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Atomistic fracture energy partitioning at a metal-ceramic interface using a nanomolecular monolayer

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ABSTRACT

We report an experimental approach to partition the fracture energy of a metal-ceramic interface into work of adhesion and plasticity, unveiling the nanomechanics of interfacial deformation and fracture. We obviate crack path uncertainties by constraining fracture to occur in an interfacial nanomolecular layer through fissure of a single bond type whose strength is varied by adjusting the chemical environment. This approach is adaptable for studying interfacial fracture and related phenomena in diverse materials systems in different thermochemical environments.

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I. INTRODUCTION

The partitioning of interface fracture energy into bond breaking energy and plastic energy in ductile layer(s) constituting the interface is a central problem in fracture mechanics. Fracture energy (or toughness) Γ_{FT} is typically comprised of multiple processes such as bond-breaking¹, corrosion², and crack-tip plastic flow and repair^{3,4}. The Griffith and Irwin model^{1,3} partitions Γ_{FT} into two terms γ_a and γ_p : $\Gamma_{\text{FT}} = \gamma_a + \gamma_p$, where γ_a is the work of adhesion, and γ_p is the plastic energy which is a function of γ_a . Cohesive fracture in homogenous bulk materials has been studied for many decades and empirical descriptions of fracture have been developed by approximating $\Gamma_{\text{FT}} \sim \gamma_a$ for brittle materials ($\gamma_p \sim 0$), and $\Gamma_{\text{FT}} \sim \gamma_p$ for ductile materials ($\gamma_p \gg \gamma_a$). However, currently there are no models that incorporate experimentally-validated γ_a and γ_p terms to obtain realistic atomistic descriptions of fracture at interfaces comprised of both brittle and ductile materials, i.e., where γ_a and γ_p are comparable and neither term can be neglected. Characterizing interfacial fracture has remained an exacting experimental challenge due to uncertainties in crack path and types of bonds broken. Although factors such as film thickness, crack tip reaction kinetics and humidity are known to influence fracture energy^{5,6}, quantitative partitioning of Γ_{FT} in terms of specific atomistic processes has been elusive. As a result, current atomistic descriptions of interfacial fracture energetics are primarily based on theoretical models and computer simulations^{7,8}.

Here, we demonstrate an experimental approach to partition Γ_{FT} of a metal-ceramic interface by using a nanomolecular monolayer (NML) at the interface. Crack path uncertainties are obviated by constraining fracture to occur through the fissure of only one bond type in the NML at the interface. Additionally, we vary the interfacial bond strength, and hence γ_a , by controlling the chemical environment, thereby allowing γ_p determination by measuring Γ_{FT} . We illustrate this approach for a model copper-silica interface treated with an organosilane NML where bond-breaking occurs by siloxane bridge hydrolysis. This method could be adapted for studying interface fracture and related phenomena in

different thermochemical environments, e.g., stress corrosion cracking, electromigration and stress-induced voiding, in a variety of materials systems through appropriate choice of NMLs.

II. EXPERIMENTAL DETAILS

Our test structures consisted of a copper-silica interface functionalized with a 0.7-nm-thick NML comprised of mercapto-propyl-tri-methoxy-silane (MPTMS) molecules. NMLs were formed on Si(001) wafers capped with an 85-nm-thick silica by dipping the wafers in a 5 mM MPTMS in toluene for 30 min in flowing N_2 as described previously^{9, 10}. A 35-nm-thick Cu layer was deposited on the NML/SiO₂/Si(001) followed by a 150-nm-thick Ta layer in a CVC DC sputter system with a 5×10^{-7} Torr base pressure. The samples were annealed at 300 °C for 30 min, and the Ta/Cu/NML/SiO₂/Si(001) was bonded to a dummy Si wafer with an epoxy, for four point bend tests. The Ta layer was used to offset poor Cu-epoxy adhesion. Our earlier works^{9, 10} have shown that the NML is strongly bound to the copper overlayer via thioligation and bond-breaking occurs exclusively at the NML-silica interface through the fissure of Si-O-Si bridges formed during molecular assembly (Fig. 1a). We vary γ_a through moisture control by exploiting the sensitivity of siloxane bonds to hydrolysis^{9, 11}.

