Subsurface microstructure evolution and deformation mechanism of Ag–Cu eutectic alloy after dry sliding wear

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**A B S T R A C T**

Dry sliding wear behavior of nanostructured eutectic Ag–Cu with grain size ~ 114 nm was investigated using pin-on-disc testing. The subsurface microstructure and texture evolution were characterized by X-ray diffraction, scanning electron microscopy, transmission electron microscopy, nano-beam diffraction, and high angle annular dark field. During sliding wear, plastic deformation transforms the subsurface material into hierarchical microstructures. At depths of 1–15 \( \mu \)m below the sliding surface, a two phase (Ag-rich and Cu-rich) nano-lamellar structure is observed, where the layer thickness decreases from 98 nm at a depth of 15 \( \mu \)m to 11 nm at a depth of 1 \( \mu \)m. Right below the sliding surface, where an equivalent strain of 7.1 was estimated, wear induced plastic deformation drives the subsurface material into a non-equilibrium super-saturated solid solution phase, with ~ 9 nm equiaxed nano-grains. These refined microstructures led to significant work-hardening in the subsurface material, as revealed by nanoindentation testing. Finally, the microstructure evolution and the underlying deformation mechanism of Ag–Cu are discussed on the basis of these characterizations.

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1. Introduction

Dual-phase (DP) engineering materials such as \( \alpha/\beta \) titanium alloys, dual phase steels, and Cu–X (X = Ag, Ni, Nb etc.) alloys have drawn considerable attention lately due to their excellent physical, mechanical, and tribological properties [1–4]. Previous studies suggest that the tribological and mechanical properties of DP materials are closely related to their composition, microstructure, and work hardening ability [1,5,6]. Generally, materials with high concentration of the harder phase, fine microstructures, and high work hardening ability of both phases exhibit excellent wear resistance. For example, Tyagi and Ray showed that both wear rate and friction coefficient decrease with increasing volume fraction of the hard martensite phase in a DP steel consists of martensite and ferrite phases [1]. In addition, the elastic/plastic incompatibility between the two phases, as well as the structure and properties of the heterophase interfaces, also plays an important role, especially for DP materials containing one or both phases at the nanometer scale [7–9].

The goal of this work is to study the subsurface microstructure evolution and deformation mechanism of DP nanocrystalline Ag–Cu alloy during wear. The alloy system is chosen for the following reasons. (a) Ag–Cu alloy is an important engineering alloy widely used for its high strength, electrical/thermal conductivity, and oxidation resistance [10–12]. So far very few studies have been performed to investigate their tribological properties, especially for Ag–Cu alloy with nanocrystalline grains [7]. (b) While Ag and Cu are immiscible at low temperatures due to their moderately positive heat of mixing (6 kJ mol\(^{-1}\)) [13], phase separated Ag–Cu alloy can transform to a supersaturated solid solution and even amorphous phase when subjected to intensive plastic deformation or under ion irradiation [13–15]. During sliding wear, the dynamical competition between the phase mixing forced by intensive plastic deformation right below the sliding surface and the phase separation promoted by the thermodynamic driving force may lead to the formation of metastable and/or dynamically stable phases in the tribolayer [16]. Thus a better understanding of the formation of these deformed microstructures could be important as their structure and properties greatly influence material’s wear and frictional response. (c) Ag and Cu are both face-centered-cubic but differ in their twinning tendency due to the difference in stacking fault energies (SFEs are 16 and 45 mJ/m\(^2\) for Ag and Cu respectively). During co-deformation of the two phases, extensive twinning in the Ag phase promotes profuse twinning within the neighboring Cu phase, changing the crystallographic texture in the later one from that of single phase Cu or Cu alloys [17]. In the present context, it will be interesting to investigate the texture evolution of the two phases of Ag–Cu alloy during sliding wear, since both wear rate and friction coefficient are...
closely related to crystallographic texture of the subsurface material [18–20].

