Transformation Textures in Ti-15V-3Cr-3Sn-3Al Alloy Sheets

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Cold rolling and recrystallization textures of the metastable β phase (bcc) and transformation textures of the α phase (hcp) precipitated during aging from the β matrices having these textures in a Ti-15V-3Cr-3Sn-3Al alloy were investigated in conjunction with their microstructures. Furthermore, the measured transformation textures were compared with the α phase textures predicted theoretically from the parent β phase textures using the crystallite orientation distribution function to discuss the mechanism of texture formation caused by the β to α transformation. In the sheets cold rolled after solution treatment, a $\langle 110 \rangle // \text{RD}$ partial fiber texture showing the maximum intensity at $\{001\} \langle 110 \rangle$ was developed at high reductions. The recrystallization textures obtained by annealing from this rolling texture consisted mainly of the $\{554\} \langle 225 \rangle$ orientation. Aging at 773 K after recrystallization resulted in the formation of somewhat coarse plate-like α precipitates at grain boundaries rather than inside grains. In this case, the quantitative agreement between the measured and predicted transformation textures were found except for the basal orientation, indicating that all variants obeying Burgers' orientation relationship appeared with nearly the same probability. On the other hand, fine needle-like α precipitates were observed in materials aged at 773 K directly from the cold rolled state. The texture of the precipitated α phase was composed nearly of orientations derived from the β phase rolling texture through Burgers' relationship. However, the results of quantitative texture analysis suggested that the variants which formed the $(\bar{1}2\bar{1}0)[10\bar{1}0]$ orientation from the main orientation $\{001\} \langle 110 \rangle$ in the rolling texture were preferentially selected in the precipitation from the deformed matrix. This variant selection can be explained by taking active slip systems operating during cold rolling into account.

(Received October 14, 1991)

Keywords: beta titanium alloy, transformation texture, cold rolling, recrystallization, precipitation, Burgers' relationship, variant selection, orientation distribution function

I. Introduction

A Ti-15V-3Cr-3Sn-3Al alloy which is well known as one of the typical β titanium alloys has excellent characteristics in cold workability, formability, age hardenability, strength and toughness. Consequently, this alloy can be cold rolled up to considerably high reductions in the single β phase condition after solution treatment, and it is usually aged to obtain high strength in the final process. In the case of β titanium alloys, the precipitation of α phase with hcp structure from the metastable β phase matrix with bcc structure takes place during aging. Since materials before aging are considered to have deformation or recrystallization textures developed by hot rolling, cold rolling and annealing, relatively strong textures in the precipitated α phase will be formed after aging. Therefore, the textures in the α phase as well as those in the β matrix might fairly affect the anisotropy of mechanical properties because of a lower crystal symmetry of the α phase, if volume fractions of the α phase are not so low. The transformation textures of the α phase in this alloy must be studied along with the β phase textures to improve the properties through texture con-

trol. However, there are only a few papers on the transformation textures in β titanium alloys including this alloy⁽¹⁾⁽²⁾.

In general, the analysis of transformation textures in titanium and its alloys has been made by the conventional pole figure method⁽¹⁾⁽³⁾, but it is difficult to obtain the exact and quantitative information on the orientation distribution by this method. Recently, quantitative texture analysis by the crystallite orientation distribution function has been vigorously carried out for transformation textures⁽²⁾⁽⁴⁾⁻⁽⁶⁾ and hot rolling textures associated with transformation⁽⁷⁾⁻⁽⁹⁾ in titanium and its alloys. Furthermore, if this analytical technique is used, transformation textures can be predicted from the parent phase textures by using the known crystallographic orientation relationship between the parent and product phases⁽¹⁰⁾⁽¹¹⁾. In the previous papers⁽⁴⁾⁽⁵⁾, the prediction of transformation textures in a commercially pure titanium sheet was done using this method and the obtained results helped to explain the formation mechanism of the textures.

