Experimental Assessment of Strain Gradient Plasticity Theories

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ABSTRACT

Classical plasticity theories generally assume that the stress at a point is a function of strain at that point only. However, when gradients in strain become significant, this localization assumption is no longer valid. These conventional models fail to display a 'size effect'. This effect is seen experimentally when the scale of the phenomenon of interest is on the order of several microns. Under these conditions, strain gradients are of a significant magnitude as compared to the overall strain and must be considered for models to accurately capture observed phenomena.

The mechanics community has been actively involved in the development of strain gradient theories for many years. Recently, interest in this area has been rekindled and several new approaches have appeared in the literature. Two different approaches are currently being evaluated. One approach considers strain gradients as internal variables that do not introduce work conjugate higher order stresses. Another approach considers the strain gradients as internal degrees of freedom that requires work conjugate higher order stresses. Experiments are being performed to determine which approach models material behavior accurately with the least amount of complexity. A key difference between the two models considered here is the nature of the assumed boundary conditions at material interfaces. Therefore, we are investigating the deformation behavior of aluminum/sapphire interfaces loaded under simple shear. Samples are fabricated using ultra-high vacuum diffusion bonding. To determine the lattice rotations near the boundary, we are examining the samples with both electron backscatter diffraction methods (EBSD) in the scanning electron microscope and with a variety of diffraction techniques in the transmission electron microscope. The experimentally found boundary conditions shall be subsequently used to determine whether the simpler internal variable model is adequately descriptive or if the greater complexity associated with the internal degree of freedom approach is warranted.

INTRODUCTION

Although not predicted by classical models, an increase in flow stress is seen during deformation when the observed phenomena is on the order of a micron and inhomogeneties are present. For example, Fleck et.al [1] showed that when loaded in torsion, a wire displays greater strength for smaller radii. Others authors have observed this type of effect in other systems, including bending [2], indentation hardness [3], and particle hardened alloys [4]. The increase in hardness under these conditions is due to additional dislocations needed for compatibility. These dislocations are commonly referred to as geometrically necessary dislocations [5]. The presence of these dislocations can be ignored and continuum theories applied at large size scales since gradients in strain are small. However, at smaller size scales, more dislocations are formed in a smaller area, resulting in large strain gradients and higher flow stresses. This effect is not captured in continuum models because these models assume that the stress at a point is a function of strain at the same point only. When strain gradients become significant, this localization assumption is no longer valid [6]. This must be accounted for in a non-local theory to accurately reflect material response. Two distinct classes of models that extend classical theories to include strain gradient effects are currently being evaluated. Fleck and Hutchinson [7] have developed one approach in which strain gradients are included in the hyperelastic work function. This type of formulation considers strain gradients as internal degrees of freedom and requires work conjugate higher order stresses, which need additional boundary conditions. Acharya and Bassani [8] have developed an alternative approach in which a strain gradient term is included in the hardening function. In this method, the strain gradients are considered to be internal variables, which do not introduce work conjugate higher order stresses or additional boundary conditions. This approach has the advantages that it is simpler overall, preserves the structure of the classical boundary value problem, and can easily be implemented into existing finite element codes. However, the additional boundary conditions in the higher order theory allow for the presence of a boundary layer. Specifically, Fleck and Hutchinson [7] determined theoretically that a boundary layer of lattice

rotation should be present at an interface between dissimilar materials loaded under remote simple shear. Since boundary layers are not allowed in Acharya and Bassani's approach, detecting the presence of these layers at interfaces will supply critical information in the continued development of strain gradient plasticity theories. Boundary layers seem likely in real materials, since dislocation motion is governed by stress fields that are strongly affected by boundaries [9]. The presence of a boundary layer, however, has not yet been definitively determined. Although previous experimental work on bicrystals by Sun et.al [10] suggests the presence of a boundary layer, the data is difficult to interpret due to the movement of the grain boundary. Therefore, experiments performed at a metal-ceramic interface are proposed to determine deformation behavior without the complication of grain boundary movement.

