CYCLIC FATIGUE-CRACK PROPAGATION ALONG CERAMIC/METAL INTERFACES

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Abstract—The integrity of ceramic/metal joints is investigated under mechanically applied cyclic stresses using double-cantilever-beam, and compact-tension, sandwich test specimens. Specifically, fatigue-crack propagation rates for interfacial cracks are characterized over a range of velocities from 10^{-9} to 10^{-4} m/s for glass/copper and alumina/aluminum-alloy interfaces tested in moist air. Compared to corresponding (stress-corrosion) results under sustained loading, it is found that true interfacial cracks in glass-copper joints are significantly accelerated under cyclic loads. In addition, crack extension force (G) thresholds for interfacial crack growth under cyclic loads are some 46% lower than under sustained loads and are typically over six times lower than the interface; under monotonic loading cracking progresses near the interface in the ceramic layer whereas under cyclic loading failure may occur either in the ceramic or in the metal. Based on a comparison with fatigue-crack growth rates in the metal are much lower than those of the bulk ceramic and show a far higher dependency on the range of G than behavior in the bulk metal.

Résumé—L'intégrité des jonctions métal/céramique a été étudiée sous contraintes cycliques appliquées de façon mécanique, en utilisant une poutre en double console et des éprouvettes sandwich compression-tension. Spécifiquement, les vitesses de propagation des fissures interfaciales, sont caractérisées dans une gamme de vitesses comprises entre 10^{-5} et 10^{-4} m/s pour des interfaces verre/cuivre et alumine/alliage d'aluminium soumises à un essai en atmosphère humide. Par comparaison avec les résultats correspondants (de corrosion sous contrainte) sous charge prolongée, on trouve que les fissures interfaciales véritables dans les jonctions verre/cuivre sont nettement accélérées sous charge cyclique. En outre, le seuil des forces (G) d'allongement des fissures pour la croissance des fissures interfaciales sous charge cyclique est inférieur d'environ 46% au seuil obtenu sous charge prolongée, et il est typiquement six fois plus faible que la dureté de l'interface (G_c). Au contraire, pour le système alumine/alliage d'aluminium, la rupture n'a jamais lieu au niveau de l'interface; sous charge monotone; la fissuration progresse au voisinage de l'interface dans la couche de céramique, alors que sous charge cyclique, il peut y avoir rupture soit dans la céramique soit dans le métal. En comparant les résultats concernant la croissance des fissures de fatigue dans de l'alumine massive d'une part et dans des alliages d'aluminium massifs de l'autre, on trouve que les vitesses de croissance des fissures proches de l'interface, dans le métal, sont bien inférieures à celles qu'on mesure dans la céramique massive, et qu'elles dépendent bien plus de la gamme de G que dans le cas du métal massif.

Zusammenfassung—Die Beständigkeit von Keramik-Metall-Verbindungen wird unter äußerer zyklischer Spannung im Umlauf-Biegeversuch und im Zugversuch an Sandwich-Proben untersucht. Insbesondere werden Ausbreitungsraten der Ermüdungsrisse an der Glas/Kupfer- und Aluminiumoxid/ Aluminiumlegierungs-Grenzfläche im Bereich zwischen 10^{-9} bis 10^{-4} m/s in feuchter Luft gemessen. Im Vergleich zu entsprechenden (Spannungs-Korrosions-) Ergebnissen unter konstanter Spannung ergibt sich, daß die echten Grenzflächenrisse in den Glas-Kupfer-Grenzflächen unter zyklischer Spannung beträchtlich schneller sind. Außerdem ist die Schwelle der Rißausbreitungskraft *G* für Wachstum der Grenzflächenrisse unter zyklischer Belastung etwa 46% niedriger als unter konstanter Last und typischwerweise sechsmal geringer als die Grenzflächenzähigkeit *G*_c. Dagegen tritt bei dem System Aluminiumoxid/Aluminiumlegierung Bruch niemals in der Grenzfläche auf; unter monotoner Belastung schreitet die Rißbildung in der Nähe der GRenzfläche fort, dagegen unter zyklischer Belastung entweder in der Keramik oder in dem Metall. Aus einem Vergleich mit dem Rißverhalten im Volumen von Aluminiumoxid oder der Legierung folgt, daß die Wachstumsraten der Risse in der Nähe der Grenzfläche im Metall viel geringer sind als in Keramikvolumen und weitaus stärker abhängen von dem Bereich von *G* als dem Verhalten im Metallvolumen.

