COMPILATION OF REFERENCES, DATA SOURCES AND ANALYSIS
METHODS FOR LMFBR PRIMARY PIPING SYSTEM COMPONENTS

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1. Introduction

A survey and review program for application of fracture mechanics methods in elevated temperature design and safety analysis has been initiated in December of 1976. This is the first of a series of reports, the aim of which is to provide a critical review of the theories of fracture and the application of fracture mechanics methods to life prediction, reliability and safety analysis of piping components in nuclear plants undergoing sub-creep and elevated temperature service conditions.

In the classical sense fracture mechanics concerns the linear behavior of materials that fail in a brittle manner (that is, below or at the yield limit) with low fracture energy and small deformations. This is called plane strain fracture for which the theory developed by Griffith has been expanded into the linear elastic fracture mechanics (LEFM) methodology, involving the crack tip stress intensity K and the single failure fracture parameter K_{Ic} , the fracture toughness. K_{Ic} is a material property depending on temperature, loading rate, stress triaxiality, etc., in a manner similar to the dependence of the yield strength on these factors. In addition, K_{Ic} is also a function of the relative magnitude of the plastic and elastic volumes in the material in front of the crack tip. When the plastic bulb is small in comparison to the crack size and the bulk of the material remains elastic, the fracture criterion is the minimum value of the stress intensity K_{Ic} , the plane strain fracture toughness.

The success of the application of LEFM methods to low temperature and/or "heavy section" structures where these conditions are satisfied, suggested the extension of the theory to ductile failures. Ductile failures are associated with high fracture energy and large scale plasticity as encountered in "tough materials" and/or in thin walled structures such as shells, plates and pipes.

Although the various attempts to find a single failure parameter of the K_{Ic} type for ductile conditions were largely unsuccessful, the methodology of plane strain fracture analysis found widespread application in fatigue and subcritical crack growth evaluation of ductile materials. The reason for the applicability of LEFM techniques to the analysis of the fatigue process of ductile materials can be traced to the degree of confinement that develop under cyclic conditions since the plastic zone size where plastic strain reversal takes place is only a fraction of the extent of the plastic "bulb" size under monotonic loading. Thus the requirements of confined plasticity are satisfied even in physical structures composed of tough structural materials.

While extensions of the LEFM based theories for crack growth and fatigue fracture to elevated temperature applications show surprisingly good correlation with continuous cycle-type test data even in elevated temperature application, the linear damage concept that is reasonably successful at subcreep temperatures becomes increasingly inaccurate for discontinuous hold time type cycles for temperatures where time-dependent crack growth mechanisms are operative.

However, the fact is that the fatigue process in most elevated temperature nuclear applications would consist of a subcritical crack growth under hold time conditions by virtue of the intermittently cyclic nature of the plant operating conditions. This indicated the necessity for the development of mechanistic damage theories for the progression of the incremental fracture under arbitrary loading histories. In other words the development of an incremental damage theory based on local stress, strain, and energy condition in the vicinity of the developing crack rather than some linear fraction of the total damage is needed. The possibility that fracture mechanics theory was so successful in applications to the evaluation of fracture under grossly elastic monotonic and continuous loadings provided the

motivation for the present survey and review. Thus, the aim of this study is to produce a critical evaluation of just what elements of the fracture mechanics methodology are useful in the development of new damage theories and reliable life prediction methods.

Since life evaluation is an integral part of the design process in a number of non-nuclear industries, a specific effort is made to collect and appraise the fracture mechanics methods that are used in turbine, jet engine, aircraft structural evaluation, and quality control procedures.

The literature on fracture mechanics is so vast that a great deal of selection and screening is needed to assemble an overview of the state of the art that is comprehensible but which at the same time does not become incomprehensible by the sheer bulk of the report. To eliminate the material that is only of tangential interest to fracture based failure criteria for elevated temperature application a number of failure mechanisms such as excessive deformation, instability due to buckling and brittle fracture have been excluded from this survey. However, in some cases a complete separation of the analytical techniques along these lines are not possible. Linear elastic solutions are, strictly speaking, only applicable to brittle fracture or high cycle fatigue problems, but some aspects of the linear theory serve as the point of departure for modifications for plasticity conditions and thus need to be included when elastic-plastic and ductile fracture are considered. Similarly, the first phase of the survey was envisioned as covering only publications related to subcreep temperature applications. However, as the selection process progressed, it became necessary to include a discussion of creep effect as the distinction between time dependent and time independent failure mechanisms even at subcreep temperatures are rather difficult to define. Thus some aspects of the application of fracture mechanics to creep conditions

have also been included; although this will be the topic of a follow-up report planned later in this program. The fundamental purpose of the current task is to provide the background material for a critical review of fracture based failure criteria as used in industrial application or as proposed for future development. The potentially applicable literature has been grouped in the following somewhat overlapping areas:

1. Linear - elastic fracture theory
2. Elastic - plastic fracture theory and ductile failure
3. Multiaxial conditions, toughness testing
4. Fatigue
5. Time dependent crack propagation (corrosion, creep-fatigue, creep)
6. Life prediction

In view of the diversity of the topics and variations in terminology, it is useful to emphasize "fracture mechanics" as meant in this context to be a generic term for continuum mechanics description of the state of stress and strain in a body containing a sharp discontinuity (crack). Fracture mechanics in this sense is most often used in a comparative way where similar crack tip mechanical states are produced in differently cracked and shaped bodies of the same material (test specimen and prototype) so that the fracture characteristics of the structure can be inferred from the behavior of the test specimen.

Therefore, a fracture mechanics method is useful only if it models the important fractures of the problem including the loading components geometry shape and size effect, stress-strain at the crack tip and the correct constitutive relations for the material behavior, i.e., non-linearities, anisotropy, rate effects, etc. The lack of established correlation between defects and service performance has hindered the effort of including fracture mechanics methodology to elevated

temperature design applications. Recently, however, a number of methods have been developed to bridge the gap. It has been found that constitutive behavior of a material in front of a propagation crack is different from the bulk behavior since the local increase of the strain rate changes the slip characteristic and the cracking mode. Thus the relationship between creep and plastic effects, for example, is influenced by the strain rate gradient as described by the strain singularity at the crack tip. Correspondingly, the concept of strain rate intensity has been developed by means of direct measurement of crack opening displacement and derivation of computational techniques are available to relate nominal strain range and the local strain intensity rate.

2. General Considerations of Fracture Based Failure Criteria

Material failure through rupture separation is an extremely complicated phenomenon. In the field one observes the failure of engineering structures or structural components at a length scale of cm, and up. At the other end of the length scale (10^{-8} - 10^{-7} cm) the basic phenomenon taking place which may be identified with fracture (in the sense that new surfaces or voids are created in the body) is the (adhesive or cohesive) rupture of interatomic bonds. Thus, on one extreme of the structural scale, the material may be modeled as a homogeneous continuum and may be analyzed by using the theories of continuum mechanics and classical thermodynamics. These are often cumbersome to apply but are workable as research tools by using certain energy considerations. Some semi-empirically based failure theories have also been developed which are quantitative and can serve as basic design tools. On the other extreme at the ionic scale, the material model is discrete and only under extremely simple and idealized conditions, but nevertheless in a physically sound manner, quantum mechanics may provide certain quantitative information regarding the rupture strength. It is well known that

the predictions of these two quantitative models regarding the rupture strength of a given material are orders of magnitude apart. Explanation of this wide difference may be found in the microstructural phenomena taking place at various length scales between ionic and continuum levels. Unfortunately, all these microstructural theories have so far been only qualitative.

Starting with the macroscopic continuum and telescoping to the location of interatomic bond rupture one may distinguish certain phenomena at various length scales which have a definite bearing on material failure. Ionic model is for a perfect lattice. In practice the lattice is never perfect. The first level of imperfection is then a point defect and may come from a vacancy (a missing atom), an interstitial (an extra atom), or an impurity (a foreign atom as a substitute or interstitial). The next level is a line (dislocations) or surface defects (e.g., twins) which are observed at a length scale of 10^{-6} cm. Coalescing defects may form clusters in the form of incipient holes (vacancy clusters) and precipitates. Thus, the next level of activity is at the subgrain level which takes place approximately at 10^{-5} cm length scale. Motion of dislocations results in slip forming surface step or slip lines. Clusters of slip lines may form (optically observable) slip bands. Thus at a scale of 10^{-4} cm one observes the phenomenon of subgrain slip bands. At a scale of 10^{-3} cm one observes the phenomenon involving grains, inclusions, and voids. Very large local plastic deformations are observed at a level of approximately 10^{-2} cm. At the scale of 10^{-1} cm one may observe contained plastic deformations in the neighborhood of small localized imperfections. At higher length scales the solid undergoes elastic-plastic or large scale plastic deformations around cracks, holes, or notches.

Since the tools available to engineers are that of continuum mechanics and classical thermodynamics which consider the phenomenon in a macroscopic scale,

on the order of inches, the failure theories which will form the basis of any particular design process must necessarily be macroscopic in nature. The advantage of considering the material failure at various levels as described above is that it could help to make the final macroscopic empirical model more effective and more meaningful. Technically the phenomena taking place in the material at various length scales are related in the sense that to qualitatively understand the phenomenon at a certain level it is essential to know the details of phenomenon at the next lower length scale. Similarly, the phenomenon taking place at the next higher length scale in a sense provide the "boundary conditions" for any quantitative analysis at a given level. In trying to relate the results of various microstructural theories to ionic or continuum models, the importance of this point cannot be overemphasized. For example, the shear stress assumed to move a line of dislocations along a crystallographic plane is almost impossible to determine (given the highly complex geometry of the polycrystalline solid, the local nonhomogeneity and anisotropy, the micro residual stress, etc.) and certainly has nothing to do with the resolved gross shear stress calculated from the continuum model.

