Article ID: 1003- 6326(2000)02- 0144- 05

Influence of B2 ordered structure on deformation texture①

MAO Wei-min(毛卫民), ZHU Guo-hui(朱国辉), ZHAO Zi-su(赵子苏), YU Yong-ning(余永宁)
Department of Materials Science and Engineering,
University of Science and Technology Beijing, Beijing 100083, P. R. China

Abstract: The formation of cold rolling textures in FeCo, CuZn and Fe_3Al based alloys with B2 structure was analyzed using X-ray diffraction technology. The difference of deformation textures obviously demonstrated the different behaviors of plastic deformation in the alloys. The boundary energy of anti-phase domains has important influence on the crystallographic behaviors of B2 ordered alloys during deformation. The activation of slip systems on the {110} planes should be the main deformation mechanism in B2 ordered alloys. The mechanical twinning on {112} planes appeared frequently in CuZn alloy with lower boundary energy of anti-phase domains, while a rather typical rolling texture like that in BCC metals was observed in Fe_3Al alloy with incomplete B2 structure, indicating that its boundary energy of anti-phase domains does not have important influence on the deformation mechanism similar to BCC metals.

Key words: ordering structure; boundary of anti-phase domain; deformation texture **Document code: A**

1 INTRODUCTION

Because of the special atom occupation in crystalline lattice, the dislocation structure of intermetallics is related to the formation of anti-phase domain boundary, for which the crystallographic behaviors of the plastic deformation is different from that of the conventional metals. In B2 ordered structure a complete dislocation has very large Burger's vector, which will be decomposed into two $a/2\langle 111\rangle$ partial dislocations connected by an anti-phase domain boundary. The two connected partial dislocations will also move connectedly during the plastic deformation, so that the ordered structure is basically not destroyed[1]. The connected moving of the couple partial dislocations will influence the plastic deformation process, mechanical properties of the materials as well as the formation of the deformation textures, on which many details are still unknown so far.

The crystallographic process of plastic deformation and the formation process of the deformation textures in FeCo and CuZn alloys with B2 structure as well as Fe_3Al alloy with incomplete B2 structure were observed and analyzed. The investigation will benefit the application of intermetallic materials as well as the physical metallurgical theory on plastic deformation and recrystallization.

2 EXPERIMENTAL

High purity Fe, Co, Cu, Zn and Al were melted in vacuum furnace and the ingots Fe-53% Co, Cu-47% Zn and Fe-28% Al (mole fraction) were obtained. The FeCo ingot was annealed at 950 ℃ for 2 h for the composition homogenization, and then forged at 950 ℃ into sheets with 5.5 mm thickness. According to Yamaguchi et al[2], the forged sample was annealed at 550 ℃ for 10 d and slowly cooled in order to obtain the maximal ordered degree. After following 37% reduction rolling at room temperature, many cracks were observed. The CuZn ingot with 22 mm thickness was annealed in a box furnace at 650 ℃ for 5 h for the composition homogenization, and then rolled at 700 ℃ down to 8.7 mm. The hot rolled sample was annealed at 440 ℃ in a salt bath for 5 h and quenched in order to obtain the ordered structure. After following 40% reduction rolling at room temperature many cracks were also observed. The Fe_3Al based alloy has B2 structure over 600 ℃, therefore the Fe_3Al ingot was 60% rolled at 600 ℃, after which a deformed structure was obtained and no recrystallization was observed. A 80% cold rolled commercial IF steel sheet was taken in order to compare the rolling texture with the intermetallics.

Using X-ray technology three incomplete pole Figs. {110}, {200} and {112} of the rolling samples were measured in the center layers of the rolling samples. Co tube was used for FeCo and Fe_3Al alloys, and Cu tube was used for CuZn alloy. The orientation distribution functions (ODFs) were calculated according to the series expansion method[3].

