Heterostructure induced dispersive shear bands in heterostructured Cu

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Here we report the formation of dispersive shear bands in a heterostructured Cu composed of coarse-grained (CG) and ultrafine-grained (UFG) domains. Microscale digital image correlation revealed that dense shear bands evolved in a stable manner over the whole gauge section. Our observation suggests that the limited strain hardening of UFG domains and deformation heterogeneity promoted shear banding, which is a major deformation mechanism in heterostructured materials. The dispersive shear banding helps with ductility retention.

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Due to their low strain hardening capability, ultrafine-grained (UFG) and nanostructured (NS) metals under tensile loading often become unstable soon after yielding by forming individual catastrophic shear bands (SBs) [1]. Efforts for regaining the ductility of these materials by annealing are usually accompanied by a significant sacrifice of strength, i.e., a tradeoff between strength and ductility [1,2]. As a common deformation phenomenon in high-strength NS metals under tension, shear banding can sustain severe local plastic strain [3]. However, for conventional homogeneous UFG or NS samples under quasi-static tensile straining, well-developed SBs can generally form only in the region near fracture, which are unstable due to the lack of strain hardening to deter their propagation [4–6]. Therefore, shear banding is often associated with failure for UFG/NS metals.

It has been recently reported that heterostructures with coarse-grained (CG) soft domains embedded in UFG or NS matrix can produce high strength while retaining reasonable ductility [7–13]. For example, the lamella Ti composed of UFG matrix and CG lamellae was as strong as UFG Ti, and at the same time its ductility was even larger than homogeneous CG Ti [8]. Heterostructures synthesized by abnormal grain growth in cryogenically deformed matrix retained excellent work hardening at limited strength loss [11–13].

A critical question arisen here is how the heterostructure accommodates the large uniform applied strain. Strain partitioning where the softer domains sustain higher plastic strain has been observed [8,10]. It has been reported that the synergistic deformation of heterostructured domains induces pile-up of geometrically necessary dislocations near domain interfaces, which contributes to the development of back stress and extra strain hardening [8–11,14–16]. However, strain accommodation mechanism in different domains has not been well studied. Our recent experiments in NS/CG laminates revealed that the synergistic constraint between layers induces highly-dispersed stable SBs in the NS layers [14]. This suggests a new synergistic deformation mechanism. But it is not clear whether this also occurs in heterostructures composed of randomly arranged heterostructured domains.

To study the above issue, the plastic behaviors of UFG/CG bimodal-grained heterostructures are probed here using microscale digital image correlation (μ-DIC). Commercial Cu (99.9 wt%) rod after equal channel angular pressing for eight passes via route Bc was cut into plates with a thickness of 4 mm, followed by rolling to a final thickness of 0.5 mm. Thereafter, three groups of as-rolled samples were annealed at 140 °C for 2 h, 140 °C for 6 h, and 230 °C for 2 h, respectively.

Microstructure of the as-processed samples was characterized using electron back-scattered diffraction (EBSD) under an FEI Quanta 3D FEG instrument. Dog-bone shaped tension specimens with gauge dimension of 12 × 2 × 0.5 mm³ were cut along the rolling direction and uniaxially tested at a quasi-static rate of 5 × 10⁻⁴ μm/min. μ-DIC was conducted in a JSM-6510LV electron microscope. Detailed speckle preparation, imaging and calculation procedures can be found in [14].

Fig. 1 shows the microstructure and grain size distribution of as-processed samples. The microstructure of as-rolled sample is characterized by homogeneous UFG with an average grain size of ~0.89 μm (Fig. 1(a1) and (a2)). During annealing treatment at 140 °C, some recrystallized grains grew up to form coarse-grains, while the matrix
maintained largely the UFG structure (Fig. 1(b1) and (c1)), i.e., partial recrystallization occurred to form a heterostructure with bimodal grain size distribution (Fig. 1(b2) and (c2)). The area fraction of recrystallized CG in samples annealed at 140 °C for 2 h and 6 h were statistically measured as ~30% and ~70%, respectively. In contrast, complete recrystallization occurred in the sample annealed at 230 °C for 2 h, forming homogeneous CG structure with the grain sizes ranging from 5 μm to 15 μm (Fig. 1(d1) and (d2)).

