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WORKING PARTY ON "FRACTURE PHYSICS"
(29 May - 16 June 1989)

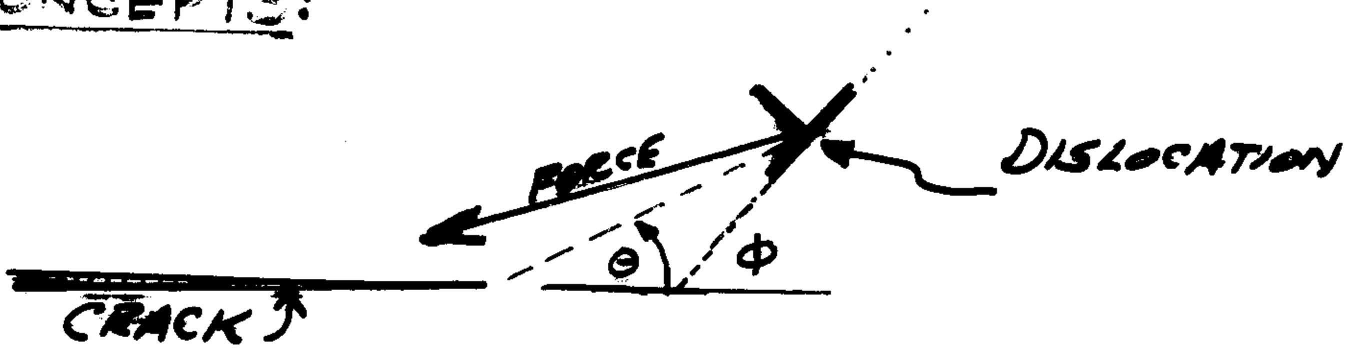
CRACK TIP SHIELDING

S. BURNS
University of Rochester
Materials Science Department
New York
Rochester 14627
U.S.A.

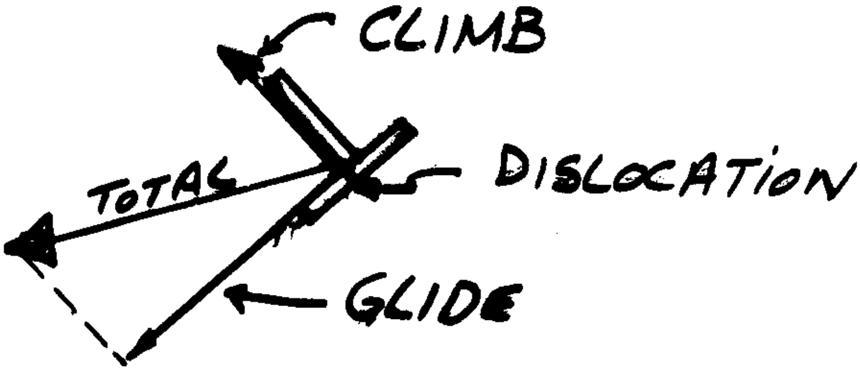
CRACK TIP SHIELDING

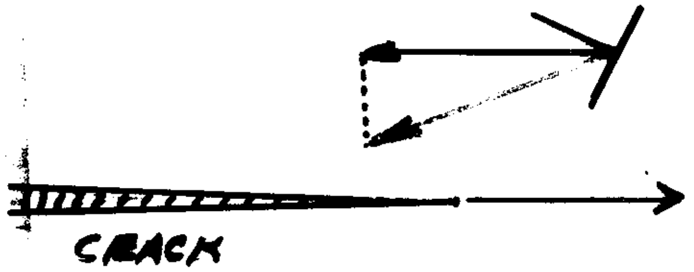
STEPHEN BURNS
MATERIALS SCIENCE PROGRAM
DEPARTMENT OF MECHANICAL ENGR.
UNIVERSITY OF ROCHESTER
ROCHESTER, NEW YORK 14627

