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Effect of texture on phase-transformation strain in CuZnAl shape memory sheets①

SHEN Fu-fa(沈甫法), YUAN Wen-qing(袁文庆)
(Key Laboratory for High Temperature Materials and Tests, Ministry of Education, Shanghai Jiaotong University, Shanghai 200030, China)

[Abstract] The textured shape memory alloys exhibits anisotropic because the property of single crystal is strongly orientation-dependent. The effect of texture on phase-transformation strain in CuZnAl shape memory sheets was investigated. The texture of parent austenite was measured by X-ray goniometer and analyzed by the orientation distribution function. Subsequently, using the texture parameters and single crystal properties, the phase transformation strains at the different directions of rolling plane by the statistically averaging method were calculated. It was showed that the experimental results are agreeable with the calculated ones. It is well explained that this anisotropy of phase-transformation strain is mainly caused by the crystallographic texture of the rolled sheets.

[Key words] shape memory alloys; texture; phase transformation strain
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1 INTRODUCTION

Shape memory effects that show thermoelastic martensitic transformation have been understood in CuZnAl shape memory alloys. It is believed that on cooling from the parent austenite, a self-accommodating microstructure is formed in the martensitic state; the martensitic variant coalescence takes place on loading, and the shape recovery ensues on heating due to the reverse martensitic transformation[1, 2]. In previous investigations the fundamental crystallographic data could be found for the single crystals. For example, the parent austenite of CuZnAl alloys is a BCC lattice of B_2 type and the lattice constant is about 0.2947 nm. The product martensite phase is a monoclinic lattice of 18R type and the lattice constants are about a = 0.4553 nm. b = 0.5452 nm, c = 3.8977 nm, and β= 87.5°[3~ 6]. Thus, the relation between crystallographic textures and shape memory effects in shape memory polycrystals could be investigated based on the direction-dependent single crystal elastic or/and plastic properties and texture parameters in the past years[7~ 15].

Orientation distribution function (ODF) of polycrystalline aggregates is increasingly used to perform texture analysis for shape memory alloys[10~ 15]. The austenitic and martensitic textures of CuZnAl alloys were studied using ODFs, and both textures were related to each other by the orientation correlation function without considering the selection of martensite variants[10]. Inoue et al[13] used ODF to study the textures of TiNi alloys and calculated the phase transformation strain of different directions in the rolling plane. A computer program was developed in terms of series expansion method for the ODF analysis of monoclinic shape memory alloy martensites[12]. Zhao et al[14] analyzed the texture of TiNiCu alloys using ODF.

In this paper, a Cu-26.1Zn-4.5Al-0.7Ti(%) is used and heat-treated into an α+ β dual phase microstructure. In the dual region the CuZnAl material is reduced by cold-rolling with intermediate annealing to the thickness of 0.5 mm. Then it is heated to 810 ℃, remained for 15 min, and quenched in water to room temperature. At room temperature, the material is in the martensitic state. Pole figures of the parent austenite are measured by texture goniometer with a simple heater, and are then used to calculate the orientation distribution functions. Thus, the texture of parent austenite is determined to be a [110] fiber texture. From the texture parameters and single crystal properties, the recoverable phase transformation strains at the different directions of rolling plane are estimated by the statistically averaging method. By comparing the experimental and calculated data, it has found that the experimental results are agreeable with the calculated ones. It is well explained that the anisotropy of phase-transformation plasticity is related with the crystallographic texture.

2 EXPERIMENTAL

In the present study, a Cu-26.1Zn-4.5Al-0.7Ti (%) alloy was used. It was heated up to 700 ℃ and kept for 30 min, then cooled with furnace down to 550 ℃ and kept for 120 min, and it was rolled into the thickness of 0.5 mm at 550 ℃. Finally it was heated up to 810 ℃ and quenched into room tempera-

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ture. At room temperature, it was in the martensitic state. For the FCC α phase of dual phase state, the (200) (220) (311) pole figures were measured by texture goniometer, using CoK$_\alpha$-radiation, in steps of $\Delta\alpha = 5°$ and $\Delta\beta = 5°$ up to 70°. For the BCC β phase of dual phase state, the (200) (110) (211) pole figures were determined with the same method. For the austenite after quenching, the same pole figures as in BCC β phase of dual phase state were measured. Their corresponding orientation distribution functions were calculated (series expansion degree $L = 22$), was shown in Figs. 1, 2 and 3.