3 RESULTS

Fig. 1 gives the ODF φ_1 sections of the rolling samples. In general, all the samples have the rolling texture characteristics of BCC metals. However in details there are obviously some differences. The rolling

① **Foundation item:** Project 59671008 supported by the National Natural Science Foundation of China and Deutsche Forschungsgemeinschafte
Received date: Jan. 19, 1999; **accepted date:** Apr. 8, 1999

textures of the intermetallics with B2 structure are actually different from that of BCC metals with A2 structure if the rolling texture in IF steel sheet is considered as the typical BCC rolling texture. Fig. 2 gives the α, γ and τ fiber analysis of ODFs. In the α fiber (φ_1= 0°, φ_2= 45°, Φ= 0°~ 90°) all grains have a ⟨110⟩ direction parallel to the rolling direction, in the γ fiber (φ_1= 60°~ 90°, φ_2= 45°, Φ= 55°) all grains have a ⟨111⟩ direction parallel to the normal direction, and in the τ fiber (φ_1= 90°, φ_2= 45°, Φ

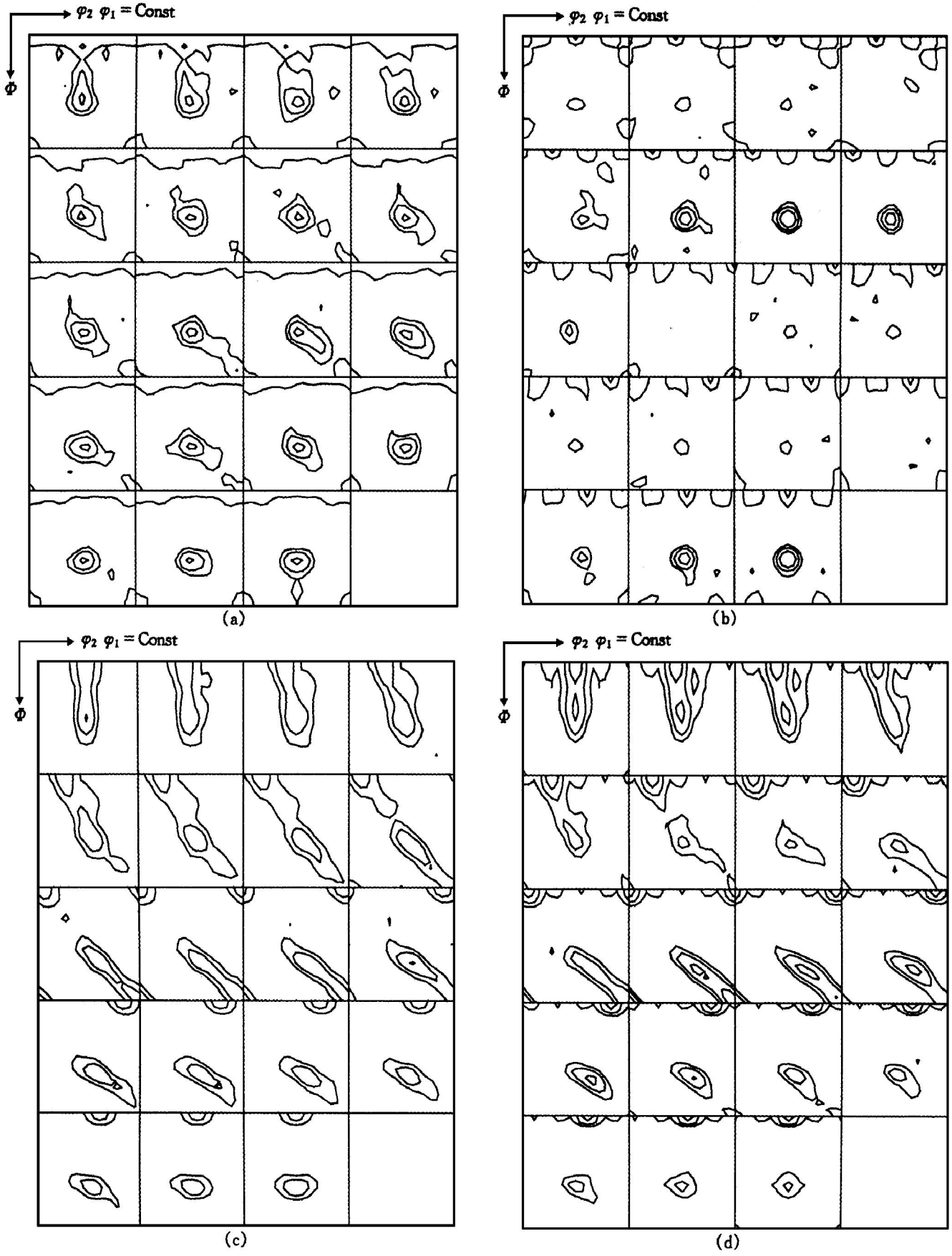

Fig. 1 ODF φ_1 sections of B2 ordered alloys and IF steel

(a) —37% rolled FeCo; (b) —40% rolled CuZn; (c) —60% rolled Fe_3Al; (d) —80% rolled IF steel