EXPERIMENTAL DETAILS

Samples have been designed to study the deformation behavior at a metal-ceramic interface. The samples consist of a metal foil sandwiched between two polished ceramic cylinders. C-axis oriented sapphire polished to $\lambda/10$ flatness as obtained from General Ruby & Sapphire is used for the ceramic. Aluminum is used for the metal because of its relatively simple deformation behavior as compared with other available choices. 99.999% pure aluminum foils are obtained from Metron Company. An ultra-high vacuum diffusion-bonding machine [11] is used to bond the sapphire cylinders to the aluminum foil. The aluminum foils are 25 or 50 microns thick and the sapphire rods measure 16 mm in diameter and 20 mm long. Prior to bonding, the surfaces are cleaned by sputtering and characterized by Auger spectroscopy. The samples are held at 600°C for 38 hours with an applied pressure of 10 MPa. These samples were tested in a prior study [12] to determine the strength of the aluminum-sapphire bond. A notch was cut in the aluminum and symmetric four-point bend tests were performed. The fracture surfaces had the dimpled morphology characteristic of ductile fracture, indicating that failure occurred in the aluminum and not at the interface. This shows that the aluminum is strongly bonded to the sapphire and is assumed to deform before debonding in this study.

DISCUSSION

In order to shed light on existing strain gradient models, a simple stress state is needed for testing. This is because all models predict length scale effects to occur during inhomogeneous deformation modes. A shear stress state can be accomplished by testing the samples under *asymmetric* four-point bending [13]. A schematic illustration of the test geometry is shown in Figure 1.

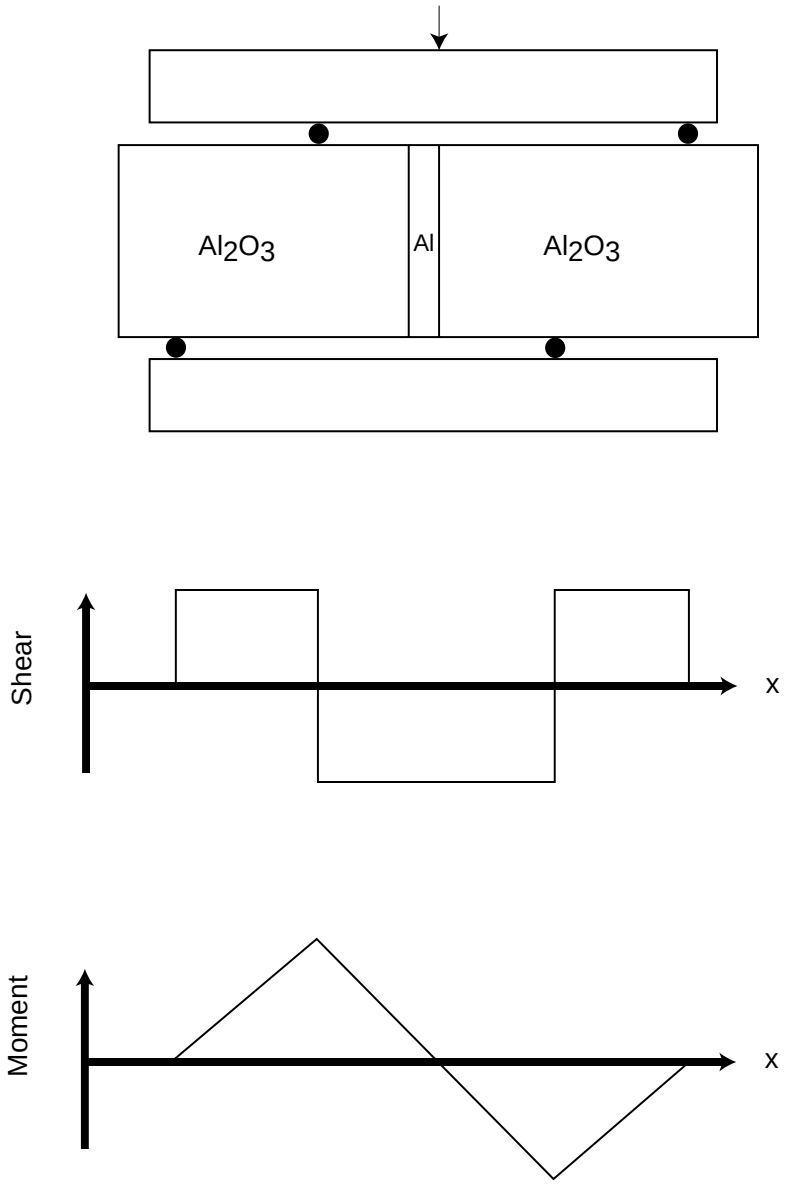

Figure 1. *Schematic of the sample, loading, and the shear and moment along the length of the sample*