From the engineering viewpoint perhaps the greatest advantage of the microstructural considerations and theories has been in the area of improvement of the mechanical properties of structural materials and in the development of the new alloys suitable for specific applications.

Macroscopically mechanical response of the material may be studied as the "strain" (or "stress") in time produced by the applied loads and the temperature. Various continuum responses are classified as elasticity, plasticity, creep, relaxation, and viscoelasticity. Thus when the loads are applied material deforms, and the application of the appropriate continuum theory is quite unambiguous. Generally, deformation may be considered as elastic or inelastic. Elasticity implies "reversibility" and in metallic materials results from the resistance of

bonds when they are stretched. During elastic deformations bonds are not severed and the work of deformation is stored as potential energy. In inelastic deformations interatomic bonds are broken and reestablished again. In crystalline solids it is generally accepted that they are caused by the generation and motion of dislocations. The inelastic deformation produces structural inhomogeneities such as slip, kink, and twin bands which play a prominent role in the formation of local cracks and voids. Thus under inelastic deformation even though macroscopically no harm is done to the integrity of the material and indeed the material may acquire some desirable mechanical characteristics, microstructurally it is difficult to assess the local damage as the material goes through this irreversible change. However, creep and high cycle fatigue failures indicate that sustained a sufficiently long time or repeated often enough, inelastic deformations are capable of macroscopic failure.

The type of considerations mentioned in this section which provide a general background for the discussion of various fracture-oriented failure theories may be found in textbooks and treatises devoted to material science. From a viewpoint of structural applications, however, the books (1.44), (1.45) and the article (1.46) can perhaps be mentioned as the key references.

2.1 Linear Elastic Theory

In the absence of any appreciable inelastic deformation at the macroscopic level, the so-called linear elastic fracture mechanics has established itself within the past two decades as a very effective tool for dealing with the low energy type of fracture of structural solids. The technique is a generalization of Griffith's early work by primarily Irwin and his followers. Methods of its application and that regarding the corresponding material characterization for the measurement of the fracture toughness of the material are thoroughly documented.

Microstructurally plane strain (or brittle) fracture may be represented as a climbing group of edge dislocations. Unstable fracture occurs by the motion of these dislocations across the fracture plane, as in cleavage or low energy tear. The intersection of a cleavage crack with inhomogeneities leads to the formation of cleavage steps which appear as river patterns. These inhomogeneities may include grown-in screw dislocations, grain boundaries, screw dislocations that form when the crack decelerates, voids, and other cracked particles. Steps may also be formed when primary and secondary cracks join together. Such step formations increase the work done during crack propagation and hence increase the cleavage resistance of the material. Detailed description of these microstructural aspects of cleavage fracture may be found in standard books on material science and on mechanical behavior of materials (see, for example, (1.44) and (1.45)). In discussing the current state of the art it needs to be pointed out that even though these microstructural theories have been quite extensive and thorough, they are almost entirely qualitative. These theories are nevertheless invaluable for the development of processes to improve the strength properties of the metallic materials.

2.2 Elastic-Plastic Theory and Ductile Fracture

The term ductile fracture is used generally to describe the rupture phenomenon in materials following appreciable inelastic deformations. It is generally accepted that in polycrystalline materials ductile fracture occurs mainly by the opening up and coalescing of holes at inclusions or other inhomogeneities, by the growth of micro cracks, by strain softening and the formation of intense shear bands, or by hole growth from dislocations or their intersections under very high triaxial tensile stress. The hole growth theory indicates that such fracture depends on the strain history and the hydrostatic stress component, on the ratio of the part size

to inclusion spacing and on the anisotropy of inclusions. These different mechanisms and the macroscopic stress and strain distributions lead to a variety of macroscopic forms of fracture: namely, "cup and cone", "shear", "normal", "rosette", "woody", and "fibrous" (1.44, 1.45, 1.46). In ductile fracture the growth is slow and stable and takes place under increasing stress that is usually above the "yield" value. Generally the accepted explanation of this behavior is that of Cottrell (1.59, 1.45). These stable ductile cracks grow by plastic deformation when dislocations are injected into slip lines ahead of the crack border. When the applied stress is less than the yield strength, the dislocations are localized near the crack and their back stresses oppose further plastic deformation at the border. Thus the crack cannot grow appreciably until the general yield is reached and the plastic zone spreads completely across the specimen.

The voids can form around grain boundary triple points, but they are usually nucleated when a nonmetallic inclusion or a second phase particle is cracked or is debonded from the matrix. The resulting crack may then open up into a void. The coalescence of voids generally occurs by the elongation of the voids and the consequent necking of the material (1.45).

Macroscopically no satisfactory failure theory seems to exist which conforms qualitatively to the microstructural observations and gives a consistent quantitative criterion. In design of the parts containing no stress raisers the yield criterion is generally used as the failure criterion. In high toughness materials containing a macroscopic crack (e.g., pipes and pressure vessels with surface cracks, or thin plate or shell structures with a through crack), there is a stable phase of the crack propagation generally termed as ductile. In this case as the crack propagates, the plastic zone ahead of the crack border grows with the crack and a steady increase in the magnitude of the applied load becomes necessary to overcome the

increasing resistance of the material to fracture. Consequently, a toughness oriented single parameter fracture criterion becomes totally inadequate to deal with the problem of ductile crack propagation.

In spite of the limitations in recent years a number of fracture criteria have been developed and proposed which may be used for ductile fracture propagation: among them the crack extension resistance curve (K_{Rc}), critical crack tip stretch, δ , net ligament plastic instability, crack opening displacement CTOD, J -Integral, and Generalized Energy Release rate are the most prominent.

Extensive discussions of the applications of such criteria to piping failure appeared within the last six months by Erdogan (3.20), Banford and Begley (2.47), Kanninen et al (2.44) and others (2.55, 2.45, 2.46). The last two are even more specifically applicable to LMFBR primary system piping as the data used in the review of the failure theories are for 304 and 316 stainless steel materials. Without attempting to anticipate the results of the evaluation of the literature, a limited comparison of the competing theories based on these reviews is shown for illustration as adopted from reference (2.44) in Table I.

A cursory look at this table shows that the problem of the selection of the "best" method is burdened with subjective judgment. Quite simply stated, at the present time the best method is the one that has the best data (or at least some data) for the particular application at hand. This will be discussed in detail in the sections that follow.

2.3 Multiaxial Conditions Toughness Testing

An additional problem area is the treatment of the differences between the fracture test specimen and actual component conditions. Cracks in an operating piping system by definition are surface cracks or part-thru crack types. The loading conditions in general induce non-axisymmetric stress conditions at the crack tip

TABLE I

COMPARISON BETWEEN PROPOSED AND CURRENT
ANALYTICAL METHODS FOR PLASTIC FRACTURE

<u>METHODS</u>	<u>MAJOR STRONG POINTS</u>	<u>MAJOR WEAK POINTS</u>
K_R Curve	<ol style="list-style-type: none"> 1. Measurements can be made easily and accurately 2. Method can cope with stable growth 	<ol style="list-style-type: none"> 1. Method completely depends on linear elastic fracture mechanics 2. K_R values may depend on geometry and initial crack size
CTOD	<ol style="list-style-type: none"> 1. Measurement has appealing physical interpretation 2. Experiments exist to show critical CTOD to be geometry independent below and above general yield 	<ol style="list-style-type: none"> 1. Interpretation and application made with linear elasticity 2. Measurement can be difficult to make 3. Application to stable crack growth is very difficult
J Integral	<ol style="list-style-type: none"> 1. Offers a well-defined straight-forward computational procedure 2. Experimental evidence exists to indicate that J_{IC} may be a material property 	<ol style="list-style-type: none"> 1. Theory rests on assumption of deformation plasticity (non-linear elasticity) which precludes unloading 2. Method cannot be applied to stable growth 3. Method cannot be applied to general yield
Generalized Energy-Release Rate	<ol style="list-style-type: none"> 1. Can take direct account of micro-mechanical processes involved in plastic crack propagation 2. Separates geometry dependent effects from material dependencies 3. Applicable for arbitrary constitutive behavior 	<ol style="list-style-type: none"> 1. Requires finite-element analysis procedure for application and interpretation of experiments
Net Section Stress	<ol style="list-style-type: none"> 1. Simplicity of application 2. Accuracy at low temperature application 	<ol style="list-style-type: none"> 1. Grossly unconservative in the creep range

which result in the variation of the stress intensity, COD, CTOD, etc., around the crack boundaries. The evaluation of the leak before break criterion requires the capability for determining the direction of the crack growth on the basis of a local fracture criterion, while in a failure analysis only the worst or critical condition need to be considered. The crack resistance in pipe under multiaxial stress conditions particularly in the elastic-plastic range is effected by constraints due to triaxiality, thermal gradients, material anisotropy due to manufacturing processes, etc. Thus the interaction between stress field and local "toughness" is a complex problem. In current design practice the simplifying assumption is made that the leak before break condition can be assured if the critical crack length is a multiple of the wall thickness. Thus the part-thru initial crack upon growth would penetrate the pressure boundary before the length would reach critical size for fast fracture initiation.

