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A MICROSTRUCTURAL APPROACH TO FATIGUE CRACK
PROCESSES IN POLYCRYSTALLINE BCC MATERIALS

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I. STRAIN-RATE SENSITIVITY AND DISLOCATION DYNAMICS

As was indicated in the proposal, we planned to evaluate two commercial LAHS steel alloys, four Fe-base systems and one model material, a BCC Ti-30 Mo alloy. It was also indicated that to fill necessary gaps in the strain-rate, temperature sensitive behavior that change of strain rate measurements on Ti-30 Mo was required. Although we did not specifically suggest that the thermal activation parameters for LAHS steels were required (Table 3 of the Program Test Plan), it has since become obvious that we must, since there is an apparent grain size effect on $\partial \ln \dot{\epsilon} / \partial \ln \sigma^*$. Whatever the reason for this variation in m^* , it may affect several thermal activation parameters, resulting in different dislocation velocity characteristics as a function of stress. Over the last year, two graduate students (Kumar Jatavallabhula and James Lucas) have been involved in this phase of the research.

I-1. Dislocation dynamics of Fe-binary alloys.

William W. Gerberich, David G. Atteridge and Y.T. Chen

In addition to the four alloys suggested in the original proposal, three Fe-Si alloys have been obtained in case the grain size phase of the program becomes difficult to address in a fundamental way.⁺ All alloy compositions are shown in Table I-1, and it may be shown that these compositions are exactly as outlined in the original proposal except for the small aluminum addition to control deoxidation practice. The iron and steel making division of Republic Steel supplied 22 kg vacuum-induction melted heats, forged to 4.4 cm thick slabs at 1230C and then rolled to 1/4" thick plate at 1200C. These plates are currently being machined into test coupons.

⁺That is, if the activation enthalpy changes with grain size (which it should not in our present understanding of a single thermally-activated process), then there are intrinsic changes in the defect interactions which may not be easily interpreted.

Although not presently working on the project, two former associates, Dr. David G. Atteridge of Battelle Northwest Laboratories and Dr. Y. T. Chen of Pitney-Bowes Corporation have collaborated on the dislocation dynamics program. Further analysis of the existing data for the Fe-binary systems has been accomplished, the resulting flow sensitive model being given in Appendix A-1. The important result is that for a given temperature, strain rate and alloy addition, the yield stress may be predicted as shown in Figure 1.

Table I-1

Heat No.	Analysis							
	C	Mn	Si	P	S	Ni	Ti	Al
Y920	.007	.01	.01	.005	.005	-	.08	.02
Y921	.02	.01	.01	.005	.005	1.08	.10	.02
Y922	.02	.01	.01	.005	.005	2.50	.09	.02
Y923	.013	.01	.01	.005	.005	4.05	.10	.021
Y924	.011	.01	.45	.005	.005	-	.09	.021
Y925	.009	.01	1.20	.005	.005	-	.10	.017
Y926	.02	.01	1.90	.005	.005	-	.12	.018
Y927	.008	-	-	-	-	-	-	.017

I-2. Strain-rate and Flow Sensitive Properties of LAHS Steels
James Lucas and William W. Gerberich

The effects of grain size and temperature on various flow sensitive properties such as σ_{ys} , m' and m^* of HSLA (high strength low alloy) steel is currently being investigated. Recent evidence strongly reveals that the aforementioned parameters are very significant in the analysis of various aspect of fatigue crack growth.^(1,2,3,4) This is especially true for threshold fatigue where microstructure, in particular grain size, play a major role. It has been proposed⁽⁵⁾ that σ_{ys} , m' and m^* are essential parameters in fatigue crack growth at low temperature in low alloy steel.

Figure 1: Effect of Si on yield stress of Fe-Si and prediction model for temperatures (77-363K)

Of three special heats obtained from Inland Steel, two are low Nb heats, one treated with rare earths for inclusion control and one not, but both having as-received yield strengths on the order of 58 Ksi (400 MPa). The third heat, with higher Nb and Mn concentrations, exhibits a higher yield strength near 87 Ksi (600 MPa). Compositions of these are outlined in Table I-2.

