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Deformation response of cube-on-cube and non-coherent twin interfaces in AgCu eutectic under dynamic plastic compression

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ABSTRACT

Split-Hopkinson pressure bar dynamic compression experiments were conducted to determine the defect/interface interaction dependence on interface type, bilayer thickness and interface orientation with respect to the loading direction in the Ag-Cu eutectic system. Specifically, the deformation microstructure in alloys with either a cube-on-cube orientation relationship with $\{111\}_{Ag}||\{111\}_{Cu}$ interface habit planes or a twin orientation relationship with $\{\bar{3}13\}_{Ag}||\{\bar{1}12\}_{Cu}$ interface habit planes and with bilayer thicknesses of 500 nm, 1.1 µm and 2.2 µm were probed using TEM. The deformation was carried by dislocation slip and in certain conditions, deformation twinning. The twinning response was dependent on loading orientation with respect to the interface plane, bilayer thickness, and interface type. Twinning was only observed when loading at orientations away from the growth direction and decreased in prevalence with decreasing bilayer thickness. Twinning in Cu was dependent on twinning partial dislocations being transmitted from Ag, which only occurred for cube-on-cube interfaces. Dislocation slip and deformation twin transfer across the interfaces is discussed in terms of the slip transfer conditions developed for grain boundaries in FCC alloys.

1. Introduction

Determining the rate limiting steps associated with transmission of dislocations across grain and bi-phase boundaries is essential to understanding and modeling the mechanical response of polycrystalline as well as multi-layered systems. Dislocation transfer across grain boundaries in face-centered cubic (FCC) and some hexagonal close packed (HCP) systems is controlled by the magnitude of the Burgers vector of the residual dislocation created in the grain boundary by the act of transmission [1]. This controlling condition appears to be independent of dislocation type as well as grain boundary character [2].

Applying the controlling condition to twin boundaries shows that they are either weak or strong barriers to the transmission of slip [3–5]. This barrier strength is attributed to: i) geometric factors of the boundary and ii) the character of the dislocation. For example, screw dislocations can cross-slip across a twin boundary without generating a residual dislocation if the lines of intersection with the boundary of the incoming and outgoing systems are collinear and parallel to the Burgers vector (line direction) of the dislocation [3,6]. When this stated condition is not satisfied, the twin boundary acts as a strong barrier. Consequently, the introduction of a high density of parallel twin boundaries can cause significant strengthening [5].

For bi-phase boundaries, a similar approach can be taken with the interfaces characterized in terms of the degree to which slip planes and Burgers vectors of the dislocations align between layers [7,8]. Interfaces across which the slip systems are well-aligned, e.g., cube-on-cube interfaces in FCC/FCC systems, have been referred to as “transparent” [8]. Interfaces across which the slip systems are misaligned, e.g., FCC/body-center cubic (BCC) interfaces, have been referred to as “opaque” [8]. The resistance to strain transfer is related to coherency strains for transparent interfaces and shear resistance of the interfaces for opaque ones. The slip transmission mechanism is dependent on the bilayer thickness (thickness of both layers bordering the bi-phase boundary), but only for bilayer thicknesses less than a few tens of nanometers [9–12]. At bilayer thicknesses above a few tens of nanometers dislocation pile-ups form at the interface and the mechanism is consistent

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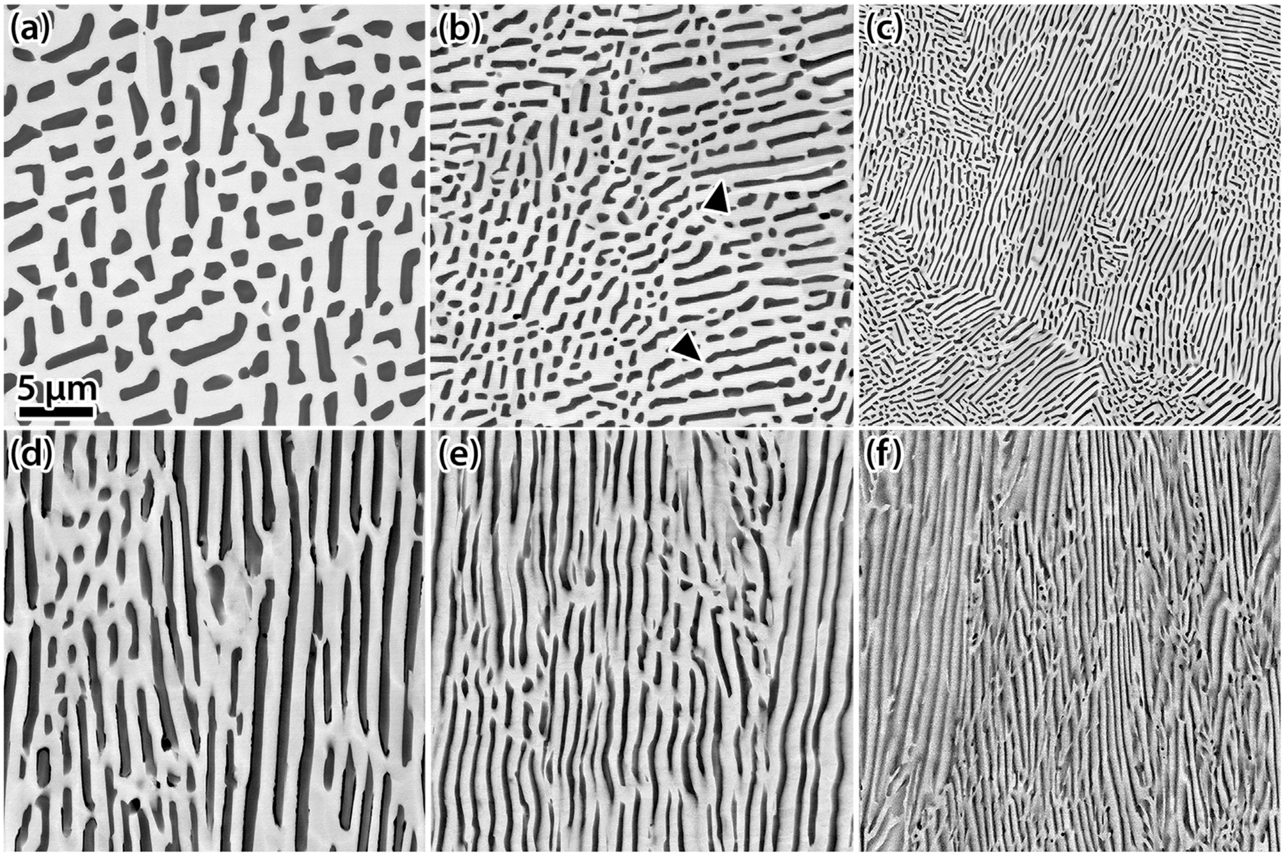

