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ESTIMATE OF THE FATIGUE
CRACKING BEHAVIOR
OF THE
FAST TEST REACTOR PRIMARY PIPING

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OF THE
FAST TEST REACTOR PRIMARY PIPING

By

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This report prepared for the FFTF Project.

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ABSTRACT

The concepts of linear elastic fracture mechanics are applied to the FTR primary piping. Estimates are made of fatigue crack propagation behavior for service conditions, and crack length versus cycles information is estimated.

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INTRODUCTION

Cyclic loads and aggressive environments, or their combination, could induce the growth of innocuous flaws in the Fast Test Reactor (FTR) primary piping system to critical sizes during system operation. The objective of this work is to estimate the fatigue crack growth behavior of typical piping components based on fracture mechanics analysis techniques coupled with the knowledge of the fatigue crack growth properties of the system construction material.

Reliable operation of the primary piping system is mandatory to ensure safe and controlled cooling of the FTR. Loss of such flow is expected to be one of the most severe of possible system malfunctions. It is possible that a decrease or cessation of flow brought about by excessive leakage or rupture within the system could damage and possibly destroy the reactor internal structure before appropriate control measures could be undertaken to shutdown the reactor. Thus, a high degree of system reliability is imperative.

Even though a large body of data exists on the high temperature tensile and creep properties of materials, design for high temperature service life of complex structures is very much an art primarily because the subcritical cracking behavior of materials in service environments is not generally known and, reliable analytical methods have yet to be developed that characterize the effects of variable amplitude loading on fatigue life. Also, the determination of structural behavior

based on simple specimens can be a complicating factor, but the severity of the problem is reduced by using fracture mechanics correlation techniques.

The primary piping system is a large complex structure and is expected to operate at combined temperature, pressure, and flow rate levels exceeding those of most sodium systems. The levels are expected to be 1200 °F with sodium pressure and flow rates of about 200 psig and 5×10^6 lb/hr. The largeness of the system is indicated by the pipe dimensions that are anticipated to be 12 to 30 in. in diameter and 3/8 to 1/2 in. thick. In addition to high temperature levels, nonuniform variations in temperature promote fatigue crack growth as a probable mode of system failure because of the relatively high coefficient of expansion and low thermal conductivity of Type 304 SS.

Crack initiation, subcritical crack propagation, and fracture are properties vital to a complete service life analysis; however, subcritical crack growth is emphasized in this report for several reasons. First, crack initiation is not a well-defined property and by eliminating it, component crack-growth analyses are more conservative. Second, flaws or cracks will either be "built-in" or generated during system operation, thus promoting fatigue crack growth as the dominant phase of the total failure process. Fracture toughness is important as it represents an upper bound to subcritical crack growth; however, it is not considered a principal design variable because Type 304 SS is highly ductile. Barring severe metal embrittlement, subcritical crack growth resulting in "leak-before-break" is expected to be the dominant system-failure mode.

It was the intent of the authors to explore the application of fracture mechanics analysis techniques to subcritical crack growth as effected by both cyclic stresses and aggressive environments. Unfortunately, very little meaningful data exist

on those properties of Type 304 austenitic stainless steel. Essentially, only limited room-temperature fatigue crack-growth data are available, and thus the title of this work was restricted to estimating the fatigue crack-growth behavior of piping components. Rough estimates of the influence of sodium on fatigue crack growth are made to estimate the possible range of component cracking behavior. Although admittedly crude, it serves the purpose of providing a design "estimate" of the fatigue crack-growth behavior of defected system components undergoing cyclic loading. No attempt is made to estimate the environmentally-induced sustained-load crack-growth behavior of 304 SS, nor have any tests been conducted to confirm its existence.

SUMMARY AND CONCLUSIONS

A survey⁽¹⁾ into probable causes and modes of failure in non-nuclear pressure piping systems was summarized and the causes categorized and discussed with respect to the FTR primary piping system. The basic principles of linear elastic fracture mechanics were reviewed, and it is shown how the existing solutions may be applied to FTR components. In general, solutions exist for straight pipes, but there are no solutions for more specialized shapes such as nozzles, elbows, tees, etc. Methods for determining the solutions for these specialized shapes are discussed.

The crack propagation properties of 304 austenitic stainless steel were surveyed, and it was found that very little work has been done at room temperature, also, no work has been done to establish the properties at elevated temperatures or in the operating environment of the FTR. Crack propagation behavior under various operating conditions was estimated, but these estimates represent only rough, "educated guesses" and it is quite possible that they are considerably in error.

Finally, based on the estimated crack propagation rate data, curves are presented that estimate the number of cycles required to grow a crack in FTR piping from a finite size to a size where it may hopefully be detected.

From this report it was concluded that FTR system failures can usually be expected to manifest themselves by leaking prior to fracture, provided that the construction materials maintain adequate ductility and corresponding fracture toughness.

Analysis of the Pipe Rupture Failure Study indicates that:

- Long-term subcritical crack growth explicitly accounted for about 54% of all the pressure piping system failures.
- Ostensibly 29% of the failures were attributed to fatigue, but in reality the value is most likely larger.
- 61% of all failures occurred in systems operating in the 600 °F to 1200 °F range; the remaining failures occurred below 600 °F.
- About 5% of the failures occurred by complete severance within the system, with fatigue being the predominant failure mode; and the failures occurred within a time span of 60 hr to more than 20 yr.
- The majority of system failures were located in runs of pipe (44%) and welds (41%).
- If welds are classed as components and the probability of failure in the components of a "typical" BWR pressure piping system were equal, failures would occur primarily in runs of pipe and welds because they comprise the largest percentage of the total system population.

Fatigue crack growth is a probable mode of system failure because of the more acute sensitivity of Type 304 austenitic stainless steel to thermal fatigue.

Other conclusions are as follows:

- Flat plate fracture mechanics solutions will satisfactorily predict the stress intensity factor K , for axial and circumferential cracks in those straight sections of FTR pipe that have been considered.
- K -solutions are not generally available for complex shapes such as nozzles, tees, etc., but methods do exist for obtaining approximate solutions for specific geometries.
- For the typical FTR pipe configurations examined in this study, estimated operating pressures and loadings appear to be low enough so that unstable fracture at room temperature and above seems improbable.
- Although unstable fracture of the straight pipe examined is unlikely at elevated temperatures, crude estimations of material cracking behavior under some conditions indicate that large stable increments of crack extension could occur during each load cycle, which would result in a fairly "rapid" pipe separation or failure.
- Only a small amount of subcritical cracking data on 304 SS exists for room temperature conditions, and no information is available for elevated temperatures and/or aggressive environments.
- It has been shown for other materials that crack-growth rates can be markedly influenced by nonroom temperature and environmental conditions. It can only be assumed that Type 304 austenitic stainless steel is not an exception to this observation.
- Results from low-cycle fatigue tests suggest that sodium containing oxygen and carbon as impurities will influence the fatigue crack-growth rate of 304.
- Until fatigue-and sustained load-tests are conducted by using notched specimens in simulated environments, an adequate estimate of cracking behavior of any system component or LMFBR alloys will not be possible.

DISCUSSION

BACKGROUND AND FAILURE SURVEY

Background

The potential results of catastrophic failure of reactor-system piping dictate a greater degree of safety assurance than in non-nuclear systems. Complete severance or fracture of the primary piping appears to be one of the most severe failure conditions. Spontaneous fractures in non-nuclear piping systems, however, are rare.⁽¹⁾ Barring severe earthquake or conditions leading to rapid and gross increases in reactor loadings or temperature, failures can be expected to manifest themselves by leaking prior to failure if the system materials remain ductile. The tentative selection of AISI 304 SS as the primary piping material virtually rules out brittle fracture, providing certain criteria are met. Such criteria will be discussed later. Spontaneous metal fracture, however, is preceded by a sequence of events which are often detectable at various stages and are potentially predictable.

Elements of Metal Fracture

Fracture of metals is an inhomogeneous process, preceded by and accompanied with localized plastic deformation. Such processes cause regions of the material to separate and load carrying capacity to decrease to zero. High, localized stresses at the tips of inhomogenities such as flaws or cracks are required to break atomic bonds. Fracture results from the passage or growth of a crack through a region of the material. Thus, factors that effect the rate of crack growth and critical crack size will influence a component's service life.

Fracture of a tensile specimen results from a monotonically increasing load which brings about local plastic flow or

necking with subsequent crack formation and growth. It is now recognized that fatigue occurs as a result of plastic deformation, both in the formation and propagation of cracks. Sustained load failures can occur in a given environment by either a general corrosive attack, which can eventually lead to crack growth caused by an increase in stress with a loss of load bearing material, or by localized attack (e.g., stress corrosion) where cracks form and grow to result in failure when the more advanced of these cracks reaches a critical size. Creep rupture occurs under static loads at temperatures generally greater than $\frac{T_m}{2}$ (T_m - absolute melting point); crack growth takes place as the material plastically elongates under constant stress. Since plastic deformation is always an essential requirement if failure is to occur, fracture is always preceded by events related to plastic deformation and subsequent crack growth. Thus, any attempt to estimate the service life of a metal component hinges on knowing the effect of loading and environment on the growth of a small or subcritical crack to a critical size. Fracture mechanics techniques have been used successfully to characterize subcritical crack growth and will be discussed later. First, however, it is important to identify that type of subcritical cracking predominant in pressure piping systems.

Pipe-Rupture Failure Study

A study⁽¹⁾ was conducted to establish the nature and apparent causes of failure in pressure piping systems; it covered steam power plants, refineries, marine applications, chemical plants, paper mills and included designers, fabricator-erectors, piping manufacturers, etc. The purpose of the study was to provide a basis for establishing the reliability of nuclear reactor primary piping systems.

The survey was successful in that causes of failure were related chronologically to problems arising from initial design through the building and operation of a pressure piping system. (See Appendix for detailed discussion of "Survey of Piping Failures for the Reactor Primary Coolant Pipe Rupture Study" GEAP-4574). Essentially, however, it can be argued that the failures were related to flaws that were either initially present or generated-in-service, and were caused to grow to critical sizes by a combination of loading and environmental effects.

Expected Failure Locations in a Reactor Primary Piping System

Comparison of the failure survey results⁽¹⁾ with a "typical" Boiling Water Reactor Primary Piping System⁽²⁾ reveals that failures in the latter can occur everywhere but are likely to be concentrated in either weldments or runs of pipe. A detailed discussion is also presented in the appendix (Table A-2) where the breakdown of components in a "typical" BWR piping system is compared to failure locations cited in Reference 1. Failures in welds (41%) and in runs of pipe (44%) were the two largest failure location categories. Table A-2 also indicates that failures in a nuclear primary piping system could occur in proportion to the population of given components but will be concentrated in pipes and welds simply because those two categories usually have the largest population.

Failure Cause and System Operation

Analyses of 399 failures as shown in Table A-1 represent a significant effort to establish and categorize the dominant causes of failure in pressure piping systems. However, a more detailed comparison of the same data [Table A-3 (Classification of Failures) and Table A-4 (Primary Causes of Failure)] shows a discrepancy in the number of failures attributed to fatigue. This is important because it is an indication of the inconsistencies in the original failure analyses.

