Anomalous deformation twinning in coarse-grained Cu in Ag₆₀Cu₄₀ composites under high strain-rate compressive loading

B.P. Eftink¹, N.A. Mara², O.T. Kingstedt³, D.J. Safarik⁴, J. Lambros³ and I.M. Robertson⁵

¹Department of Materials Science and Engineering, University of Illinois, Urbana IL 61801

²Los Alamos National laboratory, MPA-CINT, Los Alamos National Laboratory, Los Alamos, NM 87545

³Department of Aerospace Engineering, University of Illinois, Urbana IL 61801

⁴Los Alamos National laboratory, MST-6, Los Alamos National Laboratory, Los Alamos, NM 87545

⁵Department of Materials Science and Engineering, University of Wisconsin-Madison, Madison WI. 53706

Abstract:

The deformation response of a directionally solidified Ag₆₀Cu₄₀ eutectic alloy with a cube-on-cube orientation relationship between Ag and Cu subjected to high strain-rate 10^3 s^{-1} compressive loading was examined. Loading at 45° and 90° to the growth axis, near $[001]$ and $[11\bar{1}]$ local crystal orientations, respectively, resulted in deformation twinning and dislocation slip in both Ag and Cu under conditions where deformation twinning would not normally be expected in Cu. In contrast, loading at 0° and 90° to the growth axis, near $\langle 101 \rangle$ local crystal orientations, resulted in the primary deformation mode being dislocation slip. These results are

interpreted in terms of the influence of loading axis with respect to the local crystal orientation in the directionally solidified alloy and on slip transmission from Ag into Cu.

Key words: twinning, copper, silver, slip

Introduction:

The AgCu system has been of interest for applications requiring a combination of high mechanical strength and electrical conductivity including high field magnets and also rails for electromagnetic launchers [1, 2]. Achieving high mechanical strength is usually a detriment to conductivity. Twinning can strengthen materials as they provide barriers to dislocations and effectively reduce the grain size while retaining electrical conductivity [3]. Twinning has also shown potential to not only strengthen but when introduced in a high density at thicknesses on the order of a few nano-meters, increase ductility [4].

Deformation twinning in FCC metals depends on the specific material and grain size in addition to external factors such as strain-rate and temperature [5]. The stacking-fault energy, unstable stacking-fault energy, and unstable-twinning energy are intrinsic properties that impact twinnability of metals [6-8]. Although Ag deforms via twinning under most loading scenarios, Cu only twins under certain conditions. For example, deformation twinning in coarse-grained Cu has been observed under high strain-rate loading ($\sim 10^6 \text{ s}^{-1}$) above 10 GPa applied stress at room temperature [9-12], and at low temperatures, 77 K, for quasi-static tensile loading or wire-drawing [13, 14]. Additionally, the propensity for Cu to deform by twinning under high strain-rate loading is dependent on the load orientation. Deformation twinning occurs in compression at $\langle 100 \rangle$ load orientations but not $\langle 110 \rangle$ at a shock pressure of 10 GPa due to the directional

nature of the twinning process [10]. With respect to grain size, Cu has been found to have an ideal grain size for twinning of 80 nm when tested as thin films at room temperature in uniaxial tension, and in grains less than a few tens of nanometers after room temperature high pressure torsion [15, 16].

Twinning in Cu has also been reported to occur in Ag-Cu alloys in regions with a cube-on-cube orientation relationship between the phases with $(111)_{\text{Ag}}|| (111)_{\text{Cu}}$ interfaces with crystallographic directions $[\bar{1}\bar{1}0]_{\text{Ag}}||[\bar{1}\bar{1}0]_{\text{Cu}}$ and $[11\bar{2}]_{\text{Ag}}||[11\bar{2}]_{\text{Cu}}$ aligned in the interface plane at high strains, greater than 52.9% rolling reductions, for layer thicknesses less than 350 nm [17-19]. This response has been attributed to direct transmission of twinning partial dislocations from Ag through the interface into Cu [17, 18]. This particular mechanism does not derive from interface-mediated nucleation processes, and depends instead on a prolific supply of twinning partials from the Ag phase transmitting into the Cu phase. As such, we hypothesize that if this is indeed the predominant deformation mechanism, that there should be no grain size effect. That is, in the Ag-Cu system, twinning should occur in Cu at large grain sizes and moderate temperatures and strain rates. Twinning partial dislocation communication across the interface from Ag to Cu, however, has barriers arising from the presence of misfit dislocations and coherency strains to account for the lattice mismatch between phases; image forces due to differences in elastic modulus; and the necessity to leave a residual dislocation at the interface with a Burgers vector of $b_r = \frac{a_{\text{Ag}} - a_{\text{Cu}}}{6} < 112 > = 0.196 \text{ \AA}$ for each twinning partial dislocation involved in the strain transfer, where a_{Ag} and a_{Cu} are the lattice parameters of Ag and Cu, respectively [20-22]. Deformation twinning in Cu is also found in accumulative roll bonded Cu and Nb with interfacial orientation relationships of $\{112\}_{\text{Cu}}||\{112\}_{\text{Nb}}$ and $\langle 110 \rangle_{\text{Cu}}||\langle 111 \rangle_{\text{Nb}}$.