Recent work on fracture mechanics evaluation of pipe failures shows that surface cracks can propagate preferentially on the surface to many multiples of the wall thickness without breaking through. This is especially true under bending conditions and due to ovalization (as in elbows, U-bends, or out of round straight pipe under internal pressure). The reduction in triaxial constraint in front of a crack approaching a free surface is an other mechanism for the change in the uniform crack growth profile. These types of stress states and constraint variations need to be included in the analytical model as well as into the material toughness test programs to substantiate the leak before break philosophy.

A similar problem area is the evaluation of the effects of proof tests that are used to substantiate that undiscovered cracks are limited to safe size. In practice, such philosophies ignore certain inadequacies in the theory and overlook practical reality. As a result, a number of thoroughly proof-tested vessels

have been known to fail in a catastrophic manner from rather large preexisting undiscovered cracks.

There are several possible problems wrapped up in this discrepancy. The first is the possible deterioration of toughness at a crack tip due to large strain cycling. Even more significant is the question of damage occurring at elevated temperature or during shut-down affecting the ambient temperature material toughness (2.24) and allowing catastrophic fracture during pressure testing. Sometimes failure occurs because the service loading differs in some way from the proof loading (i.e., thermal conditions). A related problem area is the question of fracture mechanics assessment for cracks in regions of residual stress. References (2.26, 2.27, 2.35, 3.21) give some views on the residual stress question, but experimental validation is lacking.

In general multiaxiality has to be dealt with in all fracture mechanics applications to real problems. That is solutions must be adapted, combined or superposed to accommodate the deviation in geometry loading temperature, etc., from the idealized cases analyzed in the literature. In addition, many results are published in numerical form which requires some intermediate interpolation or curve-fitting to facilitate practical application. Some consideration has been given to this problem in references (1.5, 2.31) for 2D geometries, but it is clear that similar development is necessary for shell and piping applications.

2.4 Fatigue (below creep temperature)

Generally in actual structures fatigue is conceded to be by far the most widely encountered cause of failure. Fatigue cracks can be initiated in a number of ways; however, the important fact is that they are usually nucleated at a free surface from small flaws, scratches, and other stress raisers. Microcracks may also appear at a variety of sites. For example, soft zone existing along grain boundaries may lead to crack initiation at a triple point. Extrusions and

intrusions emanating from the slip bands under repeated loading are usually locations of microcrack initiation (1.45).

The crack initiating at a surface slip band will first continue to grow into the material along the primary slip plane and then veer into a plane perpendicular to the (local) maximum tensile stress. Crack growths before and after this transition are called Stage I and Stage II growth, respectively. Magnitude of the tensile stress generally governs the transition: the lower the stress magnitude, the larger the Stage I growth. In polycrystalline materials Stage I growth is terminated by the grain boundary intersecting the crack front. Fatigue crack propagation as is commonly referred to is the Stage II growth. One-to-one correspondence of fatigue striations and load cycles indicate that crack grows a finite increment in each load cycle. Striations are the lines or markers observed on the fatigue surface resulting from sharpening (under compression), plastic deformations and blunting (under tension), and resharpening of the crack under repeated loading. Stage II growth continues until the crack becomes large enough for unstable fracture. As key references dealing with the microstructural aspects of fatigue crack nucleation and propagation, the articles by C. Laird and by J. Schijve in (4.2) may be singled out.

From the viewpoint of design applications there have been two main difficulties in considering the fatigue failure. One is about properly distinguishing between "crack initiation" and "crack propagation" in fatigue, particularly how to make a judgment when designing a component. The second problem refers to finding a suitable quantitative model which may be used as a failure criterion or design tool. At the laboratory level these questions appear to have been resolved by separating (or classifying) the fatigue failure into high cycle fatigue, low cycle fatigue, and fatigue crack propagation. In the high cycle fatigue the net section stress is below the yield strength of the material, and the fatigue life

is (roughly) over 10^4 to 10^5 cycles. In this case the basic design tool is the S-N curve which means that the material has to be characterized for each environment and perhaps more importantly, as far as possible for each surface finish.

In the low cycle fatigue it has been shown that Stage II growth occupies most of the life (1.44, 1.45) and empirically the strain range vs. fatigue life behavior in unnotched ductile metals and alloys seems to be very similar. This observation led Coffin (1.45) and Manson (1.45) to propose the low cycle fatigue model relating the plastic strain range $\Delta\epsilon_p$ to the fatigue life N_f ($N_f^n \Delta\epsilon_p^n = C$; where C and n are constant and $n \approx 1/2$). Here it should be noted that both in high and low cycle fatigue the definition of failure has been rather vague. "Observation of first visible crack", "drop in the stress level" in constant strain tests, complete fracture, or the occurrence of some other phenomenon have variously been assumed to amount to failure. This may not be a major problem, however, since in a study (4.48) it was shown that at least 75% of the fatigue life could be accounted for as the (Stage II) crack propagation, by actually counting the striations on the fatigue fracture surface. Also by measuring the striation spacings the crack propagation rate da/dN was shown to correlate quite well with the so-called strain intensity factor $\Delta\epsilon_p \sqrt{a}$. If one also considers the fact that in low cycle fatigue below 10^3 cycles, almost the entire life is spent in Stage II propagation. It can be shown that the equations describing the process of striations lead to the Coffin-Manson model for low cycle fatigue (1.45).

There are, of course, many factors influencing the fatigue life, notably, the mean and amplitude of the cyclic stress, multiaxial nature of stress state, notch and size effects, effect of stress history, frequency, material behavior, surface conditions, temperature, and the environment. Some of these factors have been extensively studied. An adequate summary and the state of the art description

may be found in (1.44), (1.45) and (1.46). Among these factors the load biaxiality in a piping section needs attention. In the low cycle fatigue range the design rule is provided by the ASME Code (Criteria of the ASME Boiler and Pressure Vessel Code for Design by Analysis in Sections III and VIII, Division 2, 1971), which employs the maximum shear criterion modified to include the strain dependence of fatigue behavior. The review article by Kremp1 (4.5) indicates that there may be more accurate ways of modeling the phenomenon.

In contrast fatigue crack propagation below creep temperatures and in the elastic range has been very satisfactorily treated by using the tools of linear fracture mechanics. Here the stress intensity factor range appears to be a highly effective correlation parameter for the crack propagation rate. The modeling of the process is well understood and very widely used. A critical review of the development of the concept may be found in (4.6) and (4.7). (See also (4.8), (4.9) and 4.10) for recent modifications and some applications.) Even in the presence of such factors as corrosive environmental effects and temporary healing effect due to overload cycles, fracture mechanics seems to provide adequate modeling. As a key reference discussing these various aspects of fatigue crack propagation, one may mention the review articles in (4.2) (see also (4.4)).

The effect of the cyclic load profile and multiaxial stresses on the fatigue crack propagation has not been adequately studied. Regarding the load profile, in practice the tendency has been to use a linear cumulative damage rule. As for the effect of combined stress state, very limited amount of results obtained thus far indicates that it is of secondary importance. Hence, in practice, the effect of load biaxiality is ignored and the maximum principal stress (or the stress intensity factor based on the maximum stress) is assumed to be the controlling load factor.

Regarding the fatigue failure in general and the fatigue crack propagation in particular, very little work has been done to relate macroscopically observed trends and empirical models to the microstructural theories in a quantitative manner. The existing studies such as (4.11) where an attempt was made to explain Coffin-Manson model from the basic principles and (4.12, 4.13, 4.14) where the dislocation theory was applied to the problem of crack propagation, have not developed into practical methods. Nevertheless, the work along these lines as well as from the continuum viewpoint such as (4.15, 4.16, 4.17) is needed in order to improve and to base the existing empirical models on more physically sound foundations.

2.5 Time Dependent Crack Propagation (corrosion, creep fatigue, creep)

2.5.1 Mechanisms of Time Dependent Crack Growth

In this section we shall be concerned with the quantitative representation of the effects of time, temperature, and environment on crack growth rate, presuming the applicability of fracture mechanics. Time-dependent crack growth includes a variety of phenomena which have historically been studied independently and to which special characteristics have been ascribed, via, creep/fatigue interaction, corrosion fatigue, stress-corrosion cracking, hydrogen stress cracking, etc. Recently it has become acceptable to view these phenomena as a more or less continuous spectrum of mechanisms, as is illustrated schematically in Figure 1. This approach is taken here simply because the similarities between these mechanisms appear more compelling than their differences.

2.5.2 Stress Corrosion

Considering, for example, the high-temperature fatigue of Type 304 stainless steel and the process of stress corrosion cracking in oxygen-containing water, BWR bypass and core spray piping, in both cases the fracture mode is intergranular and attributed to oxygen activity. Bulk creep deformation may occur at the higher

Fig. 1. Schematic Interactions Between Creep, Fatigue, and Corrosion

BWR bypass and core spray piping, in both cases the fracture mode is intergranular and attributed to oxygen activity. Bulk creep deformation may occur at the higher temperature range, but James has shown (5.79) that air is responsible for time- and temperature-dependent fracture in fracture mechanics specimens, at least under limited conditions, and that bulk creep cavitation is not a significant factor: in the absence of air, James found crack growth to be transgranular. However, inelastic strain is required for initiation of both types of fracture. Multiple, branching cracks are produced under near static loading (thus complicating the fracture mechanics analysis). On the other hand, the precise role of the environment in the intergranular fracture mechanism is electrochemical in the aqueous environment and requires specific composition gradients at a grain boundary. Depletion of chromium by precipitation of carbide and possibly trace elements are believed to result in anodic dissolution. The analogous mechanism in the gas phase is not known but is thought to require absorption of oxygen on fresh metal surface followed by diffusion of oxygen along the grain boundary (specifically, the boundary between the carbide particles and the matrix) (5.80). Void nucleation and growth by vacancy diffusion may be involved. However, in both cases the grain boundary attack is sustained because of the inability of a protective oxide film to form at the crack tip. Thus, regardless of the differences in the exact details of grain boundary attack, it is difficult to support the separation of these two phenomena.