The present work constitutes a comprehensive characterization of the deformation microstructure, crystallographic texture, and mechanical properties of the subsurface layer of nanostructured Ag–Cu alloy after pin-on-disc wear. Plastic deformation during wear is quantified by strain and hardness measurements. Finally, the deformation mechanism of the DP Ag–Cu alloy during sliding wear is discussed on the basis of these characterizations.

2. Experimental procedure

Wear tests of eutectic Ag–Cu alloy (Ag–28.1 wt%Cu) were carried out using a Koehler 93500 pin-on-disc (POD) tester. A laboratory frame of reference was defined for the pin material by the sliding direction (SD), the transverse direction (TD) in the sliding plane, and the direction normal to the worn surface (ND), as shown in Fig. 1. Prior to the wear test, the Ag–Cu rod with an initial diameter of 10 mm was swaged along ND at room temperature to 6.5 mm and then cut into several 5 cm long pins. The hardness of the as-swaged Ag–Cu pins was ≈57.7 HRB. A 20.3 cm (8 in.) diameter martensitic stainless steel (SS) 440C disc was used as the counterface for wear testing. Wear tests were performed at room temperature, in air, under 98 N load (or equivalently 2.95 MPa normal pressure), for 1–2.5 h to ensure steady state wear. The sliding velocity was kept at 0.25 m/s to prevent flash heating at the contacting surfaces [21]. Prior to testing, the surfaces of the disc and pins were mechanically polished using SiC polishing papers with grit sizes down to 1200 to achieve an average surface roughness (Ra) less than 1000 nm. The wear rates were estimated by direct measurements of the weight loss of the pin to an accuracy of ±0.1 mg as well as from the linear displacement vs. time curve recorded in-situ by a linear variable displacement transducer (LVDT).

Microstructure characterization was performed using a JEOL 7000F SEM equipped with field emission gun (FEG), X-ray diffraction (XRD) analysis was carried out using XPert X-ray diffractometer (Cu Kα) operating at 40 kV and 30 mA. [111], [200], and [220] pole figures were acquired using a beam size of 2 × 2 cm². Orientation distribution functions (ODFs) were then calculated from the pole figures after background and defocusing correction. Transmission electron microscopy analysis was carried out using JEOL 2010F TEM with either a LaB₆ or Field Emission Gun, operated at 200 kV. TEM samples of as-swaged Ag–Cu were prepared by mechanical polishing followed by ion milling in Gatan PIPS using 5 keV Ar ion beams at an incidence angle of ±12°. Cross-section TEM samples were prepared by standard lift-out method in a FEI dual beam focused ion beam (FIB) microscopy, followed by ion-milling in Gatan PIPS to minimize Ga⁺ contamination. Nano-beam diffraction (NBD) patterns were taken in either conventional diffraction mode with ~10 nm electron beam size or by employing a ~1 nm probe in high angle annular dark-field (HAADF) imaging mode.

The hardness of the subsurface material was measured by nanoindentation on polished cross-section samples (with surface roughness Ra < 20 nm) using a Hysitron Triboindenter with a diamond Berkovich tip. Each indentation was carried out using a trapezoidal load function with a maximum load of 10 mN, a holding time of 3 s, and loading/unloading rate of 1 mN/s. The separation distance of two adjacent indents was kept larger than ten times the maximum indentation depth to prevent interactions between consecutive measurements. The hardness was then calculated using the Oliver–Pharr method [22] from the unloading curve with the instantaneous contact area determined from the calibrated area function of the Berkovich tip. Each reported hardness value was averaged over at least five measurements.

3. Results

3.1. Microstructure of as-swaged Ag–Cu alloy

Results from XRD, SEM, and TEM indicate that the as-swaged Ag–Cu eutectic alloy consists of two fcc phases: one Ag-rich, which contains 14 at%Cu, and the other one Cu-rich, which contains 5 at%Ag (hereafter referred as Ag and Cu phases for simplicity). Fig. 1(a) is a typical SEM image of the alloy, where the bright and dark contrast corresponds to the Ag and Cu phases respectively. The TEM image in Fig. 1(b) shows that both phases are comprised of nanocrystalline grains with aspect ratios ~1:1–1:4, with approximately identical grain sizes of ~110 nm, as measured using the line intercept method from over 3000 grains. Additionally, TEM analysis (results not shown here) shows that two types of special intergranular orientation relationships are frequently observed in this alloy: twin relationship, found between Ag/Ag, Cu/Cu, and Ag/Cu (hetero-twin) phases, and ‘cube-on-cube’ i.e. [100]Ag/[100]Cu, similar to findings previously reported in cast eutectic Ag–Cu [23].