Deformation twinning occurs in the Cu phase presumably by dissociation of misfit dislocations at the Cu/Nb interface or by geometrically well-aligned perfect dislocations in Nb transferring strain across the interface by twinning partial dislocation emission from the Cu/Nb interface into Cu or by a combination of the two [23]. The layer thickness influences the twinning response in multi-layered Cu-Nb with deformation twinning being prolific in Cu at layer thicknesses less than 30 nm and occurring less frequently in layers with thicknesses up to 80 nm [24].

In this paper, we report on the deformation mode in a directionally solidified eutectic $\text{Ag}_{60}\text{Cu}_{40}$ alloy following compressive loading at strain-rates of 10^3 s^{-1} . As deformation twinning in the Cu is directly related to deformation twinning in the Ag, it is posited that the slip continuity from the cube-on-cube orientation relationship between Ag and Cu allows the transfer of the twinning partial dislocations across the Ag/Cu interface from Ag to Cu. This results in twinning in Cu at a condition of strain-rate, temperature and length-scale where Cu would otherwise not deform by twinning. Crystal orientation dependence on deformation twinning is also observed and reported.

Experimental Procedure:

Directionally solidified $\text{Ag}_{60}\text{Cu}_{40}$ eutectic was cast as a 10mm diameter rod using a Bridgman furnace at a temperature of 1223 K and a removal rate of 0.46 mm/hr. This produced a microstructure that exhibited a cube-on-cube relationship between the Ag and Cu phases that grow concurrently with a common $\langle 101 \rangle$ direction parallel to the rod axis. The sample was then

annealed at 673 K for 4 hours at 10^{-6} Torr to reduce the solid solution of Ag in the Cu phase to approach a value on the order of 0.5 at. % according to the phase diagram [25].

Cylindrical samples were cut, using electrical discharge machining, from the rod at 0°, 45°, and 90° to the growth direction in order to test the dependence of the deformation response on loading orientation with respect to the Ag/Cu interfaces. The size of the samples cut 45° and 90° to the growth direction were smaller than that of the growth direction, however all samples maintained a length/diameter ratio of 1/2. The flat cylindrical surfaces were mechanically polished to remove any machining damage.

Compressive loading was conducted using a split-Hopkinson pressure bar (SHPB) at room temperature. The specimens were placed between two elastic bars, the incident bar and transmitted bar, with strain gauges mounted to their surfaces to monitor propagating stress waves. A striker bar was launched at the incident bar initiating a stress wave that traveled down the length of the incident bar. When the stress wave reached the specimen a portion of the stress wave passed through it into the transmitted bar and the remaining portion was reflected back into the incident bar. Assuming one-dimensional wave propagation, the stress, strain and strain-rate were calculated from the measured strain gauge signals. These samples were compressed at a strain-rate of $\sim 10^3 \text{ s}^{-1}$. More details about the SHPB experiments can be found in [19, 26].

TEM samples were then prepared from the interior of each rod by first cutting slices perpendicular to the load direction using electrical discharge machining. Machining damage was removed by mechanical polishing. TEM samples were produced from the slices by mechanical

polishing to a 1 μm finish then punching out 3 mm diameter disks. Thinning to electron transparency was done by ion milling at cryogenic temperatures using a Gatan PIPS. Other samples were extracted from the deformed material by using the focused ion beam machining technique using a FEI Helios 600i [27]. Diffraction contrast TEM analysis was conducted using a JEOL 2010 LaB₆ operating at 200 kV and high resolution TEM images were acquired using a FEI Tecnai F30 operating at 300 kV.