CONCEPTS:



$$\frac{\partial U}{\partial A} = -J = \text{DRIVING FORCE}$$



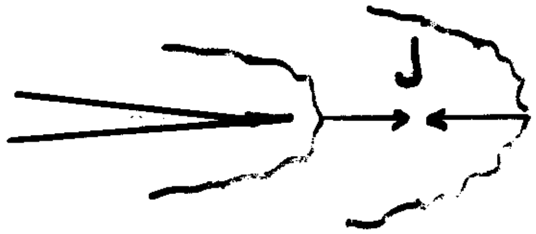


DISLOCATION

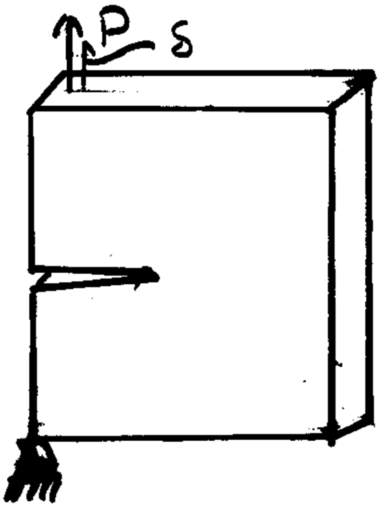
$$\frac{\partial u}{\partial x} = J$$

ΔX IS THE SAME FOR DISLOCATION AND CRACK.

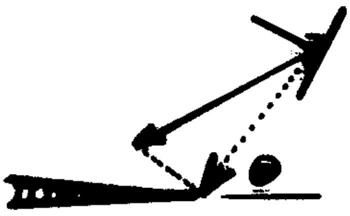
Recall:



$$du = P\,d\delta - J\,dA \qquad \left.\frac{\partial u}{\partial A}\right|_{\delta} = -J$$



ASARO'S THEOREM



$$\mathcal{G} = \frac{K_I^2}{E}(1-\nu^2)$$ FOR MODE I APPLIED

FOR DISLOCATION:

$$k_i = K_i + \sum \Delta k_i$$

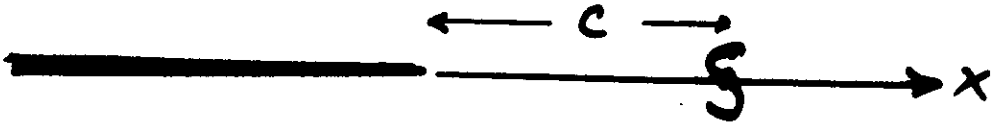
TIP ↗ APPLIED ↑ SHIELDS (NEGATIVE) ↖

$$\mathcal{G}_{TIP} = \frac{k_I^2}{E}(1-\nu^2) + $$ [illegible] $$(1-\nu^2) \quad ; \quad K_i = 0$$

MODE III STRESS

$$\tau_{yz} = \frac{\mu b}{2\pi}\frac{\sqrt{c}}{\sqrt{x}\,(x-c)}$$

$$x \to 0 \quad -\underbrace{\frac{\mu b}{2\pi\sqrt{c}}}_{\text{SHIELD}}\frac{1}{\sqrt{x}}$$



