MASTER

T/C

COO-3084-69

TECHNICAL PROGRESS REPORT

Contract EY-76-S-02-3084 A2

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

I. Review of Past Work: 1970-1979

II. Annual Progress Report: 1978-1979

R. J. Asaro, J. Gurland, A. Needleman, and J. R. Rice

Division of Engineering
Brown University
Providence, Rhode Island

U. S. Department of Energy
Contract No. EY-76-S-02-3084 A2
Technical Report No. 69
June 1979

NOTICE

This report was prepared as an account of work sponsored by the United States Government. Neither the United States nor the United States Department of Energy, nor any of their employees, nor any of their contractors, subcontractors, or their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product or process disclosed or represents that its use would not infringe privately owned rights.

DISCLAIMER

Table of Contents

Table of Contents (continued) Page

Part I

Review of Past Work

This summary presents the highlights of the work carried on, since approximately 1970, under the successive sponsorship of the AEC, ERDA and DOE on "A Combined Macroscopic Approach to the Fracture of Metals".

The major accomplishments were in the following areas:

1) Continuum Elastic-Plastic stress analysis of cracks and its coupling with microscopic fracture mechanisms for the effective prediction of crack tip opening and propagation.

2) Development of the theory of shear localization as a precursor to ductile rupture.

3) Identification of the role of particle cracking in the fracture of steels.

4) Quantitative evaluation of the loss of strength due to adsorption of impurities or environmental elements such as hydrogen, on material interfaces.

5) Development of non-equilibrium models for diffusive cavitation in elevated temperature creep rupture.

6) Analysis of the microstrucural contributions to the yield strength and strain hardening of particle strengthened alloys.

The nature of the problems and their analyses reflects the philosophy of the work on this project, namely the joint application of continuum fracture mechanics and micromechanics to fracture processes in metals and alloys. A successful approach to the solution of several of the problems was fostered by the collaboration of the participating investigators: Professor J. R. Rice, J. Gurland, R. Asaro (since 1975) and A. Needleman (since 1974) and their graduate students (see listing at end of Review Section).

1) Elastic-Plastic Analysis of Crack Tip Fields

The elastic-plastic analysis of cracks is one of the longest continuing projects under this contract and contributions have been made to a variety of topics, some of which have become part of the accepted state of the subject and a source for motivation of significant bodies of related work.

Major portions of the work include:

i) The analysis of stress and strain fields at crack tips, for materials with various hardening characteristics, and with special emphasis on finite plastic deformation associated with crack tip opening [1-3].

ii) The interpretation of results in terms of microscale mechanisms of ductile hole growth or of cleavage microcracking at a crack tip [1,2, 4-7].

iii) The determination of conditions under which the J integral provides a valid macroscopic characterizing parameter of the near tip field over wide ranges of yielding [3,5,8], and

iv) The development of effective finite element numerical procedures for elastic-plastic and fully plastic crack analysis [3,8,9].

These studies referred, for the most part, to the monotonic loading of stationary cracks in rate-independent elastic-plastic materials, and description of the onset of crack growth. As described in the body of the report, more recent studies have focused on the elastic-plastic analysis of growing cracks and the formulation of criteria for stable crack growth in ductile metals [9]. Also, recent studies have included a first analysis of time-dependent stress fields associated with the sudden loading of tensile cracks in elastic-nonlinear viscous materials [10]; the latter studies are relevant to cracks under elevated temperature creep conditions.

Contributions to the area have been made by Rice, with various coworkers (principally, Drugan, Parks, McMeeking, Sorensen, Riedel) in the full range of topics listed, and additional contributions have been made by Needleman on computational techniques, Gurland (with Rawal) on crack tip fracture processes and, recently, by Asaro (with Hermann and Odegaard) on stable crack growth.

Some of the principal publications are:

[1] J.R. Rice and M.A. Johnson, "The Role of Large Crack Tip Geometry Changes in Plane Strain Fracture" (Tech. Rept. 38 in former AEC Series), in Inelastic Behavior of Solids (ed. M.F. Kanninen et al.), McGraw-Hill, 641 (1970).

[2] R.M. McMeeking, "Finite Deformation Analysis of Crack Tip Opening in Elastic-Plastic Materials and Implications for Fracture Initiation" (Tech. Rept. 44), J. Mech. Phys. Solids 25, 357 (1977).

[3] J.R. Rice, R.M. McMeeking, D.M. Parks and E.P. Sorensen, "Recent Finite Element Studies in Plasticity and Fracture Mechanics" (Tech. Rept. 62), Comp. Meth. Appl. Mech. Engr. 17/18, 411 (1979).

[4] R.O. Ritchie, J.F. Knott and J.R. Rice, "On the Relationship Between Critical Tensile Stress and Fracture Toughness in Mild Steel" (Tech. Rept. 20), J. Mech. Phys. Solids 21, 395 (1973).

[5] J.R. Rice, "Elastic-Plastic Fracture Mechanics" (Tech. Rept. 39), in The Mechanics of Fracture (ed. F. Erdogan), ASME/AMD Vol. 19, 23 (1976).

[6] J. Gurland and S. Rawal, "Observations on the Effect of Cementite Particles on the Fracture Toughness of Spheroidized Carbon Steels", Met. Trans. 8A, 691 (1977).

[7] D.M. Parks, "Interpretation of Irradiation Effects on the Fracture Toughness of a Pressure Vessel Steel in Terms of Crack Tip Stress Analysis", Trans. ASME (J. Engr. Mat. Tech.), 98, 30 (1976).

[8] J.W. Hutchinson, A. Needleman and C.F. Shih, "Fully Plastic Crack Problems in Bending and Tension", Proc. ONR Symp. Fracture Mechanics (ed. N. Perrone et al.), 515 (1978).

[9] J.R. Rice and E.P. Sorensen, "Continuing Crack Tip Deformation and Fracture for Plane Strain Crack Growth in Elastic Plastic Solids" (Tech. Rept. 56), J. Mech. Phys. Solids 26, 163 (1978).

[10] H. Riedel and J.R. Rice, "Tensile Cracks in Creeping Solids" (Tech. Rept. 64), Presented at ASTM 12th Annual Symp. on Fracture Mech. (St. Louis, 1979) - in review by ASTM.

[11] J. R. Rice, W. J. Drugan and T-L. Sham, "Elastic-Plastic Analysis of Growing cracks" (Tech. Rept. 65), Ibid.

[12] T. K. Odegaard and R. J. Asaro, "Correlations Between Fracture Toughness and Microstructure in 4140 steel," (Tech. Rept. 66) to be submitted for publication in Met. Trans.

[13] A. Needleman and C. F. Shih, "A Finite Element Method for Plane Strain Deformation of Incompressible Solids", Comp. Meth. Appl. Mech. Eng. 15, pp. 223-240 (1978).

2) Shear Localization of Plastic Flow

The localization of plastic flow into a narrow shear band is a frequently observed precursor to ductile rupture. In some cases the localization seems to result from progressive softening of the material due to more or less homogeneously distributed cavitation, and the shear band becomes a locus of greatly intensified cavity growth, leading directly to ductile rupture. In other cases, e.g., certain age hardened alloy crystals, localization has no apparent connection with prior cavitation, and seems to result from some inherent instability of the plastic flow process. In this latter case, the intensified straining within the band may, nevertheless, lead to cavity nucleation and rupture with only little further overall strain. Often the two sources of instability are combined, with rupture cavities originating from larger inclusions and joining via a localization instability of flow in the remaining ligaments.

Although the work is now supported on other projects at Brown, a major development of the theory of such localization phenomena took place in work by Rice [1], later in collaboration with Asaro [2] and Needleman [3], on this contract. The work involved the determination of instability conditions, in the form of a bifurcation of previously uniform deformation into a localized band, in relation to parameters of elastic-plastic constitutive relations describing various plastic flow processes. The cases examined include single crystals in single and multi-slip, polycrystals, including materials exhibiting plastic dilation and potential softening through hole growth, and the somewhat analogous case of thin ductile sheets subject to local necking failure in forming operations.

Asaro and Rice [2] gave a precise formulation of incremental constitutive relations for a ductile crystal deforming in single slip and, specifically, included the effects of stresses other than the Schmid resolved shear stress on plastic flow. They pointed out that such effects should arise at the onset of cross-slip in fcc crystals, and analyzed a dislocation model for the size of the non-Schmid stress effects. They found that localization was predicted when the crystal hardening modulus h fell to values typically in the range 5×10^{-4} G to 2.5×10^{-3} G (G = elastic shear modulus), which seems to be in at least qualitative accord with observations: Stage II hardening rates are of the order of G/300 (3.3×10^{-3} G) and in Stage III, where localization is observed, h is falling in magnitude from this value. (When there are no effects of stresses other than the Schmid stress, the critical value of h is zero - i.e., an ideally plastic state must then be achieved for localization).

Studies by Needleman and Rice [3] focused on the development of macroscopic constitutive relations to describe large plastic flow and strain weakening (e.g., by micro-cavity nucleation and growth) in polycrystals, and the use of these constitutive relations to determine limiting conditions at which localized shear occurs.