In the present work, the transformation textures of α phase precipitated by aging in Ti-15V-3Cr-3Sn-3Al alloy sheets which had cold rolling textures or recrystallization textures in the β parent phase have been investigated in conjunction with the microstructures of the aged specimens. In addition, the measured textures have been quantitatively compared with the textures predicted by the above method, in order to discuss the formation mechanism of the transformation textures in this alloy.

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II. Experimental Procedure

The starting material employed in this investigation was Ti-15V-3Cr-3Sn-3Al alloy plates with a thickness of 5 mm, which was produced through hot rolling at 1223 K and subsequent solution treatment at 1073 K for 900 s. The chemical composition of this alloy is shown in Table 1. As it exhibited an inhomogeneous structure consisting of relatively equiaxed grains and significantly elongated grains, it was solution heat treated again at temperatures of 1073 K and 1223 K for 1.8 ks and water quenched. Their mean grain sizes were about 100 μm for solution treatment at 1073 K and 180 μm for that at 1223 K, respectively. These plates were cold rolled up to 90% reduction in thickness at a reduction of about 10% per pass. In order to obtain the recrystallized structure, 90% cold rolled sheets were annealed at 1073 K and 1223 K in a vacuum and air cooled. The holding time at 1073 K was 2.7 ks and that at 1223 K was 0 s. The latter means that specimens were heated up to 1223 K at a heating rate of 3.2 K/s and cooled immediately. Under such annealing conditions, the recrystallized grains with nearly the same grain size (80 μm) were obtained in all the specimens. Both as cold rolled and recrystallized sheets were aged in a vacuum at 773 K for 4.8 ks and 36 ks, respectively and air cooled. These aging times correspond with the times indicating the maximum hardness in the isothermal aging curves.

The samples for texture measurements were prepared from the central sections of the sheets by thinning mechanically and chemically. Complete pole figures of $\{110\}$, $\{200\}$ and $\{211\}$ in the β phase and those of $\{10\bar{1}0\}$, $\{10\bar{1}2\}$ and $\{11\bar{2}0\}$ in the α phase were measured using $\text{CuK}\alpha$ radiation for the back-reflection method and $\text{MoK}\alpha$ radiation for the transmission one. The selection of these pole figures in the α phase was based on the fact that (0002) and $\{10\bar{1}1\}$ reflections were close to the $\{110\}$ reflection of the untransformed β matrix. From these pole figure data, crystallite orientation distribution functions (ODF) were calculated using the series expansion method proposed by Bunge⁽¹⁰⁾⁽¹²⁾. The series expansion of the ODF in generalized spherical harmonics was performed up to $l_{\text{max}}=22$ for β phase (bcc) and $l_{\text{max}}=14$ for α phase (hcp). The ODF's of the β phase were then used to predict theoretically those of the α phase by using Burgers' orientation relationship⁽¹³⁾ with no variant selection. The predicted ODF was determined up to the same order $l_{\text{max}}=14$ as the experimental ODF.

The microstructures of aged specimens were observed by an optical microscope on the longitudinal sections perpendicular to the transverse direction. To examine the

morphology of the α phase precipitates and the orientation relationship between the β matrix and them, transmission electron microscopic observations were carried out by a JEOL JEM-2000FX electron microscope operating at 200 kV. The thin-foil samples parallel to the rolling plane corresponding to the central region of the sheet were used in this observation.

