Optimum rolling ratio for obtaining {001}<110> recrystallization texture in Ti-Nb-Al biomedical shape memory alloy

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Abstract

The rolling rate (r) dependence of textures was investigated in the Ti-26Nb-3Al (mol%) alloy to reveal the conditions required to form the $\{001\}\langle 110 \rangle$ recrystallization texture, which is a desirable orientation for the β -titanium shape memory alloy. $\{001\}\langle 110 \rangle$ was the dominant cold-rolling texture when $r = 90\%$ and it was transferred to the recrystallization texture without forming $\{112\}\langle 110 \rangle$, which is detrimental for the isotropic mechanical properties of the rolled sheet. A further increase in r resulted in the formation of $\{112\}\langle 110 \rangle$ in both rolling and recrystallization textures. Therefore, r should be controlled to form only the $\{001\}\langle 110 \rangle$ rolling texture, because the $\{112\}\langle 110 \rangle$ texture can overwhelm the $\{001\}\langle 110 \rangle$ texture during recrystallization.

Key words: texture, metallic biomaterials, shape memory alloy, superelasticity, stress shielding, titanium alloy

1. Introduction

β -titanium shape memory alloys (SMAs) are potential Ni-free alternatives to Ti-Ni in medical applications and exhibit better cytocompatibility, cold-workability and low Young's modulus of the parent phase compared to Ti-Ni [1, 2]. The shape memory effect and superelasticity appear in various β -titanium alloys owing to the β (bcc) to α'' (C-orthorhombic [3]) thermoelastic martensitic transformation [4-17]. The lattice deformation strain that determines the crystallographic maximum of the recovery strain in α'' -martensite is generally smaller than that of Ti-Ni (10.5%) and is not sufficient for practical use, even if the lattice parameters are optimized to increase the lattice deformation strain. We have proposed that the formation of a strong texture is crucial for improving the shape memory properties of β -titanium SMAs [18, 19]. By the formation of a strong texture, the recovery strain of β -titanium alloys becomes 3~6% that is mid value of R-phase (1%) and B19' martensite (10%) in Ti-Ni[20]. In addition, texture control can also decrease the Young's modulus of β -titanium biomedical alloys [21]. Extremely low Young's modulus about 30GPa that is required for the implant metallic materials for hard tissues is available in Ti-Nb based alloys without superelasticity that is detrimental for the fatigue life in general [20]. The relationships among thermomechanical treatment, texture, and mechanical properties have, therefore, been investigated in some β -titanium alloy systems. In cold-rolled β -titanium alloys, $\{001\}\langle 110\rangle$, $\{112\}\langle 110\rangle$, $\{111\}\langle 110\rangle$, and $\{111\}\langle 112\rangle$ are common cold-rolling textures, and $\{001\}\langle 110\rangle$ and $\{112\}\langle 110\rangle$ are common recrystallization textures [19, 22-31]. Here, $\{hkl\}\langle uvw\rangle$ means that the normal plane (ND) is $\{hkl\}$ and the rolling direction (RD) is $\langle uvw\rangle$. The transverse direction (TD) is defined as the direction perpendicular to both RD and ND. We have investigated Ti-Nb-Al alloys as a model

