Influence of Fe addition on annealing behaviors of a phosphorus containing brass

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Abstract:

Effect of Fe addition on annealing behaviors of cold rolled Cu-30Zn-0.1P brass was investigated in terms of hardness behavior, microstructure development and microtexture evolution. A maximum increase of about 20% in hardness was observed with the addition of Fe before and after annealing. Fe₂P particles were found dispersed in Cu-30Zn-0.1P-0.5Fe alloy before and after annealing and they showed great pinning effects and resulted in inhibited recrystallization, grains refinement and thermal stability enhancement. It is also indicated by kernel average misorientation (KAM) maps that a higher dislocation density and energy were stored in the sample with the addition of Fe. As a result, the second phase strengthening and grain boundary strengthening were found to be responsible for the hardness increment for the alloy with the Fe addition. Besides, a higher fraction of Brass texture was found at the beginning of annealing in Cu-30Zn-0.1P-0.5Fe alloy. The results also showed that Brass texture then translate into

Brass-R and (55;30;0) textures during annealing, resulting in high fractions of Brass-R and (55;30;0) textures in Cu-30Zn-0.1P alloy with the addition of Fe after annealing.

Keywords: *Cu-Zn-P alloy; recrystallization; second phase particles; EBSD; KAM*

1. Introduction

Cu-Zn brass is known for its good physical and mechanical properties, recyclability and low cost and widely used in applications such as lead frames, connectors and conducting springs. To keep up with the development in such industries in recent years, Cu-Zn brasses with higher mechanical performances and better thermal stability are demanded. Mechanisms for strengthening in alloys usually rely on strategies to improve the density of dislocations or create more internal defects (second phase particles) and boundaries (refined grains) to hinder dislocation motion [1, 2]. It's popular to enhance the strength by refining grains of the materials through severe plastic deformation (SPD) nowadays [3, 4] but the microstructures and properties usually remain unstable, particularly upon heating [5, 6]. Therefore, SPD is difficult to be applied widely in industry. By comparison, introducing second phase particles through alloying is a more effective and convenient way to achieve higher strength and thermally stable performance simultaneously of an alloy [7 - 9]. Iron (Fe) is a precipitation strengthening element for copper alloys and previous studies indicate that appropriate amount of Fe was effective in refining the microstructure and significant enhancement in strength was achieved in copper [10 - 12]. Ye *et al.* [13] have studied the influence of Fe addition on mechanical performances of Cu-Sn-Zn alloy, where Fe particles were introduced by ageing at 773 K for 4 h after solution treatment at 1023 K for 1 h. The ultimate tensile strength of Cu-Sn-Zn-Fe was increased by 45% than those without Fe addition and its thermal stability is enhanced synchronously.

Thus, the authors thought Fe could be added in brass to achieve higher mechanical performances as well as better thermal stability. However, it has been reported that each addition of 0.1wt% Fe can cause a decrease of 4% in elongation in rolled brass [14]. Meanwhile, to avoid undermining the excellent conductivity of the brass, the amount of Fe should be carefully selected [15]. In addition, recovery process from copper scraps is a common link in Cu alloy industry due to the economic and environmental-friendly demands and impurities such as phosphorus (P) is difficult to expel, which would do harm to the conductivity as well [16]. The addition of Fe, in fact, could also forms a nanoscale intermetallic compound with P impurities, which would benefit the comprehensive performance of the brass.

In the present work, Cu-30Zn brass alloy containing P is used. Effects of Fe on hardness behaviors, microstructural evolution and textures development of the Cu-Zn-P alloy were investigated. The amounts of 0.5 wt% for Fe and 0.1 wt% for P in the experimental alloy were selected, considering that intermetallic compounds such as Fe_3P and Fe_2P would be found.

2. Materials and Methods

Cu-30Zn-0.1P and Cu-30Zn-0.1P-0.5Fe alloys (all compositions are in weight percent) were prepared by induction vacuum melting and then hot rolling to eliminate some casting defects and obtain a homogenous and refined microstructure. The two alloy plates were then room temperature rolled with a total thickness reduction by 87% in several passes. The thickness reduction for each pass was about 10% and an annealing treatment at 773 K for 1 h was employed when the reduction accumulated at around 50%. Rolling experiments were conducted on a S815E2 rolling mill with a roll diameter of 420 mm and rolling speed of 0.436 m/s. The cold rolled samples were then

isothermally annealed at 573K for different times ranging from 10 to 36000 s and then immediately quenched in water.

