1.3 FACTORS INFLUENCING FATIGUE STRENGTH.

Fatigue properties obtained from a carefully polished specimen in the ideal environment of the test laboratory are rarely achieved in practice. A wide variety of factors affect the behavior of a member or assembly under conditions of fatigue loading. The most obvious parameters are those that deal with the sign, magnitude, and frequency of loading; the geometry and material strength level of the structure, and the ambient service temperature. Those processing and metallurgical factors that determine the cleanness and homogeneity of materials, the sign and distribution of residual stresses, and the surface finish often are not considered. These processing and metallurgical factors, however, may have an overriding influence on the fatigue performance of the structure, to its benefit or detriment. The factors which influence fatigue strength will be classified into four main groups for discussion in this section:

1. Metallurgical Factors

2. Processing Factors

3. Environmental Factors

4. Design Factors

1.3.1 Metallurgical Factors.

The distinction between processing factors and metallurgical factors is not always clear. In fact, it is rather arbitrary in some areas. In this section, however, the focus is on regions within the material, either at the surface or core, which adversely affect fatigue properties. These regions may arise from melting practices or primary or secondary working of the material, or may be characteristic of a particular alloy system. In nearly every instance the detriment to fatigue properties results from a local stress-raising effect.

1.3.1.1 Surface Defects.

Primary and secondary working are often responsible for a variety of surface defects that occur during the hot plastic working of material when lapping, folding, or turbulent flow is experienced. The resultant surface defects bear such names as laps, seams, cold shuts, or metal flow through. Similar defects are also noted in cold working, such as fillet and thread rolling, in which the terms lap and crest cracks apply. Other surface defects develop from the embedding of foreign material under high pressures during the working process. Oxides, slivers, or chips of the base material are occasionally rolled or forged into the surface. The surface defects in castings might include entrapped die material, porosity, or shrinkage; in the extrusion or drawing processes such surface defects as tears and seams are not uncommon.

All of the aforementioned surface defects produce a notch of varying intensity which acts as a stress-raiser under load to the detriment of fatigue properties. Because most of these defects are present prior to final processing and are open to the surface, standard nondestructive testing procedures such as penetrant and magnetic particle inspection will readily reveal their presence. If they are not detected, however, the defects may serve as a site for corrosion or crack initiations during processing (in heat treating, cleaning, etc.), further compounding the deleterious effect on fatigue strength.

1.3.1.2 Subsurface and Core Defects, Inhomogeneity, and Anisotropy.

Subsurface and core defects considered here are those which originate in the as-cast ingot. Voids resulting from gas entrapment (porosity) and improper metal fill (shrinkage) are not uncommon in cast materials. In castings (ingots) that are to be subsequently hot and cold reduced, the portion of the ingot containing the preponderance of voids is often removed and discarded. The remaining internal defects normally weld shut under the combination of temperature and pressure involved in the reduction of the ingot, resulting in a continuous, homogeneous product. Occasionally, when the surfaces of the defects are oxidized or otherwise contaminated, healing (welding) of the opposite surfaces is precluded and the defective area is retained in the wrought product. Terms such as unhealed porosity and laminations are applied to this condition. Since these defects existed before working, in the final wrought product the major diameter of the oblate or rod-shaped flaw is parallel with the direction of plastic deformation.

Fatigue testing of high-strength aluminum alloy specimens containing defects of the type discussed in this section revealed the following trends:

1. Stressing parallel to the defect plane has a small effect on the fatigue strength, provided the defect does not intersect a free surface.

2. The effect of defect size on the fatigue strength in the short transverse direction of testing (that is, with the plane of the grain flow normal to the direction of loading) is shown in Fig. E1-13.

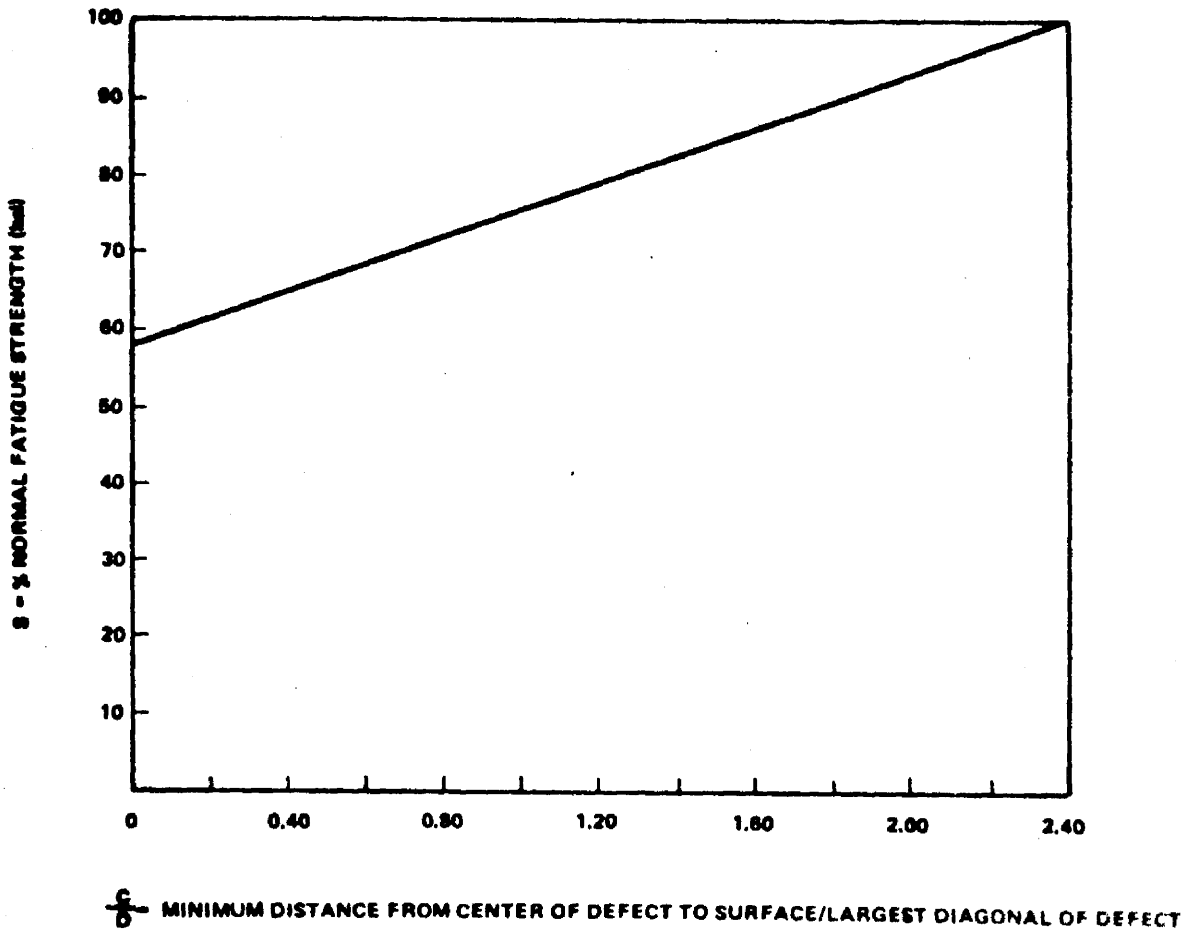


FIGURE E1-13. S VERSUS C/D

3. An internal defect adversely affects fatigue by introducing a stress concentrator into the material and reducing the load resisting cross-sectional area.

4. With respect to fatigue properties, when the edge of one defect is within approximately two diameters of the center of another defect, these should be considered as one large defect having a diameter equal to the extreme distance which will include both defects.

Inasmuch as most subsurface defects do not intersect a surface of a part, inspection is somewhat more difficult. For wrought products, ultrasonic or eddy-current testing might be used, whereas, for castings, fluoroscopic or radiographic inspection is preferred.

There are two types of inclusions in metals, nonmetallic and intermetallic. The amount and distribution of these inclusions is determined by the chemical composition of the alloy, the melting and working practice and the final heat treatment of the material. Nonmetallic inclusions are usually complex compounds of the metallic alloying elements with oxygen, nitrogen, carbon, phosphorus, sulphur, and silicon. The size of the inclusion is an important parameter in assessing its effect on fatigue properties, as shown in Figure E1-14 for 4340 steel heat treated to the 260 to 310 ksi tensile range. Although this relation does not apply to all inclusion types, it has been suggested that a separate curve exists for each predominant type of nonmetallic inclusion.

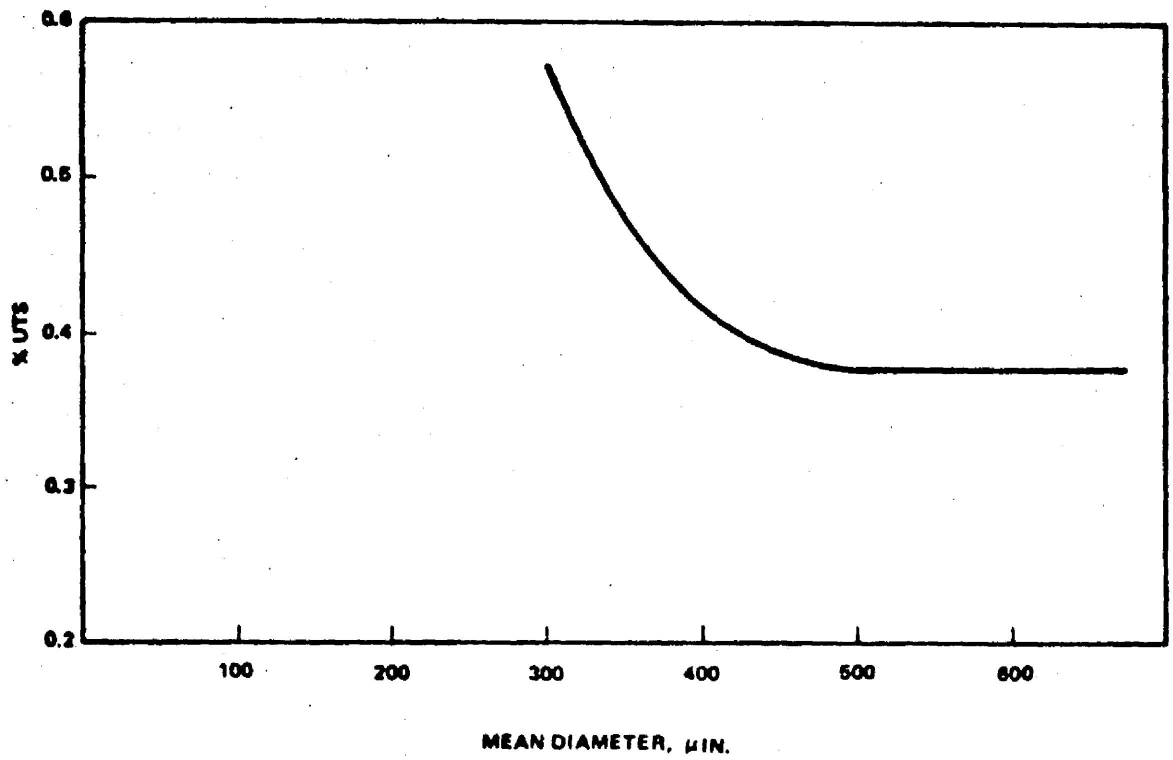


FIGURE E1-14. CORRELATION BETWEEN ENDURANCE LIMIT AS PERCENT UTS AND AVERAGE LARGE INCLUSION ARITHMETIC MEAN DIAMETER

Intermetallic inclusions may be either complex metallic compounds or second phases with variable compositions. The type of intermetallic constituent is believed to be an important consideration in determining the effect on fatigue life, although the mechanism is not clearly understood. The site of such an inclusion, however, is a discontinuous region with physical and mechanical properties different from those of the matrix phase. Under load these areas would serve as stress-raisers.

Some alloys are subject to microstructural banding which often has an adverse effect on fatigue properties. The banding is usually produced by local chemical segregration which stabilizes a phase not normally present in the alloy at room temperature. The severity of the loss in fatigue properties is dependent on the direction of the banding relative to the maximum stress direction (the banding is always in the direction of prior working) and on the degree of compatibility between the banded and matrix phases. Banded retained austenite and delta ferrite are occasionally seen in a large number of low-alloy and stainless steels. The presence of ferrite in these is intentional; in others it is not. The loss in fatigue properties produced by ferrite stringers in 431 stainless steel is shown in Fig. E1-15.

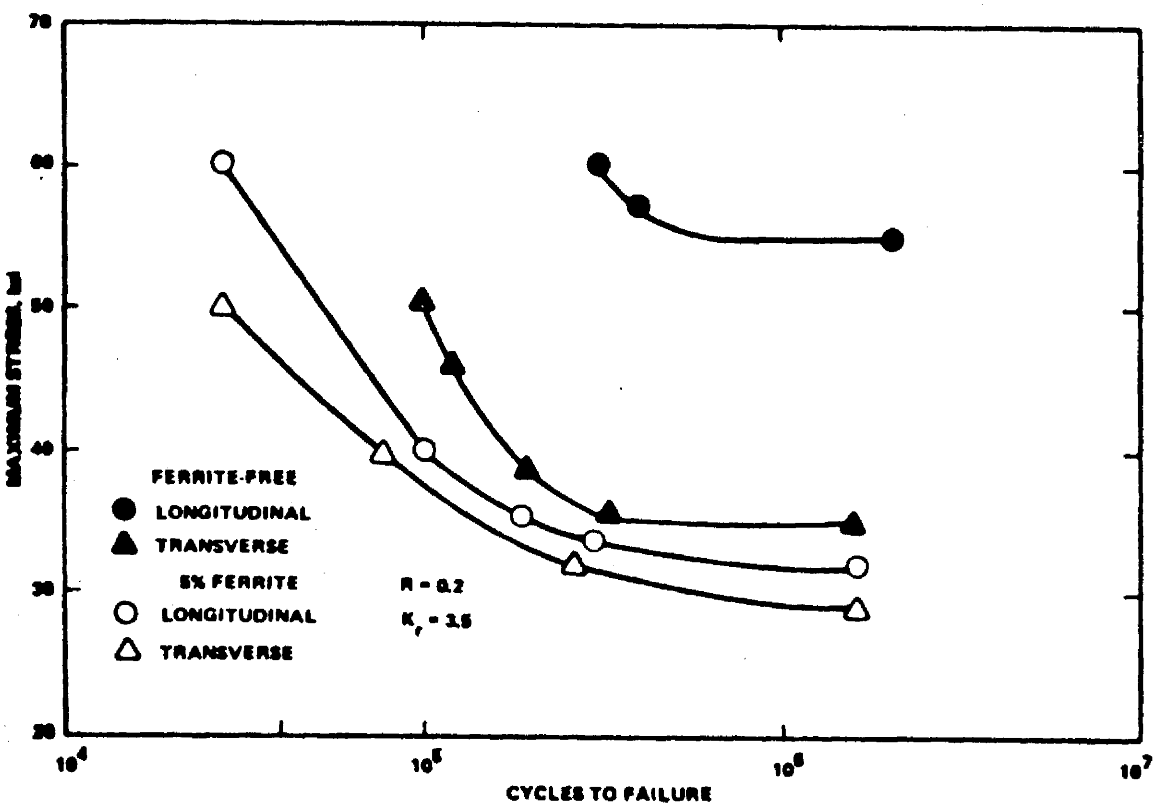


FIGURE E1-15. NOTCHED FATIGUE STRENGTH OF 431 STAINLESS STEEL HEAT-TREATED TO THE 180 to 200 ksi RANGE WITH NO FERRITE AND WITH 5 PERCENT FERRITE

Finally, the grain and subgrain structure may also reflect a preferential alignment. As previously indicated, anisotropy is most pronounced in the short transverse grain direction. It has been shown in tests on 7075-T6 aluminum alloy forgings that the endurance limit is reduced by approximately 20 percent when testing in the short transverse direction as opposed to the longitudinal direction.

