# High strength and utilizable ductility of bulk ultrafine-grained Cu–Al alloys

X. H. An,[1] W. Z. Han,[1] C. X. Huang,[1] P. Zhang,[1] G. Yang,[2] S. D. Wu,[1,a)] and Z. F. Zhang[1,a)]
[1]*Shenyang National Laboratory for Materials Science, Institute of Metal Research, Chinese Academy of Sciences, 72 Wenhua Road, Shenyang 110016, People's Republic of China*
[2]*Central Iron and Steel Research Institute, Beijing, 100081, People's Republic of China*

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Lack of plasticity is the main drawback for nearly all ultrafine-grained (UFG) materials, which restricts their practical applications. Bulk UFG Cu–Al alloys have been fabricated by using equal channel angular pressing technique. Its ductility was improved to exceed the criteria for structural utility while maintaining a high strength by designing the microstructure via alloying. Factors resulting in the simultaneously enhanced strength and ductility of UFG Cu–Al alloys are the formation of deformation twins and their extensive intersections facilitating accumulation of dislocations. © *2008 American Institute of Physics*. [DOI: 10.1063/1.2936306]

The preparation of engineering materials with high strength (hardness) and excellent ductility has been a long-standing and arduous mission for material scientists to satisfy various structural applications. Recently, tremendous enhancement of strength (hardness) has been achieved by creating nanocrystalline (NC) grains or producing ultrafine-grained (UFG) materials by severe plastic deformation (SPD) methods.[1] Disappointingly, nearly all these metals have low ductility far from the practical utility due to their artifacts and saturation of defects originated from processing.[2,3] These limitations deteriorate their ability of accumulating dislocations, and thus, increase the propensity to plastic instability in the early stage of plastic deformation, which plagues nearly all NC and UFG materials.[4] Hence, several strategies were proposed to achieve the relatively high uniform ductility in these metals.[5] However, the extent of enhancement[6,7] and rigorously processing conditions[8–10] were not practicable for the industrial application, or the ductility is a trade-off with strength.[11] Accordingly, it is imperative to improve the ductility without loss of strength through the competitively economical methods.

Except for refining grain size, alloying is another conventional strengthening mechanism by the interactions between solution atoms and moving dislocations, which could potentially be beneficial to the promotion of ductility.[12] Zhao *et al.*[6] recently reported that UFG bronze displayed simultaneously higher strength and ductility than UFG Cu processed by high pressure torsion and followed by cold rolling. However, its uniform elongation (UE) was still below the criteria for structural applications (5%).[8] In this study, three Cu–Al alloys (Cu–2.3 at. % Al, Cu–7.2 at. % Al, and Cu–11.6 at. % Al) with a wide range of stacking fault energy (SFE) (48.5, 28, and 4.5 mJ/m$^2$, respectively),[13] were used because both strength and ductility of Coarse-grained (CG) Cu–Al increase with the addition of Al.[14] Meanwhile, equal channel angular pressing (ECAP) is an effective and economical method among the SPD techniques to produce bulk UFG metals for industrial utility.[1] Thus, the purpose of this study is to investigate whether the introduction of Al element can meliorate the ductility of the bulk UFG Cu–Al alloys processed by ECAP (one pass) to exceed the criteria while retaining high strength.

Figures 1 and 2 show the mechanical properties of both UFG and CG alloys. It can be seen that the ultimate tensile strength (UTS) and Vickers Hardness (HV) of UFG and CG Cu–Al alloys were obviously enhanced through the interactions between moving dislocations and the solute atoms.[15] Compared to the CG metals, as-processed UFG alloys exhibit significantly higher UTS and harness which are consistent with the early studies.[1] It is apparent that, even only processed for one pass, the UTS of UFG Cu–2.3 at. % Al was enhanced by 37.5% (from 240 to 330 MPa), and the UTS of UFG Cu–11.6 at. % Al was nearly twice of the CG counterpart (from 355 to 670 MPa), indicating the significantly higher strengthening extent in Cu–Al alloys than that in pure Cu,[1] as illustrated in Fig. 2(a), and the similar tendency for the hardness was shown in Fig. 2(b). Figure 2(c) displays an increasing tendency of the UE of both CG and UFG Cu–Al alloys. In particular, the UE of the UFG Cu–11.6 at. % Al alloy is as high as 5.3%, which is above the critical ductility required for structural applications.[8] Consequently, static toughness (ST), a parameter comprising of both strength and ductility, was correspondingly improved for UFG and CG alloys due to the introduction of Al. Interestingly, the static toughness of UFG Cu–11.6 at. % Al is