material [27, 29]. In Ti-24Nb-3Al (mol%), the rolling texture obtained by cold-rolling with a rolling rate (r) of 99% is $\{001\}\langle 110 \rangle$ with a weak $\{111\}\langle 112 \rangle$ texture [30]. A well-developed $\{112\}\langle 110 \rangle$ recrystallization texture is formed by annealing at 1273 K and the superelasticity is dramatically improved, especially along the RD [18, 19]. However, the transformation strain along the TD is only one third of that along the RD; strong anisotropy in the mechanical properties appears in $\{112\}\langle 110 \rangle$. In contrast, the recovery strain in $\{001\}\langle 110 \rangle$, which is the major orientation of the rolling texture, is almost isotropic within the ND plane and the transformation strains along the RD and TD reach the crystallographic limit, owing to the crystallography of the transformation and anisotropy of Young's modulus [21, 29, 32]. In addition, the Young's modulus is minimum along $\langle 100 \rangle$ and is lying on the ND plane in $\{001\}\langle 110 \rangle$; extremely low Young's modulus (~ 30 GPa) that is required for the implant metallic materials for hard tissues is possible without superelasticity [21]. The stress-induced martensitic transformation hardly occurs in the as-rolled material because a large amount of dislocation disrupts the transformation [30]. The dislocation density can be reduced by recovery at 873 K while maintaining the $\{001\}\langle 110 \rangle$ orientation. However, the superelasticity is degraded by the remaining dislocations and α -phase precipitates [33]. The formation of α -phase also increases the Young's modulus and is not favored for the implant materials for the hard tissues. The strong $\{001\}\langle 110 \rangle$ texture that is formed by recrystallization is, therefore, crucial. A short annealing time for the as-rolled ($r = 99\%$) material at 1123 K produced recrystallized $\{001\}\langle 110 \rangle$ texture, although a large amount of the $\{112\}\langle 110 \rangle$ recrystallization texture was also formed [27]. It is necessary to allow the recrystallization and growth of only the $\{001\}\langle 110 \rangle$ orientation and suppress the growth of the other orientations.

The r dependence of the rolling texture was examined in Ti-26Nb-3Al (mol%) in our previous study [34]. The major orientations were $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$ at $r = 99\%$, whereas $\{001\}\langle 110 \rangle$ was the dominant orientation at $r \sim 90\%$. The $\{001\}\langle 110 \rangle$ rolling texture is the most difficult to recrystallize in the $\{hkl\}\langle 110 \rangle$ orientations (α -fiber) in bcc metals [35-37]; the growth of $\{001\}\langle 110 \rangle$ is overwhelmed by the other $\{hkl\}\langle 110 \rangle$ orientations during recrystallization. Therefore, the strong $\{001\}\langle 110 \rangle$ recrystallization texture could be formed by optimizing the rolling rate to around 90% to prevent other $\{hkl\}\langle 110 \rangle$ rolling textures from forming. The objective of this study is to find a criterion for obtaining the $\{001\}\langle 110 \rangle$ recrystallization texture by clarifying the effect of r on rolling and recrystallization textures in Ti-26Nb-3Al (mol%).

2. Experimental procedures

An ingot with a nominal composition of Ti-26 mol% Nb-3 mol% Al was fabricated by Ar arc melting with a non-consumable W-electrode in an Ar-1% H₂ reducing atmosphere. The Nb content was increased by 2 mol% compared with the alloy in the previous study [18, 19] to observe the deformation texture of the β -phase without interference from the residual martensite. High-purity starting elements of Ti (99.99%), Nb (99.9%), and Al (99.99%) were used. The change in weight caused by arc melting was less than 0.1 wt % and was judged to be negligible, requiring no chemical analysis. The oxygen content of ingots fabricated by this method was reported to be less than 400 ppm by weight; thus, we assumed that the oxygen content of the current ingot was similar. The ingot was sealed in an evacuated quartz tube, homogenized at 1273 K for 7.2 ks and then quenched in water by breaking the quartz tube. The martensite start

temperature (M_s) and reverse martensite finish temperature (A_f) of this alloy were about 150 and 250 K, respectively. The grain size of the homogenized ingot was a few hundred micrometers.

Cubes $10 \times 10 \times 10$ mm were cut from the homogenized ingot and were cold-rolled at room temperature (RT) without lubricant. The specimens with $r = 60\%$, 80% , 90% , 95% , 97% , or 99% were prepared. The cold-rolled sheets are named according to the r value; for example, “CR99” denotes the cold-rolled sheet with r of 99% . The CR materials were wrapped in Ti-foil, sealed in an evacuated quartz tube, and solution-treated at 1273 K for 3.6 ks for recrystallization. The solution-treated materials are named in the same way as the cold-rolled materials.

