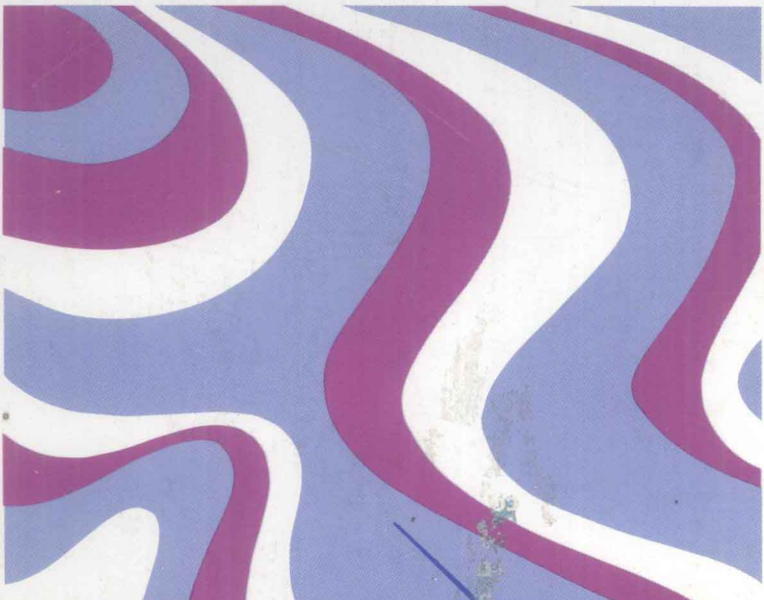
Fracture Mechanics: microstructure and micromechanisms

S. V. Nair, J. K. Tien, R. C. Bates, O. Buck



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Preface

In addition to structural design, cracked-body fracture mechanics properties are currently being used for alloy evaluation and design. Macroscopic fracture mechanics have continuously undergone mathematical and testing development in recent years through the spectrum of linear elastic (K_I) to elastic/plastic (J_I) and creep/plastic (C^*) flow regimes. These developments have been articulated and discussed in various symposia in the past, especially under ASTM auspices. The focus of this volume is the state of knowledge on the micromechanisms and metallurgical aspects of fracture mechanics.

The chapters in this book provide interpretive reviews of micromechanisms and the effects of microstructure on the most important areas of fracture mechanics, namely fracture toughness (K_{IC} or J_{IC}), fatigue or static thresholds ($\triangle K_{th}$ or K_{ISCC}) including environmental effects, and fatigue and creep crack growth rates (da/dN or da/dt), including corrosion fatigue and creep/fatigue/environmental interactions. There is also an emphasis in this book on the modeling of these micromechanisms within the context of the crack-tip region, so as to develop a certain predictive capability, based on sound physical concepts, for the macroscopically measured fracture mechanics parameters and trends.

By tradition, the chapters in this Materials Science Seminar volume have a tutorial bent, with a substantial review component by the specialists in each of the selected areas. This provides the publication a certain textbook character and will be of lasting value to both researchers and graduate students.

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INTRODUCTION

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Recent Advances in Fracture Mechanics

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Abstract

Some recent developments in theoretical fracture mechanics are reviewed to develop an understanding of yielding fracture mechanics. The approach is based on the J-integral. Issues concerning the path independence of J in elastic-plastic materials, the role of J as a dominant crack tip parameter, and the extent to which J is valid during crack growth are addressed. Simple methods for estimating J also are discussed.

INTRODUCTION

The theoretical issues reviewed in this paper were developed over a period of about 10 years up to about 1982. In that sense, they are not really very recent. However, it was considered desirable to review J-integral based elastic-plastic fracture mechanics in this paper. There have been important developments in fracture mechanics since 1983, but they are not covered in this paper.