= 0°~ 90°) all grains have a ⟨110⟩ direction parallel to the transverse direction[4]. The three fibers demonstrate the main characteristics of the orientation distributions $f(\varphi_1, \Phi, \varphi_2)$ and textures of the rolling samples[4].

The grain orientations in IF steel are accumulated around the α fiber, and especially in the area of Φ = 0°~ 50°(Fig. 2(a)), in which the orientation density is relatively high and homogeneous. The orientation density in γ fiber is also homogeneous, but very low (Fig. 2(b)). On the other hand, the rolling texture in Fe_3Al based alloy has the similar texture to the IF steel (Fig. 1(c), (d)), indicating that its deformation mechanism is similar to the BCC metals. Orientation density peaks at {558} ⟨110⟩ i. e. (φ_1= 0°, Φ= 40°~ 45°, φ_2= 45°) near {112} ⟨110⟩ were observed in the two materials (Fig. 2(a)).

In the α fiber of FeCo alloy, the orientation density is not shown very high in area of Φ= 0°~ 50°. The grain orientations are accumulated more around Φ= 35° and Φ= 55° as well as near Φ= 0°(Fig. 2(a)), which correspond to {112}⟨110⟩, {111}⟨110⟩, as well as near {001}⟨110⟩. At the same time the {111} fiber texture in the γ fiber is also stronger than that in IF steel and Fe_3Al based alloy (Fig. 2(b)).

In the α fiber of CuZn alloy the orientation density in the area of Φ= 0°~ 50° becomes more inhomo-