Table I-2: Composition of HSLA Steels (wt. %)

	C	Nb	Mn	Fe	
Heat #1	0.06	0.03	0.35	bal	<u>RET*</u>
Heat #2	0.06	0.03	0.35	bal	<u>non-RET</u>
Heat #3	0.06	0.12	1.00	bal	<u>non-RET</u>

*Rare-earth treated

Thus far, the first years efforts have centered primarily on the lower strength, low Nb heats.

Flow sensitive parameters, m' and m^* , the strain rate sensitivity exponent and stress-velocity exponent respectively, revealed a significant grain size and temperature dependence. Many investigations of the temperature dependence on m^* and m' are readily found in the literature^(6,7) for low alloy steels. However, the effects of grain size on m^* and m' are not so prevalent. The parameters m' and m^* were determined by performing strain rate cycling tests with the strain rate differing by a factor of 10. Low strain rate (1.66×10^{-4} sec) to high strain rate (1.66×10^{-3} sec) cycling was utilized at test temperatures from 300 K to 150 K. The grain size tested ranged from 10 to $\approx 150 \mu\text{m}$. Figure 1 shows that m' , $m' = \partial \ln \dot{\epsilon} / \partial \ln \sigma_{\epsilon_p} \rightarrow 0$, decreases as the temperature decreases and displays an inverse grain size dependence. In order to determine m^* , the temperature independent plateau stress, σ_i , must initially be found for each grain size. It is seen in Figure 2 that the flow stress

decreases with increasing temperature for any particular grain size. At a temperature near 375K, the plateau stress is reached. As expected, the plateau stress, σ_i , varies directly with grain size. Because $\sigma^* = \sigma - \sigma_i$, m^* can be easily calculated from the definition of m^* , that is $m^* \equiv \partial \ln \dot{\epsilon} / \partial \ln \sigma^*$ as $\epsilon_p \rightarrow 0$. In Figure 3, m^* is shown to vary with temperature in the usual fashion for low alloy steel. Also there is no doubt that the grain size effect on m^* is quite substantial. This investigation shows that m^* decreases as the grain size is increased, but the origin of this behavior is not understood. Further evaluations of substructure and independent evaluations of m^* and σ_i are required.

References (Section I-2)

1. T. Yokobori, S. Konosu and A.T. Yokobori, Jr., Fracture 1977, 1, Waterloo, Canada, p. 665.
2. T. Yokobori, ASTM Symposium of Fatigue Mechanism, May 22-24, 1978, Kansas City.
3. W.W. Gerberich and N.R. Moody "A Review of Fatigue Fracture Topology Effects on Threshold and Growth Mechanism" ASTM, STP 675, 1979.
4. N.R. Moody and W.W. Gerberich, Material Sci. & Eng., Vol. 41, p. 271, 1979.
5. W.W. Gerberich, N.R. Moody and K. Jatavalabhula, Scripta Met., Vol. 14, p. 113, 1980.
6. Y.T. Chen, Ph.D. Thesis, University of Minnesota, 1976.
7. J.T. Michalack, Acta. Met., Vol. 13, 1965.

I-3 Dislocation dynamics of the Ti-30 Mo model system.
Kumar Jatavalabhula and William W. Gerberich

The present investigations on Ti-30 Mo are concerned with determining the dislocation dynamic parameters and transition temperatures from the ductile to cleavage mode of fracture at slow and fast strain rates. A DBTT model previously proposed by Gerberich, et. al.⁽¹⁾ is employed to predict the transition

Fig.1

Fig. 2

temperatures using the experimental dislocation dynamic parameters. The predicted temperatures from the model are in good agreement with the experimental temperatures and are found to be only three to five percent higher.