Barring conditions (as noted before) that bring about rapid and gross increases in reactor loadings or temperature levels, it is very likely that failures occurring within the Fast Test Reactor Primary Piping System will result from long-term subcritical crack growth. Moreover, it is expected that stress risers and especially weld defects will be the most probable sites for crack initiation, and fatigue loading (mechanical and/or thermal) will be the most probable mode of subcritical crack growth. With this as background, a detailed study was made of all the 399 failures listed in the appendices of Reference 1. The intent was to isolate those failures that could be attributed to long-term subcritical crack growth. Failures occurring during system startup were not considered. In some cases engineering judgement was necessary, owing to the limitations of the original failure analyses. Analysis of 216 cases (54% of 399) indicated that

- 1) The failures manifested themselves as single or multiple cracks and
- 2) The failures occurred during time periods ranging from 60 to 307,000 hr (35 yr).

During system operation, subcritical cracks grew to critical* sizes.

The mode of failure or subcritical crack growth of the 216 failures is shown in Table 1.

The number of failures occurring by an unknown crack-growth mechanism suggests that the number of failures attributed to fatigue in Table A-3 is in reality probably much greater than the indicated 29.1%. This is pertinent to the FTR primary Piping System because 304 SS is more susceptible to fatigue

* *Critical, within the 216 cases, means either complete rupture, leakage or detected before leakage and the system shutdown.*

TABLE 1. Mode of Failure of 216 System Failures

<u>Apparent Mode of Failure</u>	<u>Cases</u>	<u>% of Total</u>
Fatigue	80	37
Stress-corrosion	64	29 1/2
Unknown crack- growth mechanism	72	33 1/2

(especially thermal fatigue) because its coefficient of expansion is about double while its thermal conductivity is about 1/2 that of plain carbon steel.

This discussion is not intended to detract from the work presented in Reference 1. The authors should be congratulated for piecing together a large number of obscure and often incomplete failure analyses. The "causes" of failure, as noted in Table A-1, are probably accurate, inasmuch as they describe reasons for failure based on design, material selection and operation-service. However, it is the mode of subcritical crack growth that remains suspect.

Fracture Mechanics and Subcritical Crack Growth

The primary piping system of the FTR must be designed with Table A-1 used as a very real and pertinent guide. It remains then to assure the designer that the piping system structural materials are resistant to subcritical crack growth under operating conditions characteristic of the FTR system. No matter how crack-resistant or tough a material may be, poor design, manufacturing, fabrication, erection and operation practices can drastically lower the reliability and integrity of the whole system. On the other hand, even in a well-designed system, defects will most surely be built into the system if not generated in service. Thus, the most practical recourse to ensure safe and dependable use of the system is to provide design criteria that preclude the growth of flaws by subcritical

crack growth mechanisms, to a critical size during the piping design lifetime. Linear elastic fracture mechanics analyses techniques, coupled with knowledge of the effect of loading, configuration and environment on cracking, provide the most viable approach to estimating the cracking behavior of the primary piping system.

APPLICATION OF FRACTURE MECHANICS ANALYSIS TECHNIQUES TO THE PRIMARY PIPING SYSTEM

Introduction

The theory of linear elastic fracture mechanics (Reference 3, pages 3 through 11) is based on the Griffith-Irwin relationship

$$K = \sigma\sqrt{\pi a} \quad (1)$$

which, for a flat plate of infinite width, relates the stress intensity factor K to the remotely applied gross section stress and the crack length (see Figure 1 and Notations on page 12). Brittle fracture will occur when the magnitude of K exceeds the critical stress intensity factor K_{IC} . This basic relationship has been extensively modified to cover various cases of loading and crack geometry, and also to account for the effect of finite dimensions. With few exceptions, however, the majority of the linear elastic fracture mechanics solutions are for two-dimensional problems of planar configuration. There have been some attempts to solve nonplanar problems or to extend flat plate solutions to include the effects of curvature. This section will examine some of these solutions with emphasis on how they may be applied to the primary piping system of the Fast Test Reactor.

$$K_I = \sigma \sqrt{\pi a}$$

$2a$ = Crack Length

σ = Gross Stress

K_I = Stress Intensity Factor

FIGURE 1. Crack in an Infinite Sheet

In effect, it will be an application of the state-of-the-art to the FTR. Proposed FTR configurations will be discussed, and the applicability of the various solutions to these configurations will be examined.

Circumferential Cracks and Notches

One possible failure mechanism in the primary piping system is the circumferential crack which when subjected to cyclic loading, propagates circumferentially around the pipe. The crack could initiate at a mechanical defect such as a surface flaw or at a metallurgical defect such as a weldment. Once

initiated, the crack could propagate through the pipe wall and then around the circumferential direction. The cyclic loading could result from fluctuations of internal pressure, thermal stresses or external bending moments. The possible failure modes could be characterized by the two extreme cases: ductile rupture resulting in large amounts of plastic deformation; and brittle fracture with only a small amount of plastic deformation. An actual failure would probably be a combination of the above modes and would be dependent on the material, the operating environment and, to a lesser degree, geometry and loading conditions. (Sharp notches, section changes and plastic strains at the crack tip may have an embrittling effect by the constraint they offer to plastic deformation.) It is possible that, under certain conditions, a crack of a given length could be stable at the operating temperature and then propagate in a "brittle" fashion at the same load when the temperature is decreased. This is because the critical stress intensity factor, K_{IC} , generally decreases with temperature. The effect of temperature and other operating environments on material properties will be dealt with extensively in another section.

The analysis of circumferential crack made by Folias⁽⁴⁾ is based on the coupled differential equations of Marguerre governing the stress function and the displacement function, and it is somewhat cumbersome to apply. The applicability of the Folias solution is further limited by its validity only for short cracks and large radii of curvature. The Folias solution will not be described in any more detail here because, although it gives very accurate results, a much simpler and satisfactory approach will be given later.

An analysis by Gilman⁽³⁾ allows the treatment of a larger range of problems. The limitations on the Gilman analysis are not quite as severe as the Folias solution; in addition,

it includes provisions for external bending moments as well as internal pressure. The displacements of the cylindrical shell are expressed as a Fourier series, and the strain energy due to shell bending and extension is a function of the Fourier coefficients. Strain energy is expressed as a function of crack length. The calculations are somewhat involved and a digital computer is required to obtain the solution. (Gilman solved 141 simultaneous equations, and sufficient terms must be used to insure convergence of the Fourier series.)

The above discussion would be discouraging for anyone interested in only an approximate estimate of the cracking behavior of a tubular component if it were not for the observation by Gilman that, for cracks up to about 60° , the simple flat plate relationship of Equation (1) will give satisfactory results.

A circumferential crack will be defined as shown in Figure 2. Beta is one-half the angle subtended by the crack (in radians), and ρ is the radius to the midsurface of the cylindrical shell (Gilman used the notation "a"). The results of the Gilman analysis for a cylinder of unit radius are shown in Figure 3. Three loading conditions (pressure, bending, and pressure plus bending) are shown as well as the flat plate solution represented by Equation (1). The strain energy release rate G , may be related to the stress intensity factor K , by the following equations:

$$K = \sqrt{GE} \quad (\text{plane stress}) \quad (2a)$$

$$K = \sqrt{\frac{GE}{(1 - \nu^2)}} \quad (\text{plane strain}) \quad (2b)$$

As may be seen in Figure 3, for angles up to approximately 60° ($\beta = 30^\circ$), the strain energy release rate G (and therefore K),

FIGURE 2. *Circumferential Crack*

may be satisfactorily expressed by the flat plate solution of Equation (1) regardless of whether the loading is bending or pressure or a combination of both. In effect, the cylinder may be "rolled up" so that the flat plate relationship

$$K = \sigma \sqrt{\pi a} \quad (1)$$

becomes

$$K = \sigma \sqrt{\pi \rho \beta} \quad (3)$$

If the loading condition is internal pressure only, σ in Equation (3) may be replaced by $\sigma_h/2$, and Equation (3) becomes

$$K = \frac{\sigma_h}{2} \sqrt{\pi \rho \beta} = 0.8862 \sigma_h \sqrt{\rho \beta} \quad (4)$$

FIGURE 3. Strain Energy Release Rate per Unit Cylinder Radius(3)

or alternatively, the relationship in terms of strain energy release rate becomes (assuming plane stress)

$$G = \frac{\sigma_h^2 \pi \rho \beta}{4E} \quad (5)$$

It appears reasonable that Equations (4) or (5) could be satisfactorily used in describing the fracture behavior of the primary piping in the FTR for cracks up to 60°. It is unlikely that, in a nuclear reactor, a crack one-sixth of the circumference of a pipe would go undetected. Using the proposed FTR piping sizes of

$$\begin{array}{lcl} \text{OD} = 30 \text{ in.} & \left. \vphantom{\begin{array}{l} \text{OD} = 30 \text{ in.} \\ t = 0.375 \text{ in.} \end{array}} \right\} & \text{pump suction} \\ t = 0.375 \text{ in.} & & \\ \text{and} & & \\ \text{OD} = 12 \text{ in.} & \left. \vphantom{\begin{array}{l} \text{OD} = 12 \text{ in.} \\ t = 0.375 \text{ in.} \end{array}} \right\} & \text{pump discharge,} \\ t = 0.375 \text{ in.} & & \end{array}$$

this would suggest crack lengths of approximately 15.7 and 6.28 in. for the suction and discharge piping, respectively. There have been no known leakage rate tests run on circumferential cracks in pipes of these dimensions, but it has been found that for axial through-wall cracks in PRTR pressure tubing (OD = 3.56 in., $t = 0.155$), detectable leaks occurred at crack lengths of 0.50 to 0.60 in. It would, therefore, seem reasonable that cracks would be detected in the FTR piping long before they reached the 60° limitation of the flat plate approximation and that Equation (4) or (5) may be used in analyzing such cracks. It is doubtful that it would ever be necessary to use the more complex solutions of Gilman⁽³⁾ or Folias.⁽⁴⁾

The curves shown in Figure 3 are based on a cylinder with an outside diameter to wall thickness ratio of 21. The corresponding ratios for the suction and discharge piping are

80 and 32, respectively. Although there is quite a difference in the absolute dimensions of the example shown in Figure 3 and FTR piping, the flat plate approximation would be valid since the ratios of the FTR pipes approach a flat plate ($OD/t = \infty$) even more closely than the example.

Figure 4 gives the K and G relationship for the suction tubing using the flat plate approximation. The curve is based on a maximum pressure of 5 psig, and hoop stresses are based on the Lamé' solution.⁽⁵⁾ The same information for the discharge piping is shown in Figure 5, and is based on a maximum pressure of 200 psig. In Figures 4 and 5, the K relationships are valid at all temperatures since K is dependent only upon geometry and loading, but the G relationships are valid only at the temperature shown since they are a function of the elastic modulus. This should not be too inconvenient because fracture properties are almost always quoted in terms of K_{IC} rather than G_{IC} .