The phase constitution of the materials was identified by X-ray diffraction (XRD) analysis with $\text{CuK}\alpha$ radiation using a X’pert-Pro Galaxy system equipped with an X’Celerator module (PANalytical).

Electron back scatter diffraction (EBSD) analysis was performed with a field-emission scanning electron microscope (FE-SEM; S-4300SE(N), Hitachi) equipped with an EBSD detector (INCA, Oxford Instruments). The area of analysis was about $100 \times 100 \mu\text{m}$ for the CR specimens and about $2 \times 2 \text{ mm}$ for the ST specimens. Identification of the texture orientation was made by the orientation distribution function (ODF). Texture strength was evaluated from the pole density in the inverse pole figure (IPF) with an admissible misorientation of 10° .

X-ray pole figure (XPF) technique was used to evaluate the global texture of the CR specimens at a larger scale because the substructure of the CR material was inhomogeneous. The XPF analysis (X’pert-Pro MRD, Philips) was made with $\text{CuK}\alpha$ radiation at RT by using system equipped with a Ge-monochromator. The beam size of

the X-ray was set to be 1×1 mm in the XPF and the range of elevation angle was from 0° to 85° . The 110, 200, and 112 diffraction peaks were used to measure the XPF. The analysis area depended on the elevation angle and was at least 100-fold that in the EBSD analysis.

The surfaces of the specimens for the XRD, EBSD, and XPF measurements were finished by electropolishing in a solution containing 6% perchloric acid, 35% butanol, and 59% methanol at 233 K to remove the damaged surface layer.

3. Results and discussion

3.1 Phase constitution

Figure 1 shows the XRD profiles of the CR materials. The materials were β -phase even after the cold-rolling and the α'' -martensite phase was not detected. The strongest line was 112 in CR60 and was 002 in CR80–99, whereas 110 was the strongest line in the powder diffraction indicating the existence of strong textures. The diffraction peaks were broadened by the introduction of dislocations by cold rolling. Figure 2 shows the XRD profiles of the ST materials, where only the β -phase was detected. Only 002 and 112 lines were detected, also indicating the existence of strong textures in the ST materials. The sharper diffraction peaks, compared to the broad diffraction peaks in CR materials, indicated that the materials were fully recrystallized by the solution treatment. The lattice parameter of the β -phase was determined as 0.3273 nm.

Figure 1, Figure 2

3.2 Microstructure and rolling texture of CR material

Two major orientations were identified in the CR and the ST specimens. Figure 3(a)

shows a schematic 100 pole figure of $\{001\}\langle 110\rangle$ and $\{112\}\langle 110\rangle$ orientations. These two textures are related by 35.16° rotation around RD // $\langle 110\rangle$. Figure 3(b) shows the locations of prominent orientations and bcc-fibers in a $\phi_2 = 45^\circ$ cross section of ODF. Hereafter, the ODF coordinates follow those in Fig. 3(b).

Figure 3

IPF maps of ND, 100 pole figures by EBSD, 200 pole figures by XPF, and $\phi_2 = 45^\circ$ cross sections of ODF by EBSD are shown for each rolling rate in Figs. 4(a)–(f). The black circles in the ODFs represent the positions of the orientations as indexed in Fig. 3(b). A longer exposure time was required for EBSD to obtain diffraction patterns due to the severe deformation in CR99, and the appearance of noise along the horizontal direction in the IPF map was inevitable in these measurements.

For CR60, the EBSD and XPF pole figures did not match exactly; the XPF pole figure contained other orientations that were not seen in the EBSD pole figure. XPF identified the global texture in CR60 as γ -fiber [34], although the fiber was not clear in the EBSD ODF. This indicates that the texture of CR60 was spatially inhomogeneous, as seen in the IPF map. The inhomogeneity in the texture might be due to the effect of the coarse initial grain size of the homogenized ingot [35].