$$\sqrt{2\pi x}\;\tau_{yz} = \frac{\mu b}{\sqrt{2\pi c}}$$

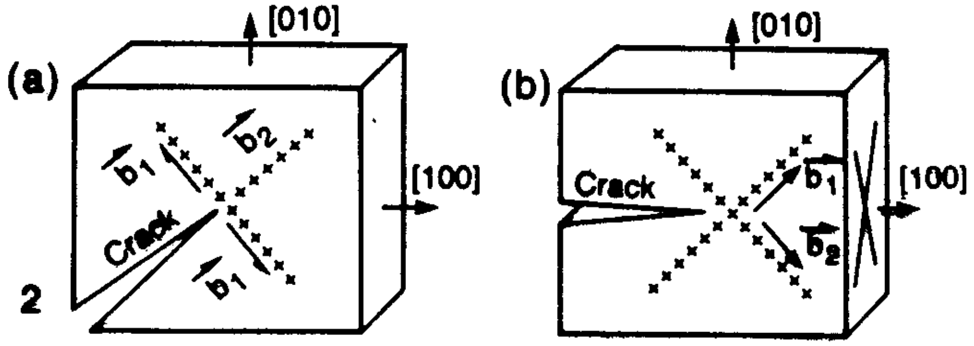


Fig. 2

A schematic representation of (a) ($\bar{1}10$) plane crack with (110) and ($\bar{1}10$) plane edge dislocations. (b) (010) plane crack with (110) and ($\bar{1}10$) plane edge dislocations. The more common screw dislocations are lightly shown on the (100) face.

. A photomicrograph of crack tip deformation in 2(b). The horizontal line is the crack. The cross are edge dislocations, some screw dislocations are also seen parallel to the crack.

RICE & THOMSON

$$\Delta K_I = -\frac{E b_e}{4(1-\nu^2)} \frac{1}{(2\pi r)^{1/2}} \left\{ 3 \sin\phi \cos\theta/2 - \sin(\phi-\theta)\cos\frac{3\theta}{2} \right\}$$

$$\Delta K_{II} = -\frac{E b_e}{4(1-\nu^2)} \frac{1}{(2\pi r)^{1/2}} \left\{ 2\cos\phi\cos\theta/2 - \sin\phi\sin\theta/2 + \sin(\phi-\theta)\sin 3\theta/2 \right\}$$

$$\Delta K_{III} = -\frac{E b_s}{2(1+\nu)} \frac{1}{(2\pi r)^{1/2}} \cos\theta/2 .$$

$K_{TIP} = K_{APPLIED} + \Delta K_{\text{shields}}$

Shields <u>ADD</u>

$\Delta K \simeq \frac{\mu b}{\sqrt{2\pi r}}$

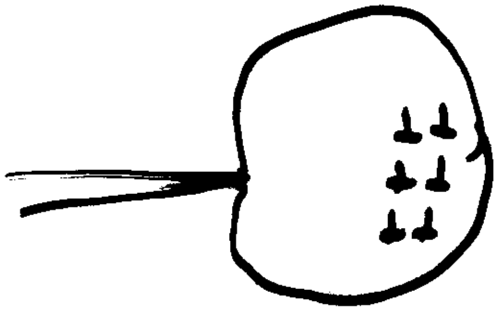
Mode III

$b = 2 \times 10^{-10}$ m

$\mu = 100 \times 10^{9}$ Pa

$r = 5$ μm

$\Delta K \approx 3$ k Pa$\sqrt{m}$ / Dislocation



N = # Dislocations

ρ = N/Area.

10 μm plastic zone $\rho \simeq 10^{14}/m^2$

$N \simeq 3 \times 10^{4}$

$\Delta K \simeq 150$ MPa$\sqrt{m}$

APPLICATION OF THERMODYNAMICS TO FRACTURE

S. J. Burns
University of Rochester
Rochester, NY 14627

Abstract

An energy balance formulation of non-linear fracture mechanics is reviewed. A thermodynamic description of any elastic cracked solid is formulated. The crack driving force, J which is assumed to be non-measurable is related to the laod, load point displacement and the crack area. In linear elastic systems, a mechanical stability criteria is related to the curvature of the compliance or stiffness versus crack area.

Slow crack growth is assumed to be a thermally activated process. The steady state crack velocity versus crack driving force in any specific form such as log log plots or semi-log plots are used to describe the crack tip processes that are rate controlling crack growth.

Introduction

Thermodynamics is concerned with energy representations and balances in physical systems. Thermodynamics methods are not predictive. However, when combined with a physical assumption it provides a powerful tool for establishing the complete consequences of the assumption. Assumptions that violate any form of the laws of thermodynamics are not useful for describing physical processes.

