

Tailoring stacking fault energy for high ductility and high strength in ultrafine grained Cu and its alloy

Y. H. Zhao and Y. T. Zhu^{a)}

Materials Science and Technology Division, Los Alamos National Laboratory, Los Alamos, New Mexico 87545

X. Z. Liao

The James Franck Institute, University of Chicago, 5640 South Ellis Avenue, Chicago, Illinois 60637

Z. Horita

Department of Materials Science and Engineering, Faculty of Engineering, Kyushu University, Fukuoka 819-0395, Japan

T. G. Langdon

Department of Aerospace and Mechanical Engineering, University of Southern California, Los Angeles, California 90089-1453 and Department of Materials Science, University of Southern California, Los Angeles, California 90089-1453

(Received 19 April 2006; accepted 8 August 2006; published online 18 September 2006)

Bulk ultrafine grained (UFG) materials produced by severe plastic deformation often have low ductility. Here the authors report that *simultaneous* increases in ductility and strength can be achieved by tailoring the stacking fault energy (SFE) via alloying. Specifically, UFG bronze (Cu 10 wt. % Zn) with a SFE of 35 mJ/m² was found to have much higher strength and ductility than UFG copper with a SFE of 78 mJ/m². Accumulations of both twins and dislocations during tensile testing play a significant role in enhancing the ductility of the UFG bronze. This work demonstrates a strategy for designing UFG alloys with superior mechanical properties. © 2006 American Institute of Physics. [DOI: 10.1063/1.2356310]

Strength and ductility are two of the most important mechanical properties for structural materials. However, they are often mutually *exclusive*, i.e., a material may be strong or ductile but rarely both at the same time.¹ This is also true for ultrafine grained (UFG) materials produced by severe plastic deformation (SPD),² which usually have high strength but disappointingly low ductility.^{3–5} The low ductility of UFG materials severely limits their practical utility. Therefore, in recent years, much attention has been paid to developing strategies for improving the poor ductility of UFG materials.^{6–17}

The low ductility of UFG materials is attributed to the lack of work hardening caused by their inability to accumulate dislocations because of their small grain sizes and saturation of dislocations.^{4,5,18} Therefore, the basic idea to improve the ductility of UFG materials is to regain the work hardening (dislocation accumulation capability), which is often accompanied with sacrifice of strength.⁷ This raises a question: Is it possible to design UFG materials that have *both* high strength and good ductility? Since the mechanical properties of a material are determined by its deformation mechanisms/behavior, any material design should be based on modifying the deformation mechanisms.

In this study, we used UFG Cu and bronze (Cu 10 wt. % Zn) to demonstrate the effect of stacking fault energy (SFE) on the strength and ductility of UFG materials. The bronze has a SFE of 35 mJ/m²,¹⁹ which is much lower than that of Cu (78 mJ/m²).²⁰ This makes the bronze deform more readily by twinning than Cu. As the data shown later, the UFG bronze exhibits simultaneously higher strength and

higher ductility than the UFG Cu, suggesting that it is possible to design UFG alloys with high strength and good ductility by tailoring their SFEs.

Copper and bronze disks with thicknesses of 0.8 mm and diameters of 10 mm were processed by high pressure torsion (HPT) for five revolutions at room temperature under a pressure of 6 GPa. These disks were further cold rolled (CR) to a thin ribbon with a thickness of 0.2 mm. Tensile specimens were cut from the ribbon to have a gauge length of 10 mm and a width of 1 mm (see Fig. 1) and were then polished to have a thickness of 0.15 mm. Shown in Fig. 2 are the mechanical tensile behaviors of the UFG Cu and bronze. Figure 2(a) shows that the UFG bronze has significantly higher 0.2% offset yield strength ($\sigma_{0.2}$ =580 MPa) and higher elongation to failure (7.1%) than the UFG copper which has $\sigma_{0.2}$ of 420 MPa and an elongation to failure of 5.1%. More importantly, the uniform elongation, determined by the Considère criterion, of the UFG bronze is 3.8%, which is 73% higher than that of UFG copper (2.2%). Thus, the lower SFE rendered the UFG bronze *simultaneously* higher in strength and ductility than the UFG copper. Figure 2(b) shows that the UFG bronze has higher normalized work hardening rate Θ than the UFG copper which is the reason for the higher