All specimens of Ti-30 Mo (compression, tensile and Charpy) were solution treated at 1000C for 4 hours in a vacuum of 10^{-5} torr. The compression and tension samples were used to obtain the dislocation dynamic parameters by strain rate cycling between $1.66 \times 10^{-4} \text{ sec}^{-1}$ to $1.66 \times 10^{-3} \text{ sec}^{-1}$. The Charpy V-notch samples were tested both under four point loading and impact to obtain the ductile to brittle transition temperatures at slow and fast strain rates respectively.

As expected for the BCC class of materials, the yield stress and strain rate sensitivity exponent in Ti-30 Mo show a strong temperature dependency. Strain rate sensitivity exponents determined from tension and compression testing are very nearly the same with the values from compression evaluations being shown in Figure 1 as a function of temperature. The Peierls stress in this material to be 1780 MPa. The activation volume of $35 b^3$ (b is the Burgers vector) suggests a thermally activated control of dislocation movement by double-kink nucleation. However, the activation energy for plastic deformation (ΔG_0) is about 1.1-1.4 eV and seems high for this process. The variability arises from how one treats a hump in the flow stress versus temperature curve, the details of which are presently being confirmed. The DBTT in slow notch bending (SNB) is the temperature at which the general yield curve and the fracture load curves intersect. This is shown in Fig. 2 for Ti-30 Mo and the temperature is 238 Kelvin. Under impact the DBTT is shifted higher and is found to be 370K. Both these temperatures are found to be in close agreement with the predicted values from the dislocation dynamics model. The final equation in this model is only mentioned as shown below.

Figure 1; Effect of temperature on strain-rate sensitivity.

Figure 2: Determination of cleavage fracture stress from slow notch bend testing at different temperatures,

$$T_{\text{SNB}} \text{ or } T_{\text{impact}} = \frac{\theta' \Delta G_o}{k} \left\{ \frac{\sigma_p^* - \left(\frac{\sigma_f^*}{pcf} - \sigma_i \right)}{\sigma_p^* + \left(\frac{\sigma_f^*}{pcf} - \sigma_i \right)} \right\} \frac{1}{\ln\left(\frac{\dot{\epsilon}_o}{\dot{\epsilon}}\right)}$$

$\dot{\epsilon}_o$ is the preexponential factor in the rate equation of the thermally activated process. This value has been found to be $4.71 \times 10^{12} \text{ sec}^{-1}$ from the slope of the activation energy and temperature plot. In addition θ' has been chosen to be 0.52 and corresponds to an $\alpha=-1$ Peierls barrier. A plastic constraint factor of 2.34 is chosen in both SNB and impact. Using $\dot{\epsilon}$ to be 200 sec^{-1} and $1.66 \times 10^{-4} \text{ sec}^{-1}$ for impact and SNB the temperatures are found to be 388 and 247 K, respectively, which correspond well to the measured values of 370 and 238K.

References (Section I-3)

1. W.W. Gerberich, Y.T. Chen, D.G. Atteridge and T. Johnson, "Dislocation Dynamics of Fe-Binary Alloys: II. Application to the Ductile-Brittle Transition," Submitted, Acta. Met., (1980).

II. FATIGUE AND FRACTURE CHARACTERIZATION

For analysis of fatigue crack propagation models, particularly at low temperatures, cyclic strain-hardening exponents and cyclic cleavage planes must be determined. The first year's effort has mainly centered on improving the cyclic stress-strain technique, using the LAHS steels. For microstructural analysis, the need for establishing the local grain orientation was discussed in the initial proposal. Since electron channelling is the best method for analysing this, initial efforts have centered on the large grain (100 μm), BCC titanium alloy. We have largely been perfecting the technique on cleaved and electroetched samples of Ti-30 Mo, although some fatigue crack samples are currently under investigation. Over the last year, two graduate students (Rex King and James Lucas) have been involved in this phase of the research.

