Title

Anisotropic dynamic compression response of a directionally-cast silver-copper eutectic alloy

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Abstract

At the eutectic composition, the silver-copper bi-phase system can be produced such that the Cu phase forms as discontinuous reinforcements within a silver matrix. The reinforcements and matrix have a $\langle 101 \rangle$ growth direction along the axial direction of the produced material. This work examines the compressive deformation response of four such unidirectional Ag-Cu materials with nominal, bilayer thickness of 3.6 μm , 2.2 μm , 1.1 μm and 500 nm. From the cast material, specimens were machined to investigate the material response parallel (0°) to the $\langle 101 \rangle$ growth direction, 45° to the growth direction, and perpendicular (90°) to the growth direction. Specimens were loaded dynamically under compression using a split-Hopkinson pressure bar. The strength response of the unidirectional material was found to increase as the bilayer thickness decreased with the 0° orientation having the greatest strength followed by the 90° orientation then the 45° orientation. Deformation mechanisms were examined using postmortem microscopy. The orientation dependent strength and strain hardening are attributed to slip system orientations, deformation twinning, and the Ag-Cu interface structure.

Keywords: Eutectic, Ag-Cu, Dynamic plastic deformation, Interface structure

1. Introduction

An attractive method to increase the mechanical properties of materials is by increasing the density of interfaces (e.g., grain boundaries, twin boundaries, phase interfaces) present throughout the material. Grain boundary density is commonly controlled through thermo-mechanical processing routes, such as rolling, equal channel angular pressing [1, 2], high pressure torsion [3] and consolidation methods [4, 5]. However, after microstructure refinement the grain boundaries are often thermally unstable resulting in relaxation of mechanical properties. Previous work on silver processed using high pressure torsion observed relaxation at room temperature over prolonged periods of time [6]. Bi-phase interfaces of low miscibility materials however have been shown to be stable at elevated temperatures at high interfacial densities. Bi-phase interface density can be controlled at the microscale in deposited multilayer systems and at the bulk scale through eutectic casting techniques or accumulative roll bonding [7].

Recent work has focused on the study of the mechanical properties of multilayered deposited metal-metal [8, 9] and metal-ceramic [10] films under nano-indentation [11, 12], tensile loading [13], rolling [14, 15], and micro-pillar compression [16]. These multi-layered thin film material systems are typically manufactured using physical vapor deposition (PVD) to produce layers that are single or polycrystalline [8]. An advantage of PVD is generating highly controlled geometries (e.g., uniform and controllable layer thicknesses), but PVD has the drawback of only being capable of producing films with total thicknesses of a few microns, limiting their structural applications. Many metal-metal and metal-ceramic multilayer systems have been seen to demonstrate a non-linear relationship of increasing hardness with decreasing layer thickness. In most cases, the hardness measure used to estimate yield stress is taken at a compressive strain value well above the traditional 0.2% offset in a uniaxial stress experiment. Therefore, the estimate of yield strength, σ_y , commonly determined through the Tabor [17] relation for hardness, H ($\sigma_y = H/3$) will also be affected by the strain hardening response of the multilayer material system. The underlying mechanisms responsible for the non-linear strength increase have been studied previously – an excellent review is provided by Wang and Misra [18]. The proposed mechanisms