Fracture mechanics and the processes taking place at the tip of a crack may be described in thermodynamic formulations (Burns, Pollet and Chow, 1978 and Pollet and Burns, 1977(a) and 1977(b)). In this text, I will briefly review the formulation of fracture mechanics from a thermodynamics point of view and then take a careful look at the kinetics of slow crack growth. This later section assumes that the fracture mechanics part of the cracked elastic solid is solved but that the fracture modified material at the tip of the crack contains an unspecified rate controlling process.

Elastic Fracture Mechanics

Figure 1 shows a partially cracked solid that is divided into two thermodynamic systems: System I contains the load, P, the load point

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Fracture Toughness of MgO-Partially-Stabilized ZrO_2 Specimens with K_R-Curve Behavior from Transformation Toughening

S. J. BURNS

Materials Science Program, Department of Mechanical Engineering, University of Rochester, Rochester, New York 14627

M. V. SWAIN*

Commonwealth Scientific and Industrial Research Organization, Division of Materials Science, Melbourne, Victoria, Australia

Transformation-toughened MgO-partially-stabilized ZrO_2 exhibits a crack-wake-induced load-point displacement in fracture samples. The equations of linear elastic fracture mechanics were self-consistently reformulated to include the residual displacement from the transformation wake. Application of these equations to double-cantilever-beam specimens of transformation-toughened ZrO_2, with the martensitic start transformation temperature close to the testing temperature, gives a decreasing K_R curve and initial toughness values near 30 MPa·m$^{1/2}$.

I. Introduction

THE advent of high-strength, high-fracture-toughness zirconia-based ceramics had led to fracture toughness measurements that show increasing toughness with crack advance, i.e., *R* curve behavior.[1] Linear elastic fracture mechanics, LEFM, applied to these ceramics have discrepancies between the work of fracture, in equivalent stress intensity units, the fracture toughness as measured from the load and the crack length, and the fracture toughness if it is measured using the load-point displacement and the crack length.[2] The fracture toughness values will be specimen and analysis sensitive since the requirements of LEFM are not fulfilled: the specimen has a residual load-point offset upon unloading. Physically, as the crack advances the zirconia material in the crack tip region transforms from tetragonal to monoclinic symmetry and leaves a crack wake that contains residual stress. The wake of transformed material is believed to be the source of the very high fracture toughness in zirconia-based ceramics;[2-6] however, it is also the source of the load-point offset. Magnesia-partially-stabilized zirconia, Mg-PSZ, specimens after fracture contain residual stresses that distort the specimens. This distortion is from the transformed wake left by the passage of the crack.[2]

In this paper, an energy balance is applied to fracture specimens that are transformation toughened during crack advance. The equations for fracture toughness values are developed for fracture specimens that exhibit load-point displacement offsets[7-9] due to transformation wakes with residual displacements. Finally, application of the equations to Mg-PSZ in double-cantilever-beam fracture specimens, with *R*-curve effects, are illustrated.

II. Fracture Mechanics Analysis

Figure 1(*A*) shows a schematic fracture specimen that contains a fracture-induced wake that gives a residual load-point displacement. Figure 1(*B*) is the anticipated load vs load-point displacement curves for this specimen. The fracture specimen is assumed to be linear elastic with a crack-length-dependence compliance and a residual displacement at zero load that is also dependent on the crack length.[7]

An incremental energy balance on this specimen during isothermal fracture is[10]

$$dU = P\,d\Delta - J\,dA \tag{1}$$

dU is the incremental change of the internal energy of the entire specimen. P is the applied load, J is the crack driving force, $d\Delta$ is the incremental load-point displacement, and dA is incremental area swept out by the crack.