III. Results

1. Cold rolling and recrystallization textures in β phase

The starting material had a very strong $\langle 110 \rangle // \text{RD}$ partial fiber texture composed of the orientations between $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$, since it was produced through hot rolling. Figure 1 shows $\varphi_2=45^\circ$ sections of the ODF's representing the evolution process of cold rolling textures for the materials solution treated at 1073 and 1223 K. The reason why the $\varphi_2=45^\circ$ section was used to represent the β phase textures is due to the fact that the orientations with relatively high density were almost included in this section. In the materials before cold rolling, different textures are observed depending on the temperature of solution treatment. The material solution treated at 1073 K has the same texture as the starting material except that the former texture is weaker than the latter one, while the material solution treated at 1223 K has a considerably different texture. This is relatively similar to the recrystallization texture shown in Fig. 2. For the material solution treated at 1073 K, a quantitative change in texture with increasing rolling reduction is observed, though the preferred orientation is not so changed. In particular, both the $\langle 110 \rangle // \text{RD}$ partial fiber component having a maximum density near $\{112\}\langle 110 \rangle$ and the $\langle 111 \rangle // \text{ND}$ fiber component are formed after 50% cold rolling. These components are quite similar to the textural components observed in low carbon steel sheets⁽¹⁴⁾. However, the $\langle 110 \rangle // \text{RD}$ partial fiber texture in which the main orientation is $\{001\}\langle 110 \rangle$ is developed again at reductions above 70%. For the material solution treated at 1223 K, the main component of texture is greatly changed as the rolling reduction increases. After 90% cold rolling, the same texture as the material solution treated at 1073 K is obtained in this material, in spite of a marked difference in the textures before cold rolling. The rolling textures at 90% reduction are, however, somewhat different with respect to orientation density.

The recrystallization textures of the 90% cold rolled sheets annealed at 1073 and 1223 K are shown in Fig. 2. These textures consist mainly of a $\{554\}\langle 225 \rangle$ orientation close to $\{111\}\langle 112 \rangle$, irrespective of the temperatures of solution treatment and annealing. Although this orientation shows the highest density in the specimen solution treated at 1073 K and annealed at 1073 K, there is almost no remarkable difference among these textures.

Table 1 Chemical composition of the Ti-15V-3Cr-3Sn-3Al alloy used (mass%).

V	Cr	Sn	Al	Fe	O	N	H	Ti
14.77	2.89	2.84	2.97	0.160	0.114	0.0064	0.0064	Bal.

Fig. 1 $\phi_2=45^\circ$ sections of the orientation distribution functions for the sheets solution treated at 1073 and 1223 K and cold rolled to various reductions.

Fig. 2 $\phi_2=45^\circ$ sections of the orientation distribution functions showing recrystallization textures in the sheets solution treated at 1073 and 1223 K, 90% cold rolled and annealed at 1073 and 1223 K.

2. Transformation textures of precipitated α phase

Figure 3 shows the textures of α phase precipitated by aging at 773 K in the sheets having recrystallization textures given in Fig. 2. In the case of α phase, both $\phi_2=0^\circ$ and 30° sections were used to represent the ODF due to the same reason as the case of the β phase. For all specimens, the textures of the precipitated α phase, namely, transformation textures consist mainly of the following three components; the orientation tube from the $(\bar{1}\bar{2}\bar{1}5)[51056]$ main orientation with maximum density to $(\bar{1}\bar{2}\bar{1}5)[9543]$ via $(01\bar{1}3)[1211]$, the secondary orienta-

tion $(\bar{1}\bar{2}\bar{1}0)[10\bar{1}0]$, and the orientation tube between $(\bar{1}\bar{2}\bar{1}0)[10\bar{1}2]$ and $(01\bar{1}0)[2\bar{1}\bar{1}4]$ which is explained by the rotation around a common $[0001]$ axis. The difference in orientation density among four specimens for these textural components corresponds well to that for the above recrystallization textures. The basal component consisting mainly of $\{0001\}\langle 2\bar{1}\bar{1}0 \rangle$ is also observed except for the specimen solution treated at 1223 K and annealed at 1073 K, but it is comparatively weak. The textures for the α phase predicted from the β phase recrystallization textures before aging, assuming that all variants obeying Burgers' relationship appear with equiprobability, are shown in Fig. 4. When comparing these textures with the measured textures given in Fig. 3, the quantitatively good agreement between the predicted and measured textures is obtained in all the specimens except for the basal component. This suggests that variant selection has hardly taken place in the precipitation of α phase from the recrystallized β matrix.

