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Regular Article

A good strength-ductility match in Cu-Mn alloys with high stacking fault energies: Determinant effect of short range ordering

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ABSTRACT

The tensile tests were conducted to investigate the influence of short-range order (SRO) on the mechanical properties and deformation microstructures in coarse-grained Cu-Mn alloys with high stacking fault energies. Tensile testing results demonstrate that a good strength-ductility match in Cu-Mn alloys can be achieved by increasing degree of SRO (namely increasing Mn content). With increasing SRO of alloys, the deformation micromechanisms were gradually transformed from wavy slip of dislocations to planar slip of dislocations and deformation twinning, resulting in a synchronous improvement of strength and ductility of Cu-Mn alloys.

As is well known, the strength and ductility are natively incompatible in most of materials, i.e., the ductility dramatically decreases with the increase in strength [1]. The low ductility always stems from diminishing work hardening capacity due to an inability to accumulate dislocations [2–4]. A larger work hardening rate can help restrain localized deformation and postpone necking, leading to a greater ductility [5]. Following this idea, there have been some attempts to improve ductility without sacrificing strength [6–11]. For example, Wang and Ma et al. [9] reported that the ductility of face centered cubic (FCC) metals may be enhanced by deforming at a low temperature (77 K) due to suppressed dynamic recovery and resulting efficient storage of dislocations. Liu et al. [10] introduced the heterogeneous planar structures in Cu through the surface mechanical attrition treatment, which results in an overall high strength and high ductility. Although these strategies have been achieved to improve both ductility and strength, the rigorous process conditions are generally required [9–11], and there are as yet not fully feasible methods in industrial applications.

Another strategy reported that the strength and ductility can be improved simultaneously by decreasing the stacking fault energy (SFE) in Cu-Al [12], Cu-Zn [13,14] and Cu-Al-Zn alloys [15]. A lower SFE inhibits cross-slip and makes the dislocation slip mode transform from wavy slip to planar slip [16–18]. Besides, a lower SFE is beneficial to the formation of deformation twins (DTs) [19]. Consequently, the improvement of strength for those alloys results from the enhanced interaction between DT boundaries and moving dislocations, while the incremental ductility is attributed to the enhancement of the strain-hardening rate coming from the suppression of dynamic recovery [12,20].

In contrast to the most Cu-based (e.g., Cu-Al, Cu-Zn, etc.) alloys, whose SFEs decrease obviously with alloy content, the SFEs of Cu-Mn alloys almost keep unchanged within a wide range of alloying composition, and they are approximately equal to the value of pure Cu [21–23]. However, in such high SFE alloys, planar arrays of dislocations were ever observed in Cu-12wt.%Mn [21,24] alloys. Gerold and Karnthaler [25] suggested that the short-range ordered structures in alloys can promote planar slip through a “glide plane softening” effect. In light of this consideration, the strength-ductility match of metallic alloys might be improved by adjusting the degree of short-range order (SRO). Consequently, the study of the deformation mechanism of Cu-Mn alloys with various Mn contents may raise a new strategy for designing the microstructure of materials to achieve a good combination of strength and ductility.

In the present work, five Cu-*x*Mn alloys ($x = 5, 7, 10, 15, 20$ at.%) with different degrees of SRO [26] were chosen as the target materials, and pure Cu was selected for a comparison. The Cu-Mn alloy ingots were fabricated using high purity Cu and Mn by melting them in vacuum in an induction furnace, and then hot rolled into plates of 10.5 mm in thickness at 770 °C. Tensile specimens were spark-cut from those plates into dog-bone shape. The gauge dimension of specimens is 16 mm × 5 mm × 3 mm with a total length of 60 mm. Before mechanical tests, the tensile specimens were annealed at 750 °C for 3–6 h in vacuum

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and water quenched to obtain a homogeneous microstructure. Subsequently, tensile specimens were mechanically ground and then electro-polished to obtain a strain-free and smooth surface.

Tensile tests were carried out on these Cu-Mn alloy specimens at a strain rate of 10^{-2} s^{-1} up to failure at room temperature using a CMT5101 machine. After tensile tests, the microstructural characterizations were performed using a FEI Tecnai G^{2}20 transmission electron microscope (TEM) operated at 200 kV. Thin foils for TEM observations were spark-cut from the gauge parts of tensioned specimens.

Fig. 1 shows the original microstructures of pure Cu and Cu-Mn alloy specimens. No precipitates and inclusions were observed in all materials. The average grain size of pure Cu and Cu-Mn alloys was measured to be in the range of 130–150 μm (excluding annealing twin boundaries). Hence, the effects of grain size and precipitation strengthening on the mechanical behavior of these materials can be ruled out.

Fig. 2a and b show the engineering and true stress-strain curves of pure Cu and Cu-Mn alloys, respectively. Obviously, the tensile curve generally moves up with the addition of Mn. The yield strength of Cu-Mn alloys is enhanced from 40 to 115 MPa as the Mn content increases from 0 to 20 at.%. This is mainly attributed to the increased solid solution strengthening caused by incremental Mn alloying [27]. Meanwhile, the ultimate tensile strength increases from 220 to 370 MPa. Interestingly, although the strength of Cu-Mn alloys is obviously enhanced, the uniform elongation of Cu-Mn alloys increases slightly with Mn content (as highlighted by the red arrow in Fig. 2a). Clearly, the Cu-20.at%Mn has the optimal mechanical properties (i.e. the highest strength and the best ductility) among all materials. The effect of alloying atoms on the improvement of strength and ductility has been reported previously in Cu-Al [12] and Cu-Zn [13,14] alloys. It was suggested that simultaneously enhanced strength and ductility of these alloys was derived from the change of slip mode and increased density of DTs, which were caused by lowering SFE in these alloys through increasing the solute content. However, this explanation is not applicable for Cu-Mn alloys for their SFE values are almost constant with increasing Mn content, as seen in Fig. 2c.