SEM analysis was conducted on a FEI Inspect F with a TSL/EDAX orientation imaging CCD for electron backscatter diffraction (EBSD). Samples were prepared for EBSD by polishing to a 1 μm finish and then ion milling at cryogenic temperatures using a Gatan PIPS with a milling angle of 3° and voltage between 1-3 kV.

Results:

The microstructure of the directionally solidified alloy consists of Cu platelets elongated in, and discontinuous along, the growth direction in a Ag matrix, Fig. 1. The discontinuous nature of the elongated Cu phase (dark phase) along the growth direction is seen in Fig. 1(a). The cross-sectional view shows the dispersion of the Cu phase in the Ag matrix, Fig. 1(b). From this image, the smallest dimension of the Cu phase is around 1 μm . The crystal orientation along the growth direction was determined by EBSD to be $\langle 101 \rangle$ in both phases, Fig. 2(a). An EBSD map of the same area with the frame of reference perpendicular to the growth direction shows a grain-like structure, Fig. 2(b). This illustrates that the growth direction is relatively constant with crystallographic rotations occurring about the growth direction. The rotations about the growth

axis result in a distribution of local crystallographic orientations with respect to the load for the samples loaded 45° and 90° to the growth direction. However, loading along the growth direction will have a constant $[101]$ along the load axis.

Figure 1: SEM backscattered electron images of the undeformed directionally solidified material. Ag appears grey and Cu appears black. (a) Growth direction vertical. (b) Growth direction into the page.

Figure 2: Undeformed directionally solidified material. (a) EBSD map with reference direction out of the page. (b) EBSD map of the same area as (a) showing the crystal orientation from a point-of-view 90° from the sample normal direction.

The as-cast material has a low dislocation density in both Ag and Cu with the dislocation density being higher in Ag than in Cu. Dislocation nucleation occurs at the Ag/Cu interfaces into the Ag as evident by the higher density of dislocation half-loops emerging from the interface into the Ag. Examples of these features can be seen in the bright-field micrograph presented in Fig. 3; examples of dislocation half-loops are indicated by arrowheads. Elastic stress concentration centers can be seen in the form of distortions extending from the interface into the Ag; examples are indicated by arrows. On examining the interface structure at higher spatial resolution, it is found that the interface between Ag and Cu is $\{111\}$ cube-on-cube with irregularly spaced $\{111\}$ steps. An example of this structure is shown in the high-resolution image presented in Fig. 4; the

estimated location of the interface is indicated by the white line. The steps on the interface vary in height with some being just one atomic plane high and others being several; these steps account for the interface curvature. Misfit dislocations, present as extra planes in Cu, occur approximately every 8-11 planes and are marked by black circles. These dislocations account for the lattice mismatch between the two phases; the lattice parameters for Cu and Ag are 3.610 Å and 4.090 Å, respectively. The spacing of the misfit dislocations matches that of previous findings of a quenched Ag₆₀Cu₄₀ material [28].

Figure 3: Bright-field TEM image of undeformed directionally solidified Ag/Cu alloy. Arrowheads mark dislocation half loops and arrows mark elastic distortion.

Figure 4: HRTEM of directionally solidified material. Misfit dislocations marked with black circles. Beam direction is $[101]$ and Ag/Cu interface is $\{111\}$.

Dynamic stress-strain curves obtained from the SHPB as a function of angle between the loading direction and the growth direction are compared in Fig. 5(a) [26]. These show that the macroscopic stress-strain response is dependent on the angle between the loading direction and the interface normal. For loading along the growth direction the measured macroscopic strength is the highest. Loading at 45° and 90° to the growth direction exhibited similar strengths, this surprising result requires further testing to eliminate any sample to sample variance due to local crystal orientation distribution with respect to the load. Macroscopically, the 45° and 90° samples show anisotropic deformation. This is seen in the optical image of the deformed samples presented in Fig. 6. In the case of the sample loaded at 45° to the growth axis,

macroscopic shear bands associated with the compression are generated and one is marked in Fig. 6(b). For the sample loaded 90° to the growth axis, the resultant cylinder has an oval cross section, Fig. 6(c). SEM analysis, not shown, confirmed that the growth direction is aligned with the short axis of the oval.

Figure 5: (a) Stress strain curves for directionally solidified $\text{Ag}_{60}\text{Cu}_{40}$ at load orientations 0° , 45° , and 90° to the growth direction at strain rates of 10^3 s^{-1} for loading with split-Hopkinson pressure bar. (b) Schematic of orientations of 0° , 45° , and 90° from left to right. Large arrows denote load, and small arrows denote growth direction.

