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TECHNICAL PROGRESS REPORT

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"A Combined Macroscopic and Microscopic Approach to the Fracture of Metals"

Period: July 1977 - June 1978

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EY-76-S-02-3084 June 1978
Technical Progress Report

CONTENTS

EY-76-S-02-3084 June 1978

Technical Progress Report

Abstract

Accomplishments during the past year are reviewed. The work reported includes studies on: (1) The role of particles and interfaces in the initiation of fracture, including fundamentals of brittle versus ductile response of interfaces and observations on cavity growth by the cracking of grain or sub-grain boundaries adjacent to carbides in spheroidized steels; (2) Environment sensitive fracture mechanisms, particularly the effect of hydrogen in reducing tensile ductility by acceleration of the crack-like mode of cavity growth along grain boundaries in steels; (3) Models for elevated temperature diffusive processes of cavity growth on grain interfaces, including non-equilibrium effects and crack-like growth modes; (4) Localization of plastic deformation and the inception of ductile rupture; and (5) Elastic-plastic stress analysis, by finite elements, of growing cracks and examination of criteria for stable crack growth.

EY-76-S-02-3084
Technical Progress Report

June 1978

A. Detailed Description of Research Program

1. Role of Particles and Interfaces in the Initiation of Fracture

1a. Void and Crack Initiation in Steels

In the past several years we have focused a good deal of theoretical and experimental effort on studying problems related to the fracture of interfaces in particular those associated with particles. This special interest is motivated by the now experimentally established fact that essentially all fractures, both ductile and brittle, initiate at interfaces of some sort. But the process of void or crack initiation at particles or interfaces is not well understood. Thus our continuing research on void and crack initiation in carbon steels remains an important part of the overall fracture program, with a main objective being to relate the macro- and micro-mechanical conditions leading to crack initiation and propagation to microstructure, in particular to the type, size and distribution of second phase particles and to the morphology of grain or sub-grain boundaries.

Most of our recent efforts have been concerned with studying void initiation in plain carbon spheroidized steels. Previous observations on uniaxially deformed specimens revealed three main modes of void initiation, namely 1) transparticle cracking, 2) cavitation of matrix ligaments between nearly contiguous particles, and 3) particle-matrix decohesion. Our results, which we have now obtained on a wide range of steels with carbon contents between 0.13 to 1.1 wt. percent, have shown that the predominant mode of void initiation is particle-matrix separation although particle fracture is observed and most frequently occurs when two or more particles have grown together; in this case voids initiate by the fracture of particle-particle boundaries. We have consistently noted important effects of particle size and, in particular, particle shape and distribution. However, we have recently made some striking and potentially very important new observations in our subsequent detailed studies of void growth.

The prevailing view has been that fracture in these ductile steels proceeds by plasticity controlled growth of voids followed by void coalescence (e.g., McClintock, J. Appl. Mech., 35, 1968, p. 363). Void coalescence may occur by either direct void-void impingement or by an accelerated process involving localized shear between closely spaced holes. Now Fig. 1 shows a series of scanning electron micrographs of a polished section of a uni-axially stretched axisymmetric rod of 1013 spheroidized steel; all three photographs are of the same specimen and contain the same field of view. The difference in micrographs 1a, 1b, and 1c is the degree of special care taken to avoid polishing over the network of fine cracks (esp. in Fig. 1c) which have propagated parallel to the tensile axis along the grain or possibly even sub-grain boundaries. In preparing the specimens, as seen in Figs. 1b and 1c, a procedure of intermittent polishing and ultrasonic cleaning in Alconox dissolved in water was utilized to prevent loose metal particles or burrs from filling in the fine interface cracks. These special precautions have now made it clear that an important contribution to the development of void volume fraction in steels deformed this way is the fracture of internal interfaces. We are now quantifying how the degree of interface fracture depends upon plastic strain, flow stress and the degree of specimen necking. However, the fact that void "growth" can occur in this way rather than solely by the normally assumed plasticity controlled growth may have important implications for modelling ductile failure in steels. For example, existing theoretical models, which assume that increases in metal porosity are controlled by plastic void growth, have often overestimated achievable ductility and one valid explanation for this is that the formation of bands of intensely localized shear can intervene and accelerate void coalescence (as discussed in Technical Report 57). Our recent findings now suggest that interfacial fracture, which we have shown can lead to rapid increases in porosity in certain specimens, may also cause reductions in ductility (that is, in addition to localized plastic flow). This supposition has, in fact, been partially confirmed by our recent work concerning hydrogen induced ductility losses in plain carbon spheroidized steels (section 2). Here again it was found, in 1030, 1045 and 1.1 wt% C steels, that interfacial fracture contributed directly to void development. The major influence of hydrogen was to accelerate boundary fracture, thus increasing the rate of void development, and thereby causing a reduction in the strains to fracture.

Our studies of interface fracture have also revealed numerous cases of essentailly complete delamination of particular grain boundaries. In addition, we have noted that the fracture surfaces are populated with equiaxed dimples. Thus interface fractures may indeed propagate over entire grain boundaries which become cylindrical with large strain and this may help to explain the deep but equiaxed dimpled topology of the fracture surface.

In addition to the experimental work just described we have continued our theoretical studies on the fracture and decohesion of carbide particles in ductile matrices. The work this past year has focused on computing the local strains and stress concentrations near the interfaces of partially contiguous second phase particles. Previously we had dealt with the limiting case of completely dispersed second phases (e.g., publication of Chang and Asaro and Technical Report 51).

(Staff: R. Asaro, J. Fisher, J. Gurland)

During the past year we have initiated a study of crack initiation and crack growth in high strength steels, namely 4140 quenched and tempered martensitic steels. The aim of this work is to understand the role of carbide particles formed in tempered martensites and bainites on crack initiation and early propagation and hence on material toughness. Our approach involves a detailed microscopic study of the microstructures that are produced using different austenitizing temperatures and tempering treatments and of the mechanisms of crack initiation in plane strain.