The texture became spatially homogeneous as r increased; the results of EBSD and XPF matched each other in CR80–99. Fig. 4(b) shows that the γ -fiber developed in CR80. The strength of the $\{111\}\langle 110\rangle$ orientation was higher than the other orientations on the γ -fiber. The γ -fiber disappeared after further cold-rolling. In CR90 and CR95, the $\{001\}\langle 110\rangle$ texture was strong, as clearly seen in the IPF maps and the pole figures in Fig. 4(c) and (d). In CR90, a strong $\{001\}\langle 110\rangle$ texture was visible in the pole figures, whereas the intensity of $\{001\}\langle 110\rangle$ in the ODF was weak. This was

because the $\phi_2 = 45^\circ$ cross section of ODF did not contain all the $\{001\}\langle 110 \rangle$ family; the strength of the texture should be evaluated by the pole figure. In CR97, the $\{001\}\langle 110 \rangle$ texture was rotated around the RD to the $\{112\}\langle 110 \rangle$ texture on the α -fiber and the texture became inhomogeneous. The purple grains in the IPF map corresponded to $\{112\}\langle 110 \rangle$ texture. The grains were elongated along the RD independent of the texture component. The dominant texture was the α -fiber, particularly the $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$ textures in CR99.

Figure 4

The r -dependence of the texture strengths of the two key orientations, $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$, are summarized in Fig. 5. The texture strength of $\{001\}\langle 110 \rangle$ increased as r increased and reached 28 at $r = 97\%$. However, the $\{112\}\langle 110 \rangle$ orientation was not formed at lower r and began to grow rapidly at about $r = 95\%$ and reached a texture strength of 25 at $r = 97\%$. The texture transition during the cold-rolling is summarized as γ -fiber $\rightarrow \{001\}\langle 110 \rangle \rightarrow \alpha$ -fiber ($\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$). Similar results have been reported in Ti-29Nb-9Ta-10Zr [26] and Ti-35Nb-9Zr-6Mo-4Sn [25].

Dillamore and Katoh showed theoretically that $\{112\}\langle 110 \rangle$ and $\{11\ 11\ 8\}\langle 4\ 4\ 11 \rangle$ are the final stable orientations in cold-rolled bcc metal, whereas a part of the α -fiber, $\{112\}\langle 110 \rangle$ and $\{001\}\langle 110 \rangle$, is a quasi-stable final orientation assuming that pencil glide occurred [38]. Raabe et al. [39] also used the relaxed Taylor model to show that the $\{001\}\langle 110 \rangle$ texture and γ -fiber develop under the activation of $\{110\}$ and $\{112\}$ slip systems in bcc metals. The rolling texture in CR99 agreed with the theoretical analyses, suggesting a pencil glide deformation mode, as observed in the Ti-15Mo-5Zr-3Al β -titanium alloy, which is a similar to the present alloy [40].

Figure 5

3.3 Microstructure and recrystallization texture of ST materials

The IPF maps of ND, 100 pole figure, IPFs, and $\phi_2 = 45^\circ$ ODF cross sections obtained by EBSD for each ST specimen are shown in Figs. 6(a)–(f). The area of the EBSD analysis for the specimens was comparable to that of XPF and was sufficiently wide to discuss the global texture of the specimens. Texture development was negligible in ST60, although 001 to 111 and 001 to 110 showed a weak alignment parallel to ND and RD in the IPFs, respectively. A weak $\{001\}\langle 110 \rangle$ texture developed in ST80 and the major component of the rolling texture, $\{111\}\langle 110 \rangle$, almost disappeared. The recrystallization texture orientation sharpened dramatically for $r \geq 90\%$, as clearly seen in the pole figures. In ST90, 110 was sharply oriented to RD and the convergence of 001 along the ND became significant in the IPFs. The major texture was $\{001\}\langle 110 \rangle$ (red area in the IPF map in Fig. 6(c)) in ST90, although a weak $\{111\}\langle 110 \rangle$ texture appeared. Further increase in r caused the development of the $\{112\}\langle 110 \rangle$ texture. The $\{112\}\langle 110 \rangle$ orientation began to develop in ST95 and became the dominant texture in ST99; the $\{001\}\langle 110 \rangle$ orientation disappeared. This change in the orientation was expressed by the rotation along RD // $\langle 110 \rangle$ and was seen in the intensity distribution on the α -fiber. The recrystallized grains were equiaxed and the abnormal grain growth of a certain orientation was not observed, as seen in the IPF maps in Fig. 6. Figure 7 shows the r -dependence of the recrystallized grain size. As r increased, the grain size decreased continuously. This because the number of recrystallization initiation sites was increased by increasing r .