1. INTRODUCTION

The problem of the bonding of dissimilar materials is important in many applications, including the decohesion of films and coatings, the mechanical properties of composite materials, and in the strength of bonds and joints, particularly in the microelectronics packaging industry [1-4]. For the joining of ceramics, bonds having strengths well in excess of the strength of the ceramic can be achieved using a ductile metal; the strength or fracture energy of the resulting

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ceramic/metal bond is then a function of such variables as the thermal and elastic mismatch across the interface, the relative metal layer thickness, plastic flow in the metal, the morphology of the interface, and the presence of defects or interfacial compounds at or near the interface [5].

Whereas much research in recent years has concentrated on understanding the fracture toughness (or fracture energy) of ceramic/metal interfaces [5-13], few studies have focused on the time dependent failure (i.e. stress corrosion or fatigue) of such interfaces [14]; in fact, we know of no studies on the strength of the interface under cyclic (fatigue) loading. This may be deemed to be important as the loss of an electronic device from the failure of a joint, for example, occurs more often at stresses far below those required for catastrophic failure by such mechanisms as thermal fatigue or environmentally assisted cracking.

Accordingly, the prime objective of the current study is to examine the structural integrity of ceramic/metal interfaces under cyclic loading by characterizing the growth behavior of interfacial fatigue cracks in model glass/copper and alumina/ Al-4 wt% Mg systems. Results are compared with corresponding stress-corrosion crack velocities for interfacial cracks, and with fatigue-crack growth behavior for the bulk metal and ceramic materials.

2. EXPERIMENTAL

2.1. Preparation of samples

Two ceramic/metal interface systems were studied, namely glass/copper and alumina/aluminum-alloy interfaces, using sandwich-type fracture mechanics specimens. Crack-growth studies on glass/copper interfaces were performed on double-cantilever-beam $[DB(M_z)$ or DCB] specimens, consisting of a thin copper film bonded between two glass substrates, with aluminum grips attached by epoxy to permit loading [Fig. 1(a)]. Samples were made by evaporating copper films into two soda-lime glass substrates



Fig. 1. Geometry of (a) glass/copper double-cantilever-beam (DCB) specimens, and (b) alumina/ aluminum-alloy compact tension C(T) specimens used for interfacial fatigue-crack growth tests. Crack monitoring on the C(T) specimens was achieved by measuring the electrical-potential changes in a 0.1 μ m metal film evaporated onto the side of the sample.

(of size $75 \times 25 \times 2$ mm), and pressure diffusion bonding the two layers together for 2.5 h at 9 MPa and 450°C. To confine crack advance to one interface, a 10-20 nm layer of chromium was evaporated onto one substrate prior to depositing the copper film. The total thickness of the copper film was $1.5 \,\mu$ m; the ratio of thicknesses of ceramic to metal, h_1/h_2 , thus exceeded 1000.

Studies on alumina/aluminum-alloy interfaces were performed using compact tension C(T) specimens [Fig. 1(b)]. The Al/Al₂O₃ bonds were formed by direct liquid-state bonding of a 220 μ m layer of an Al-4 wt% Mg alloy (prepared from 99.99% Al and 99.95% Mg) between two substrates of 99.5% purity Al₂O₃ [12]. The Al alloy was cold rolled into sheets and placed between the two diamond ground and polished alumina substrates in an alumina crucible, and heated above the melting point of the metal in a resistance furnace. To facilitate bonding, a low oxygen partial pressure was maintained by passing argon gettered with TiZr chips at 850°C through the furnace, and by placing TiZr chips within the furnace enclosure. Compact tension C(T) specimens were machined from the bonded blocks to the dimensions $25 \times 24 \times 3$ mm, and subsequently polished on both faces. The ratio of thickness of ceramic to metal, h_1/h_2 , was of the order of 50.

To improve the interfacial crack-growth resistance under cyclic loading, glass/copper interfaces were additionally tested following implantation of patterned arrays of interfacial steps and "microcracks/ voids" (typically 20 μ m long) within the metal layer. This technique involves the use of photolithography to etch channels on the surface of one glass substrate prior to evaporating the copper layer. The resulting crack bridging by bulged segments of the metal film has been shown to improve the toughness of glass/Cu bonds by as much as a factor of 80 (Fig. 2) [13, 15].

2.2. Test methods

Cyclic fatigue-crack growth tests for both specimen geometries were performed on computer-controlled high-resolution electro-servohydraulic testing machines, operating at a cyclic frequency of 50 Hz (sine wave) with a load ratio (ratio of minimum to maximum tensile loads) of R = 0.1; the testing environment was controlled room air at 22°C with 45% relative humidity. For comparison, subcritical crackgrowth rates were also measured under monotonic loading; resulting stress-corrosion (static fatigue) crack velocities (da/dt) were characterized in terms of interface fracture mechanics [16, 17], as described for cyclic fatigue below.