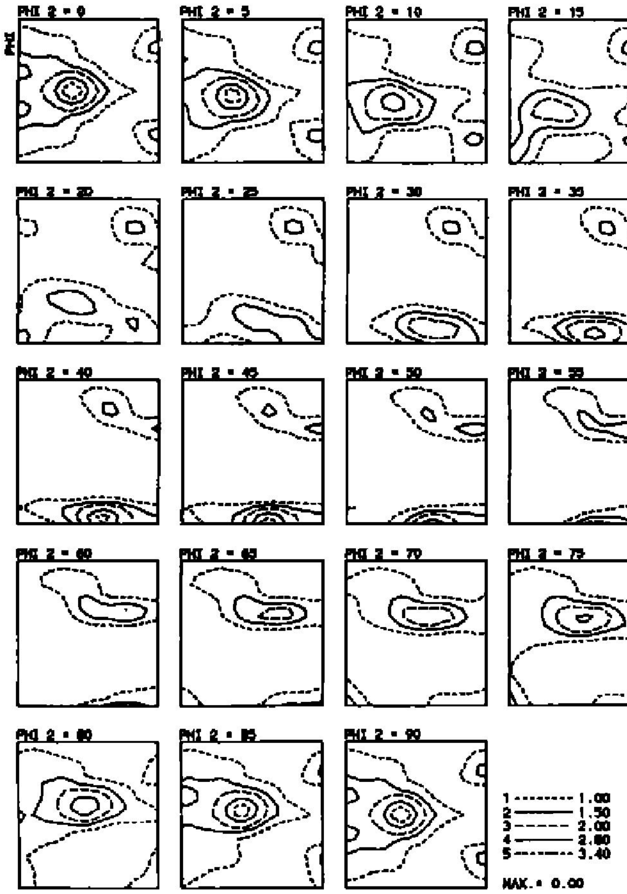

Fig. 1 Orientation distribution functions of α phases in dual phase state

Samples for tensile tests were spark-eroded from the quenched CuZnAl sheets at the angles of 0°, 15°, 30°, 45°, 60°, 75° and 90° to rolling direction. The dimension of sample is 60 mm × 10 mm × 0.5 mm. The surface of samples was polished by SiC paper up to grade 600. All the samples were tested on an INSTRON materials testing machine at room temperature. The variation of sample length was recorded by an extensometer with a gauge length of 50 mm. The strain rate was controlled at $10^{-2}\,s^{-1}$ upon loading. The unloading started when the tensile stress reached a certain value. The experimental stress—strain curve is presented in Fig. 4 for the sample tested at 15° to rolling direction. From the experiments at various angles, the variation of recoverable phase transformation