One feature of the large plastic strain response of polycrystals is that the incremental tensorial relations between strain and stress are expected, within a rate-independent model, to exhibit a "vertex" structure. This means that the direction (in an appropriate hyperspace) of the strain increment is at least somewhat dependent on the direction of the stress increment, in contrast to classical constitutive models of the Prandtl-Reuss-Mises type. Needleman and Rice showed that this feature leads to localization instabilities at sufficiently reduced strain hardening levels. By choosing the "vertex modulus" as the secant modulus to the uniaxial stress-strain curve, as suggested by analogy with the simplest slip models for polycrystal plasticity in the small strain regime, they obtain results consistent with the trend observed in highly ductile steels by Clausing, of decreasing ratios of plane strain to axi-symmetric extensional ductility with increasing strength levels (hence, with decreasing hardening). The observed strains to fracture in each case are somewhat underestimated, although subsequent work elsewhere (J.W. Hutchinson and K. Neale, Harvard) has shown that other versions of the large-strain constitutive relations can give rupture strain results in better accord with observations.

Localization instabilities have also been studied in the case of less ductile materials, exhibiting extensive microvoid nucleation and growth prior to rupture, based on an elastic-plastic constitutive model for void-containing solids developed in work of a former Ph.D. student, A. Gurson. Needleman and Rice have related instabilities in this case to the progressive weakening of material by hole growth and to stress and strain dependent models for nucleation of new voids (typically by the cracking or decohesion of inclusions). In addition, a former Sc.M. student, H. Yamamoto, working with Rice and Needleman, demonstrated a strong sensitivity to initial non-uniformities of the initial porosity of material in this case.

The principal references for this work are:

[1] J.R. Rice, "The Localization of Plastic Deformation" (Tech. Rept. 43), in <u>Theoretical and Applied Mechanics</u> (Proc. 14th IUTAM Cong., Delft, 1976, ed. W.T. Koiter), North-Holland, Vol. 1, 207 (1976).

[2] R.J. Asaro and J.R. Rice, "Strain Localization in Ductile Single Crystals" (Tech. Rept. 49), J. Mech. Phys. Solids <u>25</u>, 309 (1977).

[3] A. Needleman and J.R. Rice, "Limits to Ductility Set by Plastic Flow Localization" (Tech. Rept. 57), in <u>Mechanics of Sheet Metal Forming</u> (Proc. of G.M. Res. Lab. Symp., 1977, ed. D.P. Koistenen and N-M Wang), Plenum Press, 237 (1978).

3) Crack Initiation at Particles in Steels

The fracture of steels begins with the cracking of particles on particle interfaces. The early work on this topic by Gurland [1] was concerned with the observation of cracking in or at carbide particles in spheroidized steels. The two main modes of crack initiation at the carbide particles are particle cracking and interfacial separation. The former occurs preferentially in the high carbon steels and in the uniformly deformed sections of the tensile specimen, the latter occurs preferentially in the lower carbon steels and in the necked portion of the tensile specimen. Particle fracture takes place progressively throughout the range of plastic deformation in a direction normal to the direction of the maximum tensile strains imposed upon the particles by the deforming matrix. In general, it is observed that the cracks occur preferentially in the larger particles oriented with their largest axes parallel to the direction of tensile loading. Attempts have been made by us (1,2) and others to identify the critical fracture conditions of brittle particles in terms of internal stresses due to shear-lag load transfer and/or local stress concentration associated with slip bands or dislocation pile-ups. Current work by Gurland and Fisher deals with the nucleation process of interfacial voids, and the preliminary results reveal the important role of grain boundaries and of hydrostatic tension. This approach has been extended by Asaro and Cialone to include the effects of hydrogen and other segregants, as discussed in the next section.

The role of cracked particles in the initiation and propagation of fracture in notched and precracked fracture toughness specimens at low temperatures was convincingly shown by Rawal and Gurland, as previously mentioned in connection with the work on elastic-plastic analysis of crack tip fields [Technical Report No. 41].

The major publications in this area are:

[1] J. Gurland, "Observation on the Fracture of Cementite Particles in a Spheroidized 1.05% C Steel Deformed at Room Temperature", Acta Met. <u>20</u>, 735 (1972).

[2] C.T. Liu and J. Gurland, "The Fracture Behavior of Spheroidized Carbon Steels", Trans. ASM, <u>61</u>, 190 (1968).

[3] S.P. Rawal and J. Gurland, "Observations on the Effect of Cementite Particles on the Fracture Toughness of Spheroidized Carbon Steels, Met. Trans. <u>8A</u>, 691 (1977) (Tech. Rept. 41).

4) Segregant Induced Embrittlement of Internal Interfaces and Environmentally Assisted Fracture Mechanics

There are many known examples for which interfacial segregation of residual impurities or of environmental elements like hydrogen cause normally ductile metals to undergo brittle intergranular fractures or to suffer severe losses in ductility. In large part these represent instances where segregation induced cohesion losses are sufficient to either invoke a brittle response of interfaces, where normally they would be ductile, or significantly ease the requirements for interfacial separations such as normally occur at particle interfaces during the process of void initiation. The adverse effects on toughness are often very severe, but also usually subtle and difficult to quantify. During the past four years we have addressed several problems related to these phenomena. The aim has been to develop ways to quantify cohesion losses due to adsorption and to evaluate the effects on overall material response. This has included theoretical work on interfacial strength and experimental work on environmentally assisted fractures.

Rice [1] presented a model for the brittle versus ductile response of an interface which contains an initially sharp crack, based on his earlier work with Thompson [2]. Essentially the model posed the question, if an interface crack is loaded and the stress intensity increases, what criterion is met first, a K corresponding to the Griffith value, K_G - which is sufficient to cause brittle behavior - or a K ($<K_G$) sufficient to cause **athermal** nucleation of a dislocation? The latter case suggests that ductile response precedes brittle crack propagation possibly leading

to plastic blunting of the crack tip. 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 a cutoff radius which appears in continuum dislocation theory. $2\gamma_{int}$ and γ_{step} are the cohesive parameters and represent the works to separate the interface on which the crack lies and to create a step at the crack tip by dislocation blunting, respectively. It is these two parameters that are affected by adsorption of solutes or hydrogen and thus the model allows for an approximate quantification of ductile-brittle transitions due to segregation. An important and novel distinction was made in the work by Rice [1] between interfacial separations occurring slowly enough that near equilibrium conditions prevail, and the more usual case where separations are so rapid that no further adsorption on the separating interface takes place. It was shown that the difference between the work to separate an interface under equilibrium conditions, γ_{slow}, and where the separation is too rapid for any appreciable alteration in interface chemistry to occur, γ_{fast}, is

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

$\Gamma_{io}(\mu)$ and $\Gamma_{ioo}(\mu)$ are the excess concentrations of adsorbed specie i on the interface and on the <u>two</u> newly formed free surfaces respectively; μ_i and μ_{ioo} are the chemical potentials in the bulk phase and on the free surfaces respectively. Asaro [3] applied the above formalism to analyze cohesion losses in alloys where equilibrium adsorption data is available. Losses in grain boundary cohesion for "fast" and "slow" intergranular fractures were calculated for Fe-P alloys - the results indicated that while the cohesion losses were indeed significant for "fast" separations they were substantially less than for "slow" or equilibrium separations thereby demonstrating the importance of distinguishing between these two cases of our new thermodynamic models.

The analyses described above were used by Mason [4] to model an interesting experimental case of polycrystalline embrittlement due to adsorption of Bi on the grain boundaries of Cu. In terms of the parameters R_o and S, introduced earlier, it was shown that if

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

ductile behavior for the grain boundary is predicted. R_o and S also depend upon the orientations of the slip planes intersecting the crack front and the directions of the Burgers vectors and so important differences due to crystal structure can be, in part, accounted for. The model predicted a transition to brittle intergranular behavior, with Bi adsorption, in qualitative agreement with experiment.

Experimental work in this area has up to now focused on the effects of hydrogen on <u>ductile</u> fractures in plain carbon steels and on brittle fractures in high strength steels. Recently completed work on a 1045 spheroidized steel by Cialone and Asaro [5] has shown that hydrogen leads to accelerated void initiation during plastic deformation, although the more interesting effect leading to ductility losses was due to an accelerated void growth. The void growth process was found to involve preferential coalescence of voids along interfaces (ferrite grain boundaries) and it was this process which seemed to be most accelerated by hydrogen. Thus it was found that hydrogen assisted interfacial separations could be important in ductile as well as brittle fractures. Work by Asaro and Cialone on hydrogen assisted fractures in 4140 steels is currently in progress. They have designed a novel environmental chamber for four-point bend specimens and have developed a simple, but effective, acoustical technique for determining the onset of subsurface fracture.

The major references for this work are:

[1] J.R. Rice, "Hydrogen and Interfacial Cohesion" (Tech. Rept. 40), in Effect of Hydrogen on Behavior of Materials (eds. A.W. Thompson and I. M. Bernstein), AIME, N.Y., 455 (1976).

[2] J.R. Rice and R. Thomson, "Ductile vs. Brittle Behavior of Crystals" (Tech. Rept. 22), Phil. Mag. 29, 73 (1974).

[3] R.J. Asaro, "Adsorption Induced Losses in Interfacial Cohesion", to appear in the Proceedings of the Conference on Residuals, Additives and Material Properties, Roy. Soc. 1979.

[4] D.D. Mason, "Segregation Induced Embrittlement of Grain Interfaces" (Tech. Rept. 52), Phil. Mag., to appear, 1979.

[5] H. Cialone and R.J. Asaro, "The Role of Hydrogen in the Ductile Fracture of Plain Carbon Steel" (Tech. Rept. 59), Met. Trans. 10A, 367 (1979).

