Dynamic delamination of patterned thin films

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We investigate laser-induced dynamic delamination of a patterned thin film on a substrate. Controlled delamination results from our insertion of a weak adhesion region beneath the film. The inertial forces acting on the weakly bonded portion of the film lead to stable propagation of a crack along the film/substrate interface. Through a simple energy balance, we extract the critical energy for interfacial failure, a quantity that is difficult and sometimes impossible to characterize by more conventional methods for many thin film/substrate combinations. © *2008 American Institute of Physics*. [DOI: 10.1063/1.3056639]

Thin film adhesive failure and attendant delamination are long-standing problems hampering the performance of multilayer, microelectronic devices. High-strain rate failure of interfaces in multilayer microelectronic and micro-electrical-mechanical devices is an increasingly important reliability issue and little is known about how a thin film interface responds to dynamic forces. Interface adhesion is characterized by two properties: the interface strength, i.e., the critical traction necessary to separate the thin film from the substrate, and the interface fracture toughness, i.e., the energy required to propagate a crack along the interface. Although significant effort has been devoted to the quasistatic measurement of thin film adhesion,[1] current methods are often inadequate for quantitative analysis of interfacial failure in complex, multilayer devices.

Common test methods for film adhesion such as peel, stud-pull, scratch, blister, indentation, and superlayer delamination[1] subject the film to high stresses, resulting in complex, plastic deformations that are difficult to analyze. The inability to decouple the inelastic work from the total work of adhesion precludes accurate measurement of the interfacial fracture energy. Moreover, many of these test methods are unable to produce interface failure for strongly adhered thin films. Fracture mechanics based techniques, such as the double cleavage drilled compression[2] and four-point bend[3] tests, are limited by intricate sample preparation and bonding processes, which can cause undesirable compositional changes. In some thin film systems, it is difficult to introduce a starter crack and grow it in a controlled manner along the desired interface.

Laser spallation techniques[4–6] dynamically load the thin film interface in a precise, noncontacting manner with a laser-induced, high amplitude acoustic pulse. A nearly one-dimensional, compressive, longitudinal wave packet is generated on the back side of the substrate with a shape similar to the laser pulse and propagated through the substrate towards the film/substrate interface. Upon reaching the free surface of the thin film, the stress pulse reflects and loads the interface in tension with strain rates of the order of 10^7/s, minimizing plastic deformation within the film.[7] At a critical stress level, the interface fails and the film spalls from the substrate. A significant limitation of spallation testing, however, is that only the interface strength (critical stress for spallation) is characterized, rather than the interfacial fracture energy. The interface strength is associated with crack initiation, while interface toughness controls crack propagation, more closely associated with delamination failure. Initial strides to characterize thin film interfacial fracture energy using laser-induced stress waves involved propagating a line flaw underneath a buckled thin film.[6]

Although the significant role of kinetic energy in sustaining crack growth is well established in bulk materials, in this Letter we demonstrate the ability to effectively channel the inertial energy associated with rapid, high-amplitude acoustic waves to achieve controlled dynamic fracture of a thin film interface. Through computational analysis of the transient delamination physics, we identify an experimental protocol to directly measure the interfacial fracture energy. We build upon the results of a recent study[8] of interface failure due to dynamic loading at the edge of a patterned thin film. Our simulations revealed that the stress concentration at the corner of the film initiated an edge crack. The kinetic energy trapped in the debonded portion of the thin film near the edge caused the interface crack to extend a distance several times the film thickness and for a significant amount of time beyond the end of the loading event. However, the resulting delamination (ca. 100 μm) was too small to reliably extract interfacial energy. Here, we enhance the edge delamination of the film by introducing a weak adhesion region, which essentially serves as a precrack upon loading [Fig. 1(a)]. The new thin film pattern geometry ensures the kinetic energy in the weakly bonded portion of the film is effectively channeled to the interface, leading to controlled crack propagation several millimeters in length.