(Staff: R. Asaro, J. Gurland, T. Odegaard (MRL supported student))

EY-76-S-02-3084 June 1978
Technical Progress Report

b. Segregant Induced Embrittlement of Internal Interfaces

There are many cases where normally ductile metals can undergo brittle intergranular fractures caused by interfacial segregation of residual impurities or of environmental elements like hydrogen. In large part these represent instances where segregation induced cohesion losses were sufficient to invoke a brittle response of the boundaries where normally they would be ductile. In Technical Report 40 Rice presented a model for the brittle versus ductile response of an interface which contains an initially sharp crack. This model was further developed in Technical Report 52. Essentially the models pose the following question: when an interface crack is loaded and the stress intensity increases, what criterion is met first, a K corresponding to the Griffith value, K_G (implying that the energy deliverable to the advancing crack tip equals that stored in created surface; possible brittle behavior) or a K ($<K_G$) sufficient to cause athermal nucleation of a dislocation (meaning that the crack blunts; ductile behavior)? To answer the question the activation energy for dislocation nucleation at a sharp crack is computed and if it turns out positive then brittle response is possible, if it is negative crack blunting takes place. The analysis reduces the question of ductile versus brittle behavior to a comparison of two dimensionless parameters $R_o \approx (r_o/b)(\gamma_{int}/\gamma_{step})$ and $S \approx \gamma_{step}/(Gb)$. b is the Burgers vector of the dislocation and r_o is the so-called inner cutoff radius of the dislocation; G is the elastic shear modulus. $2\gamma_{int}$ and γ_{step} are the works to separate the interface on which the crack lies and to create a step at the crack tip by dislocation blunting respectively. R_o and S also depend upon the orientations of the slip planes intersecting the crack front and the directions of the Burgers vectors. For a given set of material parameters (i.e. γ_{int}, γ_{step}, b and the orientations of the slip systems) S and R_o may be computed and if

$$R_o > R_o^* = \frac{0.58}{S} e^{5S}$$

ductile behavior is predicted. Since γ_{int} and γ_{step} are both affected by the adsorption of impurities, the models allow for an approximate quantification of ductile-brittle transitions caused by solute adsorption or environmental interactions.

The above criterion has been used to model intergranular embrittlement, caused by solute adsorption, of face-centered cubic polycrystals (Technical Report 52). Material parameters, such as the γ's, were matched to the data of Hondros and McLean (Phil. Mag. 1974) for the adsorption of Bi on the free surface and grain boundaries of Cu. To account for the orientation dependence of R_o and S the crystallographic orientations of the grains with respect to the crack plane were randomly varied. But in the previous models the <u>crack line</u> was allowed unlimited rotation to take the direction of the intersection of the crack plane and slip plane. The model predicted a transition to brittle intergranular behavior, with Bi adsorption, in qualitative agreement with experiment but the results indicated a less brittle response than is actually observed. New work this past year has focused on relaxing the rotational freedom of the crack line. For a given orientation of the grain all possible slip systems are tested for ductile versus brittle behavior. Clearly if no slip system can undergo dislocation nucleation the interface is brittle for that orientation. On the other hand if the interface contains at least one "ductile" slip system, ductile response is possible providing the crack line is, or can be, alligned with the slip plane. Hence the more ductile slip systems there are the greater is the probability that ductile interface behavior will be observed. This new approach, which is partly statistical in nature, does in fact lead to the prediction of a more overall brittle behavior than our earlier model (for Bi in Cu) and thus places the model in better agreement with Hondros and McLean's experimental results.

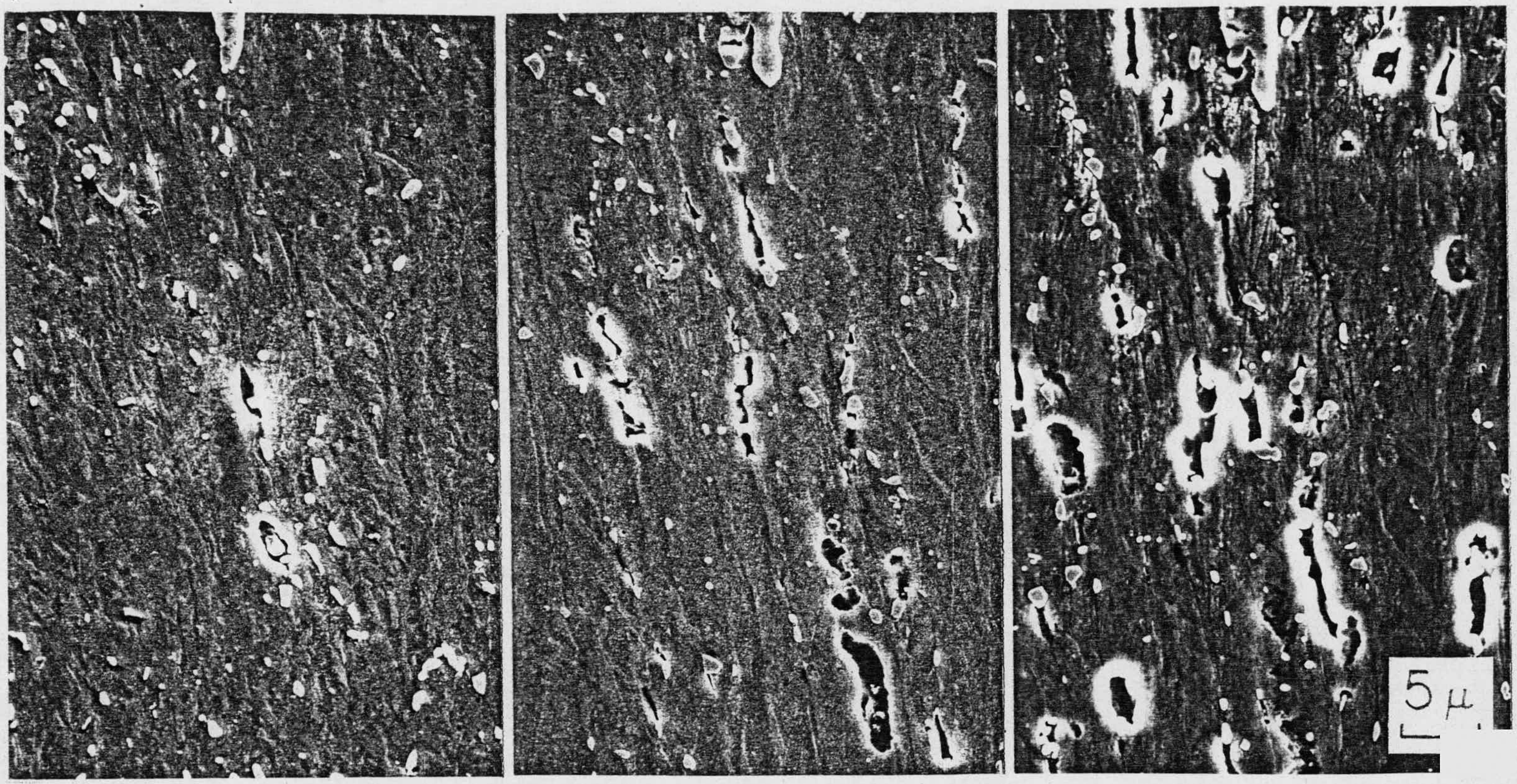
(Staff: D. Mason, J. Rice, R. Asaro)