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4 include, the Hall-Petch model [19, 20] for dislocation pileups which determines the logarithmic slope of
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6 the non-linear relation of yield strength when layer thicknesses are greater than 50 nm. For layer
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8 thicknesses below 50 nm modified versions of confined layer slip (CLS) [21] and single dislocations
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10 crossing the interface [21, 22] have been proposed.
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13 While previously conducted work on metallic multilayers has demonstrated that dislocation
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15 mechanisms are influenced by the spacing and structure of bi-phase interfaces, further experimental study
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17 and material characterization particularly of *bulk* materials with a high density of bi-phase interfaces is
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19 necessary for the following reasons: 1) bulk materials offer the ability to study loading orientations
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21 beyond predominantly normal to the interface (as in nano-indentation), and 2) the ability to study material
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23 response over the quasi-static and dynamic regimes. Through the examination of bulk materials with a
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25 high density of bi-phase interfaces additional insight can be gained on how the deformation mechanisms
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27 observed in previous thin film studies translate to bulk materials.
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31 One method to produce a bulk bi-phase material with a high density of interfaces is the casting of
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33 bi-phase eutectics [23-26] with low miscibility, such as Ag and Cu, the bi-phase system of choice for this
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35 study. In flux melt casting a highly purified melt can achieve undercooling levels necessary for the
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37 solidification of sub-micron microstructure [20, 24]. With high undercooling and rapid solidification the
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39 resulting Ag-Cu system microstructure consists of lamellar eutectic colonies. These colonies initiate at
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41 random points throughout the material, hence there is little control over their orientations. By
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43 subsequently growing these colonies directionally with the Bridgman technique [27], the Ag-Cu system
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45 forms a well-ordered directional microstructure [28-30] of Cu reinforcements aligned to the cast rod axial
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47 direction within a Ag matrix. By adjusting the rate at which the melt is removed from the furnace the
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49 length scale of the microstructure can be controlled.
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53 Previous study of directionally cast Ag-Cu material has been limited. The quasi-static material
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55 response of a Ag-Cu eutectic with reinforcement spacing between 100 nm and 3.7 μm was studied by
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57 Cline and Stein [28] with loading parallel to and at 45° to the reinforcement direction, at room and at
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59 elevated temperatures. A linear Hall-Petch type relation was obtained between the reinforcement spacing
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and yield strength, pointing to dislocation pile-ups at the bi-phase interface. Similar linear relationships of lamellar spacing and yield strength have been observed in other directional cast eutectic systems [31, 32]. A recent study by Eftink et al. [30] examined the dynamic compression response of directional Ag-Cu eutectic with microstructure features on the order of a micron. Transmission electron microscopy (TEM) found communication of deformation across the bi-phase interface in the form dislocation slip and deformation twinning. Under the loading conditions used (strain rates on the order of 10^3 s^{-1} at room temperature) deformation twinning would be expected in the Ag phase but not in Cu as Cu only exhibits deformation twinning at very high strain rates or cryogenic temperatures [33-38]. The deformation twinning observed in Cu within the $\text{Ag}_{60}\text{Cu}_{40}$ eutectic, resulted from the transmission of twinning partials from Ag across the cube-on-cube bi-phase interface into Cu [39].

The motivation for the present work is to study the response of a thermally stable directional material system with a high density of interfaces to ascertain if the deformation mechanisms and strength increases observed in deposited multilayer systems can be translated to bulk scale materials. The objectives of this work are to: 1) determine the effect of loading orientation with respect to the solidification direction on the mechanical response, 2) study the effect of decreasing microstructure feature size on yield strength and hardening, and 3) relate deformation mechanisms observed at the meso- and microscale to dislocation slip and twinning occurring at the nanoscale. The following sections of this manuscript present the processing, characterization, and dynamic mechanical response of the directional cast $\text{Ag}_{60}\text{Cu}_{40}$ eutectic material system. A discussion comparing the material response under three loading orientations, 0° , 45° and 90° to the growth, at four different microstructure bi-layer thickness, namely 500 nm, 1.1 μm , 2.2 μm and 3.6 μm , is presented.

2. Experimental Methodology

2.1 Bulk material production

Following a method similar to that of Eftink et al. [31], directionally solidified $\text{Ag}_{60}\text{Cu}_{40}$ specimens were cast. Ag and Cu at the eutectic composition ($\text{Ag}_{60}\text{Cu}_{40}$) were heated to 1223 K within a

vacuum sealed fused silica tube melting the components. The fused silica tube was removed from the furnace at a constant rate promoting directional solidification of the Ag and Cu phases. Three removal rates were used in this study, namely 0.46 mm/hr, 7.3 mm/hr and 73 mm/hr, as a method to control the resulting size scale of the solidified microstructure. These low cooling rates resulted in a material with discontinuous Cu reinforcements (i.e., globular or lamellar platelets) elongated in the axial direction surrounded by a Ag matrix. Both the Ag and Cu phases grew with a common [101] direction parallel to the rod axis [29,31]. Following solidification, the cast rods were vacuum annealed (10^{-6} torr) at 673 K for 4 hours. Annealing was conducted to reduce the solid solution solubility of the phases within each other, approaching solubilities on the order of 0.5 atomic percent [40].

2.2 Experiments: Specimen preparation and loading

Using electrical-discharge machining (EDM), specimens were cut from cast rods such that different loading orientations could be studied, namely parallel to the growth direction (0°), 45° to the growth direction (45°) and perpendicular to the growth direction (90°). A schematic detailing the three selected orientations for dynamic experiments is presented in Figure 1, with arrows indicating the compressive loading directions. Because of the somewhat variable size of the initial casting, the final dimensions of the 0° , 45° and 90° specimens vary but maintain length to diameter ratios of one-half across all orientations. The nominal dynamic specimen diameters used were 8 mm, 7 mm, and 6.8 mm for the 0° , 45° and 90° specimens, respectively. A quasi-static experiment was also conducted on the 500 nm bilayer system parallel to the growth direction (0°). The quasi-static specimens had dimensions of 8 mm diameter and 16 mm length. Prior to testing all EDM surface damage was removed the radial and flat surfaces by mechanical polishing with a final polish using a 1 μm diamond suspension. Specimens were prepared for scanning electron microscopy (JOEL 6060LV) by first sectioning the material using EDM, then casing in a conductive epoxy, followed by mechanical polishing with a final step using a 0.3 μm Al_2O_3 slurry.