II-1. Cyclic stress-strain of LAHS steels.
James Lucas and William W. Gerberich

Low cycle fatigue tests were performed on HSLA steel to investigate the effect of temperature and grain size on the cyclic strain hardening exponent, β_c . Tests were performed on threaded tensile specimens with a gage length of 1.27 cm and a gage section diameter of 0.32 cm. The range in grain size was 10-133 μm . It was found the yield stress (σ_{ys} at 0.2%) varied with grain size in accordance with the Hall Petch relationship as illustrated in Figure 1. During heat treatment to achieve the desired grain size, a conscientious effort was made to maintain constancy in other microstructural features so as to maximize the grain size effect. The cyclic stress-strain curve and hence the cyclic strain hardening exponent were determined by utilizing the "multiple step" method.⁽¹⁾ Using this method, the cyclic stress-strain curve is found by connecting the tips of stabilized hysteresis loops at constant plastic strain amplitude. It was noted during testing that the hysteresis loops stabilized rather quick, within 8 cycles or less at constant strain amplitude. In all test ran between $\pm 2.0\%$ plastic strain, cyclic strain hardening occurred. A typical σ vs. ϵ_p hysteresis loop for the "multiple step" method is shown in Figure 2. The cyclic strain hardening exponent, β_c , vs. grain size at 300 and 233 K is represented by Figure 3. Preliminary results revealed that β_c may be slightly affected by grain size at these two test temperatures. Values of β_c that lie within the scatter band range from ~0.10 to 0.20. It appears that β_c has a tendency to increase as the grain size increases. If this is verified, it would be consistent with β_c increasing with grain size as has been reported by another investigation.⁽²⁾ Investigations are planned for examining the temperature effect on β_c at 191 and 150K also, but the results are not complete at this time.

HSLA Steel

Although in the range of expected values of β_c , the scatter was sufficient to warrant reevaluation of the test method. It was noted that the initial gage length (L) to diameter (D) ratio of four was such that buckling was likely to occur at large cyclic strains. In addition, there were relatively large distances between the upper and lower crossheads necessitated by the environmental chamber for temperature control. These two factors probably combined to give:

- i) non-axiality of loading;
- ii) concomitant buckling at large cyclic strains.

It was realized that keeping the L/D to two would prevent the buckling problem. In addition, new instrumentation for strain measuring over short gage lengths and a stiff die-set for providing good concentricity of loading were obtained.⁺ With this new system we have successfully obtained reversed plastic strains of greater than four percent with corresponding cycles to failure of less than 200. In addition no anomalies were noted in the hysteresis loops and the gage length maintained its axial quality to failure.

References (Section II-1)

1. J. Dean Morrow, Internal Friction, Damping and Cyclic Plasticity, ASTM, STP 378, 1965.
2. T. Yokobori, H. Ishii and S. Koyama, Scripta Met., Vol. 13, p. 515, 1979.

II-2. Electron Channelling studies. Rex King and William W. Gerberich

For the Ti-30 Mo alloy solution treated at 1000°C for 4 hours under 10^{-5} torr vacuum, we have been developing electron channelling techniques. As the

⁺Some of the funds originally intended to purchase alloy heats were utilized to provide the new instrumentation. The Fe-binary alloys were denoted by Republic Steel Corporation, courtesy of R. Ault.

surface condition of a channelling specimen can affect pattern quality,⁽¹⁻³⁾ the specimens were mechanically polished with 0.05 micron powder and then chemically etched to remove damage. The specimens were immediately inserted into the electron microscope to avoid excessive oxide contamination.

The 40 keV electron beam of a JEOLCO 100C scanning transmission electron microscope can be made to change its angle of incidence to the specimen surface at a point with a rocking motion. This gives rise to an electron channelling pattern from an area substantially smaller than the 100 μm grain size. From a parallel study⁽⁴⁾ on hydrogen embrittlement of the same alloy, we have developed an electron channelling map on this alloy as shown in Fig. 1(a) and 1(b). This particular map was generated by rotating the specimen known amounts and recording overlapping patterns. The aim is to analyse the cyclic-cleavage path at low K and discontinuous cleavage at high K. This could be done by channelling on a surface normal to the macroscopic crack.⁽⁵⁾ It may also be possible to channel directly on fatigue fracture surfaces and we are exploring this possibility with as-fractured and electrochemically etched surfaces.

