Formation of $\alpha \rightarrow \gamma \rightarrow \alpha$ Transformation Texture in Sheet Steel

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Synopsis

For the purpose of investigating the formation mechanism of the texture which is developed by ferrite ($\alpha$) $\rightarrow$ austenite ($\gamma$)$\rightarrow$ ferrite ($\alpha$) transformation, the effects of heating and cooling rates during phase transformation and specimen thickness on the transformation texture have been studied by using an extra low carbon sheet steel with the initial texture of $\{111\}$. The results obtained are summarized in the following:

1. Formation of $\alpha \rightarrow \gamma \rightarrow \alpha$ Transformation Texture in Sheet Steel

(i) a single fiber texture with [001] or [012] orientation was developed12~ in decarburized columnar grains in carbon steel,

(ii) a developed texture was different,13~16~ depending on carburizing, decarburizing or surface condition,

(iii) the texture with $\{100\}$ component produced during $\gamma \rightarrow \alpha$ transformation was discussed in terms of $H_2S$ partial pressure,15~

(iv) the transformation texture with $\{111\}$ component was formed14~18~ by penetration of Zn or Al into steel. In most of these studies, the formation mechanism of the transformation texture was discussed in terms of the surface energy12,13,15,16~ in $\alpha$ or $\gamma$ grains. No paper considered it on the basis of the orientation relationship between $\alpha$ and $\gamma$ crystals. The transformation texture with $\{100\}$ ($uvw$) component was reported29~ to form in extra low carbon steel annealed in dry $H_2$ atmosphere, though it is not clear whether the specimen had the columnar grains.

In category (3), most papers interpreted17~22~ the transformation texture in terms of the orientation relationship between $\alpha$ and $\gamma$ crystals, based on the Kurdjumov–Sachs (KS) relation.28~ In one37~ of the papers, the transformation texture was explained by introducing the selection rule of variants based on the relation between slip systems of crystal and induced strain by hot rolling. In category (4), Horta23~ obtained the quantitative relation between the change in texture and volume fraction of $\gamma$ phase formed during intercritical annealing in $\alpha+\gamma$ duplex phase region. According to his study, $\{111\}$ texture developed with the increase in the fraction transformed. Its intensity became highest at about the $\gamma$ fraction of 60 %, beyond which with the increasing $\gamma$ fraction, $\{111\}$ intensity decreased and $\{321\}, \{100\}$ and $\{210\}$ intensities increased. It was reported by Jenkins and Wilson24~ and Saiki25~ that intercritical annealing in $\alpha+\gamma$ region with rapid heating and cooling rates increased $\{100\}$ intensity but decreased $\{111\}$ intensity. The reason for this was sought in that $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation enhanced the formation of a texture with random orientations.

Besides these studies on the transformation texture, there were many investigations30~34~ on the orientation of an individual $\alpha$ nuclei during $\gamma \rightarrow \alpha$ transformation or on the texture formed by $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation.39~ The effect of applied stress was also reported40~ on the orientation developed during $\alpha \rightarrow \gamma$

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→α transformation. In all of them, the KS relation was considered to be the reasonable orientation relationship between α and γ crystals. However this relationship was reported\(^{111}\) to change depending on applied stress.

The formation mechanisms of the γ→α transformation texture described in 1) to 4) were explained basically by the following two theories. One theory (I) considers that in order to decrease surface energy γ grains with a favourable orientation preferentially transforms to α, or α grains with the favourable orientation are preferentially formed, thereby developing the texture with {100}, {110} or their neighbouring orientations. Another theory (II) is that texture develops with a definite orientation relationship, for example the KS relation between α and γ crystals during transformation. Either theory has to explain the common phenomenon that the texture of {100}, {110} or their neighbouring orientations develops by γ→α transformation. This strongly suggests that such texture is the basic texture which develops during γ→α transformation and that this phenomenon occurs independent of these theories.