We determined the fracture energy of thin film sandwiches with Cu/NML/SiO₂ interfaces by four-point-bend tests carried out *in situ* in environments with different water activity $0.2 \leq a_{H_2O} \leq 0.8$ and temperature $278 \text{ K} \leq T \leq 358 \text{ K}$ in a CSZ PZ series chamber. We note that a_{H_2O} is identical to relative humidity, and represents the water chemical potential μ_{H_2O} when plotted on a log scale (as we have), because $\mu_{H_2O} = NRT \log a_{H_2O}$, where R is the gas constant and T the absolute temperature, and N the number of interfacial bonds per unit area. We strained 5 mm x 40 mm rectangular beams of Si/epoxy/Ta/Cu/NML/SiO₂/Si(001) stacks at a 10 nm/s displacement until a crack emerging from a fine notch scribed on the Si substrate propagated to the weakest interface (see green line in Fig. 1b), an event indicated by the first plateau in the load-displacement curve corresponding to a critical driving energy Γ_c .

At the plateau, we halted the externally applied displacement to determine the minimum sub-critical energy required for crack propagation (see red line in Fig. 1b). The delamination velocity v decreases monotonically and approaches zero as the remnant driving energy Γ diminishes to a threshold value that corresponds to the fracture toughness Γ_{FT} at equilibrium of the atomistic process that drives the crack. Thus, Γ_{FT} is extracted from v - Γ curves as $v \rightarrow 0$ and $d\Gamma/dt \rightarrow 0$ as described elsewhere¹². In our tests, the phase angle was $\sim 43^\circ$ corresponding to approximately equal parts of shear and normal loading at the interface¹³. Each Γ_{FT} value reported here is an average of at least four tests; the error bars denote one standard deviation.

Fracture surfaces were analyzed by X-ray photoelectron spectroscopy (XPS) to determine the fracture path and obtain insights into the bond-breaking mechanism. XPS measurements were carried out in a PHI 5400 instrument with a Mg K α beam at a 23.5 eV analyzer pass energy. Charging-induced spectral shifts were corrected by using the adventitious C 1s peak at 285 eV as an internal calibration.

III. RESULTS AND DISCUSSION

Fracture toughness tests show that Γ_{FT} is a strong function of both temperature and water activity a_{H_2O} (see Figs. 2a-b). Specifically, Γ_{FT} decreases with increasing a_{H_2O} at constant temperature as well as with increasing temperature at constant a_{H_2O} . The latter is due to an increase in the water vapor pressure p_{H_2O} driven by an increase in saturation vapor pressure p_{sat} with temperature ($a_{H_2O} = p_{H_2O}/p_{sat}$)¹⁴, consistent with previous studies of Cu/TaN/SiO₂ interfaces⁵. We attribute the Γ_{FT} decrease with a_{H_2O} to water-induced siloxane bridge fissure at the interface similar to stress corrosion cracking in glasses¹⁵ as corroborated by fracture surface spectroscopy measurements. XPS spectra from Cu fracture surfaces show the Si 2p core-level band at ~ 102.5 eV characteristic of silyl alkyl moieties in the organosilane, in contrast to the silica Si 2p signature at ~ 103.5 eV seen on silica fracture surfaces (see Fig. 3), confirming that delamination occurs via siloxane bond fissure at the NML-silica interface¹⁰. This delamination pathway remains

unchanged for $0.05 \leq a_{\text{H}_2\text{O}} < 0.8$, as seen from fracture surfaces showing essentially identical spectral features. At the highest water activity in our experiments, i.e., $a_{\text{H}_2\text{O}} \sim 0.8$, Cu fracture surfaces exhibit a silica sub-band along with a prominent the silyl alkyl signature, suggesting some fracture path deviation due to bond breaking in the silica substrate.