Under elastic loading, a constant amount of shear is present between the inner contact pins, but it is only directly underneath the load-point that there is no bending moment. For plastic loading, finite element modeling is used to simulate this experiment in order to evaluate the homogeneity of the deformation. The sapphire is modeled as an isotropic elastic material and the metal is modeled as a J_2 linear hardening material. The modeling shows the metal layer is under a homogeneous stress state. The shear strain is homogeneous in the center, but does exhibit some variation near the edges. This shall not affect the experimental results, because observations are only made in the center of the sample.

To prepare for testing, the cylindrical sandwiched samples are cut into rectangular rods that are approximately 40 mm long, 2.5 mm high, and 3 mm wide. The top and bottom are ground flat for ease of mechanical testing and polished to 1μm to remove surface flaws that would lead to premature fracture of the sapphire. A bending test apparatus has been made for use in an Instron mechanical test frame. The test apparatus is designed to be self-aligning, so that the load from the mechanical test frame is transmitted directly onto the center of the bending apparatus. A 5kN-load cell is used to measure the force. Displacement is measured by an LVDT as well as by a feeler gauge placed underneath one side of the interface. After testing, the surfaces of the samples are observed using a scanning electron microscope (SEM) and displacement between the sapphire rods is observed. The

displacement within the aluminum appears to be non-uniform. An average value of 48 ± 3μm of displacement has been determined by confocal optical microscopy. Undeformed samples are also scanned to ensure the accuracy of this method. These scans show that displacements between the sapphire were not present before testing.

In order to detect the presence of a boundary layer in the aluminum near the sapphire, the lattice rotations need to be measured from the center of the aluminum layer to the interface. To measure the rotation, electron backscatter diffraction patterns (EBSD) were generated in a SEM on these samples (see [14] for a description of EBSD). However, the signal was too weak to obtain accurate data under current sample preparation methods. Instead, the mechanically tested samples were cut into slices for observation in a transmission electron microscope (TEM). Lattice rotation can be measured using selected area electron diffraction patterns that display both Kikuchi lines as well as diffraction spots. The type of rotation in the diffraction pattern depends on which way the rotation occurs in the lattice. Depending on the axis of rotation within the aluminum, either the lines will move with respect to the spots or the whole pattern will rotate with respect to the optic axis. Either type of rotation can be measured with a precision of at least one degree. Unfortunately, standard TEM sample preparation methods cause the aluminum to recrystallize. Alternate sample preparation methods has been employed, and an unrecrystallized sample has been made out of an undeformed bar. From initial inspection of the diffraction patterns from these samples, no trend in rotation is detected. Undeformed samples will be fully characterized and used to establish a baseline trend in lattice rotation. Deformed samples are currently being prepared in this newly established method to avoid recrystallization complications.

Once deformed samples have been successfully prepared, lattice rotations will be measured using selected area diffraction in the TEM. Diffraction patterns will be used to determine the trend in lattice rotation as the interface is approached. The trends in rotation will be confirmed using electron backscatter diffraction methods (EBSD) in the scanning electron microscope once improved sample preparation techniques are found that will allow for a stronger signal. Also, samples using single crystal metal layers will be fabricated. Copper will be used since it is easier to polish for diffusion bonding as well as TEM sample preparation. It should also produce a stronger signal for the EBSD, since copper generates more backscattered electrons that aluminum. Single crystal aluminum may be used later if necessary.

CONCLUSIONS

We are currently evaluating two models of strain gradient plasticity. They differ from one another in their levels of complexity and concomitant computational demands. The comparison with experimentally observed deformation boundary layers at interfaces should allow for an evaluation of the level of complexity that is necessary to capture relevant materials behavior in simulations. Samples have been fabricated with interfaces strong enough for the metal layer to deform without the interface debonding. A successful preparation method for TEM samples has been found and an initial investigation performed on undeformed samples. From these first observations, no trend in rotation has been detected. The initial microstructure will be fully characterized by quantitative measurement of the diffraction patterns. The successful sample preparation method will be used on deformed samples to detect lattice rotation trends.

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