The complementary energy to U is U' with

$$U' = U - P\Delta \tag{2}$$

Differentiation of Eq. (2) with the aid of Eq. (1) gives

$$dU' = -\Delta dP - J\,dA \tag{3}$$

The Maxwell relation from Eq. (3) is

$$\left.\frac{\partial J}{\partial P}\right|_A = \left.\frac{\partial \Delta}{\partial A}\right|_P \tag{4}$$

Equation (4) upon integration yields a general expression for J.

$$J = \int \left.\frac{\partial \Delta}{\partial A}\right|_P dP + \theta(A) \tag{5}$$

The integral in Eq. (5) is computed with A held constant as prescribed by Eq. (4). $\theta(A)$ is the "constant" of integration and is needed since the left-hand side of Eq. (4) requires that A be held constant.

Figure 1(*b*) has anticipated a load vs load-point displacement relation in the form

$$\Delta = C(A)P + \phi(A) \tag{6}$$

$C(A)$ is the linear elastic specimen compliance and $\phi(A)$ is the displacement offset at zero load. Differentiation of Eq. (6), substitution into Eq. (5), and integration give the expression for J in fracture specimens that are linear elastic with residual load-point displacements.

$$J = \frac{1}{2}P^2\frac{dC}{dA} + P\frac{d\phi}{dA} + \theta(A) \tag{7}$$

Equation (7) differs from the well-established equations of linear elastic fracture mechanics in two ways: first, the middle term on the right accounts for the residual displacement offset due to the wake; second, the arbitrary function $\theta(A)$ is used to establish where $J = 0$. In LEFM, $J = 0$ when $P = 0$ for all crack lengths; note that this is the point where all unloading lines cross. In transformation-toughened ceramics, when a residual displacement is presented in the fracture specimen, unloading lines form a tangent envelope when P is negative. Recall that the crack-wake microstructure gives a negative crack tip shield as determined from crack tip stresses, that is not zero if there is no applied load. The crack surfaces due to the transformed microstructure behind the crack have a positive opening that is just as real as if forces had been applied to the cracked specimen. The untransformed part of the fracture specimen near the crack surface has tensile stresses with $P = 0$, due to the compressive stresses generated in the transformed wake. The residual strains in the wake give an offset.

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A Thermodynamic Approach to Fracture Toughness in PSZ

R. J. SEYLER, S. LEE, AND S. J. BURNS

University of Rochester
Materials Science Program
Department of Mechanical Engineering
Rochester, NY 14627

A thermodynamic analysis relating changes in fracture toughness to thermal length changes and the enthalpy of the tetragonal-to-monoclinic phase transformation in partially stabilized zirconia (PSZ) was proposed previously. The Clausius-Clapeyron equation is used to relate volume and shear phase changes to the crack-tip deformation zone. Shear transformations, which have a unique role in fracture toughening of PSZ, shield the crack tip from the applied stress intensity factor for a crack that is held stationary. Crack-tip shields from volume and normalized strain transformations are computed to be zero for stationary cracks and this deformation zone is much smaller than the shear transformation zone. Shear transformations in bulk specimens are deduced by measuring length changes vs temperature on samples that were previously subjected to a uniaxial compressive stress, i.e., with a shear component, during a thermal cycle. Thus, the constraint on the shear strain transformation in the bulk sample is directly included in the shear stress-biased thermally cycled samples undergoing tetragonal-to-monoclinic transformations. Subsequent monoclinic-to-tetragonal transformations during heating, with zero applied stress, record the shear stress-biased length changes.

The rare earth oxide ZrO_2, zirconia, is a ceramic known for its high-temperature chemical inertness. Zirconia and partially stabilized zirconia (PSZ) may be microstructurally modified by thermal treatment to improve their fracture toughness by either microcracking and/or phase transformation toughening. At present, specific guidelines for optimum phase transformation toughening of PSZ (and other materials) are not available. However, phase transformation toughening of PSZ, which is the subject of this paper, has been previously considered.[1-14]

Crack-Tip Transformation Zone

The Clausius-Clapeyron equation is used to relate a phase transformation temperature change to a hydrostatic pressure change for a specified volume change between thermodynamic phases in equilibrium.[15] Conceptually, the Clausius-Clapeyron equation represents the equality of the Gibbs free energy of both phases during transformation where pressure and volume are conjugate-state variables. A stress-induced phase transformation near a crack tip will satisfy this Gibbs equality of phases at the transformation zone boundary.