Fig. 1. SEM backscattered electron images of the undeformed directionally solidified material at each growth rate: (a) and (d) 0.46 mm/h, (b) and (e) 7.3 mm/h, (c) and (f) 73 mm/h. Cu appears darker than Ag. (a) - (c) Growth direction into the page. (d) - (f) Growth direction in the vertical direction. Arrowheads in (b) indicate regions of lamellar structure.

with that across grain boundaries in coarse-grained materials [13].

The Ag-Cu eutectic was selected for this investigation as it can be produced with interfaces with either cube-on-cube or twin orientation relationships between Ag and Cu through control of the solidification parameters [14–16]. Furthermore, the bilayer thickness can be controlled via the solidification rate [17,18]. Thus, it is possible to determine the dependence of the deformation response on the bi-phase boundary type, bilayer thickness, and loading direction with respect to the interface normal. This investigation builds on a previously published investigation on the quantitative stress/strain mechanical response of bulk AgCu eutectic with cube-on-cube and non-coherent twin interfaces to ex situ high strain rate loading [19]. The focus of this manuscript is on the differences in observed microstructures after bulk deformation in terms of: interface type, bilayer thickness, and load orientation. More specifically, in this work the deformation twinning behavior across interfaces of type $\{111\}_{Ag}||\{111\}_{Cu}$ with a cube-on-cube orientation relationship and $(\bar{3}13)_{Ag}||(\bar{1}12)_{Cu}$ with a twin orientation relationship at bilayer thicknesses between 500 and 2200 nm was determined by using high strain rate loading. It will be shown that deformation twinning was induced in the Cu layer but only if it was triggered by a deformation twin in the Ag layer, and the interface was of the cube-on-cube type. Deformation twinning in the Ag layer was dependent on the loading orientation, increasing in propensity for loading orientations away from the growth direction and decreasing in propensity with decreasing bilayer thickness.

2. Experimental methods

The materials used in this study were produced by directional solidification of $Ag_{60}Cu_{40}$ (eutectic), using a Bridgman furnace. Growth rates of 0.46, 7.3 and 73 mm/h were used to produce material with varying bilayer thicknesses and interfaces between the Ag and Cu phases. Cylindrical samples were cut at 0°, 45°, and 90° to the growth direction, using electrical discharge machining, from directionally solidified rods. The length of the specimen was always half the diameter. Machining damage was removed by mechanical polishing with a final polishing step using 1 μm diamond solution. Compressive loading at a strain-rate of $\sim 10^3\,s^{-1}$ was conducted at room temperature using a split-Hopkinson pressure bar (SHPB) [19–22].

Scanning electron microscopy (SEM) and electron backscatter diffraction (EBSD) analysis was performed using a Zeiss LEO 1530 with an electron backscatter detector and TSL/EDAX orientation imaging charge coupled device, respectively. Samples for both were prepared by mechanically polishing down to a 1 μm finish before ion milling at cryogenic temperatures with a Fischione 1050 argon ion mill. Final ion milling was performed at a 1.0 keV accelerating voltage. EBSD maps were acquired as 50 by 50 μm areas with 50 nm step size. TEM samples were prepared by either ion milling or focused-ion beam machining from the specimen interior. TEM samples were prepared such that the foil normal was close to the [101] growth direction. For samples thinned to electron transparency using ion milling, 800 μm thick slices were cut from the deformed cylinders using electric-discharge machining and mechanically polished to a final thickness of 80 μm before ion milling. The ion milling was performed using a Fischione 1050 with a liquid nitrogen cooled sample stage at a voltage of 5.0 keV with a final milling step at 1.0 keV. Focused ion beam machining was used to extract samples from both deformed and undeformed rod interiors and to thin the extracted volume to electron transparency [23]. Either a FEI Helios 600i or Zeiss Auriga was used to produce the foils. A final milling voltage of 2.0 keV was used to limit the effects of ion beam damage. Diffraction contrast TEM analysis was conducted using either a JEOL 2010 LaB_6 operating at 200 keV or a FEI Tecnai TF-30 operating at