Note that the size ranges of grains in the UFG and CG domains are approximately the same to the grain sizes of homogeneous UFG and CG samples, respectively. For simplicity, the two types of heterostructures are labeled as Cu30% and Cu70%, respectively, where the subscripts x% represent the area fraction of CG domains.

The tensile properties of the sample in varying processing states are compared in Fig. 2. The great differences in both strength and ductility between UFG and CG samples suggest significant mechanical incompatibility between the constituent domains in heterostructured samples. Remarkably, Cu30% and Cu70% samples displayed not only relatively high yield strengths, but also excellent uniform elongations. This implies that the microstructural heterogeneity induced an unusual deformation process for heterostructured samples.

The strain configurations across multiple CG and UFG domains in heterostructured samples are compared with that of homogeneous materials in Fig. 3. The μ-DIC characterization in homogeneous UFG sample was focused in an area in the necking zone but still far away from fracture, whereas in other samples it was performed in the uniform section.
As shown, dense SBs in widths of several micrometers, oriented at the −3±50° direction with respect to the loading axis, are dispersed in heterostructured samples (Fig. 3(a1–a2) and (b)). These SBs have significant strain accumulation but none of them carry enough strain localization to fail the sample. Therefore, these SBs can be referred to as stable SBs. However, such shear band configuration has never been reported in the gauge section of homogeneous FCC metals under tension. Although embryonic SBs with weak strain accumulation are also formed in homogeneous UFG at the position that experienced an ultimate strain of −4.4% (Fig. 3(c)), they have no chance to accumulate more plastic strain due to the stress relief caused by individual catastrophic SB at the most serious necking position [4–6].

No SB was observed in homogeneous CG (Fig. 3(d)) which displays high work hardening due to the uniform dislocation activity. This implies that the low strain hardening capability in UFG domain is a prerequisite for nucleating SBs. These results demonstrate that the UFG and CG domains need to co-exist in an integral bulk heterostructure to be effective in activating dispersed intense SBs and maintaining their stable evolution.

Owing to the dispersed distribution and stable evolution, SBs evolved stably during the uniform elongation of heterostructured samples, which contributed to plastic strain. Taking the Cu30% sample as an example, at the tensile strain of 12.75%, the shear banding region accounts for ~64% of gauge area, and the strain in shear banding region accounts for ~82% of total applied strain. These results indicate that dispersive shear banding is the main strain accommodation mechanism in such heterostructured material. Comparing Fig. 3(a2) with Fig. 3(b) reveals that both the density and strain intensity of SBs in the Cu30% sample are higher than those in the Cu70% sample. This might be the reason that the elongation of Cu30% sample is comparable to that experienced by the Cu70% sample although its volume fraction of CG domain is far less than that in the Cu70% sample (Fig. 2).

Note that dispersed SBs were also experimentally observed in NS/dendrite composited Ti [17] and gradient Cu [27], stimulated in the NS surface layer of gradient Cu [28] and ferrite/martensite dual phase steel [29,30]. The common features of these materials are that all of them are composed of multiple domains with significant mechanical incompatibility and low strain hardening capability in the harder domain. This suggests that dispersed shear banding might be a universal strain accommodation mechanism for heterostructures. More experimental and computational investigations are needed to verify this theory.

The nucleation of shear banding is a plastic response of local instability, which is preferentially activated at sites with stress concentration, which may be built high enough to surpass the rate of local strain hardening [19–21]. Comparing Fig. 3(c) with (a1) reveals that at similar applied strain (~4.4%) the SBs in Cu30% sample is much stronger and denser than those in homogeneous UFG. Since the partially recrystallized Cu30% sample displayed higher strain hardening rate, this may imply that the Cu30% sample sustained even higher local stress than homogeneous UFG according to the Considère criterion. Such high stress concentration should be a mechanical response induced by structural heterogeneity.