The Vickers hardness tests were measured at room temperature under a load of 1000 g for a loading period of 30 s. At least 9 measurements are made in each case to obtain an average value. Microstructure characterization were all performed primarily on the mid-thickness rolling plane (containing RD and TD) by using optical microscope (OM), scanning electron microscopy (SEM, FEI QuantaTM 650 FEG) equipped with a TSL OIMTM EBSD system and transmission electron microscopy (TEM, FEI Tecnai G² 20). Surfaces for OM observations were mechanically polished and then etched in a mixture of 12.5 g ferric chloride, 12.5 ml chlorhydric acid and 50 ml distilled water. Samples for the EBSD examinations were electropolished using an electrolyte of 70% phosphoric acid and 30% distilled water. Thin foils made for TEM observation were prepared through a conventional jet polishing technique in a solution of 25% nitric acid and 75% methanol at a temperature below -30 degree Celsius. To improve the reliability of the EBSD data, small grains comprising three or fewer pixels were automatically removed from the maps using the grain-dilation option and the map is slightly cleared. Furthermore, to eliminate spurious boundaries caused by orientation noise, a lower limit boundary-misorientation cutoff of 2° was used. A 15° criterion was used to differentiate low-angle boundaries (LABs) and high-angle boundaries (HABs). Grain size was quantified by the determination of the area of each grain and the calculation of its circle-equivalent diameter.

3 Results and Discussion

3.1 Hardness behavior

Vickers hardness-time curves of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys

during annealing at 573K are presented in Fig.1. Three stages can be observed in both curves. With the addition of Fe, higher initial hardness can be detected in Cu-30Zn-0.1P-0.5Fe alloy. Then recovery dominates at the beginning of annealing and small increase of hardness can be seen in both alloys (Stage I), which is frequently referred to as “anneal hardening”. This phenomenon has been largely reported and investigated in Cu alloys containing different alloying elements (Zn, Al, Au, Ga, Pd, Mn, Si) [17-22]. As reported in these papers, solute atoms migrate during annealing and then settle down segregating and locking to lattice defects (dislocations, stacking faults, etc.) introduced by prior cold working, which is analogous to the formation of Cottrell atmospheres in interstitial solid solution during recovery [17-22]. Thus, for the extra addition of Fe, more apparent increase in hardness is observed: the value of hardness is enhanced from 246 Hv to 257 Hv in Cu-30Zn-0.1P-0.5Fe alloy when that is increased from 233 Hv to 241 Hv in Cu-30Zn-0.1P alloy. Then, dramatic decrease due to recrystallization (Stage II) begins in Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys at around 500s and 100s, respectively, and the former alloy showed better thermal stability. Afterwards, decrease in hardness slows down at around 3600 s and 10800 s, respectively (Stage III). Compared to Cu-30Zn-0.1P alloy, the decrease in hardness of Cu-30Zn-0.1P-0.5Fe alloy experiences a two-step slowdown before the hardness reaches a steady state, which might be related to the second phase particles formed with the addition of Fe. And a higher value of 157 Hv is obtained after the long-time annealing treatment, which is enhanced by about 20% compared to that in Cu-30Zn-0.1P alloy. It can be concluded from the curves that Cu-30Zn-0.1P alloy is strengthened before and after annealing and show better thermal stability with the addition of Fe. These kinds of enhancement may be related to microstructural

differences caused by the addition of Fe, which will be discussed later on.