Fig. 2(a) shows the ODFs of the Ag and Cu phases of the as-swaged alloy represented in Euler space using Bunge convention. It can be seen that both phases exhibit {111} fiber texture (ϕ₂ = 55°, ϕ₁ = 45° and ϕ₁∈[0°, 90°]) parallel to the swaging direction (i.e. the normal direction) of the pin material, with intensities about five times random orientations. The (111) fiber, also known as γ fiber, or ‘alloy texture’, is widely observed in low stacking fault energy materials (such as Ag, brass, and Ag–Cu alloy) during rolling, drawing, or sliding at room temperature [12,24,25].

3.2. Tribological behavior

Fig. 3(a) shows that during sliding wear, the height loss of the Ag–Cu pin increases almost linearly with sliding distance. Note a
Fig. 2. Orientation distribution function in Euler space (Bunge convention) of Ag and Cu phases before ((a)-(b)) and after ((c)-(d)) wear test under 98 N load and 0.25 m/s sliding velocity at room temperature in air.

Fig. 3. (a) Wear and (b) friction coefficient measurements as a function of sliding distance during wear test. SEM image of (c) the pin surface, and (d) wear debris generated after wear test under 98 N load, at 0.25 m/s sliding velocity, for 2 h, at RT in air. Arrow in (c) indicates sliding direction of the pin.
running-in period is observed in the first 100 m of the test, where the measured pin position rapidly increases to ∼80 μm and then falls back to nearly zero. This transient is due to the sudden onset of the rotational motion of the pin, which is fixed rigidly in flat contact with the disc, and thus should be excluded from the steady state sliding. The wear rate is \((1.04 \pm 0.2) \times 10^{-6} \text{ mm}^2\), calculated by direct measurement of the weight loss of the pin averaged over three separate tests. Fig. 3(b) shows that the friction coefficient fluctuates around a constant value after the running-in period, indicating the presence of steady state wear. The average friction coefficient is 0.35 ± 0.13. It should also be pointed out that the wear rate of Ag-Cu is about three times smaller than that of a wear resistant CuNiSn alloy, and more than one order of magnitude smaller than nanocrystalline Cu under the same testing condition [26].

Fig. 3(c) shows a typical SEM image of the worn surface, where continuous ridges are observed along the sliding direction, with occasional smearing of metal layers on the surface. Wear debris particles exhibit a bimodal distribution, as shown in Fig. 3(d), where large flaky debris where lar with sizes \(\approx 20–50 \mu m\) coexist with fine granular particles of less than 10 μm. Elemental point analysis using energy dispersive X-ray spectroscopy (EDS) shows a nominal composition of \(\sim \text{Ag–}44 \text{ at}\%\text{Cu}\) of the wear debris, close to the eutectic composition of the sample bulk (Ag–40 at%Cu). In addition, occasional traces of Fe and Cr (typically less than 3 at%) have also been detected on both large flaky and small granular debris, indicating material transfer from the stainless steel disc. No evidence of oxides could be found on the worn surface or wear debris using EDS analysis.