Quasi-static compression experiments were conducted on a servo-hydraulic load frame at a strain-rate of 10^{-3} s^{-1} . Machine compliance was accounted for using the method outlined by Kalidindi et al. [44]. Dynamic high strain-rate testing was conducted on a split-Hopkinson (or Kolsky) pressure bar (SHPB) [41]. For this study, the diameter of the striker, incident and transmitted bar were identical at 12.7 mm. In the SHPB a cylindrical specimen is sandwiched between two long elastic bars, the incident bar and transmitted bar. The incident bar is impacted with a striker bar launched via a gas gun initiating a stress pulse that travels along the incident bar length. Strain gauges are mounted on the surfaces of the incident and transmitted bars to measure the propagating stress pulse. The specimen nominal engineering and true stresses, strains and strain rates may be calculated based on one-dimensional elastic wave theory from the strain gauge signals [41-43].

In the one-dimensional elastic wave theory analysis of the strain gauge signals in the split-Hopkinson bar it is assumed that a force balance develops (i.e., forces at both loading ends of the specimen are equivalent) thus placing the specimen in a state of dynamic equilibrium. The raw strain gauge signals for a typical experiment are shown in Figure 2(a) with the incident (ε_I), reflected (ε_R) and transmitted (ε_T) signals labeled. In Figure 2(b) the force balance that occurs during a typical experiment is demonstrated showing equivalent forces on both sides of the deforming specimen. The $Force_{IN}$ trace represents the force at the incident bar specimen interface and the $Force_{IN}$ trace represents the force at the transmitted bar specimen interface. $Force_{IN}$ and $Force_{OUT}$ are determined from the following,

$$Force_{IN} = E_{bar} A_{bar} (\varepsilon_I + \varepsilon_R),$$

$$Force_{OUT} = E_{bar} A_{bar} (\varepsilon_T),$$

where, E_{bar} and A_{bar} are the elastic modulus and cross-sectional area of the incident and transmitted bars, and ε_I , ε_R , and ε_T are the recorded strain gauge signals of the incident, reflected and transmitted wave respectively (see Figure 2(a). Because of the long cylindrical shape of the incident and transmitted bars dispersion of the propagating stress pulse takes place due to different frequencies traveling at

different velocities relative to the bar wave speed. As a result the addition of ε_I and ε_R results in oscillations over the first $\sim 20 \mu s$. The circular marker following these oscillations indicates the point at which equilibrium has occurred. The second circular marker indicates the point after which there is a breakdown of the equilibrium assumption. The nominal engineering strain-rate ($\dot{\varepsilon}_{eng}(t)$), strain ($\varepsilon_{eng}(t)$) and stress ($\sigma_{eng}(t)$), of the specimen can be determined from,

$$\dot{\varepsilon}_{eng}(t) = -(2c_{bar}\varepsilon_R)/l_s,$$

$$\varepsilon_{eng}(t) = \int_0^t \dot{\varepsilon}_{eng}(\tau) d\tau,$$

and,

$$\sigma_{eng}(t) = (E_{bar}A_{bar}\varepsilon_T)/A_s,$$

where, A_s and l_s are the initial specimen cross-sectional area and length, c_{bar} are the bar wave speed, elastic modulus and cross-sectional area of the incident and transmitted bars. The true strain ($\varepsilon_{true}(t)$), strain-rate ($\dot{\varepsilon}_{true}(t)$) and stress ($\sigma_{true}(t)$) are then found using the following assuming plastic incompressibility,

$$\varepsilon_{true}(t) = -\ln[1 - \varepsilon_{eng}(t)],$$

$$\dot{\varepsilon}_{true}(t) = \dot{\varepsilon}_{eng}(t)/(1 - \varepsilon_{eng}(t)),$$

and

$$\sigma_{true}(t) = \sigma_{eng}(t)[1 - \varepsilon_{eng}(t)].$$

The true strain-rate values indicated for dynamic experiments in the following section are determined from the average strain-rate over which equilibrium is maintained (i.e., the region between the black points in Figure 2(b)). Figure 2(c) shows the true strain-rate that occurs during a test with dashed lines indicating regions over which equilibrium does not occur, and the solid line indicating the duration over

which equilibrium takes place. The average true strain-rates reported in the figure legends are determined from the average strain-rate over this region.