As suggested by the strain concentration sites in Fig. 4(a) and (b), stress concentration sites are formed preferentially on or near domain boundaries. The large difference in yield strength between CG and UFG domains induces a long elastic-plastic transition stage for heterostructured samples, during which the CG domains yield plastically, while the UFG domains continue to deform elastically. The elastic/plastic domain boundary is not penetrable to the glide dislocations in CG domain, which leads to dislocation pile-ups against domain boundaries [8,11], as shown schematically in Fig. 4(c). Such dislocation pile-ups produce long-range internal stresses, i.e., back stress (σ_S) in the CG domain and forward stress (τ_f) in the hard UFG domain [Fig. 4(c)] [22,23]. The back stress acts in the opposite direction of the applied shear stress (τ_S), which strengthens CG domain by reducing the resolved shear stress (τ_S−τ_f) and permits it to sustain higher external stress. The forward stress acts in the direction of applied stress, which leads to a higher total stress (σ_T=σ_S+σ_f) in the UFG domain. Before completely yielding, the long-range internal stresses increase with increasing the number of pile-up dislocations, which contributes to extremely high stress concentration in the boundary zone near the head of dislocation pile-up [23–25].

The domain boundary will become penetrable for gliding dislocations once the total internal stress in the UFG side reaches its elastic limit [26]. This process can be promoted by the forward stress. Under such high stress concentration, local shear instability on the boundary or in the UFG side near boundary can be nucleated once a local softening is triggered by the penetration of plastic events across the heterostructured boundary [19,20]. The UFG domain does not have enough strain hardening capability to arrest the local shear instability, which permits the propagation of local shear instability along the preferred shear orientation. This leads to the nucleation of early SBs near domain boundaries.

In contrast to the development of catastrophic SB in homogeneous UFG materials, the propagating SBs in heterostructured Cu30% and Cu70% samples can be effectively arrested by the neighboring CG domain after cut through the UFG zone [17], as observed above in Fig. 4. The plastic deformation of the CG domain ahead of a propagating shear band lowers the stress intensity at the propagating front, while simultaneously producing strain hardening to stabilize the SBs [18,21]. On the other hand, the confined early SBs can only relieve the stress concentration locally, which provides opportunity to nucleate more SBs in other regions until they are dispersed over the whole gauge section. This appears to be the primary mechanism for the formation of dense and dispersed stable SBs in the heterostructured samples (Fig. 3).

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In summary, heterostructure induced the formation of dense and dispersive stable SBs in UFG/CG heterostructured Cu sample during tensile tests. The low strain hardening capability of UFG domain is a prerequisite for the nucleation of SBs. The long-range internal stress induced
Fig. 3. Strain maps at a low magnification in (a1–a2) Cu_{30\%}, (b) Cu_{70\%}, (c) as-rolled UFG, and (d) homogeneous CG samples, showing the shear banding deformation in heterostructures. $\varepsilon_y$ (the left column) is the strain in tensile loading direction and $\varepsilon_x$ (the right column) is the corresponding strain in sample width direction. The shear bands are warm-colored in $\varepsilon_y$ contour and cold-colored in $\varepsilon_x$ contour. The number in each $\varepsilon_y$ subgraph represents the average true tensile strain applied to the sample.
by elastic/plastic interaction between UFG and CG domains led to high stress concentration around domain boundaries, which promoted the nucleation of SBs from domain boundaries. The CG domain helped with stabilizing the SBs and arresting their propagation, which allowed SBs to be dispersed throughout the gauge section. This work demonstrates the possibility of achieving excellent ductility in high-strength UFG and NS materials by architecting structural heterogeneity.

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References