3.2 Microstructure development under annealing

Fig. 2 shows the OM images of Cu-30Zn-0.1P-0.5Fe (a, b) and Cu-30Zn-0.1P alloys (c, d) after cold rolling and subsequent annealing for 10800 s at 573 K. It is clear that fewer deformation bands can be seen in the cold rolled Cu-30Zn-0.1P-0.5Fe alloy with the presence of extra particles (see Fig. 2a). Unlike Cu-30Zn-0.1P-0.5Fe alloy, frequent appearance of deformation bands is found in Cu-30Zn-0.1P alloy, some of which intersected each other (as marked by circles in Fig. 2c), and ultrafine grains with a size of 2 - 5 μm are found at these intersections, which means dynamic recovery occurs during rolling. With the addition of Fe, however, few deformation bands are observed and hardly ultrafine grains can be found, which means dynamic recovery is inhibited compared to Cu-30Zn-0.1P alloy. This may be related to the uniformly dispersed fine particles existed in that alloy before and after annealing (see Figs. 2a and 2b). Considering the content ratio of Fe and P added in the alloy, these particles should be Fe_2P and/or Fe_3P particles. Corresponding TEM images for both alloys are shown in Fig. 3 after annealing for 10800 s, respectively. It can be seen that there is a uniform distribution of very fine particles lying within grains and at grain boundaries with the addition of Fe and the size of second phase particles is about 50 nm ~ 300 nm (Fig. 3a). In addition, the dislocation density in Cu-30Zn-0.1P-0.5Fe alloy is higher and the grains size is smaller and more uniform after annealing for 10800 s. To find out more microstructure details in Cu-30Zn-0.1P-0.5Fe alloy, selected area electron diffraction (SAED) patterns of the matrix and the particles are given (Figs. 3c and 3d), respectively. By calculating the angle and distance between two adjacent transmission spots and diffraction spots, it can be known that the diffraction pattern of the matrix is [110] zone

axis of the face-centered cubic (fcc) crystal, i.e. the α -Cu matrix. Thus, the main difference in microstructure after the addition of Fe elements is the presence of second phase particles. It then can be concluded that these particles should be responsible for any difference in deformation and annealing behaviors of Cu-30Zn-0.1P-0.5Fe alloy. Fig. 3d represents the SAED pattern of these particles. By using the same calibration method, the diffraction pattern is $[02\bar{2}3]$ zone axis of the hexagonal crystal. By calculating the SAED pattern of particles, the lattice constants are $a = b = 0.59$ nm and $c = 0.34$ nm, which is very close to the lattice constants of Fe_2P ($a = b = 0.58660$ nm, $c = 0.34563$ nm) [23]. It can be therefore confirmed that these particles observed in Cu-30Zn-0.1P-0.5Fe alloy are Fe_2P particles.

Typical OIM maps of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys evolved during annealing for 100 s (at around the beginning of recrystallization), 1800 s and 10800 s (at around the end of recrystallization) are shown in Fig. 4. Different colors indicate different crystallographic orientations as defined in the inverse pole figure, and lines indicate various boundaries, i.e. blue and black lines represent LABs and HABs, respectively, while red lines represent $\Sigma 3$ twin boundaries ($60^\circ \langle 111 \rangle$). Changes in average grain size and misorientation angle during annealing are concluded in Fig. 5. For Cu-30Zn-0.1P alloy, similar trends can be seen in the variation of average grain size and misorientation angle and corresponds to the variation of the hardness in Fig. 1 and can be divided into three stages: slow growth stage, sharp growth stage and the steady growth stage, representing the three stages of restoration: recovery, recrystallization and grain growth [24]. However, with the addition of Fe, when the average misorientation angle changes in the same way, recrystallized grains with initial smaller sizes in Cu-30Zn-0.1P-0.5Fe alloy grow in a much slower pace during annealing (which also