If there is a definite orientation relationship between α and γ crystals during γ→α transformation, the texture in γ phase must give some effects on the transformation texture in α phase. Except the studies on martensitic transformation\(^{1-10}\) and hot rolling texture,\(^{17-22}\) however, it has not been tried to correlate the transformation texture to the texture in its mother phase in low carbon steel.

In contrast with the texture developed during γ→α transformation, the texture formed by α→γ transformation was investigated only in Cu–Zn alloy.\(^{36}\) There has not been such a study in steel. It is a matter of fact that there has not been any research performed to study the texture formation in the whole course of α→γ and γ→α transformations in steel.

The present experiment was purposed to investigate the texture formed during α→γ transformation and thereby to elucidate the reason why {100}, {110} or their neighbouring orientation is frequently observed in sheet steel transformed from γ to α. That is, the relationship between textures prior to and after transformation was investigated in terms of heating and cooling rates during α→γ→α transformation, and the effect of specimen thickness on the difference in the transformation texture between surface and midsection of specimen was also studied. The effects of initial texture prior to γ transformation, heating and cooling rates and the effect of specimen surface on the transformation texture were considered, and a theory on the formation mechanism of α→γ→α transformation texture was proposed. With this theory for the first time we can reasonably and consistently explain the formation of the texture with {100}, {110} or their neighbouring orientations, which could not be satisfactorily explained by the former theory (I) and (II) or by the combination of them. It can be concluded by the theory, that the texture with {100}, {110} or their neighbouring orientations is the basic texture which is formed due to a crystallographic anisotropy of the transformation strain during α→γ→α transformation, even if it may be somewhat affected by decarburization, surface condition, atmosphere and others.

II. Experiment

To obtain the starting material with a low carbon content, a hot rolled steel sheet of 2.3 mm thickness was decarburization-annealed (DC). Its chemical compositions are shown in Table 1. The average grain diameter of this steel was about 100 μm (thereafter described as d=100 μm). A portion of this sheet was further subjected to strain-annealing to give grain diameter of ~600 μm (d=600 μm).

Both materials were cold rolled, cleaned and subsequently subjected to γ transformation treatment under various annealing conditions. In order to prevent specimens from sticking during annealing, the following measure was taken:

1. The stacked specimens with alumina powder between them were packed by thin sheet with the same chemical compositions with the specimens and annealed in argon atmosphere, or

2. Specimens separated with silica wool were sealed in argon atmosphere in a quartz capsule.

Heating rate was either rapid with 50~100 °C/min or slow with 14~100 °C/h. Cooling was done by quenching in water (~200 °C/s), cooling in air (~5 °C/s), or cooling in furnace (50~100 °C/h). Preferred orientation was determined by obtaining both (200) pole figure and intensity ratio (x random) of X-ray diffraction from (200), (211), (222) and (110) planes lying in parallel with the specimen surface. They were obtained using molybdenum Kα radiation after correcting extinction effect. Texture was measured on the specimen surface unless specified otherwise. The radiated area of specimen was about 20×20 mm².

III. Experimental Result

1. Grain Structure and Texture of Specimen Prior to α→γ Transformation

The specimen (d=100 μm) cold rolled by 74 % was heated at 800 °C for 15 min with either rapid (80 °C/min) or slow (50 °C/h) rate and air cooled at about 5 °C/s. The change in the recrystallization texture with distance from sheet surface is shown in Fig. 1. The {111} orientation is predominant at any distance irrespective of heating rate. In both specimens the intensity of {111} orientation is the minimum at 1/4 depth and the maximum at midsection. When a comparison is made at the same distance the {111} intensity is higher for slowly heated specimen than for rapidly heated one. While the intensities of {100}, {112} and {110} orientations are weak, showing

<table>
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<tr>
<th>Steel</th>
<th>C</th>
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<th>Mn</th>
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<td>A</td>
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<td>0.091</td>
<td>0.34</td>
<td>0.011</td>
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little change of intensity with distance. Though the intensity level is low, the {100} texture develops in slowly heated specimen, whereas the {110} texture develops in rapidly heated specimen. The microstructures of these two specimens are shown in Photo. 1, exhibiting relatively uniform grains. Slowly heated specimen has somewhat larger grains, though it is not so much distinguishable.