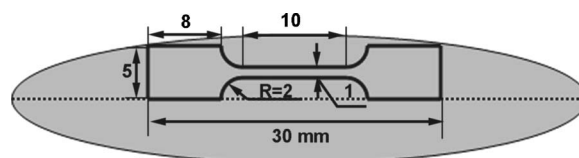


FIG. 1. Schematic representation showing cutting position of a tensile specimen from an UFG ribbon processed by HPT and CR. The thickness of the tensile specimen was polished to 0.15 mm after cutting.

^{a)}Electronic mail: yzhu@lanl.gov

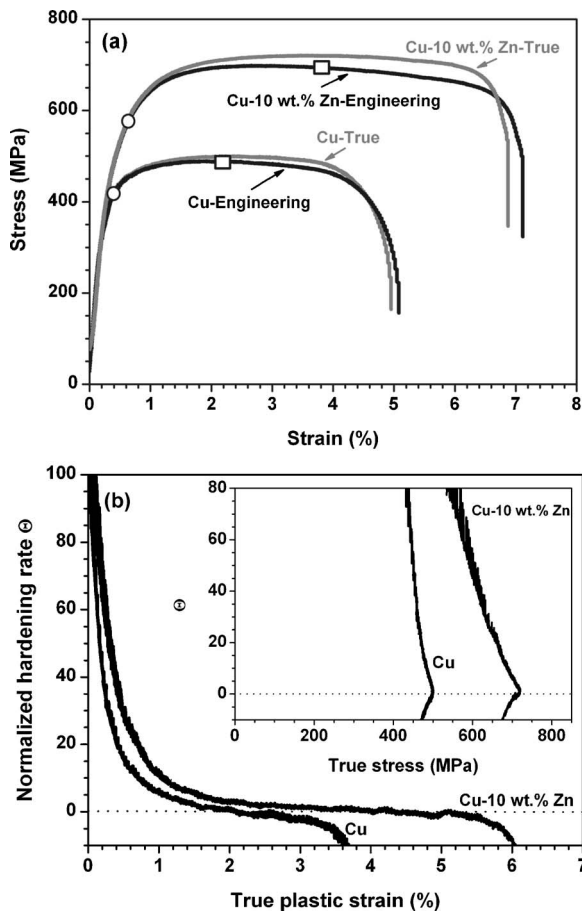


FIG. 2. (a) Tensile engineering and true stress-strain curves of the UFG Cu and bronze. The open squares mark the uniform elongations and the open circles mark $\sigma_{0.2}$. (b) and the inset show the normalized work hardening rate Θ against the true strain and true stress, respectively. The true stress-strain curves and the Θ curves are calculated from the engineering stress-strain curves by assuming a uniform deformation.

ductility in UFG bronze. Θ was defined by $\Theta = 1/\sigma(\partial\sigma/\partial\varepsilon)_\varepsilon$, where σ is true stress and ε is true strain.

To investigate the microstructure and the deformation mechanisms as well as how they are related to the mechanical behavior of the UFG copper and bronze, the gauge sections of the tensile samples were characterized using a Tecnai F30 transmission electron microscopy (TEM) and quantitative x-ray diffraction (XRD) analysis *before* and *after* tensile testing.