confirms that thermal stability of Cu-30Zn-0.1P-0.5Fe alloy is enhanced). It should also be noted that the sharp growth stage is delayed and the average misorientation angle of Cu-30Zn-0.1P-0.5Fe alloy is always lower and grows more slowly, especially for the first 100 s, when recovery dominates. Misorientation distribution details for both alloys at different annealing times are given in Fig. 6. As shown, during the first 100s, the fraction of LABs (2° - 15°) dominates in both alloys with lots of dislocation accumulations remaining. With annealing time increasing, the fraction of HABs (15° - 65°) increases and that of LABs decreases in both alloys but at a more slowly rate in Cu-30Zn-0.1P-0.5Fe alloy. It is worth noting that there are peaks around 60° in almost all the distribution charts. This phenomenon can be attributed the presence of $\Sigma 3$ twin boundaries, which happens on the $\{111\}$ plane with a misorientation angle of 60° . $\Sigma 3$ twins are known as annealing twins commonly observed during processes of recrystallization and grain growth in Cu alloys [24 - 27]. There is a big difference in the fraction of $\Sigma 3$ twin boundaries between the two alloys and fewer $\Sigma 3$ twin boundaries are detected in Cu-30Zn-0.1P-0.5Fe alloy. This difference becomes larger as annealing proceeds. The formation of $\Sigma 3$ twin boundaries are hindered in Cu-30Zn-0.1P-0.5Fe alloy. Also, there is a positive correlation between the speed of grain boundary migration and the emergence of these twins [24, 25, 27], so it can be further illustrated the Fe_2P precipitates can hinder the migration of grain boundaries. Thus, the addition of Fe and resulting Fe_2P particles might have a great effect on the whole restoration process: the Fe_2P precipitates firstly disturb dislocation accumulations from re-arranging or transforming into dislocation-free substructures with HABs and as a result, recrystallization is delayed. Also, due to the pinning effect of second phase particles, boundary migration is inhibited and grain growth is slowed down.

Kernel average misorientation (KAM), one of the functions of OIM, shows the distribution of local misorientation based on a KAM between neighbors on the scan grid. Based on EBSD data, it is a tool in OIM analysis indicating geometrically necessary dislocations (GNDs) density and strain related energy stored in different areas, and vice versa [28 - 30]. The relationship between dislocation density and energy stored [30]:

$$\rho = \alpha\theta/(b\delta) \quad (1)$$

$$E = 0.5\alpha\rho Gb^2 \approx \alpha\theta Gb/(2\delta) \quad (2)$$

Where α is a constant which depends on the geometry of the dislocation arrangement, b is Burgers vector, θ is the value of KAM, δ is step size for the EBSD analysis, ρ is dislocation density and E is stored energy per unit volume, G is the shear modulus. Generally, KAM is high in deformed grains while low in recrystallized grains [30]. The KAM distribution maps of both alloys are given in Fig. 7. Each color represent specific KAM range here, given in the bottom in the Figure. Accordingly, Blue with a KAM θ from 0° to 1° means the lowest local dislocation density areas while the red with a KAM θ from 4° to 5° represents the highest. Figs. 7a and 7d show the cold rolled specimens where lots of areas in green to red with higher KAM are found. But blue areas with low KAM expand with the increase of annealing time. Areas in blue in Cu-30Zn-0.1P alloy are wider than that in Cu-30Zn-0.1P-0.5Fe alloy but KAM in Cu-30Zn-0.1P-0.5Fe alloy is more evenly distributed. A quantitative KAM distribution is concluded in Fig. 8. It is apparent that the peak positions shift to lower KAM value but a higher and more concentrated peak is obtained with annealing time for both alloys. The KAM corresponding to peak position is always higher and the peak is more scattered in Cu-30Zn-0.1P-0.5Fe alloy. This phenomenon might be related to a higher dislocation density remained in the annealed Cu-30Zn-0.1P-0.5Fe alloy (see Fig. 3d)

and Fe_2P particles should be the main cause. As known, the most effective way to eliminate stored dislocation accumulations and strain energy is recrystallization [24]. The lower KAM is, the lower dislocation density and strain energy stored and the more fully recrystallization proceeds. On the other hand, higher KAM corresponds to higher dislocation density and higher energy unreleased. At annealing time of 10800 s, though with almost the same peak KAM position, higher number fraction at the peak was remained in the alloy with the addition of Fe, which means that recrystallization restoration is slowed down in this alloy.