2. Effect of Heating and Cooling Rates during \( \alpha \rightarrow \gamma \rightarrow \alpha \) Transformation

Sheet specimen \((d=100 \mu)\) was cold rolled to 0.48 mm thickness with 80 \% reduction. This specimen was heated to 800 °C at 50 °C/h and further to 950 °C at 14, 100 or 300 °C/h. After being held for 15 min at this temperature, one specimen was quenched in water and another was furnace-cooled (the average cooling rate between 950 °C and 700 °C was about 100 °C/h).

The relation between transformation texture and heating rate in these specimens is shown in Fig. 2, where the (200) pole figures are also shown for representative specimens. For the quenched material \{110\} component decreases with decrease in heating rate, whereas \{111\} and \{100\} components increase. As shown in the pole figure (a), the texture in the rapidly heated specimen is almost random, though it has weak \{100\} and \{111\} components. For the furnace-cooled material, \{100\} component decreases and \{110\} component increases with increase in heating rate. The \{111\} and \{211\} components decrease gradually with increasing in heating rate. Figure 2
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(b) is a pole figure obtained from the specimen rapidly heated and furnace-cooled. There is a strong component at the orientation about 20 deg off \( \{100\} \). With decrease in heating rate, the strongest component tends to move to the \( \{100\} \) orientation. Photograph 2 shows the microstructure in a longitudinal cross section of the specimen transformed at the heating rate of 14 \( ^\circ \text{C}/\text{h} \) or 300 \( ^\circ \text{C}/\text{h} \). This grain structure is different from the so-called “columnar structure” reported previously.\(^{12-16}\) From these preliminary experiments, the necessary conditions to develop the \( \{100\} \) or \( \{110\} \) component by the transformation were obtained. Further studies were performed to investigate the formation of these textures in more detail where furnace-cooled materials alone were used.

3. Change in Transformation Texture with Distance from Sheet Surface

Transformation texture has been investigated so far at the surface of specimen. This texture, however, may be different from that in the midsection of the specimen. To examine the effect of specimen thickness on the texture, a 100 \( \mu \text{m} \) thick material was cold rolled in the range from 10 to 90 \%, slowly heated to 950 \( ^\circ \text{C} \), kept for 15 min, and subsequently cooled in furnace. Pole densities at surface and midsection are shown in Fig. 3 with the specimen thickness or the amounts of the cold rolled reduction. At midsection, the texture consists of \( \{11\} \) component in the range from 1.38 to 0.48 mm, \( \{100\} \) in the thinner range and has random orientation in the thicker range. The \( \{110\} \) intensity does not change, but the \( \{100\} \) intensity decreases with increasing thickness. At surface the texture consists of \( \{100\} \) component except the specimen with the reduction less than 20 \%. The intensity of this component does not change in the reduction range greater than 60 \%. In the low reduction range, \( \{11\} \) component is relatively strong in comparison with \( \{21\} \) and \( \{11\} \) components which are always weak independent of the amount of reduction. In the range greater than 85 \%

(thickness \( \leq 0.35 \) mm), there is no difference in texture between surface and midsection, which consists mainly of \( \{100\} \) component. Photograph 3 shows the microstructures of the 0.69 mm and 0.35 mm thick specimens shown in Fig. 3. In the thick specimen, grains constituting each surface layer did not meet with grains constituting the opposite surface layer at midsection and the finer grains than those were formed there. On the other hand, in the thin specimen grains constituting both surface layers meet together at the centre of thickness. Consequently in the latter specimen the surface texture is the same as that in the midsection. This behaviour is the same for the rapidly