Recent developments on the thermodynamics of interface cohesion are presented in Technical Report 54. There the models for adsorption induced cohesion losses at interfaces, begun by Rice in Technical Report 40, were extended to include multi-component adsorption. Additional discussion of the constitutive assumptions used in the models is provided along with some alternative derivations of the formulae for the work of separation at constant interface composition. According to the models, the difference between the work to fracture initially equilibrated boundaries in the two extremes where 1) no further adsorption on the separating interface takes place, and 2) where complete equilibrium between the bulk phases and the interface is allowed to occur, is

$$\gamma_{fast}(\Gamma) - \gamma_{slow}(\Gamma) = \sum_{i=1}^{n} \int_{\Gamma_{io}(\mu)}^{\Gamma_{i\infty}(\mu)} [\mu_i - \mu_{i\infty}] d\Gamma_i \quad .$$

$\Gamma_{io}(\mu)$ and $\Gamma_{i\infty}(\mu)$ are the excesses of adsorbed specie i on the interface and on the two newly formed free surfaces respectively. Γ_i and $\Gamma_{i\infty}$ are the chemical potentials of the i component in the bulk phase and on the free surface respectively. The above formalism was used in Technical Report 54 to analyze the cohesion losses of grain boundaries of Fe due to segregation of P. It was found that a Langmuir isotherm gave an excellent fit to the equilibrium adsorption data of Hondros (Proc. Roy. Soc. 1965). The numerical results demonstrated that there was indeed a cohesion loss when no further adsorption took place to the free surfaces (i.e., fast separation) but that the cohesion loss was far less than when full bulk-interface equilibrium was assumed. Since most actual cases of brittle fracture occur on time scales much too short for diffusion to establish composition equilibrium, γ_{fast} should better characterize the quantitative effects of solute adsorption on the cohesion losses of interfaces.

(Staff: R. Asaro, D. Mason, J. Rice)



(a) (b) (c)

Figure 1

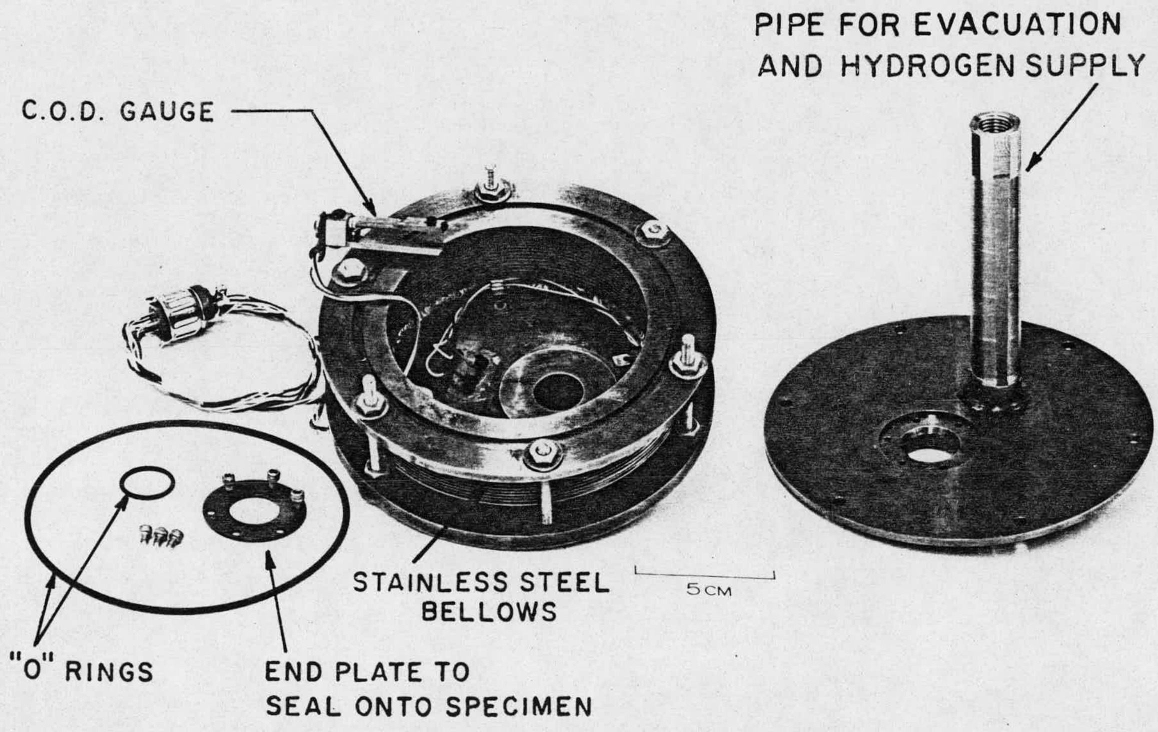


Figure 2

2. Environment-Sensitive Fracture Mechanisms

Environmentally induced or assisted fractures fall under many headings, such as "hydrogen embrittlement," "temper embrittlement," or stress corrosion cracking, but they are generally manifested in one of two ways: either there is no change in the fracture mode, with the "normal" process of fracture occurring at lower stresses and strains, or there is a transition in the fracture mode from a ductile hole initiation and growth sequence to more brittle, less energy consuming, modes such as transgranular or intergranular cleavage. The first instance is common in low and medium strength alloys when they are exposed to hydrogen containing environments and where the normal ductile mode of fracture, as would occur in the absence of an environment, is accelerated. Alternatively, it happens, especially in higher strength materials that the environment will induce a more brittle mode to set in rather early in the deformation-fracture process and thus limit the material's toughness. Our experimental program addresses both these possibilities. We have been experimentally quantifying the effect of hydrogen on ductile and brittle fracture processes as well as the conditions which lead to fracture mode transitions.

