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THE FATIGUE PROPERTIES OF MATERIALS

E. G. EELES AND R. C. A. THURSTON

PHYSICAL METALLURGY DIVISION

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THE FATIGUE PROPERTIES OF MATERIALS

by

E.G.Eeles* and R.C.A.Thurston**

ABSTRACT

Fatigue is the process responsible for the majority of service failures of engineering components and structures, and consequently its many ramifications have been intensively studied by physicists, metallurgists and engineers. The process consists essentially of two separate stages, crack initiation and crack propagation, which are affected differently by external variables. The mechanisms responsible for the development of these two stages, the parameters which control them, and the empirical relationships derived for the use of design engineers, are discussed on the basis of the most recent available information. Attention is drawn to those areas where further fundamental or applied research is required, and some of the limitations of existing theories are mentioned.

The fatigue properties of ultra-high-strength steels, aluminum alloys, titanium alloys, and composite materials (both metallic and non-metallic), are presented and discussed in terms of the foregoing review. In particular, the effects of the environment are examined in the light of the theme of the International Symposium on MATERIALS, KEY TO EFFECTIVE USE OF THE SEA, held in New York, N.Y., September 12-14, 1967.

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Direction des mines

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LA FATIQUE DES MATÉRIAUX

par

E. G. Eeles* et R. C. A. Thurston**

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RÉSUMÉ

La fatigue est la principale cause de rupture des matériaux et des charpentes. C'est pourquoi les ingénieurs, les physiciens et les métallurgistes ont minutieusement étudié les nombreuses manifestations de ce phénomène. Le déroulement de ce processus comprend deux principales étapes distinctes que les facteurs extérieurs influencent de façon différente: la naissance et la propagation des fissures. Cette étude se fonde sur les données les plus récentes et porte sur l'évolution de ces deux stades, les paramètres qui les régissent et les conséquences pratiques qu'on a étudiées à l'intention des concepteurs. On met l'accent sur les domaines qui nécessitent de plus amples recherches théoriques et expérimentales et l'on fait état de certaines des limites des théories actuelles.

Cette étude présente et analyse la fatigue des aciers à très grande résistance, des alliages d'aluminium, des alliages de titane et des matériaux mixtes (composés de matériaux métalliques et autres). Une place est réservée à l'étude de l'influence du milieu, à la lumière du thème du symposium international des Matériaux, Clé de l'Utilisation rationnelle des Océans, qui a eu lieu à New-York, du 12 au 14 septembre 1967.

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CONTENTS

LIST OF SYMBOLS

a	-	material constant
A_o	-	initial area
A_f	-	fracture area
A	-	material constant (crack growth rate coefficient)
C	-	experimental constant
Δe	-	nominal strain range
Δ_ϵ	-	local strain range at notch root
Δe_p	-	plastic strain range
e_f	-	fracture ductility in tension test
E	-	modulus of elasticity
K	-	Irwin's stress intensity factor
K_σ	-	stress concentration factor, $\Delta\sigma/\Delta S$
K_ϵ	-	strain concentration factor, $\Delta_\epsilon/\Delta e$
K_t	-	theoretical stress concentration factor
K_f	-	fatigue strength reduction or notch factor
k	-	numerical exponent
l	-	half crack length
m, n	-	numerical exponents
N	-	number of cycles
N_f	-	number of cycles to failure
n_i	-	number of cycles applied at stress "i"
N_i	-	number of cycles to failure at stress "i"
q	-	notch sensitivity in fatigue
r	-	notch root radius
σ	-	applied stress
σ_n	-	ultimate tensile strength
$\Delta\sigma$	-	local stress range at notch root
ΔS	-	nominal stress range

FIGURES

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1. INTRODUCTION

Fatigue is defined by the ASTM as the process of progressive localized permanent structural change occurring in a material subjected to conditions which produce fluctuating stresses and strains at some point or points and which may culminate in cracks or complete fracture after a sufficient number of fluctuations. It is acknowledged as being responsible for the majority of service failures and, since the early experiments of Wohler, has been the subject of intensive research. For the past few years the technical literature on fatigue has been increasing at the rate of about 300 references per year, and the aspects covered have ranged from fundamental studies of the behaviour of dislocations in single crystals of high-purity material to full-scale investigations of the effect of service loading on aircraft structures. As a result of the continuing efforts of research scientists and engineers throughout the world, there is now a much better understanding of the nature of the mechanism of the fatigue process and of the effect of the numerous operational factors which influence it, although this knowledge tends to be more qualitative than quantitative. It would be fair, therefore, to say that the state of the art is making reasonable progress but is still a long way from enabling an accurate prediction of the endurance or life of an engineering structure under given service loading and environmental conditions.