A crack-tip stress field contains significant shear stresses, and the zirconia phase change of interest, tetragonal (t) to monoclinic (m), is a shear transformation that includes a volume change. Although the shear strains have no associated volume change, a shear stress work term exists. Consequently, for the Clausius-

CRACK TIP DEFORMATION IN LiF FOR MODE I AND II CRACKS ON (100) AND (110) PLANES

by

K. Y. CHIA and S. J. BURNS

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Measurement of Localized
Deformation
by
Novel Techniques

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Edited by

W. W. Gerberich
University of Minnesota
Minneapolis, Minnesota

D. L. Davidson
Southwest Research Institute
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CRACK TIP DEFORMATION IN LiF SINGLE CRYSTALS

K. Y. Chia and S. J. Burns
Materials Science Program
Department of Mechanical Engineering
University of Rochester
Rochester, N.Y. 14627

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Introduction

Experimentally the plastic zone ahead of a crack tip has been investigated and observed (1-3). However there are very few observations of dislocation-free zones and individual dislocations at the tip of mode I propagating cracks. Recently, some materials have been investigated by in-situ experiments in an electron microscope (4-8). Dislocation-free zones have been reported in Mo, W, Al and Cu with mode III cracks; some materials such as stainless steel do not seem to have such zones (9).

Theoretically, Bilby, Cottrell and Swinden (10) in 1963 proposed a crack tip dislocation plasticity model - the BCS theory. In this theory, they introduced the concept of a continuous dislocation distribution to account for a finite plastic zone ahead of an in-plane shear crack. The BCS theory predicts an asymptotically infinite dislocation density at the immediate tip of the crack. Physically, this is not possible, since dislocations cannot be less than a unit Burgers vector apart.

An in-plane discrete dislocation pile-up ahead of a mode II/Mode III crack tip has been analyzed. Dai and Li (11) and Zhao, Dai and Li (12) used the method of Louat (13) and computer simulated crack tip dislocations. They found for most deformation examples that there is a dislocation-free zone ahead of the crack tip. An unloading case has also been simulated by the same method. It was found that the dislocations will be drawn to the crack tip by the image force after unloading (11). The extent of the draw-back depends on the lattice friction and the image force. Chang and Ohr modelled the dislocation distribution and the crack tip dislocation-free zone in terms of elliptic integrals of the first and third kinds (14). A compatibility condition between the crack and the dislocation-free zone was obtained as a function of applied stress, similar to the condition derived in the BCS theory of fracture.

The conditions for the formation of a dislocation-free zone have been discussed by several authors (12,14-18). However, it is concluded in the interactions between dislocations and crack tips that the dislocations have a substantial influence on the stress intensity factor at the crack tip (15). The applied stress intensity factor is reduced in the crack tip region by a negative contribution from each dislocation near the crack tip to the applied stress intensity factor. In other words, the crack tip is shielded by the nearby crack tip dislocations so additional applied stress is needed to propagate the crack and the material is tougher because of the crack tip dislocations.

In this report, mode I crack tip deformation in LiF single crystals has been studied. Dislocation distributions ahead of a crack tip for two different deformation modes are observed: the first deformation mode blunts the crack front; the second deformation mode jogs the crack front. These deformation modes involve different dislocations and slip characteristics. A dislocation-free zone has sometimes been found in both of these deformation modes. Unloading effects on crack tip deformation have been investigated. The results show that unloading contributes to deformation in the crack tip region, which is contrary to the usual assumption that removal of the applied load relieves the crack tip stress intensity factor.