As mentioned in section 1a, since the beginning stages of most fracture processes in metals, whether it be the initiation of voids leading to ductile fracture or the initiation and growth of brittle cracks, involve separation or decohesion along internal interfaces of some sort, it is important to quantify the effect of an environment on interfacial separation. In fact the results of our recent studies on <u>ductile</u> fracture in spheroidized plain carbon steels which have shown that interfacial fractures, in particular of grain and possibly even sub-grain boundaries (Technical Report 59, publication of Cialone and Asaro and section 1a), play a major role in void development demonstrate even further the importance of understanding the effect of environments on interfacial fracture. This has been the subject of much of our theoretical study on interfacial cohesion (e.g., Technical Report 54) and ductile-brittle response (section 1b) and our experimental program is designed to be as complimentary to these theoretical efforts as possible.

Our recent progress on the influence of hydrogen on the ductile fracture of plain carbon steels is summarized in Technical Report 59 and the publication of Cialone and Asaro. The results of our studies on 1030, 1045 and 0.95 wt. percent carbon steels have shown that, while hydrogen may affect the initiation of voids by facilitating particle-matrix decohesion, the major contribution to premature <u>ductile</u> fracture (and thus to a ductility loss) is an acceleration with strain of the development of void volume fraction. Voids initiated almost exclusively by particle-matrix separation and then propagated in a crack-like fashion along grain and possibly even sub-grain boundaries, usually linking up with other voids propagating in the same manner. The mechanistics of this process have not been appreciated before but so far are in fact totally consistent with the recent findings of Gurland and Fisher (section 1a). These "interface cracks" (see Figure 1) open laterally to form larger holes which enter directly into the final fracture process. Both the propagation of voids along boundaries and their subsequent swelling to form large holes are affected by the transverse stresses imposed by the neck as well as by the plastic strain incompatibility between the plastic matrix and elastic particles (e.g., as discussed in the earlier publication of Chang and Asaro and in Technical Report 51). Hydrogen appears to lower the stresses necessary for void growth along boundaries by either an interfacial adsorption or an internal pressure mechanism, although our calculations to date indicate no large effect from internal pressure (Thesis by Cialone and Technical Report 59). Thus we have found that interfacial fracture is an important mechanism of void development in these steels in the absence of hydrogen and that the presence of hydrogen in the lattice does not change but accelerates the "normal" ductile fracture.

It is important now to explore the generality of our findings with regard to accelerated ductile fracture of other materials containing hydrogen. We have already discussed that interface fracture certainly plays a primary role in the ductile fracture of other plain carbon steels including low carbon 1013 steels (section 1a). The appropriate experimental work has been initiated on the same ultra-high strength 4140 steels used in our studies on gaseous hydrogen embrittlement.

Our experimental work on hydrogen assisted fracture in 4140 high strength steels is currently in progress. Initially the environment is purified gaseous hydrogen. We have designed and constructed a novel environmental chamber which isolates the notched section of four-point bend specimens. The chamber, shown in Fig. 2 is "O" ring sealed to the specimen and allows for tension-compression and bending by using flexible stainless steel bellows. Hydrogen is introduced into the chamber, after repeated evacuation and purging with argon, through a palladium diffusion filter. From these experiments we expect to obtain an understanding of the effects of hydrogen on (i) accelerating hole initiation in front of blunt notches and sharp cracks, (ii) accelerating the coalescence of holes perhaps through premature localized shearing and through interface fracture, (iii) inducing transitions in the fracture mode and (iv) facilitating interfacial cleavage especially when cleavage is induced by prior temper embrittlement. In order to study the effects of hydrogen on a <u>single</u> brittle mode of fracture (item iv above) we have obtained from the Bethlehem Steel Corp. several plates of a 4140 steel each with carefully controlled phosphorous levels. The alloys with high impurity levels will be heat treated to induce intergranular temper embrittlement and then tested in a hydrogen atmosphere.

(Staff: R. Asaro, H. Cialone, J. Rice)

3. Elevated Temperature Diffusive Cavitation of Grain Boundaries

In the current reporting period we have completed a study of non-equilibrium models for diffusive cavity growth along grain interfaces. The work, reported in Technical Report 60 (Chuang, Kagawa, Rice and Sills), proceeds from the observation that creep rupture cavities are frequently rather flat and crack-like, instead of having the rounded shape that would be consistent with the quasi-equilibrium surface shapes assumed in prior studies. That is, the time scale of surface diffusion is, evidently, not always sufficiently short to validate the quasi-equilibrium assumptions.

Based on estimates of characteristic relaxation times for a variety of modes for material transport from the cavity walls and for accommodation in the adjacent grains (see Appendix of Technical Report 60), it is concluded that in typical circumstances matter is carried by surface diffusion from the cavity walls and conducted along the grain interface by grain boundary diffusion, with matter joining the crystals on either side so that the grains move relative to one another. The grains can be regarded, at sufficiently low stress levels, to separate "rigidly" -- i.e., with negligible deformations. The process was modelled for two void geometries: A long cylindrical void (for which some results are available from previous work on this project by Chuang, Kagawa, and Rice) and, more realistically, an axi-symmetric void. Two limiting growth modes are established. These are: (i) Slow, quasi-equilibrium growth (as in the classical Hull-Rimmer model) for which there is adequate time for the surface to retain its equilibrium, spherical shape and in which case grain boundary diffusion is the rate-limiting process; and (ii) Rapid, crack-like growth for which surface diffusion is inadequate to maintain a near-equilibrium shape and, instead, significant mass transfer occurs only near the advancing void tip where large gradients in curvature develop. Cavity growth in this crack-like mode is rate-limited by both surface and grain boundary diffusion processes. We develop and present guidelines for interpolating between these limiting cases on the basis of characteristic relaxation times for surface diffusion and of a "self-similar" solution for the cavity shape, in which the cavity enlarges with radius a proportional to $t^{1/4}$, where t is time.

Essentially, the results can be summarized as follows. If a is cavity radius and $v = da/dt$, then the quasi-equilibrium growth mode is a good approximation when $a^3v/B < 10$, and the crack-like mode when $a^3v/B > 30$. Here B has dimensions $(\text{length})^4/(\text{time})$ and is a kind of generalized diffusivity for mass transfer by surface diffusion, given by

$$B = D_s \gamma_s \Omega^{4/3}/kT$$

where D_s is the surface diffusion coefficient, γ_s the surface free energy, Ω the atomic volume, and kT the temperature in energy-per-atom units.