![Photo. 2. Effect of heating and cooling rates during \( \alpha \rightarrow \gamma \rightarrow \alpha \) transformation on cross-sectional microstructure in specimen heated at 950 °C for 15 min.](image)

![Photo. 3. Effect of thickness on cross-sectional microstructure in specimen shown in Fig. 3.](image)
heated specimen and slowly heated one except that in the former the surface texture consists of \{110\} instead of \{100\} (see Fig. 2). That is, in the mid-section of a thick specimen, an almost random texture having a weak \{111\} and \{100\} components is formed regardless of heating rate. At the surface, the \{100\} texture and \{110\} texture develop for slow-heating transformation and rapid-heating one, respectively. This experiment, however, may not demonstrate the exact effect of thickness on the transformation texture completely. The difference in recrystallization texture caused by a different amount of cold reduction might be superimposed on the transformation texture. However, specimens in Photo. 3 were cold rolled to 70\% and 85\%, respectively, and it is generally known that this difference in such a high range of reduction percent has little effect on the kind of main component of recrystallization texture. Accordingly, as long as a specimen is given with the cold rolling reduction of more than 60\%, the difference in transformation texture between surface and midsection is essentially due to the thickness effect which, in turn, devolves on grain structures consisting of two layers or more than three layers.

4. Relationship between Initial and Transformation Textures

Specimens (d=600 μ) with cold rolling reduction of 80\% were heated either slowly or rapidly to 950 °C, held for 15 min, and then furnace-cooled. Some of the specimens were air-cooled from 800, 850 and 890 °C to examine the texture prior to \(\alpha \rightarrow \gamma\) transformation. Changes in texture by cold rolling, heating and transformation are shown in Fig. 4. In both specimens heated to 890 °C just below transformation temperature (the textures in these specimens are called initial textures), \{111\} intensity is the strongest independent of heating rate followed by \{100\}, \{110\} and \{211\} intensity. While, in the specimens heated to 950 °C, above the transformation temperature, \{100\} and \{110\} textures develop for slow heating and rapid heating respectively, and exhibit low intensity in other components. This result is the same for the material with prior grain size of 100 μm (see Fig. 2). That is, as long as the specimens with the same initial texture are transformed at the same heating and cooling rate, transformation textures obtained are the same regardless of the difference in initial grain size. As seen in the pole figure, the texture prior to \(\sigma \rightarrow \gamma \rightarrow \alpha\) transformation consists of \{111\} \{011\}, \{111\} \{211\} and \{100\} \{001\} orientations and there is no difference in the texture between slow-heated and rapid-heated specimens. After \(\alpha \rightarrow \gamma \rightarrow \alpha\) transformation, however, the slowly heated specimen has a very strong intensity in the \{100\} and its neighbouring orientations, whereas, the specimen rapidly heated exhibits a weak intensity in the orientations within 10 deg off \{100\} and a strong intensity in the orientations of about 20 deg off \{100\}, which spreads therefrom to \{110\} orientation in the form of a doughnut around the centre of pole figure. Such behaviour is similar to the specimen with \(d=100 \mu\) which was heated rapidly to transformation temperature and cooled slowly as shown in Fig. 2 (b).

IV. Discussion

The foregoing results are summarized in the following:

1. The transformation texture in the surface layer changes with the heating rate even if an initial texture is the same.
2. The transformation texture changes with the...
cooling rate during $\gamma \rightarrow \alpha$ transformation.

(3) In slowly cooled specimens the surface texture has \{100\} component, and components of \{110\} and orientation of 20~30 deg off \{100\}, respectively for slow heating and rapid heating.

(4) In the midsection of a thick specimen the transformation texture is nearly random, exhibiting weak components of \{111\} and \{100\} orientations regardless of heating and cooling rates. This is similar to the surface texture of the specimen which is transformed with rapid heating and rapid cooling rates.