The cracks or voids originating at grain boundary junctions or triple points are usually the result of grain boundary sliding due to strain concentrations at the triple points. This process is known as the wedge type crack initiation (1.45). Grain boundary migration and recrystallization may reduce the strain concentration and hence lessen the tendency to void formation. Also, the presence of brittle impurity particles or weakly bound precipitates may facilitate the crack formation

at the triple point. The mechanism by which these wedge cracks grow and cause brittle fracture is not well-understood. The argument against a Griffith type mechanism is that at high temperatures there will always be sufficient crack blunting whereby keeping the local stress below the value of cohesive strength. Also at these temperatures the diffusion processes would be too slow to cause appreciable migration of atoms from the tip to the sides of the cracks. A critical shear displacement model (1.45) may be a possible mechanism. However, this too appears to be only a conjecture.

Cavitation or the r-type void formation starts as roughly spherical holes along the grain boundaries transverse to the local tensile stress as well as at triple points. They then flatten and link with each other. They seem to form continuously and grow at all stages of the creep process. The fact that the void density is higher near the free surfaces suggests that the growth occurs by a grain boundary diffusion process, free surfaces being the ultimate source of vacancies. Some studies indicate that the presence of second phase particles is necessary for r-type crack nucleation at the grain boundaries, the lower binding force between the particle and the matrix facilitating the void formation. The r-type voids grow by a flux of vacancies and a counterflux of atoms from the voids to the grain boundary sites (the process known as the Herring-Nabarro creep). The evidence for this vacancy diffusion mechanism has been obtained by varying the hydrostatic component of the stress state. Grain boundary voids can also grow by a continuation of the void nucleation process of grain boundary sliding. Other diffusion-controlled processes leading to steady-state creep may be the climb of edge dislocations out of slip planes to overcome obstacles and elimination of dislocation loops around precipitates (1.44, 1.45).

2.5.3 Mechanisms of Creep Fatigue

(Coffin (4.30) has taken the view that creep-fatigue interaction in air is basically stress-corrosion cracking and this approach has led to the well-known "frequency-modified fatigue lifetime" correlation for crack initiation or growth rate. One practical consequence of this hypothesis is that there is no predicted saturation of the hold-time effect, as contrasted to those correlations based upon stress relaxation, for which saturation is predicted: paramount among the latter is Manson's "strain range partitioning" method. Modifications of Coffin's method to take into account cycle shape have been proposed by Coffin, himself (4.30), and by Ostergren (5.44); they include effects of strain rate in tension versus compression. Maiya and Majumdar (5.81) have recast this method in incremental form, allowing integration of a damage law around an arbitrary strain-temperature time cycle.

The most complete crack growth algorithm for creep-fatigue, also based upon Coffin's method, is Carden's empirical parametric representation (5.67), which includes frequency, temperature and R-ratio effects (for cracks propagating in air), but does not incorporate cycle shape. Its demonstrated ability to correlate data for Type 316 stainless steel is impressive, albeit restricted to relatively short hold times. The topic of creep-fatigue has been extensively reviewed from the phenomenological and design application point of view in a recent report (ORNL 5073 i.e., 4.30) and thus these aspects will not be discussed here. In the following we will be concerned only with the limitations on fracture mechanics methods imposed by the creep effect.

2.5.4 Creep Crack Growth

The limitations of linear elastic fracture mechanics analysis to creep have been determined by Haigh and Richards (5.17) from slip-line field analysis and

empirical fatigue tests on specimens of different configurations (single-edge notched, center cracked, compact tension). Divergence of the results occurred approximately at the predicted levels of stress intensity factor. Rice (2.55) has further discussed the state of development of elastic-plastic fracture mechanics analysis and concludes there is reason to believe the J-integral approach may be able to treat the region between elastic and fully plastic behavior, as illustrated by the work of Dowling and Begley (2.61) provided crack surface closure is accounted for. Landes and co-workers at Westinghouse are actively pursuing the J-integral representation of creep crack growth, but the use of fracture mechanics to predict creep and creep-fatigue crack growth rate is more extensive in the United Kingdom than in the U.S. where the advances seem to some extent to be impeded by the ASME Code, which relies primarily on smooth specimen creep and fatigue properties.

Use of post-yield analysis for small flaws in tough material, or in the creep range, is also much more common in the U.K. (2.56, 1.58), usually following the anti-plane strain analyses of McClintock (2.57) and of Bilby, Cottrell and Swinden (2.58). In particular the BCS model has been applied to clusters or clouds of defects in which the individual flaws cannot be treated by linear elastic fracture mechanics (2.59). Tompkins (5.73) has further extended the analysis to linkup between cavities forming by creep cavitation on grain boundaries; the connection between this model and environmental effects is further discussed below. Indeed, despite the obvious deficiencies of the anti-plane strain analogy in predicting Mode I viscoplastic crack tip deformation, the BCS model appears to possess a unique versatility in treating cyclic loading (4.43), as well as lending itself to finite discretion (4.44) for computer analysis of work hardening or strain-rate dependent behavior; it appears that closure and crack face friction effects can also be simulated. The BCS model appears to be the only one that includes a Bauschinger effect.

Despite these apparent advantages, experimental evaluation of the BCS analysis for subcritical crack growth has not been reported, except for one attempt, largely unsuccessful, by Tompkins (4.45) to extrapolate results for large cracks to very small cracks. Again, the AFML-sponsored microfracture mechanics program, previously cited, represents the first known attempt to obtain fatigue crack growth laws for microcracks of known geometry at stresses near net section yield and in the creep range and to model the CTOD analytically.

Crack growth laws have been reported for specimens tested under uniform strain control at strain ranges resulting in measurable hysteresis loop widths. At Pratt & Whitney Aircraft, fracture mechanics tests are commonly conducted in fatigue on thin-walled tubes under uniaxial strain control with through-thickness cracks initiated at small holes. While data have been generated on nickel-base superalloys and titanium alloys, only the former has been reported (5.45, 5.78). Cracks are monitored by acetate replica and traveling microscope following fatigue pre-cracking to a C-dimension approximately equal to the wall thickness (0.3 - 1.5 mm). The plastic strain range is generally small compared to the elastic. Results are correlated by means of a strain intensity factor range, $B\Delta\epsilon_T\sqrt{\pi C}$, which is reduced to an effective value by considering closure effects. For the yield strengths and crack sizes tested, inelastic methods have not been found necessary and have not been applied. Attempts to extrapolate crack growth data down to flaw sizes on the order of the grain size in nickel-base superalloys, under Air Force sponsorship, have apparently proved unsuccessful.

Solomon and Coffin (5.76) have measured surface crack length during strain-controlled cycling of solid cylindrical specimens and found the rate of increase of the logarithm of crack length correlated with strain range. The depth, and consequently the strain intensity, of a surface crack are not known; Wells and

Sullivan(5.77) found that in low cycle fatigue the rate of crack growth parallel to the surface in nickel-base alloys is significantly more rapid than growth below the surface. In fact, during the early stages of growth, the growth rate is sometimes constant, independent of crack length(4.42). Description of the microcrack extension process is generally viewed as the link between fracture mechanics and crack "initiation" tests; the microcrack nucleation process is usually found to require a very small fraction of specimen lifetime in low-cycle fatigue.

The prediction of crack growth in nuclear piping components would appear to warrant the development of inelastic test methods and CTOD models along similar lines, it being recognized on the one hand that linear elastic fracture mechanics does not apply to low-cycle fatigue of low strength materials, but on the other hand that fracture consists predominantly of crack growth from flaws.

Of possible relevance to nuclear piping are crack propagation tests that simulate the load-temperature cycle experienced by the component. Again, these methods have been developed primarily at Pratt & Whitney by Rau, Hopkins and others (5.45,5.74). The procedure consists simply of simultaneously programming an induction heater (closely coupled to the tubular specimen) along with the servohydraulic actuator to repeat a desired strain-temperature history. Most of these tests are now controlled by computer. The shape of the cycle has been found to affect crack growth rate in turbine blade alloys(5.45, 5.75) primarily through influence on the effective range of strain intensity; essentially, the interpretation is based upon qualitative CTOD argument.

The successful correlation of high temperature fatigue crack growth rates with the stress intensity factor led early investigators in creep crack propagation to adopt a similar approach. However, results indicated only restricted applicability of K and later work has made use also of net section parameters. From this a rationale has begun to evolve based on consideration of creep ductility.

For the case of uncracked static creep rupture, two extremes may be identified. This so-called creep brittle behavior is associated with intergranular void formation and growth until the true stress acting on a section reaches a critical value, and failure occurs at a low value of creep strain (say $< 1\%$). Conversely, creep ductile behavior occurs mainly through a plastic flow process causing "necking" at the point of final instability and allowing a large amount of overall strain to be accumulated (say $> 10\%$). The terms "brittle" and "ductile" in this sense are not well defined but help to establish general regimes of deformation and fracture. Creep ductility is a function of material (and heat treatment, etc.), stress, and temperature.