3. Results

3.1 Bulk material microstructure characterization

Four microstructure size scales were achieved as a result of the different furnace removal rates. Two castings were conducted with a nominal furnace removal rate of 0.46 mm/hr. The first produced bi-layer thicknesses of 2.2 μm and the second 3.6 μm . Bi-layer thicknesses 1.1 μm and 500 nm were achieved using removal rates of 7.3 mm/hr and 73 mm/hr, respectively. In Figure 2, SEM micrographs of the 500 nm, 1.1 μm , 2.2 μm and 3.6 μm microstructures viewed parallel to the growth direction (Figures 3(a), 3(c), 3(e) and 3(g), respectively) and perpendicular to the growth direction (Figures 3(b), 3(d), 3(f) and 3(h), respectively) are shown. As the nominal bi-layer thickness decreases the Cu reinforcement cross-section transitions from globular platelets to lamellar platelets. Perpendicular to the growth direction all microstructure sizes show discontinuous Cu reinforcements. For clarity, specimens are referred to by their bi-layer thickness throughout this manuscript.

A TEM study of the unidirectional material revealed changes in the orientation relationship of the Ag and Cu phases as the reinforcement thickness decreases corresponding to higher cooling rate [45]. In the coarse microstructures, 2.2 μm and 3.6 μm , the dominant orientation relationship is cube-on-cube with {111} interfaces containing {111} steps [31]. At a bi-layer thickness of 1.1 μm both the cube-on-cube orientation with {111}Cu//{111}Ag interfaces and an incoherent twin orientation relationship are present. The twin orientation relationship regions account for 14.7% by area at the 1.1 μm reinforcement size. At the finest microstructure (500 nm) the incoherent twin relationship area increases to greater than 66% by area.

3.2 Loading parallel (0°) to reinforcement

The uniaxial compressive material response of specimens loaded at 0° to the growth direction, is shown in Figure 4(a); strain rates are indicated in the legend. Results of the 45° and 90° loading orientations are presented in Figures 4(b) and 4(c), respectively, and are discussed in the following sections. Two separate castings of two samples each with microstructure sizes of 500 nm are shown in green, one with up-triangle markers and the other with down-triangle markers. The black quasi-static curve with down-triangle markers is from the same casting as the down-triangle dynamic experiments. The quasi-static response is nearly identical to that of the dynamic response, indicating limited material rate sensitivity at least within this range of strain rates. Unfortunately, due to the limited amount of material available for this study, a complete investigation of the effect of strain-rate on the directionally solidified Ag-Cu material system is not possible. It can however be noted that the non-directional solidified Ag-Cu material system does not exhibit appreciable strain-rate sensitivity [26], a finding that is consistent with the unidirectional quasi-static and dynamic stress strain response in Figure 4(a). As the microstructure is refined, the material strength increases with the 500 nm bi-layers having the highest strength and the 3.6 μm bi-layers having the least strength. The yield strengths vary slightly among samples of the same type due to small changes in the microstructure size along the length of a casting. These variations are inherent to the casting process.

On the specimen radial surface of dynamically deformed specimens, vertical striations parallel to the reinforcement direction are observed. At compressive strains greater than 10% deformation localizes resulting in non-uniform radial deformation. Stereo micrographs of a 0° specimen deformed to 12% compressive strain viewed along the axial direction and from the side are shown in Figure 5(a) and 5(b), respectively. The dashed white circle in Figure 5(a) indicates the initial specimen diameter, and the solid white line indicates the deformed specimen cross-section. The white arrows in 5(a) and 5(b) indicate the position of a vertical striation. The macroscopic deformation shown in Figure 5(a) and 5(b) is typical of the 0° loading orientation and was observed for all reinforcement sizes.

SEM micrographs of the interior of a 3.6 μm specimen loaded to 16% strain are shown in Figure 6(a) and 6(b). The loading direction in both micrographs, as well as through the remainder of the

manuscript, is in the vertical direction. At the length scales observable with SEM the deformed specimen microstructure at the center of the specimen does not show noticeable deviation from the undeformed structure – compare Figure 6(a) to Figures 3(b), 3(d), 3(f), and 3(h). Near the radial surface (~100 μm) where the material is less constrained, kinking of the Cu reinforcements becomes apparent at compressive strains greater than 10% as seen in Figure 6(b). Material near the kinked region is preferentially forced outwards radially and is observed as vertical striations on the radial surface.

