

CRACK DIRECTION EFFECTS ALONG COPPER/SAPPHIRE INTERFACES

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Abstract—The Rice–Thomson model is used to predict the directional toughness of a copper/sapphire interface, consisting of a $\{221\}$ face of a copper crystal bonded to a basal sapphire surface. Specimens of this type are subsequently tested, in the form of a layered beam (having two opposing cracking directions) subject to four-point bending and confirm the theoretical predictions. A general observation is that the ductile or brittle response of a crack on a metal/ceramic interface is strongly dependent on the relative orientation of available slip planes to the existing crack plane and growth direction. This conclusion is consistent with earlier results by Wang and Anderson on symmetric tilt grain boundaries in copper and explains, for example, why brittle cracking occurs in the $[1\bar{1}4]$ direction along the $\{221\}$ interface of a tensile-loaded copper bicrystal, but along the opposite $[\bar{1}14]$ direction in bending-loaded copper/sapphire specimens with mixed mode conditions at the crack tips.

Résumé—On utilise le modèle de Rice et Thomson pour prévoir la ténacité directionnelle d'une interface cuivre/saphir, formée d'une face $\{221\}$ d'un cristal de cuivre liée au plan basal du saphir. Des échantillons de ce type sont donc testés; ils ont la forme d'une poutre constituée de couches (ayant deux directions de fissuration opposées) soumise à une flexion sur quatre points et les résultats confirment les prévisions théoriques. D'une manière générale, la réponse ductile ou fragile d'une fissure sur une interface métal/céramique dépend fortement de l'orientation relative des plans de glissement disponibles par rapport au plan de la fissure existante et à sa direction de croissance. Cette conclusion est en accord avec les précédents résultats de Wang et Anderson sur les joints de flexion symétriques dans le cuivre et explique, par exemple, pourquoi la fissuration fragile se produit dans la direction $[1\bar{1}4]$ le long de l'interface $\{221\}$ d'un bicristal de cuivre charge en traction, et dans la direction opposée $[\bar{1}14]$ dans des échantillons de cuivre/saphir chargés en flexion avec des conditions de mode mixte aux extrémités des fissures.

Zusammenfassung—Mit dem Rice–Thomson-Modell wird die Zähigkeit einer Kupfer–Saphir-Grenzfläche, die aus einer $\{221\}$ -Kupferfläche und der Saphir-Basisebene besteht, vorausgesagt. Proben in der Form eines geschichteten Balkens, die zwei gegenüberliegende Rißrichtungen aufweisen, werden im Vierpunkt-biegeversuch belastet; die Ergebnisse bestätigen die theoretischen Voraussagen. Eine allgemeine Beobachtung ist, daß das duktile oder spröde Verhalten eines Risses an einer Metall/Keramik-Grenzfläche sehr stark von der relativen Orientierung der verfügbaren Gleitsysteme bezüglich der bestehenden Rißebe und der Ausbreitungsrichtung des Risses abhängt. Diese Folgerung ist mit früheren Ergebnissen von Wang und Anderson zu symmetrischen Kippkorngrenzen in Kupfer verträglich und erklärt zum Beispiel, warum spröde Risse in der $[1\bar{1}4]$ -Richtung entlang der $\{221\}$ -Grenzfläche in einem zugbelasteten Kupfer-Bikristall auftreten, aber entlang der entgegengesetzten Richtung $[\bar{1}14]$ in biegebelasteten Kupfer–Saphir-Proben unter Bedingungen einer gemischten Mode an den Rißspitzen.

INTRODUCTION

The behaviour of metal/ceramic bonds is of great technological importance, for example, in the field of electronic device packaging, and, needless to say, in metal/ceramic composites. In this paper, experimental results from fracture tests of single crystals of copper bonded on a $\{221\}$ plane to the basal plane of sapphire are reported. At issue here is the ductile vs brittle response of the interfacial crack: If conditions are favorable for the nucleation and unstable propagation of a dislocation from an interfacial crack tip, is brittle crack propagation suppressed? An early attempt at answering this question was due to Rice and Thomson [1] in which the ductile versus brittle response of a material is assumed to be ultimately

controlled by the competition between atomic decohesion and dislocation nucleation ahead of the crack tip. Later versions of this model [2–6] extended the Rice–Thomson model to treat the crack response along an interface by evaluating the competition in terms of the parameters G_{cleave} , the energy release rate for interfacial cleavage, and G_{disl} , the energy release rate associated with the emission of a single dislocation. The competition is shown schematically in Fig. 1; if $G_{\text{cleave}} < G_{\text{disl}}$, then the crack is said to propagate in a brittle manner, and the interface is said to be intrinsically brittle; conversely, if $G_{\text{disl}} < G_{\text{cleave}}$, then a dislocation moves away from the tip thus blunting and “shielding” the crack tip from further increases in applied loading, and the interface is said to be intrinsically ductile. The most dramatic

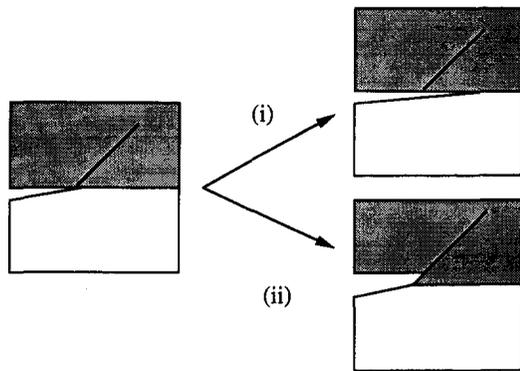


Fig. 1. Atomistically sharp interfacial crack on the left, showing the competition between cleavage decohesion (i) and dislocation emission (ii).

effect related to this competition is the directional dependence of the interfacial crack response in symmetric tilt bicrystals as predicted by the model and experimentally observed [6].