Theories on the development of a transformation texture have been classified into two categories:

(I) Ferrite grain with \{110\} or \{100\} orientation is preferentially formed during $\gamma \rightarrow \alpha$ transformation, because these orientations have a lower surface energy between matrix and atmosphere.\(^{3,15,16}\)

(II) Transformation texture is formed, because there is a special orientation relationship between $\alpha$ and $\gamma$ phases.\(^{3,17,22}\)

But the facts obtained for the first time in this experiment that the transformation texture depends strongly on heating rate cannot be explained by either theory of (I) or (II). The fact that there is a difference in texture between surface and midsection cannot be explained in terms of (II).

Abe and Ito\(^{12}\) considered that the $\gamma$ grain with a special orientation transforms preferentially into $\alpha$ phase in order to reduce a surface energy in $\gamma$ phase. In this case, however, even if there is a special orientation relationship between $\alpha$ and $\gamma$ phases, for example the KS relation, as long as all variants are operated, the transformation texture ought to have some components of more than two orientations which are obviously different from one another. Considering that the present experimental results can not be explained in terms of (I), (II) or combination of (I) and (II), we propose a new theory on the formation mechanism of the transformation texture which will be shown to explain the experimental results reasonably and consistently.

1. Hypothesis

The lattice orientation relationship between bcc and fcc is assumed to obey essentially the KS relationship, i.e., \{(111)\}_w//\{(111)\}_t, and \{[111]_w//\{[110]_t, during $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation, because many studies\(^{36-38}\) supported it in steel. In this study, the \{100\} texture or the texture consisting of \{110\} orientations and orientations of 20~30 deg off \{100\} developed solely in the surface layer of a sheet specimen. Accordingly the development of such textures is considered to be associated with a certain characteristic of a sheet surface. Taking into consideration the anisotropy of transformation strain and the easy release of the strain at a sheet surface, it can be assumed that in the surface layer the perpendicular component of transformation strain is released more easily than the other components. It is reasonable that the preferential operation of variants with the smaller ratio of virtual work in normal direction to total virtual work makes the transformation energy least.

Clarke\(^{37}\) calculated the shape and orientation of an ellipsoidal region of transformed phase that minimized the elastic strain energy by using Eshelby's method.\(^{38}\) This method is useful to determine habit planes in martensitic transformation. However, it is impossible to apply the method directly to a selection of favourable variants during $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation by a diffusional process. The following discussion is simply based on the ratio of the elastic work. The ratio of the virtual work is calculated on the basis of the ordinary elastic theory by multiplying strain by stress. $R_\alpha$ and $R_\gamma$ are the ratios for $\alpha \rightarrow \gamma$ and $\gamma \rightarrow \alpha$ transformations, respectively, where $i$ and $j$ are variants in the KS relation, the variant with the larger value of $R_\alpha$ or $R_\gamma$ transforms preferentially during $\alpha \rightarrow \gamma$ or $\gamma \rightarrow \alpha$ transformation. Let us consider the selection rule of variants and the relation between heating or cooling rate and the selection rule of variants as follows. When a specimen is heated or cooled with a constant rate between $A_i$ and $A_j$ transformation temperatures, the amount of transformation is thought to vary linearly with temperature. Its amount per unit time is defined as the transformation rate. It is reasonable that at the lower transformation rate favorable variants are selected more markedly and at the higher transformation rate every variant tends to transform, losing the selection rule of variants. Therefore, the selection rule of variants is defined as follows:

(1) In the case of the higher transformation rate every variant operates in equal probability at near sheet surface. In the midsection of a thick specimen every variant operates in equal probability regardless of transformation rate, because the strain constraint is nearly the same in the sheet normal direction as well as in the parallel direction to sheet surface.