With DCB samples, crack monitoring was achieved optically; the position of the crack was associated with interference fringes seen through the glass arm. However, in compact specimens, $\sim 0.1 \, \mu m$ thick NiCr films were evaporated onto the specimen



Fig. 2. Interfacial crack extension is illustrated by scanning electron micrographs of cross-sections revealing the crack path along a patterned glass/copper interface: (top) implanted microcrack within the copper layer prior to fracture, (center) partial development of a crack bridging bulge via lateral growth of microcrack, and (bottom) fully developed bulge after separation from the glass. Arrow indicates direction of crack growth.

surface; electrical-potential measurements of the resistance change across the film permitted in situ resolution of crack lengths to within $\pm 2 \mu m$ [Fig. 1(b)] [18]. To electrically insulate the NiCr film from the Al-sandwich layer, a ~0.4 μm thick layer of alumina was sputtered onto the specimen surface prior to film deposition.

Cyclic fatigue-crack propagation rates (da/dN)were determined over a wide spectrum of velocities from ~10⁻¹¹ to 10⁻⁷ m/cycle using test methods that were in general accordance with the current ASTM Standard E647-86A [19] for measuring fatigue-crack growth rates in metals, modified by the procedures outlined for brittle ceramics by Dauskardt and Ritchie [18]. Cyclic fatigue thresholds $\Delta K_{\rm TH}$ and $\Delta G_{\rm TH}$, below which crack growth is presumed dormant, were defined at a driving force where growth rates did not exceed 10⁻¹⁰ m/cycle, consistent with ASTM E647 guidelines.

Following completion of the fatigue tests, the fracture toughness (K_c and G_c) was determined under displacement control from resistance-curve (R-curve) behavior, using the prior fatigue crack as a pre-crack; procedures essentially conform to ASTM Standard E 399-87 [19] for measuring the plane-strain fracture toughness of metallic materials.

2.3. Stress-intensity and strain-energy release rate a calculations

Subcritical crack-growth rates were characterized in terms of linear-elastic fracture mechanics, specifically using the applied far-field stress-intensity factor K and crack-extension force, or strain-energy release rate G. Assuming first that the samples are comprised of homogeneous material, the strainenergy release rate for a symmetrical DCB specimen can be computed from beam theory provided that the ratio of crack length a to beam half-height h_1 is large, i.e. ignoring shear effects, the strain energy stored within each beam is given by [20]

$$U = \frac{2P^2 a^3}{E_1' h_1^3 B}$$
(1)

where E'_1 is Young's modulus E_1 in plane stress or $E_1/(1 - v_1^2)$ in plane strain, and v_1 is Poisson's ratio, where the subscript 1 refers to the ceramic beams; *B* is the beam thickness, and *P* the applied tensile load. For the current specimen geometry, the strain-energy release rate with crack extension thus becomes [21, 22]

$$G = \frac{12P^2 a^2}{E_1' h_1^3 B B_n} \left[1 + \frac{2h_1}{3a} \right]^2$$
(2)

where B_n is the crack-front thickness. (A plane-strain condition is assumed here because the plate geometry and grips preclude antielastic curvature of the beams.)

In general, for a homogeneous sample, the strain energy released rate G is related to the applied Mode I stress intensity K_{I} by

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$$G = K_1^2 / E_1'. (3)$$

For a C(T) sample, K_1 is given by handbook solutions in terms of the crack-length to specimen-width ratio a/W as [23]

$$K_{\rm I} = P/BW^{1/2} \cdot f(a/W) \tag{4}$$

where

$$f(a/W) = \{ [2 + (a/W)] [0.886 + 4.64(a/W) - 13.32 (a/W)^2 + 14.72 (a/W)^3 - 5.6(a/W)^4] \} / [1 - (a/W)]^{3/2}.$$
 (5)

However, although cracks in both geometries propagate under nominally pure applied Mode I loading conditions, the differences in elastic properties across the interface of the two dissimilar materials additionally induces a shear component [24]. A measure of the elastic moduli mismatch is obtained from the two parameters of Dundurs [25], which for plane strain are given by

$$\alpha = (E'_1 - E'_2)/(E'_1 + E'_2)$$
 (6a)

and

$$\beta = \frac{1}{2} \{ [\mu_1 (1 - 2\nu_2) - \mu_2 (1 - 2\nu_1)] \\ \times / [\mu_1 (1 - \nu_2) + \mu_2 (1 - \nu_1)] \}$$
(6b)

where μ is the shear modulus, and the subscripts 1 and 2 refer to materials 1 and 2 across the interface. The values of α and β both vanish when the elastic mismatch between the materials does.