Figure 6: From left to right, macroscopic deformation of samples loaded (a) 0° , (b) 45° , and (c) 90° to the growth direction. In (b) arrows mark a shear band on the sample.

Loading along the $[101]$ growth direction produced a high density of dislocations that are observable in the bright-field micrograph presented in Fig. 7 (a). The dislocation density in Ag increases near the Cu fibers with the Ag/Cu interfaces showing the highest density of dislocations. Loading at 45° to the $[101]$ growth direction, $[001]$ local orientation, resulted in dislocation slip and profuse deformation twinning in both Ag and Cu. An example of the twinned microstructure is presented in Fig. 8 (a), and the corresponding diffraction pattern with twin spots indicated in Fig. 8 (c). Inspection of the twins shows that in some cases they appear continuous across the Ag/Cu interfaces while in other regions they exist in the Ag layer and terminate at the Ag/Cu interface. Locations where Ag twins meet a Ag/Cu interface have been observed to result in emission of dislocations instead of twins into Cu, as seen in Fig. 8 (b), despite the cube-on-cube orientation relationship between Ag and Cu.

Figure 7: Bright field TEM micrograph of directionally solidified $\text{Ag}_{60}\text{Cu}_{40}$ compressed using SHPB loaded along the growth direction corresponding to a load orientation of $[101]$. Diffraction pattern for image shown in (b).

Figure 8: Bright field TEM micrographs of directionally solidified $\text{Ag}_{60}\text{Cu}_{40}$ compressed using SHPB loaded 45° to the growth

direction corresponding to a load orientation of $[001]$. Ag/Cu interfaces marked in (a). Diffraction pattern for image (a) and (b) shown in (c) and (d) respectively. (b) Planar defects and dislocations appear on the Ag side of the interface while only dislocations appear on the Cu side.

Loading at 90° to the growth direction resulted in deformation structures that were dependent on crystal orientation. For a local load orientation near $[10\bar{1}]$ dislocation slip dominated the response in both the Ag and Cu. An example of this microstructure is shown in the bright-field micrograph presented in Fig. 9 (a), which shows dislocations in both Ag and Cu, with dislocations in Ag organized into bands that run parallel to the Ag/Cu interfaces. Deformation twins were not observed in either phase using TEM for a load orientation near $[10\bar{1}]$, however a few twins, either deformation or annealing, were observed by EBSD at that orientation, Fig. 10 (a). Profuse deformation twinning in both phases is, however, observed by both TEM and EBSD for local load orientations near $[11\bar{1}]$, Fig. 9 (b) and Fig. 10 (b), respectively. The interface between Ag and Cu exhibits abrupt orientation changes thought to be due to interface rotation during the transfer of twinning defects across the interfaces as indicated by the arrowheads in Fig. 9 (b). A higher magnification image of a region of the interface showing the rotation is presented in the inset of Fig. 9 (b). Differing orientations separated by a “grain boundary” have also been observed to separate regions with and without deformation twinning. Fig. 9 (c) contains both near $[10\bar{1}]$ and $[11\bar{1}]$ local load orientations in the top left and bottom right, respectively. From the micrograph, the near $[10\bar{1}]$ load orientation exhibits

dislocation slip while the near $[11\bar{1}]$ load orientation shows both deformation twinning and dislocation slip in both phases.

Figure 9: Bright field TEM of the directionally solidified sample loaded 90° to the growth direction. (a) Local load orientation of $[10\bar{1}]$ showing dislocation slip. (b) Local load orientation of $[11\bar{1}]$ showing deformation twinning in both Ag and Cu, arrowheads mark select regions of the interface with abrupt changes indicating interfacial rotation due to the deformation process. (c) Local load orientation of $[10\bar{1}]$ for top left grain exhibiting dislocation slip, local load orientation of $[11\bar{1}]$ for bottom right grain exhibiting deformation twinning and dislocation slip. (d), (e) and (f) are the corresponding diffraction patterns for (a), (b) and (c) respectively. Diffraction spots for Ag corresponding to the left and right grains of (c) are marked with L and R subscripts respectively in (f).

Figure 10: EBSD maps of directionally solidified sample loaded 90° to the growth direction. (a) Grain loaded near the $[10\bar{1}]$ orientation. (b) Grain loaded near the $[11\bar{1}]$ orientation. (c) Inverse pole figure of (a). (d) Inverse pole figure of (b).