3.3. Loading at 45° to reinforcement

The dynamic material response of specimens loaded at 45° to the growth direction is presented in Figure 4(b); strain rates are indicated in the legend. As was observed with the 0° loading orientation there is a yield strength dependence on the microstructure feature size. Specimens with the finest structure (500 nm bi-layers) exhibited yield strengths, between 175 MPa and 200 MPa. Specimens with bi-layer thicknesses greater than 500 nm had reduced yield strengths ranging between 90 MPa and 120 MPa.

Stereo micrographs of the macroscopic deformation of a specimen deformed to 20% strain were captured showing the deformed specimen cross section viewed along the loading direction and the deformation on the radial surface, see Figure 5(c) and 5(d), respectively. When loading in this orientation, non-uniform deformation of the initially cylindrical sample results because of shearing along planes parallel to the [101] growth direction resulting in an elliptical cross-section, see Figure 5(c). The position of selected shear bands are indicated by white arrows in Figure 5(d).

Stylus profilometry was conducted to study this shear banding and to identify possible relationships between band spacing and microstructure feature size. Line contour scans (not shown here) of a 2.2 μm bi-layer reinforcement size specimen deformed to 28% strain and of a 500 nm bi-layer specimen deformed to 20% strain were conducted. The shear band spacing of the material with 2.2 μm microstructure feature size was measured at 295 μm and of the 500 nm material at 95 μm . Although the profiled specimens have different levels of strain, the results illustrate the observed trend that finer microstructured specimens showed smaller spacing of shear bands than the coarser structured materials.

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4 A SEM image of the internal deformation of a 15% compressive strain 3.6 μm reinforcement size
5 specimen is shown in Figures 7(a)-(b). SEM imaging conducted over the specimen interior revealed a
6 uniform structure with Cu reinforcements at 45° to the vertical loading direction, Figure 7(a). At
7 increased magnification, Figure 7(b), waviness of the bi-phase interface can be seen in regions indicated
8 by white arrows. These regions fall along a straight line indicating continuous shearing across both of the
9 Ag and Cu phases. The aligned slip systems in the cube-on-cube orientation throughout the 3.6 μm
10 reinforcement size specimens would allow for such transfer of deformation between phases. As the
11 fraction of twin orientation relationship Ag-Cu interfaces increases (i.e., those found in the 500 nm and
12 1.1 μm specimens) the deformation slip does not transmit across the interface as readily. A SEM
13 micrograph of a 500 nm reinforcement size specimen deformed to 21% strain shown in Figure 7(c) shows
14 straight layers hinting to the twin interface acting as a stronger barrier to the communication of slip
15 between phases.
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33 **3.4. Loading perpendicular (90°) to reinforcement**

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35 The compressive dynamic uniaxial stress-strain response of specimens loaded at 90° to the
36 growth direction is presented in Figure 4(c). As was seen in the 0° and 45° cases, the finest microstructure
37 feature size (500 nm) exhibits the greatest strength with a value of 273 MPa. The yield strength of the 3.6
38 μm and 500 nm reinforcement specimens fall between 100 MPa and 154 MPa.
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44 In response to loading, the 90° orientation specimens showed the highest levels of anisotropy,
45 deforming greatly in the direction perpendicular to the [101] growth direction, as seen in Figure 4(e). For
46 each reinforcement size, striations appear on the loading surface as seen in Figure 4(e). Crossing shear
47 bands are visible on the specimen radial surface when looking down the reinforcement growth direction,
48 see Figure 4(f). From Figure 4(e) it is clear that there is limited macroscopic deformation along the
49 reinforcement direction of the material. SEM images of the internal microstructure of specimens viewed
50 perpendicularly to the growth direction exhibit intact Cu reinforcements.
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Concentrated plastic deformation was observed internally in the form of shear bands in the 90° loading case, see Figure 7. These shear bands occur at 45 degrees to the loading direction. Shear bands initiate in regions where deformation mechanisms such as twinning or dislocation slip are difficult, or require high stresses [46, 47]. As shown in Figure 8 the shear band is present across multiple Ag-Cu interfaces causing significant thickness reductions of the Cu reinforcements and Ag matrix. Typical observed shear band widths are on the order of a few microns.