Plotting Γ_{FT} as a function of $a_{\text{H}_2\text{O}}$ for $T = 293$ K reveals two distinct regimes separable at $a_{\text{H}_2\text{O}} \sim 0.5$ - 0.4 (see Fig. 4a). For $a_{\text{H}_2\text{O}} \leq 0.4$, Cu/NML/silica interfaces exhibit Γ_{FT} ranging from ~ 2.1 to 3.6 J/m², which are comparable or significantly higher than $\gamma_a \sim 1.75$ J/m² reported for fused silica¹⁶, pointing to the existence of secondary energy dissipation processes such as copper plasticity. The twofold higher magnitude of $d\Gamma_{\text{FT}}/d\log a_{\text{H}_2\text{O}}$ for $a_{\text{H}_2\text{O}} < 0.4$ - 0.5 than that observed at higher $a_{\text{H}_2\text{O}}$ allows us to separate the relative extents of γ_a and γ_p at the Cu/NML/silica interface.

At high moisture contents, e.g., for $a_{\text{H}_2\text{O}} > 0.4$, the main energy dissipation mechanism is siloxane bond fissure via hydrolysis through the reaction $\text{Si-O-Si} + \text{H}_2\text{O} \rightarrow 2 \text{Si-OH}$, implying that $\Gamma_{\text{FT}} = \gamma_a$. The linear behavior of Γ_{FT} with $\log a_{\text{H}_2\text{O}}$ is reminiscent of the linear dependence of the equilibrium free energy of this reaction on $\log a_{\text{H}_2\text{O}}$ ¹⁷ that yields the interfacial Si-O-Si bond density $N_{\text{Si-O-Si}}$, and hence γ_a . Thus, $\log a_{\text{H}_2\text{O}}$ serves as a proxy for γ_a . Using $d\gamma_a/d\log a_{\text{H}_2\text{O}} = N_{\text{Si-O-Si}}RT$, we get $N_{\text{Si-O-Si}} = 2 \times 10^{20} \text{ m}^{-2}$, which agrees well with $N_{\text{Si-O-Si}} = 0.7 \times 10^{20} \text{ m}^{-2}$ obtained by fitting v - Γ characteristics for different $a_{\text{H}_2\text{O}}$ to a reaction rate kinetics model for water-induced fracture in orthosilicates^{15, 18}. The fit also yields a crack propagation activation energy of ~ 0.3 eV/bond reported for siloxane hydrolysis, confirming that $\Gamma_{\text{FT}} = \gamma_a$ for $a_{\text{H}_2\text{O}} > 0.4$.

The abrupt increase in the slope $d\Gamma_{\text{FT}}/d\log a_{\text{H}_2\text{O}}$ below $a_{\text{H}_2\text{O}} \sim 0.5$ - 0.4 reflects the onset of an additional energy dissipation mechanism, identified to be copper plasticity. We discount the possibility of deformation in the silica layer because this mechanism is operative only above 900°C ¹⁹. If copper

plasticity is responsible for the higher $d\Gamma_{FT}/d\log a_{H_2O}$ for $a_{H_2O} < 0.4$, the copper film should yield at $a_{H_2O} \sim 0.4-0.5$. To verify this requirement, we further examined the threshold work of adhesion γ_0 ($= \gamma_a = \Gamma_{FT}$ at yield point) value of $\sim 2 \text{ J/m}^2$ and estimated the yield stress σ_y from the strain energy per unit volume at $a_{H_2O} \sim 0.4-0.5$. Since the loading phase angle in our four point bend tests is $\sim 43^\circ$, the normal and shear mode components are roughly equal, i.e., $\Gamma_I \sim \Gamma_{II} \sim \frac{\gamma_0}{2} \sim 1 \text{ J/m}^2$. This value is in good agreement with the dislocation emission energy of $\sim 0.84 \text{ J/m}^2$ for mode I yielding according to the Rice-Thompson description²⁰, and is consistent with the expectation that Γ_I is the main contributor to interfacial bond cleavage and plasticity whilst Γ_{II} accounts for frictional effects. The yield stress σ_y calculated from $\Gamma_I \sim \frac{\gamma_0}{2}$ for different copper thicknesses by assuming copper stiffness $E=130 \text{ GPa}$ ²¹ are within $\sim 15-30\%$ of σ_y obtained from nanoindentation measurements (see Fig. 4b), confirming that $d\Gamma_{FT}/d\log a_{H_2O}$ increase for $a_{H_2O} < 0.4$ is due to copper plastic deformation. Moreover, the observed linear increase in Γ_{FT} with $\log a_{H_2O}$ for $a_{H_2O} < 0.4$ (Fig. 4a) is consistent with increased plastic deformation because the plastic zone size R_0 is expected to scale linearly with γ_a (and hence $\log a_{H_2O}$) as²² $R_0 = \frac{E\gamma_a}{3\pi(1-\nu^2)\sigma_y^2}$ for constant σ_y , where ν is the Poisson's ratio. Finally, our observation of Γ_{FT} increase with Cu film thickness (see Fig. 5c) for a fixed a_{H_2O} in the region $a_{H_2O} < 0.4$ regime also confirms copper plasticity because lower σ_y and larger plastic zones in thicker films are expected to abet plastic deformation^{21,23}.