At the crack tip, the Mode I and II components of the interface stress intensity K may be conveniently expressed in complex form $(i = \sqrt{-1})$

$$\mathbf{K} = K_1 + iK_2 = |K| e^{i\psi} l^{-i\epsilon}$$
(7)

where the phase angle, $\psi = \tan^{-1}[Im(Kl^k)/Re(Kl^k)]$ is a measure of the mixity of shear to opening experienced at distance *l* ahead of the crack tip. Using recently available interface fracture mechanics [16, 17], the extent to which the interface stress intensity factors, K_1 and K_2 , vary in proportion (or phase) to the applied far-field values, K_1 and K_{II} , calculated assuming a homogeneous specimen, depends upon the elastic mismatch and the layer thickness h_2 . Provided h_2 is smaller than *a* and all other in plane dimensions [26]

$$K_{1} + iK_{2} = \sqrt{\frac{1-\alpha}{1-\beta^{2}}} (K_{1} + iK_{11})h_{2}^{-i\epsilon} e^{i\omega(\alpha,\beta)}$$
(8)

where the oscillation index ϵ is given by

$$\epsilon = \frac{1}{2\pi} \ln\left[\frac{1-\beta}{1+\beta}\right] \tag{9}$$

and the "phase-shift" function $\omega(\alpha, \beta)$ has been tabulated elsewhere [26]. For many material couples, β , and hence ϵ , may be taken to be zero resulting in the simplified relationship between the phases of the applied and interface stress intensity factors

$$\tan^{-1}(K_2/K_1) = \tan^{-1}(K_{II}/K_I) + \omega.$$
 (10)

The strain-energy release rate per unit length of extension of the crack can be related to the interface stress intensities by [16, 27]

$$G = \left[\frac{1/E_1' + 1/E_2'}{2\cosh^2 \pi \epsilon}\right] [K_1^2 + K_2^2]$$
(11)

or using equation (8) in terms of the far-field stress intensities by

$$G = \left[\frac{1/E_1' + 1/E_2'}{2\cosh^2 \pi \epsilon}\right] \left[\frac{1-\alpha}{1-\beta^2}\right] [K_1^2 + K_{11}^2]$$
(12)

which is equivalent to

$$G \approx \frac{[K_1^2 + K_{11}^2]}{E_1^1}$$
(13)

For the glass/copper DCB samples, where the h_1/h_2 ratio is very large and $\alpha = -0.33$, $\beta = -0.15$ and $\epsilon = 0.05$, the angle of rotation is of the order of 2°; the shear component to the stress-intensity factor is therefore less than 4% of the Mode I component at a distance $l = h_2$ ahead of the crack. The

coefficients of the applied stress intensity term $[K_1^2 + K_{II}^2]$ in equation (12) for an interface crack differ insignificantly from the $1/E'_1$ coefficient for a homogeneous specimen. Conversely, for the alumina/aluminum-alloy C(T) samples, h_1/h_2 is much smaller; moreover, since the Dundurs parameters are now $\alpha = 0.68$ and $\beta = 0.14$, $\epsilon = -0.05$ such that the angle of rotation is -9° at $l = h_2$. Thus, although the applied loading is purely tensile, a significant shear component is developed in the interface (K_2 is almost 20% of K_1).

3. RESULTS

3.1. Glass/copper system

3.1.1. Plain interfaces. Results of the cyclic fatiguecrack growth tests on plain glass/copper interfaces in moist air are plotted in Fig. 3 in terms of the crack velocities (v, da/dt) as a function of the strain-energy release rate G_{max} and stress intensity K_{max} ; results are compared with those for stress-corrosion cracking (static fatigue) along the same glass/copper interfaces under monotonic loading [14]. Growth rates span 5 orders of magnitude from $\sim 10^{-9}$ to 10^{-4} m/s, and represent the first cyclic fatigue v/G data reported for ceramic/metal interfaces. It is apparent that cyclic crack-growth rates are many orders of magnitude faster than stress-corrosion crack-growth rates at equivalent G levels under monotonic loading. Moreover, compared to the interfacial fracture energy (or toughness) for "catastrophic" failure along the glass/copper bond of $G_c \approx 2 \text{ J/m}^2$, the existence of subcritical cracking, either induced by a moist environment or the presence of cyclic loading, can significantly lower the energy required for cracking of the interface. Whereas stress-corrosion thresholds reduce the effective fracture energy by almost a factor of 4, as reported previously [13, 14], the presence of cyclic loading results in a threshold G_{TH} of approximately 0.3 J/m^2 , over six times smaller than G_c . It should be noted here that such failures of glass/ copper interfaces are true interfacial fracture; in situ Auger spectroscopy of the resulting fracture surfaces reveal only glass on one side and metal on the other (Fig. 4).