The relations between the average stress σ acting across a grain boundary and the speed of cavity growth can be summarized in the following way. We consider axi-symmetric voids of diameter 2a and average center-to-center spacing 2b, and let s be the ratio of the average net stress, $\sigma/(1-a^2/b^2)$, on the uncracked portions of grain boundary to that stress which would just equilibrate voids of radius a against sintering, i.e.,

$$s = [\sigma/(1-a^2/b^2)]/[2\gamma_s \sin\Psi/a] \quad .$$

Here 2Ψ is the angle between the cavity surfaces where they meet at the cavity tip; Ψ depends on the ratio of grain boundary to surface energy and $\Psi = 75°$ is typical. Then, for cavity growth in the quasi-equilibrium mode we find that the stress and growth speed are related (for $\Psi = 75°$) by

$$s = 1 + 0.237\,\rho\,(a^3 v/B) \quad ,$$

and for growth in the crack-like mode by

$$s = 0.630\,(a^3 v/B)^{1/3} + 0.420\,\rho\,(a^3 v/B)^{2/3} \quad ,$$

where

$$\rho = \frac{D_s\,\Omega^{1/3}}{D_b\,\delta_b}\;\frac{(1-a^2/b^2)Q}{a/b} \quad .$$

Here $D_b\,\delta_b$ is the grain boundary diffusion coefficient and Q is a function of a/b which varies little for $0.1 < a/b < 1.0$ and has an average value of about 0.6 in that range. The guidelines noted earlier for quasi-equilibrium vs. crack-like growth suggest that growth can be considered to follow the quasi-equilibrium relation between s and v above when

$$s < 1.5 + 3.2\,\rho$$

and the crack-like relation when

$$s > 1.5 + 3.2\,\rho$$

Thus the quasi-equilibrium mode is favored when stress level σ is low, when D_s is large compared to D_b, and when the cavity radius a is small; growth in the crack-like mode is favored in opposite circumstances.

The most direct check of the theoretical model is provided by a recent bi-crystal study on Cu by Raj (Acta Met., 1978). From the observed rupture times and center-to-center spacings of cavities, it is possible to calculate representative values of v and of the parameter a^3v/B. Of the 6 specimens reported by Raj to fail by diffusive cavitation, 4 have values of a^3v/B, evaluated for $a = b/2$, which are of the order 10^3 or larger and hence are well into the crack-like growth regime. All of these specimens have observed rupture times which agree well with the theoretical prediction of rupture time based on the crack-like model; see Technical Report 60 for further details. In addition, a major study by Goods and Nix (Acta Met., 1978) on the creep rupture of Ag polycrystals with water-vapor filled cavities on their grain boundaries has been reported to agree well with the predictions, privately communicated, of our crack-like growth model. Here there remains a major uncertainty because D_b would have to be much larger than its accepted value for Ag to validate the version of the crack-like model which Goods and Nix use, but the stress, temperature, and microstructural size dependences predicted by the model appear to fit their data well.

(Staff: J. R. Rice, L. Sills (MRL Postdoctoral Fellow), and former Research Assistants T-j. Chuang (Ph.D., 1975) and K. I. Kagawa (Sc.M., 1976))

4. Localization of Plastic Deformation and Inception of Ductile Rupture

The localization of plastic deformation into an intense shear band is a commonly observed precursor to ductile rupture. In some cases the localization seems to result as a consequence of progressive weakening of the material due to crack and cavity nucleation. In others there seems to be some inherent instability of the plastic flow process itself which leads to localized flow.

Both sources of instability have received attention in recent studies on this project. For example, Asaro and Rice (J. Mech. Phys. Solids, 25, 1977, p. 309) formulated conditions for localization instabilities in ductile single crystals due to deviations from Schmid's rule associated with cross-slip. Also, Yamamoto (Int. J. Fracture, 1978, in press) has formulated conditions for localization in void-containing, but otherwise stable, ductile materials. Here the localization is a result of progressive softening through hole growth but, as Yamamoto showed, the strain to instability is greatly reduced by small deviations from uniformity of material properties, especially deviations having a geometric distribution that is aligned relative to possible planes of localization.

Needleman and Rice have given a general review of localization theory in Technical Report 57, both in the 3-dimensional context appropriate to shear bands, and in a two-dimensional context for the localized necks which limit formability of ductile metal sheets. Several new results on localization are presented. For background concerning these it should first be noted that an area of major uncertainty in macroscopic plastic constitutive relations is that of whether "vertex-like" effects exist on subsequent yield surfaces. Such vertex effects cause a reduced stiffness of material response for stress increments directed tangentially, for example, to what is regarded as the yield surface in classical Mises-type isotropic strain hardening formulations of plasticity theory. Models for polycrystalline plasticity, based on Schmid-like characterization of conditions for plastic flow of each single crystalline grain of a polycrystal, lead universally to the prediction that vertices develop on subsequent yield surfaces; but experimental studies have thus far been far less definitive as to whether vertices exist and how significant are their effects.

In view of this background, Needleman and Rice adopted a constitutive model for vertex yielding based on a finite-strain generalization of "deformation" plasticity (simple slip models of plasticity lead to vertices on yield surfaces and also, for a wide class of stress paths referred to as "fully active," to path-independent relations between stress and strain which are the hallmark of "deformation," versus "flow," plasticity formulations). They derive and compare predictions for the strain to localization instability under conditions of the axi-symmetric tension test and the plane-strain tension test, and show that the theoretical predictions correspond at least approximately with results of such tests by Clausing (Int. J. Fracture Mech., 6, 1970, p. 71) for seven structural steels. Features in common are that the plane-strain ductility is reduced in comparison to the axi-symmetric value, and the amount of reduction is greater for smaller values of the strain hardening exponent (typically, small strain hardening exponents are associated with high strength levels, and conversely; Clausing correlated ductility in terms of yield strength level). For example, for the simplest version of the vertex yielding model, Needleman and Rice found the ratio of greatest principal logarithmic strains at localization to be

$$\frac{(\varepsilon_I)_{\text{critical, plane strain}}}{(\varepsilon_I)_{\text{critical, axi-sym. strain}}} = \sqrt{\frac{3N}{1+3N}} \quad ,$$

for $0 < N < 1/3$, where N is the hardening exponent in a relation of the form $\sigma \propto \varepsilon^N$ between true stress and logarithmic strain in the plastic range. Also, both the theory and the experiments summarized by Clausing have in common the feature that the axi-symmetric ductility is little affected by changes in plastic flow properties, whereas the plane-strain ductility is widely variable. Further details are in Technical Report 57.