The stress intensity factor relationship for an applied external bending moment may be easily calculated by using the flat plate approximation. The stress intensity factor for a circumferential crack is [recalling Equation (3)]

$$K = \sigma \sqrt{\pi \rho \beta}$$

and the stress due to an applied bending moment is, according to elementary strength of materials theory,

$$\sigma_b = \frac{My}{I} \quad , \quad (6)$$

Substituting in the relationship for the moment of inertia of a thin tube

$$I = \pi \rho^3 t \quad (7)$$

FIGURE 4. Stress Intensity Factor and Strain Energy Release Rate Relationships for a Circumferential Crack in FTR Primary Piping (Suction) Subjected to Internal Pressure

FIGURE 5. Stress Intensity Factor and Strain Energy Release Rate Relationships for a Circumferential Crack in FTR Primary Piping (Discharge) Subjected to Internal Pressure

and combining Equations (3), (6) and (7) to solve for the stress at the remote fiber, we get

$$K = \frac{0.5642 M}{(\rho)^{3/2} t} \sqrt{\beta} \quad (8)$$

The expression given in Equation (8) is not entirely correct since the flat plate formula of Equation (3) assumed a uniform remote tensile stress field. The tensile stress of the pipe in bending varies from zero at the neutral axis to a maximum at the remote fiber. Therefore, the K predicted by Equation (8) will be somewhat high, but this will not be too significant for the short cracks considered in this section, and use of Equation (8) will result in a conservative design.

For the loading case of internal pressure plus applied bending moment, the individual stress intensity factors may be merely added together (Figures 6 and 7). As an example, if a discharge pipe were subjected to 200 psig internal pressure and a bending moment of 100,000 in-lb, the stress intensity factor for a total crack of 57.3° ($\beta = 0.5$) would be

$$\begin{aligned} K &= K_{\text{pressure}} + K_{\text{bending}} \\ &= 4688 + 7592 \\ &= 12,280 \text{ lb/in}^{3/2} \quad (9) \end{aligned}$$

The fracture toughness or critical stress intensity factor, K_{IC} , has not been defined for room temperature but will undoubtedly not be less than 100,000 lb/in^{3/2}. Although the subject will be covered in detail in a later section, it is difficult to see how brittle fracture could be a problem under the example loadings. It should be even less of a problem at elevated temperatures since K_{IC} is expected to increase with temperature.

FIGURE 6. Stress Intensity Factor for a Circumferential Crack in FTR Primary Piping (Suction) Subjected to an Applied Bending Moment

FIGURE 7. Stress Intensity Factor for a Circumferential Crack in FTR Primary Piping (Discharge) Subjected to an Applied Bending Moment

Eiber et al.⁽⁶⁾ have investigated failures of pressurized pipes with cracks at several orientations. The crack orientations were, 0, 30, 60 and 90°, to the longitudinal axis of the pipe. They developed a semi-empirical expression for the stress intensity factor at failure (K_{cr}) for the axial crack, but had little or no success formulating similar expressions for cracks at other orientations. They did observe, however, that the greater the angle between the crack and the axis of the pipe, the higher the failure stress for a constant defect length. For cracks at 90° (circumferential cracks), failure occurred at approximately the ultimate strength of the material, even though the largest crack represented over 20% of the pipe circumference.

As already discussed, the flat plate approximation should give satisfactory results for almost any circumferential crack that will be encountered. If, however, a situation arises that cannot be analyzed by the methods presented thus far (for example, a tapered pipe) another method does exist for calculating the stress intensity factor. It is called the "correlation technique" and has been successfully applied by James and Anderson⁽⁷⁾ to axial cracks in a pressurized cylinder. The method is quite general and could be easily applied to circumferential cracks. This technique correlates equivalent cracking behavior between two specimens on the basis of their crack growth rates. One specimen represents the actual component, and the other specimen is a simple, flat panel with an analytically tractable solution for K . Fatigue tests are conducted on both specimens and crack growth rates for each determined as a function of " a ". Since the K -solution (as a function of " a ") is known for the flat specimen, the specimen representing the component must be characterized by the same value of K at the same growth rate. This concept is expressed in the many "power laws" of crack growth^(8,9)

$$\frac{da}{dn} = C (\Delta K)^n \quad (10)$$

where C and n are constant for a given material and for given conditions (temperature, environment, etc.). Equation (10) is a function of the material and test (or operating conditions), and is independent of loading and/or geometry.

Although it is unlikely that it will be necessary to employ the correlation technique to find the K-solution of a length of straight pipe, the reverse application of the technique may be very useful. Cracking behavior of the actual component under experimentally inconvenient conditions (high temperature, liquid sodium, etc.) can be estimated by conducting fatigue tests on analytically tractable flat specimens under the operating environment to obtain the material behavior [Equation (10)]. The material behavior may then be combined with the K-solution [i.e. Equation (4)] to arrive at an estimate of crack length as a function of cycles (2a versus n). This procedure thus provides a reasonable estimate of the life of the component under the operating conditions and environment. This will be dealt with in more detail in a later section.

Axial Cracks and Notches

Just as the circumferential weld provided an excellent site for the nucleation and propagation of circumferential fatigue cracks, the longitudinal seam weld provides the same opportunities for the growth of axial cracks. Pressurized cylinders containing axial cracks have been the subject of several investigations,^(6,7,10-17) but to date, none has been able to arrive at a universally applicable failure criterion or a universally applicable solution for the stress intensity factor. References 7, 13, 14, and 17 deal with stress

intensity factor solutions and the other studies are concerned with critical crack lengths and/or critical pressures. All of the studies have dealt with internal pressures, and none have considered the case of applied external bending moments.

It has been established that the critical crack length for an axial crack in a pressurized cylinder is generally less than a flat sheet under the same loading conditions. Therefore, it appears that use of a flat sheet approximation, like the one used for circumferential cracks, could result, in certain instances, in an unconservative estimate of the fracture strength.

The existing stress intensity factor solutions will be briefly reviewed here and one of them will be used as an example to illustrate how it could be applied to the FTR primary piping. Reference 7 deals with the "correlation technique", and the experimental method has already been described in the section dealing with circumferential cracks. The correlation technique was applied to axial cracks in PRTR pressure tubing with a OD/t ratio of $3.56/0.155 = 22.97$. This method of solution was employed because the analytical solutions^(13,14,17) appear to be valid only for cylinders with larger ratios and, therefore, more amenable to treatment with thin shell theory. The results of the correlation technique are compared with the Catanach solution⁽¹⁷⁾ in Figure 8. As may be seen, there is considerable variation in the two solutions, indicating that the Catanach solution is not valid for cylinders of PRTR size. At the present time, it is not known if K is dependent only upon the diameter to thickness ratio, or if the absolute dimensions of these parameters are also important. The Catanach solution was experimentally verified for a thin cylinder ($OD/t = 8.0/0.050 = 160$) and could probably be used for predicting the behavior of the suction piping ($OD/t = 30.0/0.375 = 80$).

FIGURE 8. Stress Intensity Factor Relationship for an Axial Crack in a PRTR-Size Pressure Tube

Whether or not the Catanach solution would still be valid for the discharge piping ($OD/t = 12/0.375 = 32$) is questionable. It is quite possible that the correlation technique or some other method would be required to verify the stress intensity factor relationship for the discharge tubing. The Catanach solution⁽¹⁷⁾ is based on linearly superimposing the stress intensity factor due to extension (hoop stress) K_e , and the stress intensity factor due to the bending of the crack tips K_b

where

$$K = K_e + K_b \quad (11)$$

or rewritten

$$K = K_e(P_e + P_b) \text{ outer surface} \quad (12a)$$

$$K = K_e(P_e - P_b) \text{ inner surface} \quad (12b)$$

where

$$K_e = \sigma \sqrt{\pi a}$$

$$\begin{aligned} P_e = & 1.004 - 0.062432(\lambda) + 0.67419(\lambda)^2 \\ & - 1.0367(\lambda)^3 + 1.0184(\lambda)^4 - 0.49949(\lambda)^5 \\ & + 0.094455(\lambda)^6 \end{aligned}$$

$$\begin{aligned} P_b = & \frac{1}{\sqrt{1/3(1 - \nu^2)}} \left[5.6207 \times 10^{-5} - 0.0016268(\lambda) \right. \\ & + 0.146(\lambda)^2 - 0.2501(\lambda)^3 + 0.21721(\lambda)^4 \\ & \left. - 0.097711(\lambda)^5 + 0.017632(\lambda)^6 \right] \end{aligned}$$

$$\lambda = \left[12(1 - \nu^2) \right]^{1/4} \frac{a}{\sqrt{\rho t}}$$

Applying the Catanach solution to the FTR suction piping, we get the result shown in Figure 9. The solution is valid only for $\lambda < 2.2$, and this corresponds to $a = 2.85$ for the suction piping. Applying the 5 psig operating pressure (197.5 psi hoop stress), the Catanach solution predicts $K_{\max} = 1230 \text{ lb/in}^{3/2}$ on the outer surface for a half-crack length of 2.85 in. It appears that, for room temperature at least, an axial crack of total length = 5.70 in. will be far from critical at the operating pressures, and it is quite probable that a crack of this length would be easily detectable by its leakage rate.

The solution by Anderson and Sullivan⁽¹⁴⁾ also calculates K as the superimposition of hoop membrane and bulging (bending) components. [See Equation (11)]. The method of solution is somewhat different from that of Catanach, and requires the determination of empirical constants by conducting additional tests.

Eiber et al.^(6,15) have developed a semi-empirical relationship to predict the "critical" stress intensity factor, K_{cr} , for large pipeline components. The author has used the Eiber criterion on failures of PRTR pressure tubing and found the predicted values to be about twice the actual values, but other investigators have had better success with other geometries and the criterion is widely used. (The author has had somewhat better success with a similar criterion given in Reference 22.) Because of the relatively large size of the FTR primary piping, it is expected that the Eiber relationship might give reasonable failure predictions for axial cracks. The Eiber criterion is shown in Equation (13).

$$K_{\text{cr}}^2 = \frac{\pi c \sigma^2}{\cos \theta} \left[1 + \frac{5\pi \lambda^2}{32} \right] \left[\frac{4 - \kappa}{2} \right] \quad (13)$$

FIGURE 9. Stress Intensity Factor for FTR Primary Piping (Suction) Containing an Axial Crack Using the Catanach Solution

where

K_{cr} = critical stress intensity factor, ksi $\sqrt{\text{in.}}$

c = half crack length of axial through-wall crack, inches

σ = nominal circumferential stress at failure, ksi

$$\lambda^2 = \frac{c^2}{Rt} \sqrt{12 (1 - \nu^2)}$$

R = radius of vessel, inches

t = thickness of vessel, inch

$$\theta = \frac{\pi}{2} \frac{\sigma}{\sigma_c}$$

σ_c = failure stress of unflawed vessel or flow stress, ksi

$\kappa = (3 - 4\nu)$ plane strain, $(3 - \nu)/(1 + \nu)$ plane stress for axial cracks

Miscellaneous Problems

Due to the very specific nature of the designs of nozzles, tees, elbows, etc. and the mathematical complexities that would arise, there have been no general solutions for the stress intensity factor for these shapes. This lack of solutions is probably not too important in the case of tees, elbows and other standard plumbing fittings because their designs have empirically evolved over the years and satisfactory behavior has generally resulted in spite of this lack. However, nozzles and similar structures are another matter. They are usually "nonstandard" in that they are designed for a specific application and are often dimensionally different from one another and are subjected to different loadings. The nozzles are normally welded to the pressure vessels, and this produces an excellent site for a crack nucleation and propagation.