For many material forms such as sheet, light plate, and extrusion, the loading normal to the short transverse direction is low such that fatigue properties in this direction are not critical. For heavy plate, bar, and forgings, however, directionality or anisotropy can be a crucial design consideration.

1.3.1.3 Heat Treatment.

The heat-treatment processes are potentially a source of hazard to a material because at the elevated temperatures encountered many diffusion controlled mechanisms are operative that could harm the integrity of the alloy if not properly controlled. If the furnace atmosphere is not controlled, the chemical composition of the surface layer might be altered and, thus, produce a low strength or brittle surface skin. The diffusion of hydrogen into alloys during heat treatment has long been recognized as a serious problem. Hydrogen embrittlement of low-alloy steels and titanium alloys can produce disastrous results in subsequent processing or in service. Hydrogen is also suspect in the blistering mechanism in aluminum alloys. With respect specifically to fatigue properties, a brittle case will render an alloy susceptible to surface cracking. The introduction of a shallow crack produces a notch effect, so that the detriment to fatigue (life) is essentially one of a high surface stress raiser in a layer of material with low fracture toughness.

If the heat-treating temperature is not properly controlled, grain coarsening may occur which lowers fatigue properties of some alloys. Overheating of high-strength aluminum alloys is particularly disastrous, since most of these alloys are subject to eutectic melting at temperatures only marginally higher than the solution heat treatment temperature. Eutectic melting results in a gross embrittlement of the alloy coupled with reduction in strength. The difficulties with austenitizing or solution heat treating at too low a temperature are associated with a lack of hardening potential for the subsequent quench and age or temper treatments.

In order to develop full strength, most martensitic and age hardening alloys must be rapidly cooled from high temperatures by quenching into a liquid medium. There are at least two considerations in the quenching process that could affect fatigue properties. High residual quench stresses are built up in most materials and, if the geometry of the part being quenched is highly irregular, the tensile strength of the material may be exceeded at points of high stresses resulting in the not too uncommon quench cracks. On the other hand, if the quenching rate is for some reason retarded, preferential precipitation may occur which adversely affects fatigue properties.

1.3.1.4 Localized Overheating.

There are some processes that are capable of developing high, localized surface temperatures, the consequences of which are often difficult to detect and occasionally are responsible for a failure in service. Grinding is one of these processes.

The effect of severe grinding on the fatigue properties of high-strength steel is shown in Fig. E1-16. The rapid quenching of the material immediately below the grinding wheel by the large mass of cold metal can produce cracks. If actual cracking does not result, brittle, crack-prone, untempered martensite might result or, with lower temperatures, softened, overtempered martensite. High-strength steels (for which grinding is most often used) are particularly sensitive to grinding techniques.

In the electroplating processes a plating burn sometimes is observed as the result of arcing between the anode and the work piece. Such a burn generally produces a larger heat-affected zone than improper grinding and is often characterized by evidence of surface melting. The potential damage to the substrate is similar to that discussed relative to grinding.

Electrical discharge machining (EDM) is a process of metal removal that employs a spark-erosion principle. The intermittent spark produces highly localized melting on the surface of the workpiece and metal fragments which are swept away by the dielectric coolant. Although the heat-affected zone is shallow, surface cracking and untempered martensite are sometimes observed on martensitic alloys along with eutectic melting and other evidences of overheating in aluminum alloys if the process is not properly controlled.

1.3.1.5 Corrosion Fatigue.

Corrosion fatigue is that peculiar interaction of a corrosive environment with an alternating stress field which causes accelerated crack initiation and propagation, possibly where neither the environment nor the stress acting alone would be sufficient to produce a crack. In the practical application of the term, the corrosive environment usually serves to introduce stress raisers

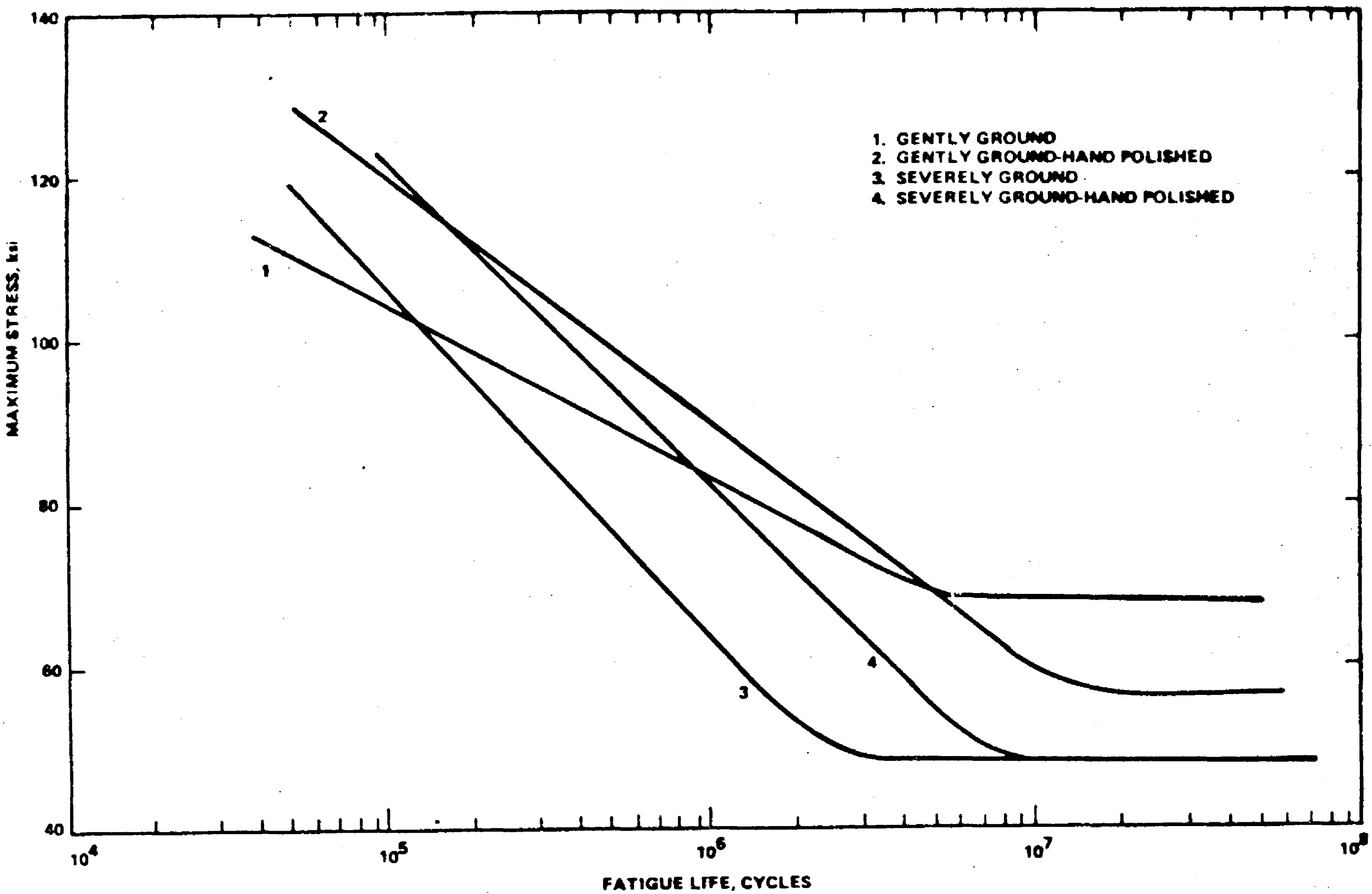


FIGURE E1-16. S-N CURVES FOR FLAT STEEL BARS (R_c-59 HARDNESS), SHOWING EFFECT OF GRINDING SEVERITY in the surface in the form of corrosive attack. The irregular surface, in turn, is detrimental to the fatigue properties of the part in a mechanical or geometric sense. For materials susceptible to embrittlement by hydrogen or for parts which are exposed to a fairly continuous corrosive environment with intermittent applications of loading, the cracking mechanism may be somewhat more complex. An example of corrosion fatigue testing is presented in Fig. E1-17, which illustrates the effect of a corrosive test environment on the fatigue properties of precipitation-hardened stainless steels.

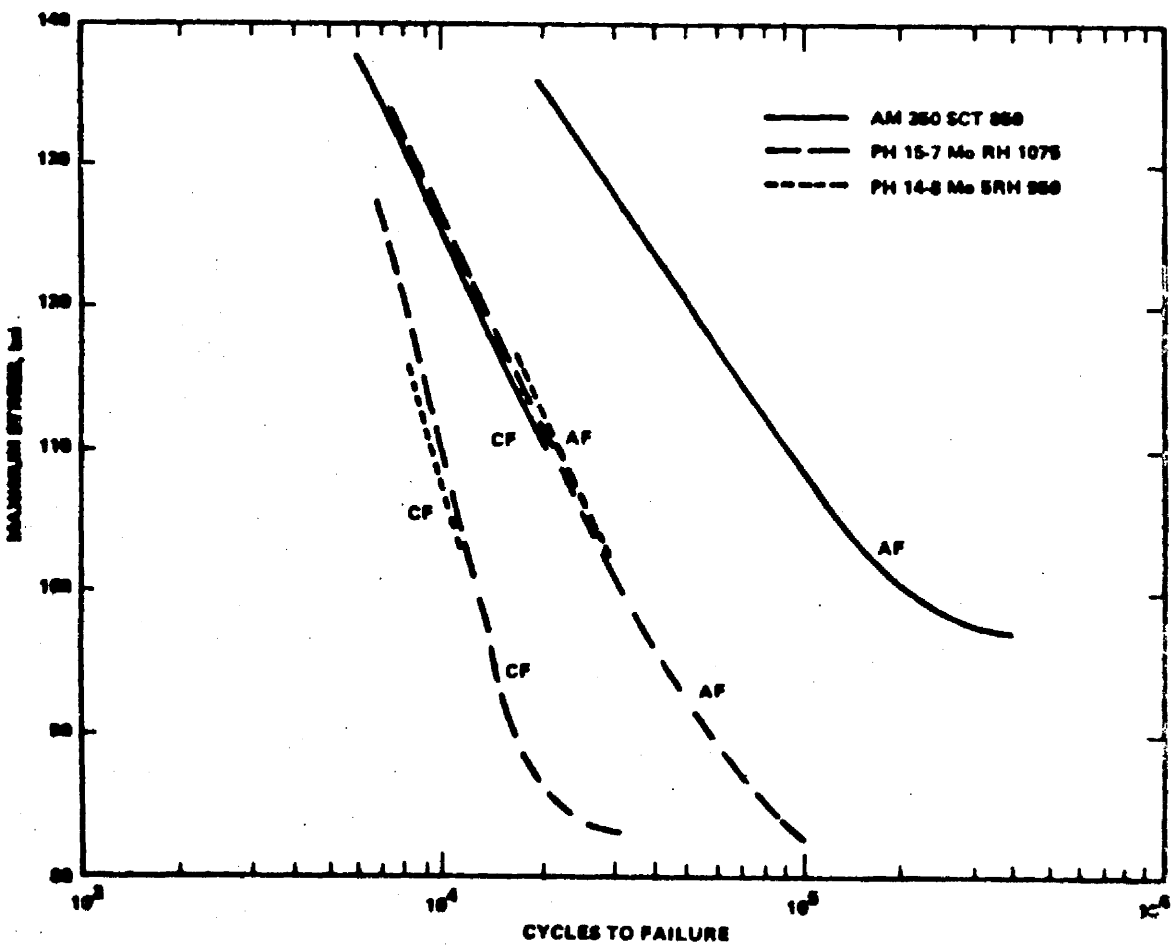


FIGURE E1-17. CORROSION FATIGUE AND AIR FATIGUE S-N CURVES FOR PRECIPITATION HARDENING STAINLESS STEEL TESTED AT ROOM TEMPERATURE

1.3.1.6 Fretting Corrosion.

The fretting corrosion phenomenon has been defined as that form of damage that arises when two surfaces in contact and normally at rest undergo relative periodic motion. In vacuum or inert atmospheres the process is completely mechanical, but in ordinary atmospheres oxidation is also involved. Fretting is potentially dangerous because it can result from extremely small surface monuments that often cannot be anticipated or even prevented. Motions with amplitudes as low as 5×10^{-9} inch are sufficient for this mechanism to be operative.

Soft metals exhibit a higher susceptibility to fretting fatigue than hard metals. Fretting corrosion increases with load-amplitude, number of load cycles, contact pressure, and an increase of oxygen in the environment. The oxidized particles that accumulate between the fretting surfaces lead to both chemical and mechanical surface disintegrations which generate nuclei for fatigue crack initiation. The presence of fretting may reduce fatigue strength by 25 to 30 percent, depending on loading conditions. When a part or assembly is known to be critical in fretting, one or a combination of the following factors will be beneficial in reducing or eliminating fretting corrosion:

1. Electroplating critical surfaces.
2. Case-hardening wearing surfaces.
3. Lubricating.
4. Eliminating or dampening vibration.
5. Increasing fastener load or closeness of fit.
6. Bonding elastic material to surface.
7. Excluding atmosphere.

1.3.1.7 Reworking.

The success of any repair or rework procedure is necessarily closely dependent on the analysis of the degrading mechanism. Only with a proper understanding of the cause of failure can a satisfactory permanent rework be accomplished. In the area of service damage caused by fatigue, in-service failure, or engineering test failure of a part usually provides the impetus to rework procedures. In general, these procedures can be separated into two categories: those parts that contain actual cracks and those that are believed to have undergone fatigue damage.

Usually, cracked structural parts are scrapped and replaced with a new part. Occasionally, however, because of the location of the crack or other circumstances, such a part is repaired. Repair would consist of removing the crack or blunting its root and supporting or strengthening the damaged area by means of doublers, straps, etc. Care must be taken in doubler design so that new sites of fatigue cracking are avoided. Factors such as fretting corrosion, dissimilar metal corrosion, detrimental stress redistribution, access, and practicality are prime considerations in establishing such a rework method.

Procedures frequently used to remove minor stress concentrators are those such as increasing a sharp edge, corner, or fillet radius, and grinding or buffing out coarse tool marks, nicks, and scratches. If assembly stresses are high, a joint having mismatched surfaces might be planed or mechanically realigned, or improved clearance could be provided. When fretting is contributing to fatigue cracking, a wear strip or lubricant may be inserted between the working surfaces, or the fasteners may be tightened to reduce or eliminate motion. Residual compressive stresses are often introduced into the critical areas of fatigue by shot peening or coining operations.