3.3. Microstructural and crystallographic texture evolution after sliding wear

TEM images of the subsurface microstructure of the Ag-Cu pin after wear test are shown in Fig. 4, for both longitudinal (Fig. 4(a)-(d)) and transverse (Fig. 4(e)-(h)) cross-section views. Well below the sliding surface, at a depth of \(\sim 15 \mu m\), plastic deformation during wear leads to the formation of forest dislocations within the grains, as shown in Fig. 4(d) and (h). Approaching the surface, at depths of 2–5 μm, the subsurface grains are deformed into nano-laminates with aspect ratios that often exceed 10 (Fig. 4(b), (c), (f) and (g)). Right below the sliding surface, the subsurface material is further refined to equiaxed...
nanograins of ∼10–20 nm (Fig. 4(a) and (e)). Remarkably, closer examination shows that right below the wear surface, sliding wear leads to the formation of a single phase with a grain size of ∼9 nm, as shown in Fig. 5. The SAD pattern in Fig. 5(c) confirms the presence of a single set of fcc phase, indicating the formation of a super-saturated solid solution of Cu in Ag, far above the equilibrium solubility limit. The lattice parameter of the mixed Ag–Cu is \( a_{\text{Ag-Cu}} = 4.0353 \) Å, about 1.1% smaller than that of the Ag phase \( a_{\text{Ag}} = 4.080 \) Å in as-swaged Ag–Cu, indicating an increase of the Cu solubility in Ag by 8.1 at% according to Vegard’s law. Note that the lattice constant measurement in TEM is calibrated using a pure Cu reference, and special attention was paid to make sure all diffraction patterns were taken at constant objective lens current. Deformation twinning and stacking faults are occasionally observed in both Ag and Cu phases, similar to those observed in the same material under cold drawing [27], rolling [17], and accumulative roll-bonding [28].

The crystallographic texture of the Ag–Cu alloy is examined using both XRD and TEM. XRD analysis uses Cu Kα radiation, and the absorption coefficients of Cu and Ag are \( \mu_{\text{Cu}} = 473 \) cm\(^{-1}\) and \( \mu_{\text{Ag}} = 2270 \) cm\(^{-1}\) respectively [29]. Thus it can be estimated that ∼95% of all the scattering intensity comes from the top 6.6 μm of the sample. Fig. 2(c) and (d) shows the ODFs of the Ag and Cu phases after wear test. It can be seen that the pre-existing {111} fiber texture is completely removed for both phases. To investigate the local crystallographic texture evolution, SAD patterns are taken from the subsurface materials at various depths in TEM, as shown in insets in Fig. 4. The deformed microstructure exhibit no preferred crystallographic orientation at depths of ∼2–15 μm, in agreement with XRD results. Right below the wear surface, a {111} texture is observed in the longitudinal cross-section (Fig. 4(a) inset), similar to those observed in the same material under accumulative roll-bonding (ARB) [28].

### 3.4. Subsurface stress and strain measurement

The hardness of the subsurface material is measured by nanoindentation on polished cross-section. Fig. 6(a) shows that the hardness increases exponentially upon approaching the wear surface from below, with a decay length of 15.3 μm. The maximum hardness increases with depth, as shown in Fig. 6(b).
During co-deformation [6], the thickness of the nano-laminates (which turned into equiaxed nano-grains) right below the wear surface, the strain can only be measured up to ~1 μm below the wear surface using this first method. The second method utilizes grain thickness reduction to calculate the local strains. According to Dautzenberg and Zaat [30], when a spherical grain of initial diameter \(D\) is subjected to simple shear, the equivalent plastic strain can be estimated as

\[
epsilon_{\text{V}_{\text{M}}} \approx \frac{D}{\sqrt{3}c} (D>c),
\]

where \(c\) is the thickness of the sheared grain after deformation. Since Ag and Cu show excellent co-deformation ability (i.e. the amount of strain experienced in Ag phase equals to that in Cu phase during co-deformation) [6], the thickness of the nano-laminates \(c\) is measured without differentiation its phase. In the present work, the values of \(c\) at various depths are measured from HAADF and TEM images over 25 μm below the wear surface (selected values are listed in Table 1). Right below the surface, grain sizes of the equiaxed grains are used instead of nanolaminate thickness. Both results are shown in Fig. 6(b). It can be seen that a sharp strain gradient is developed in the subsurface layer where the equivalent strain increases with decreasing depths, and reaches a maximum of ~7.4 (or 740%) right below the sliding surface. The strain profiles can be fitted using an exponential curve with decay length of ~2.7 μm.