Needleman and Rice also derived localization conditions based on the approximate constitutive relations given by Gurson, in earlier studies on this project, for void-containing materials. A new development is that criteria for nucleation of voids from indlusions were included in the localization analysis. Particularly, it was shown that when the void nucleation criterion is dependent on the maximum tensile stress, as proposed in studies by Argon and co-workers (Met. Trans. 6A, 1975, p. 815, 825, 839), then localization is promoted when there is high stress triaxiality, e.g., in front of a sharp notch, and when a significant fraction of the inclusions fail over a comparitively narrow range of stress. Again, Technical Report 57 may be consulted for fuller details.

A study now approaching completion by Tin, as the basis for an Sc.M. thesis, examines conditions for localization in the region adjacent to a progressively blunting, plane-strain, crack tip. Tin uses previous analyses on this project by Rice and Johnson (Inelastic Behavior of Solids, McGraw-Hill, 1970, p. 641) and McMeeking (J. Mech. Phys. Solids, 25, 1977, p. 357) to estimate the plastic strain and stress triaxiality at points near the crack tip. Hence, by using analyses like those discussed above by Needleman and Rice, he is able to determine the extent of the near-crack-tip zone in the material over which critical localization conditions are reached. Thus far, an attempt to fit the model discussed above, based on stress-dependent void nucleation from inclusions, to data of Low and co-workers on 18 Ni-Maraging and 4340 steels suggests that the nucleation model cannot explain near tip shear localization in these materials. In particular, the predicted critical levels to which the strain-hardening modulus must fall for instability, as estimated from the model and data on the rate of hole nucleation, seem to be lower than the value of the hardening modulus actually anticipated, based on McMeeking's estimates of plastic strain levels and the tensile stress-strain relation for the material. What seems to be somewhat more promising, however, is the prediction of localization conditions based on the vertex yielding model, also discussed above. The study is still in progress.

(Staff: R. J. Asaro, A. Needleman, J. R. Rice, and A. Tin)

5. Macro-mechanics of Crack Growth in Ductile Metals

a. Stress and Deformation Analysis for Stably Growing Cracks

In ductile metals such as pressure vessel steels, it is typical that the onset of crack growth does not coincide with unstable fracture. Rather, cracks are observed to grow stably under increasing load, or load-point displacement, until some terminal instability is reached. The deformation which must be imposed on the cracked element to reach instability can be very much larger than that for the onset of growth. Further, whereas the onset can be characterized in terms of parameters like J_{IC}, which are reasonably independent of specimen geometry, the final instability condition is strongly dependent on specimen type and on the elastic compliance of the loading system.

Our approach to the problem of stable crack growth is at the level of macroscopic elastic-plastic stress analysis. The goal is to describe accurately the stress and deformation fields near a growing crack tip and, from this description, to arrive at a physically viable characterization of the growth process.

The approach taken in our work (e.g., Technical Reports 55 and 56) has involved a combination of asymptotic studies of the form of the near-tip stress and deformation fields, and large scale numerical finite-element studies of crack growth. The asymptotic studies have been carried out based on the elastic-perfectly plastic material model and consider crack growth under plane-strain conditions. The near-tip stress field is presumed to be given by the Prandtl slip line construction and, when this stress field is moved through the material with a growing crack tip, non-compatible elastic strain alterations occur and induce singular plastic strain increments of logarithmic character.

As a principal result, e.g., from Technical Report 56, it was shown that the rate of change of displacement δ between upper and lower crack surfaces at a small (compared to plastic zone size) distance r behind the crack tip has the form

$$\dot{\delta} = \frac{\alpha}{\sigma_o} \dot{J} + \beta \frac{\sigma_o}{E} \dot{\ell} \log \frac{R}{r} \quad .$$

Here J is the far-field value of the J integral, σ_o the tensile yield strength, E the tensile elastic modulus, ℓ is crack length, and β is a pure number (3.93 when Poisson's ratio $\nu = 0.3$). The parameters α and R cannot be determined by the asymptotic analysis and for this purpose we have attempted to correlate the theoretical expression against numerical results from finite-element studies of growing cracks (e.g., Technical Report 55, and Ph.D. Thesis by E. P. Sorensen).

The finite-element mesh is such that the crack can be advanced only element-by-element. Also, for economic reasons, the plastic zone was not caused to grow beyond a maximum radius of approximately 10 element dimensions, and this leaves considerable uncertainty in the interpretation, since the above theoretical result is only asymptotically valid very near the tip, well within the plastic zone. Nevertheless, by matching numerical results for crack surface opening to the theoretical expression, for various combinations of increases in far-field J followed by increases in crack length, Rice and Sorensen found that for well-contained plastic yielding the parameter α is constant to within about 7% (and hence has the same value as for monotonic load increase on a stationary crack), and the length parameter R seems to scale with the plastic zone size. In particular, a tentative fit to the numerical results for small scale yielding is that

$$R \simeq 0.16 \; E \; J/\sigma_o^{\;2} \quad ,$$

which coincides closely with the maximum plastic zone radius.

Work in progress by S. Sham, in collaboration with Rice and Needleman, has the aim of performing a definitive numerical analysis of crack growth under small scale yielding conditions. As of this date, the progress has consisted of code re-writing as necessary to change from the present program, written by Sorensen for IBM machines, to a program which can be run on a CDC machine made available for use through collaboration with Dr. C. F. Shih at General Electric Research Laboratories. The aim is to grow the plastic zone to a considerably larger size, with maximum radius encompassing approximately 50 element dimensions or more, before various increments in crack length and load are imposed to identify the parameters α and R in what we hope will be an unambiguous manner.

In addition to its obvious contribution for understanding crack growth under rising load, the highly detailed description of the near-tip stress and strain field resulting from this solution should also provide useful input to formulations of environmentally assisted crack growth. This has, for example, a close connection with the study of hydrogen embrittlement discussed earlier.

(Staff: A. Needleman, J. R. Rice, E. P. Sorensen, and S. Sham)

b. Criterion for Stable Crack Growth

Based on the representation of crack surface openings given earlier, Rice and Sorensen (Technical Report 56; also, J. Mech. Phys. Solids, 1978, in press) have proposed a model criterion for crack growth. This criterion requires that at a distance $\Delta\ell$ behind the growing tip, where $\Delta\ell$ is considered characteristic of the material and represents a measure of the fracture process zone size, the crack surfaces must be opened to a critical displacement δ. Indeed, the ratio $\phi \equiv \delta/\Delta\ell$ defines a kind of critical crack opening angle, namely a "secant" angle associated with the material distance $\Delta\ell$, required for growth. In terms of ϕ and $\Delta\ell$ the crack growth process is described by

$$\frac{\alpha}{\sigma_o}\,\frac{dJ}{d\ell} = \phi - \beta\,\frac{\sigma_o}{E}\log\frac{eR}{\Delta\ell}$$

where e is the Naperian logarithm base.