Unfortunately, a mathematically rigorous analysis of the components is beyond the present state-of-the-art of linear elastic fracture mechanics. There are, however, two techniques that could be used to estimate the stress intensity factor. The first is the "correlation technique"⁽⁷⁾ in which a flaw (surface or through-wall) is introduced in the nozzle and fatigue-cycled to obtain the crack growth rate data. This requires that a test specimen be built that is dimensionally representative of the component, also, it must be subjected to similar loadings. It is, however, not necessary for the specimen to be made of the same material as the component, or to be subjected to the operating environment. This is described in more detail in Reference 7a.

The second method that could be employed to analyze a nozzle is one used by Watwood.^(18,18a) He analyzed a suspected "delamination" in the nozzle area of the Containment Systems Experiment vessel. The defect was assumed to be an internal "plate" approximately four inches in diameter and appeared to be a delamination of the A-212B steel. The analysis is basically a "compliance" calculation⁽¹⁹⁾ utilizing the "direct stiffness matrix" method.⁽²⁰⁾ The computer program used was AXISOL which is fairly typical of a number of such codes available. This program solves axi-symmetric problems and Watwood therefore assumed that the defect was axi-symmetric (i.e., the defect extended all of the way around the nozzle). This assumption would probably predict a higher value of K and would, therefore, be a "conservative" solution.

It is quite possible that either the experimental correlation technique or the finite elements approach used by Watwood could be usefully employed to examine the influence of flaws on some of the areas of the FTR primary piping system otherwise analytically intractable. The existing design analysis techniques and the existing pressure vessel codes are not presently applicable to the case of a flawed component.

Surface Flaws

Surface flaws are frequently encountered in service and may take the form of cracks, notches, grooves, gouges, etc. If they remain undetected, they may grow into through-wall cracks, or they may be severe enough to cause brittle fracture themselves. The stress intensity factor of a surface flaw is a function of the flaw depth as well as the length and is, therefore, a more complicated relationship than that for a through crack. The basic analysis may be found in Reference 21; pages 72a-72c of Reference 22 illustrate the solution applied to axial surface flaws in Sch 40, 80, and 160 pipe. Another treatment of surface flaws in pressure vessels is shown in Reference 23.

Eiber et al.⁽⁶⁾ have developed a simple empirical relationship that predicts the failure stress for axial surface flaws in pressurized pipes. In addition, the relationship gives some indication of whether the flaw will leak or whether unstable rupture will result. The flaw is illustrated in Figure 10, and the formula is given in Equation (14).

$$\sigma_{fs} = \sigma_c \frac{A_o - A}{A_o - \frac{A \sigma_{ft}}{\sigma_c}} \quad (14)$$

where

σ_{fs} = failure stress for surface flaw, ksi

σ_c = failure stress for unflawed vessel or flow stress, ksi

$A_o = Lt, \text{ in.}^2$

L = length of surface flaw, in.

t = wall thickness, in.

A = the area of the defect, in.²

d = depth of flaw, in.

σ_{ft} = failure stress for through-wall flaws of length L , ksi

FIGURE 10. Surface Flaw Configuration

Spherical Shells

It is not certain whether or not the need will arise to investigate flaws in pressurized spherical shells. However, it may be possible to approximate some of the system components by a spherical shell. A solution for this problem has been developed by Folias⁽²⁴⁾ and is similar in form to his solutions previously discussed.^(4,13)

SUBCRITICAL CRACK GROWTH

The primary causes of subcritical flaw growth are cyclic loading (fatigue) and aggressive environments. For both fatigue⁽²⁵⁾ and environmental cracking,⁽²⁷⁾ experiments have shown that the stress intensity factor K is the primary factor controlling the subcritical flaw growth rate.

Crack Growth Under Cyclic Loading and the Stress Intensity Concept

The usual method for describing the fatigue properties of metals is to plot an alternating stress value $\Delta\sigma$, versus the number of cycles to failure, N . The stress or mechanical

factor usually takes the form of a maximum gross or maximum net alternating stress value. In fracture mechanics however, the corresponding factor characterizing fatigue crack growth is the stress intensity factor K , which is a measure of the stress condition in the neighborhood of a crack tip.

Comparison of subcritical growth rates under cyclic stressing show clearly that the stress intensity factor K is the rate controlling parameter in fatigue crack growth. This is shown in Figure 11⁽²⁶⁾ for two different loading configurations using a single aluminum alloy 7075-T6, where crack growth rate $d(2a/dN)$ is plotted against the range of the stress intensity factor ΔK or $(K_{\max} - K_{\min})$. The results suggest that the fatigue crack growth rate is a single valued function, within experimental scatter, of the range of the stress intensity factor encountered in cyclic loading and is independent of specimen geometry and loading condition. With end-loaded specimens under constant load, both the stress intensity factor [Equation (15a)] and the net section stress increase with increasing crack length. Under wedge force loading (loads applied near crack surface) however, an increase in crack length decreases the stress intensity factor [Equation (15)] while the net section stress increases

$$K_{I(\text{end load})} = \sigma \sqrt{\pi a} \quad (15a)$$

$$K_{I(\text{wedge force load})} = \frac{P}{\pi t \sqrt{\pi a}} \quad (15b)$$

where

I = stands for an opening mode crack

a = half crack length

t = thickness

σ = gross or remote stress

P = load

FIGURE 11. *Correlation of Crack Growth Rates in 7075-T6 Aluminum Alloy for Two Different Configurations Under Sinusoidal Loading(26)*

The results from the two different loading configurations (Figure 11), thus allow a clear separation of the roles of net section stress and the stress intensity factor. In addition it should be noted that the same K-versus-rate relationship should hold for complex geometry multiloading structures providing that the loadings are planar and thus induce only an opening mode crack. For cracks that are short with respect to some piping or component dimension, the opening mode should predominate. For longer cracks, say where the crack length approaches a pipe dimension such as a diameter, the tearing or out-of-plane cracking mode may become significant. Thus for relatively short cracks this means that different test specimens and flaw geometries and load configurations should yield identical results and those results should in turn be directly applicable to the primary piping structural components such as curved or tapered pipe, tees, wyes, complex joints, etc.

Variables Which Affect the Rate of Fatigue Crack Growth

Many other factors in addition to the range of the stress intensity factor, ΔK , can affect crack growth rate. Mean load level, loading frequency, specimen thickness, test temperature, heat treatment or material condition, variations in alloy content and environment may cause variation in da/dN and will be outlined below. In addition estimates of the effects of these factors on da/dN in austenitic stainless steels will also be presented.

Effect of Mean Load and Crack Tip Plasticity

In contrast to the effect of mean load on fatigue properties, the mean load level in fatigue crack growth testing appears to play a lesser or secondary role and is related to the nature of the deformation processes occurring at a crack tip.

First the ability to correlate data on crack growth rates (Figure 11) is not net-section stress limited providing the remaining or net section is not undergoing cyclic plastic strain. Although the linear elastic-stress intensity analysis is deficient once the net section is plastic-cycled, some investigators here had limited success empirically by using a strain intensity factor in this range.⁽²⁸⁾ Because of the extreme stress concentration at a crack tip, plasticity is always present in an elastically-loaded cracked specimen. Thus, upon unloading to zero, it is expected that the yielded material or plastic enclave near the crack tip will be forced back to its original position and thus yield in compression. Under these conditions the compressed plastic zone is expected to assume a size about $1/4^*$ the previous tension size because the material experiences a stress change in the order of $2\sigma_{ys}$. Equations developed by Rice⁽²⁹⁾ indicate that the reversed

* Plastic zone size is inversely proportional to $C\sigma^2$

plastic zone and change in plastic deformation depend only on the decrease in load and is independent of the original load level. Furthermore, during cyclic loading, his model also predicts a cyclic reversed plastic zone dependent only on the amplitude of load fluctuation and independent of mean load. For 7075-T6 the rate of crack growth under fully reversed loading does not vary significantly compared to pulsating tensile loading. Therefore, it is anticipated that under nonreversed external loading, the mean load will be of secondary influence. Crack growth rate data (Figure 12) for 2024-T3⁽³¹⁾ aluminum alloy indicate the mean load to be of secondary influence for various related mean load levels γ , where

$$\gamma = \frac{K_{\text{mean}}}{\Delta K} = \frac{P_{\text{mean}}}{\Delta P}$$

A limited amount of crack growth data for 304 SS and related steels are shown in Figure 13 where the GEAP and BNWL data vary considerably. The BNWL 304 (20% cold rolled) and 304L data

FIGURE 12. Mean Load and Fatigue Crack Growth Rates in 2024-T3 Aluminum Alloy

FIGURE 13. ΔK Versus da/dN for Various Stainless Steels

have similar slopes; whereas, the GEAP 304 data display a much steeper slope. It is not obvious why the variation is so significant. The variation in da/dN , for equal K -levels, is as great as a factor of 6 between 304 and 304L which vary ostensibly only in carbon content. Too little data is available to see if mean load influences crack growth rates in 304 SS.

Effect of Loading Frequency

Frequency is normally regarded as a secondary variable in fatigue testing. However, where strain rate affects a materials yield stress, it is expected to have some influence on fatigue crack growth rate. There are no data describing the effect of frequency on crack growth properties of 304. The effect of frequency on da/dN for 2024-T3 is shown in Figure 14.⁽³¹⁾ For

FIGURE 14. Frequency and Fatigue Crack Growth Rates in 2024-T3 Aluminum Alloy. (31)

a given value of ΔK , da/dN increases with decreasing cycle rate and varies by a factor of about 2 to 3 between 20 and 12,000 cpm. Similar data on 7075-T6 aluminum alloy show no significant frequency effect. This has been attributed to the lesser strain rate sensitivity of 7075-T6. Little frequency-effect data are available on other materials, particularly 304. This lack of data is apparently caused by most investigators using certain test frequency ranges. Very low and very high frequency testing requires long test times and, often, specially made equipment, respectively, and both can be costly. Thus, for the purpose of fatigue crack-life estimation, da/dN is assumed to vary by a factor of 2 to 3 for 304 SS also.

It should be noted that frequency of loading and environmental effects in combination may have strong influence on da/dN . If it is generally true that environmental crack extension primarily occurs near the maximum cyclic stress intensity, then the time at maximum load could have a very substantial effect in da/dN or more generally da/dt . No estimates of da/dN (or da/dt) are possible since the effect of environment (viz, N_a), has not been characterized. Further discussion of this potential problem is deferred until later.

Thickness Effect and Variation in Chemical Composition

It is possible that either thickness effect or chemical composition variation, or both, will have an effect on da/dN . Both are lumped together simply because the variation in one property may interact or affect the others, thus each case is difficult to assess by itself. An analogous problem occurs in creep since both grain size and heat treatment can influence creep, but variations in heat treatment are often used to

influence grain size. In addition, variations in chemical composition can also affect grain size. Thus, it becomes difficult to single out the effects of changes caused by a single variable when others change simultaneously. With respect to the primary piping system, variation in thickness, and chemical composition resulting from weldments can be expected to occur.