5) Processes of Diffusive Cavitation of Grain Boundaries in Elevated Temperature Creep Rupture

Classical models of the grain boundary diffusive cavitation process, following from the work of Hull and Rimmer, were based on quasi-equilibrium assumptions: surface diffusion was assumed to be rapid enough that the cavities retained a rounded, near-equilibrium shape, and hence cavity growth was assumed to be rate-limited only by grain boundary diffusion. Further, deformability of the adjoining grains was, for the most part, neglected. For example, in the Hull-Rimmer analysis the grains were assumed to move apart in an effectively rigid manner.

The assumption of negligible grain deformability is often appropriate. However, creep cavities frequently have narrow crack-like shapes, implying that quasi-equilibrium assumptions as to cavity shape are inappropriate. The work by Chuang and Rice on this contract was the first to examine the effects of non-equilibrium surface shapes. Their first paper [1] provides a solution for the cavity profile, according to the equations of surface diffusion, for the opposite limiting case of (comparatively) rapid cavity growth in a narrow crack-like mode.

Subsequent work, initiated in the Ph.D. thesis by Chuang and Sc.M. thesis by Kagawa, led to a comprehensive understanding of non-equilibrium effects in diffusive cavitation. The work was reported by Chuang, Kagawa, Rice and Sills [2], as the second of a recently initiated series of "overview" papers in Acta Met. Their work compared the limiting cases of quasi-equilibrium and crack-like growth models. Criteria for choosing between the two were given on the basis of representative relaxation times for the surface diffusion process, and also by examining the properties of a "self-similar" solution for cavity shape. By a suitable choice of parameters which measure the growth rate, this solution could be made to give results corresponding to either limiting case, and aided the interpolation between them. The results suggested that if s is the ratio of the applied stress to that which just equilibrates cavities against sintering, then for circular cavities on a grain boundary with diameter equal to a quarter of their average center-to-center spacing, the quasi-equilibrium mode applies when $s < 1 + 6\Delta$ and the crack-like mode when $s > 2 + 9\Delta$. Here Δ is the ratio of surface to grain boundary diffusivity. Also, the stress dependence of the growth rate and rupture lifetime were established in each case, and the results were discussed in relation to the interpretation of experimental data, especially by Nix and Goods on Ag polycrystals and Raj on Cu bi-crystals.

The work developed in Chuang's Ph.D. thesis gave the first solution to the coupled equations of elasticity and diffusion for a creep cavity growing along a grain boundary in a narrow crack-like mode. Further, our recent work [3]. described in the body of this Progress Report, has developed significant insights into the interaction between plastic creep flow and diffusion processes in grain boundary cavitation. Although the creep effects are, typically, only important at the higher ranges of stress and temperature, we have found that the two processes acting together can lead to substantially greater rates of cavity growth than either process would if it acted in isolation. This is a subject under continuing study by Needleman and Rice, who have devised a finite element procedure to analyze this novel class of coupled problems.

The major references for this work are:

[1] T-j. Chuang and J.R. Rice, "The Shape of Intergranular Creep Cracks Growing by Surface Diffusion" (Tech. Rept. 21), Acta Met. 21, 1625 (1973).

[2] T-j. Chuang, K.I. Kagawa, J.R. Rice and L.B. Sills, "Non-equilibrium Models for Diffusive Cavitation" (Tech. Rept. 60), Acta Met. (Overview Paper No. 2) 27, 265 (1979).

[3] J.R. Rice, "Plastic Creep Flow Processes in Fracture at Elevated Temperature", Proc. Feb. 1979 DOE Workshop on Time-Dependent Fracture of Materials at Elevated Temperature - in press.

6) Yield Strength and Strain-hardening in Particle-Strengthened Alloys

Our contribution to the understanding of the mechanism of strengthening by dispersed hard particles has revealed both the simplicity and the complexity of the process. The simplicity is associated with the general applicability of the Hall-Petch equation for the yield strength of two-phase microstructures,

$$\sigma_y = \sigma_{o,y} + k_y \lambda^{-1/2} \tag{1}$$

where σ_y is the yield stress, λ is the spacing of structural barriers to the propagation of slip, and $\sigma_{o,y}$ and k_y are experimental constants. The Hall-Petch equation has been used as an empirical approximation to describe the yield strength, and hardness over a wide range of microstructures in carbon steels [1,2] and cast aluminum-silicon alloys [3].

The complexity arises when considering the details of the strengthening mechanisms and the identification of the relevant microstructural spacings in the alloys during yield and strain hardening. Successive studies of tempered and spheroidized steels by Liu and Gurland [4], Anand and Gurland [5] and Chang and Asaro [7] have shown that the strengthening effect attributed to the carbide particles is predominantly due to the boundaries and dislocation structure with which the particles are associated. For instance, the yield strength of spheroidized steels is controlled by the mesh size of the subgrain boundary - or grain boundary - network stabilized by the cementite particles and of mesh size similar to the particle spacing. Equation 1 applies if λ is taken as the boundary spacing [5]. Isolated carbide particles contribute only a small amount to the yield strength by slightly increasing the friction stress $\sigma_{o,y}$.

The particles play a more direct role during strain hardening. A large fraction (approximately 20%) of the strain hardening increment arises from the internal stress due to the presence of unrelaxed plastic strain discontinuity between the ferrite matrix and the cementite particles. The internal stresses increase up to a plastic strain range of 3-5% as a result of the increasing impediment offered by the strain - generated dislocation density around the particles to the secondary dislocation processes which are necessary to relieve the stresses in and around the particles. The limiting value of the internal stress corresponds to the attainment of critical particle or particle-matrix interface stresses which are sufficient to nucleate the secondary stress relaxation processes [6,7]. The transition was proposed to be responsible for the observed "double-n" strain-hardening behavior; this was later confirmed by Chang and Asaro [7] on 1045 and 1090 steels.

Chang and Asaro [7] also made direct measurements of the residual internal stresses. They found the experimentally determined Bauschinger effect in good accord with that predicted by a continuum analysis of the internal stress fields based upon a multiple slip model. The cementite particle pinned subgrain boundaries, formed during a post-quench annealing treatment, were found to lower the internal stress levels, thus indicating that they assisted the relaxation processes of entrapped Orowan loops by acting as sources of dislocation.

The flow stress curves of these spheroidized carbon steels were found to be approximately described by a modified mean-square-root addition law of the form

$$\sigma_f = \sigma_{\epsilon p=0} + \sigma^i + [(\Delta\sigma^s)^2 + (\Delta\sigma^{ss})^2 + (\sigma^f)^2]^{1/2}$$

where σ^i is the "internal stress", σ^f a local "forest stress", $\Delta\sigma^{ss}$ the "source shortening stress", and $\Delta\sigma^s$ is the stress due to the matrix dislocation density [7].

The major publications describing this work are:

[1] J. Gurland,"Correlation Between Yield Strength and Microstructure of Some Carbon Steels", in ASTM, STP 504, 108 (1972).

[2] P.C. Jindal and J. Gurland, "On the Relation of Hardness and Microstructure of Tempered and Spheroidized Carbon Steels", Met. Trans. 9, 1649 (1974). (Tech. Rept. 27)

[3] E. Erginer and J. Gurland, "The Influence of Composition and Microstructure on the Strength of Cast Aluminum-Silicon Alloys", Z. Metallkde, 61, 606 (1970).

[4] C.T. Liu and J. Gurland, "The Strengthening Mechanism in Spheroidized Carbon Steels", Trans. Met. Soc., AIME, 242, 1939 (1968).

[5] L. Anand and J. Gurland, "Effect of Internal Boundaries - The Yield Strengths of Spheroidized Steels", Met. Trans. 7A, 191 (1976). (Tech. Rept. 35)

[6] L. Anand and J. Gurland, "Strain Hardening of Spheroidized High Carbon Steels", Acta Metallurgica, 24, 901 (1976). (Tech. Rept. 37)

[7] Y.W. Chang and R.J. Asaro, "Bauschinger Effects and Work Hardening in Spheroidized Steels", Met. Sci. 12, 277 (1978). (Tech. Rept. 51)

List of Past Graduates

Tze-jer Chuang (Ph.D.), Engineer, Advanced Systems Technology, Westinghouse Electric Company, Pittsburgh, Pennsylvania.

Arthur L. Gurson (Sc.M., Ph.D.), Research Engineer, Avco Corp., Lexington, Massachusetts.

Keith I. Kagawa (Sc.M.), Mechanical Engineer, Hawaiian Fluid Power Corporation, Honolulu, Hawaii.

Robert M. McMeeking (Sc.M., Ph.D.), Assistant Professor of Theoretical and Applied Mechanics, University of Illinois, Urbana, Illinois.

Deborah D. Mason (Sc.M.), Mechanical Engineer, Polaroid Corporation, Waltham, Massachusetts.

David M. Parks (Sc.M., Ph.D.), Assistant Professor of Mechanical Engineering Massachusetts Institute of Technology, Cambridge, Mass.

E. Paul Sorensen (Ph.D.), Engineer, Hibbitt and Karlsson Inc., Providence, Rhode Island.

Hiro Yamamoto (Sc.M.), Engineer, Nippon Steel Corporation, Tokyo, Japan.

Chain-Tsuan Liu (Ph.D.), Metals and Ceramics Division, Oak Ridge National Laboratory, Oak Ridge, Tennessee.

Prem C. Jindal (Ph.D.), Staff Metallurgist, ABAR Corporation, Feasterville, Pennsylvania.

Erk Erginer (Ph.D.), King, North Carolina.

Lallit Anand (Ph.D.), Research Scientist, Research Laboratory, U.S. Steel Corporation, Monroeville, Pennsylvania.

Suraj Rawal (Ph.D.), Chief Metallurgist, New England Metals Company, Manville, Rhode Island.