4. Discussion

The bulk quasi-static and dynamic testing conducted captured a strength dependence on the microstructure feature size and loading orientation. Figure 9 shows the yield strength of each loading orientation with respect to the bi-layer thickness. While previous studies [28, 29, 31,32] have presented yield strengths against the inverse root of a constituent phase thickness it has been chosen to present yield strength here in this manner to steer away from any assumption of a Hall-Petch strength relationship in the Ag₆₀Cu₄₀ material system. TEM study of the Ag₆₀Cu₄₀ material system has found that the occurrence of deformation twinning is dependent on the loading orientation and the bi-phase interface structure [30]. When loading parallel to the growth direction (0°) deformation twinning was not observed, but was prevalent in 45° and 90° loading orientations. In regions where the cube-on-cube interface exists, deformation twinning occurs readily while it is suppressed in regions of the incoherent twin interface [45]. In the following discussion we provide reasoning for the trends in yield strength, macroscopic deformation and strain hardening with respect to loading orientation and bi-layer thickness, as well as address the development and spacing of shear bands in the 45° and 90° loading orientations. Due to limited material availability, a complete investigation of the strain-rate sensitivity of the Ag-Cu material system was unable to be conducted. Future studies involving the use of micro-tension and micro-compression specimens could be used to analyze the effect of strain rate on the mechanical properties reported herein.

It is well known that deformation via dislocation slip occurs prior to deformation via twinning [47, 48]. Previous studies on low stacking-fault energy materials found it necessary to reach strains on the order of 5% before twinning occurs [48]. Therefore the following discussion of the material yield strength is rationalized through arguments of dislocation slip. Through comparison of the slip system Schmid factors of each loading orientation, including crystallographic rotations about the [101] direction, the general trend of the 45° loading orientation showing the lowest yield strength followed the 90° and 0° orientations exhibiting increasing strength can be rationalized. Furthermore, the general anisotropic specimen deformation is also predicted, see Figure 5.

Figure 10 shows the Thompson tetrahedron oriented such that the [101] direction coincides with the vertical loading direction (0°). The slip plane orientations, α (dashed outline), β (coincidental with the front cube face), γ (dotted outline) and δ (dash-dotted outline) are shown. From Figure 10 it can be seen that the β and γ planes are parallel to the loading direction, while α and δ planes are inclined to the interfaces at a constant angle regardless of crystallographic rotation. From the Thompson tetrahedron and Schmid's law it can be deduced that there are only four slip systems experiencing a resolved shear stress, two on the α plane and two on the δ plane, namely $(\bar{1}1\bar{1})[110]$, $(\bar{1}1\bar{1})[011]$, $(111)[\bar{1}10]$, $(111)[10\bar{1}]$. In the cases of loading at 45° and 90° loading orientations the determination of the Schmid factor on each slip system becomes more complex due to the crystallographic rotations about the [101] direction. Due to the symmetry of the FCC structure, Figures 11(a)-(c) provide the Schmid factor for each loading orientation over a rotation of 180°. Vertical lines indicating the crystallographic loading direction are provided in Figure 11(b) (45°), and Figure 11(c) (90°). A comparison of the Schmid factor magnitudes in Figure 11 points to the 45° loading orientation requiring the lowest applied stress to promote slip, followed by the 90°, and lastly the 0° orientation. Thereby recovering the yield strength trend observed at each bi-layer thickness. With decreasing bi-layer thickness each loading orientation saw an increase in the material strength response, see Figure 9. For each loading orientation a reduction from bilayer thickness of 3.6 μm to 1.1 μm resulted in an increase of ~ 50 MPa. Further reduction to 500 nm bilayer thickness

saw an additional yield strength increase of ~50 MPa or greater. The abrupt increase in the yield strength response is proposed to be accounted for by the transition in the structure of the Ag-Cu interface from cube-on-cube to incoherent twin. The cube-on-cube interface acts to hinder slip due to image forces, coherency strain and dislocation content at the interface from dislocation communication from one phase to another [22, 28]. Although the cube-on-cube interface will act as a barrier to dislocation motion the aligned slip systems across the interface are amenable to deformation communication as was seen for deformation twins. In the incoherent twin orientation there is only one shared slip plane and it would therefore be more difficult for deformation communication than in the cube-on-cube structured interface. Based on TEM results, we propose that there is greater difficulty in the communication of slip and twinning across the incoherent twin interfaces than the cube-on-cube interface [45]. For the twin interfaces, dislocations in Cu layers are confined to localized regions indicating strain is not communicated uniformly from the neighboring silver phase. When the orientation relationship is cube-on-cube these localized regions of high strain accumulation are not observed. Therefore it follows that the greater difficult of strain communication between phases could result in the increasing material strength response as seen for all loading orientations in Figure 9.