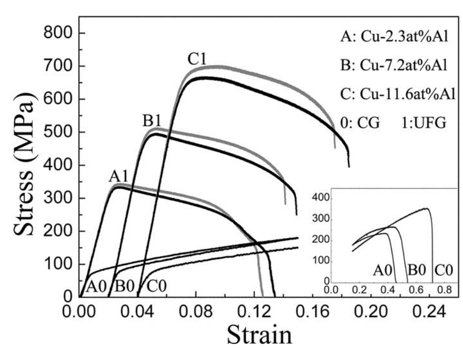


FIG. 1. Typical tensile engineering (black) and true (gray) stress-strain curves of UFG and engineering stress-strain curves (black) of CG alloys with the average grain size of about 350 $\mu$m. Uniaxial tensile tests were performed at a strain rate of $5\times10^{-4}$ s$^{-1}$ at RT for both the CG and UFG specimens. Dog-bone shape specimens have a gauge dimension of $8\times2\times1$ mm$^3$.

a)Authors to whom correspondence should be addressed. Electronic addresses: shdwu@imr.ac.cn and zhfzhang@imr.ac.cn.

 

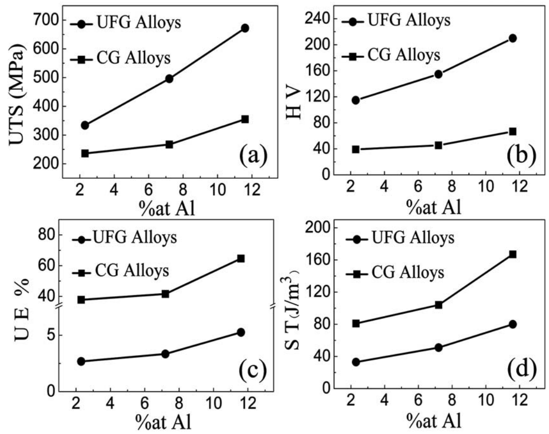


FIG. 2. The dependence of strength, hardness, ductility, and static toughness on the Al concentration in the CG and UFG Cu–Al alloys: (a) UTS, (b) HV, (c) UE, and (d) ST. The HV was measured with loads of 4.9 N for CG alloys and 9.8 N for UFG alloys holding for 10 s.

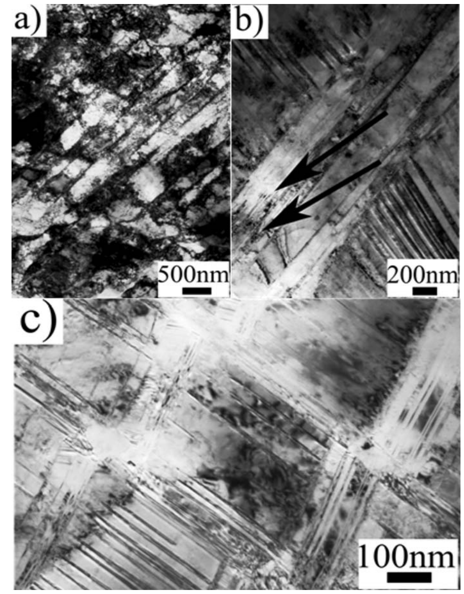


FIG. 3. TEM micrographs of microstructures of UFG alloys processed by ECAP for one pass: (a) Cu−2.3 at. % Al, (b) Cu−7.2 at. % Al, and (c) Cu−11.6 at. % Al. The specimens for TEM were cut from the *Y* plane in the centers of the pressed rods.[16]

nearly the same as that of CG Cu−2.3 at. % Al, as shown in Fig. 2(d). The results above indicate that, with the addition of Al element, the UFG alloys have a simultaneously increasing tendency for main mechanical properties including UTS, HV, UE, and ST. It can be conjectured that this simultaneity may be closely related with the corresponding microstructures.