This same classification is useful when describing the creep response of cracked structures. In the extreme, brittle fracture is considered to be controlled by the crack tip stress field, whereas ductile fracture is predominantly dependent on net section deformations. Two approaches to the problem can be made: one via fracture mechanics for the brittle case and one via conventional creep theory for the ductile case. In general, however, a transitional form of behavior would be expected and a little work is also now being directed towards this end.

2.5.5 Creep Crack Growth Correlation with Fracture Mechanics Methods

For a cracked structure exhibiting low creep ductility, there is little redistribution of stress due to creep outside of the immediate crack tip region. Thus the initial stress distribution may remain valid, and creep crack propagation will be by a process of local rupture at the crack tip. It is possible under these conditions, therefore, that crack growth might be correlated with the stress intensity factor, K .

Creep crack growth data for a 2 1/4 Cr1Mo simulated weld metal were obtained (5.49) in quenched low ductility form, at 565°C (1050°F) using single edge notch

(SEN) specimens. Results were presented in a similar manner to fatigue crack propagation results, plotting growth rate $\frac{da}{dt}$ against stress intensity factor (K), Figure 2. The correlation obtained suggested a relationship of the form:

$$\frac{da}{dt} = AK^q$$

where A, and q, are material constants. The authors suggested that the exponent q (in this case q = 5.5) may be approximated to the stress exponent n in the conventional creep law:

$$\dot{\epsilon} = B_n \sigma^n$$

where $\dot{\epsilon}$ denotes creep strain rate and B_n is a constant. Good correlation with K was found (5.37) on two simulated HAZ (heat affected zone) CrMoV steels at 565°C (1050°F) also using SEN specimens but of varying thicknesses and widths. They concluded that linear elastic fracture mechanics (LEFM) could be applied providing general creep displacements were small. Recently the K dependence of crack growth for low ductility quenched 1/2 CrMoV steel was confirmed (5.35) indicating that final failure is crack tip and not net section controlled. In all the above investigations fracture was intergranular and crack growth occurred on a straight front. However, 1/2% CrMoV steel was also investigated (5.37) in the normalized and tempered state at 565°C (1050°F), but correlation with K was not obtained. Similarly, poor correlation was found (5.54) between crack growth rate and K for 1% CrMoV tempered steel at 565°C (1050°F), using both SEN tension and bend specimens. A "family" of curves was obtained and a form for the crack growth rate equation was derived empirically which included the propagation time t_p , vis: $\frac{da}{dt} = AK^q t_p^r$. The exponents q, r were found to differ for the tension and bending cases. There was evidence of crack tunneling in all specimens that had not been machined with side-grooves. This characteristic of creep crack propagation seems to be universal in ductile materials where the

FIG. 2 CRACK GROWTH RATE VS. STRESS INTENSITY FACTOR FOR A
QUENCHED 2½ CrMoV STEEL

specimen thickness is large enough to allow plane strain conditions to prevail at mid-thickness (in contrast to plane stress at the surface), indicating an effect of different multiaxial stress state.

A 0.2% C cast steel was tested (5.48) at 450°C (840°F), using SEN and CT specimens, and a 0.2% C wrought steel, also at 450°C using CT specimens; results were interpreted on an LEFM basis. The plot of crack growth rate against K correlated the data reasonably over the limited range of K employed (30-50 MN/m^{3/2}) - (27-45 ksi/m^{3/2}); values of the K exponent in the expression $\frac{da}{dt} = AK$ were 14 for the cast steel and 20 for the wrought steel. The effects of temperature variation were also investigated and an Arrhenius relationship is indicated.

James (5.25) found good correlation between crack growth and K for both CT and CCP (centre crack plate) specimens in 316 stainless steel at 538°C (1000°F). However, the range of K in this case was very small (approx. 30-40 MN/m^{3/2}) - (27-36 ksi/m^{3/2}). Thornton (5.52) also used K to correlate results from SEN bend tests on 1% CrMoV and 1/2% CrMoV steels at 565°C (1050°F) with apparent success. The generality of the James and Thornton results to other geometries, constraint conditions, and a wider variation of stress intensity need to be confirmed.

The creep fracture behavior was studied (5.12) for several nickel-base alloys. Crack propagation rates and total failure times of CT specimens were examined on both a stress intensity and a net section stress basis. In general, more consistent results were obtained from the stress intensity correlations. There was no evidence of through-thickness deformation nor of excessive crack tunneling these tests, suggesting the ductility was limited.

One of the first attempts to relate crack growth under creep conditions to a net section parameter was made for a CrMoV turbine steel at 538°C (1000°F) in (5.20). Data was initially correlated with stress intensity factor with fair success. However,

much better correlation was achieved when growth rate was plotted against net section stress. The specimen geometry used was CCP, and results were presented for three different stress levels. Recently, good correlation of crack growth rate with net section stress was found (5.38) for creep of 316 stainless steel at 740°C (1364°F). Two thin sheet geometries were tested, viz. SEN tension and notched center hole specimens, and it was suggested that with creep deformation the sharp crack situation becomes indistinguishable from the blunt notch case.

Creep crack growth in CT and SEN specimens (of 1% CrMoV steel at 550°C , 1022°F) was investigated in reference (5.16) where some account of bending effects should be taken. The approach was based on the reference stress and correlation of total specimen creep displacement rates was preferred to the normal crack propagation rate correlation. It was found that the stress intensity factor was totally unsuitable for this purpose but that the reference stress yielded good results. Furthermore, reasonable agreement was found between deformation rates in cracked specimens and uniaxial unnotched creep test strain rates, considering the difficulty of defining a suitable gauge length for the cracked specimen geometry. Crack opening displacements (COD) were related to overall displacements by assuming suitable hinge points calculated from rigid plastic theory, and crack growth rates correlated well with COD rates during propagation when plotted on a log-log basis (Figure 3). Further, it was found that crack growth did not proceed until a critical COD was attained, thus explaining the incubation period of crack propagation often observed in creep ductile situations (Figure 4).

The application of COD to problems where large creep and plastic strains are present has been suggested in (5.34) following the successful use of strain intensity factors in correlating results where small inelastic strains are present (5.45). A linear relationship was also found (5.51) between propagation rate and notch opening displacement rate which was independent of applied stress level.

FIG. 3 CRACK GROWTH RATE VS. RATE OF COD INCREASE FOR TEMPERED 1 CrMoV STEEL

FIG 4 SCHEMATIC DIAGRAM SHOWING PROPAGATION OF CRACKS AND DAMAGE FRONTS FOR BRITTLE AND DUCTILE BEHAVIOUR

Creep crack growth was investigated (5.42) in vacuum at 550°C (1022°F) on a 1/2Cr 1/2Mo 1/4V steel heat treated so that it was in a relatively embrittled state. Discontinuities in the plots of crack propagation with time appeared to coincide with the grain size, and it was suggested that when sufficient cavitation had occurred, the main crack moved forward in a step-wise manner from one grain boundary to the next. It was interesting to note that even for this brittle alloy and the localized cavitation damage at the crack tip, the correlation of the stress intensity factor with propagation rates was poor; correlation appeared to be better with the accumulation of displacement at the crack tip.

2.6 Life Prediction Methods

2.6.1 Methodology

A methodology of lifetime prediction, or risk evaluation, based upon fracture mechanics has been developed for aircraft gas turbine disks and is currently being reduced to practice for steam turbine rotors. The interplay between nondestructive evaluation (NDE), metallography, mechanics of materials and structural analysis adopted from a paper to be presented at the March 1977 ASME Gas Turbine Conference (6.24), is shown in Table II.

While a review of NDE methods is beyond the scope of this survey, the probabilistic evaluation of flaw severity is at least equal in importance to the analytical prediction of crack growth rate and instability. This evaluation is obviously governed ab initio by the probability that the NDE procedure will be able to represent the extent of cracking in the pipe component. The methodology outlined here is similar to that under development for prediction of stress corrosion cracking in natural gas pipe lines, in which cracks initiating at the O.D. must be detected by an inspection from the I.D.

TABLE II.

Methodology of Structural Lifetime Assurance

The reliability of NDE is undergoing considerable analytical development. Johnson (6.25) defines the "uncertainty" of examination as the standard deviation of the normal distribution of indicated defect area for a specified true area. He shows that for any given uncertainty there exists an optimum "inspection size" (i.e., defect area corresponding to a rejection probability of 0.50) which minimizes the total cost. While the cost/risk optimization process described by Johnson (6.26) may be less applicable to nuclear components, it appears highly relevant to in-service inspection and repair/replace decisions. In particular, the impact of inspection uncertainty on the cost of rejection (and unnecessary repair) of sound components is usually much greater than on the cost of acceptance of unsound components. Also, a high degree of inspection uncertainty obviously offsets the accuracy of fracture mechanics analysis.