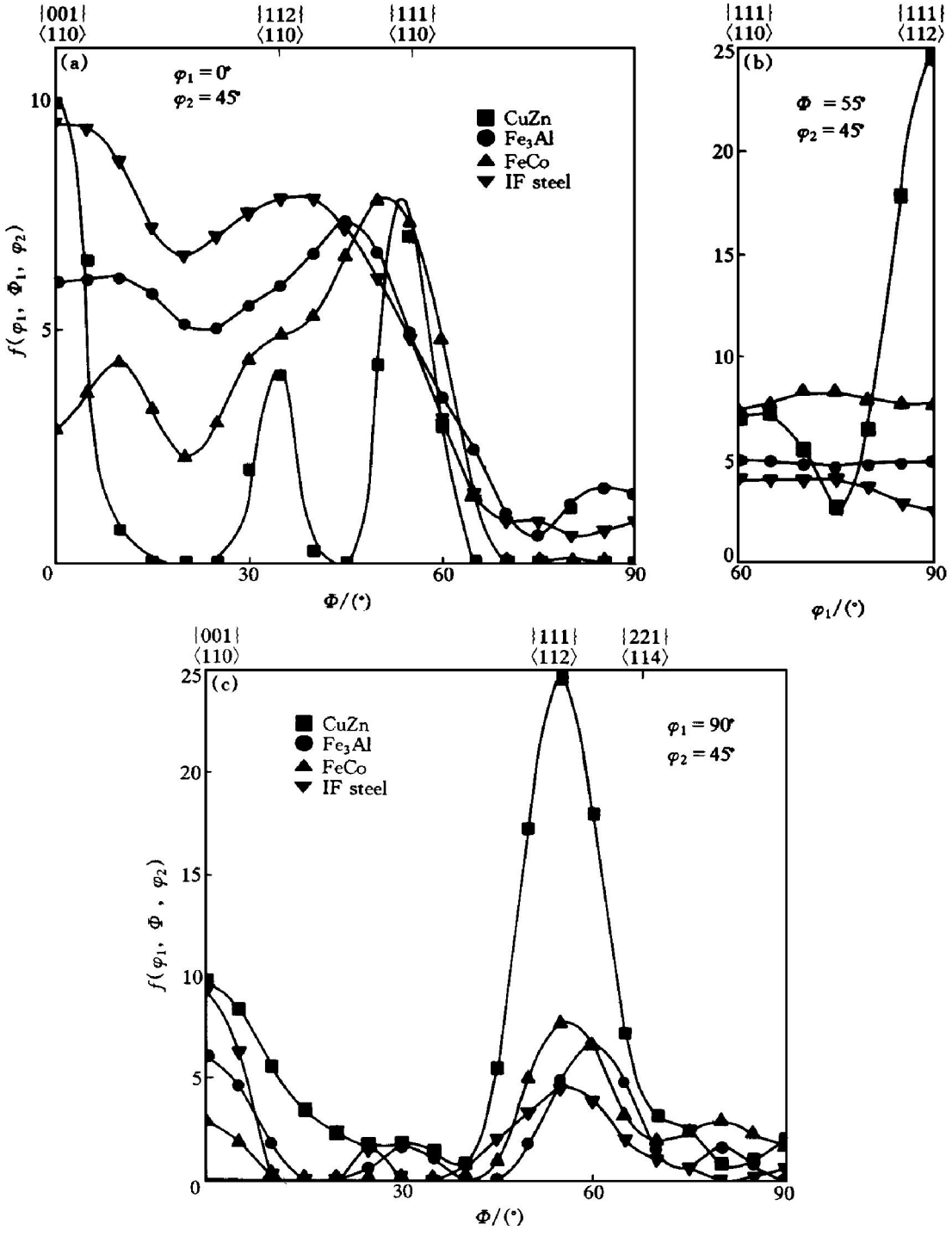

Fig. 2 Characteristics of deformation texture in cold rolled B2 ordered alloys
(a) —α fiber; (b) —γ fiber; (c) —τ fiber

geneous. The grain orientations are accumulated generally around $\Phi = 0^\circ$, $\Phi = 35^\circ$ and $\Phi = 55^\circ$ (Fig. 2 (a)), which correspond to {001} ⟨110⟩, {112} ⟨110⟩ and {111} ⟨110⟩. At the same time the orientation distribution in the γ fiber is very inhomogeneous, i. e. the distribution is emphasized more around {111} ⟨112⟩ (Fig. 2(b)), of which the density value reaches 24. 5.

As above-mentioned, the rolling texture in the alloys with B2 structure has the general characteristics of that in BCC metals, and could be described briefly by the α and γ fibers. However the orientation distribution of the alloys with perfect B2 structure is quite different from that of BCC metals in the α fiber (Fig. 2(a)). Except CuZn alloy, the orientation density are homogeneously distributed in the γ fiber in all other alloys (Fig. 2(b)). The distinguished {111} ⟨112⟩ component in CuZn alloy is related to its special plastic deformation mechanism.

4 DISCUSSION

The texture components in Fig. 2 are those observed frequently in rolling sheets of BCC metals. The {001} ⟨110⟩, for example, is the stable rolling texture component in commercial purity iron, and the {112} ⟨110⟩, {111} ⟨110⟩, {111} ⟨112⟩ etc could also be observed in high purity IF steels[5]. But in general it can be noticed that the orientation density in α fiber demonstrates a continuous distribution with high value in the area of $\Phi = 0^\circ \sim 50^\circ$, which could also be observed in the case of IF steel in Fig. 2(a). However the alloys with B2 structure have quite different density and continuity in orientation distribution, though the grain orientations are also accumulated in the α and γ fibers (Fig. 2(a)).

The rolling deformation simulations indicated[6] that under the condition of stress continuity the {112} ⟨110⟩ should be the most stable orientation in BCC polycrystalline metals, and under the condition of strain continuity the orientation near {111} ⟨112⟩ should be stable by means of activation of 5 independent slip systems. If the strain continuity is a little relaxed, the {111} ⟨110⟩ could also become stable by means of activation of 3 independent slip systems. The theories are based on the fact that the plastic deformation is completed solely by the dislocation slips.