To decipher the effect of ECAP and Al element on the mechanical behaviors of Cu–Al alloys, it is necessary to investigate their microstructures and the corresponding deformation mechanisms with respect to these properties. Figure 3 shows typical transmission electron microscope (TEM) images of three UFG Cu–Al alloys processed by ECAP. As mentioned above, the addition of Al element lowers the SFE, changing the dislocation activities and facilitating deformation twinning. Since the SFE of both pure Cu and Cu−2.3 at. % Al is relatively high, the microstructure of UFG Cu−2.3 at. % Al is analogous with that in UFG Cu,[16] as displayed in Fig. 3(a), where mainly parallel laminar boundaries with the average subgrain size along the minor axes about 200−500 nm existed. For Cu−7.2 at. % Al with moderate SFE, thin deformation twins with average thicknesses of 40−60 nm and shear bands were frequently observed, and there were some intersections between twins and shear bands. Meanwhile, smaller deformation twins were localized in the shear bands because of a high local stress which resulted from the ECAP process,[16] as indicated by the arrows in Fig. 3(b). For UFG Cu−11.6 at. % Al with the lowest SFE, its microstructure mainly consisted of high density of deformation twins with an average thickness of 20−30 nm and extensive intersections of deformation twins, as observed in Fig. 3(c). Moreover, well-developed subgrains were not clearly observed due to the presence and interactions of the profuse twins in Cu−7.2 at. % Al and Cu−11.6 at. % Al. The transition of microstructures can be attributed to the effect of SFE on dislocation movement and deformation twinning.[17]

In close-packed structures, the SFE determines the extent of dislocation dissociation, which influences the ease of cross slip and subsequent microstructure.[14] In high-SFE materials, dislocations can easily cross slip, resulting in extensive dynamic recovery. Dynamic recovery plays a crucial role in annihilating dislocations and rearranging them in a lower energy configuration of cell walls. Nevertheless, lowering SFE inhibits cross slip and limits dynamic recovery, which restricts the dislocation motion to form the planar-type dislocation configuration and enhances the propensity to deformation twinning. Also, the lower SFE, the higher density of twins. In addition, during ECAP process, a large plastic strain was imposed, resulting in the formation of shear bands traversing the whole region with profuse twins and their intersections.[17]

The presence of profuse twins can effectively strengthen materials because twin boundaries block the propagation of dislocation slip and decrease the slip barrier spacing, leading to an increased applied stress for further plastic flow.[18] Thus, it is reasonably believed that the elevation of strength of UFG Cu–Al alloys could be primarily attributed to deformation twins. Similarly, high tensile ductility of UFG Cu−11.6 at. % Al alloys with a high strength can also be ascribed to the influence of deformation twinning[9] on strain hardening rate (SHR). A high SHR is crucial for good UE because it can help delay localized deformation[14] and it is controlled by a dislocation storage (hardening) component and a dynamic recovery (softening) component related to dislocation annihilation processes. The presence of profuse deformation twins renders ample room[9] for the storage and accumulation of additional crystalline defects, cuts down the mean free path of slip dislocations, and restricts dynamic recovery, leading to more effective blockage of dislocation slip,[6] and thus, enhancing SHR. Aside from the influence of deformation twinning on SHR, the extensive intersections of deformation twins in a grain will extraordinarily inhibit slip on essentially all slip systems (unlike the primary twins that only affect the non-coplanar slip systems), which promotes higher SHR.[16] Resembling with the influence of higher density of twins on UTS (hardness), it can further enhance SHR (Ref. 19) and improve the ductility. Thus, the good properties of UFG Cu–Al alloys originated from the influence of deformation twins on dislocations activities.

In summary, high strength can be achieved by producing UFG alloys through SPD process, as shown in Fig. 4; how-

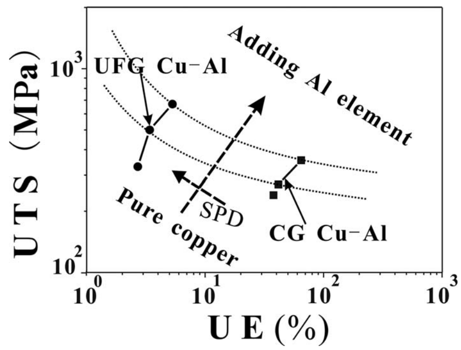


FIG. 4. Relationship between UTS and UE of CG- and UFG–Cu–Al alloys showing the effects of adding Al element and SPD processing.

ever, it is inevitable to deteriorate the ductility only using this technique.[1] Fortunately, the introduction of Al element combining with ECAP rendered simultaneous enhancement in strength (hardness) and ductility. Such a concurrent increase was closely relative to unique features of the formed microstructures. Deformation twins and their intersections play significant roles in simultaneously enhancing the strength and ductility. Designing the microstructure, through increasing the passes of pressing or posttreatment after ECAP, may result in superior properties of UFG Cu–Al alloys. Owing to easiness and relatively economical efficiency of process, it can be applied to produce UFG materials for various structural applications.

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