When the J value for onset of growth is specified, namely as J_{IC}, and R is expressed as a function of J (e.g., as previously for the case of well-contained yielding), this expression becomes a differential equation describing the relation between J and the increase of crack length ℓ as required for stable growth. Instability under increasing load occurs when an amount of growth is reached at which the parameter describing the load intensity must stop increasing, and subsequently decrease, in order to meet the crack growth criterion. As remarked, this instability point is strongly dependent on specimen geometry and the nature of the loading system.

As has been shown in recent work by P. C. Paris and co-workers (Wash. Univ. report to Nucl. Reg. Comm., 1977) a useful dimensionless measure of the resistance to crack instability is

$$T \equiv (E/\sigma_o^{\,2})\, dJ/d\ell \quad .$$

By using the crack growth criterion outlined above, and identifying the process zone $\Delta\ell$ with the crack tip opening, $\alpha\, J_{IC}/\sigma_o$, at the onset of growth, the initial value of T for crack growth under well-contained yielding conditions is (with $\alpha \simeq 0.5$)

$$T \simeq 2.0\,\frac{E}{\sigma_o}\,\phi - 7.9\log\left(0.87\,\frac{E}{\sigma_o}\right) \quad .$$

To evaluate this initial T value and other parameters descriptive of crack growth with small scale yielding, Rice and Sorensen (J. Mech. Phys. Solids, 1978, in press) made use of data on pressure-vessel steels tested in small, fully plastic bend specimens (Clark, Soudani, Furguson, Smith and Knott, Inst. Mech. Engrs. preprint, 1977), for which the opening displacement at the original crack tip was measured as a function of the amount of stable growth, over growth distances on the order of 1 mm or less. The data is approximately linear over this range and its slope provides a value for ϕ. In this way, T values were found to range from 100 to 550, depending on details of heat treatment, for the A533B reactor pressure vessel steel, to have a value of 174 for a particular treatment of HY-80 but only 10 after the same HY-80 is subjected to 20% pre-strain, to have the value 15 for the higher strength HY-130, and to be negative for certain weld deposits. In general, it was suggested that fracture in practical sized specimens or structural configurations is not possible for T values of the order 50 or more, but quite possible for values in the range of 10 or less.

(Staff: J. R. Rice and E. P. Sorensen)

c. Fully Plastic Crack Analysis

Needleman and Shih (Technical Report 58; also, Comp. Meth. Appl. Mech. Engr., 1978, in press) have developed a new finite element formulation for non-linear plane-strain problems of incompressible material behavior. The formulation is developed principally with rigid-plastic analysis of fully plastic flow fields in mind. The technique is also applicable to problems of non-linear viscous deformation as in steady-state creep at elevated temperatures. The finite-element formulation is based on the principle of virtual work, as written in terms of deviatoric stress quantities. The incompressibility constraint is imposed by direct elimination of nodal displacements, and the resulting stiffness matrix is symmetric and positive definite. Once a convergent solution for displacements, and hence deviatoric stresses, is determined, the hydrostatic stress is obtained by direct use of the principle of virtual work.

Hutchinson, Needleman and Shih (Proc. 1978 ONR Int. Symp. on Fracture Mech., to appear) have applied the new program to the fully plastic analysis of two pre-cracked fracture test configurations: an edge-cracked bar loaded in bending and a center-cracked bar in tension. The specimens were chosen because, according to ideally plastic analysis, they result in very different near tip stress and deformation fields at general yielding. It is known, on the other hand, that strain hardening tends to produce a unique near tip field (the "HRR" field) of a one-parameter character with intensity given by the J integral. Hence the purpose of the numerical solutions was to determine conditions under which this one-parameter field prevails over some non-negligible size scale near the crack (which is a requirement necessary for correlation of the onset of crack growth in terms of J). It was found that for the bend specimen the one-parameter HRR field prevailed near the crack at all levels of strain hardening examined, but not for the center cracked tension specimen. In that case hardening exponents $N \geq 0.3$ seem to lead to the one-parameter field but it prevails only over a small region near the tip; for $N = 0.1$ the results from even the smallest elements adjacent to the tip are significantly different from the HRR field. Thus, among fully plastic specimens, the bend geometry seems to be capable of simulating near tip states as would exist in larger specimens with contained plastic yielding, but the fully plastic center cracked geometry does not.

(Staff: A. Needleman)

B. Reports, Publications, Theses, Oral Presentations, and Other Related Activities

1. Technical Reports

C00-3084-53 J. Gurland, J. R. Rice, R. J. Asaro, and A. Needleman, Technical Progress Report for July 1976 to June 1977, July 1977.

C00-3084-54 R. J. Asaro, Adsorption Induced Losses in Interfacial Cohesion, July 1977.

C00-3084-55 E. P. Sorensen, A Numerical Investigation of Plane Strain Stable Crack Growth under Small Scale Yielding Conditions, September 1977.

C00-3084-56 J. R. Rice and E. P. Sorensen, Continuing Crack Tip Deformation and Fracture for Plane Strain Crack Growth in Elastic-Plastic Solids, October 1977.

C00-3084-57 A. Needleman and J. R. Rice, Limits to Ductility Set by Plastic Flow Localization, November 1977.

C00-3084-58 A. Needleman and C. F. Shih, A Finite Element Method for Plane Strain Deformations of Incompressible Solids, November 1977.

C00-3084-59 H. Cialone, The Role of Hydrogen in the Ductile Fracture of Plain Carbon Steels, April 1978.

C00-3084-60 T-j. Chuang, K. I. Kagawa, J. R. Rice, and L. Sills, Non-Equilibrium Models for Diffusive Cavitation of Grain Interfaces, June 1978.

2. Publications

*R. J. Asaro and J. R. Rice, Strain Localization in Ductile Single Crystals, J. Mech. Phys. Solids, 25, 1977, pp. 309-338.

*Y. W. Chang and R. J. Asaro, Bauschinger Effects and Work-Hardening in Spheroidized Steels, J. Metal Sci., in press.

*R. M. McMeeking, Blunting of a Plane Strain Crack Tip into a Shape with Vertices, Trans. ASME (Jour. Engr. Materials and Technology), 99, 1977, pp. 290-297.

