-url)

**Figure 8** $\phi = 45^\circ$ sections of the crystallite orientation distribution functions observed in Fe-0.004%C-0.20%Ti and Fe-0.023%C-0.13%Ti alloys after 70% cold rolling.
Figure 9  Orientation distributions along $\psi = 90^\circ$ line on $\phi = 45^\circ$ sections given in Figure 8.

Figure 10  Orientation distributions along $\psi = 0^\circ$ line on $\phi = 45^\circ$ sections given in Figure 8.
in Fe-0.004%C-0.20%Ti alloy. Orientation distributions along $\psi = 90^\circ$ line on these $\phi = 45^\circ$ sections are shown in Figure 9. It should be noted that orientations in the range between (001)[110] and (111)[110] were not well developed in Fe-0.023%C-0.13%Ti alloy. In this alloy, orientations in the range between (001)[110] and (112)[110] were even weaker than in 0.05%C rimmed steel.

Figure 10 shows orientation distributions along $\phi = 0^\circ$ line on $\phi = 45^\circ$ sections given in Figure 8. Orientation distributions in Fe-0.023%C-0.13%Ti alloy had a maximum at (111)[112] orientation. However, both the height and position of this maximum were the same as those observed in the 0.05%C rimmed steel.

From these results, it is evident that the addition of excess amount of C smears out the strong development of $\{112\langle110\}$ and $\{554\langle225\}$ rolling texture components observed in Fe-0.004%C-0.20%Ti alloy).

4 DISCUSSION

With the crystallite orientation distribution function analysis, it has already been found that the addition of Ti enhances the development of the $\{112\langle110\}$ component of the rolling texture in low carbon steels cold rolled 70 to 80% (Willis, 1975, Matsuo et al., 1972, V. Schlippenbach et al., 1986). Similar observations have been made also in Nb stabilized steels (Willis et al., 1975 and Hook et al., 1975). Hook, Heckler and Ellias studied the effect of Nb addition on the development of the rolling texture in Fe-0.005%C alloy cold rolled 60%. They found that the $\{112\langle110\}$ rolling texture component increased remarkably with the increasing Nb content. Since the rolling reduction adopted in these investigations have been limited between 60 and 80%, it is not clear at present how the whole rotation pathes are affected by the addition of Ti or Nb. By varying the rolling reduction in wide range and by following the rotation of various rolling texture components systematically, it was found in the present investigation that not only the development of the $\langle110\rangle||RD$ fiber rolling texture, but also the development of the $\langle110\rangle||TD$ fiber rolling texture were strongly affected by the addition of Ti. Since these effects of Ti are widely different
between these two fiber rolling texture components, they will be
discussed separately below.

4.1 \( \langle 110 \rangle || RD \) fiber rolling texture

Development of the strong \{112\} \( \langle 110 \rangle \) component in the rolling
texture of Ti and Nb steels have been attributed in the past
investigations to the presence of strong \{112\} \( \langle 110 \rangle \) transformation
texture in hot hands (Willis et al., 1975, V. Schlippenbach et al.,
1986 and Hook et al., 1975). It has been considered that cold rolling
results in the reinforcement of this component, since it is the stable
end orientation of cold rolling (Hook et al., 1975). It is wellknown
in the case of control-rolled low carbon microalloyed steels that
such transformation textures can be developed appreciably, if
recrystallization of austenite is suppressed by the precipitation of
NbC or TiC, and if heavy rolling reductions are given to the
unrecrystallized austenite (Inagaki, 1977). The austenite rolling
texture developed through this processing is supposed to be Cu
type, and the main orientations of the transformation texture
derived from this austenite rolling texture via Kurdjmov-Sachs
relationship were identified, in the case of 0.1%C-1.3%Mn-
0.04%Nb steel, to be \{311\} \( \langle 011 \rangle \) and \{322\} \( \langle 113 \rangle \) (Inagaki, 1977).
Although previous investigations have identified one of the main
orientations of the transformation texture as \{211\} \( \langle 011 \rangle \),
\{311\} \( \langle 011 \rangle \) seems to be better description of this component. This
is also evident from Figure 6. In very low C steel, \{332\} \( \langle 113 \rangle \)
component seems to be better described as \{554\} \( \langle 225 \rangle \). This is also
evident from Figure 7, although spread of orientation distribution
about this orientation is large.