The complexity of the fatigue phenomenon is due in part to the fact that progressive fracture is essentially a sequence of at least two processes - crack initiation and crack propagation - which may be controlled by two different sets of criteria. Furthermore, these processes are inherently statistical in nature. The deformation on the microscale can be expected to be erratic or intermittent, whether the observations be those related to dislocation movements, slip bands, or vacancies, etc. The overall or gross behaviour of a macroscopic zone thus becomes an integrated effect of the progressive changes in local regions (Dolan, 1964). The models in Figure 1 are intended to illustrate the fields of observation of those studying the fatigue strength characteristics of materials, namely the solid-state physicist, the metallurgist and the materials engineer, respectively. The physicist may hypothesize from the dislocation model (No.1) the effects of vacancies or interstitial atoms, and even predict a pile-up of dislocations, but the presence of grain boundaries, foreign atoms and alloy constituents in the metallurgical model (No.2) will significantly modify the behaviour of the material. Similarly, the predictions of the metallurgist may be influenced to a large extent in the engineering model (No.3) by heterogeneous yielding across a section, rolling texture, or the presence of non-metallics, not to mention specimen preparation and surface effects.

The materials engineer makes his observations generally on machined and polished test-pieces of commercial materials, each containing a random distribution of "weak links" induced by such factors as inclusions, difference in grain size, anisotropy, spacing of hardening particles, residual microstresses, etc. Since fatigue is essentially a localized phenomenon, the endurance at a given stress will vary from specimen to specimen, and should be studied on a statistical basis. At low stresses, the number and type of weak links affected will normally vary widely between specimens and the scatter will be large. At high stresses, on the other hand, the scatter will decrease. Figure 2 shows a set of typical S/N curves for various probabilities of failure obtained from bending-fatigue tests on 7075-T6 aluminum alloy (Sinclair and Dolan, 1953). The apparent loss in fatigue strength of large metal sections, often referred to as the size or scale effect, is caused in part by the greater probability of occurrence of a weak link in a large volume of metal than in a small volume, other things being equal.

From an engineering viewpoint, it would be most useful if the fatigue strength of a material were uniquely related to some more readily determinable property, such as the ultimate tensile strength. Regrettably, such is not the case, although for most groups of materials a general trend exists. In steels, the fatigue strength is about 50% of the ultimate tensile strength up to tensile values of about 180 ksi, but above this level the fatigue strength remains sensibly constant. Similarly, for aluminum alloys the ratio is about 35% up to a tensile strength of 50 ksi and then levels off.

The situation is further complicated by the large number of processing and engineering variables which influence the fatigue characteristics of a material and to some of which reference has already been made. Grain size is an important factor for materials in general, and has an inverse relationship to the fatigue strength. Changes in microstructure may not affect the fatigue strength and the tensile strength to the same degree. For example, in steels the ratio of the former to the latter is significantly higher for ferrite and tempered martensite than it is for pearlite or untempered martensite (Waisman, 1959). Wrought materials show directionality in fatigue strength as well as in their other mechanical properties. In addition, there is the effect of the ever-present inclusions, which will be discussed in more detail with reference to the ultra-high-strength steels.

Engineering parameters, which may or may not be controllable but which must be considered, are the surface finish, geometry, stressing conditions (type, mean stress, residual stress), speed of cycling, ambient temperature, and environment. It is an unfortunate but accepted fact that most of these parameters are inter-related. For example, commercial methods of finishing components, such as milling, shaping, turning, grinding and honing, result in different degrees of cold work, surface roughness, and residual stress in the surface. The effect of these on the fatigue behaviour will vary with the material and its hardness. In general, a rough surface will have a more deleterious effect on a hard material than on a soft material. However, shot-peening results in both a rough surface and beneficial residual compressive stresses, and the latter more than compensate for the former and can be particularly effective on hard steels. Surface roughness may be regarded as a mild form of engineering notch or geometric discontinuity, e.g. holes, fillets, keyways, screw threads, etc. The theoretical stress concentration factor (K_t), which is the ratio of the elastic stress at the discontinuity to the nominal stress, is seldom fully realized in fatigue due to local plasticity effects. The deviation between K_t and the experimental value (K_f), which is the ratio of the unnotched fatigue strength to the notched fatigue strength, is therefore used as a measure of the notch sensitivity (q) of the material:

$$q = \frac{K_f - 1}{K_t - 1} \quad . . . . (1)$$

Obviously, for those materials which are fully notch-sensitive ($K_f = K_t$), $q = 1$; and, for insensitive materials ($K_f = 1$), $q = 0$.

The effect of frequency or cycling speed is also complicated by a number of auxiliary time-dependent effects. As the frequency increases, heating due to mechanical hysteresis will tend to increase and the time the specimen is allowed to dwell at low stresses will decrease; furthermore, atmospheric corrosion will be less effective. As a general comment, it may be said that the fatigue strength of structural materials tends to increase with frequency above a speed of about 5000 rpm.

The foregoing is intended to present an introduction to the nature of fatigue and to give some idea of the complexity of the phenomenon. In subsequent Sections, modern concepts of the more important metallurgical and engineering aspects of the subject are reviewed in the light of the theme of the Symposium, and finally, data on the more promising light alloys, ultra-high-strength steels, reinforced plastics and composites are discussed.