Discussion:

In these directionally solidified alloys, the exclusively cube-on-cube orientation relationship between Ag and Cu is in agreement with a previous report [29], but at variance with Ag-Cu eutectic cast at higher solidification rates, which show both a twin orientation relationship as well as the cube-on-cube relationship [30-32]. This is consistent with the concept that the twin orientation relationship is more likely to occur at faster solidification rates [31, 32]. At a fast cooling rate, the proportion of cube-on-cube and twin Ag/Cu orientation relationships are similar whereas at a slower cooling rate, as used in this study, the cube-on-cube orientation relationship dominates [30]. Britzman et al., after directionally solidifying Ag-Cu eutectic at growth rates two orders of magnitude higher than used here, observed morphologies of near circular fibers and a lamellar structure [33]. Observing both fiber and lamellar morphologies relates to volume fraction of the phases in the Ag-Cu system, which straddles the boundary between lamellar and

fiber morphology for eutectics [33]. This, however, differs from the morphology observed at the much slower growth conditions used here, which results in a platelet like Cu phase showing preferential $\{111\}$ interface planes.

Loading along the growth direction, $[101]$, and hence the Cu platelet elongation direction, showed the highest mechanical strength of the three orientations tested, Fig. 5. From a resolved shear stress argument eight of the twelve possible slip systems are shut down when loading in this orientation: $(\bar{1}\bar{1}1) \pm [101]$, $(1\bar{1}\bar{1}) \pm [101]$, $(111) \pm [10\bar{1}]$, $(\bar{1}\bar{1}\bar{1}) \pm [10\bar{1}]$, $(\bar{1}\bar{1}\bar{1}) \pm [110]$, $(1\bar{1}\bar{1}) \pm [0\bar{1}1]$, $(\bar{1}\bar{1}1) \pm [\bar{1}10]$, and $(\bar{1}\bar{1}1) \pm [011]$. The slip systems that can be activated have a relatively small mean free path to encounter Ag/Cu interfaces which impedes dislocation slip. By this argument the growth direction will be the strongest orientation. When loading 45° to the growth direction, regardless of local crystal orientation the $(1\bar{1}\bar{1}) \pm [101]$ and $(\bar{1}\bar{1}1) \pm [101]$ slip systems are aligned to have the maximum Schmid factor and a large mean free path to encountering a Ag/Cu interface, this should result in this being the weakest load orientation. The sample loaded 90° to the growth direction should have a strength between loading 0° and 45° to the growth direction with a strength dependent on the distribution of local crystal orientations. Further testing is underway to understand the dependence of strength on the local orientation distribution for samples loaded in the 90° orientation.

As compared to the initial state of the material, when loaded along either the growth direction, $[101]$, or normal to the growth direction, near $[10\bar{1}]$, the dislocation density of the material increases in both the Ag and the Cu phases, Fig. 7 (a) and 9 (a). This deformation behavior is consistent with that of the constituents if tested separately. Additionally, the increased

dislocation density near the interfaces indicates they act as barriers to dislocation slip, Fig. 7(a). Consistent with prior high strain-rate compressive loading of Cu, deformation twins were not common in Cu, or Ag, at this local crystal orientation with respect to the load [10]. A few twins, either deformation or annealing twins, however, were identified in the EBSD map in a $[10\bar{1}]$ oriented grain, Fig. 10 (a).

Twinning in Cu was observed, but not without twins across the interface in the Ag, in certain crystallographic load orientations. Twinning occurred after loading 45° to the growth direction near a $[001]$ load orientation, and loading 90° to the growth direction near a $[11\bar{1}]$ load orientation. However, twins also exist in the Ag phase without corresponding twins in Cu. Deformation twinning is expected under these loading conditions for Ag but not for Cu. This result supports the concept that deformation twinning in Ag is necessary to initiate twinning in Cu. This conclusion raises the important question about the interaction of twinning partial dislocations in Ag interacting with the interface and causing deformation twinning in the copper. One possible explanation is that the interface is effectively transparent to the Ag twinning partial dislocations. Support for this mechanism is provided by molecular dynamics computer simulations of a Ag-Cu interface with a cube-on-cube orientation relationship, i.e., $(111)_{\text{Ag}} \parallel (111)_{\text{Cu}}$ with crystallographic directions $[1\bar{1}0]_{\text{Ag}} \parallel [1\bar{1}0]_{\text{Cu}}$ and $[11\bar{2}]_{\text{Ag}} \parallel [11\bar{2}]_{\text{Cu}}$ aligned in the interface plane [22]. Two sets of twinning partial dislocations were considered to form the twin in Ag; $b_{\text{Ag}} = \frac{a}{6}[112]$ or a combination of $b_{\text{Ag}} = \frac{a}{6}[\bar{2}1\bar{1}]$ and $b_{\text{Ag}} = \frac{a}{6}[1\bar{2}\bar{1}]$. Interaction of either set of partial dislocations resulted in the formation of intrinsic interfacial dislocations and a rotation of the interface plane. Interactions involving twinning partial dislocations with a Burgers vector of $b_{\text{Ag}} = \frac{a}{6}[112]$ were found to generate