(2) In the case of the lower transformation rate, the variant with larger $R_\alpha$ and/or $R_\gamma$ value preferentially transforms and grows one after another, and the variant with lower $R_\alpha$ and/or $R_\gamma$ value is suppressed to transform.

2. Strain, Stress and Work During Transformation

The strain accompanying the structural change from bcc to fcc or vice versa, is calculated as homogeneous small deformation for all the 24 variants belonging to the KS relation on the basis of the ordinary elastic theory. Prior to the transformation, three mutually perpendicular vectors are set up; the first vector perpendicular to the plane defining the KS relation, the second parallel to the direction defining the KS relation and the third perpendicular to the both vectors. Then they are deformed to the corresponding vectors in a transformed crystal. This strain is calculated as strain tensor, where lattice constants of 3.6458Å\(^{40}\) and 2.9019Å at 916°C are used for $\gamma$ phase and $\alpha$ phase respectively, by taking account of thermal dilation. This tensor is converted into the tensor on the coordinates based on the crystal principal axes. Stress tensor is also calculated by using this tensor and the following elastic constants\(^{39}\): $C_{11}=2.42$, $C_{12}=1.46$, and $C_{44}=1.12$ ($\times 10^{12}$ dyn/cm\(^2\)
on the crystallogical coordinates. The strain and stress tensors are obtained on the crystal coordinates of bcc and fcc; $E_i$ and $S_i$ are stress and strain of $i$ variant on bcc coordinates during $\alpha \rightarrow \gamma_i$ transformation and $E_j$ and $S_j$ are those of $j$ variant on fcc coordinates during $\gamma \rightarrow \alpha_j$ transformation, where $i$ and $j = 1 \sim 24$. Further they are converted into the tensors on the coordinates of a specimen, assuming that a sheet specimen is made of a single crystal in which the normal direction to the sheet surface is $X$ axis, the rolling direction $Y$ axis and the perpendicular direction to them $Z$ axis. They are designated as $E_a$, $S_a$, $E_j$ and $S_j$. Elastic works are expressed by the following equations:

$$W_a = \frac{1}{2} \sum_{n=x}^{m=x} (E_a)_{nm} (S_a)_{nm} \quad (1)$$

$$W_{ij} = \frac{1}{2} \sum_{n=x}^{m=x} (E_i)_{nm} (S_j)_{nm} \quad (2)$$

Where, 24 orientations are obtained in $\gamma$ phase from one orientation of $\alpha$ phase alone. Accordingly after $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation there are theoretically 576 orientations in $\alpha$ phase, starting from one orientation in $\alpha$ phase. Then $W_{ij}$ value in Eq. (2) expresses the elastic works of 576 kinds ($i=1 \sim 24$, $j=1 \sim 24$). The ratios of the work in normal direction to total work, $R_a$ and $R_{ij}$ values are given by:

$$R_a = \frac{1}{2} \sum_{n=x}^{m=x} (E_a)_{xx} (S_a)_{xx} W_a \quad (3)$$

$$R_{ij} = \frac{1}{2} \sum_{n=x}^{m=x} (E_i)_{xx} (S_j)_{xx} W_{ij} \quad (4)$$

Figure 5 shows a three-dimensional change in $R_a$ and $R_{ij}$ values with an initial orientation in $\alpha$ phase. They are plotted on the stereographic triangle. In this figure the curved surfaces with maximum and minimum values of $R_a$ and $R_{ij}$ are shown in relation to initial orientations in $\alpha$ phase. The maximum and minimum values of $R_a$ change remarkably with the initial orientation, the maximum value being small in the vicinity of the (111) initial orientation. The minimum value of $R_a$ is almost zero if the initial orientation is in the range of 30~40 deg off (100) orientation and it increases with approaching (100) orientation. According to the assumption, variants with large $R_a$ value (for example $R_a \geq 15 \%$) transform preferentially in the surface layer in slowly heated specimens. Even if prior $\alpha$ phase has a preferred orientation in [111] to some extent, grains with the orientations {100}, {110} and their neigh-