Now let's discuss about the enhanced hardness performance in Cu-30Zn-0.1P-0.5Fe alloy based on microstructural development. Considering the materials and deformation processes, several strengthening mechanisms in Cu-30Zn-0.1P-0.5Fe alloy can be proposed to draw a conclusion: solid solution strengthening, second phase strengthening and grain boundary strengthening. Baker H *et al.* [31] show that the solubility of Fe in Cu at room temperature is less than 0.0004%, which means that Fe is almost insoluble in Cu. As a result, the solid solution strengthening effect can be ignored. According to the difference in microstructures, it can be inferred that the main mechanisms responsible for the strengthening are particles strengthening mechanism due to the dispersed second phase particles and grain boundary strengthening mechanism with fine grains in Cu-30Zn-0.1P-0.5Fe alloy. It's worth noting that extra strengthening effects of dislocation substructures introduced by the particles should not be neglect, since there are still some differences in these substructures between two alloys after being annealed for 10800 s, as seen from the TEM images and KAM results.

3.3 Microtextural changes during annealing

Textures developed in both cold-rolled alloys and their evolution during annealing were studied in this section. Fig. 9 shows distributions of the typical texture components of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys annealed at 573K for different times. The corresponding crystallographic faces and directions of these components are listed in Table 1. These texture components can be classified into two types: rolling texture and recrystallization texture. Changes in area fraction of main texture components and sum area fraction of each type of textures in Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys are concluded as a function of annealing time in Fig. 10. As shown in Fig. 10, with the increase of annealing time, the fraction of recrystallization textures increases and that of rolling textures decreases. Moreover, more remnant of rolling textures is found in Cu-30Zn-0.1P-0.5Fe alloys. This shows again that recrystallization is inhibited with the addition of Fe, where Fe_2P is responsible.

From Fig. 9 and Fig. 10, it is noted that both alloys under annealing are mainly comprised of Brass (red), Brass-R (cyan) and (55;30;0) (pink) textures (as fractions of other rolling textures were small after annealing and little Cube-R (green) texture can be found in Cu-30Zn-0.1P-0.5Fe alloy, seen from Figs. 10a and b). At beginning of annealing, Brass texture dominates in both alloys. For the first 100 s, fractions of $\{011\}\langle 211 \rangle$ Brass, $\{236\}\langle 385 \rangle$ Brass-R and (55;30;0) texture in Cu-30Zn-0.1P-0.5Fe alloy are 33%, 3% and 3% (Fig. 10a), respectively, while in Cu-30Zn-0.1P alloy are 20%, 5% and 6% (Fig. 10b). Then after annealed for 10800 s, fractions of these three textures that are 9%, 22% and 12% in Cu-30Zn-0.1P-0.5Fe alloy (Fig. 10a), which are higher than those in Cu-30Zn-0.1P alloy (7%, 17% and 8%, respectively) (Fig. 10b). Their correlation might be responsible for the special phenomenon that the sum

fractions of these three components are always higher during annealing with the addition of Fe.

Brass texture is usually found as the dominant texture in the deformed microstructures and its formation can be accelerated when dislocation movements are under suppression in Cu alloys [27, 32 - 35] as in the present case with the addition of Fe in the Cu-30Zn-0.1P alloy. The pinning effect of dispersed particles on the dislocation movements in the Cu-30Zn-0.1P-0.5Fe alloy could result in more fraction of Brass texture in the deformed state compared with the deformed Cu-30Zn-0.1P. Upon annealing, these microstructures with Brass type rolling textures would mostly transformed into Brass-R textures [34]. Besides, it is worth mentioning that the (55;30;0) orientation, which is found as a recrystallization texture, is close to that of Brass texture and might also derive from orientation spread of the Brass component during annealing [33]. Therefore, it is reasonable to see that a larger amount of Brass-R texture and (55;30;0) grains is obtained in the annealed Cu-30Zn-0.1P-0.5Fe alloy than in the annealed Cu-30Zn-0.1P one.

4. Conclusions

The effect of Fe addition in the Cu-30Zn-0.1P alloy on grain size and annealing behavior of brasses was studied and can be summarized as follows:

(1) Hardness was increased with dispersed second phase particles due to the addition of Fe, whose main strengthening mechanisms are second phase strengthening and grain boundary strengthening.

(2) Due to the pinning effect of second phase particles on dislocation motion and boundary migration, recrystallization in Cu-30Zn-0.1P-0.5Fe alloy is hindered. Dislocation density decreases slowly, strain energy releases in a low speed and grains

hardly grow during annealing. As a result, significant enhancement of thermal stability was found in the Cu-30Zn-0.1P-0.5Fe alloy.