References (Section II-2)

1. E.M. Schulson, C.G. Van Essen and D.C. Joy, Scanning Electron Microscopy/1969, Chicago, Illinois, p. 45 (1969).
2. G.R. Booker, Modern Diffraction and Imaging Techniques in Material Science, Ed. by S. Amelinckx, et al., North Holland Publishing Co., p. 613, (1970).
3. E.M. Schulson, in Electron Microscopy and Structure of Materials, ed. by G. Thomas, et al., Berkeley, CA, p. 286 (1972).
4. W.W. Gerberich, "Macro/micro studies of hydrogen embrittlement mechanisms in titanium and aluminum alloys," Report No. 0637-5901-79-1, AFOSR, Washington (1979).
5. R. Stickler and G.R. Booker, Electron Microscopy and Structure of Materials, ed. by G. Thomas, et al., Berkeley, CA, p. 301 (1972).

Figure 1(a): Electron Channelling Map of Ti-30 Mo.

Fig. 1b - Electron channelling map for Ti-30 Mo.

III. FATIGUE CRACK GROWTH AT INTERMEDIATE ΔK

In the first year, significant gains in the understanding of the cyclic cleavage process have been made. This has culminated in two publications which still represent initial thinking on the physical mechanism and the micro-mechanical modelling, but nevertheless show:

- i) the high exponents on ΔK at low temperatures in Fe and its binary alloys depends on both cleavage and dislocation dynamics;
- ii) cyclic cleavage is the dominant mechanism promoting large growth rates below the DBTT;
- iii) cyclic cleavage striation measurements demonstrate that fatigue crack growth rate may be at least 100 times faster than the corresponding ductile striation mechanism occurring at the same temperature and stress intensity level.

Two students, James Lucas and Kumar Jatavallabhula, have contributed towards this phase of the project this last year.

III-1. Experimental Observations of Cyclic Cleavage Neville R. Moody, James Lucas, Kumar Jatavallabhula and W. W. Gerberich

Fractographic analysis of fatigue crack propagation in alloys similar to those in Table I-1 has been underway since the inception of the program. For the first time, this has produced clear evidence of step-wise advances of cyclic cleavage in polycrystalline microstructures of these binary systems. Even before it was entirely clear that this was the controlling mechanism, we had hypothesized that the combined effects of alloying elements on solid solution softening or hardening and the cleavage fracture stress affected the exponent on ΔK by a cyclic cleavage mechanism. That is, the value of n in

$$da/dN = C(\Delta K)^n$$

which controls the sensitivity of fatigue crack growth rates to stress intensity in region II, can be increased drastically by cyclic cleavage below the ductile-brittle transition temperature (DBTT). This is emphasized in Fig. III-1 where the value of n can increase from 4 to 12 by dropping the test temperature 100°C below the DBTT. An analysis of data from seven alloys tested at temperature ranging from 123 to 296K in the intermediate K regime of 12 to 30 $\text{MPa}\cdot\text{m}^{\frac{1}{2}}$ are given in Ref. 1 and shown in Appendix A-2.

III-2 Theoretical Modelling of Cyclic Cleavage

William W. Gerberich, N.R. Moody and K. Jatavallabhula

The initial theoretical model in Reference 1 was based upon dislocation dynamics and the assumption that the formation of cleavage rivers was the rate limiting step, i.e.

$$\underset{\text{microcleavage}}{da/dN} = \alpha(\bar{h})^2$$

With experimental verification that cyclic cleavage was controlling, further analysis of the cleavage step height problem demonstrated that the above equation could be derived from first principles in terms of geometrically necessary dislocations.⁽²⁾ That is, the macroscopic crack opening displacement had to be accommodated by local plastic displacements along each river and there in turn could be calculated in terms of dislocation emission along each river. This is schematically shown in Fig. III-2. By analysing the local crack a range in terms of the local displacement as related to the average cleavage step height along n_R rivers, the microcleavage growth rate was

$$\left. \frac{da}{dN} \right|_{\text{microcleavage}} = \frac{\pi \rho_E b E n_R}{4 \sigma_{ys} (1-\nu^2) \omega^2} [\bar{V}]^2$$