The primary purpose of this paper is to test the idea that the competition between dislocation emission and cleavage decohesion controls the ductile versus brittle behaviour of a metal/ceramic interface and that the crack response is direction-dependent, similar in spirit to the efforts of Wang and Anderson [6]. In their work, symmetric tilt bicrystals of copper with different bismuth segregation levels were tested. The most dramatic effect relating to the directionality occurred in a $\Sigma 9[110]/(\bar{2}\bar{2}1)$ symmetric tilt copper bicrystal, in which two specimens were cut and notched along the grain boundary so that a crack would run in the opposite directions $[\bar{1}14]$ and $[11\bar{4}]$,

respectively. The specimens were fatigued under a mode I loading of increasing amplitude. The specimen with the $[1\bar{1}4]$ cracking direction broke along the interface when the maximum normal stress reached 28.1 MPa, corresponding to $G \approx 28 \text{ J/m}^2$. An intergranular fracture surface with cleavage "tongues" was observed. The other specimen, with a cracking direction of $[\bar{1}14]$, was loaded under identical conditions and eventually fractured at a normal stress of 76.7 MPa. The fracture surface contained large regions of ductile transgranular fatigue fracture and plastic tearing, and the G value, $>210 \text{ J/m}^2$, was beyond the reliably measurable range for elastic fracture mechanics. The only difference between these two specimens was the cracking direction, hence it was concluded that the difference in ease with which dislocations could be nucleated at each crack tip was the cause of this behaviour.

ANALYSIS OF SPECIMEN

The specimens used in this work are the four-point bend specimens consisting of a metallic layer bonded to a thinner ceramic layer (see Fig. 2). This type of specimen was originally proposed by Charalambides *et al.* [7] and Suo and Hutchinson [8] for the testing of linear elastic interfacial fracture mechanics concepts. The usefulness of the specimen for a possible directional effect resulting from the use of an asymmetrically aligned metal single crystal was pointed out in the work of Rice *et al.* [5]. An advantage of this type of specimen is that when the crack is long compared with the thickness of the notched layer h , the complex stress intensity factor

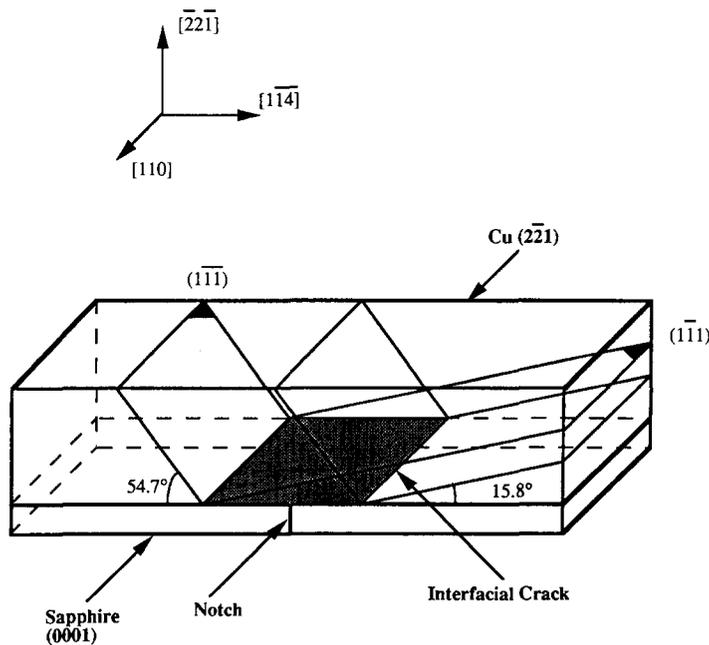


Fig. 2. Schematic of the specimen analyzed: a copper single crystal with its $\{221\}$ face bonded to sapphire; loaded in bending with crack tips parallel to the $[110]$ direction.