Henry Cialone (Sc.M.), Graduate Student, Brown University.

Young Won Chang (Sc.M.), Graduate Student, Brown University.

Teng-Ling Sham (Sc.M.), Graduate Student, Brown University.

Trond K. Odegaard (Sc.M.), Research Staff, General Electric Laboratories, San Jose, California.

PART II

Annual Progress Report (1978-1979)

EY-76-S-02-3084 A2
Annual Technical Progress Report

June 1979

Abstract

The accomplishments during the past year include progress in the following areas: 1) Microscopic fracture mechanisms, including studies of void initiation at dispersed particles and void growth in the absence and presence of hydrogen, providing a description of the effects of hydrogen on ductile fracture in medium and high carbon steels; 2) Elastic plastic crack growth, including the analytical description of the quasi-stable growth of cracks in ductile solids under increasing load and the conditions of instability; and 3) Elevated temperature rupture, including an analysis of the stress field near a crack tip in an elastic-nonlinear viscous material under tensile load, as well as consideration of the processes of diffusive cavitation of grain boundaries in plastically creeping materials.

EY-76-S-02-3084 A2
Technical Progress Report

June 1979

A. Detailed Description of Research Program

1. Microscopic Fracture Mechanisms

1a. Void and Crack Initiation in Steels

The main objective of the study of void initiation in plain carbon spheroidized steels is to relate the macro-and micro-mechanical conditions of crack initiation and propagation to the microstructure, in particular to the type, size and distribution of the second phase particles. The study encompasses a wide range of carbon steels (0.13 to 1.1 wt. percent carbon) heat treated to produce a microstructure consisting of spheroidized iron carbide particles embedded in matrix of ferrite.

The improved metallographic techniques described in the preceding report have revealed convincingly that the preferred cavity nucleation sites are located at the intersections of grain boundaries and cementite particles. The study of cavity initiation in spheroidized carbon steels has been continued both experimentally and analytically.

Much of the experimental work consisted of the quantitative measurement of the number density of voids on transverse and longitudinal sections in the necked portion of broken tensile specimens. For the purposes of determining number and area fraction, the transverse sections are much preferred over the longitudinal sections which suffer from polishing and etching problems associated with the shallowness of the voids in the highly deformed grain structure. The quantitative results confirm the previously noted preferential location of voids at the particle-matrix interface at grain boundaries, also the frequent cavitation of matrix between closely spaced neighboring particles on a grain boundary. These observations concerning the role of grain boundaries in the ductile fracture process are in general accord with our parallel studies, carried out by Cialone and Asaro (section 1c), on the effects of internal hydrogen on ductile fracture in spheroidized steels. There it was found that an important influence of hydrogen leading to a loss in ductility was to facilitate the process of void growth and coalescence along grain boundaries. In addition, the analysis of the data indicates a notable effect of hydrostatic tension on the frequency of void initiation. The results also indicate that voids nucleate preferentially at larger particles.

In connection with the quantitative metallography consideration was given to the effect of plastic deformation on the microstructural parameters. It is shown both theoretically and experimentally that the average area fraction and the average spacing parameters (for instance, the mean free path) are invariant during deformation, but that the orientation of non-equiaxed particles is influenced by the deformation.

A rigid tensile testing attachment to the mechanical "Instron" testing machine was built and is currently under evaluation. The purpose of the device is to permit precise control of the specimen elongation at the moment of incipient fracture and thereby to ascertain experimentally the critical volume fraction of voids resulting from nucleation and growth required for deformation instability and fracture of the specimen in comparison with the theoretical predictions of Gurson, Rice and Needleman.

An analytical model of the interfacial void initiation is being developed. The model combines linear elastic fracture mechanics with the Eshelby transformation theory in order to calculate the conditions required to form a stable void cap on a spherical particle. It shows that as a result of the nucleating process there is both a critical minimum size of the void cap, due to the simultaneous requirement of a critical stress and a critical minimum energy, and an equilibrium maximum size, due to the decrease of the energy release rate as a function of growth of the void cap. When applied to isolated particles inside the grains, the model predicts the requirement of much higher values of the unrelaxed plastic strains than were experimentally determined by Chang and Asaro (Technical Report 51) and therefore indicates that void nucleation is unlikely at isolated particles.

The model was extended to spherical particles at grain boundaries by considering the large incompatibility associated with the particle-grain boundary intersection, and combines the incompatibility effects with the results of the void cap growth model. The detailed analysis is very complex and somewhat approximate, involving both Mode I and Mode II crack opening along the particle-matrix interface, and, simultaneously, Mode I and Mode II crack extension on the grain boundary near the particle. The calculations to date support two important conclusions: 1) the energy conditions favor crack initiation at the particle-grain boundary site as opposed to isolated particles, and 2) the voids form by a nucleation process with a minimum critical size. In addition, the results of the model agree with the observed effects of the externally imposed hydrostatic tension in the neck of the tensile specimens and those due to the interaction of neighboring particles; also the model predicts a particle size effect on nucleation, which is also observed experimentally.

It is expected that this phase of the study will be completed during the first half of the next contract period.

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

1b. Load Transfer to Dispersed Particles in Two-Phase Alloys

In connection with the fracture of particles on particle interfaces, it is pertinent to evaluate the average and maximum stresses in the particles due to an externally applied load. One of the contributions to the particle stress is the load transferred from a plastic matrix to hard particles across the particle-matrix interface. An approximate expression for this contribution was derived for a dispersion of irregularly shaped particles by adopting the shear lag model used for discontinuous fibers. It was shown that the equality

$$\frac{\bar{S}_{\alpha\beta}}{4\bar{A}_\beta} = \frac{\bar{L}_\beta}{\bar{\ell}_\beta}$$

exists between the ratio of mean surface area $\bar{S}_{\alpha\beta}$ to mean cross-sectional area $\bar{A}_\beta$ and the ratio of mean caliber length $\bar{L}_\beta$ to mean intercept length $\bar{\ell}_\beta$ of the particles. The particle stress due to load transfer is proportional to $S_{\alpha\beta}/A_\beta$ and, in the absence of particle yield or fracture,

$$\bar{\sigma}_\beta = \frac{\gamma}{2} \frac{\bar{L}_\beta}{\bar{\ell}_\beta} \sigma_{\alpha y}$$

where $\bar{\sigma}_\beta$ is the average stress in the dispersed β phase, γ is a geometrical orientation factor ($\gamma \simeq \frac{1}{2}$), and $\sigma_{\alpha y}$ is the in-situ yield strength of the ductile matrix. Appropriate, but approximate relations have also been developed to represent the average stresses in particles which yield to the transferred load. In that case also, the average stress is a function of the aspect ratio $\bar{L}_\beta/\bar{\ell}_\beta$. These considerations permit a more accurate estimate of the particle stresses, as needed for instance for the evaluation of the fracture strength of the particles. A technical report is in preparation.

(Staff: J. Gurland)

c. Environment-Sensitive Fracture Mechanisms

There are numerous ways in which environments adversely affect the mechanical properties of solids, especially fracture toughness and ductility. In metals, such as the steels that were, and are being, studied in our research, it is useful to separate the specific manifestations of adverse environments into two categories: either there is no apparent change in the fracture mode, and the "normal" process of fracture is made to occur at lower stresses or strains, or there is a transition in fracture mode from one which is ductile, involving void initiation, growth, and coalescence, to one characterized by cleavage and intergranular fracture. In lower strength, ductile materials the former process is common, whereas in high strength materials, which often have rather modest rates of strainhardening, it happens that cohesion losses at grain boundaries or other internal interfaces can cause either transgranular or intergranular cleavage to occur at lower stresses and strains so as to preempt what might have been a more ductile process. In our work to date we have focused on both types of effects involving the influence of hydrogen on the fracture of low-to-medium strength plain carbon steels and high strength alloys like 4140. It is important to note that these studies into the influence of hydrogen on ductile and brittle-like fracture processes have direct application to many instances of stress corrosion cracking in ferrous alloys. This is so since many corrosion induced fractures are in fact due to a combination of localized oxidation reactions, hydrogen evolution in the corrosion reactions, and subsequent hydrogen embrittlement.

The normal fracture mode in a spheroidized plain carbon steel, like 1030 or 1040 is ductile but when they are subjected to severe hydrogen environments--in our case this was by electrolytic charging--they undergo large losses in ductility, as high as 40 percent. However, as discussed in Technical Report

59 and by Cialone and Asaro (Met. Trans. 1979) there are no important changes in the microscopic or macroscopic processes of fracture when these 1030 or 1040 steels are tested with or without hydrogen. The effect of introducing hydrogen is instead to accelerate, with strain, both void initiation at carbide interfaces and void growth. In Technical Report 59 it was shown that void coalescence took place by the propagation of voids, in a ductile fashion, along interfaces which interlinked the carbides at which the voids initiated. This was a novel observation which helped explain the greatly increased void growth rates observed with hydrogen. The results are consistent with the idea that hydrogen lowers the cohesive strength of these interfaces as described, for example, by the adsorption models of Asaro, Mason and Rice discussed in Technical Reports 40, 52, and 54. Other mechanisms for increased void growth, such as internal pressure effects due to transport of hydrogen to the voids, were considered but shown to be inadequate to explain the observations. Subsequent work, discussed in the publication by Cialone and Asaro (Met. Trans. 1979), was aimed at studying void <u>initiation</u> in more detail. The new experiments showed that internal hydrogen, introduced by electrolytic charging, had a small but definite effect on void initiation. Voids initiated at smaller values of strain--and therefore smaller values of stress acting locally at the particle-matrix interfaces. The effect was quantified approximately by modifying the analysis of Argon and co-workers (Met. Trans. 1975) for void initiation at spheroidal carbides and by using the ideas and results discussed by Gurland and co-workers in Technical Reports 5, 37, and 51 on strainhardening and void initiation in spheroidized steels. Void initiation and the morphology of initial void growth, which occurred along the tensile axis, was further interpreted in terms of the models developed by Chang and Asaro (Metal Sci. 1978) for the normal and hydrostatic stresses acting at particle interfaces after plastic deformation. Again, the results suggested that hydrogen

induced cohesion losses at the particle-matrix interfaces, and of the ferrite grain boundaries on which the cavitated carbides lie, were the main contributting factor leading to ductility loss. However, as described above, it was the greatly increased void growth rates which accounted for the large losses in ductility.