Estimating the depth of fatigue damage on a surface or below the tip of a fatigue crack is difficult and should be experimentally determined for all alloy-forming-heat-treating conditions and the load spectrum. Preliminary data indicate that the depth of fatigue damage beneath cracks for 7075-T6 aluminum alloy is approximately 0.003 inch. However, for high strength steel it may be many times greater than 0.003 inch.

1.3.2 Processing Factors.

Fatigue usually initiates at a surface because stresses are normally higher there, particularly since most parts undergo bending loads resulting in substantially higher stresses in the outermost fibers. The detrimental (or beneficial) effect of processing on fatigue properties is usually manifest in its effect on the strength level or residual stress condition, or both, of the surface material.

1.3.2.1 Hardness.

Strength of metals commonly used for engineering purposes is generally higher with increased hardness, up to a point. In steel, for example, increased hardness does not necessarily indicate a higher fatigue limit because the fatigue limit is also affected by the surface finish. Curves of average fatigue limit values for a range of surface finishes are shown in Fig. E1-18. Because these curves represent average values, allowance should be made for size effect (larger size generally means lower fatigue limit).

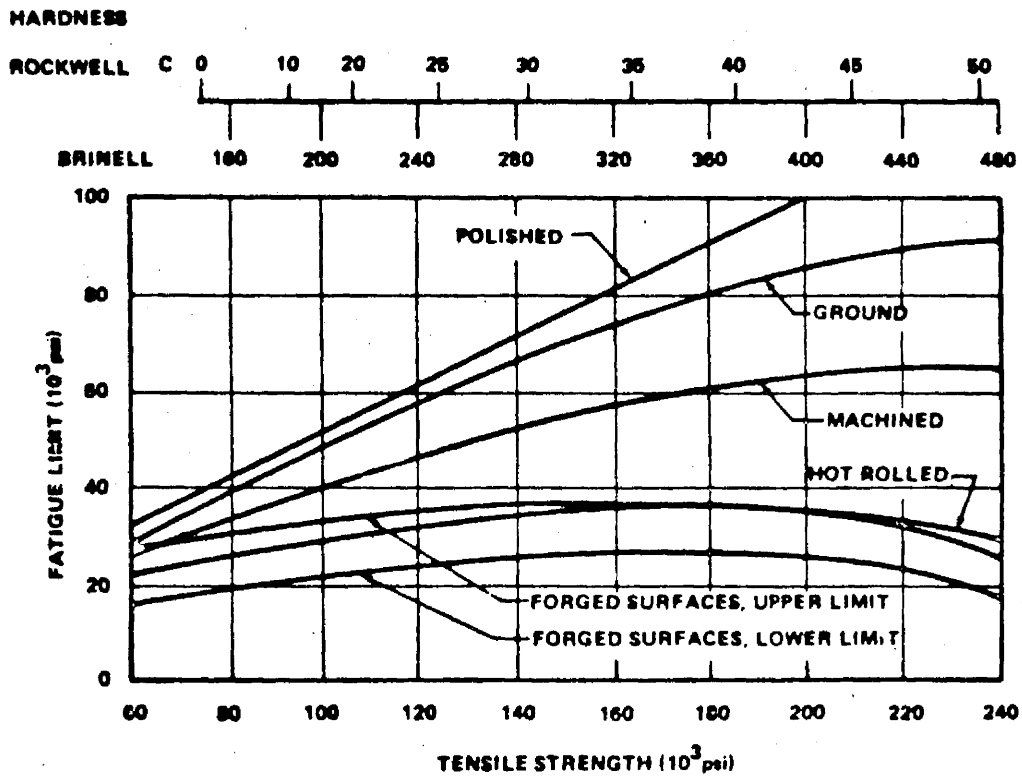


FIGURE E1-18. EFFECT OF HARDNESS AND SURFACE FINISH ON FATIGUE LIMIT OF STEEL IN REVERSED BENDING (0.3-INCH-DIAMETER SPECIMEN)

1.3.2.2 Forming.

By definition, the forming process produces plastic deformation (and residual stresses) in a part to achieve a permanent change in configuration. Occasionally these residual stresses may prove beneficial; however, usually there is some loss in fatigue life. Consequently, the residual stresses produced in forming (and their effect on fatigue) often dictate the forming limits for materials.

Residual forming stresses in the completed part are dependent on at least three additional factors: The heat-treatment-forming sequence in processing, the temper of the material, and the forming temperature. Parts formed and subsequently completely heat treated are free of prior forming stresses. Parts formed and stress relieved contain reduced forming stresses, depending upon the stress relieving temperature. The forming temperature and the material temper, e.g., AQ, T-4, or T-6 for aluminum alloys, also influence the magnitude of forming stresses to the extent that they affect the yield strength of the material at the forming temperature. In general, the lower the yield strength when forming occurs, the weaker the residual stress field generated.

1.3.2.3 Heat Treatment.

Residual stresses are both produced and relieved in many of the common heat treat cycles for both ferrous and nonferrous alloys. The principal source of residual stress occurs in quenching from high temperature solutioning or austenitizing treatments. Residual stresses are built up by nonuniform cooling rates between surface and core. For aluminum alloys, differential cooling produces residual surface compression and core tensile stresses. These surface compressive stresses are of sufficient magnitude to produce slightly higher fatigue strengths.

Aging temperatures for aluminum alloys are too low to produce any appreciable stress-relieving; however, most steels are tempered at temperatures sufficiently high to affect residual quench stresses. Consequently, for steels after tempering, quenching stresses are not recognized as a detrimental factor. Quenching stresses in aluminum alloys, however, persist after completion of heat treatment, as indicated by distortion in machining, increased susceptibility to stress corrosion and possible detrimental effects on fatigue life. To minimize these effects in aluminum alloys, special processing techniques have been developed, such as reducing section sizes by rough machining before heat treatment, use of less severe quenches where possible, and stress relief/equalization by cold working of quenched materials (for example, stretch-stress relief tempers).

1.3.2.4 Surface Finish.

A given surface-finishing process influences the fatigue properties of a part by affecting at least one of the following surface characteristics: smoothness, residual stress level, and metallurgical structure. The effects of surface finish on fatigue life for 7075-T6 extrusions are shown in Fig. E1-19. Here it can be seen that, in general, fatigue life increases as the magnitude

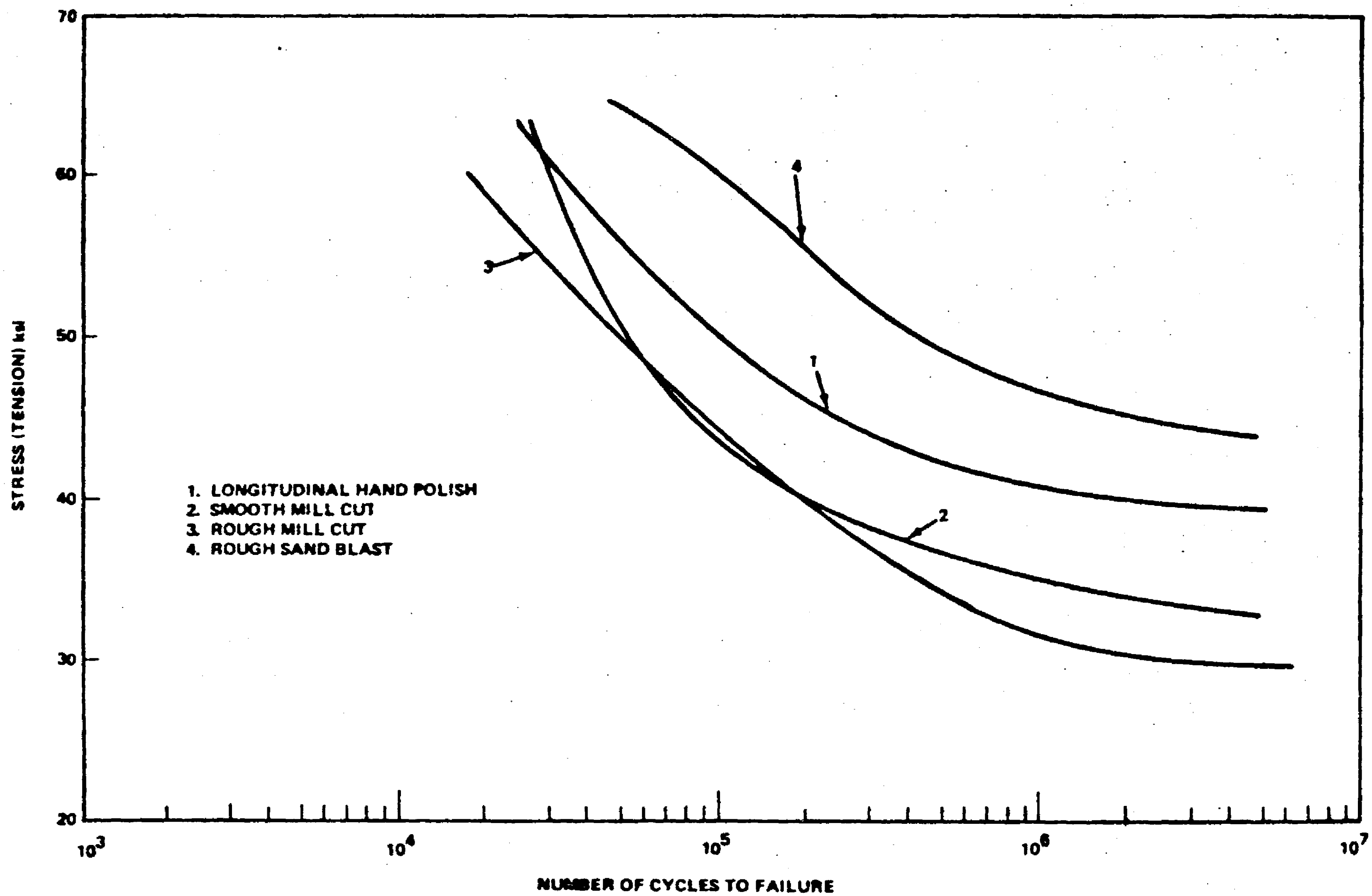


FIGURE E1-19. SURFACE FINISH VERSUS FATIGUE LIFE FOR 7075-T6 EXTRUSIONS of surface roughness decreases. Decreasing surface roughness is seen as a method of minimizing local stress raisers.

Aside from effect on surface roughness, the final surface finishing process will be beneficial to fatigue life when it increases the depth and intensity of the compressively stressed layer and detrimental when it decreases or removes this desirable layer. Thus, sandblasting, glass-bead peening, and other similar operations generally improve fatigue life properties. Conversely, processes, such as electropolishing, chem-milling, and electrical-chemical machining (ECM), which remove metal without plastic deformation at the tool point may reduce fatigue life properties.

Many local surface defects and irregularities occur in fabrication and service that are difficult to anticipate, inspect for, or control. Stress concentrations resulting from small indentations on an otherwise smooth surface in the form of accidental tool marks, grinding scratches, corrosion pits, or service-related minor damage are occasionally as effective in reducing the fatigue life of structural parts as large scale stress concentrations resulting from design deficiencies, depending on the material, heat-treat range, design margins, etc. Such unintentional stress-raisers are damaging in a structure only if their notch effect is more severe than the most severe stress concentration arising from design, unless they are located so as to intensify the stress-raising effect of the critical design feature. Usually the effect of design stress concentrations far outweigh the effect produced by surface finish.

1.3.2.5 Cladding, Plating, Chemical Conversion Coatings, and Anodizing.

Clad sheets become progressively weaker than the bare sheets as the lifetime increases. However, the reduction in fatigue life for plain fatigue specimens is not present to the same degree in built-up assemblies, where the fabrication stress raisers overshadow the effect of claddding on fatigue strength.

Aluminum metal spraying is sometimes applied to extrusions and forgings for added corrosion protection. The effect on fatigue properties is quite similar to that observed for cladding. However, it has been found that sprayed zinc finishes on aluminum alloys do not produce any measurable reduction in fatigue life. Also, it is generally agreed that neither zinc nor cadmium plating has any appreciable effect on fatigue properties.

Chemical conversion and anodic coatings which are applied to aluminum alloys for corrosion protection or wear resistance usually produce a reduction in fatigue life ranging from a negligible amount up to 10 or 15 percent of the endurance limit. The method of producing the coating further affects fatigue properties. As an example, fatigue tests run on chromic acid anodized 7075-T6 indicated that anodizing with a 5-percent dichromate seal offered a slight but definite lowering in fatigue life as compared to that of unanodized metal, whereas, a sulfuric type anodizing treatment resulted in a substantial reduction in fatigue life as shown in Fig. E1-20.

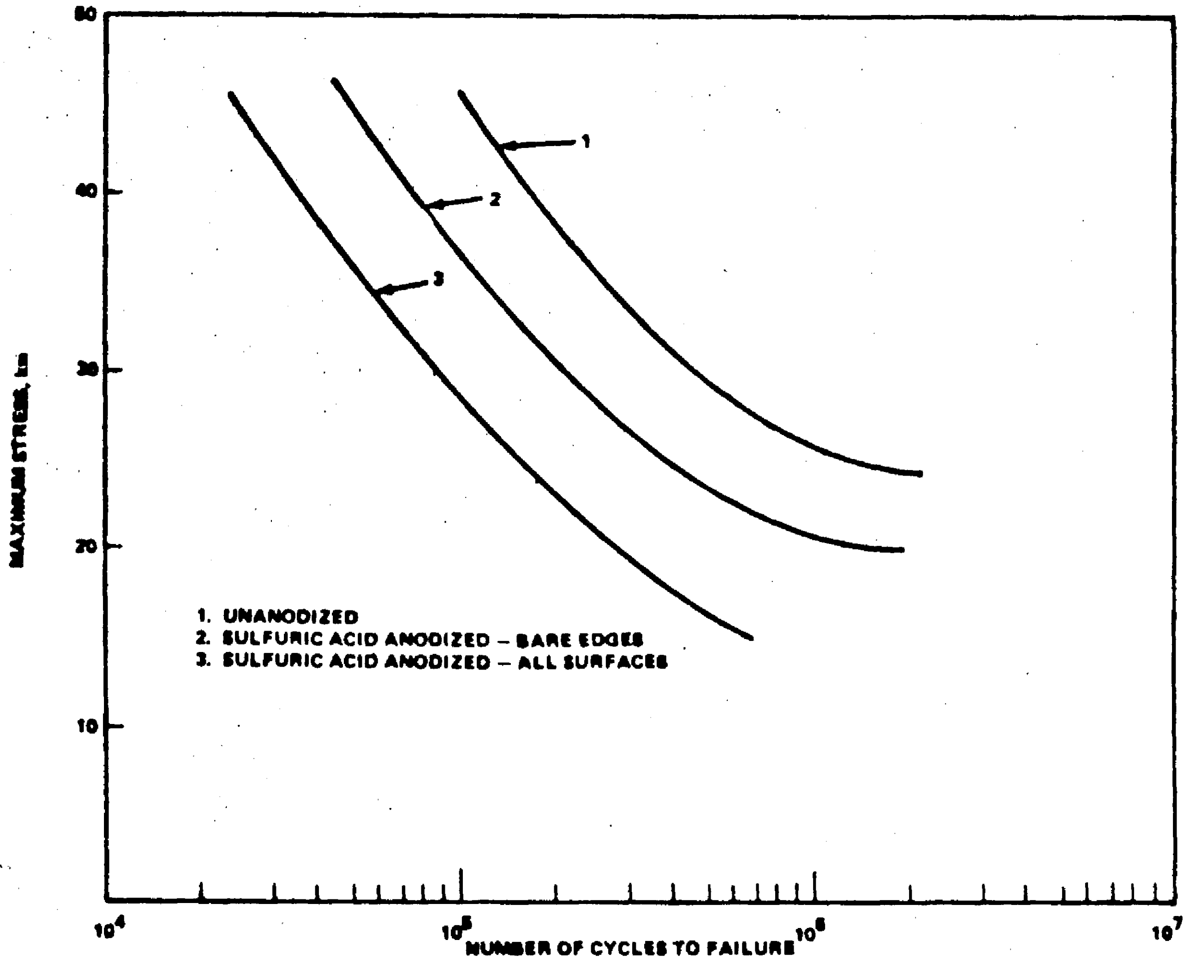


FIGURE E1-20. EFFECT OF SULFURIC ACID ANODIZE ON FATIGUE PROPERTIES: 7075-T6 ALCLAD SHEET 0.090 IN. THICKNESS, $R = +0.2$

Fatigue tests of 7075-T6 bare sheet with an Alodine #1200 coating showed no loss in fatigue properties compared to untreated material.