2.6.2 Flaw Severity

The "severity" of a flaw refers to its effect on initiation and propagation of a crack. In turbine rotors (6.24) the flaw is considered to be a fatigue crack or a concentration of porosity or inclusions. In the United Kingdom, it is common to regard a "cluster" of pores or inclusions in Cr-MO-V forgings as a linked up crack if the flaw diameter to spacing is sufficient for the ligaments between the flaws to yield upon loading (2.59). In the current EPRI program on Reliability of Steam Turbine Rotors, the severity of NDE indications is being determined by testing large strain-controlled low-cycle fatigue specimens containing the areas of interest within the gage section while monitoring the development of cracking by compliance and ultrasonic methods. After a macrocrack is indicated, the specimen is cooled below the nil ductility temperature and fractured to avoid excessive plastic deformation. Since striations are not formed on intergranular fracture surfaces, it is necessary to have an auxiliary measurement of the crack growth law at the

same frequency and temperature and in the same environment, usually from a fracture mechanics specimen prepared from adjacent material with the fracture plane in the same orientation. The electron channeling technique (6.32) provides a measure of plastic deformation on the scale of electron microprobe analysis and has the potential of identifying regions of varying crack tip plasticity on an intergranular fracture surface, at least in austenitic stainless steel (5.18). The crack growth law can then be integrated over the specimen lifetime to determine the initial flaw area corresponding to the final crack size in the specimen. If the calculated flaw area is smaller than the projected area of the physical defect, the time or number of cycles required to propagate the crack from its calculated to its true size is considered an initiation lifetime (at the strain range applied to the specimen): if the calculated area is larger than the true projected area, the presence of initial crack extension or of trace elements associated with the flaw is indicated.

In considering the effects of very small flaws, the scale of applicability of fracture mechanics is of concern. Similarly, in any discussion of crack initiation there is a need to define the limit of naturally occurring defect which can be considered the boundary between "initiation" and "propagation". While it is held by some that the crack growth stage in a smooth, unflawed specimen can be extrapolated via fracture mechanics back to some characteristic microstructural dimension, such as the grain size, direct verification of this proposition has not been made and is, in fact, the object of a current Air Force program (1.57). This investigation is being carried out on an extremely fine-grained nickel-base superalloy, forged from a powder metallurgy billet, which is used for advanced turbine disks. The specimens are in effect razor blades, with cracks initiated at the apex of a 60 deg. included single edge. Crack growth at room temperature is followed on both flanks by in situ fatigue cycling in a scanning electron microscope (SEM) at room temperature and by replication at 700C (1292°F), both on a scale of a few microns.

On the other hand, it has been well established that fracture mechanics applies to fatigue crack propagation from very small inclusions in steel, viz, down to a 10 micron surface crack length in AMS 4340 (4.46) and to 25 microns in A 533B (4.47). Fatigue cracks have also been observed to initiate preferentially at sulfide inclusions in 2 1/4 CR1Mo steam generator material in simulations of departure from nucleate boiling (DNB). These tests were typically conducted at 360C (770°F) and a total strain range of 0.003 at 1 Hr. The wide range of inclusion size in several combinations of melting and heat treating practice resulted in extensive scatter in fatigue lifetime. In view of the difficulty of representing the behavior of large structures by laboratory tests on small specimens, a statistical approach based upon fracture mechanics and measured inclusion size distributions has been suggested.

The propagation of low-cycle fatigue cracks from inclusions in aircraft gas turbine compressor disks has been described by Hill(6.27). This investigation, one of the most comprehensive lifetime assurance projects documented to date, contained most of the aspects of Table II, such as the determination of flaw severity from smooth specimen fatigue tests and the use of Monte Carlo procedure (6.28) to combine the effects of variable crack growth properties and initial inclusion area distribution. The problem involved failure of AMS 6304 disks from suspect heats found to contain trace amounts of lead. Fatigue crack initiation occurred at subsurface slag-type oxide inclusions at the bore and propagated to failure in about 20,000 hr. of service. Over 80 parts were ultrasonically inspected, and flawed regions in 31 disks were characterized by metallographic sectioning. Both center cracked and smooth bolt hole specimens were tested to determine the statistical distribution of crack growth data for material with different lead contents at elevated temperature; no creep/fatigue interaction was involved. Specimens removed from ultrasonically "dirty" regions were tested in fatigue, and the lifetime

correlated with initial defect area. The observed failure rate was found to be correctly predicted by fracture mechanics. Hill concluded that accurate distribution functions for flaw area and crack growth data are required to predict component life at very low failure rates.

2.6.3 Creep Damage Concepts

In creep life prediction to define a time for local rupture at the crack tip, the manner in which the material deteriorates must be known. Two approaches to the problem have evolved: (a) consideration of the physical processes involved in damage accumulation; and (b) invoking a phenomenological model of material behavior which is consistent with the observed results. The latter approach is more directly applicable to continuum models.

The manner in which cavities and cracks are formed is still subject to debate but two mechanisms may be identified: void formation and growth, and wedge crack growth by grain boundary sliding. Analyses of crack growth based on these types of mechanisms were carried out (5.56, 5.32). An important aspect of cavity formation and growth is its enhancement by prior plastic straining which was studied by Dyson and Rogers (5.5). This is particularly relevant to the creep crack situation where the initial high stresses within the plastic zone relax with deformation - the plastic zone area is effectively prestrained. The influence of multiaxial states of stresses on cavity formation is also being studied by Dyson.

Considering grain boundary strains and microcrack nucleation rates, crack initiation lives, propagation rates, and rupture lives for both smooth and notched specimens were derived (5.50). It was shown that crack growth can be expressed in terms of specimen net section stress and the model was substantiated by experimental results from 1CrMoV steel tested at 600°C (1112°F).

The phenomenological approach is used to model the development of damage in the rupture theory of Khachanov (5.28). The basis of the theory is the concept of a continuity function ψ which decreases from unity to zero as damage accumulates and failure occurs. The assumption is made that the rate of decrease of continuity (analogous to rate of increase of damage) is a power function of the 'true' stress, i.e.,

$$\frac{d\psi}{dt} = -C \left(\frac{\sigma_o}{\psi} \right)^\nu$$

where C, ν are material constants and σ_o is the applied stress. The continuity approach has been used to describe the initiation of propagation of a failure front under the action of a multiaxial stress system (as would prevail at the tip of a macroscopic notch or crack), Khachanov (5.29), Odqvist (5.40). In these analyses of creep crack growth from a notch the stress relation at crack tip was not taken into account. Also, it is to be noted that ψ is not necessarily set to zero at the failure front. There is a good argument in favor of studying damage propagation rather than crack propagation in that it more accurately represents the deterioration in load carrying capacity of a structure.

Employing the Khachanov theory of the Eqn shown above, a creep crack propagation model was developed by To (5.36) based on damage accumulation and rupture within the plastic zone ahead of the moving crack. A major conceptual assumption is that the Dugdale model of plane stress crack tip plasticity is valid under creep conditions. A linear relation between crack growth rate and stress intensity is predicted.

A recent report (5.14) indicates another manner in which the problem may be approached by way of continuum methods. A finite element computer program was used to model the advance of "tertiary" creep zones from a crack tip under static loading.

It was found that the damage front growth characteristics were very similar to those of experimentally observed crack growth, and that the damage front is "blunt" in comparison to the original sharp crack, Figure 5.

2.6.4 Special Considerations for Life Evaluation of Welds

The most difficult locations for lifetime prediction, and possibly the most important, are welds, and it is worth some discussion of their special requirements at this point. The poor accuracy of weld residual stress analysis is well known, as is the fact that no nondestructive residual stress measurement technique exists for nonmagnetic materials (X-ray diffraction may be possible for the inner surface of large pipe, but is limited by texturing associated with grain growth in the HAZ and the steep axial gradients characteristic of welding stresses). Process control and destructive evaluation of samples appear necessary, but obviously are of little use in in-service inspection. LASER pulse heating combined with acoustic emission, developed by Ford Motor Company, used in a circumferential and axial scanning mode, offers some potential for residual stress level measurement, fusion line and HAZ defect detection, or both. For magnetic piping materials, adaptations of the acoustic Barkhausen effect for residual stress mapping are possible (1.60).

The weld geometry, residual stress, grain structure and grain boundary composition (carbide morphology and trace elements) probably influence the shape and growth rate of circumferential cracks in fatigue and creep, as they do in stress-corrosion cracking. The rate of surface growth and linkup relative to through-wall propagation is known to govern the "leak before break" behavior of a gas line pipe and may be relevant to nuclear piping. Multiple initiation along weld beads or scratches has been reported to result in a high aspect ratio, plane strain crack that opens rapidly to a running, through-wall crack. In this connection, 2 1/4 Cr1Mo appears to be very scratch sensitive in long-life LCF ($\approx 10^5$ cycles),

Fig. 5.

The Propagation of Tertiary Creep Damage in a Plane Stress Notched Plate

despite its very low yield strength. Test results on full-thickness weld specimens, containing ID and OD surfaces, would be extremely valuable to the selection of the appropriate crack geometry for analytical fracture mechanics.

It is possible to map out the shape of such cracks with fair accuracy, if they are surface connected. Eddy current and magnetic perturbation techniques are capable of defining crack depth over a lateral dimension of a few mils.

2.6.5 Damage Gage Concept and Crack Initiation

While biaxial capacitance strain gages are being developed for nuclear piping systems (6.29), such instrumentation provides only a very indirect measure of crack growth. The Air Force has developed a fatigue crack growth gage that integrates that effects of load history and environment on the material of the structure (6.30). A precracked specimen is attached near the critical region of the structure with the crack normal to the maximum principal stress range. The compliance of the cracked gage relates the strain history of the structure to the strain intensity of the gage crack. If the structural crack size and local geometry are known or assumed, the observed crack growth in the gage can be numerically integrated to predict the crack growth in the structure. There are questions of nonlinear overload and R-ratio effects in applying such a transfer function to aircraft spectrum loading, but the concept merits consideration for monitoring nuclear piping components. The crack growth can be measured remotely by ac potential techniques (6.31). The gage material is the same as that of the structure and ideally would contain a weld. Several precracks can be introduced.