OVERVIEW

A starting point for understanding the utility of the J-integral of Rice[1] as a fracture criterion is small-scale yielding. In that case:

$$J = (1 - \nu^2)K_I^2/E \tag{1}$$

where K_I is the mode I (tensile opening) linear elastic stress-intensity factor (throughout the paper, we will be concerned only with the mode I situation), E is Young's modulus, and ν is Poisson's ratio. Thus, if K_I is a fracture criterion, then so is J. In small-scale yielding, any plastic zone at the crack tip is very small compared to any specimen dimensions. Such a plastic zone is embedded in material behaving elastically, and all stresses and deformations in the zone are controlled by the elastic field. The elastic field is, in turn, characterized by the stress-intensity factor in the region outside the plastic zone where the near singular stress field is dominant. Thus, K_I is a characterizing parameter and, due to Eq 1, so is the J-integral.

The J-integral has the general definition in planar problems:[1]

$$J = \int_\Gamma W dx_2 - n_i \sigma_{ij} \frac{\partial u_j}{\partial x_1} ds \tag{2}$$

where Γ is a contour around the crack tip from the bottom surface to the top. The contour is traversed in an anti-clockwise sense. The coordinate system is such that x_1 is aligned with the straight crack. The strain energy density W is such that:

$$W = \int_o^\varepsilon \sigma_{ij} d\varepsilon_{ij} \tag{3}$$

$\mathbf{n}$ is the outward pointing unit normal to the contour Γ; σ is the stress; $\boldsymbol{\varepsilon}$ is the strain; $\mathbf{u}$ is the displacement; and ds is an infinitesimal element of arc length on the contour. An equivalent definition for finite strains exists given by Eshelby.[2] For elastic materials, linear or nonlinear, J is path independent.

In particular, an isotropic power-law elastic material has a constitutive law:

$$e_{ij}/\bar{\varepsilon} = 3s_{ij}/2\bar{\sigma} \tag{4}$$

and

$$\bar{\varepsilon}/\varepsilon_o = (\bar{\sigma}/\sigma_o)^n \tag{5}$$

where the deviatoric strain

$$e_{ij} = \varepsilon_{ij} - \frac{1}{3}\delta_{ij}\varepsilon_{kk} \tag{6}$$

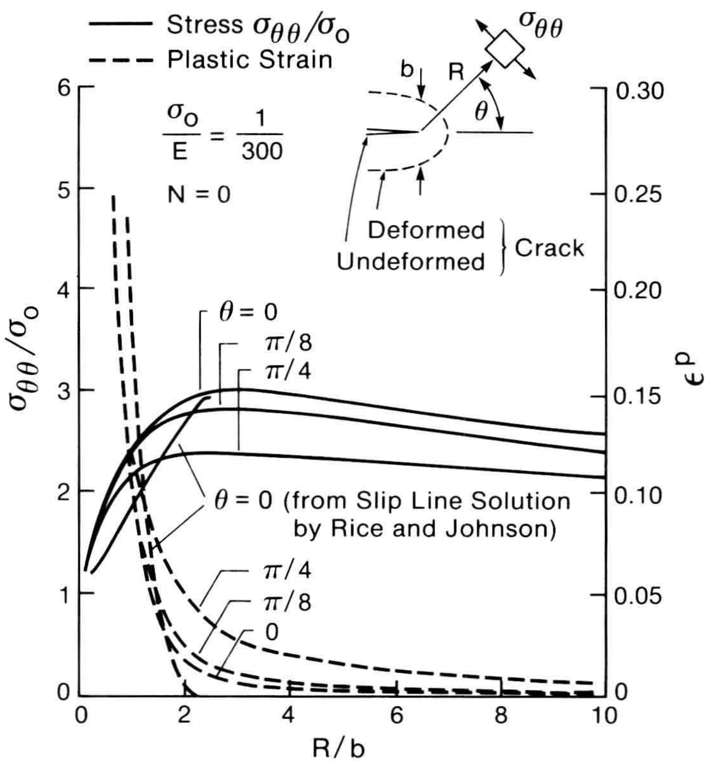

Fig. 1 Stresses and tensile equivalent plastic strains near a blunting crack tip in plane-strain small-scale yielding in an elastic perfectly plastic material.