Figure 5 shows the textures of α phase precipitated by aging at 773 K in the 90% cold rolled sheets having the $\langle 110 \rangle // \text{RD}$ partial fiber textures given in Fig. 1. The transformation textures of the materials solution treated at 1073 and 1223 K are similar to each other because of the similarity between the rolling textures before aging. The main components of these textures are mostly concentrated on the $\phi=90^\circ$ line of the $\phi_2=0^\circ$ section, and they are represented by $(\bar{1}\bar{2}\bar{1}0)[10\bar{1}0]$, $(\bar{1}\bar{2}\bar{1}0)[30\bar{3}2]$ and $(\bar{1}\bar{2}\bar{1}0)[0001]$ in order of the orientation density. The first orientation with the highest density in which a $[0001]$ axis is parallel to the transverse direction is accompanied with an orientation spread on the $\phi_1=0^\circ$ line, i.e., a $\langle 10\bar{1}0 \rangle$ fiber component. The textures for the α phase predicted from the β phase cold rolling textures before aging without variant selection according to Burgers' relationship are shown in Fig. 6. By comparing this figure with

Fig. 3 $\varphi_2=0^\circ$ and 30° sections of the orientation distribution functions of precipitated α phase showing transformation textures in the sheets solution treated at 1073 and 1223 K, 90% cold rolled, annealed at 1073 and 1223 K and aged for 36 ks at 773 K.

Fig. 4 $\varphi_2=0^\circ$ and 30° sections of the orientation distribution functions of α phase predicted from the β phase recrystallization textures shown in Fig. 2.

Fig. 5, it is found that the predicted textures are nearly consistent with the measured textures in respect to peak positions, whereas the quantitative agreement between them concerning peak heights is not obtained. In the predicted textures, a $[0001]\parallel\text{RD}$ fiber texture between $(\bar{1}2\bar{1}0)[0001]$ and $(01\bar{1}0)[0001]$ located at $\varphi_1=90^\circ$, $\phi=90^\circ$ on the $\varphi_2=30^\circ$ section is much stronger than the

$(\bar{1}2\bar{1}0)[10\bar{1}0]$ orientation which showed the maximum density in the measured textures. The $(\bar{1}2\bar{1}0)[30\bar{3}2]$ orientation as well as this orientation is weaker in the predicted textures. These results suggest that variant selection has taken place in the precipitation of the α phase from the deformed β matrix, though the textures of the α phase are composed nearly of orientations obeying Burgers'

Fig. 5 $\varphi_2=0^\circ$ and 30° sections of the orientation distribution functions of precipitated α phase showing transformation textures in the sheets solution treated at 1073 and 1223 K, 90% cold rolled and aged for 4.8 ks at 773 K.

relationship. In particular, the variants forming the $(\bar{1}2\bar{1}0)[10\bar{1}0]$ orientation seem to have been preferentially selected.

3. Microstructures of aged materials

Figure 7 shows microstructures observed on the longitudinal sections of the specimens aged at 773 K after recrystallization (a), (b) and after cold rolling (c). The precipitation of α phase from the recrystallized structure occurs preferentially at grain boundaries rather than inside grains, as seen in Fig. 7(a) and (b). Such a tendency is more pronounced in the specimens solution treated at 1223 K than in the specimens solution treated at 1073 K. In the former, large precipitation-free ranges within β grains exist mainly in the central region of the sheet. The α phase precipitates in recrystallized specimens can be thus observed by an optical microscope, but those in cold rolled specimens are too small to be observed as shown in Fig. 7(c). However, it is considered that the α phase precipitates are easy to be formed in regions of higher dislocation density such as shear bands observed as black

Fig. 6 $\varphi_2=0^\circ$ and 30° sections of the orientation distribution functions of α phase predicted from the β phase rolling textures of the 90% cold rolled sheets shown in Fig. 1.

striations in this photograph.