has the following form, which is independent of crack length

$$K = Y(\alpha, \beta, h/H) M h^{-3/2-\epsilon} e^{i\psi(\alpha, \beta, h/H)} \quad (1)$$

where M is the applied moment per unit width, and Y and Ψ are dimensionless functions of the thickness ratio and the Dundurs parameters α and β , which are given by

$$\alpha = \frac{(1 - \nu_2)/\mu_2 - (1 - \nu_1)/\mu_1}{(1 - \nu_2)/\mu_2 + (1 - \nu_1)/\mu_1}, \quad (2)$$

$$\beta = \frac{1}{2} \frac{(1 - 2\nu_2)/\mu_2 - (1 - 2\nu_1)/\mu_1}{(1 - \nu_2)/\mu_2 + (1 - \nu_1)/\mu_1}$$

where μ and ν refer to the shear modulus and Poisson's ratio, respectively. Subscript 1 refers to the material on top, which is taken to be the metal, and subscript 2 refers to the ceramic phase. The functions Y and ψ may be found in [8] or [9]. By definition, ψ is the phase angle of the complex quantity Kh^{ϵ} , and the quantity $\tan \psi$ is a relative measure of the ratio of loading modes on length scales of order h from the tip. The parameter ϵ is known as the oscillation index, and is given by

$$\epsilon = \frac{1}{2\pi} \ln \left(\frac{1 - \beta}{1 + \beta} \right). \quad (3)$$

The stress intensity factor K reduces to the familiar $K_I + K_{II}$ when $\epsilon = 0$.

The intent of the four-point bend specimen is to compare crack growth in two opposing directions in the *same* specimen during the *same* mechanical test. The orientation of the copper in these experiments is the same as that in the S9 bicrystals tested by Wang and Anderson: the (221) face of the copper crystal is bonded to the sapphire surface, with the short dimension of the specimen parallel to the [110] direction in the copper. Once a crack runs through the sapphire and branches onto the interface in both directions, the crack fronts at both ends lie along the intersection of a pair of {111} slip planes of the copper crystal with the interface. The crack growth directions are hence $[\bar{1}14]_{Cu}$ and $[1\bar{1}4]_{Cu}$. Due to the asymmetry of the copper crystal, the crack oriented to run in the $[\bar{1}14]_{Cu}$ direction encounters slip planes which are aligned at 15.8° and 125.3° , respectively, while the crack oriented to run in the $[1\bar{1}4]_{Cu}$ direction encounters slip planes inclined at 54.7° and 164.2° (see Fig. 2).

For the study of dislocation emission from the tip of an interfacial crack, a parameter known as the *atomic scale phase angle*, ψ' , has been introduced [5], which is the phase angle of the quantity Kb^{ϵ} , and thus is a measure of the ratio of local mode II to mode I conditions on atomic length scales from the crack tip. It is a simple exercise to show that the two angles are related by

$$\psi' = \psi - \epsilon \ln(h/b) \quad (4)$$

where b represents the magnitude of the Burgers vector. Unlike the typical situation when the local

phase angle is defined based on a "laboratory" length scale, the distinction between ψ and ψ' in this context becomes important because the oscillatory stress field can give rise to a significant shift in the ratio of mode II to mode I type conditions when considering atomic-scale distances from the crack tip. The loading phase angle ψ is about -52° , while the local phase angle, ψ' , is about -79° under four-point bending for the specimens analyzed in this work [5], indicating a significant mode II component of loading. Under these conditions, using a Rice-Thomson model which had been specialized for bimaterial interfaces, Rice *et al.* [5] predicted that G_{disl} is about 0.86 J/cm^2 for cracking in the $[\bar{1}14]_{Cu}$ direction (dislocation emission is easiest on the slip plane inclined at 125.3°) and 4.9 J/m^2 for cracking in the $[1\bar{1}4]_{Cu}$ direction (dislocation emission is easiest on the slip plane inclined at 164.2° in this case). The parameter G_{cleave} , a measure of the coherence of the interface, is assumed to be identical in either direction, hence brittle crack growth should be favored in the $[\bar{1}14]_{Cu}$ direction. It is interesting to point out that in the work of Wang and Anderson on bicrystals, the ductile cracking direction was $[\bar{1}14]$ and the brittle direction was $[1\bar{1}4]$, the *opposite* to that predicted here for the bend specimen. This is attributable entirely to the phase angle effect. Here, the atomic scale phase angle ψ' is -79° , while for a mode I loading of a symmetric tilt bicrystal it is 0° . Figure 3(a) which gives G_{disl} vs θ for copper/sapphire interfaces loaded in such a manner as to give an atomic scale phase angle of -79° , highlights the effect of the slip plane inclination angle.

The phase angle effect is further illuminated by Fig. 3(b), which gives G_{disl} vs ψ' for the various slip plane inclination angles of interest. The dashed lines correspond to angles associated with the $[\bar{1}14]_{Cu}$ direction, and the solid lines correspond to angles associated with the $[1\bar{1}4]_{Cu}$ direction. Comparison of the curves at $\psi' = 0$ and $\psi' = -79^\circ$ shows that the favored direction for dislocation emission reverses when the phase angle is altered.