These findings suggest the need to critically assess the role of impurities, as well as environments, in ductile fracture processes in general. Temper embrittlement, caused by the segregation of solutes to interfaces, is a phenomenologically well understood example of this but where the fractures are often quite brittle like. However, as essentially all fractures, ductile or brittle, begin at interfaces of some sort it seems entirely plausible that there are other cases of ductile fracture where losses interfacial strength due to the adsorption of impurities are also important.

The work summarized above seems to provide a reasonably thorough description of the effects of hydrogen on ductile fracture in the plain carbon steels we studied. This work is being currently extended to lower carbon steels, partly for the purpose of determining the generality of our findings.

As previously mentioned, in high strength materials environment effects can lead to a change in the fracture mode from one which is ductile to one involving intergranular cleavage. We are currently performing experiments on a high strength 4140 steel in gaseous hydrogen environments which are designed in their longer term goals to document the conditions under which these transitions take place. The more immediate aims, though, involve temper embrittling the alloys so that the fracture mode will remain unchanged as intergranular cleavage. We have varied the compositions of a series of 4140 alloys, in particular the phosphorous content, to facilitate complete embrittlement of the prior austenite grain boundaries. Hydrogen then introduced at known chemical activity will augment this single fracture mode and our plan is to quantify the effect. We have designed a novel specimen chamber which isolates the notch region of pure bend specimens and have developed a simple but effective acoustic technique to determine the onset of subsurface fracture. Several experiments have been already conducted and we are convinced that our apparatus functions quite adequately. To summarize then, the goals of those experiments are to obtain an understanding of the effects of hydrogen on (i) accelerating void initiation in front of blunt notches and sharp cracks, (ii) accelerating the coalescence of voids perhaps through premature localized shearing brought on by the larger microvoid density and through the interface propagation mechanism discussed earlier, (iii) facilitating interfacial cleavage, especially when cleavage is induced by prior temper embrittlement, and (iv) inducing transitions in fracture mode when prior embrittlement treatments are not given.

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

2. Mechanics of Crack Growth

2a. Elastic-Plastic Crack Growth in Ductile Metals

Studies by a combination of analytical and finite element methods have been continued on near-crack-tip stress fields in rate-independent elastic-plastic materials. These are intended to model fracture at temperatures below the range of substantial creep effects, and are aimed particularly at relatively tough ductile metals such as pressure vessel steels. Studies on crack tip stress analysis, and on associated problems of finite-element computation at large plastic strain, have been summarized in a paper written by Rice (Technical Report 62) in collaboration with former students, R. M. McMeeking (University of Illinois), D. M. Parks (M.I.T.) and E.P. Sorensen (Hibbitt-Karlsson, Inc.). The paper discusses various studies of elastic-plastic crack tip fields, giving particular emphasis to the finite openings at the tip of a loaded stationary crack and to the development of large plastic strains and triaxial stressing near the tip. Further, results are presented which show the extent to which plane strain near tip fields are determined in an essentially unique manner by the J integral, irrespective of the extent of yielding. For an edge-cracked bend specimen the stress and strain fields near the crack tip (over the size scale of the tip opening, which is known to coincide roughly with the fracture process zone size at the onset of growth) define nearly unique distributions when plotted against $r/(J/\sigma_o)$, where σ_o = yield strength, at all load levels from those of small scale yielding to those of the general yielding range. Hence in this type of specimen, which is also representative of typical fracture test specimens of the compact type, J seems to be a valid measure of the near tip state. On the other hand, for the center cracked tension specimen (which does not maintain a plastically constrained state of high triaxiality at the tip in the fully plastic range) the correlation of the near tip field with J becomes invalid only slightly beyond the small scale yielding range, even before plastic flow has spread over the uncracked ligament.

The major focus of our more recent studies of cracks in elastic-plastic materials has been on the description of the quasi-stable growth of cracks in ductile solids under increasing load, with an aim towards determining conditions under which the crack growth process becomes unstable in favor of a rapidly propagating fracture. While the stress and deformation fields at the tips of stationary cracks have been studied extensively, similar fields at the tips of quasi-statically growing cracks are not well understood, and the problem is then considerably more complicated because of the strain-path dependence of elastic-plastic stress strain relations. A recent combination of analytical and numerical finite element studies, by Rice, Drugan, and Sham (Technical Report 65) had led to the determination of the form of the stress and deformation fields very near the tip of a growing crack in an elastic-ideally plastic solid under plane-strain small scale yielding conditions. The work completes and refines initial studies of the problem by Rice and Sorensen, published in 1978, and provides a fremework for describing stable crack growth. One portion of the recent work involves an asymptotic solution for the near tip stress field, and it is found that a stress state similar to the well known

"Prandtl field" (which results for the case of a stationary crack) results very near the tip of a growing crack. This modified Prandtl field involves an angular sector of elastic unloading occupying the angular range between approximately $\theta = 115^{o}$ and $\theta = 162^{o}$ (θ is the angle measured from the plane along which the crack advances). Remarkably, in spite of this relatively large zone of elastic unloading, the stress immediately ahead of the crack is only slightly less than 1% smaller than that of the Prandtl field. Further, the asymptotic analysis shows that the rate of crack opening at small distances r behind the crack tip is given by the same form of expression proposed by Rice and Sorensen,

$$\dot{\delta} = \alpha\dot{J}/\sigma_o + \beta(\sigma_o/E)\dot{a} \, \ell og \, (R/r)$$

where J is the far-field value of the J integral, σ_o is yield strength, E is the elastic modulus, and a is crack length, but now β has the revised value of 5.08 for Poisson ratio of 0.3. The parameters α and R are undetermined by the asymptotic analysis.

Rice, Drugan and Sham made approximate estimates of α and β by correlating the theoretical formula against results of a finite-element solution done by Sham in consultation with Needleman (this solution was initiated under DOE support, and completed in calculations carried out by Sham at General Electric, Schenectady, under support from an EPRI funded GE subcontract to Brown). This finite element solution, which was limited to the plane strain small scale yielding range, suggests α to be in the range of approximately 0.65 to 0.70, and to be approximately unaffected by prior crack growth, at least for the limited amounts of growth studied. Also, R is found to scale approximately with the extent of the plastic zone, with

$$R \approx 0.2\ EJ/\sigma_o^2$$

for the small scale yielding cases studied. However, the finite element solution still leaves ambiguities in the determination of α and R (and of their possible dependence on prior crack growth). Hence, a new finite element solution with a refined and improved mesh has been initiated by Sham, to be carried out under DOE funding on the Brown computer.

An approximate crack growth criterion was examined, based on the Rice and Sorensen proposal that a critical opening of the crack surfaces be maintained at some distance behind the tip for continuing growth. This criterion was shown to be of a form similar to one based instead on the attainment of a critical accumulated "equivalent" plastic strain everywhere within a small "fracture process zone channel" as the crack tip approaches. The implications of the fracture criterion were shown by determining, with the above values of α and R, the variation of J with $a-a_o$ (a_o = initial crack length) for growth under small scale yielding condition. It was noted that the growth curve depends on the two parameters J_{IC} and T_o, which is the initial value of Paris' "tearing modulus"

$$T = (E/\sigma_o^2)\ dJ/da$$

at the onset of growth under small scale yielding.

Some speculations were also made on the form of the J versus $a-a_o$ relation for large scale yielding, from estimates of R based on demensional considerations for fully yielded specimens (i.e., R approximately proportional to the uncracked ligament size) and of α based on rigid-plastic slip line solutions. It was also indicated that various definitions of J are possible at large scale yielding, and it was shown for fully plastic bend specimens that J based on a far-field contour integral, and not J as defined by the "deformation plasticity" formula, has appropriate properties. The results, although certainly tentative in character and tied to the ideally plastic model, suggest that there may be an appreciable dependence of the J versus $a-a_o$ relation on the extent of yielding. This is a subject on which much further work, both analytical and experimental is needed.

On the experimental side, Asaro, Hermann and Odegaard have begun experimental studies of the J versus a-a relation in 4100 series steels with various heat treatments; some results are described in Section 2b. These materials hold the promise of allowing study of stable growth under a wide range of conditions, from those of small scale yielding to those well into the plastic range. The effort is currently being extended, with the aim of correlating observed results against the theoretical work. The work has led to the development of novel experimental techniques for measuring small increments of crack growth and the "J" integral. The techniques are based upon differential compliance measurements and utilize various types of special "frictionless grips" designed by us for this project. A preliminary description of the apparatus and electronics is given in Technical Report 67.