It is readily understood that strong \{112\} \( \langle 110 \rangle \) rolling texture
component can be developed by cold rolling hot bands containing
strong \{311\} \( \langle 011 \rangle \) transformation texture, since \{211\} \( \langle 011 \rangle \) and
\{311\} \( \langle 011 \rangle \) are only 10° apart, and since \{211\} \( \langle 011 \rangle \) is located on
the rotation path toward the \{322\} \( \langle 011 \rangle \) stable end orientation
(Inagaki, 1987). In the case of Fe-0.004%C-0.20%Ti alloy, how-
ever, \{311\} \( \langle 011 \rangle \) transformation texture component was not well
developed in hot bands, Figures 2 and 6, since finishing temperature
of hot rolling was intentionally selected to be above the re-
crystallization temperature of the austenite. In spite of this,
development of the strong \{112\}\langle110\rangle rolling texture component could be observed in this alloy cold rolled above 50% reduction in thickness. Development of the strong \{112\}\langle110\rangle rolling texture component cannot be explained therefore with the mechanism described above. It seems that either absence of solute carbon atom, or presence of fine TiC or NbC particles in hot bands enhances the slip rotation which results in the development of the strong \{112\}\langle110\rangle rolling texture component.

In ultra low carbon Ti stabilized steels, it has been confirmed that very fine TiC particles are already precipitated in hot bands (Matsuoka and Takahashi, 1971, Fukuda and Shimizu, 1972). Also, the results of internal friction measurements have indicated that solute C atoms are absent in these hot bands due to the scavenging effect of Ti atom (Fukuda and Shimizu, 1972).

In the case of Fe-0.023%C-0.13%Ti alloy, it is interesting to note that, although strong \{311\}\langle011\rangle transformation texture component was developed in the hot band, \langle 110\rangle||RD fiber rolling texture observed after 70% rolling was quite weak. It has been observed that, in Ti stabilized steel with higher C content, TiC particles precipitated in the hot band are relatively coarse (Matsuoka and Takahashi, 1971). Presence of these coarse particles might have randomized the cold rolling texture.

From these results, it might be concluded that, although the presence of \{311\}\langle011\rangle transformation texture component in the hot band may be favourable to the development of \{112\}\langle110\rangle component in the rolling texture, it is not a necessary condition, at least at higher rolling reductions.

4.2 \langle110\rangle||TD fiber rolling texture

From Figure 3, it is clear that, in Fe-0.004%C-0.20%Ti alloy, the rotation of crystals from \{554\}\langle225\rangle to \{111\}\langle112\rangle orientations was strongly suppressed below 70% rolling reduction. As a result, a strong stable peak was formed at the \{554\}\langle225\rangle orientation. Since crystals in this orientation would be severely strained, it is expected that, on annealing these specimens, strong \{554\}\langle225\rangle recrystallization texture would be developed. This is in fact just what has been observed (Willis et al., 1975, V. Schlippenbach et al., 1986, Akisue and Takashima, 1973). The mechanism through which
the rotation of the \( \{554\}<225> \) orientation into the \( \{111\}<112> \) orientation is suppressed, is not clear at present. However, presence of very fine TiC precipitate particles seems to play here an important role.

Above 70% rolling reduction, such barrier to the crystal rotation from \( \{554\}<225> \) to \( \{111\}<112> \) was completely eliminated, and since all of these strong \( \{554\}<225> \) orientations are rotated into \( \{111\}<112> \), and further into \( \{111\}<110> \) orientations, \( \langle 111\rangle \parallel ND \) fiber texture which is much stronger than that observed in the 0.05%C rimmed steel was developed.

Thus, it might be concluded that, below 70% rolling reduction, the addition of Ti enhances the development of the \( \{554\}<225> \) orientation by suppressing the rotation of the \( \{554\}<225> \) orientation into the \( \{111\}<112> \) orientation.

In Fe-0.023%C-0.13%Ti alloy, however, such effect of Ti could not be observed. This might be also ascribed to the large amount of coarse TiC particles precipitated in the hot band of this alloy.

5 CONCLUSIONS

The addition of Ti significantly affects the development of the rolling texture in high purity iron. This effect is widely different among various components of the rolling texture.

(1) \( \langle 110\rangle \parallel RD \) fiber rolling texture components; The addition of Ti enhances the rotation of these components about the \( \langle 110\rangle \parallel 11RD \) axis. This leads to the development of the strong \( \{112\}<110> \) rolling texture component.

(2) \( \langle 110\rangle \parallel TD \) fiber rolling texture components; In the orientation range between \( \{110\}<001> \) and \( \{554\}<225> \), the addition of Ti does not significantly affect the rotation of these components about the \( \langle 110\rangle \parallel TD \) axis. However, the rotation of the \( \{554\}<225> \) orientation into the \( \{111\}<112> \) orientation is strongly suppressed by the addition of Ti. This results in the development of the strong \( \{554\}<225> \) rolling texture component.

Thus, the rolling texture of the Ti-stabilized steel can be characterized by the strong \( \{554\}<225> \) and \( \{112\}<110> \) components.
EFFECT OF Ti ON THE TEXTURES OF IRON

References