DISLOCATION NUCLEATION BASED ON A PEIERLS MODEL

The methods for determining G_{disl} as discussed in the previous discussion were based on a modified Rice-Thomson model [5], which treats a dislocation as an elastic singularity which exists ahead of the crack tip prior to loading. An unfortunate consequence of this usage is that a *core cutoff parameter* must be introduced into the analysis, which is usually assumed to be on the order of one atomic spacing. Argon [10] and, more recently Schoeck [11] have recognized that a full dislocation is likely to emerge unstably from an incomplete, incipient dislocation at the tip, but a reasonably exact treatment has been given only recently by Rice [12]. That treatment, discussed in further detail by Beltz and Rice [13, 14], solves the elasticity problem of a traction free crack

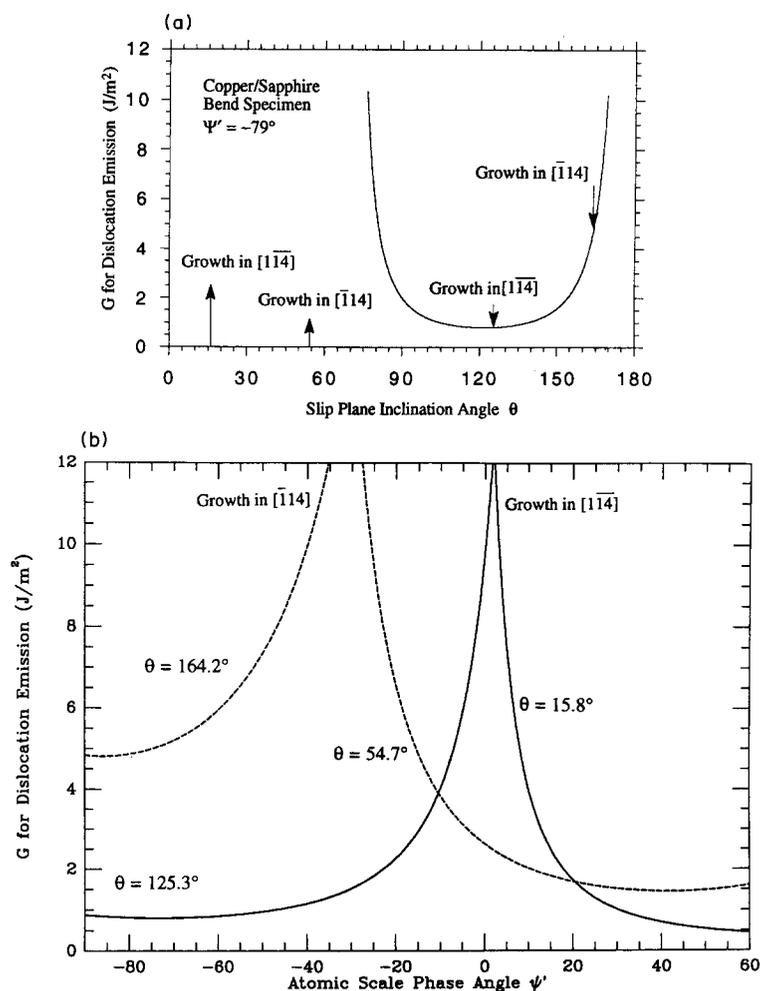


Fig. 3. (a) Plot of G_{disl} vs the slip plane inclination angle, as predicted by the Rice–Thomson model for a copper/sapphire interfacial crack; and (b) G_{disl} vs the atomic scale phase angle for various angles of interest. The solid lines give the relevant angles for crack propagation in the $[114]_{\text{Cu}}$ direction, and the dashed lines represent angles relevant to propagation in the $[\bar{1}\bar{1}4]_{\text{Cu}}$ direction.

with a Peierls-type stress versus displacement relation being satisfied as a boundary condition along a slip plane ahead of a crack tip. Once this problem is solved for a suitable constitutive relation for material sliding and perhaps opening along a slip plane, there is no need for the core cutoff parameters. The advantage of this method is that it allows for the existence of an extended dislocation core during nucleation, and eliminates uncertainty in choosing the core parameters. Numerical work by Beltz and Rice [14] agrees with the directional and phase angle effects given by the Rice–Thomson model [5] for the case of a metal/ceramic interface.

The directional effects observed here and in [6] are consistent with the fact that the ductile crack tip orientation involves slip planes which are more favored for dislocation motion (i.e. they have a higher resolved shear stress) than near the opposing crack tip. In light of the recent improvements in the modeling of dislocation nucleation, we feel it is appropriate to preview our experiments in terms of

the newer theory. The approach requires a knowledge of the Peierls-type shear stress τ ($\equiv \sigma_{r\theta}$, on $\theta = 0$ in the case now discussed) vs relative atomic displacement (denoted Δ_r) relation such as the sinusoidal representation in Fig. 4(a); Δ_r denotes the shift of one atomic plane relative to another at the slip surface. This curve gives the shear stress needed to locally shear atoms with respect to one another on a given slip plane, and is the fundamental input to the Peierls–Nabarro dislocation model [15,16]. The initial slope of such a curve corresponds with an appropriate shear modulus. The parameter b is the magnitude of a Burgers vector and represents the periodicity of the stress–displacement relation. This type of data has been calculated through the use of pair potentials or the embedded atom method by several researchers [17–20]. The integral of such a curve from $\Delta_r = 0$ to the unstable equilibrium at which the shear stress next vanishes (at $\Delta_r = b/2$ in simple cases) has been called [12] the *unstable stacking energy*, denoted γ_{us} . An estimate of this solid