Experiments and Results

A Harshaw single crystal of LiF was radiation-hardened to the color of light yellow by a cobalt 60 source. The crystal in this condition can be cleaved without introducing too many

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VIEWPOINT SET No. 10

CRACK TIP DISLOCATION NUCLEATION OBSERVATIONS in BULK SPECIMENS

S.J. Burns
Materials Science Program
Department of Mechanical Engineering
University of Rochester
Rochester, NY 14627

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Introduction

The brittle to ductile fracture transition is an important material property that is poorly understood. Materials are often subject to service applications near the fracture transition temperature so this property has practical importance. In addition, chemical and radiation effects cause materials to become embrittled and change the transition temperature. This fracture transition is controlled, at least in part, by the ability of a crack tip to exhibit local ductility. Crack tip nucleation or generation of dislocations is a necessary condition for crack tip ductility. Engineering, fracture tough, materials must, of course, have extensive crack tip ductility. This paper reviews some of the background concepts in dislocation crack tip nucleation and provides quantitative observations of etched dislocations in cracked single crystals of LiF.

Crack tip generated dislocations produce two effects. First, they reduce the applied stress intensity factor at the true crack tip by 'shielding' the true crack tip (1-3). The dislocation crack tip 'shield' is a negative stress intensity factor that reduces the applied stress intensity factor at the crack tip. Crack tip dislocation shields add because stresses add. Second, crack tip dislocation emission changes the geometry of the crack tip: edge dislocations with a component of the Burgers' vector normal to the fracture plane blunt the crack tip (2,4,5); screw dislocations with a Burgers' vector component normal to the fracture plane jog the crack plane (4-7). Thus, crack tip dislocation generation helps reduce some crack tip stresses and changes the crack tip geometry and bonding. Gilman (8,9) observed crack generated dislocation half-loops on (100) cleavage planes of LiF single crystals after crack propagation. The proposed model, at that time, was that the propagating crack nucleated dislocation loops, and subsequently it cut through the dislocation structures, leaving on the cleavage plane half-loops with screw dislocations intersecting the fracture surface (10). The crack velocity associated with loop deposition was thought to be sensing the dislocation nucleation rate (11). Subsequent interpretation of the observed half loops indicated that the bands of half loops were places where the crack driving force oscillated because of mechanical vibrations in the dynamically fractured specimens (12). The dislocations left by the crack slightly below the cleavage plane were released by image forces leaving only the half loops that had been propagated deeper into the crystal by the oscillating crack driving force (13). Thus, half loops were observed.

The first analytic criteria for crack tip dislocation nucleation resulted from the dislocation half loop observations. Friedel (14) assumed a dislocation loop would be nucleated by a crack when the dislocation loop had a critical curvature on the size scale of the Burgers' vector. Armstrong (15) considered a dipole loop nucleated in the plane of a penny shaped crack but neglected crack image forces. Burns and Webb (7) considered the competition between a positive force on the dislocation caused by the stress field of the crack and the negative image force. The maximum force on the dislocation was set equal to the flow stress. The problem assumed a dislocation was already at the crack tip and calculated when it would collapse back into the crack from a moving crack. Rice and Thomson (2) also considered the competition between crack forces and image forces for dislocation nucleation. The criteria was that the stress on the dislocation at one dislocation core radius from the crack tip would have to be just slightly positive. In their paper the crack-dislocation interactions were derived. Chang and Ohr (16) and Ohr and Chang (17) have modified Rice and

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CRACK TIP SHIELDING—AN ELASTIC THEORY OF DISLOCATIONS AND DISLOCATION ARRAYS NEAR A SHARP CRACK[†]

B. S. MAJUMDAR[‡] and S. J. BURNS

The Materials Science Program, Department of Mechanical and Aerospace Sciences,
University of Rochester, Rochester, N.Y. 14627, U.S.A.