The above discussion indicates that measuring Γ_{FT} as a function of $\log a_{H_2O}$ (or μ_{H_2O}) could be an attractive alternative way of determining thin film yield stress. The σ_y values extracted from Γ_{FT} - $\log a_{H_2O}$ characteristics are similar to, but consistently lower than, that from nanoindentation (Fig. 4b), and higher than that predicted by molecular dynamics simulations²⁴ of nanograined bulk copper as well as extrapolations of nanoindentation results from thicker films²⁵. These results suggests that obtaining σ_y by

our method mitigates the so-called substrate effect invoked to explain high σ_y obtained from nanoindentation of sub-200-nm-thick films.

For $a_{\text{H}_2\text{O}} < \sim 0.4$, fracture toughness Γ_{FT} includes non-negligible contributions from both γ_a and γ_p . We can calculate γ_p by calculating the difference between the measured toughness Γ_{FT} and the γ_a obtained by extrapolating the γ_a vs. $\log a_{\text{H}_2\text{O}}$ linear fit for high $a_{\text{H}_2\text{O}}$ to the $a_{\text{H}_2\text{O}} < 0.4$ regime. This extrapolation is justified because γ_a is sensitive only to $a_{\text{H}_2\text{O}}$, as described earlier. Thus, we can plot γ_p as a function of γ_a (see Fig. 5a) by the equation $\gamma_p = \tau(\gamma_a - \gamma_0)$, where $\gamma_0 = \gamma_a$ at yield point. For $\gamma_a < \gamma_0$, the plastic energy $\gamma_p = \tau = 0$, and for $\gamma_a > \gamma_0$, γ_p increases linearly with γ_a with $\tau \sim 1.2$. Therefore, besides separating the contributions of the work of adhesion and plasticity, our results reveal the quantitative dependence of γ_p on γ_a . While a multiplicative effect has been postulated²⁶, and analytical models predict $0.2 < \tau < 8$ for hetero-interfaces²¹, our experimental results here provide a new means for directly determining τ . Earlier experimental works²⁷⁻³⁰ based on a dislocation model and measurement of Cu films with different thicknesses have also related γ_p and γ_a . But, our approach obviates complications of yield stress variations in films of different thicknesses and offers versatility for studying the effects of environmental degradation, segregation, and molecular functionalization. This validation provides a means to refine Griffith-Irwin descriptions beyond idealized interfaces to realistic hybrid materials and heterointerfaces where neither γ_p nor γ_a can be neglected.

Finally, by carrying out experiments similar to described above at different temperatures, we can obtain deformation mechanism maps for the interfacial delamination as a function of thermochemical variables such as temperature and water activity (see Fig. 5b). At all temperatures, γ_a increases with decreasing $a_{\text{H}_2\text{O}}$ due to decreased hydroxylation-induced interfacial siloxane bridge weakening. At low $a_{\text{H}_2\text{O}}$, γ_a is high, and Γ_{FT} increases with increasing temperature due to increased γ_p . At high $a_{\text{H}_2\text{O}}$, $\Gamma_{\text{FT}} \sim \gamma_a$ because the interfacial bonds are too weak to support copper plasticity; γ_a decreases with increasing

temperature due to facile bond breaking at elevated temperatures. Generation of interfacial deformation maps by our approach for other materials systems would open up new ways of studying the effects of thermochemical parameters on interfacial fracture to model phenomena such as stress corrosion cracking, stress- and electro-migration that are accentuated in devices and systems comprised of nanomaterials.