Examining Figure 11(a) it can be inferred that under a given load, slip is equally likely to occur in any radial direction. Macroscopically, isotropic radial deformation was observed in the 0° loading orientation up to compressive strains of 10%. Beyond 10% strain deformation kinking occurs resulting in non-uniform radial deformation. In the 45° loading orientation the slip systems with the greatest Schmid factor both occur in the [101] direction, namely $(1\bar{1}\bar{1})[101]$ and $(\bar{1}\bar{1}1)[101]$ (see Figure 11(b)). In Figure 4(d) it is shown that slip lines occur on parallel planes near the [101] direction resulting in the elliptical cross-section. Lastly, in the 90° loading orientation the two slip systems in the [101] growth direction have Schmid factors of zero. Other slip systems with non-zero Schmid factors can have a projection in the [101] direction. Calculations of the contribution of the projections in the [101] direction, assuming an arbitrary Schmid factor cutoff of 0.2 to be ‘active’, point to an anisotropic deformation response. The sum

of the ‘active’ slip system projections in the [101] direction is significantly lower than in the direction perpendicular to both the crystallographic loading orientation and the [101] growth direction. As shown in Figure 5(e) the deformation along the [101] direction is limited, with deformation primarily occurring in the direction perpendicular to the [101] direction. In fact, the specimen deformation along the [101] direction routinely showed strain on the order of 0.5% compared to an isotropically predicted strain of 7% assuming volume conservation. The shear strength of the Ag-Cu interface would also limit deformation in the [101] direction. However, this has yet to be established.

The strain hardening index of each of the stress-strain curves shown in Figure 4, with the exception of the quasi-static curve is presented in Figure 12. The strain hardening response of the $\text{Ag}_{60}\text{Cu}_{40}$ system is attributed to the reduction in bi-layer thickness, deformation twinning, and the structure of the bi-phase interface. The trend of decreasing strain hardening with decreasing bi-layer thickness is somewhat expected because of the increasing difficulty of dislocation accumulation within either phase as their thickness decreases [50]. As is seen in the 0° loading orientation, in which twinning does not take place, there is a reduction in strain hardening with decreasing bi-layer thickness. Deformation twinning plays a role in the hardening response in the 45° and 90° loading orientations. In the Ag-Cu system twinning will initiate first in the Ag phase [51]. Previous studies of other low stacking-fault energy materials, such as MP35N [48], found two regimes of deformation applicable to the strains explored in this study. When strains are less than $\sim 5\%$, deformation is dislocation slip driven. Upon additional straining, twinning begins to occur. In cases where there is delayed activation of twinning, there may not be evidence of the transition in deformation modes in the stress-strain curves [52], as was observed in this study. In the $3.6\ \mu\text{m}$, $2.2\ \mu\text{m}$, and $1.1\ \mu\text{m}$ microstructured materials deformation twins were observed in both phases. It is noted that as the bi-layer thickness is reduced there is greater difficulty in the formation of deformation twins [52]. The deformation twins have an effect similar to reducing the grain size for slip and twin directions that are not coplanar to the deformation twin plane [53] because the deformation twins are strong barriers to dislocation motion [54]. While it is possible that propagating slip and twinning dislocations can cross-slip into the observed deformation twins, stress concentrations due to

pile-ups at the deformation twin-matrix interface would be necessary [55,56]. The offset that occurs between the 0° loading orientation and the 45° and 90° loading orientations can be attributed to the presence of deformation twins in the 45° and 90° loading orientations. Previous work by Mahajan and Chin [54] reported that systems that deform by both dislocation slip and twinning have greater strain hardening than those that deform by only dislocation slip. Karaman et al. [57] also came to the same conclusion.

The shear banding observed in the 90° orientation is most likely tied to the structure of the bi-phase interface as it was not observed in the 2.2 µm or 3.6 µm bilayer thickness, but it was in the 1.1 µm and 500 nm bilayer thicknesses which have a greater density of the incoherent twin interface. Recent work by Zheng et al. [58] points to the transmission of dislocations at the interface leading to crystallographic band formation, as was seen in the 45° case, see Figure 7(b). When there is resistance to transmission, non-crystallographic shear banding occurs, see Figure 8. TEM analysis and geometric alignment of slip systems arguments were used to conclude that the cube-on-cube orientation would lead to crystallographic slip and that the twin relationship prevents communication of twinning partial dislocations from Ag to Cu would cause tilting of the bi-phase interface. The tilted interface is then proposed to trigger 1) layer parallel slip, 2) large crystal reorientations, and 3) shear banding [58]. Although, microscopy of the slip band regions was not been performed in this study, Zheng et al. [58] provides a plausible explanation for the shear banding observed here.

5. Conclusions

The unidirectional Ag-Cu system provides a bulk material system of study that consists of discontinuous globular or lamellar copper reinforcements within a silver matrix. In this study the unidirectional system was studied at four nominal bilayer thicknesses namely, 500 nm, 1.1 µm, 2.2 µm, and 3.6 µm. Each of the specimens was subjected to compressive loading via the SHPB under three different loading orientations, parallel to the [101] growth direction 0°, at 45°, and perpendicular 90° to the growth direction.