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Abstract—An elastic solution has been found for a screw dislocation near a crack in the absence of any external stresses. The dislocation produces a stress intensity factor on the crack even without external stresses. The stress intensity factor at the crack tip plus the dislocation's contribution to the stress intensity factor add to the stress intensity factor due to the applied stresses. Also the interaction between the crack and dislocations, and between dislocations in the presence of the crack can be solved by simply considering the associated stress intensity factors at the crack tip. Several additional solutions derived are two dislocations at the crack tip, a dislocation off the crack plane, a circle of dislocations around the crack tip and a finite length crack with external stresses.

Résumé—Nous avons trouvé une solution élastique pour une dislocation vis près d'une fissure en l'absence de toute contrainte externe. La dislocation produit un facteur d'intensification de la contrainte sur la fissure, même en l'absence de contraintes externes. Le facteur d'intensification à l'extrêmité de la fissure et la contribution de la dislocation au facteur d'intensification de la contrainte s'ajoutent au facteur d'intensification de la contrainte appliquée. On peut également trouves les interactions entre la fissure et les dislocations, et entre dislocations en présence de la fissure, en considérant simplement les facteurs d'intensification de la contrainte associés à l'extrêmité de la fissure. Nous avons également obtenu les solutions correspondant à deux dislocations à l'extrêmité de la fissure, une dislocation en dehors du plan de la fissure, un cercle de dislocations autour de l'extrêmité de la fissure et une longueur finie de fissure en présence de contraintes externes.

Zusammenfassung—Für eine Schraubenversetzung in der Nähe eines Risses wurde eine elastische Lösung bei Abwesenheit äußerer Spannungen aufgestellt. Die Versetzung erzeugt einen Spannungsintensitätsfaktor sogar ohne die äußere Spannung. Der Spannungsintensitätsfaktor an der Rißspitze plus der Beitrag der Schraubenversetzung addieren sich zur angelegten Spannung. Des weiteren kann die Wechselwirkung zwischen Riß und Schraubenversetzung und zwischen Versetzungen bei Anwesenheit eines Risses einfach durch eine Betrachtung der entsprechenden Spannungsintensitätsfaktoren an der Rißspitze gelöst werden. Weitere zusätzliche Lösungen werden abgeleitet für zwei Versetzungen an der Rißspitze, für eine Versetzung außerhalb der Rißebene, für einen Versetzungsring um die Rißspitze herum und für einen endlichen Riß unter äußerer Spannung.

1. INTRODUCTION

Material behavior at the tip of a crack will ultimately form the basis for all rational fracture theories. Orowan [1] and Irwin [2] modified Griffith's [3] totally brittle, elastic fracture theory by adding a plastic energy term, to account for local crack tip yielding of the material. Such a treatment is unsatisfactory from a materials standpoint since plasticity aspects are ultimately explained in terms of individual dislocation phenomenon and other defect theories. The Bilby, Cottrell, Swinden (BCS) [4] theory introduced the concept of a continuous distribution of dislocations to account for a finite plastic zone ahead of an inplane shear crack (Mode II). However, the individual interaction of dislocations with one another and their specific role in modifying the crack tip stress field remains unclear. Further, the BCS theory predicts asymptotically infinite dislocation densities at the immediate tip of the crack. Physically this is not possible, since dislocations cannot be less than a unit Burgers' vector apart. In the case of a propagating crack, the drag forces associated with moving dislocations has been treated in detail in [5].

Several recent observations indicate that crack tips can remain atomically sharp. Lawn and Hockey [6, 7] obtained transmission electron micrographs of crack tip regions in SiC and Al_2O_3. There are no crack generated dislocations and these materials are ideally brittle. TEM observations of *in situ* shear cracking of Mo and W [8], and stainless steel [9], show crack tip dislocations. The micrographs provide excellent evidence of dislocation distributions ahead of sharp crack tips in Mode III (antiplane shear) separation. TEM observations have also been made on

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[‡] Present address: Systems Research Laboratories, Inc., Dayton, OH 45440, U.S.A.