Transmission electron micrographs and selected area diffraction patterns at an early stage of precipitation for the specimen aged at 773 K after recrystallization are shown in Fig. 8. In Fig. 8(a), plate-like precipitates are observed inside a β grain. Since the precipitates marked with a circle indicate a $(\bar{1}2\bar{1}0)[10\bar{1}0]$ orientation from the diffraction pattern, the longitudinal direction of the precipitates might be parallel to the $[21\bar{3}0]$ direction rather than the $[0001]$ direction, assuming that the habit plane of the plate-like precipitates is parallel to a $(\bar{4}5\bar{1}0)$ plane as described by Rhodes and Williams⁽¹⁵⁾. The transverse direction of the plate-like precipitates is, however, considered as the $[0001]$ direction. When the precipitation of α phase occurs obeying Burgers' relationship, the $[21\bar{3}0]$ direction becomes nearly parallel to the $\langle 322 \rangle$ direction in the β matrix. Regularly arranged plate-like precipitates formed at a grain boundary are observed in Fig. 8(b). From the diffraction pattern, it is found that Burgers' relationship holds within 5° (0.087rad) between these precipitates and the β matrix. In this case, it is also reasonable to regard the longitudinal direction of the precipitates as the $[1\bar{3}20]$ direction from the result of

Fig. 7 Optical micrographs of the aged materials. (a) Solution treated at 1073 K, 90% cold rolled, annealed at 1073 K and aged for 36 ks at 773 K. (b) Solution treated at 1223 K, 90% cold rolled, annealed at 1223 K and aged for 36 ks at 773 K. (c) Solution treated at 1223 K, 90% cold rolled and aged for 4.8 ks at 773 K.

Fig. 8 Transmission electron micrographs and SAD patterns of the specimen solution treated at 1073 K, 90% cold rolled, annealed at 1073 K and aged for 7.2 ks at 773 K.

analysis of the diffraction pattern. In general, the α phase precipitates formed from the recrystallized β matrix are relatively coarse.

A typical microstructure of the specimen aged for 4.8 ks at 773 K after 90% cold rolling is shown in Fig. 9. Numerous fine needle-like precipitates are distributed uniformly throughout the specimen in comparison with the precipitation from the recrystallized matrix. However, more precipitates are likely to exist in regions of higher dislocation density from the results of some observations.

IV. Discussion

It is obvious from the experimental results that the amount of strain in the β matrix before aging largely influences the formation of the α phase textures and the morphology and distribution of the α phase precipitates. Therefore, the formation mechanism of transformation textures in this alloy is discussed separately in each case that the β phase before aging is in recrystallized state or in cold rolled state.

1. α phase transformation textures formed from β phase recrystallization textures

Recrystallization textures of β phase having bcc structure are composed mainly of a $\{554\}\langle 225 \rangle$ orientation as seen in Fig. 2. In general, the main components of recrystallization textures in low carbon steels are

Fig. 9 Transmission electron micrograph of the specimen solution treated at 1073 K, 90% cold rolled and aged for 4.8 ks at 773 K.

$\{111\}\langle 110 \rangle$ and $\{554\}\langle 225 \rangle$, and it has been considered that these orientations were nucleated preferentially near grain boundaries from $\{111\}\langle 112 \rangle$ and $\{112\}\langle 110 \rangle$, respectively⁽¹⁴⁾. In this case, the $\{111\}\langle 110 \rangle$ orientation is far weaker than the $\{554\}\langle 225 \rangle$ orientation, because the $\{111\}\langle 112 \rangle$ orientation necessary to form $\{111\}\langle 110 \rangle$ is very weak in rolling textures observed after 90% cold rolling.