(Staff: R. J. Asaro, A. Needleman, J. R. Rice, L. Hermann, W. J. Drugan, T. Odegaard, T-L. Sham)

2b. Correlation of Fracture Toughness and Stable Crack Growth with Microstructure

As part of our general study of combined microscopic and macroscopic aspects of fracture in ductile solids we have initiated research this past contract period on correlations between microstructure and fracture toughness in high strength 4140 and 4340 steels. The work involves measurements of fracture toughness and other standard mechanical properties, along with extensive microscopic studies of crack tip deformation and mechanisms of crack initiation. The research makes important specific contact with the previously mentioned work of Rice and co-workers on continuum models for ductile crack growth (section 2a) through direct measurements of such micromechanical features of the crack growth process as crack tip openings, void initiation and coalescence ahead of the growing crack tip, and crack tip opening angles. Along with the determination of the J versus $a-a_o$ relation for conditions of both small and large scale yielding, this information provides the direct experimental counterpart to the research outlined in section 2a. Another aim of this research is to provide direct evaluation of the toughness of particular microstructures which should be of considerable value for alloy designers who seek to achieve optimum properties of strength versus toughness.

To date our research has been concerned with quenched and tempered 4140 steels. There have been recent reports in the literature of significant improvements in plane strain fracture toughness in such steels when they are austenitized at temperatures as high as 1200^{o}C as opposed to more conventional temperatures near 870^{o}C. High austenitizing temperatures generally lead to considerable coarsening of the microstructures in these materials - for example, a tenfold increase in austenite grain size along with a nearly corresponding increase in martensite packet size is commonly reported. Such increases in grain size are generally thought to be detrimental since they are well known to reduce Charpy impact energy and ductility, to increase ductile-brittle transition temperatures, and to increase environmental sensitivity to brittle intergranular fracture. Our steels were tested after austenitizing at 870°C and 1200°C in the as-quenched condition and after tempering at 200, 300, 400, and 500°C. In the as-quenched and 200°C temper conditions the plane strain K_{IC} was about 40 percent larger for the coarse grained microstructures, but with tempering temperatures much above this the finer grained microstructures had higher plane strain fracture toughness. On the other hand, for all tempers the coarse grained specimens displayed significantly reduced ductility and impact energy suggesting that they have a reduced "intrinsic" resistance to fracture. The results were tentatively rationalized by a fracture model akin to that used by Ritchie, Knott and Rice, in Technical Report 20, for cleavage fracture in mild steel. The fracture criterion was modified for the present case by assuming that a critical strain, ε_f^*, must be achieved over a critically large microstructural distance, δ^*. ε_f^* was taken equal to the plane strain tensile ductility, and δ^* as the mean martensite colony size. The latter choice was based upon our microscopic studies of secondary cracks as described in Technical Reports 66 and 67. This leads to a K_{IC} given as

$$K_{IC} = \left[\frac{E\,\delta^*\sigma_{ys}}{0.48(1-\nu^2)I^*}\right]^{1/2}$$

where I^* is the intercept on the distance/crack tip opening axis of McMeeking's plots (Technical Report 44) for plastic strain ahead of blunting crack in mode I opening; σ_{ys} is the yield strength. The intercept corresponds to a plastic strain equal to ε_f^*. I^* increases as ε_f^* decreases and displays a very rapid increase for $\varepsilon_f^* \leq 0.05$; this generally results in a decreasing K_{IC} with increasing strength for ultra high strength steels in opposition to the apparent trend indicated by the σ_{ys} term alone. The effect of increasing δ^* on K_{IC} is evident in the formula. This model provided a very good description of the relative toughness differences following high and low temperature austenitizing temperatures. A full report, in addition to Technical Reports 66 and 67, and which describes our more recent studies of stable crack growth in this alloy is forthcoming.

As described above our results have indicated that high temperature austenitizations do not lead to a general improvement in properties. However, this may well be due primarily to the severe coarsening of the grain size which invariably occurs in commercial steels. On the other hand, our electron replica studies have indicated that a refined carbide and particle substructure may be possible using high austenitizing temperatures and, in itself, this can be expected to have a beneficial effect on strength and toughness. Thus, providing grain size can be kept refined, high temperature solution treatments may be of practical use. To explore these possibilities we are currently evaluating the use of titanium in the deoxidation process of 4340 steels to take advantage of the grain refining of TiN. To date we have been successful in maintaining a grain size of less than 90 μm after austenitizing at 1200°C, but our expectations are to reduce this figure to about 40 μm. Fracture toughness, along with stable crack growth, will then be studied with varying heat treatment but always with a relatively fine grain structure. It should be mentioned that the above procedures should find use in other types of steels such as medium to high strength pressure vessel steels where toughness is also important.

(Staff: R. J. Asaro, J. R. Rice, L. Majno, and T. Odegaard)

2c. Other Studies on the Mechanics of Crack Growth

In a paper (Technical Report 63) based on his general lecture at the 8th U.S. National Congress of Applied Mechanics, Rice reviewed a number of recent studies on the mechanics of quasi-static crack growth. These included studies of the geometry and stability of crack growth paths in brittle solids, thermodynamics of brittle cracking including effects of surface adsorption, and crack growth in materials of non-elastic rheology, especially elastic-plastic materials. They are reviewed briefly here:

Geometry and stability of crack growth in elastic-brittle solids: This involves the use of a first order perturbation procedure developed by Cotterell and Rice for calculating elastic stress intensity factors at the tips of slightly curved or kinked two-dimensional cracks. The results were found to be accurate when compared to exact solutions, up to considerable deviations from straightness, and allowed them to formulate equations governing the path of crack growth (on the assumption, appropriate for isotropic brittle materials, that the shear-mode stress intensity factor always vanishes at the advancing crack tip). The results lead to an experimentally verified criterion for stability, under tensile loading, of a straight crack path. In particular, when the non-vanishing but non-singular stress term at the tip, corresponding to tension parallel to the crack surface, is positive, the crack path is found to curve away from a straight line, but not when the term is negative.

Thermodynamics of elastic-brittle crack growth: Motivated by reported discrepancies between the classical Griffith criterion and results from lattice models for elastic-brittle crack growth, the work reviewed restrictions placed by thermodynamics on phenomena such as lattice trapping of cracks and kinetic crack growth by thermally activated processes, including the case of growth in a surface-adsorbing environment. Under quasi-static isothermal conditions, the principles of thermodynamics were shown to require only that

$$(G-2\gamma)V \geq 0$$

where V is crack growth speed, G the Irwin continuum elastic energy release rate and, for a surface-inactive environment, $2\gamma = 2\gamma_o$, the reversible work of lattice separation. For a surface adsorbing environment,

$$\gamma = \gamma_o - \int_{-\infty}^{\mu} \Gamma(\mu)d\mu$$

where $\Gamma(\mu)$ is the surface excess adsorbed at potential μ of the environmental species, so that 2γ corresponds to Gibbs' expression for the reversible work of surface formation at constant potential. The thermodynamic restriction is compatible with lattice trapping provided that, contrary to speculations otherwise in the literature on lattice models, $G \leq 2\gamma \leq G^+$ (where G^+ corresponds to crack growth and G^- to crack healing). The results also show for the case of thermally activated crack growth, again contrary to speculations otherwise, that the equilibrium value of G (corresponding to V=0 in the kinetic relation V = function of G) must correspond to $G = 2\gamma$.

Paradoxes in energy balance approaches to crack growth: In recent years continuum mechanical solutions have been developed for the growth of mathematically sharp tipped cracks (i.e., crack models which include no account of details of the near-tip processes of separation) in materials with a variety of rheologies. These include linear visco-elastic, non-linear creeping, rate-independent elastic-plastic, and fluid-infiltrated porous elastic solids. Various attempts have been made to extract crack growth criteria from these solutions by generalizing the Griffith approach, setting the energy G released to the crack tip per unit area of crack growth (as calculated from the non-elastic continuum solution) to a characteristic work of fracture (representing work to achieve near tip separation processes ignored in the continuum model). For elastic solids this procedure gives results which are known to be compatible with models based on a small near-tip zone of gradual decohesion. However, a review of these attempts shows the procedure to lead to paradoxical results in every case of crack growth in non-elastic solids thus far examined. For linear visco-elastic and fluid-infiltrated solids, the results are at variance with known limiting cases (e.g., of slow or rapid growth), and differ qualitatively from solutions based on crack models with small near-tip zones of gradual decohesion. For strongly non-linear creeping and rate-independent elastic-plastic solids, the result of the energy flux calculation is G=0 at all load levels and crack speeds, suggesting a paradoxical impossibility of crack growth. The analysis of these various cases shows, of course, that there is no defect in the basic notion of energy balance during growth, but rather that the paradoxes arise from the tacit assumption that the calculation of energy transfers can be decoupled at the continuum level from details, over a small but finite size scale, of the decohesion process.

(Staff: J. R. Rice)

EY-76-S-02-3084 A2 June 1979
Technical Progress Report

3. Fracture at Elevated Temperatures

3a. Stress Fields at Macro-Crack Tips in Creeping Solids

H. Riedel and J.R. Rice (Technical Report 64) developed a first analysis of the stress field created near a crack tip in a tensile loaded solid of elastic-nonlinear viscous material. The non-linear viscous terms involve a power law relation between plastic strain rate and stress, and are intended to represent creep at elevated temperature in metals.