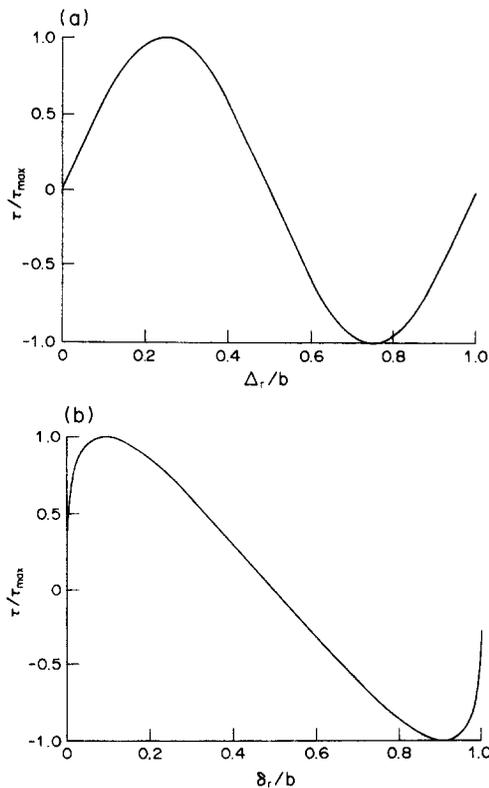


Fig. 4. Expected form of the shear stress τ vs: (a) relative atomic displacement Δ_r , and (b) displacement discontinuity δ_r .

state parameter for copper, based on Frenkel theory (which models the τ vs Δ_r curve as a sine wave), gives a value of about 0.22 J/m^2 .

Define δ_r as the displacement discontinuity on a mathematical cut coincident with the slip plane. We relate δ_r to the displacement Δ_r of the atomic planes at $\pm h/2$ from the cut by

$$\delta_r = u_r^{(+)} - u_r^{(-)} = \Delta_r - \frac{\tau h}{\mu} \quad (5)$$

where h is the interplanar spacing. This idealized cut represents the slip plane, and by adding to the displacement discontinuity δ_r across the cut (in what is otherwise considered a linear elastic continuum) the additional "elastic" displacement $h\tau/\mu$, we simulate approximately the relative displacement $\Delta_r = \delta_r + h\tau/\mu$ between atomic planes a distance h apart. If τ is now plotted versus δ_r , the curve becomes skewed so as to give an infinite slope at the origin [see Fig. 4(b)]. The integral of τ over half of a cycle remains equal to γ_{us} , however.

A J -integral calculation may be used to derive the following result for a mode II shear crack with a coplanar slip plane (see [21] for details)

$$G_{\text{disl}} = \frac{1-\nu}{2\mu} (K_{\text{II}}^{\text{disl}})^2 = \gamma_{us}. \quad (6)$$

With the exception of the nonlinear behavior along the slip plane, the material in this simple case is taken as an isotropic, linear elastic solid with shear modulus

μ and Poisson's ratio ν . As discussed below, G_{disl} may be calculated for more realistic situations involving inclined slip planes, mixed mode loadings, and bimaterial crack tip fields; however, the above result illustrates a feature that pervades these complicated cases: the energy release rate for dislocation nucleation scales with the recently identified solid state parameter γ_{us} .

Further complexities arise when we include in the model the effects of normal tractions and dilatant openings across the slip plane. This situation occurs if a model I-type loading is added to the mode II situation just discussed, or in more realistic cases when the slip plane is inclined with respect to the crack plane. There are no reasons to assume that a given τ vs δ_r curve retains its shape if tension is superposed; hence the effect of superposed tension on the "effective" γ_{us} must be investigated. Argon [10] and Cheung *et al.* [22] have already noted the importance of softening in shear due to large tensile stresses across a slip plane. In recent analyses by Beltz and Rice [13, 14], more realistic configurations, including ones involving slip planes inclined at the angles 125.3° and 164.2° have been treated. The effect of tension/compression normal to the slip plane was treated by assuming a tensile stress vs normal component of separation relation consistent with the well-known fit, with energy proportional to $-(L + \Delta_\theta)\exp(-\Delta_\theta/L)$, to the universal bonding correlation of Ferrante *et al.* [23]. The parameter L is the characteristic length associated with the decohesion process (the tensile stress reaches a maximum, at $\Delta_r = 0$, when $\Delta_\theta = L$).

Results given in [14] for a simplified set of slip plane constitutive relations $\tau(\delta_r, \delta_\theta)$ and $\sigma(\delta_r, \delta_\theta)$ for copper are now summarized. In Fig. 5, the applied energy release rate G/γ_{us} is plotted as a function of the crack tip opening displacement δ_r^{tip}/b for the two angles of interest. Here b is the Burgers vector of the partial (i.e. $b_{\text{total}}/\sqrt{3}$). When $\theta = 125.3^\circ$ (i.e. crack growth in the $[1\bar{1}4]_{\text{Cu}}$ direction), unstable nucleation of a Shockley partial dislocation occurs at $G/\gamma_{us} = 1.839$. When $\theta = 164.2^\circ$ (crack growth in the

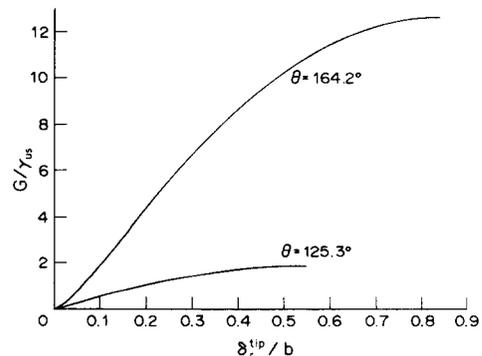


Fig. 5. The applied energy release rate versus the amount of slip at the crack tip for the two angles of interest, as predicted by the Peierls-type theory.