Figure 6, Figure 7

The r -dependence of the texture strength of the two major orientations in ST specimens is summarized in Fig. 8. The $\{001\}\langle 110 \rangle$ texture strength reached a maximum around ST90 and then decreased as r increased. However, the $\{112\}\langle 110 \rangle$ orientation developed instead of the $\{001\}\langle 110 \rangle$ orientation at the higher r . In ST99, the $\{001\}\langle 110 \rangle$ orientation completely disappeared and $\{112\}\langle 110 \rangle$ was the dominant orientation with a strength of 38. Comparing Figs. 8 and 5 shows that the increase in $\{112\}\langle 110 \rangle$ orientation in CR95 was linked to the decrease of the $\{001\}\langle 110 \rangle$ orientation and the increase of the $\{112\}\langle 110 \rangle$ orientation in ST95.

Figure 8

The recrystallization texture follows the rolling texture in bcc metals and alloys, unlike fcc metals [35]. Furthermore, the $\{001\}\langle 110 \rangle$ deformation texture does not recrystallize along the α -fiber orientations [35-37]. The development of the $\{112\}\langle 110 \rangle$ orientation in the CR material became significant when r was higher than 95% in the present alloy. Therefore, the competition between the development of $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$ orientations occurred during the recrystallization for $r > 95\%$; the $\{112\}\langle 110 \rangle$ orientation overwhelmed the $\{001\}\langle 110 \rangle$ orientation. This is a plausible explanation for the strength of the $\{001\}\langle 110 \rangle$ texture in the ST specimen reaching a maximum at around $r = 90\%$ and $\{112\}\langle 110 \rangle$ becoming the dominant orientation at $r = 99\%$. The optimum r for obtaining the $\{001\}\langle 110 \rangle$ recrystallization texture depended on the alloy system. In the Ti-24Zr-(8-12)Nb-2Sn alloy [41], $\{001\}\langle 110 \rangle$ is the only orientation even at $r = 98.5\%$. Our results indicate that a fully recrystallized $\{001\}\langle 110 \rangle$ texture, which is desirable for the β -titanium SMAs, is obtained by controlling r to produce only the $\{001\}\langle 110 \rangle$ texture in CR material, preventing the $\{112\}\langle 110 \rangle$ texture from overwhelming the $\{001\}\langle 110 \rangle$ texture during the

recrystallization.

4. Conclusions

The rolling rate dependence of the rolling and recrystallization texture was investigated in Ti-26Nb-3Al (mol%) and we obtained the following results.

1. The texture transition during the cold-rolling was γ -fiber $\rightarrow \{001\}\langle 110 \rangle \rightarrow \alpha$ -fiber ($\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$). $\{001\}\langle 110 \rangle$ is the dominant orientation when the rolling rate is 90% and $\{112\}\langle 110 \rangle$ appears above 95% in the rolling rate.
2. The $\{001\}\langle 110 \rangle$ recrystallization texture is well developed when the rolling rate is 90%. The $\{112\}\langle 110 \rangle$ recrystallization texture overwhelms the $\{001\}\langle 110 \rangle$ texture when the rolling rate is higher than 95%. The recrystallization grain is equiaxed and the average grain size decreases with increasing r .
3. To obtain the $\{001\}\langle 110 \rangle$ recrystallization texture, which is the desirable orientation for β -titanium shape memory alloys, r must be controlled to form only the $\{001\}\langle 110 \rangle$ rolling texture, preventing the $\{112\}\langle 110 \rangle$ texture from overwhelming the $\{001\}\langle 110 \rangle$ texture during recrystallization.