The material is assumed to obey the non-linear "Maxwell" relation

$$\dot{\varepsilon} = \dot{\sigma}/E + B\sigma^n$$

in uniaxial tension, where E and n are constants and B is either a constant or a function of time (intended to represent transient creep, in a very approximate way, by a "time hardening" formulation). The law is generalized to combined stress in the form

$$\dot{\varepsilon}_{ij} = \frac{1+\nu}{E} \dot{\sigma}_{ij} - \frac{\nu}{E} \delta_{ij} \dot{\sigma}_{kk} + \frac{3}{2} B \sigma^{n-1} \sigma'_{ij}$$

where σ'_{ij} is the deviatoric stress and now σ is the equivalent tensile stress

$$\sigma = \sqrt{\frac{3}{2} \sigma'_{ij} \sigma'_{ij}}$$

Cases of stationary cracks, subject to sudden load application at t=0 were considered for plane stress and for plane strain.

Their work answers the question of what loading parameter determines the stress and deformation field near a crack tip, after sudden load application, and hence that of what parameter governs the onset of macrocrack growth under creep conditions. An important result of their analysis is that "small scale yielding" conditions (analogous to those of rate-independent elastic-plastic fracture mechanics) may be defined for sufficiently short times after load application. On this time scale the nonlinear crack tip field is governed, in a way determined by their analysis, by the elastic stress intensity factor K_I, and elastic strains then dominate creep strains everywhere except in a small creep zone, which expands in time. For long times, when the estimated creep zone size is no longer small compared to length scales of the cracked body, the deformation field approaches a state of steady creep flow, with no further alterations of the stress field. In this long time limit the intensity of the near tip field is governed by the value of a path independent integral, denoted by C*, which is the analogue for nonlinear viscous materials of the J integral of nonlinear elasticity (or of rate-independent elasto-plasticity, modelled by a total strain formulation).

In particular, they show that in r,θ coordinates about the crack tip the stress and strain rate fields behave, as $r \to 0$, as

$$\sigma_{ij} = \left[\frac{A(t)}{Br}\right]^{1/(1+n)} \Sigma_{ij}(\theta;n)$$

$$\dot{\varepsilon}_{ij} = B \left[\frac{A(t)}{Br}\right]^{n/(1+n)} E_{ij}(\theta;n)$$

where $A(t)$ is a time-dependent amplitude factor and Σ_{ij}, E_{ij} are functions determined by the analysis (of the same form as the Hutchinson-Rice-

Rosengren fields for power-law strain hardening in the time-independent model), and normalized such that $A(t) \to C^*$ in the long time limit of steady creep, $t \to \infty$. For short times they prove that A(t) must vary in proportion to K_I^2/Et and show, by an approximate analysis based on the J integral, that

$$A(t) \simeq K_I^2/(1+n)Et$$

for plane stress; a factor $1-\nu^2$ must be inserted for plane strain. They use this solution to present calculations of the creep zone boundary for short times; the creep zone is defined as the zone in which the Mises equivalent creep strain exceeds the corresponding equivalent elastic strain, and grows in time about the crack tip in proportion to

$$K_I^2 \; (EBt)^{2/(n-1)}$$

For the case of time-dependent B, the term Bt is merely replaced by $\int B(t)dt$.

Based on the short and long time solutions they define a characteristic time

$$t_1 \approx K_I^2/[(1+n)E\ C^*]$$

was defined, such that K_I controls the crack tip field for times that are small compared to t_1, and C* for times that are large compared to t_1.

In continuing studies, still underway, Riedel, Rice, and Drugan are examining analogous near-tip fields for quasi-statically growing cracks in non-linear viscous materials. These studies suggest that for sufficiently large creep exponents ($n>3$), a new type of singular field, in which stress and strain vary in proportion to $r^{-1/(n-1)}$, where r is distance from the tip, is created. This field is thought to have, under typical conditions, a very limited spatial domain of validity, with a transition to field analogous in character to those associated with stationary cracks taking place over a length scale in the material which increases non-linearly with crack speed.

(Staff: J. R. Rice, W. Drugan, in collaboration with H. Riedel - MRL Postdoctoral Fellow, on leave from Düsseldorf)

3b. Processes of Diffusive Cavitation of Grain Boundaries Under Creep Conditions

Existing models for the diffusive cavitation of grain interfaces during elevated temperature creep have for the most part neglected grain deformability through plastic creep flow. The classical Hull-Rimmer model treats the grains as rigid bodies and a limited amount of work (e.g., Chuang, Raj, Vitek, J. and J.R. Weertman) has been done on models in which the grains are assumed to deform elastically. The general conclusion from these latter studies is that elastic deformability is relatively unimportant under sustained load conditions (although it can be very significant in transient stressing following load alterations), and this has led to the suspicion that grain deformability is relatively unimportant for the sustained load case.

However, in recent work Rice (paper at DOE Workshop on Time-Dependent Fracture, February 1979) has demonstrated an important coupling between plastic creep flow and grain boundary diffusion. The coupling results, under sufficiently high stress and temperature conditions, in greatly increased rates of cavity enlargement by comparison to predictions from models of the cavitation process mentioned above in which grain deformation is neglected and the cavities grow simply by diffusive transfer of matter from the cavity surfaces into the grain boundary. The coupling arises because deformability of the grains allows matter diffused from the cavity surfaces to be accommodated by highly localized separation velocities across the grain boundary, limited to the immediate vicinity of the cavities. Hence the diffusion path length can be much shorter than a representative half-spacing between cavities, resulting in far more rapid removal of material from the cavity surfaces at a given level of applied stress. Further, at sufficiently high stress levels the voids can grow at a rate that becomes independent of grain boundary diffusion, since the plastic creep flow process enlarges the cavity volume and surface diffusion (generally rapid compared to grain boundary diffusion) continually maintains a spherical-caps cavity shape, thus enlarging the cavity radius. The coupling effect can be represented through a stress level and temperature dependent parameter L, of length dimensions, where

$$L = [D_b \delta_b \; \Omega\sigma / \dot{\varepsilon} kT]^{1/3}$$

(here $D_b \delta_b$ is the grain boundary diffusion coefficient, Ω the atomic volume, kT the energy measure of temperature, σ the applied tensile stress, and $\dot{\varepsilon}$ the corresponding plastic creep flow rate).

Rice presented tabulations of numerical estimates of L for various pure metals at 0.5 T_m (T_m = melting temperature) and 0.8 T_m. For example, at a stress level σ of 10^{-3} μ (μ = shear modulus) L is typically in the range of 0.2 to 5 μm, and of the order of 13 μm for Ag and Ni; it increases by approximately a factor of 20 when σ is decreased to 10^{-4} μ. At a given stress level L decreases by factors ranging from 4 to 20 when the temperature is increased from 0.5 T_m to 0.8 T_m.

The significance of L is as follows: when L is much larger than the cavity radius (a), the classical Hull-Rimmer model applies and plastic creep flow is unimportant. When L is comparable to cavity radius (say, 20a > L > 0.2a) the coupling between creep and grain boundary diffusion mentioned above is important and leads to cavity growth rates which become, with decreasing L, very much larger than predictions of the Hull-Rimmer model. Finally, when L is much smaller than a, grain boundary diffusion becomes unimportant and growth is dominated entirely by creep flow, at rates which exceed the Hull-Rimmer predictions by factors on the order of $(a/L)^3$. For example the results reported by Rice based on a preliminary analysis of a very approximate "shear plate" model suggested that in the case of cavities spaced at 10 times their diameter, the growth rate a would exceed the Hull-Rimmer predictions by a factor of about 3 when $L = 3a$, and about 10 when $L = a$.

Thus, using the values for L noted above and choosing L/4 as the least value of a for which creep effects can be neglected, these tentative results suggest that at $\sigma = 10^{-3}\mu$ and $0.5\ T_m$, a must typically be smaller than 0.05 to 1 μm (but 3 μm for the more creep resistant materials, Ag and Ni). When the stress is lowered to $10^{-4}\mu$, a must typically be smaller than 1 to 20 μm. These last results, although tentative in character, suggest that at the very low stress levels representative of long-lifetime components for high temperature applications, there may frequently be only very minor contributions to cavity growth from plastic creep flow.

A more precise analysis of the effects of combined power-law creep flow and grain boundary diffusion is now being carried out by Needleman. He has developed a finite element program for this class of problems on the basis of a variational principle developed by Rice. The problem under study consists of a spherical-caps cavity on a grain boundary at the mid section of a round bar loaded in tension. Boundary conditions imposed on the lateral surfaces of the bar are intended to model the effect of neighboring cavities on a grain boundary, with the bar diameter corresponding to the mean cavity spacing. In particular, the radial contraction rate on the lateral surface is taken to be uniform along the length of the bar, and the lateral surface is free of shear stress. Although the work is still underway, a family of meshes has been designed, a number of coarse-mesh preliminary computer runs have been made, and the numerical results of a fine-mesh run agree very closely with the known Hull-Rimmer limit for the case when L is much larger than a.

A number of other studies on diffusive rupture are also underway. For example, Rubenstein and Rice are studying the diffusive growth of thin crack-like cavities (the limiting case of low surface diffusivity) along a grain boundary in a material with, for mathematical simplicity, a linearly viscous rheology. Chuang (now at Westinghouse) and Rice are working on a manuscript on the discussion of boundary values of chemical potentials and calculations of energy release rates for the diffusive growth of grain boundary cavities. These are subjects on which there is not full agreement in the published literature, and an attempt is being made to clarify and examine the validity of different approaches and interpretations.

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

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

1. Technical Reports

COO-3084-61 R. J. Asaro, J. Gurland, A. Needleman and J. R. Rice, Technical Progress Report for July 1977 - June 1978, (June 1978).

COO-3084-62 J. R. Rice, R. M. McMeeking, D. M. Parks and E. P. Sorensen, "Recent Finite Element Studies in Plasticity and Fracture Mechanics" (August 1978).