It is seen that the emitted density of dislocations, ρ_E , the burgers vector, b , the modulus to yield strength ration, E/σ_{ys} , cyclic frequency, ω , and the average dislocation velocity near the crack tip, $\bar{V}$, all control the rate of cyclic cleavage. One point of caution is that da/dN is not proportional to $1/\omega^2$ since $\bar{V}$ also depends on rate through the cyclic frequency. A first approximation gives

$$\left. \frac{da}{dN} \right|_{\text{microcleavage}} \propto \omega^{\left\{ \frac{2m^*}{m} - 2 \right\}}$$

For pure polycrystalline Fe this varies from about ω^{-1} to $\omega^{-0.2}$ but there is no experimental evidence to confirm or deny such a variation. There is some indirect evidence⁽³⁾ and this will be further discussed in next year's proposal.

A full solution to the microcleavage growth rate problem may be obtained from the above equations by incorporating an equation of state appropriate to the crack tip. Such an approach which includes strain, strain rate, strain hardening and strain-rate sensitivity was derived and is presented in detail in Appendix A-3.

References (Section III)

1. N.R. Moody and W.W. Gerberich, "Fatigue Crack Propagation in Iron and Two Iron Binary Alloys at Low Temperatures," Materials Science and Engineering, 41 (1979) pp. 271-280.
2. W.W. Gerberich, N.R. Moody and K. Jatavallabhula, "A Proposed Cyclic Cleavage Crack Growth Model for Fatigue of BCC Iron and its Alloys," Scripta Metallurgica, 14 (1980) pp. 113-118.

3. P. Neumann, H. Fuhlrott, and H. Vehoff, "Experiments Concerning Brittle, Ductile and Environmentally Controlled Fatigue Crack Growth," Fatigue Mechanisms, Proc. ASTM-NBS-NSF symposium, J.T. Fong, ed., ASTM STP 675, Am. Soc. Test. and Nat'l., 1979, pp. 371-395.

IV. FATIGUE CRACK GROWTH AT HIGH ΔK

Fatigue crack growth at high ΔK may include static modes of fracture such as burst cleavage and microvoid formation.^(1,2) The first year's program involved the analysis of such mechanisms in terms of microscopic cleavage or ductile hole growth criteria. With this understanding we hope to apply such physical insight to fatigue crack propagation at high ΔK . This far, we have:

- i) formulated, in its final stages, a ductile-brittle transition model for burst cleavage in Fe, Fe-Ni and Fe-Si alloys;
- ii) shown that 2nd phase particles are responsible for cleavage nucleation in these Fe-base systems;
- iii) shows that the same ductile-brittle transition model applies to our model material, BCC Ti-30 Mo;
- iv) analysed microvoid hole growth and void sheet instabilities in higher strength alloys to demonstrate the role of second phase particles in the ductile fracture process.

Several of these are in various stages of publication^(3,4) as aided by Dave Atteridge, Y.T. Chen, Tom Johnson, Kumar Jatavallabuhla and Rex King. Only the later two have been financially supported on the project this last year.

IV-1. Burst Cleavage in BCC Fe

W.W. Gerberich, Y.T. Chen, D.G. Atteridge and T. Johnson

We now have a working model for those alloying and microstructural effects which control the DBTT in Fe and Fe-binary systems.^(3) Without going into

detail, which will be presented in the next progress report, it is sufficient to state that the model includes fundamental dislocation parameters such as line energy, burgers vector, solute atom misfit, solute concentration, Hall-Petch slopes, Peierls stress and cleavage fracture stress. The approach is to use the plastic constraint factor to raise the temperature and strain rate dependent flow stress to the cleavage fracture stress. What remains to be done is to apply this to the quasi-static crack-tip problem much as Pisarenko and Krasowsky have attempted and then modify this for the fatigue crack propagation problem at high ΔK .