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References

- [1] S. Miyazaki, H.Y. Kim, H. Hosoda, *Materials Science and Engineering: A*, 438-440 (2006) 18-24.
- [2] A. Yamamoto, Y. Kohyama, H. Hosoda, S. Miyazaki, T. Hanawa, *Materials Transactions*, 48 (2007) 361-366.
- [3] A.R.G. Brown, D. Clark, J. Eastabrook, K.S. Jepson, *Nature*, 201 (1964) 914-915.
- [4] J.I. Kim, H.Y. Kim, T. Inamura, H. Hosoda, S. Miyazaki, *Materials Science and Engineering: A*, 403 (2005) 334-339.
- [5] H.Y. Kim, Y. Ikehara, J.I. Kim, H. Hosoda, S. Miyazaki, *Acta Mater.*, 54 (2006) 2419-2429.
- [6] E.G. Obbard, Y.L. Hao, T. Akahori, R.J. Talling, M. Niinomi, D. Dye, R. Yang, *Acta Mater.*, 58 (2010) 3557-3567.
- [7] Y. Al-Zain, Y. Sato, H.Y. Kim, H. Hosoda, T.H. Nam, S. Miyazaki, *Acta Mater.*, 60 (2012) 2437-2447.
- [8] E. Takahashi, T. Sakurai, S. Watanabe, N. Masahashi, S. Hanada, *Materials Transactions*, 43 (2002) 2978-2983.
- [9] C. Baker, *Metal Science*, 5 (1971) 91-100.
- [10] T.W. Duerig, J. Albrecht, D. Richter, P. Fischer, *Acta Metall.*, 30 (1982) 2161-2172.
- [11] T. Grosdidier, M.J. Philippe, *Materials Science and Engineering a-Structural Materials Properties Microstructure and Processing*, A291 (2000) 218-223.
- [12] D. Sugano, M. Ikeda, *Materials Science and Engineering: C*, 25 (2005) 377-381.
- [13] T. Maeshima, M. Nishida, *Materials Transactions*, 45 (2004) 1096-1100.
- [14] T. Maeshima, S. Ushimaru, K. Yamauchi, M. Nishida, *Materials Transactions*, 47 (2006) 513-517.
- [15] T. Furuhashi, S. Annaka, Y. Tomio, T. Maki, *Materials Science and Engineering: A*, 438-440 (2006) 825-829.
- [16] Y. Tomio, T. Furuhashi, T. Maki, *Materials Transactions*, 50 (2009) 2731-2736.
- [17] T. Sugimoto, K. Kamei, S. Komatsu, K. Sugimoto, *Titanium Science and Technology*, 3 (1985) 1583-1590.
- [18] Y. Fukui, T. Inamura, H. Hosoda, K. Wakashima, S. Miyazaki, *Materials Transactions*, 45 (2004) 1077-1082.
- [19] T. Inamura, Y. Fukui, H. Hosoda, K. Wakashima, S. Miyazaki, *Materials Transactions*, 45 (2004) 1083-1089.
- [20] K. Otsuka, C.M. Wayman, *Shape Memory Materials*, Cambridge University Press, 1999.
- [21] T. Inamura, H. Hosoda, K. Wakashima, S. Miyazaki, *Materials Transactions*, 46

(2005) 1597-1603.

[22] H.Y. Kim, T. Sasaki, K. Okutsu, J.I. Kim, T. Inamura, H. Hosoda, S. Miyazaki, *Acta Mater.*, 54 (2006) 423-433.

[23] J. Fu, A. Yamamoto, H.Y. Kim, H. Hosoda, S. Miyazaki, *Acta Biomater.*, 17 (2015) 56-67.

[24] L. Wang, W. Lu, J. Qin, F. Zhang, D. Zhang, *Materials Science and Engineering: A*, 490 (2008) 421-426.

[25] S. Dai, Y. Wang, F. Chen, X. Yu, Y. Zhang, *Materials Science and Engineering: A*, 575 (2013) 35-40.