$[\bar{1}14]_{\text{Cu}}$ direction), instability occurs at $G/\gamma_{\text{us}} = 12.55$. If we assume a value of 0.22 J/m^2 for the unstable stacking energy, the relevant energy release rate values are $G_{\text{disl}} = 0.405 \text{ J/m}^2$ for the ductile direction and $G_{\text{disl}} = 2.76 \text{ J/m}^2$ in the brittle direction. There is more than a factor of 6 difference in G_{disl} for crack growth in the opposing directions; hence, it is concluded that dislocation nucleation is preferred in the $[\bar{1}14]_{\text{Cu}}$ direction, and blunting should be favored in this growth direction.

EXPERIMENTAL

The single crystals of copper used in these experiments were grown by the vertical Bridgman technique in a graphite mold in an atmosphere of flowing argon gas. The crystal seeds were cut from pure copper crystals also grown by way of the Bridgman technique. The orientations of the seeds and the bicrystals were determined by the back-reflection Laue method with an accuracy of approximately $\pm 2^\circ$. The crystals used to make specimens were cut from the master ingot via spark-cutting, and were 47.6 mm by 2.5 mm by 4.8 mm. The copper pieces were diffusion bonded to 1 mm thick, commercially-obtained sapphire slides on the basal plane, at 1040°C , for 72 h, under a rigid graphite support clamp. The atmosphere consisted of a flowing hydrogen/80% argon (by volume) mixture. Several specimens were annealed after the bonding process in order to reduce the size and number of smaller grains which appeared near the copper/sapphire interface during diffusion bonding.

The surface of the sapphire layer in the specimens was "pre-notched," either by use of a diamond scribe or a diamond saw. The purpose of this was to insure that the initial crack through the sapphire would start at the midpoint of the specimen. The initial crack was induced by subjecting the specimen to either three- or four-point bending. After pre-cracking, the interfaces were observed through the sapphire layer to verify that the initial crack had branched onto the copper/sapphire interface in both directions. The specimens were then loaded under four-point bending until a nonlinear load versus time relation was noted. The specimens were unloaded and reloaded several times, and the interfaces of some specimens were observed after each unloading step. Five identical specimens were tested.

RESULTS AND DISCUSSION

The copper/sapphire interfaces appeared to be homogeneous, well-bonded, and free of any potential reaction products. This is consistent with the work of Heidt and Heimke [24] and Mulder and Klomp [25], which showed that the interdiffusion distance in the copper/sapphire system is negligible and that no additional chemical products are formed, when the interface is created under conditions similar to ours. When viewed via optical microscopy through the

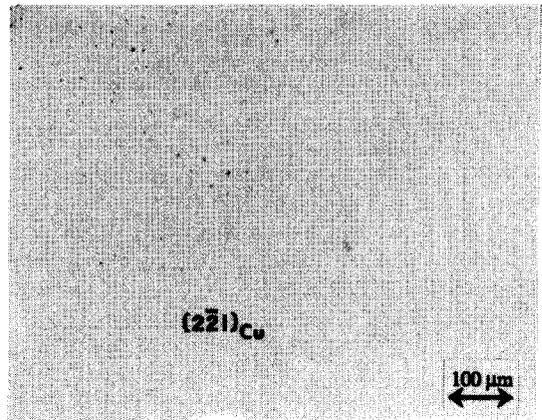


Fig. 6. View through the sapphire layer of the copper/sapphire interface, which shows the presence of pores after the diffusion bonding.

sapphire layer, pores were observed on the interface, as shown in Fig. 6; this is not uncommon in metal/ceramic interfaces involving a ductile metal and have been discussed in [26–28] and references therein. We note that the atomistic models for dislocation nucleation we utilize say nothing about the details of the bonding process or conditions at the interface in general, except that the interface be "atomically sharp." Effects relating to the interfacial strength enter the Rice–Thomson model through the parameter G_{cleave} , which we assume to be identical in the two crack growth directions. As an estimate of the order of magnitude of G_{cleave} , we quote the result of 0.475 J/m^2 , for the work of adhesion of $\text{Cu}/\text{Al}_2\text{O}_3$, as measured by Nicholas [29] at a temperature near that of the melting temperature of copper. This value may be interpreted as a lower bound approximate value to the ideal work of fracture $2\gamma_{\text{int}}$ for the $\text{Cu}/\text{sapphire}$ interface at room temperature.