3.1.2. Patterned interfaces. Corresponding cyclic fatigue-crack propagation data for patterned glass/copper interfaces, containing $\sim 20 \,\mu\text{m}$ micro-cracks implanted within the metal layer, are shown in Fig. 5; again results are compared with those for stress-corrosion cracking (static fatigue) along the same interfaces under monotonic loading [15]. As with the plain interfaces, the presence of cyclic loading leads to a significant increase in subcritical crack



Fig. 3. Comparison of experimental crack velocity vs G data for plain glass/copper interfaces subjected to monotonic (stress-corrosion) and cyclic (fatigue) loading. Results obtained on sandwich DCB samples tested at R = 0.10 (50 Hz) in moist room air.

velocities and to a marked reduction in the fracture energy to cause interface cracking. However, compared to the plain interfaces (Fig. 3), the presence of the implanted microcracks leads to much improved crack-growth resistance; v/G curves under both monotonic and cyclic loading are shifted by approximately a factor of ~10 to higher G values, representing many orders of magnitude reduction in growth rates.

Fractographically, crack bridging of the interfacial crack by bulged metal-film segments provides a significant contribution to the improved subcritical crack-growth resistance under monotonic loading (Fig. 6); the extent of such bridging is limited by interface failure (see Fig. 2 for an example). Under cyclic loads, such bridging still prevails but it is now limited by failure of the metal-film ligaments themselves. Failure involves a combination of cracking or tearing of the bridging metal films plus interface fracture between the bridging ligaments. However, the amount of deformation of the ligaments prior to fatigue failure diminishes as the cyclic G range, ΔG , is reduced, as can be seen by comparing Figs 7 and 8.

3.2. Alumina/aluminum-alloy system

A representative scanning electron micrograph (SEM) of the as-polished, uncracked interface formed between the alumina and the Al-Mg alloy is shown in Fig. 9(a). Subsequent crack propagation, initially under cyclic fatigue and then monotonic loading, revealed that cracks propagate through regions close to the interface [Fig. 9(b)], differing from the glass/Cu system in which crack propagation was purely inter-

[†]This fracture energy is representative of the tensile failure of glass/copper interfaces pressure diffusion bonded at 450°C. However, G_c values as high as 10 J/m² have been measured for bonding temperatures up to 650°C [15].



Fig. 4. In situ Auger spectroscopy data on the chemical composition of the two fracture surfaces generated by interfacial fracture of plain glass/copper interfaces, showing *true* interfacial failure.

facial. During initial propagation in the metal, cyclic fatigue cracks were generally observed to continue growing in the metal layer in close proximity to one of the interfaces [Fig. 9(c)]. Under subsequent monotonic loading, crack growth occurred in the ceramic often after crack reinitiation on the opposite side of the sandwiched layer. A number of possible crack path morphologies have been observed and are schematically illustrated in Fig. 9(d).

The metallic fatigue fracture surface is fibrous in appearance and shows evidence of fatigue striations [Fig. 10(a)]; its proximity to the interface was indicated by the observation of regions of exposed ceramic surface. The fracture surface corresponding to monotonic loading is shown in Fig. 10(b) revealing primarily intergranular failure of the ceramic [Fig. 10(c)]. The proximity of the monotonically loaded crack to the interface is indicated by regions of ductile dimpling formed by microvoid coalescence in the metal [Fig. 10(d)].

The cyclic fatigue results were obtained by initiating crack growth in the metal and then reducing the loads such that cracking proceeded at a decreasing rate until the fatigue-crack growth threshold was achieved. Results of the cyclic fatigue-crack growth tests for cracks located in the metal layer of the alumina/Al-Mg interface are illustrated in Fig. 11 in terms of crack-growth rates (da/dN) as a function of the fracture energy range, ΔG ; for comparison results are also shown for a bulk alumina ceramic [28] and bulk aluminum alloys [29]. Interfacial growth rates are much lower than those in the ceramic, yet similar to a range of data obtained for bulk metal. Only very limited straining of the metal layer was apparent during fatigue loading. Subsequent monotonic loading to determine R-curve behavior is accompanied by extensive plastic deformation resulting in significant crack-tip blunting. The effect of blunting is to diminish the crack-tip stresses to a small multiple of the yield stress of the metal. Corresponding metal slip lines radiate out from the blunted crack tip across the metal film; this is apparent in the optical micrograph in Fig. 12, which shows the fracturing of a thin NiCr layer on the specimen surface with microcracks in