Variation in Thickness

It is now well-known that the stress condition at the leading edge of a crack can have a significant effect on monotonic-loading fracture-toughness values. Specimens which are thick with respect to the critical or fracture plastic zone size develop a high degree of constraint at the leading edge of the crack. Under these conditions, once the load reaches a critical level, the specimen fails in a catastrophic manner, and the fracture path becomes primarily flat (as opposed to a shear failure). Also, the fracture toughness value for that material in a given metallurgical condition and temperature is referred to as the plane strain fracture toughness and is designated K_{IC} . The same material, however, would show a more ductile or shear type fracture if tested in thinner sections and thus result in a higher fracture toughness value. At a certain thickness, the fracture would be 100% shear, with the higher toughness value referred to as the plane stress fracture toughness (K_{CR}).

The effect of thickness on fatigue crack growth, however, is not entirely clear. Some investigators have found no significant effect of stress condition (i.e. plane stress or plane strain) in fatigue crack growth. Swanson⁽³²⁾ found a small decrease in growth rate in the plane stress-plane strain transition region; whereas, Schwenk⁽³³⁾ found a variation of da/dN with thickness in 7178-T6 aluminum (Figure 15).

Five thicknesses varying from 0.58 to 0.093 in. were obtained by milling equal amounts from both sides of 0.58 in. thick plate. The value of da/dN at $K = 14,000$ varied from 3.2×10^{-5} to 2.5×10^{-5} in./cycle. The variation was not consistent with thickness because the 0.093, 0.188 and 0.58 in. thickness specimens grew at about the same faster rate while the intermediate thickness 0.250 and 0.375 in. grew essentially at the same slower rate. Although these differences might be caused by a thickness effect, they could also be influenced by chemical and mechanical deformation properties that vary through the plate thickness. It is even possible that a thickness dependent state of residual stress could be influencing the rate. Unfortunately, no known tests have been conducted to assess potential residual stress effects.

Variation in Chemical Composition. The effect of variation in alloy composition on the fracture toughness of some alloy systems can be significant. Because environment may sometimes play a significant role, it is not entirely clear how the same variations affect crack growth rate.

Large variations in fracture toughness of high strength 7000 series-type aluminum alloys (as much as a factor of 3) of nearly equivalent strength have been related to variations in alloy composition, notably within the range of allowable alloy specifications⁽³⁴⁾ (variations in the order of 0.1 to 2%); even larger variations were found when certain elements (namely, Zn and Fe) were extended beyond their normal limits. Such large variations in K_{cr} are probably related to the morphology of complex precipitates and their behavior as obstacles to dislocation motion. However, such severe changes in K_{cr} are probably rare, but they do indicate that small alloy composition variations can have a severe embrittling

FIGURE 15. Crack Length Versus Cycles 7178-T6 Aluminum Alloy

effect. However, it is also possible that wide alloy variations in some metal systems could also have a beneficial effect.

The relatively high ductility and low strength of 304 indicate a high potential for subsequent hardening without a substantial loss in toughness. One can only speculate that slight changes in alloy composition would not have a severe embrittling effect on as-received unexposed 304. However, environmental exposure (at temperature and in sodium) could have a severe effect and is discussed further in subsequent sections on the effect of time, and temperature and environment.

Fatigue crack-growth tests on high strength aluminum alloys, similar to those noted above, indicate that variations in chemical composition do not have a significant effect on da/dN in the absence of environmental effects.⁽³⁵⁾ Over a limited range of crack growth rates in dry air, da/dN did not vary substantially but when tested in humid air, crack-growth rates were accelerated by as much as factor of 10. With all other things ostensibly being equal, this latter effect may partially explain the apparent discrepancies in K versus da/dN data that sometime exist between different investigators. Thus, crack propagation laws for specific materials proposed by Head,⁽³⁶⁾ McEvily and Johnston,⁽³⁷⁾ Frost,⁽³⁸⁾ Liu,⁽³⁹⁾ etc., should be re-examined in the light of potential environmental effects hitherto considered to be innocuous.

Recent fatigue crack-growth tests of plane carbon steel specimens indicate a possible chemical composition effect on da/dN .⁽²²⁾ Fatigue crack growth tests were conducted on A-212B plate materials and A-106 Sch 80 pipe material. The latter designation is an ASTM standard for seamless carbon-steel pipe while the former is for carbon silicon steel plate.

Each have identical alloy composition specifications. Figure 16 shows the da/dN for the two different specimen shapes varies by as much as 15 to 30 (e.g., 33×10^{-6} in./cycle for A-106 to 500×10^{-6} in./cycle for A-212B both at an apparently equal ΔK level of 50 ksi $\sqrt{\text{in.}}$). The investigators stated that, "the difference in characteristics of the two materials may be caused either by differences in material properties resulting from compositional and microstructural differences or possibly from the different specimen configuration employed." Providing that all loadings are axial and symmetric and that the proper pipe specimen K-solution was used, the configuration difference should have only a secondary effect. They did find that the mechanical properties did differ as shown in Table 2.

TABLE 2. Tensile Properties of Plain Carbon Steel A-106 Pipe and A-212B Plate Specimens

Property	A-106 Pipe	A-212B Plate
Ult. Tens. Str., ksi	62.5	75.9
0.02% Yield Str., ksi	31.2	48.8
0.2% Yield Str., ksi	36.3	45*
% Elongation (2 in. gage length)	33	30.5

**This value may be incorrect but no indication was made to a possible yield drop that might have occurred at the indicated strain level.*

In subsequent crack growth test on 2-1/4 Cr-1 Mo and 304 plate and pipe specimens they found essentially the same systematic behavior.⁽²²⁾ This suggests that their K-solution for the pipe is incorrect, rather than a chemical effect. However, since the intervening document (10th Quar. Rep. GEAP 5554) is unavailable for reference to a chemical analysis on the above materials (Table 2), it is not possible to exclude chemical composition variations as having no effect on da/dN .

FIGURE 16. Crack Growth Rates in A-106 Schedule 80 Prototype Pipe Test Material at Room Temperature

The effects of thickness variation and chemical composition variation on fracture toughness and fatigue crack growth by themselves are not clear. Differences in mechanical forming processes and environmental effects may, in some cases, play a dominant role. Ultimately, however, such variations in cracking properties must be rationalized in terms of the size and distribution of obstacles to dislocation motion. Unfortunately, not enough data exist that relate the latter to such things as thickness and alloy variations thus the effect of similar variations in stainless steels is pure speculation. Although no general relationship exists between fatigue crack growth and fracture toughness, they both involve large strains at a crack tip and therefore should depend on the total plastic behavior of the material.

Effect of Mechanical Properties

Significant improvement of the resistance to fatigue crack growth by alteration of mechanical properties do not appear likely. Furthermore, it does not presently appear feasible to predict the general effect of mechanical property changes on da/dN .

The result of an analysis of crack growth in an elastic-plastic solid [Equation (16)] was developed from a continuum dislocation model of plastic yielding at a notch.⁽⁴⁰⁾ The rate of growth given by this analysis based on a nonstrain hardening material is

$$\frac{da}{dN} \propto \frac{(\sigma \sqrt{a})^4}{\gamma G \sigma_y} \quad (16)$$

where

γ = surface work term

G = shear modules

σ_y = yield stress.

This has been empirically modified to incorporate the effect of strain hardening⁽³⁷⁾ as shown in Equation (17):

$$\frac{da}{dN} \propto \frac{(\sigma \sqrt{a})^4}{\left[\left(\frac{\sigma_y + \sigma_u}{2} \right)^2 \right]^E \sigma_u^2 \epsilon_u} \quad (17)$$

where

σ_u = ultimate tensile strength, ksi

E = Youngs modulus

ϵ_u = strain at ultimate tensile strength.

The room temperature data shown in Figure 17 are modified by using Equation (17) as shown in Figure 18. Except for the aluminum alloy 7075-76, which may be unstable because of an interaction between dislocations cutting G. P. zones, the results fall into a narrow band. Since designers either increase design loads or decrease structural dimensions in proportion to increases in mechanical property levels, it appears that significant improvements in the resistance to crack growth by alteration of mechanical properties are not likely. However, exposure-induced changes in mechanical properties of a given component will probably alter its crack growth properties. If Equation (17) were valid, da/dN could be either increased or decreased, depending on the nature of the mechanical property changes. Too little information exists to do anything but speculate the effects of such changes particularly for stainless steels.

Effect of Low Stress Intensity Values

The phenomenon of endurance limit in smooth specimen fatigue suggests the possibility of a corollary effect with fatigue crack growth in notched specimens. However, not enough testing has been done to either confirm or deny the existence of a crack-growth endurance limit.

FIGURE 17. Rate of Fatigue Crack Growth for a Number of Alloys⁽³⁷⁾

FIGURE 18. Normalized Fatigue Crack Growth Rates for Alloys Shown in Figure 17.⁽³⁷⁾

Early investigators of fatigue crack growth^(25,26,41) often obtained growth rates in the order of 10^{-7} in./cycle for several materials. Lower rates were not usually obtained for two reasons: excessive testing time on ordinary fatigue equipment or crack growth delays brought about by initiating cracking at high loads and subsequently running the tests at lower loads. Kinse⁽⁴⁵⁾ reported that a 7075-76 aluminum notched specimen initially cracked at a ΔK level of about $3.5 \text{ ksi } \sqrt{\text{in.}}$ and subsequently lowered to about $1.8 \text{ ksi } \sqrt{\text{in.}}$ did not grow up to the time the test was halted after 80 million load cycles.

Recently, average crack-growth rates as low as about 10^{-10} in./cycle (Figure 19) were observed in 7075-76 aluminum by using a high cyclic rate test machine.⁽⁴⁴⁾ Thus, for

FIGURE 19. Fatigue Crack Growth Rates in 7075-T6 Aluminum Alloy at very Low Cyclic Stress Intensities.⁽⁴⁴⁾

7075-76, there appears to be no threshold K-level corresponding to a so-called endurance limit. This may be related to the lack of an apparent endurance limit of 7075-76 in smooth specimen fatigue testing. Typical S-N curves for aluminum alloys have the appearance shown in Figure 20 where with decreasing stress, the curve continues to fall steadily but at a decreasing rate. In contrast certain materials such as iron, which contain carbon as an interstitial element, usually display a fatigue limit which is thought to be caused by strain aging during cyclic loading.⁽⁴²⁾ It is important to note that the data in Figure 19 indicate that for a K-level less than 5% of the plane strain fracture toughness (K_{IC}), fatigue crack growth still exists. Type 304 austenitic stainless steel appears to display an endurance limit at elevated temperatures but not at room temperature.⁽⁴³⁾ However, since no corresponding cracking data exist for 304 SS it is assumed that no threshold limit exists. The possible lack of a crack endurance limit in 304 may not be significant. Should, however, high cyclic rate loading occur from high frequency mechanical vibration and/or rapid variations in the mixing of hot and warm sodium (thermal fatigue), fatigue and fatigue crack growth are possible modes of piping failure.

Effect of Temperature and Time

The effect of time and temperature are often intimately related and not easily separable. In addition, creep effects and changes in mechanical properties with exposure make it difficult to predict their effect on da/dN .