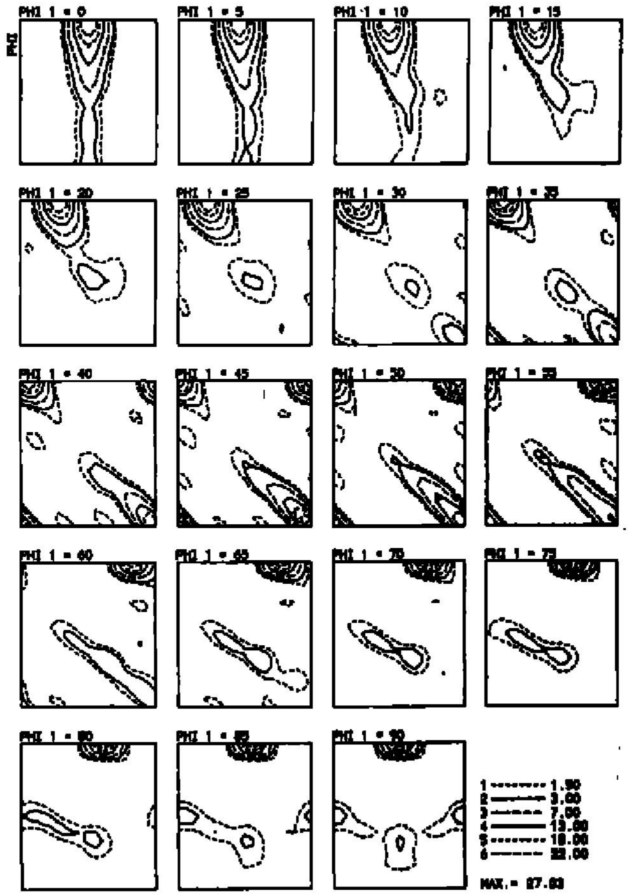

Fig. 2 Orientation distribution functions of β phases in dual phase state

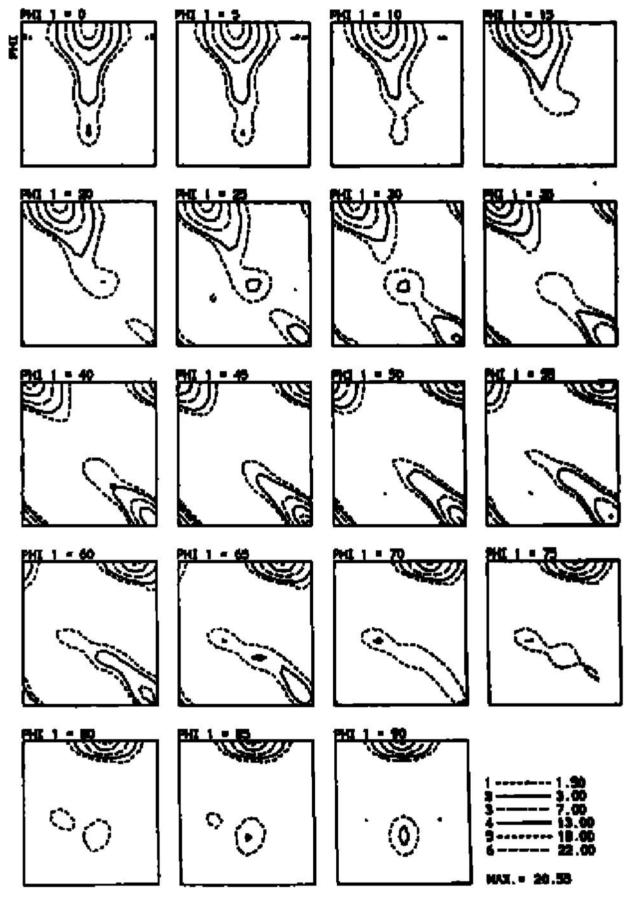

Fig. 3 Orientation distribution functions of parent austenite phases

strains with angles to rolling direction is summarized and shown in curve 1 of Fig. 5.

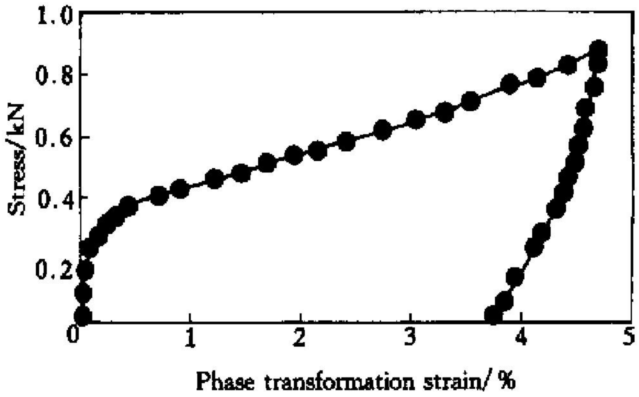

Fig. 4 Stress —strain curves of phase transformation of CuZnAl alloy

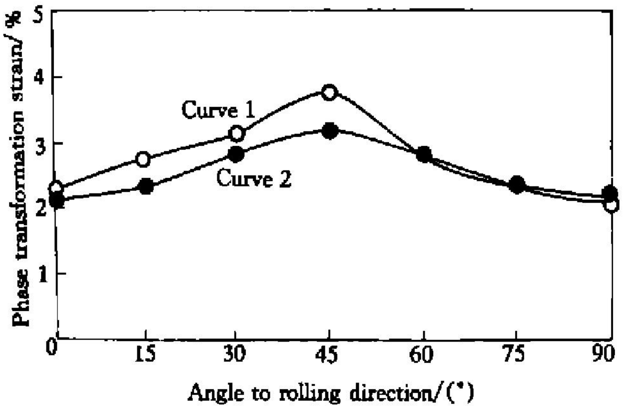

Fig. 5 Relation between phase transformation strain and angle to rolling direction