1.3.2.6 Cold Working.

Cold working of parts to induce residual surface compressive stresses has been found to be an effective tool for improving the fatigue life of both simple and complicated shapes. Methods of imparting cold work include coining of holes, thread rolling, fillet rolling, peening, hole expansion, prestressing, tumbling, and grit blasting.

It should be realized that residual surface compressive stresses are most beneficial under loading conditions which produce high surface tensile stresses, as in bending. Residual compressive stresses, however, are also beneficial to the fatigue life of axially loaded specimens.

For the desirable effect of surface cold working to be maintained, the cold-working process must be accomplished in the final heat-treated condition; subsequent thermal treatment must be eliminated when feasible, but closely controlled when it is essential.

Perhaps the most widely used process for inducing residual compressive stresses in surface material is shot peening. Its use is widespread because of its low cost and relatively easy application to a wide variety of materials and parts of varied size or configuration. As an example, peening might be used before chrome plating or hard anodizing, or in special situations, after a heat treatment which produced decarburization of a steel part.

When a part undergoes loads that are much higher in one direction than in the opposite direction, prestresssing the part by applying an excessive load in the direction of major service loading will often produce beneficial residual surface compressive stresses and work hardening that will significantly improve fatigue life properties. In reverse loading this effect is lost and a detriment to fatigue life may result. In the aircraft industry, prestressing is often used for torsion bars and bomb hooks, to name two examples. It has been shown in some specimens that tensile prestressing to 90 percent of the tensile strength increased the endurance limit about 100 percent.

Techniques involving other types of cold working can be found in Ref. 2.

1.3.3 Environment Effects.

The fatigue properties of a metal can be greatly influenced by the nature of the environment, especially in long-time tests at low stresses and at low frequencies. Even normal air has a deleterious effect on fatigue life properties, and recent indications show that moisture and oxygen are the two principal adverse components. Moisture decreases the fatigue strength of copper, aluminum, magnesium, and iron by 5 to 15 percent. An example of the influence of environmental air moisture on the fatigue properties of an aluminum alloy is given in Fig. E1-21.

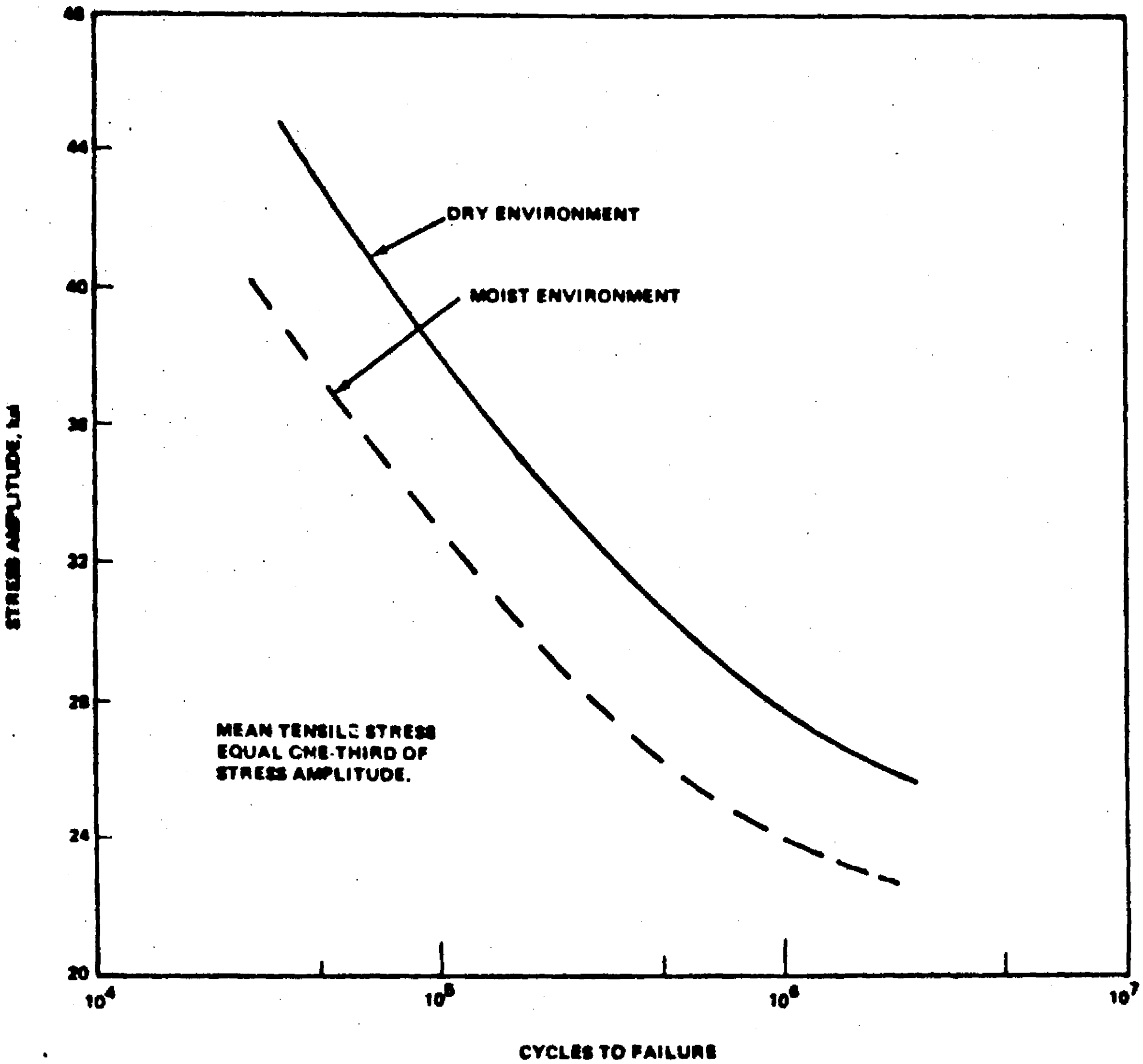


FIGURE E1-21. S-N CURVES FOR CLEAN SPECIMENS OF ALUMINUM ALLOY 6060-T6 IN HIGH- AND LOW-HUMIDITY ENVIRONMENTS

It is generally agreed that the main effect of a normal atmosphere with respect to fatigue life is on the propagation of cracks rather than in their initiation. The principal effect of the environment will be in its reaction to the clean metal exposed at the tip of a fatigue crack. Cracks that would be of a nonpropagating variety in vacuum propagate in the presence of oxygen or moisture; thereby, they lessen the fatigue resistance.

In environments that are more corrosive than air and that substantially reduce the resistance of the surface layers to fatigue crack initiation, both initiation as well as propagation may be affected. Environments that lead to stress-corrosion cracking may also adversely affect fatigue properties. (Refer to Paragraph 1.3.1.5.)

1.3.3.1 Irradiation.

Some investigations have been made with respect to the effect of nuclear irradiation on mechanical fatigue properties of metals. Radiation-induced changes are usually referred to as radiation damage since in many cases the effects have been detrimental in one way or another. Damaging effects such as loss in ductility have been noted for many metals. On the other hand, beneficial effects evidenced increased yield and ultimate strengths, fatigue strength, and surface hardening also have been noted. Rotating beam fatigue tests on 7075-T6 aluminum alloy (Ref. 3) indicate an improvement in life due to the effects of a total integrated flux of 2×10^{18} fast neutrons per cm^2. Although the amount of irradiation received by the specimens in these tests is believed sufficient to alter mechanical properties, it is not known if the results are significant for the characteristics of metals during the conjoint action of fatigue straining and irradiation. Additional studies need to be undertaken for the simultaneous action of irradiation and mechanical straining at low temperatures.

1.3.3.2 Vacuum.

It has been demonstrated experimentally that fatigue cracking under uniaxial direct stressing of an aluminum alloy occurs in hard vacuum almost as readily as within atmospheric pressure. This is contrary to general belief. Experiments have disclosed trends toward increased crack growth rates in vacuum which may eventually result in characteristics more critical than growth rates obtained from in-air tests. Results on aluminum continuously held in vacuum for periods greater than a week yielded drastic reductions in fatigue properties. This strongly indicates the time dependency of the phenomenon. Results from short-time test exposures in vacuum have shown improved properties. This is the usually accepted belief. However, for many cases an extension of the vacuum outgassing time provides a more realistic environment than do the short-time test exposures previously investigated. Because of this anomaly, prolonging the vacuum exposure is suggested as the only reliable procedure at present for evaluating metals for service in space environment.

1.3.3.3 Meteoroid Damage.

In the environment of space, tiny meteoroids and micrometeoroids may strike the surfaces of a space vehicle. At present, no information is available as to this effect on fatigue properties. However, it can be reasoned that indentations caused by these meteoroids provide ideal sites for the initiation of fatigue cracks by reason of the increased stress concentration.

1.3.3.4 Solar Irradiation.

Solar irradiation will cause differential heating on orbiting spacecraft. Over long periods of time, the thermal cycling and the resulting fatigue could become a serious problem.

1.3.3.5 Temperature.

Elevated temperatures may cause metals to lose strength; low temperatures may cause some metals to become notch sensitive and fail by brittle fracture under load conditions that would be common at normal (standard) temperature.

In general, the fatigue life of most metals will decrease with an increase in temperature (Fig. E1-22). Studies have indicated that as the temperature decreases, the crack nucleation period (or time to produce an observable crack) increases. Moreover, as temperature decreases, the period of observable fatigue crack growth decreases.

The rate of fatigue-crack propagation as affected by rate of cyclic loading and test-load frequency is an additional parameter to be considered. In elevated-temperature fatigue testing, it is known that the number of cycles to fracture decreases and the crack-growth rate increases as the speed or frequency of cyclic loads is decreased. The damaging, thermally activated mechanism of creep or creep cracking, acting conjointly with fatigue, is responsible for this behavior. In general, this is true because, in the accumulation of stress cycles, slower rates of load cycling result in exposure of the metal to temperature for longer periods of time than do fast rates of load cycling. However, from cryogenic temperatures to room temperature, no damaging, thermally activated mechanisms are active. It is believed that fatigue lives and crack-growth rates at cryogenic temperatures will be independent of load frequency for most metals.

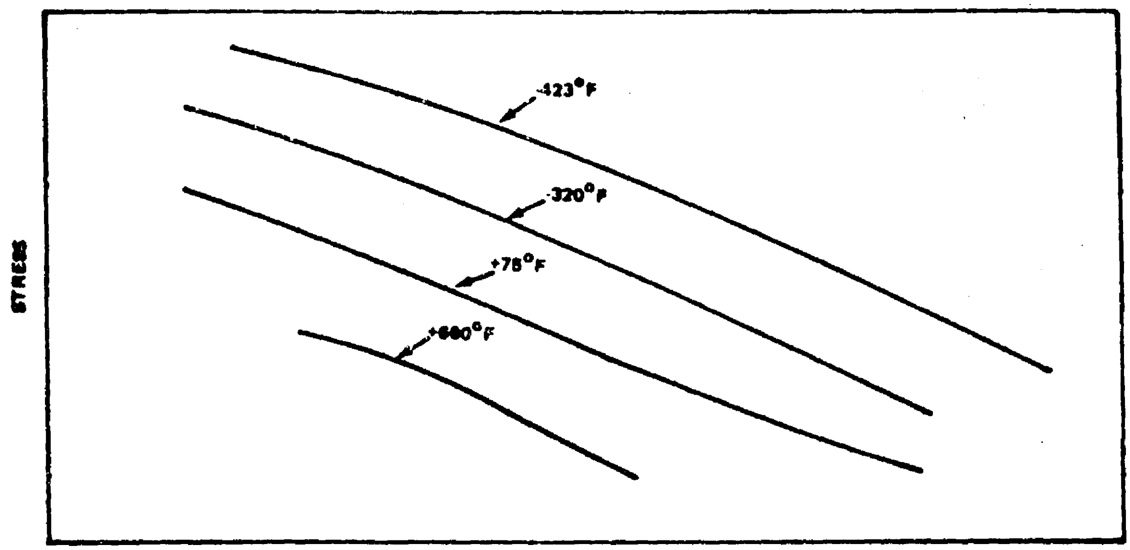


FIGURE E1-22. FATIGUE LIFE AS A FUNCTION OF TEST TEMPERATURE (TYPICAL FOR MOST METALS)

Additional information on thermal and creep effects will be given in Paragraph 1.4.3.

1.3.4 Design Effects.

Manufactured parts often include abrupt changes in cross section such as grooves, fillets, holes, and keyways; these cause a nonuniform distribution of elastic strains and stresses. The maximum elastic stress at an abrupt change in cross section is greater than the nominal stress obtained by calculation with either the direct stress, flexural, or torsional formulas.

The geometric stress-concentration factor, K, is defined as the ratio of the maximum stress in the section to the nominal, or average, stress. Nominal stress is usually based on the net remaining section. Because of exceptions, it is necessary to ascertain the basis upon which the nominal stress is established before applying a stress concentration factor.

The full geometric stress-concentration factor is not always applicable in fatigue loading. The actual fatigue limit of a notched member is frequently higher than would be expected from the geometric stress-concentration factor. This phenomenon is called notch sensitivity. Use of the notch-sensitivity factor results in a fatigue stress-concentration factor that is determined from:

$$K_f = q(K_t - 1) + 1$$

where K_f is the fatigue stress-concentration factor for direct tension or bending; K_t is the geometric stress concentration factor for direct tension, or modified geometric stress-concentration factor for bending (Ref. 4); and q is the notch sensitivity. Typical values of q versus notch radius for steels of various strengths are shown in Fig. E1-23.