Temperature. McEvily and Johnston⁽³⁷⁾ claim that Equation (17) will provide a reasonable estimate of da/dN as a function of temperature so long as the alloys remain stable and the appropriate stress-strain and modulus values are available. This may be so but little elevated temperature data

FIGURE 20. *Fatigue Strength of Rotating Beams, 24S-T Alloy, 0.300-in. diam, Superfinish. (46)*

are available to either refute or support their argument. Recent fatigue crack-growth tests on 304 at room temperature and 550 °F⁽⁴⁷⁾ indicate essentially no effect between the two noted temperature levels. If the equation is valid, da/dN at 500 °F should be about 4 times as great as the room temperature value and 30 times as great at 1200 °F. No data are available to either refute or support such a large temperature effect. The corollary effect in reversed-bending fatigue tests of high temperature alloys is that the fatigue strength (for a given number of load cycles) usually decreases with increasing temperature.⁽⁴⁸⁾ It is possible that strain rate or cyclic rate at high temperatures will have a pronounced effect on da/dN . The interaction of creep or time-dependent deformation and slip-type, or low-temperature cyclic deformation is not clear but is expected to be related to the cyclic rate at a given temperature.

Time at Temperature. Thermal aging can increase and decrease the notch sensitivity of stainless steels as well as alter their mechanical properties. Austenitic stainless steel creep tested in the temperature range, 1100 to 1300 °F, undergo notch strengthening.^(49,50)

Specifically, AISI 304 notch strengthens at 1500 °F but weakens at 1100 °F (Figure 21). The corresponding tensile properties, however, change during the course of such tests. After exposure at 1200 °F and up to 13,000 hr only slight changes occurred in yield stress whereas ultimate stress and % elongation values changed as much as +14% and -33%, respectively (Figure 22).⁽⁵⁹⁾ No impact test data at elevated temperature were found, but room temperature charpy tests on 304 SS exposed 3000 hr at 1000, 1300, and 1500 °F showed a consistent fracture energy decrease with increasing temperature to a level of about 1/2 (for 1500 °F exposure samples) the room temperature unexposed sample.⁽⁶⁰⁾ Such behavior may be related to the size, shape and number of precipitate particles forming in grain boundaries as discussed by Garofalo.⁽⁵¹⁾ However, such variations in strength, notch rupture strength ratio and fracture energy suggest that variation in da/dN might occur too.

Effect of Intermittent Loading and Intermittent Temperature on Creep Crack Growth

It should also be noted that intermittent load and temperature application during creep testing of 321 altered the creep properties. Since creep "fracture" occurs through the coalescence and growth of voids, cyclic loading and cyclic temperature testing affects their "rate of growth."

Intermittent Loading Effects. The results for both constant load and intermittent load tests at 1200 °F are shown in Figure 23. Two sets of curves are included: the upper curve

FIGURE 21. Notch Rupture Strength Ratio Versus Time AISI 304 at 1100 °F (49)

FIGURE 22. Change in Mechanical Properties with Time at 1200 °F
304 Stainless Steel (59)

FIGURE 23. Stress Versus Elapsed Time at Load for Constant Load and Intermittant Load Creep Tests on AISI 321(50) 1200 °F

is a plot of the stress-time values corresponding to elongation of 5%; the bottom set show the same thing corresponding to total rupture (parenthesized numbers indicate rupture elongations). In both sets of curves, intermittent loading produced the same creep rates at stress levels varying from 5 to 15% less than the constant load stress levels, (or an increase in average creep rate of about 200% - determined by dividing % elongation by time to achieve that elongation). In addition the rupture elongation values were slightly reduced by the intermittent loading.

Intermittent Temperature Effects. The results for intermittent temperature application as well as constant temperature (and constant load) are shown in Figure 24. Intermittent temperature like intermittent loading increased creep rates at 1200 °F but more so at 1350 °F (not shown). In contrast to intermittent loading, intermittent temperature application decreased rupture elongation substantially at both 1,200 and 1350 °F. Such temperature cycling produced equivalent constant-load-constant temperature creep rates at 1200 °F at stress levels varying from 8 to 19% less than the former with similar reductions at 1350 °F (or an increase in creep rate of about 200%, based on percent elongation). Rupture elongation at 1200 °F decreased by as much as a factor of about three (e.g. 13 to 4%).

In addition (1200 and 1350 °F), failure in both static and intermittently loaded specimens was preceded by the formation of intergranular cracks progressing from the surface toward the interior. Such cracks could obviously be propagated more rapidly by cyclic loads and thus leading to reduced rupture ductilities and accelerated fracture.

No attempt will be made to link the combined effect of creep and fatigue together as it is beyond the scope of this paper. It is obvious, however that "fatigue loading" during

FIGURE 24. *Stress Versus Elapsed Time at Temperature for Constant Load and Intermittant Temperature Creep Tests on AISI 321(50) 1200 °F*

creep testing of 321 had a deleterious effect on creep rupture properties but it is not clear if the reverse is necessarily true. However, creep effects cannot be separated from fatigue crack-growth effects because the higher primary piping temperatures are in the creep range of 304 (e.g. 1200 °F).

Summary of the Variables That Affect Fatigue Crack Growth Rate

In the preceding sections, factors affecting fatigue crack-growth rate were discussed. The dominant crack-growth rate controlling variable is the range of the stress intensity factor, ΔK , and is manifested in alternating plasticity at a crack tip. To a first approximation it is proportional to the range of the alternating stress, $\Delta\sigma$. K is the local elastic stress intensity "surrounding" the crack tip plasticity. Characterization of fatigue crack-growth rate in terms of K is realistic as it has been used with success over a wide range of crack growth-rate values (e.g., up to 8 orders of magnitude, 10^{-10} in./cycle to 10^{-2} in./cycle in 7075-76) for many different alloy systems.

The effect of mean-load and thickness have a minor influence on da/dN . The effect of frequency is secondary but could be significant at higher temperatures and in aggressive environments. Nominal variations in material chemical composition can affect da/dN in some cases, but environment appears to have an overriding influence. K versus da/dN room temperature data have been normalized in terms of mechanical properties, but insufficient data are available to speculate on elevated temperature behavior. Lack of data does not allow correlation of a fatigue crack growth endurance limit to a corresponding fatigue endurance limit. Time-temperature exposure is expected to affect da/dN and may be reflected by changes in mechanical properties brought about by that exposure. At higher temperatures where creep is involved, time-dependent and cycle-dependent

deformation effects are expected to interact, and strain rate or frequency effects will most likely play a more significant role.

From a practical point of view, the fatigue crack-growth behavior of a given material in a given environment can be best established in terms of a plot of da/dN versus ΔK . Thus, evaluation of the fatigue cracking behavior of primary piping materials must be obtained from specimens tested in anticipated FTR environment(s).

Cracking in Aggressive Environments

Crack-growth rates of materials are increased by exposure to particular environments. Impure sodium has a severe affect on the low-cycle fatigue life of 304 SS. No explicit data are available to quantify the effects of impure sodium on da/dN , but low-cycle fatigue data suggest that it could be significant.

Effect of Environment on Fatigue Properties

Normal air can have an effect on fatigue properties. The fatigue life of copper in dry air is about 1/10 of that in a high vacuum;⁽⁵²⁾ the introduction of moisture further reduces the fatigue life by another 1/3. Cold-worked 18-8 austenitic stainless steel displayed an endurance limit of about 50,000 psi in both air and distilled water but decreased to 25,000 psi in salt water.⁽⁵³⁾

Investigators have generally agreed that the main effect of a normal atmosphere on fatigue is on the propagation of cracks rather than in their initiation. The principal effect of environment will be in its reaction to the clean metal exposed at the tip of a fatigue crack. Crack initiation is not affected because the initially present oxide films are not significantly altered by normal environments. In

environments that are more corrosive than air and that substantially reduce the resistance of the surface layers to fatigue crack initiation, both fatigue crack initiation and propagation may be affected.

A great deal of effort has been expended in examining the properties of reactor materials as they relate to specific disciplines such as mechanical properties, irradiation effects, sodium corrosion effects, physics, chemistry, etc. However from a structural engineering viewpoint, no sustained or cyclic-load tests of notched specimen in simulated FTR environments have been attempted. It is possible that 304 SS may be the most crack resistant structural material for sodium reactor environments. However, it should be noted that fatigue crack-growth tests in a salt water environment were responsible for redirecting the alloy selection for the Supersonic Transport Airplane (SST) and the Deep Diving Submersible.⁽⁵⁴⁾

A major part of the alloy selection program for SST titanium alloys involved mechanical testing. Smooth specimens had been fatigue tested in salt water environments to high percentages of their yield strength and were found not to be significantly different than the same materials tested in air. In addition, some fatigue crack-growth tests were conducted but only in ambient air. Only some time after Ti-8Al-1Mo-1V had been essentially "selected" as the candidate alloy did some investigators examine the effects of salt water on notched specimens of titanium. Fatigue crack-growth tests on the same and similar alloys previously unaffected when tested in a smooth condition, displayed increased crack-growth rate (compared to air) by a factor of $10^{(55)}$ to about 25,⁽⁵⁶⁾ depending on the type of crack specimen geometry and mode of loading. Also, sustained load tests on wedge-force loaded specimens indicated the presence of environmental static-load cracking phenomenon that was thickness dependent. Essentially, cracks initiated

and grew at lower K-levels in thick specimens than in thin specimens.⁽⁵⁷⁾ In addition, the toughness of the alloys were unaffected by the environment; the original fatigue cracks extended under sustained loads until they achieved a critical crack length associated with the "environment"-free K_{CR} of the particular alloys.

Fractographic studies showed that the increased crack-growth rates were associated with a change from a ductile mode of failure to a brittle quasi-clearage mode normally associated with stress-corrosion cracking. No comparable studies have been conducted on 304 SS or related alloys in either impure or pure sodium to examine the possibility of a similar cracking behavior. However, recent tests on the low-cycle fatigue behavior of 304 in impure sodium indicate a potential environmental cracking problem.⁽⁵⁷⁾ Briefly, reversed-bend low-cycle fatigue tests were conducted in 1200 °F helium, high-oxygen sodium (Hi O₂ Na), air and high-carbon sodium (Hi C Na), respectively. The data shown in Figure 25 have been modified in comparison to those shown in Reference 57. (The reason for this will be discussed later.) The data plot in Figure 25 indicates that fatigue life decreases essentially in the order of exposure from helium to air to impure sodium. Unfortunately, changes in the sodium environment occurred during testing, which make it difficult to compare the effects of Hi C Na only and Hi O₂ Na only to air and helium. However, the environment excursions do shed some light on the possible severity of impurity fluctuations.

First, the data indicate that the combination of Hi O₂ Na and carbon pickup decreased the high-strain fatigue life by a factor of 5 with respect to helium. This indicates that fatigue crack-growth rates would be increased by the environment because Stage II crack growth (i.e. where crack growth is normal to σ_1) generally occupies more than 75% of the

FIGURE 25. % Cyclic Strain Versus Strain Cycles AISI 304 SS at 1200 °F

total fatigue lifetime⁽²⁸⁾ of low-cycle smooth fatigue specimens. This also suggests that in a notched, elastically loaded specimen, the plastic enclave and "fracture-level" strain associated with a large, growing crack may be more sensitive to the environment than small cracks in "uniformly" plastic strained low-cycle fatigue specimens. A notched specimen then could provide a better means of measuring the effect of environment on fatigue crack growth; it would also provide a better means of rating or comparing the "crack growth sensitivity" of various reactor structural alloys.