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**Fig. 5.**
Effect of initial orientation on ratio of elastic work in normal direction to total elastic work performed during $\alpha \rightarrow \gamma_i (R_a)$ or $\gamma_i \rightarrow \alpha_j (R_{ij})$ transformation; maximum and minimum values for $R_a$ or $R_{ij}$ are shown by curved surfaces on stereographic triangle.
hobs transform in preference to the grains with \{111\}. As the result, unless prior a phase has very strong \{111\} orientation, it has the same texture as that having only the initial components of \{100\}, \{110\} and their neighbours. This means that even if there is some difference in the initial texture, \(\gamma \rightarrow \alpha\) transformed texture shows nearly the same one.

On the other hand, in rapidly heated specimens every variant transforms in the equal probability independent of its \(R_s^\gamma\) value even in the surface layer. Therefore, the initial texture ought to give a large effect on the transformation texture compared with specimens slowly heated. In the case of rapid heating or in the midsection of thick specimen, texture tends to be random through the transformation caused by the operation of many variants.

The above explanation was mainly for the \(\alpha \rightarrow \gamma\) transformation. During \(\gamma \rightarrow \alpha\) transformation the selective transformation also occurs in slowly cooled specimens. The \(R_s^\gamma\) value of each variant in \(\gamma \rightarrow \alpha\) transformation can be shown on a stereographic triangle of fcc for each orientation in \(\gamma\) phase as Fig. 5 (a). Here, Fig. 5 (b) shows the \(R_s^\gamma\) values of variants during \(\gamma \rightarrow \alpha\) transformation, starting from 24 kinds of orientations of \(\gamma\) phase which are transformed from each orientation of \(\alpha\) phase on the stereographic triangle of bcc. In this figure the maximum value of \(R_s^\gamma\) is somewhat small near \{111\} orientation, and the minimum value of \(R_s^\gamma\) is almost zero in the whole range of the initial orientation in \(\alpha\) phase. Observed change in the value of \(R_s^\gamma\) with the initial orientation in \(\alpha\) phase is small, reflecting a number of variants involved. Figure 5 (b) is constructed on the assumption that during \(\alpha \rightarrow \gamma\) transformation all the 24 orientations are formed from each initial orientation of \(\alpha\) phase in equal probability. It is very difficult to show all orientations of a final (transformed) \(\alpha\) phase exhibiting each initial \(\alpha\) orientation and its \(R_s^\gamma\) value during the \(\alpha \rightarrow \gamma\) transformation on the same stereographic triangle. However, \(\alpha \rightarrow \gamma \rightarrow \alpha\) transformation texture can be predicted as a pole figure by considering the selection mechanism of variants operated during \(\alpha \rightarrow \gamma \rightarrow \alpha\) transformation starting from every orientation composing the initial texture in \(\alpha\) phase. Therefore, by limiting the number of orientations in the initial \(\alpha\) phase, and taking into account the selection rule of variants, the texture formed after \(\alpha \rightarrow \gamma \rightarrow \alpha\) transformation is calculated. This is shown in the following section.