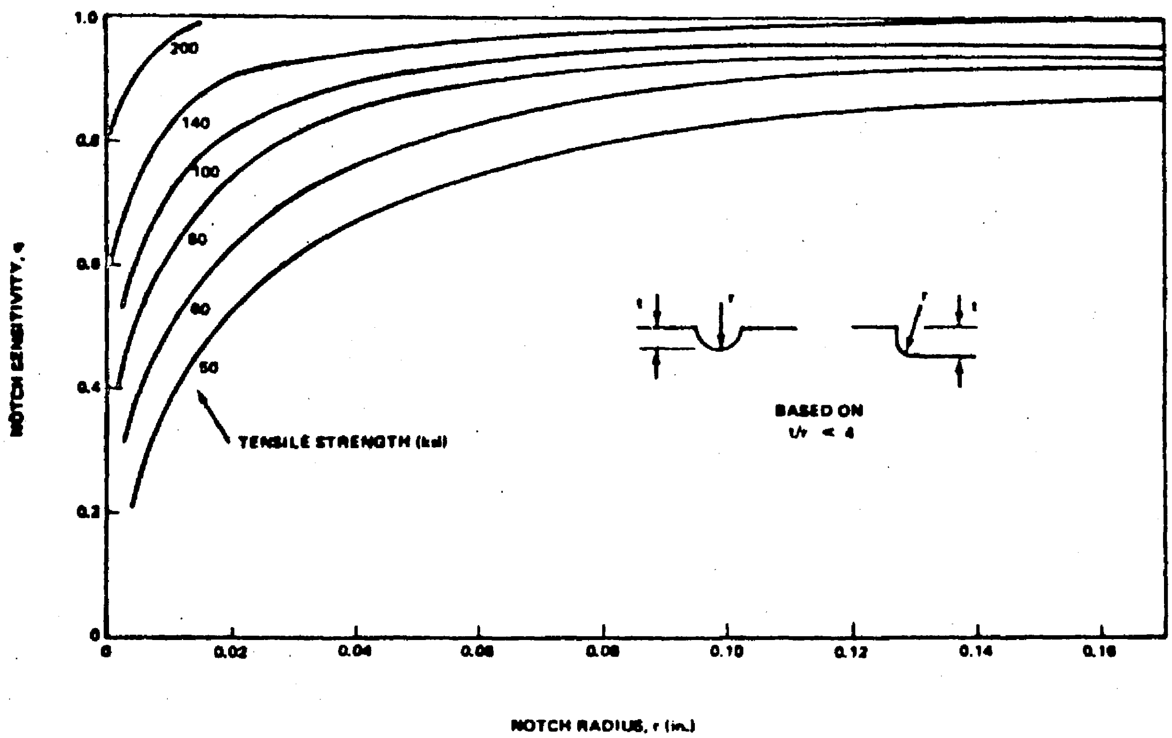


FIGURE E1-23. NOTCH-SENSITIVITY CURVES FOR POLISHED STEEL SPECIMENS OF VARIOUS STRENGTHS

The fatigue-stress concentration factor, K_f, represents the extent to which a notch can be expected actually to reduce the fatigue limit of a part. It is the ratio of the fatigue limit of the specimen without the notch to its fatigue limit with the notch. Specimens must have the same effective section when K_f is evaluated experimentally so that only the effect of the notch is determined without including an effect which is due to reduction in the section.

When one notch is located in the region of maximum influence of another notch, the resulting stress-concentration factor is determined by multiplying the two individual stress-concentration factors. Whether the maximum influence of one notch encounters the maximum influence of the other depends on the stress patterns developed by the two notches and the extent of overlap. When stress concentrations are adjacent to each other, but are not superimposed to the extent that one is placed in the region of maximum influence of the other, the resulting stress concentration factor lies between the value of the larger factor and the product of the two stress-concentration factors. In steel members with numerous notches and discontinuities, a value of $K_f = 3$ is applicable in most cases, and a value of $K_f = 4$ is the maximum likely to be encountered.

1.3.5 Welding Effects.

There exists a considerable amount of information on the fatigue properties of various welding effects in steel (Ref. 5). However, the amount of information on welding effects in nonferrous materials is relatively scarce, primarily because of the late development of welding of aluminum alloys. As a consequence, the use of welded aluminum in fatigue-loaded structures has been limited; however, it is becoming more extensive, particularly for applications where weight savings is of importance. This frequently implies that fatigue strength is the design criterion.

It has been found that by far the most important factor in the fatigue behavior of welded joints in aluminum and steel is the geometrical shape, both of the joint as a whole and of the weld bead. Parent material seems to be a relatively unimportant variable.

The fatigue limit of welded high-strength steels is only slightly higher than for low-strength steels if the weld bead is left in place. However, if the weld bead is removed, the fatigue limit is increased; the amount of increase depends upon the smoothness of the finished surface. Even partial removal of the weld reinforcement increases the fatigue limit considerably. For welded steel with the weld undisturbed, fatigue limits of approximately 12 000 psi in reversed bending and 23 000 psi in zero to maximum tension are common. These are average values, and an allowance for a lower failure rate should be made. Removal of weld beads also improves fatigue characteristics of aluminum.

1.3.6 Size and Shape Effects.

Usual specimens for fatigue tests are in the range of 0.2- to 0.5-inch diameter. Some reduction in fatigue limit with increasing size of specimen has been observed. This reduction is generally attributed to the fact that larger sizes stressed in bending or torsion have a larger volume of material in the region of high stress. This increases the probability that a defect will occur in the high stress region, leading, in turn, to a higher probability of failure in larger sizes.

Fatigue tests in axial loading show relatively little size effect, probably because of the uniform stress across a section under an axial load.

Size has little effect upon the failure stress at 1000 cycles. Thus, a size correction is neglected at the 1000 cycle point of the S-N curve and applied only to the fatigue limit.

Shape of cross section has some effect on the flexural fatigue limit of some materials. Steel beams of circular cross section have fatigue limits of 8 to 10 percent higher than square beams. Square beams loaded so that the maximum stress occurred at the corners of the square exhibited fatigue limits from 4 to 8 percent less than round beams.

1.3.7 Speed of Testing.

The effect of the speed of testing has been studied by a number of investigators. In the range of from 200 to 7000 cycles per minute there appeared to be little effect on the fatigue properties, except when temperature increases in the specimens were encountered. At very low speeds a slight decrease in fatigue strength was noted, while for very high speeds some increase in strength was found.

Of course, the speed of testing is of great importance at the low-cycle side of the stress-versus-life curve and also when thermal cycling or creep is involved. These effects are covered in Paragraphs 1.4.2 and 1.4.3.

1.4 LOW-CYCLE FATIGUE.

A complete S-N curve may be divided into two portions: the low-cycle range and the high-cycle range. There is no sharp dividing line between the two. We might arbitrarily say that from 0 to about 10^3 or 10^4 cycles is low cycle and from about 10^3 or 10^4 cycles to 10^7 or higher is high cycle.

Until World War II little attention was paid to the low-cycle range, and most of the existing results were for high cycles only. Then it was realized that for some pressure vessels, missiles, spaceship launching equipment, etc., only a short fatigue life was required. Consequently, the low-cycle fatigue phenomenon began to gain attention.

In the low-cycle range of fatigue life below approximately 10 000 cycles, the primary parameter governing fatigue life appears to be plastic strain per cycle as measured on a gross scale. For higher (cyclic) lives, elastic strain also assumes importance; perhaps the governing variable is still plastic strain per cycle, but the plastic strain is highly localized at imperfections in the structure and is difficult to measure or compute. Beginning at approximately 10 000 cycles and continuing upward, it becomes more appropriate to regard total strain (elastic plus plastic) as the primary variable. Alternatively, the fatigue life can be regarded as being governed by stress range, and the ensuing simple relationship between fatigue life and stress range appears to be valid for most of the materials that have been tested to date, over a very large range of life on both sides of 10 000 cycles. However, for most thermal-stress problems the stress range is less likely to be known than is the strain range; hence, for practical purposes it is desirable to express life in terms of total strain range. The relationships as applied to problems of both thermal and mechanical loading are discussed below.

The failure mechanism in the low-cycle range is close to that in static tension, but the failure mechanism in the high-cycle range is different and may be termed "true fatigue." A comparison of the failure mechanisms in the two ranges is made in Table E1-1.

Table E1-1. Comparison of Low Cycle and High Cycle Fatigue

	Low Cycle	High Cycle
Internal stresses and strain hardening	High	Low
Net sum of plastic flow	Micro size	Micro size
Gross sum of plastic flow	Small	Large
X-ray disorientation	Large	Small
Slip	Coarse (10^3-10^4Å)	Fine (10Å)
Slip plane distortion	Normal	Persistent
Crack origin	Interior	Surface
Crack path	Along maximum shear	Cross maximum tensile stress
Fracture	Delayed static	Structural deterioration

The terms used in this section are defined below and are shown in Figure E1-24.

E = modulus of elasticity

N = cycles to failure

ϵ_t = total strain range

ϵ_e = elastic portion of strain range

ϵ_p = plastic portion of strain range

C = a material constant

ϵ_f = true strain at fracture, commonly known as fracture ductility

$S = \frac{1}{2} E \epsilon_t$

ΔS = stress amplitude

S_e = endurance limit

S_u = ultimate tensile strength in ordinary tensile test

S_y = yield strength

S_{ut} = true stress at fracture in tensile test

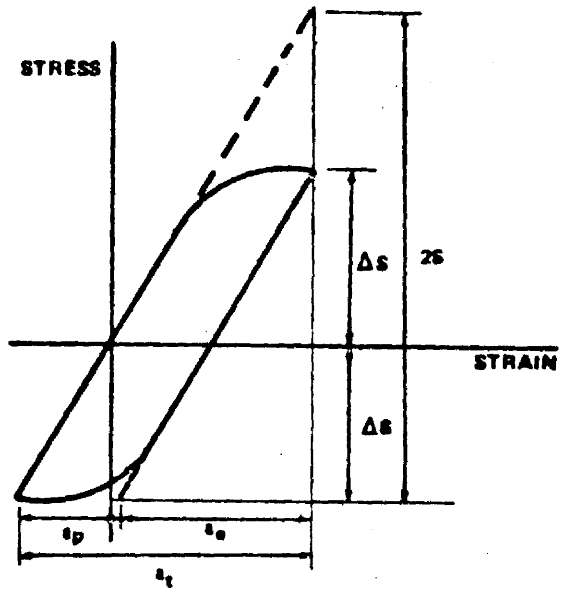


FIGURE E1-24. STRESS-STRAIN CYCLE

1.4.1 Below Creep Range.

In recent years, low-cycle fatigue has been studied extensively under conditions where little attention has been paid to temperature and temperature effects. There now exists a fairly large body of information on the subject obtained from laboratory experiments under room temperature conditions (Ref. 6). The general procedure in such investigations is to control the strain between fixed limits and, if desired, to measure the stress, while cycling the specimen until failure is determined. Such tests may be under conditions of controlled diametral strain, longitudinal strain, bending strain, or torsional strain. The results can be represented in terms of total strain range versus cycles to failure, or, if desired, of plastic strain range versus cycles to failure. The former is most often used for direct design procedures, while the latter is applied when one is primarily interested in real material behavior. One of the more fascinating features of representing the results in the form of plastic-strain range versus cycles to failure is the observation that the curve (on logarithmic scales for both $\Delta\epsilon_p$ and N_f) is of constant slope of value 0.5 to 0.65 (depending on the experimenter) and that the vertical position of the curve relates to the tensile ductility of the material. A plot of this curve is shown in Fig. E1-25. The line in this figure can be expressed by the equation

$$\Delta\epsilon_p \sqrt{N} = C$$

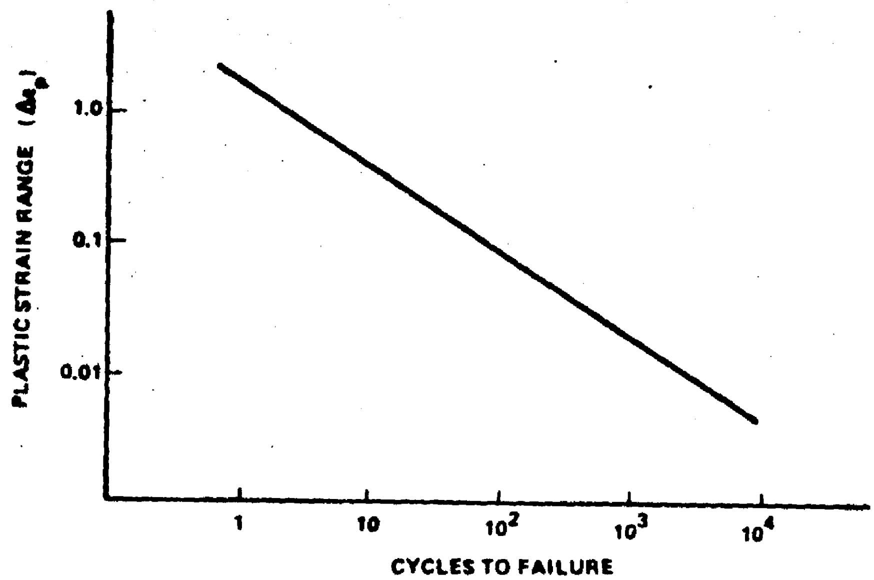


FIGURE E1-25. PLASTIC STRAIN RANGE VERSUS CYCLES TO FAILURE

where the constant C may be related to the reduction of area RA in static tension test.

$$C = \frac{1}{2} \ln \frac{100}{100-RA} .$$

A fair degree of information and understanding now exists for the low-temperature (below the creep range), low-cycle fatigue problem, although there are a number of unresolved questions yet to be answered.

1.4.1.1 Practical Problem Solutions.

The fatigue curve or equation needed by the engineer or designer is one which shows stress versus cycles and one that contains sufficient safety factors to give safe allowable design stresses for a given number of operating cycles; or, conversely, allowable operating cycles for a given value of calculated stress. The stress values on the fatigue curves or in the equations should be directly comparable to the stress values which the designer calculates using his usual methods of analysis for pressure stress, thermal stress, stress concentration, etc.

For fatigue below the creep range, Langer (Ref. 7) has proposed an equation for plotting the fatigue curve for any material:

$$S = \frac{E}{4\sqrt{N}} \ln \frac{100}{100-RA} + S_e \quad ,$$

(refer to Fig. E1-24 for definitions of terms).

For a design stress to represent a lower bound on the data, it is recommended that the equation above be used with a safety factor of either 2 on stress or 10 on cycles, whichever is more conservative at each point. It is believed that these safety factors are sufficient to cover the effects of size, environment, surface finish, and scatter of data.

It also has been shown that in the low cycle range (below 10 000 cycles), mean stress has little or no effect on fatigue life; thus, this effect can be ignored.

Some typical values of E and RA for various materials are given in Table E1-2. If information on the endurance limit (S_e) is not available one-half the value of the ultimate strength is often used.

If a more detailed fatigue analysis is required the techniques proposed in References 6 and 8 should be used.