Fig. 5. Experimental crack velocity vs G data for glass/copper interfaces implanted with arrays of "voids/microcracks", showing a comparison of results subjected to monotonic (stress-corrosion) and cyclic (fatigue) loading. Note how interfacial crack-growth resistance is enhanced compared to plain glass/Cu interfaces (cf. Fig. 3). Results obtained on sandwich DCB samples tested at R = 0.10 (50 Hz) in moist room air.

response to the strain field. The crack was often observed to renucleate *in the ceramic* adjacent to the opposite interface. The resulting R-curve is shown in Fig. 13. The toughness of the interface approaches





Monotonic Loading

Fig. 6. Scanning electron micrograph of the metal side of the fracture surface for cyclic fatigue of a plain glass/copper interface, showing schematic diagram of the principal operating crack-tip shielding mechanism of crack bridging by the bulged metal-film segments. Arrow indicates direction of crack growth.







Cyclic Loading

low AK levels

Fig. 7. Scanning electron micrographs of (a) the ceramic side and (b) the metal side of the fracture surface for cyclic fatigue at low G levels ($<5 \text{ J/m}^2$) of a patterned glass/copper interface, showing schematic diagram of the fatigue cracking of the bulged metal-film segments, which limits the extent of crack bridging. Arrow indicates direction of crack growth.

400 J/m², well in excess of the toughness of the ceramic $\sim 20-100 \text{ J/m}^2$), yet small compared to the alloy toughness, $\sim 10^4 \text{ J/m}^2$.

4. DISCUSSION

The two types of ceramic/metal interfaces are quite different in terms of the dimensions used for the sandwich samples and the physical and chemical properties of the interface and the adjoining materials. Important characteristics are summarized in Table 1. For both material combinations the metal has a much higher thermal expansion coefficient than does the ceramic which leads to in-plane residual tensile stresses in the metal after cooling from the diffusion bonding temperature. In both instances,



Cyclic Loading

high **AK** levels

Fig. 8. Scanning electron micrographs of (a) the ceramic side and (b) the metal side of the fracture surface for cyclic fatigue at high G levels ($\sim 6.5 \text{ J/m}^2$) of a patterned glass/copper interface, showing schematic diagram of the tearing of the bulged metal-film segments, which limits the extent of crack bridging. Arrow indicates direction of crack growth.

the thermal strains exceeded the yield point and the metal is at approximately the yield stress. This has been measured by X-ray diffraction for the Cu/glass samples and has been estimated from microhardness measurements for the Al/alumina samples. The yield stress of the Cu (320 MPa) is higher than expected for bulk Cu owing to the fine grain microstructure (grain size about 1 μ m), a high density of twins, and perhaps impurities from evaporation; the Al has a coarser microstructure (grain size about 55 μ m) and a yield stress which is influenced by the Mg alloying addition.

The plain Cu/glass samples used here invariably failed at the ceramic/metal interface under monotonic loading at a relatively low value of G_c , 2 J/m²; this low value in part results from organic and other impurities at the interface. In contrast, the

Al/alumina samples virtually never failed exactly at the interface; the crack location depends upon the loading mode, but was generally within the ceramic after sufficient monotonic loading, but depended upon the loading history and other details. The much higher G_c values for these materials suggest the better interfacial chemical bonding, which can be inferred from the lack of interfacial fracture, but also from the greater plastic deformation resulting from the better interfacial bonding, the lower yield stress for the Al and from the thicker metal film used. To permit rough comparisons, the approximate width of the crack-tip plastic zones (Table 1) have been calculated using a simple continuum expression for homogeneous materials

$$r_{\rm y} \approx \frac{G_{\rm c} E_1' E_2'}{\pi (E_1' + E_2') \sigma_{\rm y}^2}$$
 (14)

where $\sigma_{\rm v}$ is the yield stress (estimated as discussed above) and G_c is for monotonic loading which gives interfacial failure for the Cu/glass and failure in the ceramic several microns from the interface for the Al/Al_2O_3 . Clearly, the plastic zone is comparatively much smaller for the Cu/glass samples; however, even in these samples, it has been estimated that the plastic deformation constitutes most of the fracture energy. For the Al/alumina samples, the extensive plastic deformation is evident from optical or scanning electron microscopy of the previously polished side of the metal. For the Cu/glass samples toughened by microcrack-induced bridging, the plastic deformation is much greater than for the plain Cu/glass samples, but the deformation is in the bridging ligaments of Cu and is not simply that anticipated from the plastic deformation around a crack tip under plane strain.