Figure 3 shows typical TEM images of (a) the UFG Cu and (b) bronze *before* the tensile tests. The average grain size measured from the TEM micrographs are 180 nm for the UFG copper and 110 nm for the UFG bronze. XRD analysis yielded an average grain size of 70 nm for UFG copper and 50 nm for UFG bronze. It is known that XRD analysis often yields a smaller grain size because it measures the sizes of coherent-diffraction domains.²¹ Under the same HPT+CR processes, the smaller grain size of the UFG bronze than that of the UFG copper indicates that the lower SFE contributed to the grain refinement. Importantly, twins were frequently observed in the UFG bronze [see Fig. 3(b)], but few twins were observed in the UFG copper. In addition, some nano-sized grains in the UFG bronze exhibit a high density of wide stacking fault ribbons formed by dissociated dislocations (see Fig. 4), similar to the wide stacking faults formed in nanocrystalline aluminum.²² By tilting numerous grains to

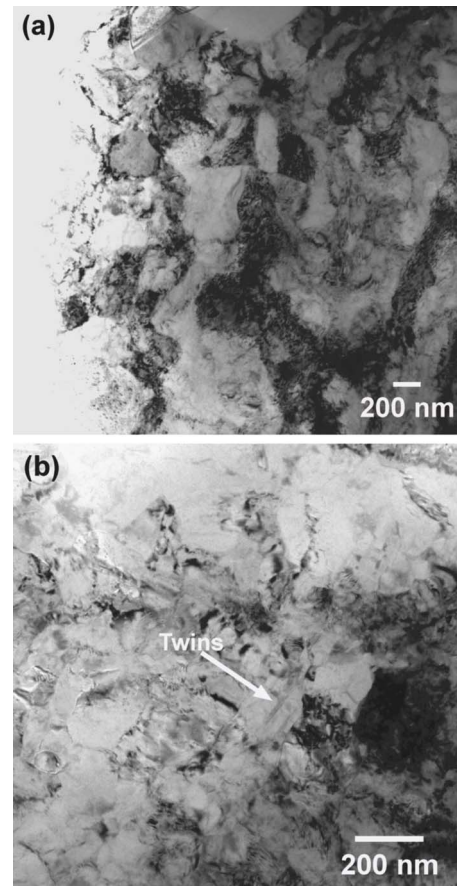


FIG. 3. Typical bright-field TEM images from the gauge sections of tensile samples of (a) the UFG Cu and (b) bronze *before* the tensile tests.

a $\langle 110 \rangle$ zone axis and checking the angle difference between neighboring grains, it was found that the grain boundaries (GBs) of both UFG Cu and bronze are mainly high-angle type, which was further verified by the strong contrast difference among the dark grains (close to zone axis) and their neighboring bright grains (far away from zone axis), as shown in Figs. 3(a) and 3(b). The high-angle GBs of the UFG Cu and bronze were mainly formed during HPT process,²³ and the subsequent CR deformed the GBs, making

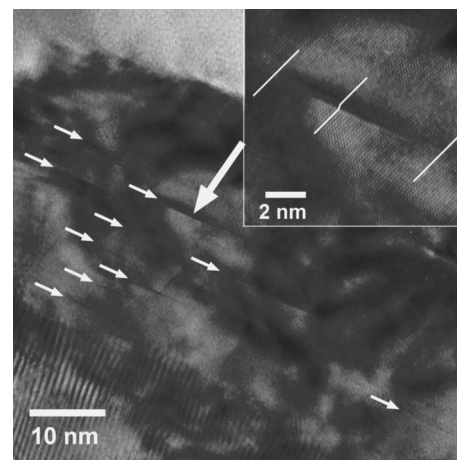


FIG. 4. High-resolution TEM images viewed from the $[110]$ direction showing stacking faults (indicated by small white arrows) formed by dissociated dislocations in the UFG bronze. The inset shows an enlarged image of the stacking fault pointed by a large white arrow.

them wavy. TEM did not reveal a significant change in microstructure after the tensile tests.

XRD analysis on sample gauge sections before and after tensile tests revealed that the dislocation density in the UFG Cu sample increased from 2.3×10^{14} to $2.8 \times 10^{14} \text{ m}^{-2}$ while the twin density β remained constant at 0.1%. β is defined as the probability of finding a twin boundary between two neighboring (111) planes. In contrast, in the UFG bronze, the dislocation density increased from 5.9×10^{14} to $7.3 \times 10^{14} \text{ m}^{-2}$ and β increased from 4.8% to 5.5%. The dislocation and twin densities were calculated according to Refs. 24–26. The above XRD analysis indicates that in the UFG copper the work hardening was caused by dislocation accumulation, while in the UFG bronze both dislocation and twin accumulations contributed to its higher work hardening.