1.4.2 In Creep Range.

The high-temperature, low-cycle fatigue problem is in quite a different state. The material behavior becomes very much more complicated at elevated temperatures because of the occurrence of creep and other diffusion processes. At temperatures in the creep range of a material, the test results will be very strongly frequency dependent. Citing the fact that frequency is important in high-temperature, low-cycle fatigue is the same thing as saying that the strain rate of the test is important. A somewhat related situation occurs if one introduces a hold time into the cycle, say, at the point where the tensile stress is a maximum. While the strain is held constant, the peak stress relaxes for some period of time. Results of prolonged hold times during strain cycling show that the fatigue life decreases with increasing hold times, other conditions being the same. A common observation is that, as the temperature is raised and the rate decreased (or hold time increased), the character of

Table E1-2. Mechanical and Elastic Properties of Materials

Material	Yield Strength S_y (ksi)	Ultimate Strength S_u (ksi)	Reduction in Area, RA (percent)	Modulus of Elasticity (E)	Poisson's Ratio (μ)
4130 (soft)	113.	130.	67.3	32×10^6	0.29
4130 (hard)	197.	207.	54.7	29	0.28
4340 (ann.)	92.	120.	43.4	28	0.32
4340 (hard)	199.	213.	38.1	29	0.30
52100	279.	292.	11.2	30	0.29
304 ELC (ann.)	37.	108.	74.3	27	0.27
304 ELC (hard)	108.	138.	68.8	25	0.34
310 (ann.)	32.	93.	63.5	28	0.30
350 (ann.)	64.	191.	52.1	28	0.32
350 (hard)	270.	276.	20.3	26	0.30
Inconel X	102.	176.	19.7	31	0.31
Titanium (6 Al-4V)	172.	179.	41.0	17	0.33
2014-T6 Al	67.	74.	25.0	10	0.33
5456-H311 Al	34.	58.	34.6	10	0.33
1100 Al	14.	16.	87.6	10	0.33
Beryllium	38.	46.9	1.7	42	0.024

the fracture changes from one which is transgranular in nature to one which is largely intergranular. This suggests very strongly that the problem is, to a large degree, metallurgical in nature. A close resemblance can be seen between these observations and those found in creep and stress rupture.

A second important, and certainly related, effect which must be appreciated when dealing with high-temperature fatigue is the change in metallurgical structure which occurs with time at temperature and stress.

Plastic strain and, in the present instance, cyclic plastic strain can have an important bearing on the precipitation and strengthening effects which take place in the metal structure. The term "strain-age hardening" describes the effect that plastic strain has on developing additional strength. It is well known, for example, that the level of strength reached by applying a fixed amount of strain to a metal which strain ages is greater if the strain is applied at the aging temperature. In designing an alloy for creep, advantage can be taken of strain-age hardening provided the strain involved is the creep strain selected for the design.

1.4.2.1 Ductility Versus Creep Strength.

As mentioned above, creep strengthening can be accomplished for specific alloys at various regions of temperatures. However, it should be noted that, simultaneously with strengthening, these same regions show reduced fracture ductility.

This well-known behavior of metals, the inverse relationship between strength and ductility, introduces a dilemma into the selection of materials for high-temperature service. It is necessary, to optimize the strength, to employ alloy additions and heat treatments so as to develop the precipitate structure in the metal, but this will lower the ductility. Ductility is of prime consideration in low-cycle fatigue and, to optimize fatigue resistance, it is desirable to have the material in as ductile a condition as possible (a condition which results in low creep strength). However, one wants both creep strength and fatigue resistance, and it is apparent that the best that can be achieved is some compromise condition.

1.4.2.2 Procedure for Estimating High-Temperature, Low-Cycle Fatigue.

The procedure given here is taken from References 9 and 10. It is a procedure for estimating the high-temperature, reversed-strain-cycling fatigue characteristics of laboratory specimens. It should be emphasized that the method is intended to give only first-approximation estimates as a guide to material selection. Final design of important equipment should be based on actual fatigue data generated under conditions which simulate as closely as possible those to be encountered in service.

I. Basis.

Figure E1-26 illustrates the method of universal slopes. The total strain range is divided into its elastic and plastic components, and each component is plotted against cyclic life on log-log coordinates. Straight lines usually result for each of these components. The method of universal slopes prescribes that the slopes of these lines are assumed to be the same for all materials. The plastic components are taken to have an average slope value of -0.6, and the elastic components an average slope value of -0.12. Again, it should be emphasized that these values are not always -0.6 and -0.12 for all materials; values between -0.4 and -0.8 have been obtained for various materials for the plastic component, and values between -0.08 and -0.16 have been found for the elastic components. When average values of -0.6 and -0.12 are chosen for the plastic and elastic components, reasonable results are obtained.

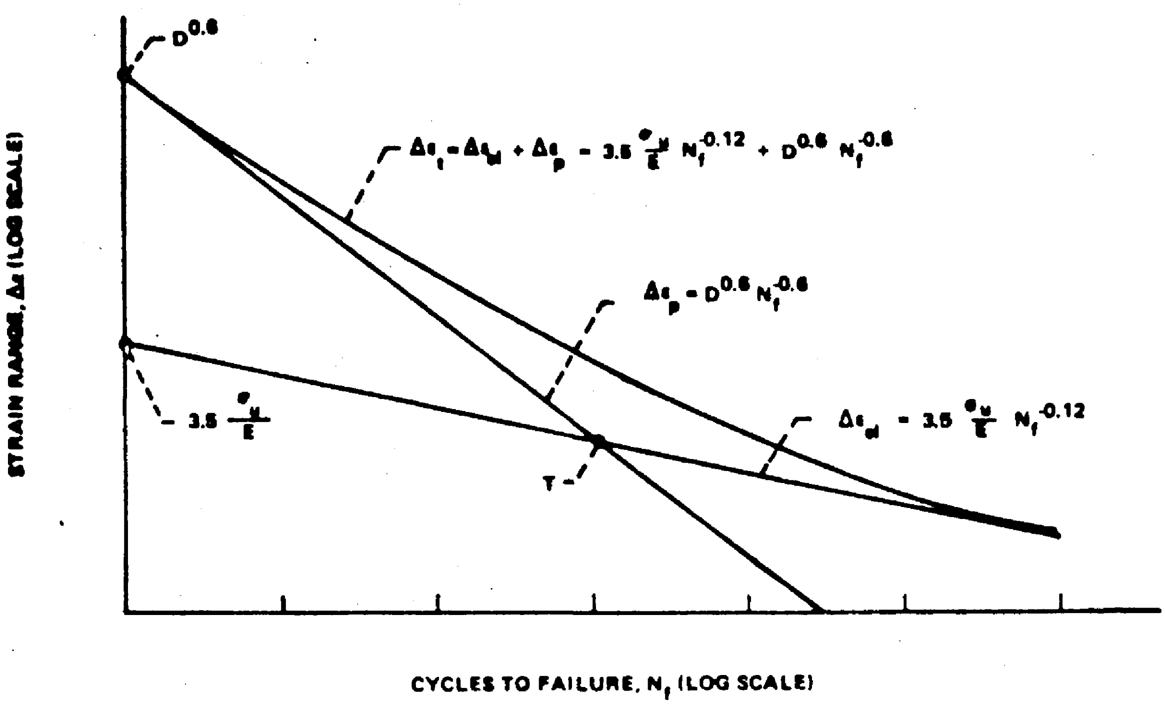


FIGURE E1-26. METHOD OF UNIVERSAL SLOPES FOR ESTIMATING AXIAL FATIGUE LIFE

Having decided on the slopes, it is necessary to determine the intercepts of the two straight lines. In general, it has been found that the property which most significantly governs the intercept of the plastic line is the ductility D, where $D = \ln \frac{100}{100-RA}$.

For the elastic line, the governing property for the intercept is σ_u/E, where σ_u is the ultimate tensile strength and E the elastic modulus. The total strain range, given as the sum of the elastic and plastic components, thus becomes

$$\Delta\epsilon_t = \frac{3.5\,\sigma_u}{E} (N_f)^{-0.12} + D^{0.6} (N_f)^{-0.6}$$

as shown in Fig. E1-26.

Good agreement has been obtained with this formula for many materials (Ref. 9). However, the question arises: Can one use this procedure at high temperatures within the creep range of the material? In general, it has been found that the procedure almost always yields fatigue life predictions higher than those actually obtained by testing. The explanation for the unconservative results is, of course, very complex. The relationship between high-temperature, low-cycle fatigue and intercrystalline cracking is very important.

It has been shown that intercrystalline cracking due to a creep effect occurs early in the fatigue life, and thereby bypasses the transcrystalline crack-initiation period; thus, the total fatigue life is approximately equal to only the crack-propagation period.

This concept has yielded the 10-percent rule. In this method, crack initiation and propagation are not regarded as functions of fatigue life; rather it is assumed that, on the average, only 10 percent of the fatigue life computed by the method of universal slopes will actually be realized at temperatures within the creep range.

Generally, it has been found that such a computation yields conservative results; that is, in most cases, it leads to a lower bound on fatigue life. However, there are cases in which even the 10-percent rule predicts lives that are higher than those actually achieved in test. This observation (together with the fact that the 10-percent rule approach inherently excludes the possibility of taking into account frequency effects, hold-time effects, mean load, etc.) causes one to seek further for simple methods of estimation that are not as limited.

The simplified analysis that has been adopted can be explained by the following. The creep damage effect is taken as the ratio of the time actually spent at stress to the time required to cause rupture at that stress value. Since the stress and temperature are presumably known, this rupture time can be obtained directly from the creep-rupture curve of the material. The fatigue damage effect is taken as the ratio of the number of cycles actually applied to the number that would be sustained in the absence of creep effect according to the method of universal slopes. Since the test frequency yields a definite relation between the time of the test and number of cycles sustained, a closed-form analytical expression for the number of cycles to failure can be obtained. This description is shown in the following derivation:

$$(\text{Creep-Rupture Damage}) + (\text{Fatigue Damage}) = 1$$

$$\left(\frac{t'}{t_r}\right) + \left(\frac{N_f'}{N_f}\right) = 1$$

where

$$t' = \frac{k}{f} N_f'$$

$$t_r = A\left(\frac{\sigma_r}{1.75}\right)^{1/m} = A(N_f)^{-0.12/m} \quad ;$$

hence,

$$\left(\frac{kN_f'}{AF(N_f)^{-0.12/m}}\right) + \left(\frac{N_f'}{N_f}\right) = 1$$

or

$$N_f' = \frac{N_f}{1 + k/AF(N_f)^{(m + .12)/m}}$$

Here

N_f' = number of cycles to failure under combined fatigue and creep

N_f = number of cycles to failure in fatigue, based on method of universal slopes, using ductility, ultimate tensile strength and elastic modulus from uniaxial tests at strain rates comparable to that undergone by the metal in the fatigue test. Where appropriate data are not available, use may be made of data from conventional tensile tests.

k = empirical constant, assumed to be 0.3, but adjustable as more information becomes available.

m = slope of straight line-creep-rupture curve on log-log coordinates, that is, representing curve by $\sigma_r = 1.75\ \sigma_u\ (t_r/A)^m$ as shown in the insert in Fig. E1-27.

A = coefficient in creep-rupture relation

F = frequency of cycling, cycles per unit time.

Table E1-3 contains some of the information above, obtained from various test conditions (Ref. 9).

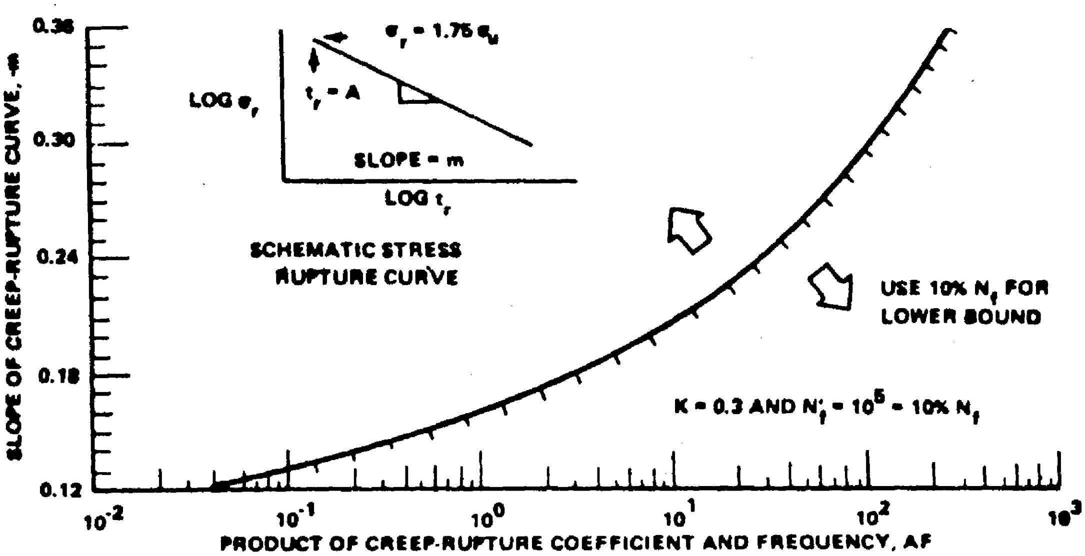


FIGURE E1-27. CRITERION TO ESTABLISH NEED FOR CREEP-RUPTURE CORRECTION

Table E1-3. Alloys, Test Conditions, and Pertinent Properties

Alloy Designation	Test Temperature (°F)	Test Frequency (cpm)	ΔA (%)	Tensile Strength (ksi)	Elastic Modulus (10^3 ksi)	Creep-Rupture		
						Slope (-m)	Intercept, A (min)	Intercept, A (sec)
A-286 — CR 34%	1000	5 to 50	31.	182.	26.8	<0.12		
	1200	5 to 50	38.	167.	23.1	<0.12		
A-286 — Aged	1200	5 to 50	18.	115.	23.0	<0.12		
A-286	1400	20	57.6	75.	19.0	0.20	1.3	78.
Inco 901	1400	20	16.1	104.	20.8	0.135	0.028	0.17
D-979	1260	5 to 50	42.0	179.	22.0	<0.12		
Cr-Mo-V Steel	1050	1, 10	70.0	69.5	24.0	<0.12		
L-605	1000	5, 50	48.2	106.0	27.0	<0.12		
	1200	5, 50	47.1	95.4	25.0	<0.12		
304 Stainless Steel	1000	10	67.3	55.5	23.0	<0.12		
	1200	3 to 18	58.0	41.0	22.0	0.18	13.0	900.
	1500	3 to 27	42.0	20.0	19.0	0.18	0.3	18.0
347 Stainless	1110		67.7	55.0	22.0	<0.12		
Nimonic 75	1200	10	29.5	81.5	20.3			
	1380	10	42.0	53.1	15.4			
	1600	10	62.5	28.0	12.2			
	1800	10	66.0	16.2	9.7			
Nimonic 90	1200	10	24.5	144.0	26.5	<0.12		
	1380	0.1, 10	12.0	119.0	25.0	0.135	0.16	9.6
	1600	0.1, 10	14.0	76.0	23.0	0.24	2.5	150.
	1800	0.1, 10	94.0	16.0	21.0	0.25	4.0	240.
Nimonic 105	1200	10	13.7	142.0	25.0	<0.12		
	1380	10	18.0	151.0	23.5	<0.12		
	1600	10	32.5	99.5	22.0	0.16	0.08	4.8
	1800	10	68.0	39.2	19.0	0.28	3.7	222.
Udimet 700	1400	1 to 2	31.0	155.0	23.0	<0.12		
Astroloy	1400	20	29.6	141.0	23.0	<0.12		
Inconel	1500	0.017, 0.1	63.3	25.0	18.0	0.15	1.0	60.
In 100 (PWA 47 Coat)	1000	6	9.3	121.0	27.1	<0.12		
	1700	6	11.0	96.0	22.7	0.17	0.11	6.6
	2000	6	10.3	26.8	19.3	0.22	1.3	78.
MAR M 200 (PWA 47 Coat)	1300	6	5.	120.0	26.5	<0.12		
	1700	6	4.	77.0	23.7	<0.12		
B 1900 (PWA 47 Coat)	1300	6	5.	145.0	25.5	<0.12		
	1700	6	3.	95.0	23.0	0.17	0.75	45.
	2000	6	9.	35.0	20.5	0.26	7.3	440.