The material response was loading orientation and bi-layer thickness dependent. An increasing strength trend of $45^\circ < 90^\circ < 0^\circ$ was obtained at all bilayer thicknesses examined. At the finest bilayer thickness (500 nm) a greater yield strength increase was observed due to a transition of the interface structure from cube-on-cube to incoherent twin. The transition of the interface also affected the strain hardening response. A general decrease in the strain hardening was observed with decreasing bilayer thickness consistent with limited accumulation of dislocations at finer microstructures. The 45° and 90° loading orientations exhibited a higher strain hardening response than the 0° orientation because of the presence of deformation twins in both phases.

Internal deformation examined through SEM showed that each loading orientation had unique microscale deformation mechanisms. The 0° orientation displayed reinforcement layer kinking within $100\text{ }\mu\text{m}$ of the radial surface under compressive strains greater than 10%. The 45° loading orientation specimen displayed a change in the planarity of the interface pointing to slip communication across the bi-phase interface. The 90° orientation showed a number of mechanisms: Crossing shear bands were observed on the radial surface when viewing down the growth direction. Internally shear bands with widths of a few microns were observed under SEM.

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Figure 1: Schematic of the 0° , 45° and 90° , specimens cut from a cast rod. The loading compression direction for each orientation is indicated by large arrows.

Figure 2: One dimensional analysis of split-Hopkinson pressure bar experiments (a) raw strain gage signals from a typical experiment (b) establishment of force equilibrium with black points indicated when equilibrium develops and is lost during an experiment and (c) the corresponding region of equilibrium conditions (bold red line) of the true strain-rate of the same experiment from which the average strain-rates presented are determined.

Figure 3: SEM micrographs of the 500 nm, 1.1 μm , 2.2 μm and 3.6 μm microstructures. Micrographs (a), (c), (e), and (g) view coincidental to the growth direction. Micrographs (b), (d), (f) and (g) view perpendicular to the growth direction

(a)

(b)

(c)

Figure 4: The stress-strain response of the $\text{Ag}_{60}\text{Cu}_{40}$ cast eutectic. (a) 0° (b) 45°, and (c) 90° loading orientation. Bi-layer thickness and strain-rates for each curve are provided in the respective legend. Two separate castings of the 500 nm bi-layer microstructure were examined at 0° the loading orientation, one casting indicated by up-triangles, the other with down-triangles.

Figure 5: Macroscopic deformation of the $\text{Ag}_{60}\text{Cu}_{40}$ subjected to dynamic loading viewing along the loading direction, (a), (c) and (e), and the radial surface (b), (d), and (f), for the (a)(b) 0° loading orientation, (c)(d) 45° loading orientation, and (e)(f) 90° loading orientation. Dashed white lines in (a), (c), and (e) indicate the initial specimen diameter. Solid white lines in the (b), (d), and (f) indicate the deformed specimen diameter. The orientation of the $[101]$ growth direction is provided as a dashed arrow. Solid white arrows indicate the position of important features on the radial surface, (a) striations, (b)(c) shear bands.

Figure 6: Internal structure of a 3.6 μm microstructure size specimen deformed to 16% compressive strain. (a) At the center of the deformed specimen the microstructure is similar to the as-cast microstructure. (b) Kinking of the copper reinforcements near the radial surface.

Figure 7: SEM micrographs of the internal deformation of the 45° fiber orientation to loading of at increasing magnifications (a,b) of a 3.6 μm bilayer specimen. Internal deformation of a 500 nm bilayer specimen is shown in (c). Waviness of the Ag-Cu interface was observed in deformed specimens with bilayer thicknesses greater than 1.1 μm.

Figure 8: SEM micrograph of a region of shear banding through multiple Ag-Cu interfaces. Severe reductions in thickness occur over the width of the shear band.

Figure 9: A comparison of the unidirectional material yield strength of each loading orientation with respect to bilayer thickness.

Figure 10: The 0° loading orientation showing with an imposed Thompson tetrahedron to indicate the position of the slip planes.

Figure 11: Schmid factors for each of the 12 FCC slip systems rotated about the [101] direction for the (a) 0°, (b) 45°, and (c) 90° loading orientations. The rotations of the colonies correspond to changes in crystal load orientation for the 45° and 90° cases. In the 0° case all rotations correspond to [101] load direction.

Figure 12: Strain hardening index of each of the loading orientation with respect to bilayer thickness.

Reduced bi-layer thickness with
increased solidification rate

Mechanical response is loading orientation and
microstructure dependent

Mechanism based
approach for
interpreting yield
strength and
hardening trends