The lower SFE in the bronze affects its mechanical behavior in several ways. First, the lower SFE makes it easier for a full dislocation to split into two partials with a wider stacking fault ribbon between them. This makes it difficult for the full dislocation to cross slip or climb when it encounters a barrier, which hinders the dislocation recovery via cross slip and climb. Therefore, a lower SFE usually leads to a higher Θ . This has been recently predicted by molecular dynamics simulations²⁷ and observed by experiments²⁸ in nanocrystalline materials. Second, a lower SFE makes it easier for deformation twins to form, which is why β in the UFG bronze is 47 times higher than in the UFG copper. Recently, it was found that boundaries of *growth* twins in electrodeposited Cu could act as locations for dislocation accumulation which improved the Θ .^{11,12} Therefore, the high β in the UFG bronze should help with raising its Θ . However, their effect should be less than the growth twins because a high density of dislocations already exists in the as-processed UFG bronze. In addition, since the twins act as dislocation barriers, they also contribute to higher strength. Third, the wide stacking faults (see Fig. 4) formed by dissociated dislocations in the UFG bronze should also help in raising Θ by interacting with other slipping dislocations. In addition, the zinc alloy element used to lower the SFE also causes solution hardening which increases the strength of the UFG bronze.

It is important to point out that the UFG bronze has smaller grain size and higher dislocation density than the UFG copper but still has higher Θ despite the general observations that smaller grain size and higher dislocation density usually lead to a lower Θ in UFG materials. This demonstrates the high effectiveness of the low SFE and deformation twins in increasing Θ . The GBs may have influence on mechanical behavior of UFG materials; however, in present study, they have minor contribution because of the similar high-angle GBs of both UFG Cu and bronze.

The increase in β in the UFG bronze during tensile testing is also significant because the accumulation of twins leads both to more effective blockage of dislocation slip and to more dislocation accumulation sites, both of which will result in a higher Θ . Because the UFG structure was produced by SPD, it is reasonable to anticipate that β may reach a saturation level in the as-synthesized state if the SFE is so low that twinning becomes a primary mechanism for grain refinement during SPD. This will then make it impossible to further increase β during the tensile tests. Indeed, research in progress suggests that there is an optimum SFE energy that yields the best ductility in UFG Cu alloys.

It has been reported that in coarse grained alloys the strength increases with decreasing SFE, but this increase in strength is accompanied by slightly decrease in Θ and ductility.²⁹ This is due to the fact that the coarse grained materials with high SFE can maintain high Θ via dislocation accumulation. In contrast, the UFG Cu studied here lost most of its work hardening capability, while the UFG bronze can maintain some work hardening capability because its lower SFE activated deformation twins and stacking faults during its processing and tensile testing.

In summary, because of its low SFE, the UFG bronze has both higher strength and higher ductility than the UFG copper. The higher ductility of the UFG bronze is derived from its higher Θ , which is caused by its low SFE, high twin density, wide stacking faults formed by dissociated dislocations, and continuous twin and dislocation accumulations. The higher strength of the UFG bronze is derived from its smaller grain size, higher twin density, and dislocation density, and solution hardening. This study indicates that tailoring SFE by alloy design can be an effective strategy to produce UFG materials with both high strength and ductility. It also indicates that deformation twinning is a deformation mechanism that could improve Θ and ductility of UFG materials.

¹R. Z. Valiev, *Nature (London)* **419**, 887 (2002).

²R. Z. Valiev, Y. Estrin, Z. Horita, T. G. Langdon, M. J. Zehetbauer, and Y. T. Zhu, *JOM* **58**, 33 (2006).

³C. C. Koch, *Scr. Mater.* **49**, 657 (2003).

⁴Y. T. Zhu and X. Z. Liao, *Nat. Mater.* **3**, 351 (2004).