Although the results are only preliminary, a few notable trends emerge already. For the Cu/glass samples, which exhibit the smallest amount of plasticity, the crack-growth rates under cyclic loading significantly exceeded those found under monotonic loading at $G = G_{max}$ for the cyclic loading. [To expedite such comparisons, crack-growth rates for cyclic loading in Figs 3 and 5 are plotted as average velocity versus G_{max} which is $\Delta G/(1-R)$ with R = 0.1 for these tests.] Although the crack-tip geometry (i.e. the degree of blunting) has not been adequately characterized for interfacial failure under monotonic loading in such couples, it may be tempting to suppose the cyclic failure entails alternate sharpening and blunting leading to plastic deformation induced extension per cycle for these flat interfaces.

For the toughened Cu/glass samples and for the Al/alumina samples, the situation is more complex in that the fracture location is changed by switching from cyclic for monotonic loading. For the microcrack toughened Cu/glass, prior work [30] has shown that the failure location is dictated by the parameter t_f/λ , where t_f is the thickness of the bridging metal film and λ is the periodic spacing of the microcracks; for large values of this ratio, failure tends to be purely



Fig. 9. Scanning electron micrographs of crack-path profiles of interfacial fatigue cracks in Al₂O₃/Al-Mg, showing (a) the as-polished uncracked interface, (b) cyclic fatigue-crack growth in the metal within a few microns of the interface, (c) the transition in crack path at the initiation of cracking on the R-curve from a fatigue pre-crack, and (d) schematic diagram of typical crack morphologies. Bold arrow indicates direction of crack growth.

at the interface whereas for small values failure entails alternate tearing of the bridging metal films and interfacial failure between the bulging bridges. For the sample tested here, monotonic loading led to entirely interfacial failure whereas under cyclic loading failure switched to involve failure of the metallic bridges. Moreover, the extent of ductile deformation in the metal prior to failure was reduced by inducing failure by cyclic fatigue at a lower ΔG level.

The situation is analogous but even more complex in the Al/alumina couples. Again cyclic loading switches the bias toward failure in the metal; however, here under monotonic loading the failure switches to fracture in the ceramic rather than at the ceramic/metal interface. For the several tests conducted to date, the C(T) samples were cut with a chevron pre-crack (notch); initial loading was under cyclic conditions in order to encourage formation of a sharper crack. In all instances, the fatigue crack runs near the interface, but $10-20 \,\mu m$ into the metal. Upon switching to monotonic loading the situation is unexpectedly complex. Eventually the crack tends to run in the ceramic; the exact location may be dictated by the crack seeking a path giving $K_{II} = 0$, which would be dictated by the far-field geometry which tends to drive the crack out from the center of the sample, and perturbations in the stress field owing to the compliance difference between the more com-

geometry; moreover, the cracks meander and branch enough that a careful characterization of the spacing is not possible for this relatively coarse-grained alumina. However, the transitions between cyclic and monotonic loading are surprisingly varied. At the

pliant metal film and the ceramic. However, compu-

tations of this path have not been done for the present

monotonic loading are surprisingly varied. At the conversion to monotonic loading, the fatigue crack blunts, and sometimes the extension involves propagation into the ceramic on the same side of the metal as the fatigue crack. In other instances, damage develops at the other interface which leads to formation of the crack at the opposite interface, which may extend. However, in the example shown (Fig. 12), after some extension the crack switched back to the original side of the sample. The extensive deformation of the metal associated with this transition led to development of very high loads (a steeply rising and then falling apparent R-curve) prior to development of an unbridged crack on one side of the metal film (Fig. 13). It is not entirely clear what leads to the tendency of the crack on both sides of the metal film. However, it is reminiscent of the bulging bridges observed in tests of Cu/glass films of $1 \,\mu m$ thickness (without Cr on either side) [10]. Again in samples with both interfaces nominally similar, and with a cracktip plastic zone of the order of the film thickness, the



Fig. 10. Scanning electron micrographs of the fracture surfaces for interfacial cracking along alumina/ aluminum-alloy interfaces, illustrating (a) cyclic fatigue failure in the aluminum metal (by ductile fatigue striations), (b) low magnification view of the monotonic fracture surface showing intergranular cracking in the ceramic and regions of ductile fracture in the metal, (c) monotonic failure in the Al₂O₃ (by brittle intergranular fracture), and (d) monotonic failure in the metal (by microvoid coalescence). Bold arrow indicates direction of crack growth.