3 DISCUSSION

3.1 Texture of dual phase state

CuZnAl alloy is a technically applicable shape memory alloy due to many advantages, such as widely adjustable transformation temperatures, narrow hysteresis loops, low cost as well as low electric resistance. It is easy to grow up to coarse grain, therefore, being brittle. The suitable chemical composition and heat treatment are usually selected for fabricating the CuZnAl shape memory alloy to solve this problem. For this reason, a composition of Cu-26Zn-5Al-0.7Ti(%) was used in this experiment, as which the alloy was heated up to 700 ℃ and kept for 20 min, then, cooled with furnace down to 550 ℃ and kept for 120 min. Thus, this material is in the dual phase region with the α and β phases of equal amount and goes through cold-rolling with intermediate annealing. In previous investigations it was proved that the α phase has a strong texture, but in our experiment there is a weak texture. From Fig. 1 we can see that the maximal orientation density is 3.92 at (011)[112]. The texture type is different from that found in former references. This may be caused by the following factors. On one hand, the intermediate annealing leads to the difference in texture type and intensity. In annealing, recrystallization takes place, lowering the texture intensity in α phase as indicated by the study of Engler[16]. On the other hand, the α phase deformation is influenced by β phase, which takes 50% (volume fraction). In deformation, the external stress mainly acts on β phase because β phase is harder than α phase and distributes in netlike structure. Its deformation plays a governing role and proceeds more or less like the deformation of single grains. The shape of α grains is passively changed in accommodation to the shape of β grains. The deformation of α grains may be to some degree randomly. Therefore, in our case, β phase resists the texture formation in α phase, the result is that α phase has a weak texture. In opposite, β phase has a strong texture. From Fig. 2, it is seen that there is a fiber texture from orientation (001)[110] to (111)[110]. For this fiber texture, the fiber axis [110] is parallel to rolling direction and the maximum orientation density is 27.03 at orientation (001)[110]. The above-mentioned recrystallization annealing intensifies the texture of β phase[16], and the existence of α phase is beneficial to the texture formation of β phase because α phase is soft[11].

3.2 Texture of austenite phase

The shape memorized quenching must be carried out to get shape memory effect. Firstly the dual phase alloy was heated to 800 ℃ and kept for 20 min and cooled with water to room temperature. At that time it is in the martensitic state because the martensite finish temperature is 66 ℃. Thus, the austensite pole figures were measured at 80 ℃ by texture goniometer. The orientation distribution function is shown in Fig. 3. From Fig. 3, it can be seen that the austenite texture is a fiber texture from orientation (001)[110] to (111)[110]. The fiber axis [110] is parallel to rolling direction and the maximum orientation density is 20.55 at orientation (001)[110]. It is inherited from β phase texture in the dual phase state, at the same time, α phase disappears. The austenite phase texture resembles the β phase texture in the dual phase state. By comparing Fig. 2 and Fig. 3, it is concluded that the texture type hardly changes significantly while the intensity somewhat changes during quenching. It was reported that the texture of such type is beneficial to the shape memory effects[8]. Based on this texture, the phase transformation plasticity at different directions of rolling plane is calculated for this shape memory alloy as paragraph 3.3.

3.3 Calculation of anisotropic phase-transformation plasticity

It is well known that for single crystals, the recoverable phase transformation strains could be calculated under tensile loading from the shape strain, lattice strain or Taylor's factor for various cases, and were in agreement with the experimental values[3~5]. In these studies, the single crystal shape was not considered and the strain was assumed to be axis-symmetric, so that the recoverable transformation strain was only dependent on the tensile axis relative to the crystal (Fig. 6). In the polycrystalline texture analysis, the inverse pole figure of tensile direction indicates the distribution of various crystal planes normal to the tensile axis, which is the relative volume fraction of various crystal planes normal to the tensile axis. If the individual grain shape is not considered and the interaction between the constituent grains is ignored, the mean value of axial recoverable transformation strain for polycrystalline shape memory alloys can be estimated by linearly averaging the strains of all the constituent grains in the tensile axis by Eqn. (1).