intrinsic interface dislocations with a larger Burgers vector and cause a larger interface rotation than interactions involving the other set of twinning partial dislocations. Consequently, it was proposed that reactions involving the other set of Ag twinning partial dislocations were more favorable [22]. Abrupt changes in the Ag/Cu interface orientation in twinned regions, shown in Fig. 9 (b), are believed to be from transfer of twinning partial dislocations across the interface. Characterizing the exact rotation was not performed as the original orientation of the interfaces is unknown. Eftink et al. have during *in situ* TEM straining experiments of this alloy seen directly the rotation of the interface as deformation twins move rapidly through it [34]. Additionally, Wang et al. suggested that the total energy of the accumulated intrinsic interfacial dislocations could be reduced by the interface ejecting other dislocations in addition to the Cu twinning partial dislocations [22]. Again, Eftink et al. have observed such a response from an interface but not always in relation to the transfer of twinning partial dislocations across the interface. Based on the molecular dynamics computer simulations, it would appear that the process of deformation twin transfer across a cube-on-cube interface is controlled by the shear stress on the Ag twinning partial dislocations, minimization of the accumulation of intrinsic interfacial dislocations and rotation of the interface [22]. The simulations, however, did not provide insight as to why some Ag deformation twins do not cause a transfer event despite being in close proximity to ones that do.

The molecular dynamics computer simulations of Ag-Cu support an interfacial strain minimization condition for predicting the strain transfer process across the hetero-phase interfaces. However, in the Ti-48 at.% Al alloy system, stress-induced activation of interface sources was found to be more determining than minimization of strain energy at the hetero-phase interfaces for causing the transfer of strain from the γ -TiAl to the α_2 -Ti₃Al phase lamellae [35].

In that system, the lack of strain transmission is explained by the need for a high density of dislocations for twins to transfer across the boundary [36], for the stress-induced activation of interface dislocation sources [36] and by an elastically-mediated transfer process, which is restricted to the thinnest sections of the hard α_2 -phase [37]. Eftink and coworkers have reported stress-induced activation of deformation twinning transferred elastically across the Cu layer in Ag-Cu such that deformation twinning is activated in the next Ag layer [38]. This occurs only in Cu layers tens of nanometers thick and if the orientation relationship between Ag and Cu is a twin.

Deformation twins in Cu were observed to extend for a few microns, which is an order of magnitude larger than the 250-350 nm reported in previous studies [17, 18]. The occurrence of these deformation twins in Cu across layers microns wide is in contrast to the Cu-Nb system in which deformation twinning occurs in layers less than or equal to 80 nm thick [24]. The difference in length-scale dependence could be due to differing transfer mechanisms, direct transmission from Ag to Cu compared to emission from Cu/Nb interfaces or dislocation transfer from Nb to twinning dislocations in Cu [39].

Conclusions:

- Deformation twinning in the Cu phase was observed at room temperature at a strain rate on the order of 10^3 s^{-1} , conditions under which coarse-grained Cu would not exhibit deformation twinning.

- Ag and Cu with slip continuity between the phases, in this case from a cube-on-cube orientation relationship, in addition to a stress state to induce twinning in Ag was necessary for twinning partials to transfer from Ag into Cu.
- The propensity for deformation twinning to be the dominant deformation mode in both Ag and Cu is dependent on the local crystal orientation with respect to the load. Samples loaded 45° and 90° to the growth direction and with local crystal orientations of [001] and [11 $\bar{1}$], respectively, contained deformation twins. Samples loaded 0° and 90° with $\langle 101 \rangle$ local load orientations did not exhibit deformation twinning but deformed by dislocation slip exclusively.

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