$\boldsymbol{\varepsilon}$ is the infinitesimal strain tensor; δ_{ij} is the Kronecker delta; and s_{ij} is the deviatoric stress:

$$s_{ij} = \sigma_{ij} - \frac{1}{3}\delta_{ij}\sigma_{kk} \tag{7}$$

The tensile equivalent stress is defined as:

$$\bar{\sigma} = \sqrt{\frac{3}{2}\sigma_{ij}\sigma_{ij}} \tag{8}$$

The tensile equivalent strain is:

$$\bar{\varepsilon} = \sqrt{\frac{2}{3}e_{ij}e_{ij}} \tag{9}$$

and $\varepsilon_o\sigma_o^n$ and n are material parameters. The volumetric part of the strain ε_{kk} is linear with the hydrostatic stress $\sigma_{kk}/3$ in the material of interest. As shown by Rice and Rosengren[3] and by Hutchinson[4] by different means, a consequence of the path independence of J is that the stresses near the crack tip for this material (in the planar situation of interest) are:

$$\sigma_{ij} = (J/R)^{1/(n+1)}\tilde{\sigma}_{ij}(\Theta,n) \tag{10}$$

where R and Θ are cylindrical polar coordinates measured from the crack tip, as shown in the inset of Fig. 1, and $\tilde{\sigma}$ are functions given in Ref 3 and 4. Such a tip field is known as the HRR singularity. It is thus clear that, because stress in the material around the crack is determined by J, then J should control fracture directly. Equivalent fields for stress-strain curves of a non-power-law character can be developed to achieve the same effect, but the evidently simple nature of the power-law fields make them extremely useful as a conceptual tool.

Clearly, elastic-plastic materials are not elastic in that the constitutive law is history dependent with:

$$\dot{\varepsilon}_{ij}^{p}/\dot{\bar{\varepsilon}}^{p} = 3s_{ij}/2\bar{\sigma} \tag{11}$$

replacing Eq 4 in the isotropic case with von Mises yielding. Here, the plastic strain rate is $\dot{\varepsilon}^p$ and $\dot{\bar{\varepsilon}}^p$is the tensile equivalent plastic strain rate such that:

$$\dot{\bar{\varepsilon}}^{p} = \sqrt{\frac{2}{3}\,\dot{\varepsilon}_{ij}^{p}\dot{\varepsilon}_{ij}^{p}} \tag{12}$$

The plastic deformation is incompressible and the total deviatoric strain rate is the sum of the plastic strain rate and an elastic deviatoric strain rate proportional to and coaxial with the deviatoric stress rate. In addition to the history dependence of the plasticity law, the behavior is irreversible in that Eq 11 applies only when the plastic work $\sigma_{ij}\dot{\varepsilon}^p_{ij}$ is positive. If the plastic work cannot be positive, then $\dot{\varepsilon}^p = 0$, and the response (unloading) is purely elastic. An additional constraint, implicit in the foregoing, is that the stress lies on the yield surface when plasticity is taking place. In the case considered above, the yield surface is the von Mises criterion that requires that the tensile equivalent stress equals the value for it determined by the strain-hardening law like Eq 5.

Because the elastic-plastic material is history dependent and irreversible, J cannot be guaranteed to be path independent. However, in a situation of monotonic, proportional loading, the elastic-plastic material is indistinguishable in its response from the equivalent nonlinear elastic material, known as the deformation plasticity theory material.[5] As Rice[1] has pointed out, this situation prevails ahead of a nongrowing crack in a specimen loaded proportionally and monotonically. Thus, the deformation theory results, being for nonlinear elasticity, can be used for the crack tip fields, and the HRR solution (Eq 10) is valid near the crack tip for the power-law hardening material.