II. Method.

In summary, two procedures for computing the lower bound on fatigue life have been outlined. Proceed:

1. By the 10-percent rule. The universal slopes equation is first used to determine cyclic life, and this life is divided by a factor of 10.

2. By the combined creep and fatigue effect.

The problem of determining which of the equations to use, and how to interpret the results has been given considerable study and the following conclusions can be drawn:

a. Determine life by both methods, and use the lower of the two calculated values. Figure E1-27 is an auxiliary plot that minimizes the computations needed to determine which value to use. It is merely necessary to determine the product, AF, and the slope, m, of the creep-rupture curve at the test temperature. If the point representing the coordinates lies above the curve, use the equation for N_f'. If it lies below the curve, use the 10-percent rule.

b. The lower of the two values determined in a. serves as a lower bound on estimated life.

c. As an estimate of average or most probable life, use twice the lower bound determined in b.

d. As an estimate of the upper bound on life, use 10 times the lower bound determined in b.

It has been found that, in most cases encountered in the laboratory, the 10-percent rule is the applicable one.

As tempting as it is to conclude that the method gives results good enough for many engineering uses, it is perhaps of greater importance to emphasize some of the cautions involved in its use. It must first be emphasized that the data analyzed relate to constant amplitude-strain cycling under constant temperature. The method appears to place the high-temperature, low-cycle fatigue behavior in the proper range. Thus, it would seem that while the method may be regarded as very good for screening materials, important material choices should be made on the basis of actual tests from among the more promising materials. And, whenever possible, the complexities of stress and temperature history expected in service should be included in the test evaluation.

1.4.2.3 Method of Strain-Partitioning.

In their continuing examination of high-temperature, low-cycle fatigue, Manson and co-workers (the 10-percent rule) have developed a method which attempts to analyze directly the effects of creep and plastic flow on cyclic life (Ref. 11). This method was developed because of the highly time-dependent nature of the creep effect and because the life-reducing effects of creep depend, to a large extent, upon where within a cycle the creep is introduced and whether it is reversed by plastic flow or by creep.

In this method, the completely reversed cyclic inelastic strain is divided into four parts. These are (1) $\Delta\epsilon_{pp}$ – tensile plastic flow reversed by compressive plastic flow, (2) $\Delta\epsilon_{cp}$ – tensile creep reversed by compressive plastic flow, (3) $\Delta\epsilon_{pc}$ – tensile plastic flow reversed by compressive creep, and (4) $\Delta\epsilon_{cc}$ – tensile creep reversed by compressive creep.

In any arbitrary hysteresis loop, such as shown in Fig. E1-28, the tensile inelastic strain ($\overline{AD}$) can be separated into plastic ($\overline{AC}$) and creep ($\overline{CD}$) components. Likewise, the compressive inelastic strain ($\overline{DA}$) can also be separated into its plastic ($\overline{DB}$) and creep ($\overline{BA}$) components. In general, neither the two plastic components ($\overline{AC}$ and $\overline{DB}$) nor the two creep components ($\overline{CD}$ and $\overline{BA}$) will be equal. It is necessary only that the entire tensile inelastic strain ($\overline{AD}$) be equal to the entire compressive inelastic strain ($\overline{DA}$) since we are dealing with a closed hysteresis loop. The partitioned strain ranges are obtained in the following manner. The completely reversed plastic strain range, $\Delta\epsilon_{pp}$, is the smaller of the two plastic components and in this example is equal to $\overline{DB}$. The completely reversed creep strain range, $\Delta\epsilon_{cc}$, is equal to the smaller of the two creep components and becomes $\overline{CD}$. As can be seen graphically, the difference between the two plastic components must be equal to the difference between the two creep components, or $\overline{AC} - \overline{DB} = \overline{BA} - \overline{CD}$. This difference is then equal to $\Delta\epsilon_{pc}$ or $\Delta\epsilon_{cp}$ in accordance with the notation just stated. For this example, it is equal to $\Delta\epsilon_{pc}$ since the tensile plastic strain is greater than the compressive plastic strain. It follows from the preceding procedure that the sum of the partitioned strain ranges will necessarily be equal to the total inelastic strain range or the width of the hysteresis loop.

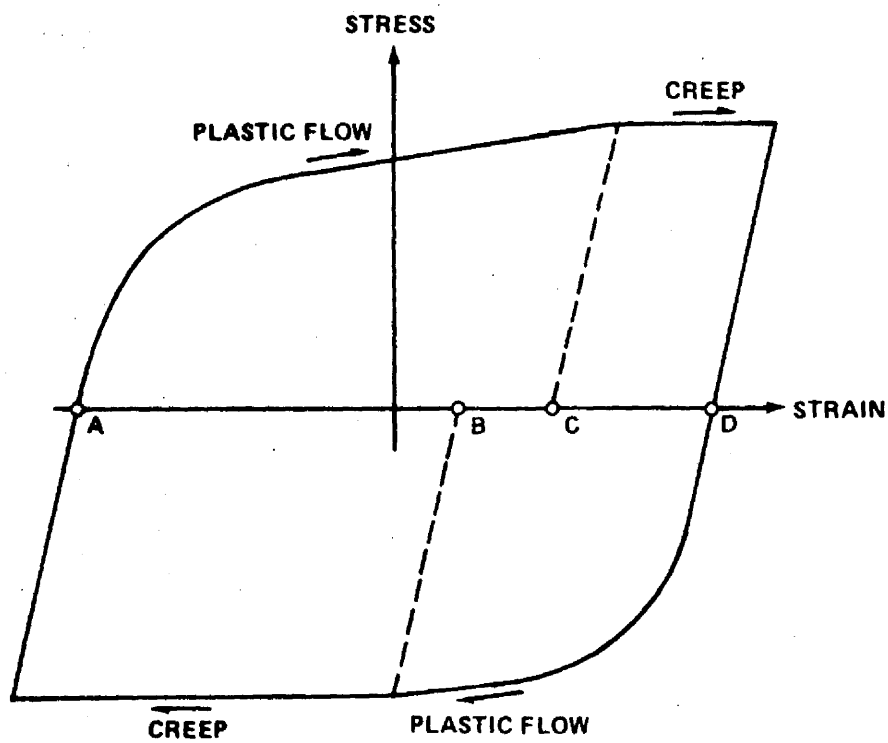


FIGURE E1-28. HYSTERESIS LOOP

The effect of each of these strain ranges on cyclic life is noted in Ref. 11, and the total cyclic life is determined using a linear cumulative damage relationship,

$$\frac{N}{N_{pp}} + \frac{N}{N_{cp} \text{ or } N_{pc}} + \frac{N}{N_{cc}} = 1 \quad ,$$

where N_{pp} is the cyclic life associated with the strain range of $\Delta\epsilon_{pp}$, N_{cp} or N_{pc} is the cyclic life associated with $\Delta\epsilon_{cp}$ or $\Delta\epsilon_{pc}$, and N_{cc} is the cyclic life associated with $\Delta\epsilon_{cc}$.

This method differs from the 10-percent rule in that it attempts to deal directly with the phenomena occurring at high-temperature, low-cycle fatigue rather than using an estimate based on low-temperature, low-cycle fatigue findings. It differs from other methods which partition damage effects at high temperatures in that all relations used are strain range versus cyclic life rather than a combination of that and stress versus time life. The generation of strain range versus cyclic life data is required in this method where the 10-percent rule requires only static tensile properties. This method agrees well with the 10-percent rule for the materials tested and may prove a basis for that rule. This method is also useful in explaining frequency effects and hold-time effects. Finally, to quote the authors of this method, "The method is in an early stage of development, and many questions must be answered before its merits can be evaluated" (Ref. 11).

1.4.2.4 Two-Slope Fatigue Law.

Coffin and Manson have proposed the following fatigue law for plastic strain:

$$N_f^{\beta} \Delta\epsilon_p = C \quad , \tag{1}$$

where N_f is the cycles to failure, $\Delta\epsilon_p$ is the plastic strain range, and β , C are the constants. This has been modified to include the effect of frequency in high-temperature, low-cycle fatigue,

$$\left(N_f \nu^{k-1}\right)^{\beta} \Delta\epsilon_p = C_2 \quad , \tag{2}$$

where ν is the frequency, k , C_2 are the constants, and $N_f\nu^{k-1}$ is defined as the frequency-modified fatigue life. The frequency effect was extended further to include the stress range,

$$\Delta\sigma = A \left(\Delta\epsilon_p\right)^{\eta} \nu^{k_1} \tag{3}$$

and

$$\Delta\sigma = \text{stress range} \quad ,$$

where A, η, and k are constants. Combining equations (2) and (3) and generalizing for the total strain range

$$\Delta\epsilon = (AC_2{}^{\eta}/E)\ N_f{}^{-\beta\eta k_1 + (1-k)\beta\eta} + C_2 N_f{}^{-\beta}{}_{\nu}{}^{(1-k)\beta} \quad , \qquad (4)$$

where E is the elastic modulus.

Coffin reports in Ref. 12 that with increasing temperature the exponent β increases approaching one and the constant C correlates less well with the tensile ductility. Because of the high exponent for β in equation (1) reported for many materials at elevated temperature, extrapolation to low lives could lead to plastic strain ranges exceeding the tensile ductility. Because this situation is highly unlikely, there must be a break in the curve of the $\Delta\epsilon_p$ versus N_f relationship so that it will pass through the tensile ductility value at $N_f = 1/4$ (Fig. E1-29). This change in slope is due to a change in the mode of crack propagation. In less ductile materials this change in slope is due to a change in the fracture mode resulting from environmental interaction. In ductile materials this change in slope at high temperatures results from a change from intergranular to transgranular fracture and as can be seen in Fig. E1-30, as temperature increases the change in slope is more drastic.

The main benefit of the two-slope fatigue law is that it gives a more accurate description of the occurrences in plastic strain and leads to a clearer recognition of the operative physical process. A recognition of the causes of the two-slope phenomenon emphasizes the importance of the environment and of time in the environment (frequency and hold time). However, the constants of the equations are highly material dependent and temperature dependent and if experimental data are not available, these constants cannot be determined.

1.4.3 Thermal Cycling.

The previous sections have considered mechanical strain or stress cycling at a constant temperature; this section will consider mechanical

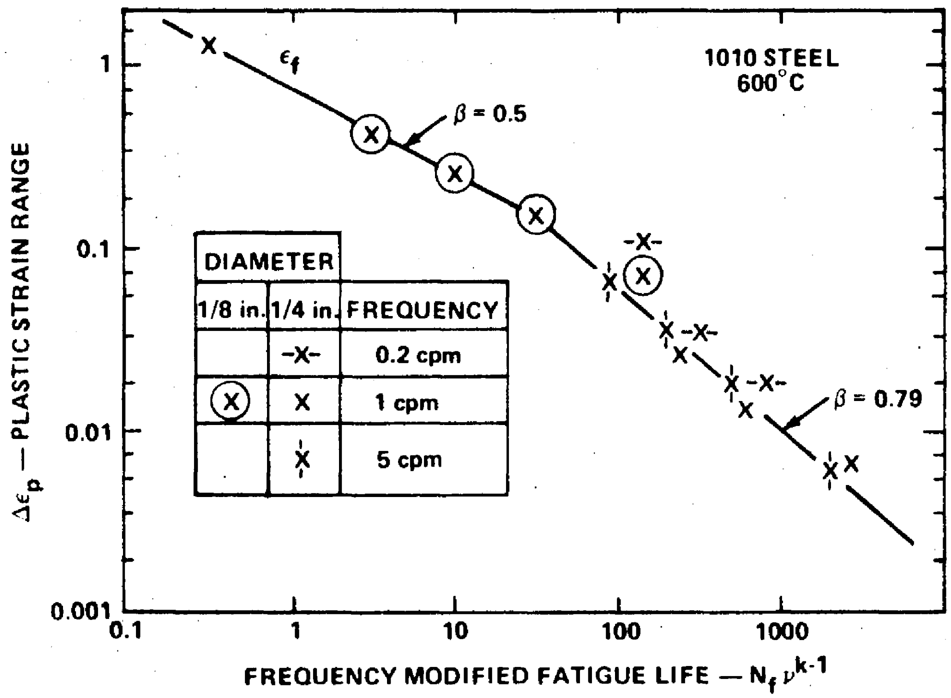


FIGURE E1-29. PLASTIC STRAIN RANGE VERSUS FREQUENCY MODIFIED FATIGUE LIFE — AISI C1010 STEEL, 600°C

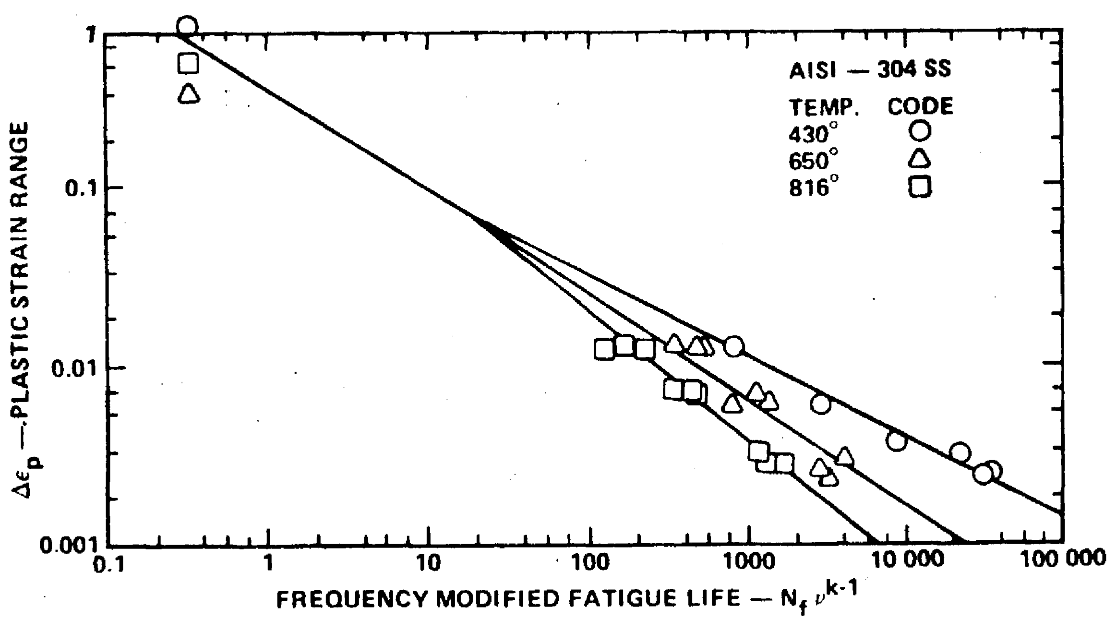


FIGURE E1-30. PLASTIC STRAIN RANGE VERSUS FREQUENCY MODIFIED FATIGUE LIFE — AISI 304 STAINLESS STEEL strains induced by cyclic thermal stresses. Failure under repetitive application of thermal stress has been termed thermal stress fatigue.