As noted earlier, the curves shown in Figure 25 were modified from those shown in Reference 57. During fatigue testing in $\text{Hi O}_2 \text{ Na}$ an unscheduled addition of carbon had a pronounced effect on the number of cycles to failure. The investigators determined the amount of carbon increase of the fracture face for each specimen and "corrected" the resulting cycles-to-failure to reflect the effect of $\text{Hi O}_2 \text{ Na}$ in the absence of carbon. They also noted that the high-strain short-life specimen fracture faces suffered a larger carbon increase than the faces of lower-strain long-life specimens (e.g., about 1,000 ppm C @ $\pm 1.9\%$ strain as opposed to about 300 ppm C @ $\pm 0.5\%$ strain). This indicates that the crack surface associated with a highly strained specimen is more effective at picking up carbon than that with lower applied strain specimens, even though the latter were exposed to the $\text{Hi O}_2 \text{ Na}$ for much longer times. Therefore, the "correction" of the data to reflect a zero % increase in carbon is probably not realistic because of the carbon absorption process taking place at crack tip. The data also suggest that the effect of lower impurity levels of carbon and oxygen in sodium could be intensified by the large volume of crack-tip plasticity in a notched specimen that otherwise might be less sensitive in a smooth fatigue specimen. This again indicates that a prenotched specimen would be more useful than smooth specimens for evaluating the crack growth sensitivity of reactor alloys.

In contrast to the nil effect of salt water on the mechanical and fracture toughness properties of titanium alloys, exposure to impure sodium at temperature can alter the mechanical properties of 304 SS.⁽⁵⁷⁾

In general, high oxygen sodium:

- Reduces creep rupture life
- Increases creep rates
- Severely decreases rupture ductility and impact energy.

While high carbon sodium:

- Increases strength levels
- Increases rupture life
- Decreases creep rates
- Decreases creep rupture ductility and impact energy.

Most of the property changes noted above are not beneficial from a structural viewpoint, but their change in magnitude do suggest that changes in da/dN can be expected. Figure 26 shows the change in impact energy of 304 exposed to helium, $Hi\ C\ Na$ and $Hi\ O_2\ Na$, respectively. The curves are a measure of the decrease in room temperature fracture toughness brought about by the exposure. The fact that the specimen notches were machined only and not subsequently fatigue-cracked indicates that the impact energy values are not conservative. The impurity levels referred to are obviously high and are not expected to exist in the FTR sodium pool. Generally, the data do indicate that the plasticity associated with a large growing crack is probably more sensitive to sodium impurity level than heretofore realized.

Estimation of Fatigue Crack Growth Behavior of 304

The accurate prediction of the service lifetime of a complex structure from simple tests is still not a reality owing to the fact that reliable analytical methods have not been developed which include the effects of variable amplitude (or

FIGURE 26. Impact Energy Versus Exposure Time AISI 304

random-type) loading on fatigue. Nevertheless, testing of simple specimens under simulated service conditions remains as the most realistic way in which to base estimates of structural component behavior.

Estimation of the fatigue crack-growth behavior of primary piping components is based on knowledge of the basic fatigue crack-growth behavior of the construction material. As noted earlier, available ΔK versus da/dN data for 304 exist for room temperature air and some limited data at 550 °F.⁽⁴⁷⁾ A very rough estimate of the fatigue cracking behavior might appear as shown in Figure 27. Room temperature data are used as references although it should be noted that the data displayed a large scatter band (Figure 13). It is further assumed that a cycle rate-temperature sodium-impurity relationship exists whose effect is large for low crack-growth rates reducing to a small value for high crack-growth rates. Admittedly, it is crude and could be three to four orders of magnitude in error. However, subsequent calculations in the next section are based on the estimated data for purposes of illustration.

ESTIMATE OF PIPE LIFE UNDER OPERATING CONDITIONS

This section will, in effect, combine the results of the previous two sections, and an attempt will be made to estimate the crack extension behavior of typical flaws under various operating conditions. It cannot be emphasized enough that the crack extension curves shown in this section are based on estimated material properties that have absolutely no experimental verification, and, therefore, errors of several orders of magnitude are not inconceivable. (The only exception to this statement is the low cyclic rate da/dN versus ΔK data for 70 °F, and even in this case it was necessary to extrapolate experimental data into the region of interest).

FIGURE 27. Estimated ΔK Versus da/dN Curves for 300 SS

The assumed crack-growth rate behavior used in these estimations is shown in Figure 27. This behavior, in general, follows the "power law of crack propagation" as expressed by Equation (10). Examples of axial and circumferential cracks are worked for the FTR discharge piping. Examples were not worked for the suction piping because the applied pressures and loads were much smaller, and the suction piping is considerably larger. Therefore, the discharge piping represents a much more severe condition. The assumed sizes and loadings used in this study are shown below:

Outside diameter	= 12.00 in.
Wall thickness	= 0.375 in.
Internal pressure	= 200 psig maximum
Applied bending moment	= 100,000 in-lb maximum

These represent the expected sizes and loadings that were current when these calculations were made. Although they have probably changed since that time, they will doubtlessly change again. Unless the pipe size and/or loadings change drastically in the future, the estimated crack length versus cycles curves presented in this section will probably remain a reasonable indicator of estimated behavior. The rationale behind this statement is that the estimated crack-growth rate data from the previous section may have errors of several orders of magnitude (in either direction) and, therefore, small changes in pipe size and/or loadings should have no effect on the curves shown in Figures 28 and 29.

Figures 28 and 29 give the estimated crack length versus cycles behavior of the discharge piping for circumferential cracks and axial cracks, respectively. The curves show the number of cycles that are estimated to grow a through-wall crack from an initial length of $2a = 0.50$ in. to a length of $2a = 2.00$ in. These curves were stopped at a crack length of 2 in. because the leakage from such a crack probably would

FIGURE 28. Estimate of Crack Extension Behavior of a Circumferential Crack in FTR Discharge Pipe Under Internal Pressurization and Applied Bending Moment

FIGURE 29. Estimate of Crack Extension Behavior of an Axial Crack in FTR Discharge Pipe Under Internal Pressurization

have been detectable. It should be pointed out that in some of these cases, the crack is propagating quite rapidly as it approaches a length of 2 in. For example, in Figure 28, for the case of sodium environment at 1200 °F (low cyclic rate), it is estimated that 1000 cycles were required to grow the crack from 0.50 to 0.60 in., while only 132 cycles were required to grow the crack from 1.90 to 2.00 in. Thus, it may be seen that if the crack-growth rate estimates are reasonably correct for this condition, then cracks that grow much beyond a total length of 2 in. could propagate quite rapidly. Therefore, it is very important that cracks be detected before they exceed 2 in. in length. "Failure" of the pipe under conditions of high crack-growth rate would probably be a "ductile tearing" mode as opposed to unstable "brittle" crack growth. For this reason, consideration of K_{IC} values (fracture toughness) at these temperatures is meaningless.

It was pointed out in the section on application of fracture mechanics techniques that it may not be possible to apply the Catanach solution⁽¹⁷⁾ to an axial crack in the FTR discharge pipe because of the relatively low ratio of OD to thickness ($OD/t = 32$). In spite of this possible limitation, the Catanach solution was used to obtain the crack extension behavior (Figure 29) because no other solution was available.

The preceeding discussion has shown that very little is known of the crack growth behavior of 304 SS under the FTR operating conditions and, until more data becomes available, any predictions of the expected cracking behavior must remain quite qualified.

NOTATIONS

a	half crack length, inch
β	one-half the angle subtended by a circumferential crack, radians
da/dN	crack growth rate, inch/cycle
ΔK	stress intensity factor range, $K_{\max} - K_{\min}$, lb/in. ^{3/2}
E	modulus of elasticity, lb/in. ²
G	strain energy release rate, also called crack-extension force, (in this report assumed to be opening mode G_I unless otherwise noted) in-lb/in. ²
ν	Poisson's Ratio, dimensionless
K	stress intensity factor, also called stress-field parameter (in this report assumed to be opening mode K_I , unless otherwise noted), lb/in. ^{3/2}
K_{Ic}	critical fracture toughness in plane strain, lb/in. ^{3/2}
ρ	radius to mid-surface of cylindrical shell, inch
σ	gross section tensile stress, lb/in. ²
σ_h	hoop stress, lb/in. ²
t	thickness of tube wall, inch

APPENDIX

REVIEW OF PIPE RUPTURE FAILURE STUDY

APPENDIX
REVIEW OF PIPE RUPTURE FAILURE STUDY

DATA AND INFORMATION COLLECTED

The survey provided general failure information (Table A-1), but essentially no data were taken that could be applied to fracture mechanics analysis techniques. Basic information such as crack length, geometry or configuration, and type of fracture were usually not recorded. Personnel responsible for the plain maintenance were neither stress analysts or metallurgists. Thus, their evaluation of failures often did not contain pertinent information such as precise dimensions, material history and loadings. Immediate repairs to minimize shutdown time (and an unawareness of the future usefulness of such data) often precluded extensive failure analyses. Consequently, the survey data must be treated with care and providence.

FAILURE CAUSES

Pressure piping system failure causes were categorized as shown in Table A-1.⁽¹⁾ They are:

- Design - Relates to failures caused by design inadequacy such as fatigue, expansion, restraint, notches, joint design, etc.
- Material Selection - Caused by material application (or misapplication) or metallurgical design such as wrong specification, graphization, stress rupture, etc.
- Manufacturing - Caused by manufacturing of base materials, and/or base metal defects, wrong material, material not to specification, etc.
- Fabrication-Erection - Caused by events during fabrication and/or erection such as, surface defects, welding defects, heat treatment, cleaning, and incorrect installation.

TABLE A-1. Failure Data Summary - Number of Failures

(1) Primary Cause of Failure	(2) Up to 600 °F	(3) 600 to 1200 °F	(4) Total-All Temp.	(5) Percent of Total	(6) Occurred Without Leakage	(7) Complete Severance Cases	(8) "Thermo- well"	Location of Failure					
								(9) Nozzle	(10) Fitting or Com- ponent(a)	(11) Pipe	(12) Cast- ing	(13) Weld	
Design													
Mech. fatigue	8	10	18	4.5	2	6	7	2	3	6			
Expansion/flexibility	5	5	10	2.5		1				7			3
Section transition	1	1	2	0.5						1			1
Support		2	2	0.5						2			
Design-operation	4	28	32	8.0	3	1		1	1	28			2
Thermal-fatigue	5	4	9	2.3			1		1	3		1	3
Miscellaneous	23	50	73	18.3	5	8	8	3	5	47	1	1	9
Subtotal													
Material Selection													
Misapplication		9	9	2.3		1			3	1			5
Metallurgical													
Type 347 SS		59	59	14.8	(b)					59(c)			
Graphitization		41	41	10.3	11(b)	1			1	38(d)	1		1
"Stress-rupture"		5	5	1.2	1					3			2
"885" embrittlement		2	2	0.5		1		1		1			
Miscellaneous	1	3	4	1.0						4			
Subtotal	1	119	120	30.1	13+	3		1	4	105	1		8
Manufacturing													
Wrong material supplied		5	5	1.2									
Base Material defects	17	14	31	7.8	1	1			1	3			1
Weld defects		1	1	0.3					2	16	13		
Associated with welding	1	3	4	1.0						1	3		1
Materials not to spec.	1	1	2	0.3						1			
Subtotal	18	24	42	10.6	1	1			3	21	16		2
Fabrication-Erection													
Weld defects	19	8	27	6.8	1	1							27
Associated with welding	4	6	10	2.5				2	1	4	1		2
Wrong materials used		4	4	1.0		2				2			1
Miscellaneous	2	7	9	2.3	1	3	1		1	7	1		1
Subtotal	25	25	50	12.6	2	3	1	2	2	13	1		31
Operation-Service													
Misoperation	4	5	9	2.3	1	2			4	4			1
Thermal shock and shock	1	5	6	1.5	1	1			2	2	2		
Corrosion and/or erosion	82	12	94	23.4	2	3			5	80	7		2
Subtotal	87	22	109	27.2	4	3			11	86	9		3
Unknown	1	4	5	1.2		1		1		3			1
	155	244	399	100.0	25+	19	9	7	25	276	28		54

(a) Elbow, tee, wye, etc.

