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High strain-rate interfacial behavior of layered metallic composites

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ABSTRACT

The high strain-rate interfacial behavior of layered aluminum composite has been investigated. A dislocation-density based crystalline plasticity formulation, specialized finite-element techniques, rational crystallographic orientation relations, and a new fracture methodology for large scale plasticity been used. Two alloy layers, a high strength alloy, aluminum 2195, and an aluminum alloy 2139, with high toughness, were modeled with representative microstructures that included precipitates, dispersed particles, and different grain boundary (GB) distributions. The new fracture methodology, based on an overlapping element method and phantom nodes, along with a fracture criteria specialized for fracture on different cleavage planes is used to model interfacial delamination. Dislocation-density evolution significantly affects the delamination process, and this has a directly related to the strengthening, toughening, and failure of the layered composite. It is also shown that brittle alumina (Al_2O_3) platelets in the interface region played an important role in interfacial delamination and overall composite behavior.

1. Introduction

Layered metallic composites are widely used in applications, which require high strength, toughness, and damage tolerance. The interfaces between constituent layers play an essential role in thermo-mechanical deformation and failure. Interfacial delamination is quite common in layered metallic composite for both quasi-static and dynamic deformations. Delamination in the layered composite can absorb more energy in the deformation process as compared to the energy absorbed by a homogeneous layer of the same thickness comprised of either of the constituents. The extrinsic toughening mechanisms that help to improve the fracture resistance of a layered composite include crack deflection, crack blunting, crack front convolution, stress redistribution, and local plane stress deformation. Interfacial delamination can activate these extrinsic toughening mechanisms (Lesuer et al., 1996). Interface delamination can change the failure mode from shear localization to a mode characterized by energy absorption due to bending and stretching for higher strain rate ballistic impact test (Lesuer et al., 1996). Cepeda-Jiménez et al. have also experimentally shown, for low-strain rate experiments, that the extrinsic toughening mechanisms of interface delamination in a roll-bonded Al-alloy layered composite significantly improves impact toughness by crack renucleation in the next layer (Cepeda-Jiménez et al., 2009). Higher ductility of the layer materials inhibits delamination, whereas the delamination is more likely to happen with a brittle interface (Cepeda-Jiménez et al., 2008). Brittle surface oxide layer, which can have alumina (Al_2O_3) particles, due to roll-bonding, and these brittle platelets can accelerate delamination by initiating cracks during deformation of layered aluminum composites (Cepeda-Jiménez et al., 2008; Barlow et al., 2004).

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Therefore, in this article, we will investigate the interfacial effects on dynamic behavior of layered aluminum composite for different delamination scenarios. The layered composite configuration to be used here is a roll-bonded layered composite of aluminum alloys 2195 and 2139 (Khanikar and Zikry, 2014). In this study, we will use a configuration with the high strength 2195 aluminum alloy on top to investigate how high strain-rate delamination evolves in the layered composite under different conditions. The role of interfacial platelets of alumina on delamination behavior will also be studied. A new fracture approach, which is based on modifying the overlapping element method, will be used with a fracture criteria specialized for different cleavage planes, microstructures, and large scale plasticity. This method surmounts some of the issues associated with other fracture methods, such as XFEM and cohesive fracture.

This paper is organized as follows: the microstructurally-based dislocation–density crystalline plasticity formulation and computational approach are presented in Section 2; the microstructural representations are described in Section 3, the computational representation of the fracture surface and the failure criteria are presented in Section 4, the results and discussions are outlined in Section 5 and finally the conclusions are given in Section 6.

2. Constitutive formulation and computational approach

2.1. Dislocation-density based crystalline plasticity formulation

Formulations for the rate-dependent multiple-slip crystalline plasticity, which are coupled to the evolutionary equations for the dislocation densities, were used in this analysis. Detailed presentation can be found in the references (Orsini and Zikry, 2001; Zikry and Kao, 1996; Ashmawi and Zikry, 2002). Only a brief outline of the approach will be presented here.

The velocity gradient can be decomposed into a symmetric deformation rate tensor D_{ij} and an antisymmetric spin tensor W_{ij}. D_{ij} and W_{ij} can then be additively decomposed into elastic and plastic components as

$$D_{ij} = D^*_{ij} + D^p_{ij}, \tag{1a}$$

$$W_{ij} = W^*_{ij} + W^p_{ij}. \tag{1b}$$

The inelastic parts are defined in terms of the crystallographic slip rates as

$$D^p_{ij} = P^{(\alpha)}_{ij}\dot{\gamma}^{(\alpha)}, \tag{2a}$$

$$W^p_{ij} = \omega^{(\alpha)}_{ij}\dot{\gamma}^{(\alpha)}, \tag{2b}$$

where α is summed over all slip systems, and $P^{(\alpha)}_{ij}$ is the symmetric and $\omega^{(\alpha)}_{ij}$ is the antisymmetric parts of the Schmid tensor in the current configuration.