The foregoing discussion of crack tip fields for elastic-plastic materials made no mention of small-scale versus large-scale or general yielding. The HRR theory is independent of this. In small-scale yielding, J still retains its value of equivalence to K_I, as in Eq 1, with K_I generally being computed from the loads applied to the specimen. However, in large-scale and general yielding, where the plastic zone is comparable in size to the specimen dimensions and fully covers a specimen ligament, respectively, J becomes the primary independent variable for characterizing the crack tip fields. It is not a trivial task to compute J from the loads in large-scale or general yielding, as is documented by Shih and Needleman[6] and Parks, Kumar, and Shih.[7] However, all of these aspects of the characterizing behavior of J discussed above form the foundation for the theory that initiation of crack growth is controlled by J. The question of continued growth, however, requires a more elaborate treatment.

Begley and Landes,[8,9] and independently Broberg,[10] proposed that the J-integral could be used as a fracture criterion in large-scale and general yielding. In addition, Begley and Landes[8,9] provided experimental evidence suggesting that crack growth initiation could be correlated with the attainment of a critical J value. This evidence came in the form of plots of J versus the length of crack growth that had taken place in center-cracked panels and compact tension specimens of a high-strength steel.[11] Although the data for the two different specimens had different slopes, when they were extrapolated back to zero crack growth, the results coincide, indicating a unique value of J for initiation of crack growth. The subsequent divergence of the data shows that there is a breakdown of J-dominance of the crack tip field as the crack grows.

The J-dominance issue is most plainly stated in the observations of McClintock,[12] who noted the widely divergent crack tip stress and deformation states in non-hardening, fully plastic plane-strain fields computed by slip lines. For example, deep double-edged notched tension and edge-cracked bend specimens exhibit high triaxial tension on the plane ahead of the crack, whereas the center-cracked panel develops no such triaxiality and straight slip lines at 45° to the crack plane proceed from the tip to free surfaces. McClintock[12] noted that these differences would show up in greatly different crack growth characteristics in the two types of geometry.

However, it is usually argued, following Begley and Landes,[8,9] that strain hardening will enforce the uniqueness characteristic of the HRR[3,4] field and, consequently, uniqueness of the crack growth behavior. In contradiction of this, the later data of Begley and Landes[11] show that this is not always true. J-dominance in strain-hardening materials has been investigated theoretically by McMeeking and Parks,[13] who show that unconstrained configurations like center-cracked panels can lose J-dominance even when substantial strain hardening is present in the material. They also suggested limits for the geometry that would ensure J-dominance in constrained and unconstrained configurations.

Following the pioneering work of Begley and Landes,[8,9,11] Paris and coworkers[14-18] have built up an empirical and theoretical foundation for J-controlled crack growth in large-scale yielding based on the J-resistance curve (the R-curve). This curve is a plot of J versus the length of crack growth that has occurred. However, the data must be collected from J-dominated configurations such as compact tension or bending specimens. Paris *et al.* have shown that the data are unique for a given metal if the J-dominance limits are respected and the specimen is tested in a state in which its response is stable. J-dominance is confined to a limited amount of crack growth and runs out after excessive propagation.

This phenomenon is rationalized in a model due to Hutchinson and Paris,[14] where the crack tip stresses are considered to change due to increments of J and due to changes in R and Θ in Eq 10 caused by changes of the origin position tied to the moving crack tip. As long as J rises fast enough with respect to crack length, the stresses change proportionally at a material point in a way dominated by J and preserving its characterizing features. If dJ/da (where a is crack length) is not large enough, then the nonproportional changes in stress become important to the detriment of the validity of J as a fracture criterion. Thus, it is apparent that the slope dJ/da of the data—usually recast into the tearing modulus,[15] $T = (E/\sigma^2_o)\, dJ/da$, where σ_o is the yield stress in tension—is highly significant to the J-dominance issue.