*R. M. McMeeking, Finite Deformation Analysis of Crack Tip Opening in Elastic-Plastic Materials and Implications for Fracture Initiation, J. Mech. Phys. Solids, 25, 1977, pp. 357-381.

*R. M. McMeeking, Path Dependence of the J Integral and the Role of J as a Parameter Characterizing the Near Tip Field, ASTM Special Technical Publication, 631, 1977, pp. 28-41.

E. P. Sorensen, A Numerical Investigation of Plane Strain Stable Crack Growth Under Small Scale Yielding Conditions, ASTM STP (Proceedings of the ASTM Symposium on Elastic-Plastic Fracture, Atlanta, November 1977), in press.

*Listed as "in press" or "submitted" in last year's report.

J. R. Rice and E. P. Sorensen, Continuing Crack Tip Deformation and Fracture for Plane Strain Crack Growth in Elastic-Plastic Solids, J. Mech. Phys. Solids, in press.

A. Needleman and J. R. Rice, Limits to Ductility Set by Plastic Flow Localization, in Mechanics of Sheet Metal Forming, Plenum Press, in press.

A. Needleman and C. F. Shih, A Finite Element Method for Plane Strain Deformation of Incompressible Solids, Computer Methods in Appl. Mech. and Engr., in press.

J. W. Hutchinson, A. Needleman, and C. F. Shih, Fully Plastic Crack Problems in Bending and Tension, to be published in Proceedings of the ONR International Symposium on Fracture Mechanics (Washington, DC, September 1978).

D. D. Mason, Segregation Induced Embrittlement of Grain Interfaces, submitted to Philosophical Magazine.

E. L. Exner, J. R. Pickens, and J. Gurland, A Comparison of Indentation Crack Resistance and Fracture Toughness of Five WC-Co Alloys, Met. Trans., in press.

H. Cialone and R. J. Asaro, The Role of Hydrogen in the Ductile Fracture of Plain Carbon Steels, submitted to Met. Trans.

R. J. Asaro, Adsorption Induced Losses in Interfacial Cohesion, to appear in the Proceedings of Conference on Residuals, Additives and Material Properties (The Royal Society, London, May 1978).

H. Yamamoto, Conditions for Shear Localization in the Ductile Fracture of Void-Containing Materials, Int. Journal of Fracture, in press.

3. Theses (All Degrees awarded in June, 1978)

E. P. Sorensen, Ph.D., September 1977, Some Numerical Studies of Stable Crack Growth.

H. Cialone, M.S., April 1978, the Role of Hydrogen in the Ductile Fracture of Plain Carbon Steels.

4. Oral Presentations

R. J. Asaro presented work on interfacial cohesion and grain boundary fracture at the conference on Residuals, Additives and Material Properties held in London, May 1978 and will present similar lectures at Stanford University, Sondia Laboratories, Livermore, California, and at the ARPA sponsored workshop on Fundamentals of Cohesion, Adsorption and the Strength of Material Interfaces in La Jolla, California in July 1978. He also gave lectures on localized deformation and fracture in ductile single crystals at the U. S. Naval Research Laboratory (Washington, DC), Westinghouse Research Labs, U. S. Steel Research Lab, Yale University, and at the A. I. M. E. symposium, "Forming, Analysis and Experiment".

J. Gurland has presented work during 1978 on the strength and deformation of spheroidized steels at the University of Glasgow and the University of Strathclyde, Scotland; at the University of Caen, France; and at the Chalmers University of Technology, Göteborg, Sweden. In addition, he will present work on indentation cracking at the 5th European Symposium on Powder Metallurgy, Stockholm, Sweden, 1978.

J. R. Rice presented work with E. P. Sorensen on stable crack growth in ductile metals at a Nuclear Regulatory Commission (NRC) sponsored meeting at Harvard University, September 1977, and at the conference on Concepts in Elastic-Plastic Fracture Mechanics, Washington University, May/June 1978. He also presented work with A. Needleman on plastic flow localization and ductile rupture at the General Motors Conference on Mechanics of Sheet Metal Forming, Warren, Michigan, October 1977, and gave a review lecture on computational problems on elastic-plastic fracture mechanics at the Conference on Numerical Methods in Fracture Mechanics, Swansea, Wales, January 1978. In addition, Rice will give a Special Lecture on the mechanics of quasi-static crack growth at the 8th U. S. National Congress of Applied Mechanics, UCLA, June 1978, and he will present work on adsorption and cohesion at an informal ARPA/University of Michigan conference at La Jolla in July 1978, and on diffusive mechanisms of grain boundary cavition at the Gordon Conference on Ceramics, New Hampshire, August 1978.

E. P. Sorensen presented work on finite element computational analyses of growing cracks at the ASTM Symposium on Elastic-Plastic Fracture, Atlanta, November 1977.

5. Related Activities

R. J. Asaro is a member of the Physics and Chemistry of Solids Committee of A. I. M. E. and has recently organized the conference on Fundamental Aspects of Plasticity and Fracture, at the Fall 1977 AIME Meeting.

A. Needleman was the recipient of a John Simon Guggenheim Memorial Fellowship in support of research, now being carried out with R. Hill at the University of Cambridge, on the stability of plastic flow, localization, and ductile rupture.

J. R. Rice is co-chairman of the newly formed task group for Research on Advanced Concepts, ASTM Subcommittee E24.08 on Elastic-Plastic and Fully Plastic Fracture Mechanics Technology. Also, Rice and J. P. Hirth (Ohio State University) have organized an ARPA-sponsored informal conference on Fundamentals of Cohesion, Adsorption and the Strength of Interfaces, to be held in La Jolla, California during July 1978. In addition, Rice has been appointed a member (inactive, thus far) of the Advisory Committee for the High Temperature Materials Laboratory of Oak Ridge National Laboratory, a member of the Publications Committee for the International Conferences on Fracture, and a member of the Editorial Boards for the Quarterly of Applied Mathematics and the Journal of the Mechanics and Physics of Solids. He was also elected as a Fellow of the American Academy of Arts and Sciences in 1978.

C. Personnel

1. Personnel Connected with Contract

1. Professional Staff: R. J. Asaro, J. Gurland, A. Needleman, and J. R. Rice

2. Research Engineer: L. Hermann

3. Technical Assistant: H. Stanton, W. Rebello

4. Research Assistants (Graduate Students): H. Cialone, J. R. Fisher, D. Mason, E. P. Sorensen*, A. Tin, and S. Sham**

5. Undergraduate Research Assistant: E. Exner**

*Supported for portion only of contract period

**Not supported by contract