⁵D. Jia, Y. M. Wang, K. T. Ramesh, E. Ma, Y. T. Zhu, and R. Z. Valiev, *Appl. Phys. Lett.* **79**, 611 (2001).

⁶R. Z. Valiev, I. V. Alexandrov, Y. T. Zhu, and T. C. Lowe, *J. Mater. Res.* **17**, 5 (2002).

⁷Y. Wang, M. Chen, F. Zhou, and E. Ma, *Nature (London)* **419**, 912 (2002).

⁸E. Ma, *JOM* **58**, 49 (2006).

⁹K. M. Youssef, R. O. Scattergood, K. L. Murty, J. A. Horton, and C. C. Koch, *Appl. Phys. Lett.* **87**, 091904 (2005).

¹⁰B. Q. Han, Z. Lee, D. Witkin, S. Nutt, and E. J. Lavernia, *Metall. Mater. Trans. A* **36**, 957 (2005).

¹¹L. Lu, Y. Shen, X. Chen, L. Qian, and K. Lu, *Science* **304**, 422 (2004).

¹²E. Ma, Y. M. Wang, Q. H. Lu, M. L. Sui, L. Lu, and K. Lu, *Appl. Phys. Lett.* **85**, 4932 (2004).

¹³Z. Horita, K. Ohashi, T. Fujita, K. Kaneko, and T. G. Langdon, *Adv. Mater. (Weinheim, Ger.)* **17**, 1599 (2005).

¹⁴Y. B. Lee, D. H. Shin, K. T. Park, and W. J. Nam, *Scr. Mater.* **51**, 355 (2004).

¹⁵Y. H. Zhao, X. Z. Liao, S. Cheng, E. Ma, and Y. T. Zhu, *Adv. Mater. (Weinheim, Ger.)* **18**, 2280 (2006).

¹⁶H. W. Kim, S. B. Kang, N. Tsuji, and Y. Minamino, *Acta Mater.* **53**, 1737 (2005).

¹⁷H. W. Höppel, J. May, and M. Göken, *Adv. Eng. Mater.* **6**, 781 (2004).

¹⁸Z. Budrovic, H. Van Swygenhoven, P. M. Derlet, S. V. Petegem, and B. Schmitt, *Science* **304**, 273 (2004).

¹⁹A. Howie and P. R. Swann, *Philos. Mag.* **6**, 1215 (1961).

²⁰M. H. Loretto, L. M. Clarebrough, and R. L. Segall, *Philos. Mag.* **11**, 459 (1965).

²¹Y. T. Zhu, J. Y. Huang, J. Gubicza, T. Ungár, Y. M. Wang, E. Ma, and R. Z. Valiev, *J. Mater. Res.* **18**, 1908 (2003).

²²X. Z. Liao, S. G. Srinivasan, Y. H. Zhao, M. I. Baskes, Y. T. Zhu, F. Zhou, E. J. Lavernia, and H. Xu, *Appl. Phys. Lett.* **84**, 3564 (2004).

²³Y. H. Zhao, X. Z. Liao, Y. T. Zhu, Z. Horita, and T. G. Langdon, *Mater. Sci. Eng., A* **410–411**, 188 (2005).

²⁴Y. H. Zhao, H. W. Sheng, and K. Lu, *Acta Mater.* **49**, 365 (2001).

²⁵Y. H. Zhao, K. Zhang, and K. Lu, *Phys. Rev. B* **56**, 14322 (1997).

²⁶J. B. Cohen and C. N. J. Wagner, *J. Appl. Phys.* **33**, 2073 (1962).

²⁷V. Yamakov, D. Wolf, S. R. Phillpot, A. K. Mukherjee, and H. Gleiter, *Nat. Mater.* **3**, 43 (2004).

²⁸F. Ebrahimi, Z. Ahmed, and H. Li, *Appl. Phys. Lett.* **85**, 3749 (2004).

²⁹J. W. Simmons, *Acta Mater.* **45**, 2467 (1997); A. Rohatgi, K. S. Vecchio, and G. T. Gray III, *ibid.* **49**, 427 (2001).