Fig. 11. Experimental cyclic fatigue-crack growth rate (da/dN) vs ΔG data for alumina/aluminum-alloy interfaces, compared corresponding fatigue data for bulk alumina [28] and bulk aluminum alloys [29]. Results obtained on sandwich compact-tension samples tested at R = 0.10 (50 Hz) in moist room air.



Fig. 12. (a) Scanning electron micrograph and (b) schematic representation of the deformation and fracture events during the determination of the R-curve, showing initial blunting of the fatigue pre-crack in the metal layer (within microns of the interface), extensive yielding across the metal, and re-nucleation of the crack in the ceramic on the other interface. Bold arrow indicates direction of crack growth.

samples had a strong tendency to alternately crack on both interfaces. Further investigation and modeling is proceeding under the supposition that blunting leads to stress reduction near the tip of the main crack, and that the stresses on the opposite interface are high enough to trigger damage owing to variabilities in the strength of the interfaces owing to heterogeneous impurities (in the case of the Cu/glass) or the variabilities in the residual interfacial porosity in the case of the Al/alumina samples.

5. CONCLUSIONS

Based on a study of subcritical crack propagation by cyclic fatigue along glass/copper and alumina/ aluminium-alloy interfaces using nominally Mode I

Table 1. Comparative features of ceramic/metal couples exmained

		Cu/glass	Al/Al ₂ O ₃
Young's modulus of metal	E_2 , GPa	130	70
Young's modulus of ceramic	$\overline{E_1}$, GPa	70	400
Ratio of shear moduli	μ_1/μ_2	0.59	6.3
Dundurs parameters	α	-0.33	0.68
-	β	-0.15	0.14
Oscillation index	ε	0.05	-0.05
Phase-shift function	ω	2°	-9°
Metal yield strength	$\sigma_{\rm v}$, MPa	320	150
Metal-layer thickness	$h_2, \mu m$	1.5	200
Interfacial fracture energy ^a	$G_{\rm c}, {\rm J}/{\rm m}^2$	2	100
Plastic-zone size	$r_{\rm v}, \mu {\rm m}$	0.3	95

 ${}^{a}G_{c}$ is for monotonic loading which gives interfacial failure for the Cu/glass and which gives failure in the ceramic several microns from the interface for the Al/Al₂O₃.



Fig. 13. Resistance-curve (R-curve) fracture toughness behavior for alumina/aluminum-alloy sandwich sample, compared to R-curves for coarse- and fine-grained polycrystalline alumina.

double-cantilever-beam and compact-tension sandwich samples, the following specific conclusions can be made:

1. Crack-velocity/strain-energy release rate (v/G) data are presented, for the first time, for cyclic fatigue-crack propagation along ceramic/metal interfaces.

2. For true interfacial failure along (plain) glass/copper interfaces, cyclic crack-growth rates (primarily in Mode I) are found to be many orders of magnitude faster than subcritical (stress-corrosion) crack-growth rates at equivalent G levels under monotonic loads, and to occur above a threshold G level almost 50% lower than under monotonic loading.

3. Compared to the interfacial fracture energy (or fracture toughness) of the glass/copper interface of $G_c \approx 2 \text{ J/m}^2$, the presence of cyclic loading reduces the G level to cause interfacial cracking by over a factor of 6 to $\sim 0.3 \text{ J/m}^2$.

4. By implanting patterned arrays of "microcrack/ voids" within the copper layer by prior photolithography techniques, crack-tip shielding is induced by crack bridging of bulged metal-film segments across the crack, such that cyclic crack-growth rates along glass-copper interfaces are reduced by many orders of magnitude; moreover, threshold $G_{\rm TH}$ levels to cause cracking are increased by approximately an order of magnitude compared to that measured for plain glass/Cu interfaces.

5. For alumina/aluminum-alloy interfaces tested with sandwich compact-tension samples (i.e. loaded in ~80% tension and ~20% shear), cyclic crackgrowth rates are markedly less than those for bulk alumina, yet are on the fast side of the scatterband of results for bulk aluminum alloys. Similarly, the interface fracture toughness G_c approaches 400 J/m², well in excess of the toughness of alumina (~20-100 J/m²), yet small compared to the toughness of Al alloy (~10⁴ J/m²).

6. In the alumina/aluminum-alloy system, crack growth, whether under monotonic or cyclic loads, does not occur in the interface; under monotonic loads, cracks primarily proceed in the ceramic within 50 μ m or so of the interface, whereas under cyclic loads crack growth may occur either in the ceramic or the metal within a few microns of the interface.

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