IV-2. Burst Cleavage in BCC Ti-30 Mo
Kumar Jatavallabuhla and William W. Gerberich

Both slow notch bend and impact tests of Charpy V-notch specimens have been run to establish the DBTT values for the vacuum annealed (22 ppm H) Ti-30 Mo alloy. As outlined in Section I-3, these DBTT temperatures may be quantitatively modelled with the dislocation dynamics approach. The next phase of our research will be to apply this to the quasi-static crack-tip problem.

IV-3. Microvoid Processes
W.W. Gerberich, M. Stout, K. Jatavallabuhla, D. Atteridge and R. King

Detailed measurements of acoustic emission from particle fracture processes in conjunction with fractography have established that large particle-matrix decohesion followed by void sheet instabilities is the microscopic failure mechanism. Critical strains for both uniaxial tensile testing and plane strain fracture toughness testing, as calculated from a McClintock hole growth model, compare favorably with ductilities and toughness for eleven heat treatment conditions of IN-718 superalloy and 4340 steel. Such models

indicate that hole growth around Ti(C,N) particles in IN-718 enlarge the initial diameter, D_i , by about a factor of two prior to instability, i.e.

$D^*/D_i \approx 2$. Hole growth around these larger 4.3 μm size particles is followed by a void sheet instability between smaller Laves phase particles ($<1 \mu\text{m}$ diameter). This, however, only applies to the initial instability associated with J_{IC} since much higher loads (15 percent higher) and crack growth (1mm) are encountered to J_{max} . The detailed analysis of both the acoustic emission data and applicability to fracture toughness models is presented in Reference 4 and included here as Appendix A-4.

Assuming that hole growth develops sequentially between the large second-phase Ti(C,N) particles, it is possible to model equilibrium at each stage of growth in terms of a cohesive, non-linear zone (the plastic zone), a semi-cohesive, non-linear zone (the void-filled hole growth region), and the crack. The size of each growth step is governed by a local strain criterion which determines how many of the ligaments in the developing void sheet will undergo fracture. Preliminary calculations indicate that the average growth step near J_{max} approaches the average spacing between large particles. The model, in its preliminary stages, is more descriptive than analytical since it does not accurately portray the boundary-value problem nor does it consider any history effects. Nevertheless, it would appear to be a microscopically appealing approach that addresses the sequence of events as they are observed to happen. An unexpected outcome is that this approach predicts reasonably well not only the increase in load (stress) with increasing crack growth to J_{max} but also the decrease in stress with increasing crack growth beyond J_{max} . Such a result is shown in Figure 1 where the experimentally determined crack extension versus applied stress is compared to two numerical solutions. The modelling parameters are hold growth coefficients, α , volume fraction of Laves

phase, f_{v_0} , length of the void sheet, l_v , initiation strain for void nucleation; ϵ_i , and flow stress, σ_{flow} . The decrease in stress may be due to net section yielding, but, alternatively, may be related to the changing void fraction in the semi-cohesive zone.

References (Section IV)

1. R.O. Ritchie and J.F. Knott, Acta Met., 21 (1973) p. 639.
2. W.W. Gerberich and N.R. Moody, "A Review of Fatigue Fracture Topology Effects on Threshold and Growth Mechanisms," Fatigue Mechanisms, Proc. ASTM-NBS-NSF Symp., J.T. Fong, ed., ASTM STP 675, Am. Soc. for Test. and Mat's., (1979) pp. 292-341.
3. W.W. Gerberich, Y.T. Chen, D.G. Atteridge and T. Johnson, "Dislocation Dynamics of Fe-Binary Alloys: II. Application to the Ductile-Brittle Transition," Submitted, Acta Met., (1980).
4. W.W. Gerberich, M. Stout, K. Jatavallabuhla and D. Atteridge, Int. J. of Fracture, 15, No. 6, (1979) pp. 491-514.