(3) A higher fraction Brass texture component was seen in Cu-30Zn-0.1P-0.5Fe alloy after cold rolling. And high fractions of Brass-R and (55;30;0) textures were obtained in Cu-30Zn-0.1P-0.5Fe alloy after annealing due to the correlation between Brass, Brass-R and (55;30;0) textures.

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Table caption

Table 1 Several ideal rolling and recrystallization textures for face-centered cubic

metals.

Figure captions

Fig. 1 Variations of Vickers hardness with annealing time at 573K for Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys.

Fig. 2 OM images of Cu-30Zn-0.1P-0.5Fe alloy (a) before and (b) after annealing and Cu-30Zn-0.1P alloy (c) before and (d) after annealing for 10800 s.

Fig. 3 Microstructures of specimens annealed for 10800s: TEM images of (a) Cu-30Zn-0.1P-0.5Fe and (b) Cu-30Zn-0.1P alloys; (c) a [110] selected area electron diffraction (SAED) pattern of matrix and (d) second phase particles in (a).

Fig. 4 Typical OIM maps of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys evolved during annealing at different holding times.

Fig. 5 Variation in average grain size and average misorientation angle of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys with different annealing times.

Fig. 6 Misorientation angle distributions for ranges of 2-65° of Cu-30Zn-0.1P-0.5Fe (a, b, c) and Cu-30Zn-0.1P alloys (d, e, f) annealed for different times.

Fig. 7 Kernel average misorientation maps of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys with different annealing times.

Fig. 8 Kernel average misorientation distributions of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys with different annealing times.

Fig. 9 Distributions of the main texture components within the OIM maps for Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys annealed for different times.

Fig. 10 Annealing time dependence of area fraction of main texture components (a, b) and sum fraction of rolling and recrystallization textures (c, d) in Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys.

Table 1 Several ideal rolling and recrystallization textures for face-centered cubic metals.

Texture types		Euler angles (degree)			Miller indices	
		φ_1	ϕ	φ_2	Rolling plane	Rolling Direction
Rolling textures	Brass	35	45	0	{011}	$\langle 211 \rangle$
	Goss	0	45	90	{011}	$\langle 100 \rangle$
	Cube	0	0	0	{001}	$\langle 100 \rangle$
	Dillamore	90	27	45	{4;4;11}	$\langle 11;11;8 \rangle$
Recrystallization textures	(55;30;0)	55	30	0		
	Cube-RD	0	22	0	{013}	$\langle 100 \rangle$
	Brass-R	80	31	35	{236}	$\langle 385 \rangle$

Fig. 1 Variations of Vickers hardness with annealing time at 573 K for Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys.

Fig. 2 OM images of Cu-30Zn-0.1P-0.5Fe alloy (a) before and (b) after annealing and Cu-30Zn-0.1P alloy (c) before and (d) after annealing for 10800 s.

Fig. 3 Microstructures of specimens annealed for 10800 s: TEM images of (a) Cu-30Zn-0.1P-0.5Fe and (b) Cu-30Zn-0.1P alloys; (c) a [110] selected area electron diffraction (SAED) pattern of matrix and (d) second phase particles in (a).

Fig. 4 Typical OIM maps of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys evolved during annealing at different holding times.

Fig. 5 Variation in average grain size and average misorientation angle of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys with different annealing times.

Fig. 6 Misorientation angle distributions for ranges of 2-65° of Cu-30Zn-0.1P-0.5Fe

(a, b, c) and Cu-30Zn-0.1P alloys (d, e, f) annealed for different times.

Fig. 7 Kernel average misorientation maps of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys with different annealing times.

Fig. 8 Kernel average misorientation distributions of Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys with different annealing times.

Fig. 9 Distributions of the main texture components within the OIM maps for Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys annealed for different times.

Fig. 10 Annealing time dependence of area fraction of main texture components (a, b) and sum fraction of rolling and recrystallization textures (c, d) in Cu-30Zn-0.1P-0.5Fe and Cu-30Zn-0.1P alloys.