After an initial crack was introduced on to the interface, crack growth occurred significantly in all specimens upon further loading in only *one* direction namely, the $[\bar{1}14]_{\text{Cu}}$ direction with respect to the copper crystal (see Fig. 7). In one case the effect was particularly dramatic: the crack propagating in the $[\bar{1}14]_{\text{Cu}}$ direction ran unstably all the way to the end of the specimen, causing half of the sapphire layer to become entirely detached. Figure 8(a) shows the notched area of the latter specimen after the bend-test; note that there was some crack growth in the ductile direction in this case. No future growth was observed, however, of the crack tips pointed in the $[1\bar{1}4]_{\text{Cu}}$ (ductile) direction. The crack fronts were typically rounded as schematically shown in Fig. 9, thus the crack tip pointed in the ductile direction is not visible from the side of most specimens (e.g. Fig. 7). Table 1 gives an estimate of the critical energy release rate for fracture of the brittle crack, as given by the bimaterial Irwin-type expression associated with equation (1) (as given, e.g. in [5]); the values hover in the range of $5\text{--}7 \text{ J/m}^2$.

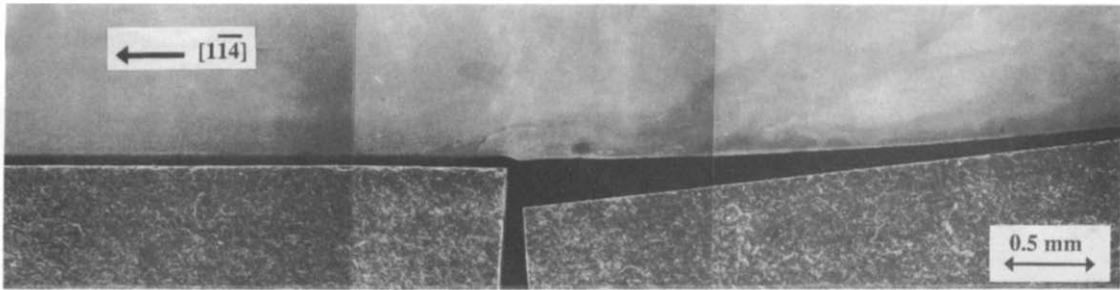


Fig. 7. Side view of a copper/sapphire specimen after mechanical testing, showing that the crack propagated only in the $[1\bar{1}4]_{Cu}$ (brittle) direction.

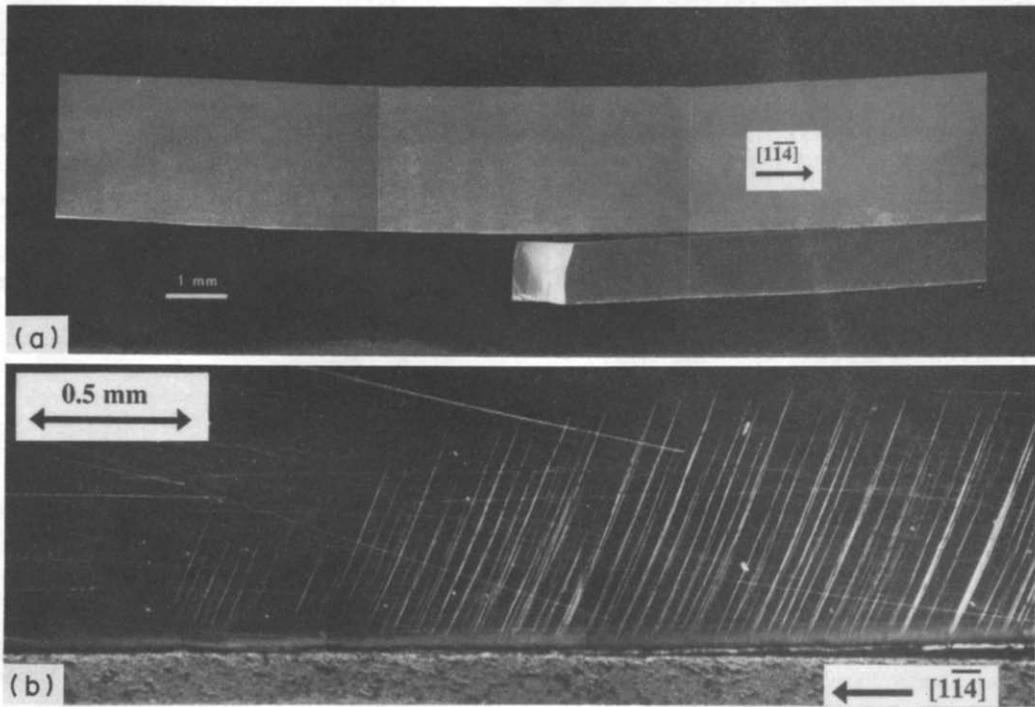


Fig. 8. Side view of a copper/sapphire specimen after mechanical testing: (a) shows complete decohesion in the $[1\bar{1}4]_{Cu}$ (brittle) direction accompanied by crack growth in the $[1\bar{1}4]_{Cu}$ (ductile) direction and (b) shows the relatively high density of slip lines in the vicinity of the crack which propagated in the ductile direction.