Conservative design analysis would assume the initial presence of flaws in the welds and base metal. The distribution of initial flaw size can be determined for the specific components by NDE with the aid of artificially-flawed standards or from the fabrication process control of similar components. Flaw characterization

by destructive metallography is usually required in the latter case. If the defect size distribution is known to be well below the practical resolution of NDE, it may be overly conservative to assume the presence of flaws with a low probability of detection by the NDE methods and procedures. In this case an "initiation lifetime" can be added based upon experimental data, or an equivalent initial sharp flaw size can be determined by extrapolation of crack growth data via a small flaw fracture mechanics. Where the component is subjected to high stress gradients, either in operation or service, the latter approach is preferable, since laboratory "initiation" tests are generally incapable of including the effects of such gradients.

2.6.6 Multiaxial Loading Effect on Creep Lifetime

In the creep range, it is generally found that rupture ductility decreases significantly with the increasing ratio of hydrostatic tension to effective stress: the effect upon rupture lifetime is much less. Unfortunately, little quantitative crack initiation or propagation data are available. However, from studies of CrMoV forgings it is believed that triaxiality governs the transition between transgranular and intergranular fracture, and in this event the lifetime can be substantially reduced. It is thought that the effect described here has more to do with the global stress state than with the microscopic stress state ahead of a crack. Investigations with blunt notched as opposed to fracture mechanics specimens may lead to a better understanding of these effects. The consequence of triaxiality appears to be the development of extensive intergranular cavitation through rapid void growth and linkup, rather than through nucleation by grain boundary sliding. It also appears that fracture mechanics may be less applicable in the presence of such cavitation, since the metallurgical factors that increase resistance to stress rupture are thought to result in higher crack growth rates.

On the other hand, crack growth tests conducted under plane strain conditions (e.g., plate bending) have shown lower growth rates than under plane stress (beam bending), presumably because the transverse stress decreases to CTOD (5.93). It is not known whether such tests have been carried out in the range of intergranular fracture.

Since the effect of triaxiality on other steels has not been determined, it is not known whether this possibility poses a threat to the reliability of piping components in the creep range.

3. Summary and Review of Fracture Mechanics Methodology for Piping Application

The survey of the assembled literature shows a great variety of theoretical and experimental strategies by which fracture mechanics concepts are used in the study of crack growth and the determination of the load carrying capacity of components containing flaws. In relatively low temperature applications, linear elastic fracture mechanics (LEFM) has been developed on the basis of plane strain assumption to characterize the state of material at the edge of a sharp straight through crack. This assumption is central to LEFM theory as the relationships between stress intensity factors K_I , K_{II} , K_{III} , and the energy release ratio G , the equivalence of the J integral to G , and the correlation of fracture toughness G_c and K_c all hinge on the plane state of the stress or strain field surrounding the crack tip. This means that crack extensions have to occur in the linear load deflection regime without any out-of-plane deformation taking place. These conditions are satisfied in thin walled piping only for very limited ranges of fracture toughness and loading conditions.

In the more general cases of surface cracks, embedded cracks, or curved cracks (such that occur in elbows), three-dimensional effects have to be accounted for. Also application of fracture mechanics theory to the tough and relatively low yield

strength materials used in the LMFR piping designs require modification because even low global stress levels can result in significant crack tip plasticity which must then be included in the analysis. In the general case of non-straight cracks and embedded or surface cracks, the stress intensities vary along the crack front and the direction of crack growth cannot yet uniquely be determined.

Similarly, geometry and loading conditions encountered in piping components such as combined pressure, in-plane, out-of-plane, and torsional moment loading on an elbow, represent a mixed mode condition for which unique relationships between fracture toughness and crack size or direction have not yet been developed. For certain elbow geometries of long radii, experimental studies have demonstrated that under simple tensile membrane loading, the surface cracks tend to propagate through the wall rather than in the axial direction thus qualitatively justifying the leak before break philosophy.

The same studies, however, show that more complex bending or combined loading can produce long part-through cracks, thus, demonstrating the need for more research in the area of the effect of crack front shape and direction of propagation for the general loading case. In addition to the complexity of the stress state in piping components, high fracture toughness and relatively low yield strength of the stainless steel material makes it inevitable that plastic zones develop in front of the crack tip. LEFM methods have been successfully extended for such elasto-plastic conditions provided the plasticity is completely contained by the surrounding elastic material. A number of concepts have been developed for the assessment of fracture toughness when extensive plasticity precedes failure. Some are modifications of the elastic parameters K_{IC} , others are empirical relationships desired for specific test configurations which also included a large number of cylindrical and nozzle geometrics applicable to piping components.

The J-contour integral has been utilized extensively for some piping component studies, although general fracture criteria in terms of critical value J_c , that would be applicable to complex loading conditions (requiring 3-D stress analysis) are not available. This is because the generalization of the J-analog as an energy change in a 3-D plastic regime has not yet been successfully derived.

Crack opening displacement (COD) concepts provide another method for plastic zone modification of LEFM. This method uses a correction factor based on the relationship between extension force and crack opening displacement. The COD method depends on calibration tests for determining the opening at the crack tip as a function of the applied strain and load evaluated either experimentally or by finite element techniques. This method also permits the inclusion of residual stress effects (the J-integral is insensitive to local stress fields when these are self-equilibrating because it evaluates the global loads and deflections which are not effected by residual stresses). The crack tip conditions are strongly effected by residual stresses particularly if, in the worst case, the crack lies in the tensile part of the residual stress distribution. Thus, the COD approach, although semi-empirical is capable of evaluating many geometry and loading conditions encountered in piping systems.

These elastic-plastic fracture mechanics concepts present the methods by which the fracture process is characterized by a single parameter. The J-integral examines the rate of change of the energy as the crack extends using the stress strain conditions along an arbitrary boundary far away from the crack tip. In the COD method, fracture is defined by a critical displacement δ_c , right at the crack tip. Both are consistent with the linear elastic fracture criteria based on the critical energy release rate G_c and fracture toughness K_{Ic} :

$$G_c = \frac{K_c^2}{E} = J_c = \sigma \delta_c$$

These methods, however, are applicable only to the onset of crack extension; and their general utility is limited by the restrictions imposed on the size of the plastic zone.

3.1 Subcritical Crack Growth

As described by LEFM, crack growth occurs in brittle materials when G reaches the critical value G_c with no relationship describing the increasing crack length below G_c . In case of ductile materials, a crack may grow under increasing load with the crack extending, but remaining stable until a critical crack length is reached. In LEFM, the crack tip stress intensity is represented by K as the driving force for crack growth. This concept is conveniently applicable to high cycle fatigue where the stress range is close to elastic and the plastic zone is completely contained by an elastic field. The crack growth rate $\frac{da}{dn}$ was found to follow a power law of the stress intensity range ΔK as follows:

$$\frac{da}{dn} = c\Delta K^n$$

This relationship has been extended by including terms for a lower threshold value of ΔK_0 below which crack growth is unmeasurable and for the range of increasing non-linearity when ΔK approaches the upper limit K_c , the critical stress intensity for monotonic loading. The life of a cracked component can be evaluated by integrating the crack growth equation between initial crack size a_0 and the critical crack length a_c determined from fracture toughness data. Thus,

$$N = \int_{a_0}^{a_c} \frac{da}{c\Delta K^n}$$

Alternately, the crack growth in N cycles may be determined by a summation of the crack extension increment:

$$a = a_0 + \sum_{i=1}^n c \Delta K^n$$

The subcritical crack growth relationships are reasonably accurate for essentially constant amplitude loading sequences. In general loading histories, the growth rate is reduced when a large amplitude is followed by a smaller load variation. Modeling of the sequence effects within the framework of fracture mechanics has been also developed with some success describing the process of yield zone reduction or increase due to the residual crack tip stress following an overload or load drop.

These methods are useful in evaluating piping component life histories if the fracture mechanics parameters a_c , K_c , and ΔK are available. As discussed in the previous section for many crack shapes, geometrics and load combination values will have to be determined.

The extension of the power law of crack propagation to elevated temperature conditions has also been described in a number of references. Surprisingly, good correlations can be found in spite of the reduced yield strength with temperature implying a correspondingly increased size of the plastic zone thus violating the conditions for LEFM application. The probable explanation for the extended applicability of the power law lies in the differences between monotonic and cyclic plasticity. Experimental and theoretical work indicates that the size of the plastic zone in which plastic strain reversal takes place under cyclic loading is much smaller than the extent of the plastic area under the same peak load when applied monotonically. Therefore, fracture mechanics techniques are expected to be useful in LMFBR piping evaluation particularly in high cycle life

prediction of components under relatively low stress amplitudes. For low cycle fatigue and for the evaluation of hold-time effects and other time dependent damage mechanisms, further considerations have to be introduced.