4. Discussion

Sliding wear of eutectic Ag–Cu alloy leads to high shear strain at the contacting surfaces, resulting in intense plastic deformation and work-hardening in the subsurface materials. The measured plastic strain increases exponentially and reaches a maximum of ~7.4 at the wear surface. As a result of this high strain gradient, a hierarchy of microstructures is developed below the wear surface. Far away from the wear surface, at depths ~15–25 μm, the subsurface materials contain high dislocation densities with little grain shape change; at depths ~1–15 μm, the subsurface grains are deformed into nano-laminates with aspect ratios well above 10; right below the wear surface, equiaxed nanocrystalline grains and occasionally, supersaturated single phase Ag–Cu nanocrystalline are observed. These refined microstructures lead to significant work-hardening and a maximum hardness of ~3 GPa at the sliding surface.

For the Ag and Cu eutectic phases with cube-on-cube or twin relationship, their lattice mismatch can be estimated as 
\[
\delta = \frac{(2(a_{\text{Ag}} - a_{\text{Cu}}))/(a_{\text{Ag}} + a_{\text{Cu}}) = 12\%},
\]
where the lattice parameters \(a_{\text{Ag}} = 4.080\ \text{Å}\) and \(a_{\text{Cu}} = 3.628\ \text{Å}\) were measured by XRD from as-swaged Ag–Cu. Besides lattice mismatch, the shear modulus of Ag (\(\mu_{\text{Ag}} = 30\ \text{GPa}\)) and Cu (\(\mu_{\text{Cu}} = 48\ \text{GPa}\)) differ by 60%. These incompatibilities result in a deformation behavior characteristically different than that of the single phase fcc materials. During plastic deformation of Ag, Cu or their single phase alloys at low temperatures, dislocations often rearrange to form subgrain boundaries perpendicular to the slip planes in order to lower the strain energy. These subgrain boundaries then evolve into high angle grain boundaries at higher deformation. Grain subdivision is however not observed in the present study until a strain of ~600%, as shown in Figs. 4 and 6. To confirm the absence of subgrain boundaries in the deformed material, nano-beam electron diffraction (NBD) analysis is carried out at depths of 2 μm below the wear surface (corresponding to a strain of ~350%), as shown in Fig. 8. It can be seen that little variation in crystallographic orientations is present within either the Ag or Cu nano-laminates. Analysis of low magnification TEM and HAADF micrographs indicates that these nanolaminates typically extend over about 400–600 nm at depths of 1–2 μm below the wear surface. Similar microstructures have been reported in two phase Ag–Cu alloys after drawing [27,31], ARB [28], and HPT [3,32]. For example, Ohnsi et al. [28] performed ARB of eutectic Ag–Cu and showed that the Ag and Cu phases are strongly elongated along the rolling direction, and their interphase spacing quickly decreases to a few tens of nm after only two cycles of
deformation, which corresponds to a strain of 4.41. These nano-laminates are present even after five cycles of deformation, or a strain of 6.84, which is the maximum amount of deformation used by these authors.

Another interesting observation is that large scale shear bandings form in the transverse, but not in the longitudinal cross-section of the eutectic Ag–Cu, as shown in Fig. 4. Similar observations have also been reported in wire drawn pearlite, where ‘swirled’ microstructures are observed in the transverse plane (perpendicular to the drawing direction) but not in the longitudinal plane (parallel to the drawing direction) [33,34]. During plastic deformation of a DP composite, shear bands are often observed when the constituting phases exhibit different crystal structure, such as those in Cu/Zr multilayers [28]. For DP materials with similar crystal structure within each phases, the shear banding tendency increases with increasing difference in flow stress and hardening behavior of the two phases [28,35]. It originates from the necking instability of the harder phase and the cooperative deformation of the softer one. In single phase material, the necking instability takes place when \( \frac{d\sigma}{dx} = \sigma \). For materials that work harden following a power law \( \sigma = K\varepsilon^n \), the strain at necking equals the work hardening exponent, i.e. \( \varepsilon = n \), where \( n < 1 \). For DP composite materials, however, the onset of shear banding does not take place at the initial stages of deformation, but often at an intermediate strain. For example, the onset of necking is observed after a strain of 4.41 in cold rolled eutectic Ag–Cu [28]. In the present work, the onset of shear banding takes place at a depth \( \sim 5 \mu m \), or a strain of \( \sim 1.2 \). This delay of shear banding can be rationalized as follows. During the co-deformation of the two phases, in order to maintain strain compatibility, the stress states of the two adjacent layers will be different. When this difference is larger than the shear stress of the harder phase (Cu phase), shear banding takes place. It should also be pointed out that shear banding is suppressed in the longitudinal cross-section where the deformation is dominated by shear along the SD direction, unlike the transverse cross-section where the shear stress component is out of the plane.