Dense slip lines were present on the $(110)_{Cu}$ (side) surfaces in the vicinity of the crack tip oriented in the $[1\bar{1}4]_{Cu}$ direction, indicating a large amount of dislo-

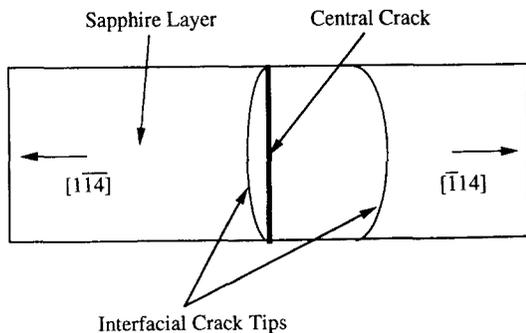


Fig. 9. Schematic showing the typical shape of the crack front as it propagated along the copper/ceramic interface.

cation activity. In contrast, interfacial debonding occurred along the $[1\bar{1}4]_{Cu}$ (brittle) direction; slip lines were much less dense in this region of the specimen, indicating that a debonding process dominated. Figure 8(b) shows the slip traces observed in the vicinity of the ductile crack tip. These lines originally corresponded to the slip planes inclined at 125.3° with the crack plane, which is the predicted slip system (see Figs 3 and 5); because the specimens were tested until

Table 1. Experimentally obtained values of G_{cleave} in copper/sapphire

Specimen number	G_{cleave} (J/m ²)
1	7.85
2	5.13
3	5.03
4	5.86
5	5.86

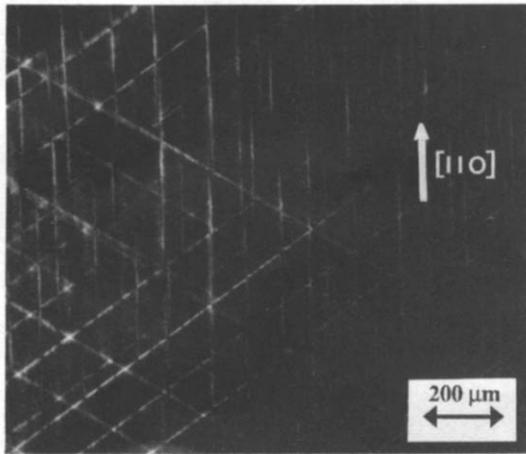


Fig. 10. Slip traces in the vicinity of the crack running the $[1\bar{1}4]_{\text{Cu}}$ (ductile) direction, as seen in this view of the interface. Note that slip lines are present for systems which do not intersect the crack front.

macroscopic bending set in, these slip planes have been distorted somewhat within the plane, hence they are tilted slightly curved. Slip traces corresponding to the favored plane for dislocation emission from the brittle crack tip, at 164.2° , were not observed; in fact, the only system that appears to have been activated of the slip systems that contain a slip plane coplanar with the crack front is the system which contains a slip plane oriented at 125.3° with respect to the crack running in the $[1\bar{1}4]_{\text{Cu}}$ direction. It is evident from Fig. 10, however, that additional slip systems, having $\{111\}$ type slip planes which are *not* coplanar with the crack front, were active, and provided further means for the accommodation of plastic flow and the dissipation of energy.

The results here have been explained in terms of theories which have only taken into account the emission of a single Shockley partial dislocation for the copper crystal orientation considered. Whereas the second partial dislocation to nucleate possesses a Burgers vector which is perpendicular to the crack front, the first does so at an angle which is 30° from the crack front. The numerical procedures associated with the Peierls-type nucleation model are not yet adequate to deal with the latter case, but a calculation presented in [12] suggests that the relevant values of G_{disl} will increase by a factor of $(4 - 3\nu) \approx 3$ to nucleate the first partial, thus retaining the factor of 6 difference between nucleation levels for the two cracking directions considered. The emission of partial dislocations within the earlier Rice–Thomson framework has been worked out in considerable detail by Anderson [30].

As alluded to in the experimental section, it was observed that crack growth proceeded initially in a symmetric fashion, i.e. crack growth occurred in both directions for a short time. The crack growing in the predicted ductile direction never grew longer than about 0.5 mm, however. This result can be

rationalized by at least one explanation: the elasticity solution of equation (1) for this specimen assumes that the crack length is much longer than the thickness of the thin layer (i.e. the thickness of the sapphire). When the crack is short (i.e. of a comparable length scale to the sapphire thickness or shorter), the phase angle differs as given by an analysis by He and Hutchinson [31]. The local phase of the incipient cracks is less than the steady-state value of -79° , which leads to a condition where the values of G_{disl} at the two crack tips are more similar.

SUMMARY

Predictions of a modified Rice–Thomson model for the ductile versus brittle behaviour of cracks, which takes into account a bimaterial crack-tip field, are in general agreement with experimental observations of beam-type layered specimens consisting of copper bonded on a $\{221\}$ face to sapphire. The agreement indicates that the different fracture behaviour of these bimaterial interfaces may be understood, at least qualitatively, by comparing the crack tip energy release rate for dislocation emission from the crack tip against that for cleavage decohesion of the boundary. The observations here are also in agreement with an improved method for calculating G_{disl} which takes into account the nonlinear core effects associated with extended dislocations.

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