The principal activity in which analysts have, until recently, engaged whenever thermal-stress problems were encountered has been the computation of elastic stresses. The treatment of this subject may be found in Section D, Thermal Stresses. While such computations constitute a necessary and desirable first step in any practical analysis, they unfortunately do not provide sufficient information for a final evaluation when dealing with ductile materials. Thus, it is important not only to take into account plastic-flow effects that occur when the yield point is exceeded but to consider how such plastic flow might change during progressive thermal and load cycling of the material. Recent contributions have made a valuable start toward an understanding of this problem. Not only have new computational techniques been developed for taking into account inelastic effects such as creep and plastic flow, but a start has been made toward including cyclic effects both in computational procedure and in the interpretation in terms of material behavior. The failure criterion for brittle materials is, of course, different from the criterion for ductile materials. From an engineering point of view, the problem is to determine suitable working stresses.

1.4.3.1 Idealized Thermal-Cycle Model.

A model will be chosen in which all the thermal strain is converted to mechanical strain. The case is shown in Fig. E1-31 in which a bar is fixed at its ends between two immovable plates so that the length of the bar must remain constant. To approach the problem realistically, the model should show strain hardening, and the effects of stress relaxation that occurs during any hold period that may be imposed at high temperature.

Referring to Fig. E1-31, the bar is assumed clamped when hot, so that tensile stress is developed along OAF during the first cooling. Plastic flow is initiated at A, but the stress continues to increase to F because of strain hardening. At F the specimen is held for a dwell period, but no stress change occurs (relaxation), since the temperature is presumed to be low enough to make creep and anelasticity negligible. Upon reheating of the specimen, the stress-strain relation proceeds along FGE, yielding at G, at which time there is a much lower stress than there is at A because of the Bauschinger effect. (The Bauschinger effect states that plastic flow in one direction reduces the stress at which yielding will occur in the opposite direction.)

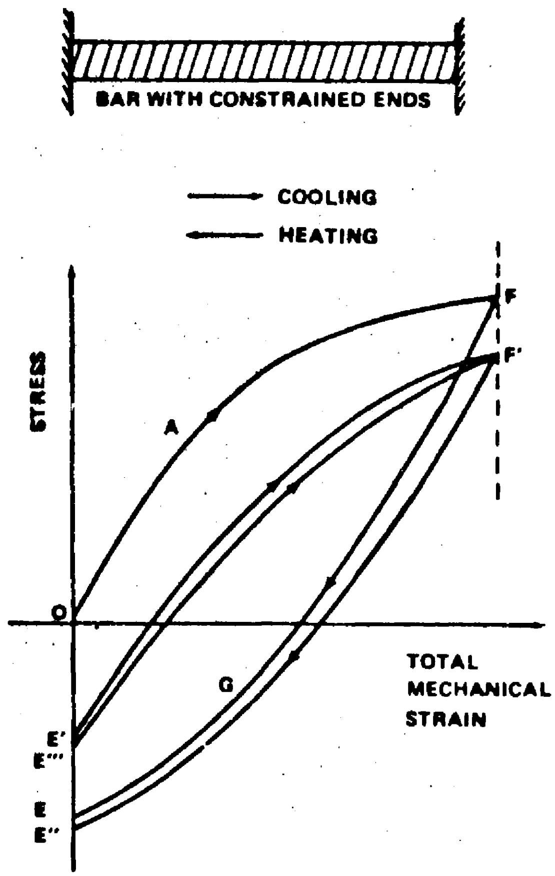


FIGURE E1-31. SCHEMATIC STRESS-STRAIN RELATIONSHIP FOR FIRST FEW CYCLES OF THERMAL CYCLING OF CONSTRAINED BAR

When the initial high temperature is restored, the state of the material is at point E, where a compressive elastic strain is necessary to offset the tensile plastic flow that occurred during AF and, thereby, return the specimen to a net strain of zero. This elastic strain introduces the stress OE. Any hold period at the high temperature and stress may convert this elastic strain to inelastic strain-creep and anelasticity, thereby reducing the stress. Thus, point E moves to point E' by the time the specimen starts to cool again. The cooling causes the path E' F' to be traversed. Reheating results in the path F' E''. The hold period at the high temperature converts e'' to E''', etc. After a few cycles, the stress-strain path may settle down to an essentially unchanging loop. For illustrative purposes this loop may be taken as E'''F'E''E'''. This corresponds to an asymptotic hysteresis loop and takes into account not only all the hardening or softening characteristics but also creep and anelasticity effects as well as effects due to changing temperature and metallurgy.

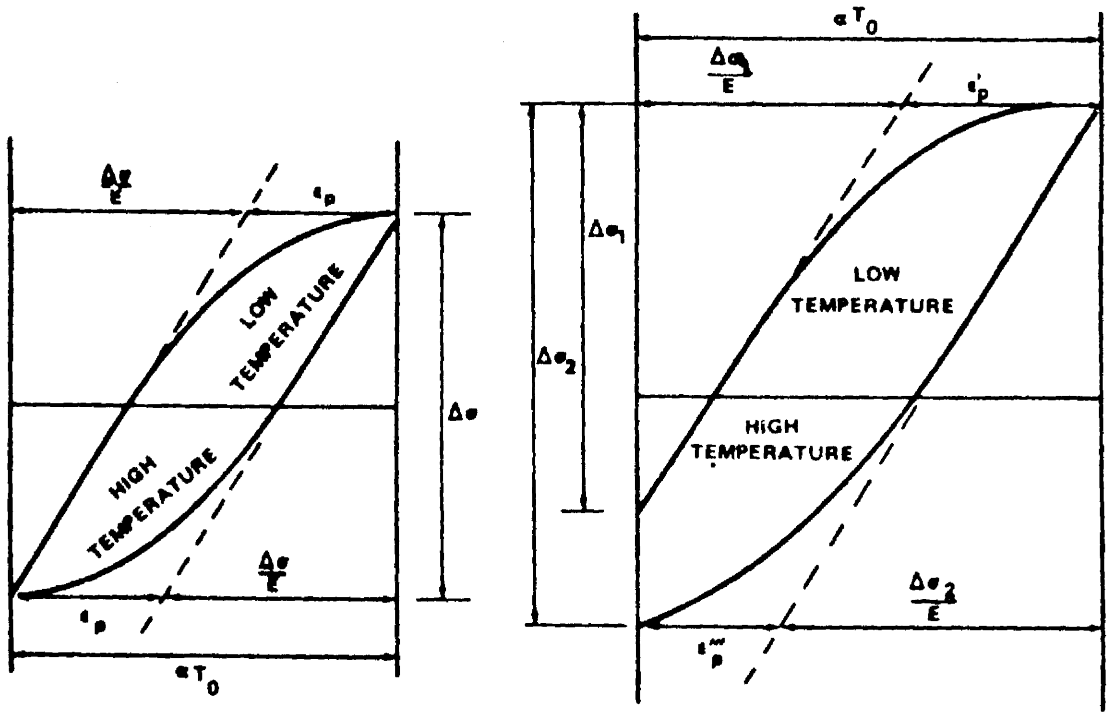
1.4.3.2 Effect of Creep.

In the experiments that have been performed to date, analysis of the thermal-stress fatigue behavior has been made largely by relating the plastic strain per cycle to the number of cycles to failure. Normally, the plastic strain is determined by subtracting the elastic-strain range from the total-strain range, which in turn is computed by dividing the stress range by the elastic modulus. When creep occurs during the high temperature portion of the cycle, this procedure produces inaccuracies. Not only is the plastic flow improperly computed if creep is neglected, but the omission of the creep can seriously affect the computation of cyclic life because the effect of creep strain on life differs appreciably from the effect of slip-type plastic flow.

Consider Fig. E1-32, which shows the stress-strain relationships for two idealized cases. In Fig. E1-32 (a) no creep is assumed to take place. Then,

$$\epsilon_p = \alpha T_0 - \frac{\Delta\sigma}{E}$$

where $\Delta\sigma$ is the stress range and E is the elastic modulus. (The elastic modulus is assumed to be constant over the entire temperature range;



(a) (b)

FIGURE E1-32. PLASTIC-STRAIN AND STRESS-RANGE RELATIONSHIP FOR (a) NO CREEP RELAXATION DURING HOLD PERIOD AND (b) CREEP RELAXATION DURING HOLD PERIOD assumption of a temperature-dependent elastic modulus complicates the problem even further.)

However, when creep occurs during a hold period [Fig. E1-32(b)], the plastic strain per cycle is not directly related to the total stress range. For the tensile portion of the cycle, the plastic strain ϵ_p' is

$$\epsilon_p' = \alpha T_0 - \frac{\Delta\sigma_1}{E}$$

where $\Delta\sigma_1$ is the stress range developed during the cooling portion of the cycle only. In the heating portion of the cycle, the stress range is larger, thus reducing the direct compressive plastic strain ϵ_p'''. However, the total plastic strain ϵ_p'' also includes the creep ϵ_p''' during the hold period. This creep strain replaces the elastic strain relaxed during the creep and is, therefore, equal to $(\Delta\sigma_2 - \Delta\sigma_1)/E$. Thus,

$$\epsilon_p'' = \epsilon_p''' + \epsilon_p'''' = \alpha T_0 - \frac{\Delta\sigma_2}{E} + \frac{\Delta\sigma_2 - \Delta\sigma_1}{E} = \alpha T_0 - \frac{\Delta\sigma_1}{E} = \epsilon_p' \quad .$$

Therefore, the plastic strain in compression is equal to that in tension, but these strains are not directly related to the total stress range.

1.4.3.3 Comparison of Thermal-Stress Fatigue with Mechanical Fatigue at Constant Temperature.

Several experiments in which mechanical fatigue at various temperatures was compared with thermal-stress fatigue have been conducted (Refs. 13 and 14). Fatigue tests were conducted at several constant high temperatures by loading mechanically, and the results of the tests were compared with thermal-stress fatigue tests in which the temperature naturally varied over an appreciable range. The results of a test are shown in Fig. E1-33 and are analyzed in terms of cyclic life at a given measured plastic strain. The mechanical-fatigue tests are shown at 350°, 500°, and 600° C, whereas in the thermal-stress tests the specimen was completely constrained and cycled between 200° and 500° C, so that the mean temperature was 350° C. As can be seen from the figure, at equal values of cyclic plastic strain the number of cycles to failure was much less for the thermally cycled specimen than for one mechanically cycled at 350° C and even for one mechanically cycled at 600° C, even though no part of the specimen in the 200° to 500° C thermal-stress

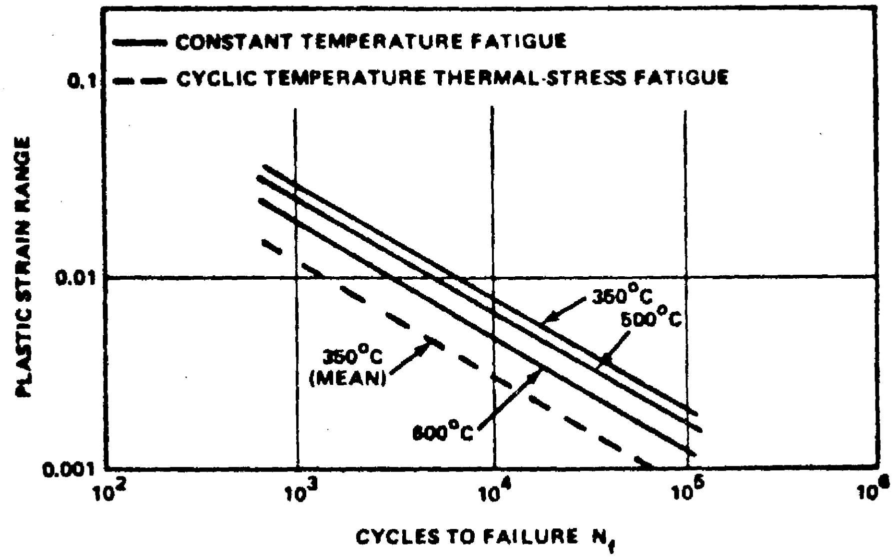


FIGURE E1-33. CYCLES TO FAILURE IN THERMAL-STRESS FATIGUE COMPARED WITH CYCLES AT CONSTANT TEMPERATURE IN SIMILAR PLASTIC STRAIN RANGE

fatigue test ever reached 600° C. Reasons for this discrepancy have been generally attributed to nonuniformities of temperature along the specimen and the sensitivity of the properties of the materials to temperature in the range traversed by the thermal-cycling tests.

For this reason it would appear that, until further tests clarify the problem, the temperature for the mechanical-fatigue tests should be taken as the highest temperature of the thermal-stress test. In this way, metallurgical phenomena will not be overlooked and pessimistically low values of life will be indicated; this appears to be necessary on the basis of most tests conducted to date.

Spera (Ref. 15) has presented a method for calculation of thermal-fatigue life based on accumulated creep damage. He proposes that thermal-fatigue life can be determined from the basic mechanical properties of a material by calculating lines for each of two distinct and independent failure modes: (1) cyclic creep-rupture, using a modification of the well-known life-fraction role proposed by Robinson and Taira, and (2) conventional low-cycle fatigue, using the empirical equations of the Method of Universal Slopes developed by Manson (Refs. 9 and 10). Equations are presented in sufficient detail to define completely the analytical procedure.

1.4.3.4 Summary.

When plastic strains are introduced by constraint of thermal expansion, fatigue ultimately results. The number of cycles that can be withstood depends on the plastic strain and the temperatures at which these strains are induced. Whether tension or compression occurs at the high temperature seems to have little effect on fatigue life. Probably the most important single variable is the maximum temperature of the cycle, particularly if it is high enough to cause metallurgical effects to take place. Increasing the maximum temperature for a given temperature range will generally cause a much greater reduction in cycles to failure than increasing the temperature range by the same amount and maintaining the same maximum temperature. The time at which the maximum temperature is maintained can also have an effect on fatigue life, but of smaller magnitude. The effect can be beneficial or detrimental to fatigue life, depending on the material, temperature, and life range.

Limited data indicate that fatigue life in a thermal-stress fatigue test is sometimes considerably less than the fatigue life of a mechanically strain-cycled specimen having the same total strain as the thermally fatigued specimen and tested at a constant temperature equal to the average temperature of the thermal-stress fatigue test.

The effect of prior thermal cycling on specimens subsequently evaluated for stress rupture depends greatly on the material and the number of prior cycles. In general, the effect can be expected to be detrimental; however, in some cases it may actually be beneficial.

Each of the materials that has been studied to date — type 347 stainless steel, the cobalt-base alloy S-816, and the nickel-base alloys Inconel and Inconel 550 — displays characteristics considerably different from the others in thermal-stress fatigue, and prediction of distinctions in the behavior of one from that of another is difficult. Hence, more classes of materials must be investigated before a reasonably complete picture of the thermal-fatigue behavior of materials, in general, can be understood. For specific information on behavior of the various types of materials, investigate Refs. 13, 14, 16, 17, 18, 19, and 20.