The mechanical properties of the subsurface material is investigated by calculating the yield stress from the hardness results in Fig. 6(a) using Tabor’s relationship \( \sigma = H/3 \). Since both stress and strain are measured as a function of depth, the stress–strain relationship can be obtained, as shown in Fig. 9. A power law relationship is found to best describe the work hardening as

\[
\sigma = K\varepsilon^n,
\]

where the fitting parameter \( K = 0.78 \) GPa and the strain hardening exponent \( n = 0.14 \) are obtained by linear fitting of \( \ln(\sigma) \) vs. \( \ln(\varepsilon) \) in Fig. 9 inset. The strength of a variety of crystalline materials as well as multilayered materials varies with the structural length scale following the Hall–Petch relationship, i.e.

\[
\sigma = \sigma_0 + k/\sqrt{d},
\]

where \( \sigma_0 \) is the lattice friction stress, \( d \) is the grain size or layer thickness, and \( k \) is the Hall–Petch constant. For example, Langford showed that the strength of wire drawn pearlite is a function of the square root of the interlamellar or carbide spacing [33]. A similar behavior is also observed for the DP Ag–Cu alloy in the present study. Fig. 10 shows the H–P plot for the nanoeutectic Ag–Cu, overlayed with data from Refs. [11,36,37]. Two regions with different H–P constant are observed. For structural length scales

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**Fig. 8.** (top) Bright-field TEM image of longitudinal cross-section of Ag–Cu alloy \( \sim 2 \mu m \) below the sliding surface, and (bottom) convergent beam diffraction of regions marked 1–8 on the TEM image.
greater than ~50 nm, the Hall–Petch constant $k$ is 0.165 MPa/$\sqrt{m}$, and $\sigma_0=144$ MPa (linear fit I in Fig. 10). When the structural length scale becomes smaller than 50 nm, the stress continues to increase with decreasing of structure length scale, but at a smaller rate with H–P constant of 0.0134 MPa/$\sqrt{m}$, $\sigma_0=860$ MPa (linear fit II in Fig. 10). A summary of Hall–Petch constants of eutectic Ag–Cu, Ag/Cu multilayer, Ag, and Cu is listed in Table 2 for comparison. It can be seen that the H–P constant of the eutectic Ag–Cu in region I is close to that of coarse grained Ag (grain size of 0.67–480 $\mu$m) [38], and is 42% larger than Ag/Cu multilayer, and 24% larger than nanocrystalline Ag. The H–P constant of region II is much smaller than region I, similar to those observed in multilayered Ag/Cu with grain sizes from 1.2 to 50 nm [39] and in multilayered Cu/Nb with layer spacing smaller than 20 nm [40]. It is remarkable that our results cover both regions I and II and fit existing data, which were however confined to only one of these two regions. This suggests that there is a cross-over length scale =50 nm, separating regions with different work hardening mechanisms. Indeed as shown in a recent review by Wang and Misra [41], for multilayers with layer thickness greater than ~50–100 nm, the dislocation pile-up based model is applicable where the stress follows Hall–Petch scaling. For multilayers with layer thicknesses from a few nm to a few tens of nm, dislocation pile-ups cannot form and plastic deformation advances by Orowan bowing of single dislocation loops between interfaces, also known as the confined layer slip (CLS).