As already described, the textures of α phase derived by aging at 773 K from such recrystallization textures almost satisfy Burgers' relationship without variant selection, but the $\{0001\}\langle 2\bar{1}\bar{1}0 \rangle$ basal orientation which

cannot be obtained from the recrystallization textures through Burgers' relationship is somewhat observed. In the paper of Rath *et al.*⁽²⁾, the basal orientation has been also observed as a weak component in the experimental α phase ODF of a Ti-3Al-8V-6Cr-4Mo-4Zr alloy sheet aged at 783 K after annealing at 1089 K, though it has not in the predicted α phase ODF. Nevertheless, they have not described the difference between these ODF's. Two reasons can be considered for the appearance of this orientation. In the first place, if this orientation was formed through Burgers' relationship, a parent orientation in the β phase would be probably $\{110\}\langle 001\rangle$. Although this is almost absent in recrystallization textures, there is the possibility of existence of a few $\{110\}\langle 001\rangle$ oriented grains in the recrystallized structure, for this orientation is related with the $\{554\}\langle 225\rangle$ main orientation by the 30° (0.524rad) rotation about a $\langle 110\rangle$ axis parallel to the transverse direction. Moreover, since grain boundaries are easy to become the preferential site of precipitation, the $\{0001\}\langle 2\bar{1}\bar{1}0\rangle$ oriented precipitates may be formed preferentially at the boundaries of $\{110\}\langle 001\rangle$ grains surrounded by grains with other orientations. However, the fact that the formation of $\{0001\}\langle 2\bar{1}\bar{1}0\rangle$ from $\{110\}\langle 001\rangle$ must be accompanied with variant selection is in conflict with the experimental results indicating no variant selection for textural components other than the basal orientation. Secondly, the existence of the α phase precipitates disobeying Burgers' relationship is considered. Rhodes and Williams⁽¹⁵⁾ have reported that two distinct types of α phase consisting of Type 1 α obeying Burgers' relationship and Type 2 α disobeying it were observed in the precipitation process of metastable β phase titanium alloys, and Type 2 α precipitates were formed after longer aging times and/or at higher aging temperatures, compared with Type 1 α . The texture of the α phase precipitated during aging at 873 K from the recrystallized β matrix is shown in Fig. 10. This texture is considerably different from the transformation textures obtained by aging at 773 K, but it consists of the orientations shifted about 15° (0.262rad) from the textural components shown in Fig. 3. This reveals that Type 2 α precipitates are formed at higher aging temperatures and have some relation with Type 1 α . In Fig. 10, the orientations of Type 2 α which seem to have been shifted from the $\{0001\}\langle 2\bar{1}\bar{1}0\rangle$ orienta-

tion are observed. This fact reduces the possibility of existence of Type 2 α precipitates having the $\{0001\}\langle 2\bar{1}\bar{1}0\rangle$ orientation in the specimens aged at 773 K. In this way, it is difficult to clarify the formation mechanism of the basal component in the present work. However, except for the basal component, it can be concluded that the transformation textures of α phase formed by aging at 773 K from the β phase recrystallization textures consist of the orientations obeying Burgers' relationship and no variant selection has taken place in this transformation process.

2. α phase transformation textures formed from β phase cold rolling textures

The textures of α phase precipitated in the as cold rolled matrix coincide qualitatively but do not coincide quantitatively with the textures predicted from the rolling textures before aging through Burgers' relationship. This implies that the formation of the α phase which satisfy Burgers' relationship with respect to the β matrix is accompanied with variant selection. The comparison between Fig. 5 and Fig. 6 especially indicates the preferential selection of variants forming $(\bar{1}\bar{2}\bar{1}0)[10\bar{1}0]$ and $(\bar{1}\bar{2}\bar{1}0)[30\bar{3}2]$ orientations. The formation mechanism of the $(\bar{1}\bar{2}\bar{1}0)[10\bar{1}0]$ orientation which has the highest density in the transformation textures is now discussed in detail.