IV. CONCLUSIONS

In summary, we have demonstrated an experimental approach using molecular functionalization and environmental control to quantitatively partition the fracture energy Γ_{FT} of a heterointerface into work of adhesion γ_a and plastic energy γ_p . We employed a model metal-ceramic interface with a nanomolecular monolayer to force fracture to occur in a confined nanoscopic plane by breaking exactly one kind of interfacial bond whose strength can be adjusted by altering the chemical environment. This approach obviates uncertainties in crack path, and number and type of bonds broken, and allows the determination of the work of adhesion γ_a and extraction of plastic energy γ_p and its dependence on γ_a . Our results provide a means to quantitatively describe fracture energetics using the Griffith-Irwin model without neglecting either γ_p or γ_a . This framework obviates errors arising from neglecting either term and paves the way for rational design and accurate stability modeling of materials and structures comprised of heterointerfaces, e.g., composites, coatings, material joints, nanoelectronics devices and packaging. Additionally, it provides an alternative means for determining the ductile layer yield stress σ_y . Finally, our approach is attractive for adaptation for studying the effects of environmental degradation, segregation, and molecular functionalization on the nanomechanics of interfacial fracture in diverse materials systems through appropriate choice of nanomolecular layers and chemical environments.

V. ACKNOWLEDGEMENTS

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FIGURE CAPTIONS

Figure 1. (a) Schematic sketch depicting water-induced siloxane bond breaking in an organosilane NML at a Cu/NML/silica interface. (b) A typical load-relaxation curve from a four point bend test with a schematic showing the crack path. A crack initiated at the notch tip on the substrate traverses to the weakest interface and propagates laterally at a constant load corresponding to Γ_c (green). When displacement is arrested, the crack is driven sub-critically by the remnant load (red) which continually decreases until crack growth stops. The remnant load at the crack arrest point corresponds to the equilibrium fracture toughness Γ_{FT} .

Figure 2. (a) Crack velocity v vs. the driving energy Γ plots obtained at $T=278$ K at different a_{H_2O} . The red solid line represents the reaction rate kinetics model fit¹⁵ for $a_{H_2O} = 0.25$. (b) Plot showing Γ_{FT} decrease with a_{H_2O} (blue) and temperature (green).

Figure 3. (a) Survey XPS spectra from silica fracture surfaces show Si and O peaks with no traces of Cu, while Cu fracture surfaces exhibit prominent Cu peaks (top). (b) Multiplex scans on Cu fracture surfaces show Si 2p signatures of silyl alkyl moieties (~ 102.5 eV) in the organosilane, in contrast to the silica band (~ 103.5 eV) on silica fracture surfaces (bottom left). This result, and the exclusive presence of S 2p peak on the Cu fracture surface (bottom right), indicate that delamination occurs via siloxane bond fissure at the NML-silica interface (top left—where filled green and red spheres denote Si and S, respectively, in the MPTMS NML). The essentially identical spectral features for $0.05 \leq a_{H_2O} < 0.8$ (bottom left) indicate that the delamination path is unchanged. At $a_{H_2O} \sim 0.8$ —the highest water activity in our experiments— Cu fracture surfaces exhibit a silica sub-band along with the prominent silyl alkyl signature, suggesting some fracture path deviation due to bond breaking in the silica substrate.

Figure 4. (a) Γ_{FT} as a function of a_{H_2O} at 293 K. Inset schematically illustrates γ_a and γ_p ; plasticity is depicted as wiggles in the delaminated Cu film. (b) Cu yield stress σ_y extracted from Γ_{FT} - $\log a_{H_2O}$ plots (red) compared with that obtained by nanoindentation (blue), molecular dynamics simulations of nanograined bulk Cu²⁴ and extrapolation of an analytical model developed from nanoindentation of thicker films²⁵. (c) The Γ_{FT} increase with Cu film thicknesses at constant a_{H_2O} where $\gamma_p \neq 0$.