Both continuities of stress and strain will remain in the actual plastic deformation of polycrystalline metals[7]. But only the orientation {001} ⟨100⟩ and {001} ⟨110⟩ could keep the both continuities simultaneously, in which 4 slip systems with the same orientation factor are equally activated during which the corresponding grains will undergo the identical strain and stress to the external strain and stress of the whole rolling sample. However the grain orientation will deviate rapidly when the activation of the 4 slip systems becomes a little unequal. Therefore in the view of stress continuity the {001} ⟨100⟩ and {001} ⟨110⟩ should be only the metastable orientations during rolling, especially the {001} ⟨100⟩ could be hardly observed in rolling sheets[8]. The {001} ⟨110⟩ demonstrated higher stability only for example in the low purity commercial iron because of the impurity atom effect impeding the activation of dislocations and the limitation of strain continuity[5].

For most BCC pure metals the stress continuity could remain if the grain orientations reach the stable α fiber while the grains are deformed by the activation of two slip systems, in which the most strain components are identical to that of the rolling sheet. Supposing that the subscript 2 and 3 are transverse direction and normal directions respectively, only the strain component ε_{23} is not identical to the rolling sheet. In the case of ductile metals, the ε_{23} can be modified or compensated by the activation of additional slip systems in the boundary area under the reaction stress, after which the continuity of ε_{23} can be obtained. If the dislocations could move also on the {112} or {123} planes, the ε_{23} can be modified more easily, which leads to a homogeneous and continuous orientation distribution in α fiber. That is the case of IF steel shown in Fig. 2(a), in which the orientation distribution does not emphasize the {112} ⟨110⟩ very much or is deviated obviously from the {111} ⟨110⟩, which are the theoretically relative stable orientations.

In B2 ordered alloys the boundary energy magnitude of anti-phase domain induced by dislocation slip decides the moving ability of superlattice dislocations and the choice of slip planes[9]. In the three B2 ordered alloys FeCo shows the highest boundary energy of anti-phase domain, which has the lowest value on {110} plane among the {110}, {112} and {123}[10]. Therefore it is commonly believed that the dislocation slips should appear mainly on the {110} planes for FeCo alloy. Basically the slips appear also on {110} planes for the B2 ordered CuZn alloy[11]. The slip mechanism leads to a bad ability of strain modification in the boundary area, and only the orientations which keep the continuities of strain and stress at same time could demonstrate a high stability. The {001} ⟨110⟩[5] and the {111} ⟨110⟩, whose deformation is conducted by the activation of 3 slip systems[6], belong to the orientations which show therefore higher stability in FeCo and CuZn alloys, whose density peaks approach the {111} ⟨110⟩ more than that of IF steel (Fig. 2(a)). The similar analysis is also valid for the higher stability of {111} ⟨112⟩ in the two ordered alloys than that in IF steel (Fig. 2(b)). The grains oriented around {112} ⟨110⟩ have very good stress continuity and stability during rolling[6], but the strain ε_{23} could be relatively high and incompatible,

which could lead to the cracks during rolling, though a density peak around {112}〈110〉 is obtained.

CuZn alloy has very low boundary energy of anti-phase domain, which is about 30% lower than that of FeCo alloy[10]. The anti-phase domain boundary can be an unstrained plane of mechanical twinning. If the dislocation movement is impeded during deformation, the twinning becomes possible in order to change the grain orientation and promote the plastic deformation, especially the low anti-phase domain boundary will reduce the energy increase induced by mechanical twinning. The calculation according to the crystallography of plastic deformation indicates[12], that for {001}〈110〉 the orientation factor is rather low for dislocation slip but very high for mechanical twinning, and the twinning towards {221}〈114〉 has the highest orientation factor and the closest strain components to those of the rolling sheet. In Fig. 2(c) it can be seen that the relative stable orientation {001}〈110〉 has been transformed into {221}〈114〉, and then moved into {111}〈112〉 nearby. Therefore the very strong {111}〈112〉 component is formed in CuZn alloy (Fig. 2(c)).