Actually, alloy atoms in a solid solution are distributed over the lattice positions in a not totally random way, and SRO structures are always formed. Usually the degree of SRO in solid solution alloys consisting of A and B atoms is described by [26]

$$\alpha_i = 1 - \left(P^i_{AB}/c_B\right) \tag{1}$$

where α_i is the SRO-parameter for the ith coordination shell, P^i_{AB} the probability to find a B-atom in the ith coordination shell around an A-atom, and c_B is the concentration of B-atoms ($c_A + c_B = 1$). Pfeiler [26] has measured the first short-range order parameters for Cu-Mn alloys by electrical resistivity measurements. As seen in Fig. 2c, there is an obvious increase in the first SRO-parameters of the present Cu-Mn alloys from 0.037 to 0.12 as the Mn content increases from 5 to 20 at.%. Therefore, it is reasonable to speculate that the improvement in the strength and ductility should be primarily attributed to the increase in SRO degree for the present Cu-Mn alloys, whose SFEs keep almost unchanged with the Mn content.

To gain an insight into the effect of SRO on the mechanical behavior, the deformation microstructures of Cu-Mn alloys were investigated by TEM, as representatively shown in Fig. 3. The dislocation structures of Cu-Mn alloys with low Mn contents (5 and 7 at.%) are featured by elongated dislocation cells with a low dislocation density (Fig. 3a, b), which is a typical kind of wavy-slip type dislocation structure [18]; in this case, the Mn content added does not increase the degree of SRO to an enough level to change the dislocation slip mode. A similar result has been reported by Steffens and Schwink [28]. As a result, the enhancement of work hardening rate of these two alloys with less SRO formation is not conspicuous, coinciding well with the subtle improvement of ductility (Fig. 2b).

As the Mn content increases to 10 at.%, the dislocation structures mainly comprise dislocation cells with an increased dislocation density, and the size of these cells becomes smaller (Fig. 3c). The evolution of dislocation density is mainly attributed to the increments of the SRO degree, which hinders cross slip and thereby the recovery of dislocations [24]. Furthermore, a direct evidence is that planar-like slip structures (marked by "A" in Fig. 3d) have been observed in some local regions. Although the dislocation slip mode in Cu-10 at.%Mn alloy is still dominated by wavy slip, it has provided a substantial evidence for the transition of slip mode from wavy slip to planar slip. Meanwhile, it is worth noting that some typical DTs, which are conventionally believed not to form in high SFE metals [29,30], were observed in Cu-10 at.%Mn alloy, as marked by a red cross in Fig. 3d. Quite recently, Hamdi and Asgari [24] observed planar slip structures in Cu-12wt.%Mn single crystals deformed to a 20% strain amount under compression, but no DTs were found. It is well known that a sufficiently high local stress is required to nucleate deformation twinning. Narita and Takamura [31] expressed the twinning shear stress, τ_T as a function of SFE, γ_{SF}

$$\tau_T = \frac{\gamma_{SF}}{2b_p} \tag{2}$$

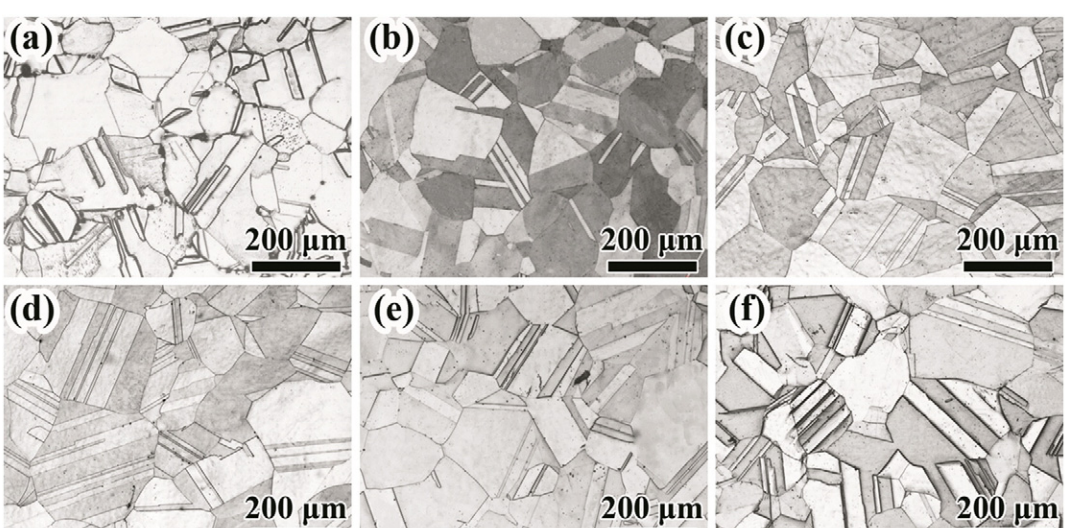

Fig. 1. Optical microscope (OM) images showing the original microstructures of Cu (a), Cu-5 at.%Mn alloy (b), Cu-7 at.%Mn alloy (c), Cu-10 at.%Mn alloy (d), Cu-15 at.%Mn alloy (e) and Cu-20 at.%Mn alloy (f) investigated.