$$\bar{\varepsilon}_y = \frac{\int_s R_y(hkl)\,\varepsilon(hkl)\,\mathrm{d}S}{\int_s R_y(hkl)\,\mathrm{d}S} \qquad (1)$$

where $\bar{\varepsilon}_y$ is the mean recoverable transformation strain of polycrystal in the tensile axis y. $\varepsilon(hkl)$ is the recoverable transformation strain of single crystal in the normal direction of crystal plane (hkl). $R_y(hkl)$ is the pole density at the pole point (hkl) in the inverse pole figure of the tensile axis y, which is the volume fraction of the crystal plane (hkl) normal to the tensile axis. $\mathrm{d}S$ is the spherical area element around the (hkl) pole point, and S is the integral area (i. e. the spherical triangle of inverse pole figure).

In order to solve the above-mentioned equation, the equal spherical area method is used to discrete the integral area. The whole spherical area of inverse pole figures is first divided into the n small domains of equal spherical area[15,17]. In each small domain, the values of $R_y(h_ik_il_i)$ and $\varepsilon(h_ik_il_i)$ are assumed to be uniform, where h_i, k_i, l_i are the coordinates of the center of small domain i. Hence, Eqn. (1) is expressed by averaging the values of n discrete points as shown in Eqn. (2).

$$\bar{\varepsilon}_y = \frac{R_y(h_1k_1l_1)\cdot\varepsilon(h_1k_1l_1) + R_y(h_2k_2l_2)\cdot\varepsilon(h_2k_2l_2) + \cdots + R_y(h_nk_nl_n)\cdot\varepsilon(h_nk_nl_n)}{R_y(h_1k_1l_1) + R_y(h_2k_2l_2) + \cdots + R_y(h_nk_nl_n)} \qquad (2)$$

If the Taylor factors for single crystals are used, the mean Taylor factor of polycrystal in tensile axis y is given in Eqn. (3)

$$\bar{M}_y = \frac{R_y(h_1k_1l_1)\cdot M(h_1k_1l_1) + R_y(h_2k_2l_2)\cdot M(h_2k_2l_2) + \cdots + R_y(h_nk_nl_n)\cdot M(h_nk_nl_n)}{R_y(h_1k_1l_1) + R_y(h_2k_2l_2) + \cdots + R_y(h_nk_nl_n)} \qquad (3)$$

The recoverable phase transformation strain in the tensile axis y is expressed as Eqn. (4).

$$\bar{\varepsilon}_y = \frac{\eta(1-\bar{\upsilon})}{\bar{M}_y} \qquad (4)$$

where $\bar{\varepsilon}_y$ is the mean recoverable strain of polycrystal in the tensile axis y, $\bar{\upsilon}$ is the volume fraction of retained austenite, η is the maximum shear magnitude in habit planes, and $\bar{M}_y$ is the mean Taylor factor in tensile axis y.

For the various tensile axes, we have the respective inverse pole figures, therefore, it was possible to calculate the recoverable transformation strains in the different tensile axes. Finally, the recoverable strains as a function of angles to rolling direction can be obtained as shown in curve 2 in Fig. 5. Comparing curves 1 and 2 in Fig. 5, we discover that the experimental and calculated values changes agreeably with the angles to rolling direction although there is a difference between their absolute values. Thus, we can easily conclude that this textured shape memory alloy has good shape memory effect in 45° direction whereas in the rolling and transverse directions the shape memory effect is low. From this comparison, it is proved that such a calculation method can very well predict the anisotropy of phase transformation strain for shape memory alloys, therefore, estimate the shape memory ability at the different directions. It makes us be able to use shape memory alloys optimally.