Figure 5. (a) Plastic energy γ_p in copper plotted as a function of the interfacial work of adhesion γ_a , at $T=293$ K. (b) Deformation mechanism map created from Γ_{FT} measurements on structures with Cu/NML/SiO₂ interfaces with a 100 nm-thick Cu film for $278 \text{ K} \leq T \leq 358 \text{ K}$ and $0.6 \leq a_{H_2O} \leq 0.1$. In yellow/orange/red regions γ_p is a major contributor, while green and blue regions represent conditions where $\Gamma_{FT} \sim \gamma_a$.

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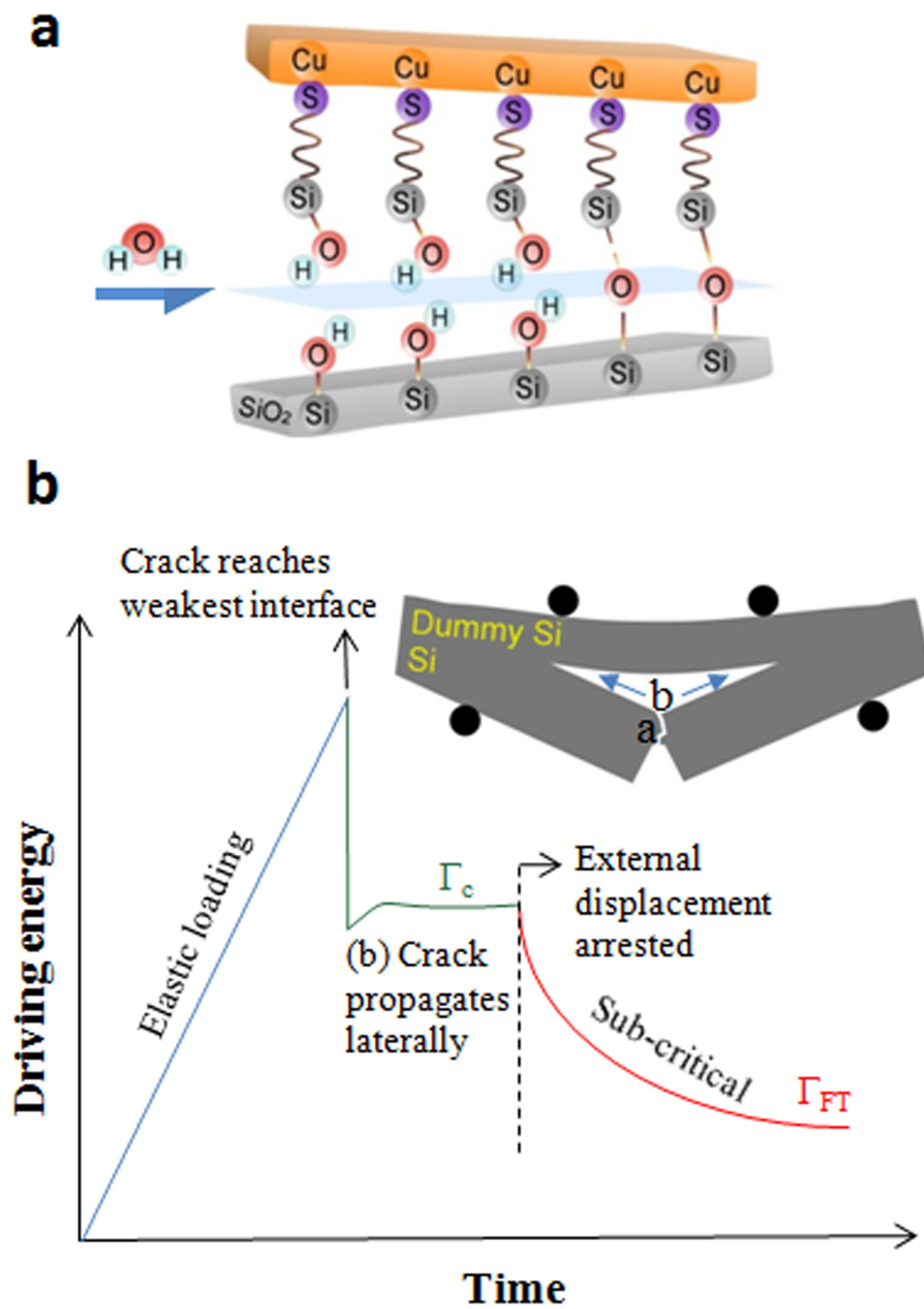