Patterned aluminum (Al) thin film specimens are produced by depositing on a silicon (Si) substrate, a 310 μm wide, 380 μm long, and 250 nm thick gold (Au) rectangular film as a weak adhesion layer followed by a 310 μm wide and 2.8 μm thick aluminum (Al) strip as shown in Fig. 1(a). For the generation of laser-induced stress pulses, we deposit a 400 nm thick Al absorbing and 10 μm thick water glass

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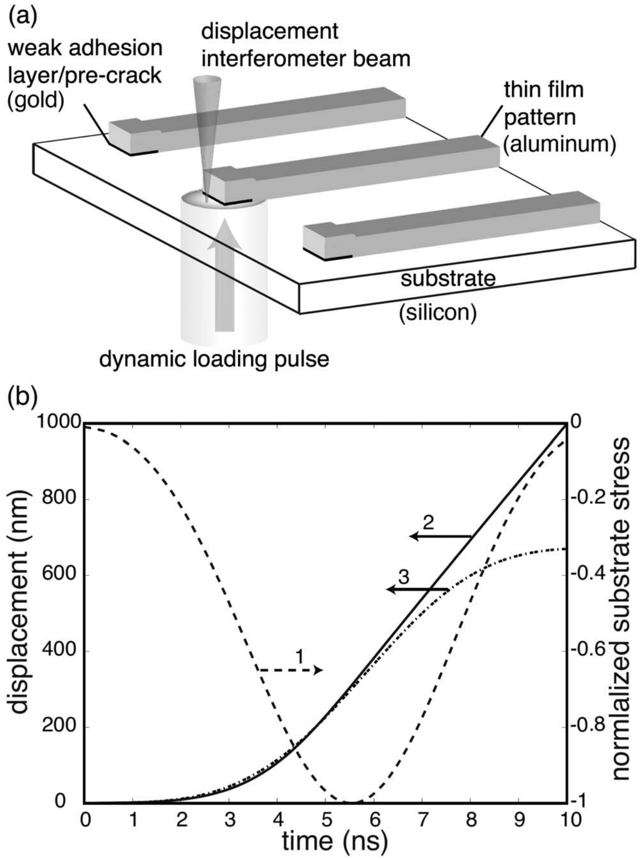

FIG. 1. (a) Schematic of dynamic adhesion protocol. (b) Measured substrate stress profile (1) and corresponding experimental displacement histories of the weakly (2) and strongly bonded (3) portion of the thin film during experiments.

constraining layers[4] on the back of the substrate. A Nd: yttrium aluminum garnet (YAG) laser pulse ($\lambda = 1064$ nm) is focused to a 1.5 mm diameter spot on the absorbing layer to launch a high amplitude, compressive, acoustic pulse with a 5 ns rise time. The stress pulse profile [Fig. 1(b)] is characterized through interferometric measurements of the out-of-plane displacements of the substrate surface.[4] Figure 1(b) also compares the corresponding displacement histories on the free surface of the film above the weakly and strongly bonded regions.

To provide insight on the dynamic delamination process, a numerical scheme is developed to study the dynamic failure of thin films and layered structures. The scheme is based on a combination of an explicit spectral scheme used to model the dynamic response of the substrate[9] and a dynamic finite element representation of the thin film. The spectral solver is coupled to the explicit finite element model through a rate-independent, state-dependent cohesive model that relates the cohesive tractions (T_n and T_t) acting along the interface to the associated displacement jumps (δ_n and δ_t). As described in the supplemental information, the bilinear cohesive model[8] adopted in this study is characterized by four key parameters: the tensile ($\sigma_{\max}$) and shear ($\tau_{\max}$) failure strengths obtained from spallation experiments[4] and the opening/tensile mode I (G_{Ic}) and in-plane shear mode II (G_{IIc}) fracture toughnesses, i.e., the area under the traction-separation curve[8] extracted from the current delamination experiments.