COO-3084-63 J. R. Rice, "The Mechanics of Quasi-Static Crack Growth" (October 1978).

COO-3084-64 H. Riedel and J. R. Rice, "Tensile Cracks in Creeping Solids" (February 1979).

COO-3084-65 J. R. Rice, W. J. Drugan, and T. L. Sham, "Elastic-Plastic Analysis of Growing Cracks" (May 1979).

COO-3084-66 T. K. Odegaard and R. J. Asaro, "Effects of Austenitizing Temperature on Fracture Toughness of 4140 Steel" (June 1979).

COO-3084-67 T. K. Odegaard, "Correlation Between Fracture Toughness and Microstructure in 4140 Steel" (June 1979).

COO-3084-68 J. Gurland, "The Contribution of Load Transfer to Internal Stresses in Dispersed Particles" (June 1979).

2. Publications

*Y. W. Chang and R. J. Asaro, "Bauschinger Effects and Workhardening in Spheroidized Steels," Metal Sci. 12 (1978) p. 277.

*E. L. Exner, J. R. Pickens and J. Gurland, "A Comparison of Indentation Crack Resistance and Fracture Toughness of Fine WC-Co Alloys," Met. Trans. 9A (1978) p. 736.

*E. L. Exner, J. R. Pickens and J. Gurland, Indentation Crack Resistance by the Palmquist Method, Proceedings, 5th European Symposium on Powder Metallurgy, Jernkontoret, Stockholm, Sweden (1978) p. 63.

*A. Needleman and J. R. Rice, "Limits to Ductility Set by Plastic Flow Localization", in Mechanics of Sheet Metal Forming (Proceedings of General Motors Research Laboratories Symposium, October 1977, ed. D.P. Koistinen and N-M. Wang), Plenum Press (1978) pp. 237-267.

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

A. Needleman and C. F. Shih, "A Finite Element Method for Plane Strain Deformations of Incompressible Solids," Comp. Meth. Appl. Mech. Eng'g., 15, pp. 223-240 (1978).

*J. R. Rice and E. P. Sorensen, "Continuing Crack Tip Deformation and Fracture for Plane-Strain Crack Growth in Elastic-Plastic Solids", Journal of the Mechanics and Physics of Solids, 26, 1978, pp. 163-186.

T-j. Chuang, K. I. Kagawa, J. R. Rice and L. B. Sills, "Non-equilibrium Models for Diffusive Cavitation of Grain Interfaces", Acta Metallurgica, Overview Paper No. 2, 27, 1979, pp. 265-284.

J. R. Rice, R. M. McMeeking, D. M. Parks, and E. P. Sorensen, "Recent Finite Element Studies in Plasticity and Fracture Mechanics", in Proceedings of the FENOMECH '78 Conference (Stuttgart, edited by K.S. Pister et al.), North Holland Pub. Co., Vol. 2, 1979, pp. 411-442; also Computer Methods in Applied Mechanics and Engineering, 17/18, 1979, pp. 411-442.

H. Riedel and J. R. Rice, "Tensile Cracks in Creeping Solids", presented at ASTM 1979 Symposium on Fracture Mechanics, in review for publication by ASTM.

J. R. Rice, "Plastic Creep Flow Processes in Fracture at Elevated Temperatures", Proceedings of Department of Energy Workshop on Time-Dependent Fracture of Materials at Elevated Temperature (Germantown, Md., Feb. 1979) - in press.

*D. D. Mason, "Segregation Induced Embrittlement of Grain Interfaces", Philosophical Magazine - in press.

*H. Yamamoto, "Conditions for Shear Localization in the Ductile Fracture of Void-Containing Materials", International Journal of Fracture, 14, 1978, pp. 347-365.

*E. P. Sorensen, "A Numerical Investigation of Plane Strain Stable Crack Growth Under Small Scale Yielding Conditions", ASTM Special Technical Publication, ASTM, Philadelphia, STP 668, 1979.

*H. Cialone and R. J. Asaro, "The Role of Hydrogen in the Ductile Fracture of Plain Carbon Steels," Met. Trans. 10A (1979) p. 367.

T. K. Odegaard and R. J. Asaro, "Correlations Between Fracture Toughness and Microstructure in 4140 Steel," to be submitted to Met. Trans.

J. R. Rice, "The Mechanics of Quasi-Static Crack Growth," in Proc. 8th U. S. National Congress of Applied Mechanics, ed. R. E. Kelly, Western Periodicals Co., No. Hollywood, 1979.

J. R. Rice, W. J. Drugan, and T-L. Sham, "Elastic Plastic Analysis of Growing Cracks," presented at ASTM 1979 Symposium on Fracture Mechanics, in review for publication by ASTM.

J. W. Hutchinson, A. Needleman and C. F. Shih, "Fully Plastic Crack Problems in Bending and Tension," Proc. of the ONR Symposium on Fracture Mechanics (ed. N. Perrone et al.) pp. 515-527, (1978).

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

3. Theses

T. L. (Sam) Sham, Sc.M., June 1979, Thesis: "A Finite Element Analysis of Quasi-Static Crack Growth in Elastic-Perfectly Plastic Solids."

T. K. Odegaard, Sc.M., Thesis (June 1979): "Correlations Between Fracture Toughness and Microstructure in 4140 Steel."

4. Oral Presentations and Seminars

a. Oral Presentations by J. R. Rice

Lead Lecture at ASTM 12th Annual Symposium on Fracture Mechanics (St. Louis), May 1979; topic: Elastic-Plastic Analysis of Growing Cracks.

John E. Dorn Memorial Lecture, Dept. of Materials Science, Northwestern University, May 1979; topic: Studies on the Mechanics of Crack Growth.

General Lecture, 20th Polish Solid Mechanics Conference (Kozubnik, Poland), September 1978; topic: Elastic-Plastic Fracture Mechanics.

General Lecture, British Theoretical Mechanics Colloquium (Bath, England), April 1979; topic: Studies on the Mechanics of Crack Growth.

Invited Lecture, Conference on Finite Elements in Non-linear Mechanics (Stuttgart), August/September 1978; topic: Recent Finite Element Studies in Plasticity and Fracture Mechanics.

General Lecture, U.S. National Congress on Theoretical and Applied Mechanics (Los Angeles), June 1978; topic: Mechanics of Quasi-Static Crack Growth.

Talk at Rockwell/AFOSR Conference on Non-Destructive Evaluation of Materials (La Jolla), July 1978; topic: Long Wavelength Defect Characterization.

Talks at ARPA Materials Research Council Meeting on Interfaces (La Jolla), July 1978; topics: Adsorption and Embrittlement of Material Interfaces; and, Crack-like Modes of Diffusive Rupture.

Talk at Gordon Conference on Ceramics (Tilton, New Hampshire), August 1978; topic: Non-equilibrium Models for Diffusive Cavitation of Grain Interfaces.

Talk at DOE Workshop on Time-Dependent Fracture (Germantown, Md.), February 1979; topic: Plastic Creep Flow Processes in Fracture at Elevated Temperature.

Talk at ASTM E24:08:01 Task Group Meeting (Pittsburgh), February 1979; topic: Tensile Cracks in Creeping Solids.

b. Seminars by J. R. Rice

On the Topic: Studies on the Mechanics of Crack Growth, at:

Institute for Mechanical Problems, Moscow; Sept. 1978
Brown University; November 1978
SUNY, Stony Brook; October 1978
University Houston; November 1978
Institute Fundamental Tech. Research, Warsaw; April 1979
University of Illinois-Urbana; May 1979

On the Topic: Surface Thermodynamics and Grain Boundary Cracking, at:

Cornell University, October 1978

On the Topic: Localization of Plastic Flow; and, Finite-Element Analysis at Large Plastic Deformation, at:

Institute Fundamental Tech. Research, Warsaw; April 1979

On the Topic: Processes of Elevated Temperature Creep Cavitation, at:

Brown University, January 1979

c. Oral Presentations by R. Asaro

"Role of Hydrogen on the Ductile Fracture of Plain Carbon Steels" ARPA Materials Research Council Meeting on Interfaces (La Jolla) July 1978.

"Environmental Effects in Ductile Fracture" Sandia Laboratories, Livermore, California, July 1978.

"Effect of Hydrogen on the Ductile Fracture in Plain Carbon Steels", U. S. Department of Energy Workshop on the Effects of Hydrogen on the Mechanical Behavior of Metals, Gaithersburg, November, 1978.

"Localized Plastic Deformation and Fracture in Single Crystals", 1979 EuroMech Colloquium on Flow and Fracture in Solids, Jablonna, Poland, June, 1979.

"Localization of Plastic Deformation in Single Crystals", Stanford University, July 1978.

d. Oral Presentation by J. Gurland

"Strength of Two-Phase Alloys and Load Transfer to Particles", Brown University, January 1979.

e. Oral Presentation by A. Needleman

"Incompressible Finite Elements in Plain Strain", Can. Congress Appl'd Mechanics, Sheabroure, Quebec, June 1979.

EY-76-S-02-3084 A2
Technical Progress Report

June 1979

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 Assistants: H. Stanton, W. Rebello.

4. Research Assistants (Graduate Students): H. Cialone, W. J. Drugan, J. Fisher, T. Odegaard, S. Sham** and A. Rubenstein*.

5. Undergraduate Students (in 4th year of 5-year B.S. - Sc.M. program): A. Szewcyk* and L. Majno*.

* Supported for portion only of contract period.

** Partially supported by contract.