The tearing modulus is also crucial to the stability of the crack growth. This behavior is analogous to the stability of crack growth in small-scale yielding in plane stress. The resistance curve for this behavior was first introduced by Irwin in terms of K[19] and adapted for J-methods by Paris *et al.*[14-18] The idea is that the material has a unique response characterized by the material resistance curve, for our purposes J versus crack length. The manner in which the specimen or component is loaded provides a relationship between J and crack length determined by the mechanics of the problem. For example, if dead loads are employed, J will generally rise as the crack grows at fixed load according to the mechanics. If displacement control is employed, J will generally fall as the crack grows under total constraint according to the mechanics. Of course, crack growth occurs when the applied J inferred through the mechanics is equal to the critical J (a function of the amount of crack growth known as J_R). However, an instability in the crack growth will occur if the applied J tends to exceed J_R as growth continues. That is, if the tearing modulus according to the mechanics tends to become greater than the tearing modulus of the material, then the crack growth will destabilize to rapid propagation.

Based on the technology described above, a standard for toughness testing in terms of J has been developed[20] and a potential design method against crack growth in large-scale yielding proposed.[21] The analysis involved relies heavily on approximate estimation techniques, which obviate the need to compute line integrals like Eq 2. These techniques rely on the fact that the compliance definition of J:

$$J = -dU/da \tag{13}$$

where U is the area under the load displacement curve, is entirely equivalent to the line integral form.[1] Consequently, J can always be converted to a formula involving the area under the load deflection curve, which is then the only quantity that need be measured or estimated for the purposes of evaluating J. These methods were first proposed for strain-hardening materials by Rice, Paris, and Merkle.[22] They have been generalized by McMeeking.[23]

With this development, the foundation for J-based elastic-plastic fracture mechanics is substantially complete. The ideas have found use in the testing of extremely tough alloys, where it has allowed engineers to abandon the enormous specimens required for valid plane-strain fracture toughness (K_{IC}) testing.[24] Less common is the use of the methodology for design, because conservative standards for initial flaw sizes and fatigue are enforced in any case.[25]

However, it is likely that the design activity based on J-methods will increase as more efficient and higher performance components will be required in the future. At the same time, improved models for crack propagation must be developed to draw the methodology and the more exact theories closer together. Examples of this include the work of Rice, Drugan, and Sham,[26] in which the mechanics of the near-tip elastic-plastic deformation near a growing crack tip is rigorously treated. At the time of this writing, plane-strain and plane-stress growing crack problems with perfect plasticity are solved for the asymptotic tip behavior for the small-scale yielding case.[27] However, there is no consensus as to the validity of proposed asymptotic solutions for strain hardening.[28] Large-scale or general yielding asymptotic solutions[27] are currently becoming available, otherwise only finite-element approximations are available to inform in these cases.[29-32]

A more extensive and complete review of elastic-plastic fracture mechanics is available in Kanninen and Popelar.[33] At this stage, we will proceed to discuss some particular theoretical issues concerning elastic-plastic fracture mechanics.