[26] V.D. Cojocaru, D. Raducanu, T. Gloriant, D.M. Gordin, I. Cinca, *Materials Science and Engineering: A*, 586 (2013) 1-10.

[27] H. Hosoda, Y. Kinoshita, Y. Fukui, T. Inamura, K. Wakashima, H.Y. Kim, S. Miyazaki, *Materials Science and Engineering: A*, 438-440 (2006) 870-874.

[28] T. Inamura, Y. Fukui, H. Hosoda, K. Wakashima, S. Miyazaki, *Materials Science and Engineering: C*, 25 (2005) 426-432.

[29] T. Inamura, Y. Kinoshita, J.I. Kim, H.Y. Kim, H. Hosoda, K. Wakashima, S. Miyazaki, *Materials Science and Engineering: A*, 438-440 (2006) 865-869.

[30] T. Inamura, Y. Fukui, H. Hosoda, K. Wakashima, S. Miyazaki, *Mater. Sci. Forum*, 475-479 (2005) 2323-2328.

[31] V.-D. Cojocaru, D. Raducanu, T. Gloriant, I. Cinca, *JOM*, 64 (2012) 572-581.

[32] T. Inamura, J.I. Kim, H.Y. Kim, H. Hosoda, K. Wakashima, S. Miyazaki, *Philos. Mag.*, 87 (2007) 3325-3350.

[33] T. Inamura, H. Hosoda, K. Wakashima, J.I. Kim, H.Y. Kim, S. Miyazaki, *Materials Transactions*, 48 (2007) 395-399.

[34] T. Inamura, R. Shimizu, H. Hosoda, S. Miyazaki, *Mater. Sci. Forum*, 706-709 (2012) 1899-1902.

[35] A. Rollett, F.J. Humphreys, G.S. Rohrer, *Recrystallization and Related Annealing Phenomena*, Elsevier Science, 2004.

[36] O. Hashimoto, N. Ohashi, K. Fujimoto, *The Iron and Steel Institute of Japan*, 57 (1971) 1155-1166.

[37] W.R. Hibbard, W.R. Tully, *Transactions of the Metallurgical Society of Aime*, 221 (1961) 336-343.

[38] I.L. Dillamore, H. Katoh, *Metal Science*, 8 (1974) 21-27.

[39] D. Raabe, K. Lucke, *Z. Metallk.*, 85 (1994) 302-306.

[40] S.-H. Lee, K. Hagihara, T. Nakano, *Metallurgical and Materials Transactions A*, 43 (2011) 1588-1597.

[41] L. López Pavón, H.Y. Kim, H. Hosoda, S. Miyazaki, *Scripta Mater.*, 95 (2015) 46-49.

Figure captions

Fig. 1: XRD profiles of the CR specimens

Fig. 2: XRD profiles of the ST specimens

Fig. 3: Ideal positions of (a) 100 pole in $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$ orientations and (b) some key orientations in ODF at $f_2 = 45^\circ$

Fig. 4: Rolling texture of the CR specimens

Fig. 5: Rolling rate dependence of the texture strength for the major orientations in the CR specimens

Fig. 6: Recrystallization texture of the ST specimens

Fig. 7: Rolling rate dependence of the average grain size in the ST specimens

Fig. 8: Rolling rate dependence of the texture strength for the major orientations in the ST specimens

Fig. 1 XRD profiles of CR specimens

Fig. 2 XRD profiles of the ST specimens

Fig. 3 Ideal positions of (a) 100 pole in $\{001\}\langle 110 \rangle$ and $\{112\}\langle 110 \rangle$ and (b) some key orientations in ODF at $\phi_2=45^\circ$.

Fig. 4 Rolling texture of the CR specimens

Fig. 5 Reduction rate dependence of the texture strength for the major orientations in CR specimens

Fig. 6 Recrystallization texture of the ST specimens

Fig. 7 Reduction rate dependence of the averaged grain size in the ST specimens

Fig. 8 Reduction rate dependence of the texture strength for the major orientations in the ST specimens