The rate-dependent constitutive description on each slip system can be characterized by a power law relation, for strain rates below a critical value of $\dot{\gamma}_{critical}$ as

$$\dot{\gamma}^{(\alpha)} = \dot{\gamma}^{(\alpha)}_{ref}\left[\frac{\tau^{(\alpha)}}{\tau^{(\alpha)}_{ref}}\right]\left[\frac{|\tau^{(\alpha)}|}{\tau^{(\alpha)}_{ref}}\right]^{(\frac{1}{m}-1)}, \tag{3}$$

where $\dot{\gamma}^{(\alpha)}_{ref}$ is the reference shear strain rate, which corresponds to a reference shear stress $\tau^{(\alpha)}_{ref}$, and m is the rate sensitivity parameter. Above the critical strain rate $\dot{\gamma}_{critical}$, where the phonon drag is assumed to dominate, m is taken as 1. The reference stress τ^{α}_{ref} that was used here is a modification of widely used classical forms (Mughrabi, 1987) that relate the reference stress to a square-root dependence on the dislocation-density ρ_{im} as

$$\tau^{(\alpha)}_{ref} = \left(\tau^{(\alpha)}_y + G\sum_{\beta=1}^{nss} a_{\alpha\beta} b^{(\beta)}\sqrt{\rho^{(\beta)}_{im}}\right)\left(\frac{T}{T_0}\right)^{-\xi}, \tag{4}$$

where $\tau^{(\alpha)}_y$ is the static yield stress on slip system α, G is the shear modulus, $b^{(\beta)}$ is the magnitude of the Burgers vector for slip system β, and the coefficients $a_{\alpha\beta}$ are the slip system interaction coefficients, T is the temperature, T_0 is the reference temperature, and ξ is the thermal softening exponent.

The rate of change of temperature due to the high strain-rate deformation of the crystal is a function of adiabatic heating. The temperature evolution relation can be obtained from the balance of energy. Assuming adiabatic conditions, the thermal conduction rate can be considered as negligible and the rate of the plastic work can be given by

$$\dot{T} = \frac{\chi}{\rho C_p}\sigma^{dev}_{ij}D^p_{ij}, \tag{5}$$

where χ is the fraction of the plastic work converted to heat, σ^{dev}_{ij} is the deviatoric stress, ρ is the material density, and c_p is the specific heat of the material. For adiabatic high strain-rate applications, it is assumed that thermal conduction is negligible.

2.2. Dislocation density evolution

The crystalline plasticity constitutive formulation has been coupled with dislocation density evolutionary equations to bridge microscopic dislocation activities with macroscopic deformation process. For a given deformed state of the material, it is assumed that the total dislocation density $\rho^{(\alpha)}$ can be additively decomposed into a mobile and an immobile dislocation density $\rho^{(\alpha)}_m$ and $\rho^{(\alpha)}_{im}$ as

$$\rho^{(\alpha)} = \rho^{(\alpha)}_m + \rho^{(\alpha)}_{im}. \tag{6}$$

It is assumed that an increment of strain results in a change in the dislocation structure. The balance between generation and annihilation of dislocation densities as a function of strain was thus taken as a basis for the following equations that describe the evolution of mobile and immobile dislocation densities as

$$\frac{d\rho^{(\alpha)}_m}{dt} = |\dot{\gamma}^{(\alpha)}|\left[\frac{g_{sour}}{b^{(\alpha)}b^{(\alpha)}}\left(\frac{\rho^{(\alpha)}_{im}}{\rho^{(\alpha)}_m}\right) - \frac{g_{minter}}{b^{(\alpha)}b^{(\alpha)}}\exp\left(-\frac{\Delta H}{kT}\right) - \frac{g_{immob}}{b^{(\alpha)}}\sqrt{\rho^{(\alpha)}_{im}}\right], \tag{7a}$$

$$\frac{d\rho^{(\alpha)}_{im}}{dt} = |\dot{\gamma}^{(\alpha)}|\left[\frac{g_{minter}}{b^{(\alpha)}b^{(\alpha)}}\exp(-\frac{\Delta H}{kT}) + \frac{g_{immob}}{b^{(\alpha)}}\sqrt{\rho^{(\alpha)}_{im}} - g_{recov}\exp(-\frac{\Delta H}{kT})\rho^{(\alpha)}_{im}\right], \tag{7b}$$