Finally, we discuss the driving force for mechanical mixing of Ag and Cu during wear. Right below the wear surface, a supersaturated Ag–Cu phase with ~9.2 nm grains is observed. The alloying of Ag and Cu in the tribolayer is interesting since Ag and Cu has a moderately positive heat of mixing of $\approx 6$ kJ/mol and are immiscible at the solid state [13]. Thus complete mixing is not predicted by equilibrium phase diagram at the wear testing temperatures. In the present work, the mixed Ag phase extended the solubility limit by a factor of ~2.9 compared to the thermodynamic equilibrium. From a more general point of view, it is well known that materials tend to be driven into non-equilibrium status under constant wear and friction [42–44], similar to those in ball-milling [45], under ion irradiation [46], or during severe plastic deformation [28,47]. For example, Shen et al. reported the formation of a nanocrystalline solid solution Cu–50 wt% Ag with an average grain size ~8–12 nm by mechanical alloying and/or severe plastic deformation at room temperature by Shen et al. [50]. During plastic deformation of immiscible systems, two competing forces are at play simultaneously: the thermodynamics driving force that favors phase separation, and plastic deformation that forces atomic scale mixing. The amount of randomization of the mixing depends strongly on the magnitude of heat of mixing [48]. Our present results indicate that, under the sliding wear conditions we employed, forced mixing dominates at the sliding surface, thus resulting in the formation of a chemically mixed, nanocrystalline, single phase. This information can be potentially applied toward the design of engineering metal coatings that will self-develop into a non-equilibrium tribolayer with high hardness and excellent wear resistance.

### 5. Conclusions

The microstructure evolution of eutectic Ag–Cu is studied by advanced characterization techniques. The main conclusions are:

1. Wear induced hierarchical deformation microstructure in the top 15 $\mu$m below the wear surface, Ag and Cu formed nanolaminates due to severe plastic deformation imposed by wear.
2. Wear induced complete mixing between Ag and Cu and the formation of a supersaturated solid solution.
3. SPD during wear tests erased pre-existing texture in Ag–Cu and developed local shear texture.
4. The refined subsurface microstructures lead to a two stage work-hardening and a maximum hardness of ~3 GPa at the sliding surface, which is ~1/3–1/2 of the theoretical strength of Ag and ~1/4 of the theoretical strength of Cu.

### Table 2

Summary of Hall–Petch constants of eutectic Ag–Cu, Ag/Cu multilayer, and microcrystalline and nanocrystalline Ag.

<table>
<thead>
<tr>
<th>Material</th>
<th>Testing</th>
<th>Structural length scale (nm)</th>
<th>H–P constant (Mpa/(\sqrt{m}))</th>
<th>Source</th>
</tr>
</thead>
<tbody>
<tr>
<td>Eutectic Ag–Cu</td>
<td>Nanoindentation</td>
<td>50–180</td>
<td>0.165</td>
<td>Current work</td>
</tr>
<tr>
<td>Eutectic Ag–Cu</td>
<td>Nanoindentation</td>
<td>10–50</td>
<td>0.0134</td>
<td>Current work</td>
</tr>
<tr>
<td>Ag/Cu multilayer</td>
<td>Nanoindentation</td>
<td>50–200</td>
<td>0.116</td>
<td>Mckeown et al. [39]</td>
</tr>
<tr>
<td>Ag</td>
<td>Tension</td>
<td>670–480,000</td>
<td>0.169</td>
<td>Aldrich and Armstrong [38]</td>
</tr>
<tr>
<td>Ag</td>
<td>Microhardness</td>
<td>16–80</td>
<td>0.133</td>
<td>Qin et al. [49]</td>
</tr>
</tbody>
</table>

![Fig. 9](image9.png)

**Fig. 9.** Stress–strain curve for eutectic Ag–Cu. The inset shows the stress and strain in natural log scale.

![Fig. 10](image10.png)

**Fig. 10.** Yield strength vs. layer thickness for Ag–Cu alloys from the current work and Refs. [11,36,37]. The alloy compositions are: $Ag_8Cu_{24}$ multilayers in Verdier et al. [36]; $Ag_{12}Cu_{10}$ multilayers in Huang and Spaepen [37]; eutectic $Ag_{80}Cu_{20}$ in the present work and Benghalem and Morris [11].
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