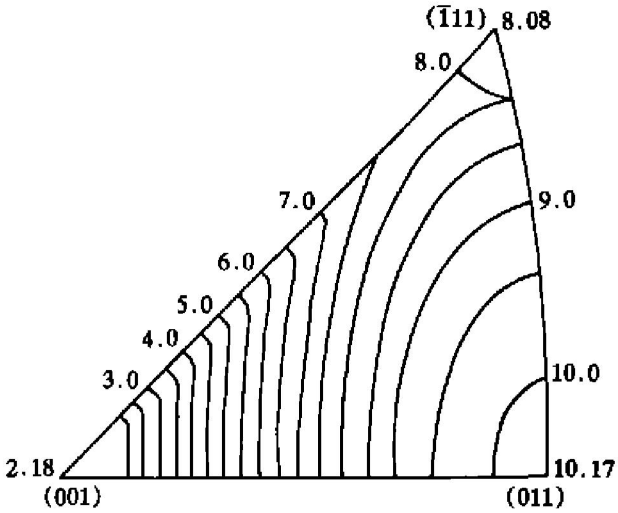

Fig. 6 Orientation-dependent Taylor factor in CuZnAl single crystals

4 CONCLUSION

Quantitative texture analysis shows that Cu-26.1Zn-4.5Al-0.7Ti(%) alloy austenite has a fiber texture of fiber axis [110] parallel to rolling direction, and the orientation density extends from 20.0 at (001)[110] orientation to 0 at (111)[110]. In this alloy, there is only a single fiber texture which is caused by cold-rolling with repeated annealing. Second, the method of statistically averaging the strains of all the constituent grains in the tensile axis is effective for predicting the phase transformation strain anisotropy of CuZnAl textured shape memory polycrystals. The predicted and experimental results change agreeably with the angles to rolling direction. This calculation method can clearly indicate where there is a high shape memory effect.

[REFERENCES]

[1] Otsuka K, Shimizu K. Shape memory alloys and its applications [J]. Inter Mater Rev, 1986, 31: 93- 130.

[2] Delaey L, Aernoudt E, Roos J R. Crystallography of martensitic transformation [J]. Metall, 1977, 31: 1325 - 1350.

[3] Miyazaki S, Otsuka K, Wayman C M. The shape memory mechanism associated with the martensitic transformation in Ti-Ni alloys- Ⅰ. Self-accommodation [J]. Acta Mater, 1989, 37: 1873- 1883.

[4] Miyazaki S, Otsuka K, Wayman C M. The shape memory mechanism associated with the martensitic transformation in Ti-Ni alloys- Ⅱ [J]. Acta Mater, 1989, 37: 1884- 1894.

[5] Ono N, Satoh A, Ohta H. A discussion on the mechanical properties of shape memory alloys based on a polycrystal model [J]. Trans Jpn Inst Metals, 1989, 30: 756- 764.

[6] Miyazaki S, Kimura S, Otsuka K, et al. The habit plane and transformation strains with the martensitic transformation in TiNi single crystals [J]. Scripta Metall, 1984, 18: 883- 890.

[7] Monasevich L A, Paskal Y I, Prib V E, et al. Investigation of TiNi-based shape memory alloys [J]. Metalloved Term Obrab Met, 1979, 9: 62.

[8] Eucken S, Hirsch J, Hornbogen E. Anisotropy and texture of CuZnAl shape memory alloy [J]. Texture and Microstr, 1988, 8- 9: 415.

[9] Mulder J H, Thoma P E, Beyer J. Anisotropy of the shape memory effect in tension of cold-rolled 50.8 Ti 9.2 Ni (mole fraction, %) sheet [J]. Z Metallkd, 1993, 84: 501- 508.

[10] Park N J, Bunge H J. Texture transformation of CuZnAl shape memory alloys [J]. Z Metallkd, 1990, 81: 636- 642.

[11] Yuan W Q, Park N J. Development of the dual phase CuZnAl shape memory alloys [J]. Mater Sci Forum, 1998, 273- 275: 541.

[12] Yuan W Q, Yi S. Orientation distribution functions of a TiNi alloy an austenite and a TiNiCu alloy martensite [J]. Scripta Mater, 1999, 41: 1319- 1325.

[13] Inoue H, Miwa N, Inakazu N. Texture and tensile transformation strain of cold-rolled TiNi sheets [J]. Acta Mater, 1996, 44: 4825- 4834.

[14] Zhao L, Willemse P E, Mulder J H, et al. Texture development and transformation strain of a cold-rolled Ti50-Ni45-Cu5 alloy [J]. Scripta Mater, 1998, 39: 1317- 1325.

[15] Yuan W Q, Yi S. Pseudo-elastic strain estimation of TiNi shape memory alloys [J]. Mater Sci and Eng, 1999, A271: 439- 448.

[16] Engler O, Hirsch J, Lueck K. Texture development of dual phase CuZn alloys [J]. Z Metallkd, 1995, 7: 465 - 474.

[17] Bunge H J. Texture Analysis in Materials Science [M]. London: Butterworths, 1982.

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