3. Simulation of \(\alpha \rightarrow \gamma \rightarrow \alpha\) Transformation Texture

As a model of an initial texture in \(\alpha\) phase the recrystallization texture is used which was obtained in the specimen annealed at 890 °C, shown in Fig. 4. Considering the experimental result that this texture consists of \{111\}, \{100\}, \{110\} and \{211\} with the ratio of \(4:2:1:1\), and the location of accumulation and the symmetry in the pole figure, this texture is assumed to be represented by the following nine orientations: \(\{111\} [011], \{111\} [011], \{111\} [211], \{111\} [211],\ 100\} [011], \{101\} [011], \{110\} [111], \{755\} [011] and \{755\} [011], where the amount of cumulation of each orientation is one in the pole figure but \{100\} [010] orientation is twice as strong as others. Starting with these initial orientations, the texture after \(\alpha \rightarrow \gamma \rightarrow \alpha\) transformation which obeys the KS orientation relationship and the selection rule of variants, is computer-simulated. For the calculation, the variants with the larger values of \(R_s^\gamma\) and \(R_s^\alpha\) than certain threshold values are assumed to operate. Their value is zero (%) for rapid transformation rate, and is 35 (%) for slow transformation rate, which is about one half of the maximum value. The \(200\) pole figures thus obtained are shown in Fig. 6; the initial texture (a), the \(\alpha \rightarrow \gamma \rightarrow \alpha\) transformation texture under the condition of slow heating and slow cooling (b), rapid heating and slow cooling (c), and rapid heating and rapid cooling (d).

The threshold values of variants are shown in the figure. In these figures the amount of cumulation is shown by the number of each orientation which is one for one initial orientation in (a), and by the root of the ratio of the number of each orientation to its number in the texture calculated as a random distribution of an orientation in (b), (c) and (d). These three kinds of pole figures are qualitatively in good agreement with the experimental results. That is, the initial specimen with \{111\} texture has:

i) \{100\} transformation texture after slow heating and slow cooling as shown in Fig. 6 (b) which corresponds to Fig. 2 (d) or Fig. 4 (a),

ii) the transformation texture with components of \{110\} and 20~30 deg off \{100\} orientation after rapid heating and slow cooling as shown in Fig. 6 (c) which corresponds to Fig. 2 (b) or Fig. 4 (b),

iii) the weak transformation texture with components

Fig. 6. (222) pole figures showing simulated initial texture; (a), transformation textures; (b) slow heating and slow cooling; (c) rapid heating and slow cooling; (d) rapid heating and rapid cooling.
of {111} and {100} orientations after rapid heating and rapid cooling as shown in Fig. 6 (d) which corresponds to Fig. 2 (a).

As is clear from the comparison, there is a very good accordance in the texture between experimental results and computer-simulation.

Therefore, the theory proposed by the present authors has demonstrated its validity in the most popular case of sheet steel with {111} preferred orientation. The transformation texture in the specimen with the initial texture of {100} or {110} has been further confirmed to be well explained by this theory in both cases of rapid and slow transformation rates, which will be shown in the following paper.

There has been no consolidated theory which can explain various experimental results on the transformation texture. This is because heating and cooling rates were not intentionally and sufficiently controlled during transformation and the relation between initial texture and transformation texture was not examined. For the first time, the present authors examined the effects of heating and cooling rates and initial texture on a transformation texture. This is because heating and subsequent slow cooling showed a texture with components {110} and the orientation shifted 20~30 deg away from {100}.

Slow heating and subsequent slow cooling produced a texture with distinct {100} component in the surface layer.

(4) There was a marked difference in texture between surface and midsection of thick specimen. In the midsection, the similar texture as described in (i) was formed, independent of the heating and cooling rates. While, the specimen surface exhibited the texture as mentioned in (1) to (3), depending on each transformation condition.

To explain these results a new theory on the formation mechanism of transformation texture was proposed, which was based on the following two assumptions; (i) the orientation relationship between bcc and fcc obeys the Kurdjumov–Sachs relation during $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation, and (ii) the orientation change proceeds by the preferential operation of such variants that have larger values of $R_\alpha$ and $R_\gamma$, ratios of an elastic work in the sheet normal direction to total elastic work, during a slow progress of $\alpha \rightarrow \gamma$ and $\gamma \rightarrow \alpha$ transformations in the surface layer of a specimen, respectively. Whereas, every variant transforms equivalently independent of its $R_\alpha$ or $R_\gamma$ value in the case of a rapid progress of transformation in the surface layer or regardless of heating and cooling rate in the midsection of thick specimen. The textures computer-simulated by this theory were shown to be in good agreement with the textures obtained in the experiments.

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