The Fe-28% Al alloy has incomplete B2 structure[13], which can hardly differ from the BCC structure. $a/2\langle 111\rangle$ dislocation could move rather independently without the obvious limitation of boundary energy of the anti-phase domain, and the {112} as well as {123} could also be the slip planes[14]. Therefore the deformation behavior of Fe-28% Al alloy is very similar to that of BCC pure metals.

5 CONCLUSION

The anti phase domain boundaries could be produced in B2 ordered alloys during cold deformation, which results in the slips of dislocations basically on the {110} planes. Therefore the {001}〈110〉, {111}〈110〉 and {111}〈112〉 texture components were formed, which can provide roughly the continuity of both strain and stress, in which the {111}〈112〉 component was more emphasized by the mechanical twinning in CuZn alloy. The high stable {112}〈110〉 component maintaining the stress continuity could result in cracking during rolling. The structure characters and the crystallographic mechanism of plastic deformation in Fe-28% Al alloy with incomplete B2 structure are similar to those of BCC metals, therefore Fe-28% Al alloy has the similar rolling texture to BCC metals.

ACKNOWLEDGEMENT:

The authors are very grateful to Prof. Gottstein G, Dr. Hu W and Dr. Brückner G for the helps they offered.

REFERENCES

[1] Koehler J S and Seitz F. Proposed experiments for further study of the mechanism of plastic deformation [J]. J Appl Mechanics, 1947, 14(3): A217~224.

[2] Yamaguchi M, Umakoshi Y, Yamane T, et al. Slip systems in an Fe-54% Co alloy [J]. Scripta Metallurgica, 1982, 16(5): 607~609.

[3] Bunge H J. General outline and series expansion method [A]. In: Bunge H J and Esling C eds. Quantitative texture analysis [C]. Oberursel: DGM-Informationsgesellschaft, 1981: 1~72.

[4] Lücke K and Hölscher M. Rolling and recrystallization textures of BCC sheets [J]. Textures and Microstructures, 1991, 14~18: 585~596.

[5] Mao W, Wang X, Zhang S, et al. Formation of annealing texture in rolling IF steel sheet [A]. In: Zhang S, et al eds, HSLA'95, Conf Proc [C]. Beijing: China Sci & Tech Press, 1995: 175~179.

[6] MAO Wei-min. Theoretical calculation of rolling texture in cubic metals [A]. In: Mao W and Zhang X eds. Quantitative texture analysis of crystalline materials [C], (in Chinese), Beijing: Metallurgical Press, 1993, 157~170.

[7] MAO Wei-min. Rolling texture development in aluminum [J]. Chin J Met Sci & Technol, 1991, 7(2): 101~112.

[8] Ridha A A and Hutchinson W B. Recrystallization mechanisms and the origin of cube texture in copper [J]. Acta Metall, 1982, 30(10): 1929~1939.

[9] Takeuchi S. The mechanical properties of ordered alloys [J]. Phil Mag, 1980, A41(4): 541~553.

[10] Stoloff N S and Davies R G. Computer simulation of motion of 〈111〉 superlattice screw dislocation in CsCl-type lattice [J]. Progress in Materials Science, 1966, 13: 12~27.

[11] Yamaguchi M and Umakoshi Y. Dislocation in BCC metals and ordered alloys and compounds with BCC based ordered structures [A]. In: Paidar V and Lejcek L eds. The structure and properties of crystal defect [C], Libkice, Czechoslovkia: Elsevier Science Publisher, 1984, 20: 131~147.

[12] ZHU Guo-hui. Investigation on recrystallization behaviors of cold rolled B2 ordered alloys [D]. (in Chinese), Beijing: University of Science and Technology Beijing, 1998.

[13] Davies R G. An X-ray and dilatometric study of order and "K-state" in iron-aluminum alloys [J]. J Phys Chem Solids, 1963, 24(8): 985~992.

[14] Raabe D and Mao W. Experimental investigation and simulation of the texture evolution during rolling deformation of an intermetallic Fe-28% Al-2% Cr(mole fraction) polycrystal at elevated temperatures [J]. Phil Mag, 1995, 71A(4): 805~813.

(Edited by HUANG Jin-song)