3.2 Time Dependent Crack Growth

Structural failures in components operating in aggressive environments or at elevated temperatures occur even under sustained loads. Generally, such failures are preceded by extensive cracking, and attempts were made to correlate stress intensity factors with subcritical crack growth rates. There is a large body of information available on this type of LEFM application to corrosion studies but the literature on creep-crack evaluation is very limited. Results of crack extension tests under constant loads have been presented as $\frac{da}{dt}$ versus applied stress intensity K in a manner similar to cyclic crack growth representation. This information can be used in life evaluation by integrating over the predicted service period or alternately by evaluating the time required for reaching a specified crack size. This topic of stress corrosion, however, will not be covered in this preliminary survey since chemically pure sodium has been shown by a number of studies to have little effect on crack growth in stainless steel LMFBR piping material. A possible concern, however, is that upset or emergency load conditions fast fracture can be initiated from the sites of stress corrosion cracks as these have a tendency to develop both in the longitudinal and circumferential directions.

On the other hand, the creep behavior of cracked or notched pipes and tubes will be of primary interest in the later phases of this investigation. References on work carried out report two types of behavior, the first identified mainly by ductile behavior and the second by brittle material response. Tests on small tubes containing axial defects of varying length and depth showed that the material behaved in a ductile manner in a fairly low constrained geometry. This is a

"creep ductile" condition as the main cracks did not propagate axially until the heavily creep damaged material failed by tear after the penetration of the tube wall.

Tests were reported on larger pressure pipes containing circumferential notches with diameter to thickness ratios closely approximating LMFBR geometry conditions. The specimens first showed asymmetry in crack growth which caused bending of the pipe. After appreciable crack growth that showed almost complete penetration, the final failure was by fast ductile shear fracture. Net-section type analysis overestimated the life by an order of magnitude. But fracture mechanics evaluation using stress intensity factors K derived from notched plate tests gave reasonably close agreement with the observed crack growth rates and fracture life of the specimens.

Elastic-plastic analysis for small flaws as induced by creep in ductile materials has been pursued by a number of investigators. In particular, an anti-plane strain analogy has been developed and applied to clusters or clouds of defects where the individual flaws cannot be treated by LEFM methods. Work in progress is also reported on J-integral type representation of the creep crack growth but all these efforts are still in the preliminary stage. These topics will be reviewed in the second phase of this investigation.

3.3 Life Prediction Methods in Industrial Use

Life prediction and risk evaluation are an integral part of the design and quality control procedure in a variety of industrial fields, such as gas and steam turbine manufacturing, natural gas pipelines, aircraft and engine production, etc. The life evaluation methods based on fracture mechanics theory (that deals with crack extension and propagation but not with crack initiation) by definition has to be closely related to the techniques of initial flaw detection and continuum

mechanics analysis methods. The interplay required between metallography, fracture mechanics, structural analysis, and non-destructive evaluation (NDE) providing crack information is illustrated schematically in Table II (pp.33). The impact of the sensitivity and reliability of NDE methods on life evaluation is very pronounced. Correspondingly, NDE are in the process of rapid development. Concurrently, analytical treatment of flaw severity, minimum inspection size, cost/risk optimization, and repair/replace decision algorithms are developed. For example, the "severity" of flaws refers to shape, orientation, proximity, etc., effect on the initiation and propagation of cracks and are extending the single parameter modeling of the fracture process. In turbine rotors, the flaw is considered to be a sharp fatigue crack, in castings, a cluster of inclusions and in pipelines, weld porosities, etc. Statistical analysis is used to determine the probability density functions for spacing, orientations, and sharpness of the flaws, etc. which then are used with Monte Carlo procedures to combine the effects that control the crack growth properties. The number of disciplines that have to interact to produce quantitative risk information is shown in Table II, which also illustrates how broadly the term "Fracture Mechanics" is interpreted in this newly developing field.

The problems encountered in nuclear applications will require substantial modifications of the model analysis, and inspection details on account of radiation effects and accessibility limitations, etc. It is quite clear, however, that the fracture mechanics methodology is a most promising approach to develop physically sound fatigue, creep-fatigue, creep damage evaluation, and reliable life prediction methods.

4. Conclusion, Assessment of Survey and Outline of Future Work

The purpose of this section is to present conclusions and to give a preliminary assessment of the ways in which the information presented could be analyzed and

brought forward to the point where applicability of current fracture mechanics procedures could be clarified. The objective is to evaluate the fracture mechanics application strategies in the continuation of this program.

4.1 Application to Shell Problems

It is clear that many of the cracks appearing in piping systems will be classified as "shell problems". There is a need to sort out the existing literature to establish ranges of thickness/crack length/shell diameter ratio for which the various theories are valid. (Some confusion arose initially because the Kirchhoff theory was used in many cases instead of the Reissner theory which gives a more accurate description of the near-crack stresses.) Some of this work has been carried out in some previous "reviews" such as (1.14, 1.19, 1.28) but this should be examined and any differences reconciled. Similarly, the experimental shell data on shell application in references (1.6, 1.8, 1.18, 1.21, 1.24, 1.31, 1.35, 2.23) need to be evaluated. This kind of clarification is even more necessary for crack tip plasticity or general yielding situations which are significant in the thin section/ductile regime where most piping systems operate.

4.2 The Part-Through Crack Problem

Most cracks occurring in service will be of the part-through variety and the validity of various treatments needs to be examined critically with respect to the variation of stress intensity/COD around the perimeter of the crack. This is especially true with respect to plasticity/limit load/triaxial complications which arise at the minimum section of the pipe wall (references 1.1, 1.2, 1.6, 1.7, 1.8, 1.22, 1.25, 1.28, 1.29, 1.36, 1.40, 1.41, 1.43, 2.8, 2.9).

4.3 Leak-Before-Break

The development and justification of the leak-before-break criterion will be highly desirable for any piping systems. In the past it was maintained that such a

condition could be achieved simply by ensuring that a given vessel would behave in a fracture tough mode with a crack which was long compared with the pressure-tight thickness (say 4 times). Unfortunately, there is plenty of practical evidence that part-through cracks in shells can propagate to substantial lengths without break-through occurring. This is especially true where shell bending occurs (e.g., oval pipes under pressure) or due to the loss of triaxial constraint at the break through point. In order to answer the leak-before-break question, references (1.21, 1.23, 1.24, 1.31, 1.32, 1.34, 1.38, 1.40, 1.42) should be further evaluated.

4.4 The General Yield Fracture/Limit Load

The operating regime for piping systems is likely to place fracture modes in the large strain/general yield regime because the components are in general thin and made of tough materials. It must be said that theoretical approaches in this regime are inspired rather than rigorous. This is also a topic of some controversy between the proponents of J integral, COD, net-section stress, etc. Recent reports by Rice (2.55), Kannien, et al. (2.44), Banford and Begley (2.47) Erdogan (3.20), bring this into a sharp focus. These publications and others such as (2.45, 2.46, 2.19, 2.24, 2.25, 2.26, 2.28, 2.29, 2.34, 2.37, 2.40, 2.51, 2.52, 2.61) should be reviewed in depth.

4.5 The Triaxial Constraint Problem

It was emphasized in Section 2 that the mechanical model which is to be employed must reflect the real features of the problems if it is to be useful in the practical sense. Too little attention has been paid to the problems of assessing triaxial constraint in practical situations and to reproducing such constraint in material toughness tests.

In general multiaxiality has to be dealt with in all too real problems. That is solutions must be adapted, combined, or superposed to accommodate the deviations in geometry loading temperature, etc. from the idealized cases analyzed in the

literature. In addition, many results are published in numerical form which require some intermediate interpolation or curve-fitting to facilitate practical application. Some consideration has been given to this problem in references (2.44, 2.47, 3.20, 1.5 and 2.31 for 2d geometries, but it is clear that similar development is necessary for shell and piping applications.

4.6 Fatigue Crack Propagation

It is clear that nuclear piping systems will be subject to cyclic loading and therefore, fatigue crack propagation assessment will be required. Given " ΔK " based propagation rate data for materials (and considerable data exists which was not included in the references for brevity), the application of the data to predict service life requires some linking effort (e.g., the curve-fitting procedures and plasticity correction procedures outlined in the previous section).

Again it is not clear that data exists for high strain fatigue crack propagation, and it may be that the transition from linear elastic to general yielding needs to be examined.

4.7 The Non-Linear Problem

Perhaps the most important analytical problem which has been neglected (due no doubt to great difficulty) concerns the description of cracked bodies made of entirely non-linear material. This has practical relevance to general yielding in the time-independent case and to creep. It has frequently been argued that creep crack propagation in practical situations can be described adequately by linear theory (i.e., materials are normally specified for elevated temperature so that the strain levels in the bulk of the component apart from the crack region are essentially elastic). However, this point has not so far been successfully resolved with regard to typical components operating in the creep regime, and therefore, a high priority should be given to determination of the relative levels of elastic and creep strain in piping structures to be considered.

4.8 Conclusion

The application of fracture mechanics to practical problems raised by design or casualty analysis is rarely easy. It requires considerable analytical effort to relate and adapt available information to the cases which arise. Some of the areas which need to be resolved are identified above and no doubt, others could be added. These topics may provide a framework for the analysis and assessment. It is worth pointing out that it is very easy for analytical effort to be directed at hypothetical problems, which in the event turn out to be of small significance. Therefore, one of the tasks of this program will be to channel the analytical and test developments along lines that are directly applicable to NRC safety evaluations. It is our assessment that these problems can be solved but it is necessary to bring forward current problems and designs at an early stage so that the evaluation effort may be channeled along meaningful lines.

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(To save space the titles of some journals are abbreviated as follows:

- IJFM - International Journal of Fracture Mechanics
- EFM - Engineering Fracture Mechanics
- IJPVP - International Journal of Pressure Vessels and Piping
- ICF - International Conference on Fracture
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