Figure 1 LN12213B 10NOV2010

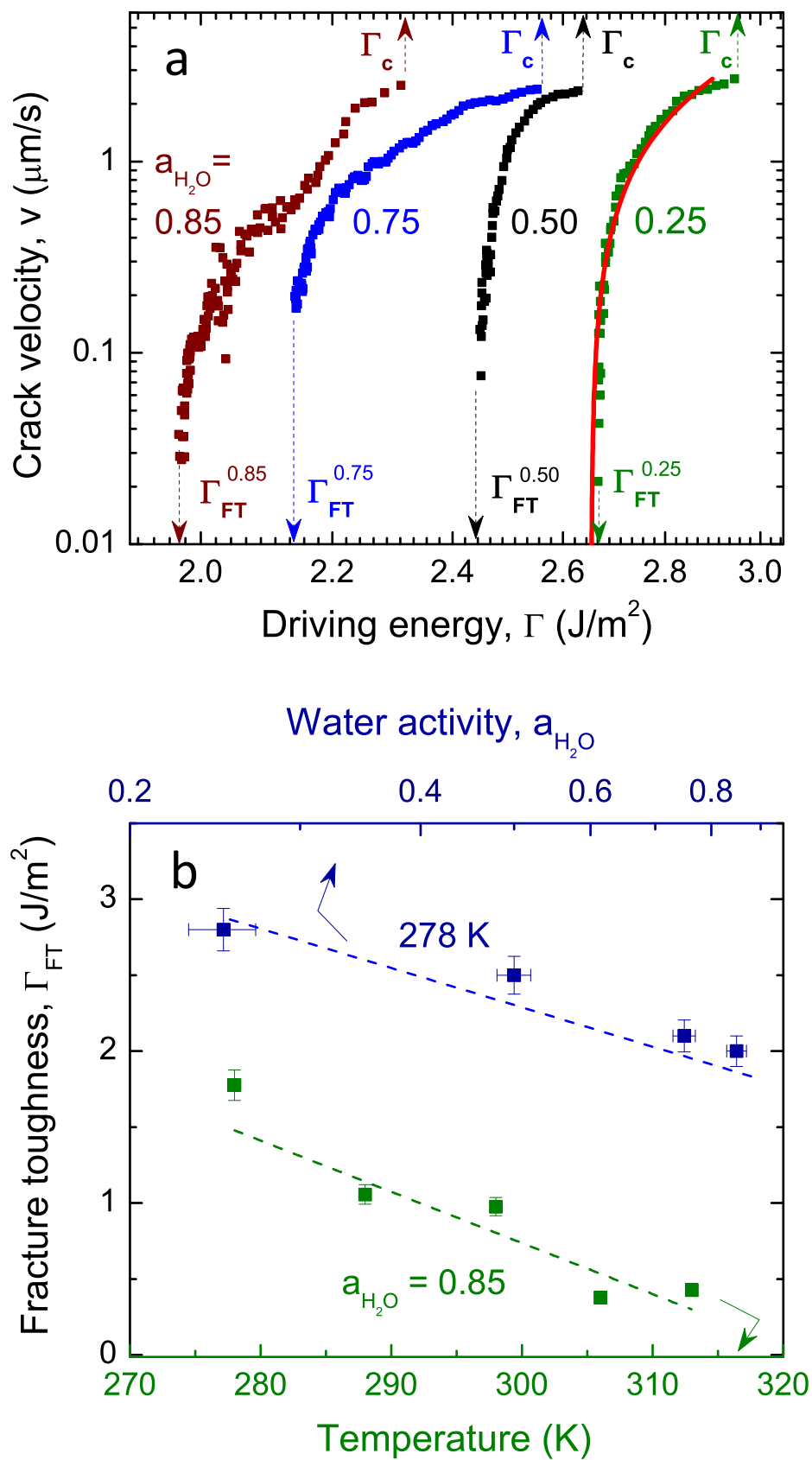
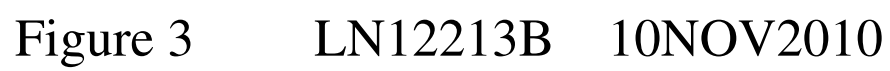