(b) Some of these occurred without leakage; precise data was not supplied.

(c) Most of these occurred in the weld heat-affected zone, some were in components elbows; specific data distribution was not supplied.

(d) Most of these were located in the weld heat-affected zone.

- Operation - Service - Caused by conditions that were excessively severe and not readily foreseen, such as shock freezing, corrosion, etc.
- Unknown - Sufficient documentation but no specific conclusion as to cause is supported.

The survey results are important in that they relate to the chronological sequence of circumstances that often lead to failure from initial design through the building and operation of pressure piping systems. Essentially, however, all groupings (as defined above) can be related to flaws that were either initially present or generated-in-service and caused to grow by a combination of loading events and environmental effects leading to eventual fracture.

Fracture is the last stage of subcritical crack growth. Thus, the quantitative assessment of system rupture or fracture hinges on knowledge of the fundamental behavior of subcritical crack growth. Linear elastic fracture mechanics techniques have been successfully used in some instances to quantitatively relate subcritical crack growth to the stress intensity factor, K . Fracture-mechanics crack-analysis techniques are discussed in the body of the report but it is important that it be determined where cracking is most likely to occur within the primary piping system, the predominate modes of failure, and their relation to the system mode of operation. Certain portions of Reference 1 provide a basis for such an analysis and are paraphrased and annotated as noted by brackets [].

From Table A-1 some of the more pertinent observations concerning the 399 failures are shown below.

FAILURE OBSERVATIONS

Temperature

Sixty-one percent of all failures (244 out of 399) occurred in systems operating in the 600 to 1200 °F range: the remaining failures (155) occurred below 600 °F.

Materials

Of those failures attributed to materials selection (120 out of 399), only one occurred at temperatures below 600 °F; however, all the remaining "materials selection failures" (119 or 30% of all failures) occurred in the 600 to 1200 °F range.

Failures Without Detected Leakage

The number of failures that occurred without detected leakage was about 6% (25 out of 399). [This probably reflects the inadequacy of routine inspections and that most NDT equipment has limited sensitivity when used on complex shapes. In addition, piping systems are often covered with insulation which makes detection of leaks more difficult when they are small.]

Complete Severance Failures

Almost 5% of the failures (19 out of 399) occurred by complete severance within the system. [From the failure tables compiled in Reference 1, it appears that of the 19 cases noted, four failures were due to overstress (e.g. overdesign loads, too high a temperature or material with too low strength properties) and four cases attributed to low ductility. In these cases either rapid deformation or rapid crack growth are favored. Of the remaining 11 cases, fatigue was described as the predominant mode, but no information was available to estimate fatigue crack-growth rates. Interestingly, in 18 of the 19 cases the failed components were in service prior to complete failure from as low as 60 hr to more than 20 yr; no service time was reported for the 19th case.]

Failure Locations

Columns 8 through 13 (Table A-1 indicate that 85% (337 out of 399) of the failures were located either in the pipe or in pipe welds and weld areas. If we subtract those failures connected with welding in the piping from the 276 pipe failures,

[276-(59+38+1+4*) = 174] about 44% of all the failures occurred in the pipe per se. The next largest failure location as noted in Table A-1 was in the welds.⁽⁵⁴⁾ If we then include the heat-affected zone [Item (b) and (c), Table 1] and the "associated with welding" classifications as belonging to the weld, the number increases from 54 to 163 [(54+59+38+1+3+2+1+4+1* = 163)] or about 41%. Although 85% of all failures fall into these two categories, one should not discount failure in other components. For example, the primary piping system of a typical boiling-water reactor has been classified as having approximately 690 sections of straight or curved pipe, 500 welded joints, 76 tees, and 79 hanger attachments.⁽²⁾ If we assume that probability of failure within all classifications are equal, then the number of failures will be proportional to the number of elements within a given classification. Based on this hypothesis, failures occurring in a typical BWR would be distributed as shown in Table A-2; the distribution of failure locations for the 399 cases cited in Table A-1 are also shown. Interestingly, the data indicated that the failure-location and component-distribution percentages are similar. Such a comparison, however, is not completely valid because the "typical" BWR system is not necessarily the same as the average non-nuclear pressure piping system. The results indicate that failures in a nuclear primary piping system may

TABLE A-2. *Pressure Piping System Failure Location*

System	Pipe	Weld	Tees (miscellaneous components)	Hangers	Castings	Others
Breakdown of components in a "typical" BWR pri- mary pipe system, %	$51 = \left\{ \frac{690}{1345} \times 100 \right\}$	37	6	6		
Failure distribution by percent in compo- nents of a general primary pressure piping system	44	41	6		6	3

occur in proportion to their population of pipes, welds, components, etc., but that failures will be concentrated in pipes and welds primarily because those two categories have the largest population. This indicates, too, that engineering design codes for individual components of primary piping systems are well-distributed. In addition failures in components such as tees, castings, etc., should not be completely discounted just because fewer numbers of failures occur in that classification.

Failure Causes and System Operation

The data in Table A-1 represent a significant effort to establish and categorize the dominant causes of failure in pressure piping systems. However, as the author noted, "it becomes quite apparent during the survey and subsequent follow-up, that failure investigations are often quite cursory in nature."

Similarly, the mode of subcritical cracking remains even more questionable and is demonstrated by comparing two different data tables such as Tables A-3 and A-4 which were reproduced directly from Reference 1. Table A-3 is a compilation of failures based on observations of the type of fracture or failure mode while Table A-4, although prepared from the same data, is a compilation of failures as related to the classification noted in Table A-1. Table A-4 ostensibly differs from A-3 in that the more nondescript and infrequent failure causes (see Table A-1) are apparently eliminated for brevity.

A dominant mode of fracture as indicated by Table A-3 "Classification of Failures by Fracture or Failure mode," is fatigue-high-cycle-elastic and low-cycle-plastic, and accounts for 116 failures. However, Table A-4 indicates mechanical and thermal fatigue as the primary cause of only 50 failures. Therefore, this difference of 66 failures is apparently due to the

TABLE A-3. *Classification of Failures by Fracture or Failure Mode*

Mode ⁽¹⁾	No. Cases	%
Cleavage	7	1.7
Fibrous	7	1.7
Creep Rupture		
General	15	3.8
Graphitization	41	10.2
Type 347 SS	59	14.8
Fatigue		
High Cycle-elastic	20	5.1
Low Cycle-plastic	93	23.3
Not clear in the two above	3	0.7
Corrosion		
Austenitic-general	5	1.2
Transcrystalline	3	0.7
Intercrystalline	2	0.5
Ferritic-General	18	4.6
Transcrystalline	-	-
Intercrystalline	2	0.5
Erosion	82	20.4
No fracture, defect leaked	24	6.2
Insufficient data to attempt classification	<u>18</u>	<u>4.6</u>
Total	399	100.0

method used for failure categorization in Table A-1 and could also result from inconsistencies in the original failure analyses. For example, failures are often stated as being caused by processes where a notch is either manufactured (e.g., voids in material, weld defects, fabrication-erection defects, etc.) or produced in service (e.g. erosion, corrosion, etc.); whereas, the resulting mode of crack growth is not always distinguished and can be caused by one of several mechanisms. By further analyzing the detailed tabulated failure data of Reference 1,

TABLE A-4. Primary Causes of Failure Noted in Table A-1

<u>Primary Cause</u>	<u>No. of Failures</u>	
	<u>600 °F or less</u>	<u>All Temperature</u>
Design		
Mechanical fatigue	8	18
Thermal fatigue	<u>4</u>	<u>32</u>
Total fatigue	12	50
Materials		
Graphitization	0	59
Type 347 SS	<u>0</u>	<u>41</u>
Total creep-rupture	0	100
Manufacturing		
Base metal defects	17	31
Fabrication - Erection		
Weld defects	19	27
Service		
Corrosion-erosion	<u>82</u>	<u>94</u>
	130	312

$$\frac{312}{399} \approx 76\% \text{ of total failure cases}$$

fatigue cracking accounts for eight additional failures and is defined as a secondary cause of failure. Although this still leaves 58 unaccountable fatigue fractures, the author does not reconcile this difference. The reason for the discrepancy is not entirely clear. However, it must be related to both inconsistencies in the original failure analyses and the desire to lay the blame for piping system failure on design-fabrication type procedures that assisted crack formation as opposed to the subsequent operation-induced growth of these cracks.

The most revealing characteristics of all the failures are not found in the attached tables (A-1, A-3, or A-4) but in the failure analysis summaries in the appendices of Reference 1 where detailed study was made of all 399 failures. The intent was to isolate those failures that could be attributed to long-term subcritical crack growth. Failures occurring during system startup were not considered. In some cases engineering judgment was necessary because of the limitations of the original failure analyses. Analyses of 216 cases indicated that the failures manifested themselves as single or multiple cracks, and the failures occurred during time periods ranging from 60 to 307,000 hr (35 yr).

During system operation, subcritical cracks grew to critical sizes. Critical, within the 216 cases, means complete rupture or leakage (or detected before leakage) and the system was shut down. Often such cracks apparently formed and grew from defects such as pits, holes, abrasions, etc., which are readily identifiable even to an untrained eye. The mode of subcritical crack growth, on the other hand, often requires high magnification electron fractography to clearly identify the type and nature of the crack path (e.g., fatigue stress-corrosion, clearance, hydrogen embrittlement, etc). Considering that the majority of the failure analyses were conducted by people whose knowledge of cracking behavior was limited, the number of failures attributed to fatigue and stress-corrosion could be in error. For example, of the 216 failures cited above, the mode of failure or subcritical crack growth appeared to fall into categories as shown below in Table A-1.

The number of failures falling in the "unknown crack growth mechanism" category suggest that the number of failures attributed to fatigue (and also environmentally-assisted fatigue) in Table A-3 is in reality probably greater than the

indicated 29.1%. This is pertinent to the FTR Primary Piping System because 304 SS is more susceptible to fatigue (especially, thermal fatigue) because its coefficient of expansion is about double, while its thermal conductivity is about 1/2 that of plain carbon steel.

This discussion is not intended to detract from the work presented in Reference 1. The authors should be congratulated for piecing together a large number of obscure and often incomplete failure analyses. The "causes" of failure as noted in Table A-1 are probably accurate inasmuch as they describe reasons for failure based on design, material selection ... and operation-service. It is the mode of subcritical crack growth, however, that remains suspect.

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