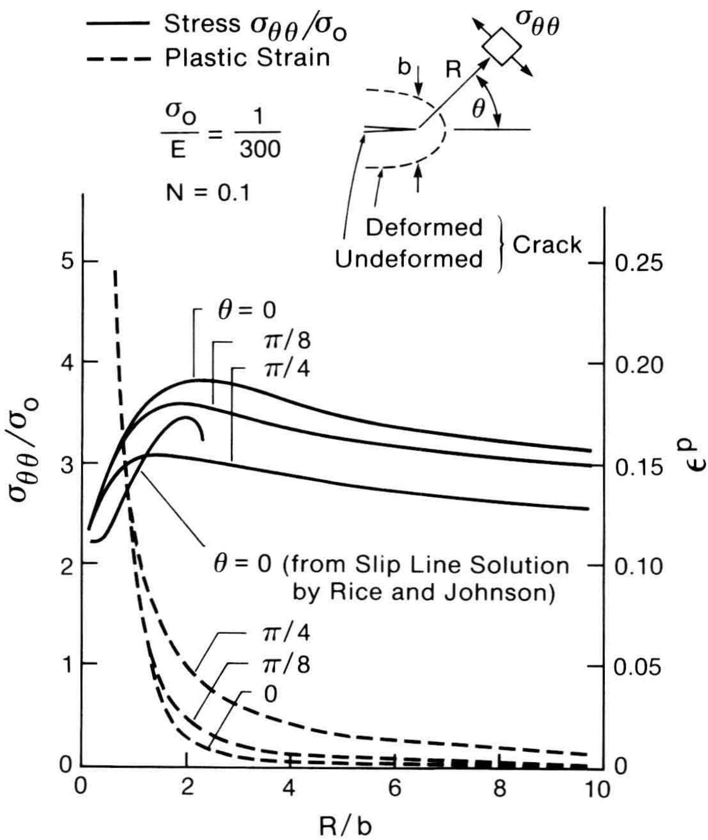

Fig. 2 Stresses and tensile equivalent plastic strains near a blunting crack tip in plane-strain small-scale yielding in an elastic-plastic material with strain hardening.

PATH INDEPENDENCE OF J

As noted above, J is path independent for nonlinear elastic materials and, by analogy, for elastic-plastic materials with nongrowing cracks loaded in a proportional manner. Indeed, Rice and Tracey[34] confirmed this through the use of finite-element calculations and demonstrated the presence of the HRR fields in their solutions as well. However, these calculations were carried out for mathematically sharp cracks, and a question arises concerning the effect of crack-tip blunting on the path independence of J. Crack-tip blunting causes distinct nonproportional straining near the crack tip, as was shown by Rice and Johnson[35] using slip-line theory for perfect plasticity. This nonproportionality causes deviations from the conditions necessary to ensure path independence of J.

McMeeking[36] carried out further investigations on crack-tip blunting. He used large deformation finite-element calculations for this purpose and thus was able to solve the strain-hardening problem, as well as the perfectly plastic case addressed previously by Rice and Johnson.[35] In small-scale yielding, the near crack tip fields have the form shown in Fig. 1 for perfect plasticity. Beyond about two current crack-tip opening displacements, the plastic strains are quite small, less than 5%. They rise rapidly to the crack tip and, for an initially mathematically sharp crack, the strains will actually reach infinity. The stresses are limited on the surface of the crack due to the traction-free condition there and also due to the yield condition for the perfectly plastic material.

However, equilibrium along with the state of deformation requires that the stresses rise from the crack surface until a peak level is reached. This elevation of the stress is due to elastic dilatation giving rise to high hydrostatic stresses; the deviatoric stresses are still limited by the perfectly plastic yield condition. The peak level of triaxiality, which involves tensile stresses three times the uniaxial yield stress, is associated with the constraint of a Prandtl punch-type slip-line field.[5] Beyond the peak stress location, the stresses fall off to unconstrained levels near the plastic zone boundary. For comparison with the finite-element solutions of McMeeking,[36] the results of the slip-line analysis of Rice and Johnson[35] are shown in Fig. 1, up to the peak stress level, which is essentially the limit of their validity.

The effect of crack-tip blunting in a strain-hardening material is very similar to the perfect plasticity case. An example is shown in Fig. 2 for a strain-hardening exponent $N(= 1/n)$ of 0.1. The plastic strains are essentially unchanged from the perfectly plastic case, which reflects the very constrained type of deformation that takes place as the crack tip blunts. On the other hand, the peak stresses are higher than in the perfectly plastic case because of the effect of strain hardening over and above the elevation of hydrostatic stress.

It is not apparent from the finite-element results of Fig. 2, but an implication is that, for an initially mathematically sharp crack, there would be an upturn to very high stresses near the surface of the blunt crack. This would arise because of the hardening resulting from the extremely large strains at the surface of the blunt tip. As before, the predictions of Rice and Johnson[35] are shown for comparison. The tendency to high stresses very near the crack tip surface is apparent in the Rice and Johnson calculations.

The contrast between the HRR field for a mathematically sharp crack and the blunting solutions is self-evident. In the HRR field, the stresses are