The β phase rolling textures observed in the 90% cold rolled sheets before aging consist mainly of a $\langle 110\rangle//\text{RD}$ partial fiber texture which can be described by three ideal orientations $\{001\}\langle 110\rangle$, $\{112\}\langle 110\rangle$ and $\{111\}\langle 110\rangle$. Figure 11 shows the α phase textures predicted from these ideal orientations with the Gaussian distribution⁽¹⁰⁾ without variant selection using Burgers' relationship. It is found from this figure that the $(\bar{1}\bar{2}\bar{1}0)[10\bar{1}0]$ orientation is included in the α phase orientations inherited from $\{001\}\langle 110\rangle$ which is the strongest component in the rolling textures. Therefore the strong $(\bar{1}\bar{2}\bar{1}0)[10\bar{1}0]$ orientation is probably formed through variant selection from the $\{001\}\langle 110\rangle$ orientation in the deformed β matrix. The mechanism of the variant selection is shown in Fig. 12. The left-hand micrograph was taken from three sections perpendicular to the rolling, normal and transverse directions for the specimen aged at 773 K after 90% cold rolling. On the section perpendicular to TD, shear bands and slip bands inclined at about 35° (0.611rad) to RD are observed over wide ranges. They are formed during cold rolling and exhibit many striations nearly parallel to TD on the sections perpendicular to RD and ND. These bands can be accordingly regarded as the trace of the slip plane with an angle of 35° (0.611rad) to RD, as shown in the right-hand illustration. If the deformed β matrix before aging has a $(001)[110]$ orientation, this slip plane will be a $\{112\}$ plane. As the precipitation of α phase in the deformed matrix generally occurs at dislocations, it is considered that slip deformation during rolling greatly influences variant selection. Dillamore and Roberts⁽¹⁶⁾ have assumed that a biaxial stress system composed of the compressive stress σ parallel to the rolling plane normal and the tensile stress $-\sigma$ parallel to the rolling direction

Fig. 10 $\phi_2=0^\circ$ and 30° sections of the orientation distribution function of precipitated α phase observed in the sheets solution treated at 1073 K, 90% cold rolled, annealed at 1073 K and aged for 18 ks at 873 K.

V. Conclusions

Transformation textures of the α phase precipitated by aging from the β matrix after cold rolling and from that after recrystallization in a Ti-15V-3Cr-3Sn-3Al alloy have been investigated by comparing the experimental texture with the theoretically predicted texture using crystallite orientation distribution functions. The main conclusions obtained in the present study are as follows:

(1) Rolling textures in 90% cold rolled sheets consist of a $\langle 110 \rangle // \text{RD}$ partial fiber texture between $\{001\} \langle 110 \rangle$ and $\{111\} \langle 110 \rangle$, irrespective of the temperature of solution treatment before rolling. Recrystallization textures developed from these rolling textures through annealing have the $\{554\} \langle 225 \rangle$ main orientation, and almost no effect of annealing temperature on the recrystallization textures is observed under the applied condition.

(2) Aging at 773 K after recrystallization leads to the formation of relatively coarse plate-like α precipitates at grain boundaries rather than within grains. The textures of the precipitated α phase coincide well with transformation textures predicted from the recrystallization textures of the β phase through Burgers' relationship, except for the basal orientation. This indicates that variant selection has hardly taken place in the precipitation of the α phase from the recrystallized matrix.

(3) By aging at 773K after 90% cold rolling, fine needle-like α precipitates are formed at dislocations. The textures of the precipitated α phase are composed mostly of the orientations obeying Burgers' relationship, but are quantitatively inconsistent with transformation textures predicted from the rolling textures of the β phase because

of the occurrence of variant selection. The preferential formation of the $(\bar{1}2\bar{1}0)[10\bar{1}0]$ orientation can be explained by considering active slip systems in cold rolling.

Acknowledgment

The authors wish to express their appreciation to Iron & Steel Research Laboratories, Kobe Steel Ltd. for preparing rolled sheets.

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