The presence of the Au weak adhesion region is modeled by using a lower failure strength and fracture toughness between the crack tip and the end of the film. To reduce the computational time, we simulate a model geometry with a 50 μm long, 200 nm thick Au weak adhesion layer introduced below a 2 μm thick Al film.

The simulation is initiated by the arrival of the pressure pulse at the film/substrate interface. The pulse shape (Gaussian), amplitude (0.7 GPa), and duration (10.6 ns) are obtained from experiments [Fig. 1(b)]. The evolution of the normal traction at various locations along the interface is shown in Fig. 2(a). After the pressure pulse reflects from the top surface of the film as a tensile wave, the interface traction quickly increases and failure ensues. The interface stress is the same at all locations beyond the crack tip region during the initial stress wave loading phase (denoted by open circles). The Au/Si interface in the weak adhesion region debonds at ca. 6.5 ns which parallels the experimental change in displacement between the weakly and strongly bonded portions of the film (curves 2 and 3) in Fig. 1(b). Along the Al/Si interface, which is the interface of interest, failure initiates at the crack tip due to the local stress concentration and quickly propagates ahead. The delamination process continues past the end of the laser-induced pulse, as apparent from the curves corresponding to observation points located at 6, 9, and 12 μm ahead of the initial crack tip. This process is associated with the aforementioned inertial effect, i.e., with the transfer of the kinetic energy stored in the debonded region of the thin film located above the weak adhesion layer.

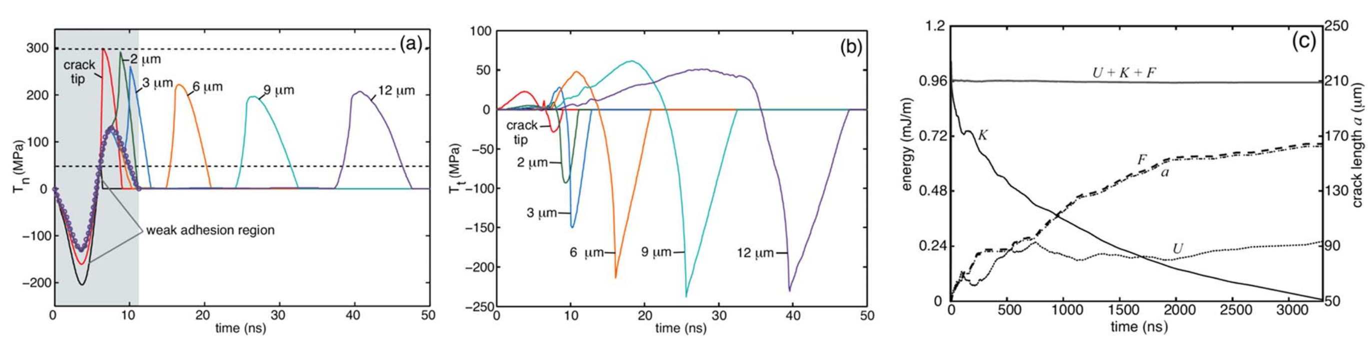

FIG. 2. (Color online) Evolution of traction stresses along the Si/Al interface: (a) tensile T_n and (b) shear T_t. The curves are labeled by the distance of the point of observation to the corner. The dashed top and bottom horizontal lines in (a) denote the strengths ($\sigma_{\max}$) of the Al/Si and Au/Si interfaces, respectively. The shaded region in (a) corresponds to the duration of the stress wave loading. (c) Evolution of fracture (F), kinetic (K), and strain (U) energy components in the film (left axis) and of the crack length (a) (right axis).

As illustrated in Fig. 2(b), the failure process is of mixed-mode nature, with the tangential traction switching sign as the crack tip approaches the point of observation. The level of mode mixity (i.e., the ratio of in-plane shearing to pure tensile opening) evolves during the delamination event, especially during the initial phase, where the failure process takes place primarily under tensile conditions. However, during the second (and much longer) phase of the delamination event, which is driven by the kinetic energy stored in the debonded portion of the film, the mode mixity appears nearly constant until the crack runs out of driving force and finally stops.