Figure 2

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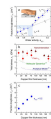


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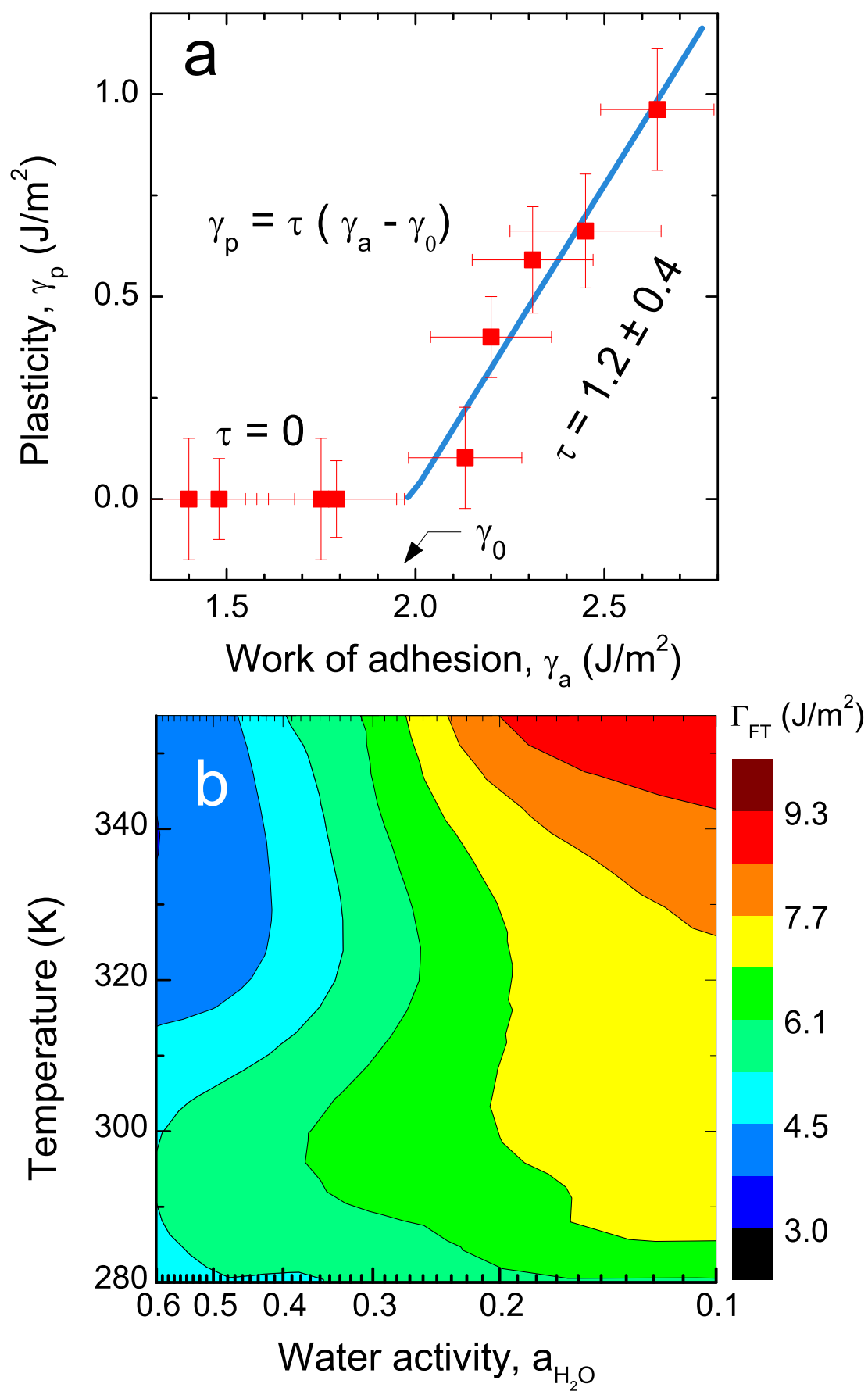


Figure 5

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