Figure 2(c) shows the evolution of the various components of energy[8,10] during the entire failure process, i.e., from the initiation to the arrest of the delamination front. The failure initiation event is characterized by the rapid accumulation of kinetic energy in the thin film above the weak adhesion region followed by the debonding of the Au/Si interface, starting at about 6 ns after the arrival of the substrate pulse at the interface (corresponding to $t=0$). The spallation event initiates the delamination process along the Al/Si interface, with the failure energy F reflecting a decrease in the kinetic energy K, while the strain energy U stored in the debonded portion of the film remains at about 20% of the total energy. As the length of the crack increases, the strain energy represents an even smaller portion of the total energy imparted to the system and the fracture energy follows more closely the decrease in kinetic energy. The total energy $K+F+U$ remains remarkably constant, indicating that very little energy leaks from the film to the substrate during the delamination event. Finally, as apparent from the evolution of the fracture energy F, which closely follows that of the crack length for the rate-independent cohesive model used in this study, the crack propagation event is quite unsteady, with periods of crack arrests followed by rapid crack extensions. In this illustrative simulation, the final crack extent is more than three times the initial length of the weak adhesion region.

The dynamic delamination protocol is demonstrated for patterned Al films on Si as shown in Fig. 3. The final delamination for the Al thin films extend to 6 ± 0.2 mm in length, which is three orders of magnitude greater than the film thickness, twenty times greater than the width of the weak adhesion region, and eight times larger than the spot size of the laser pulse. All four tests were conducted at the same laser fluence ($28.5\ \mathrm{mJ/mm^2}$) and the resulting delamination lengths were repeatable. Based on the numerical analysis, we extract the interface toughness G by assuming all the kinetic energy trapped in the thin film at the onset of spallation of weak interface of length a_o (precrack length) is expended into fracture energy leading to a final crack extension a_f-a_o, through the simple relation $G=Ka_o/(a_f-a_o)$.

The kinetic energy per unit area (K) is determined by using the substrate stress pulse information in a one-dimensional analysis of the wave propagation in the multilayer specimen (substrate, weak adhesion layer, and thin film system). Following this procedure, we are able to extract the Al/Si interfacial fracture energy as $5.6\ \mathrm{J/m^2}$. This result agrees well with the reported value of interfacial energy for an Al film on a Si substrate measured using superlayer test.[1]

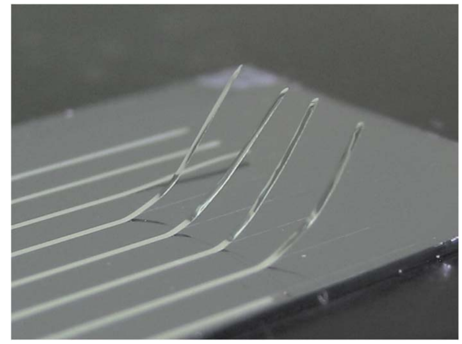

FIG. 3. (Color online) Al film patterns (2.8 μm thick and 310 μm wide) after dynamic delamination from a Si (100) substrate. A final delamination length of ca. 6 mm develops from a 380 μm long initial precrack (weak adhesion layer).

In summary, we have achieved controlled dynamic crack growth along a specific thin film/substrate interface and extracted the associated interfacial fracture energy. By selectively modifying the substrate surface to create a weak adhesion region beneath the patterned film, the kinetic energy developed by the initial spallation of the weakly adhered region is shown to effectively transfer to the interface, resulting in controlled delamination of the film long after the initial stress pulse has passed. The interfacial fracture energy of an Al film patterned on a Si substrate is calculated directly from delamination measurements with good repeatability. Overall, the combined experimental and computational methodology offers tremendous potential to address the critical problem of adhesive failure and delamination in multilayer thin film devices.

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