2. METALLURGICAL ASPECTS

It was recognized as long ago as 1903 (Ewing and Humphrey) that alternating stresses produced structural effects different from those associated with simple uni-directional loading. Although normal slip lines are noted initially, certain of these progressively thicken into coarse slip bands, in which fatigue cracks finally appear. With the development of dislocation theory, the attention of many workers has been drawn towards a satisfactory understanding of the processes taking place.

The coarsening of certain slip lines with continued cyclic deformation is apparently a uniform characteristic, becoming more pronounced as the cyclic amplitude is reduced (Wood, 1959). The formation of these so-called persistent slip bands continues over a prolonged period in spite of the fact that cyclic strain-hardening reaches a saturation value after about 1% of the total fatigue life (Alden and Backofen, 1961; Coffin and Tavernelli, 1959). The dislocation distributions formed by cyclic stressing have been examined extensively by transmission electron microscopy, and in general the dislocation distributions, as for example Figure 3 (Laufer and Roberts, 1966), are characteristic not only of the strain amplitude but also of the intrinsic stacking fault energy of the metal (Ham, 1966). This latter property is related to the ease of cross-slipping of screw dislocations, a phenomenon which is of considerable importance in interpreting cyclicly produced dislocation patterns. The wide variations in dislocation distribution that have been noted are clearly the reason why a simple cumulative damage law, such as that proposed by Palmgren and Miner (1924:1945), is not capable of accurately assessing the effective fatigue damage due to variable load spectra.

The mechanism by which a fatigue crack initiates from a persistent slip band is still not unequivocally established, even for the well-studied f.c.c. metals, but it is generally agreed that the crack develops at a suitable topographical notch at the surface. Although a simple model was proposed (Wood, 1958) for such a process, the widespread observation of surface intrusions and extrusions, probably involving some form of cross-slip, suggests that these phenomena are an important precursor of the formation of a fatigue crack (Low, 1963). The possible mechanisms responsible for these topographic features have been the subject of much discussion (Cottrell and Hull, 1957), which has not satisfactorily accounted for all of the phenomena.

The free surface of the material is of considerable importance, as it has been shown that periodic removal of as little as 20μ from the surface will indefinitely increase the low amplitude fatigue life of copper (Thompson et al., 1956). This phenomenon also extends to the field of commercial materials. Although a thick anodic film was found (Alden and Backofen, 1961) to prolong the life of pure aluminum, work by the authors (Eeles and Thurston; Eeles, 1967) has shown that the endurance of a commercial aluminum alloy is controlled by the properties of the thin adherent surface oxide layer.

Once a fatigue crack has been initiated in a material, further cyclic stressing will normally propagate the crack through the section until final rupture occurs. Propagation takes place in two stages (Forsyth, 1963). In so-called Stage I, cracking occurs on a plane at 45° to the tensile axis at a rate of the order of angstroms per cycle. It has been proposed (Grosskreutz, 1962) that this stage of propagation takes place by continuous re-initiation of the crack within slip bands ahead of the advancing crack, but it has recently been argued (Laird, 1966) that Stage I takes place by a process similar to that for Stage II, namely a continuous relaxation and resharpening of the crack tip during successive tensile and compressive parts of the stress cycle (Figure 4). In Stage II the rate of advance of the crack front is much greater, of the order of microns per cycle. In f.c.c. materials the crack propagation rate is substantially controlled by the stacking fault energy (Miller et al., 1966), with lower rates in materials of low energy being attributed to increased difficulty in forming a suitable substructure to facilitate cracking at sub-grain boundaries.

The progressive nature of the advancing crack front produces the characteristic macroscopic appearance of a fatigue crack. On a microscopic scale, there is strong evidence that each load cycle is shown by a separate striation on the fracture face. The spacing of the striations depends on the load and, with spectrum loading, the cumulative effects of the individual cycles at the various loads can clearly be seen, as shown in Figure 5 (Weibe, 1966). A consequence of this observation has been the rapid development of microfractography into one of the most useful techniques for the study and interpretation of service failures.

It has already been pointed out that the fatigue strength is not in any fixed proportion to the ultimate strength characteristic of the material. In addition, certain materials, virtually all ferrous based, exhibit a sharp limiting stress below which, under normal conditions, failure will not take place. There are several divergent opinions as to the reasons for this limiting value and it appears (Adair and Lipsitt, 1966) that no one mechanism is responsible for the complete fatigue limit behaviour. It probably originates as a basic structural property of the b.c.c. lattice, with its characteristics controlled by the behaviour of interstitial atoms.

With the knowledge that fatigue takes place by localized structural processes, it can readily be understood why there are often substantial changes in endurance with only minor changes in metallurgical characteristics. As the mechanical properties of many materials are determined by the precipitation or dispersion of secondary phases, fatigue and other deformation-dependent characteristics are related to the ease with which dislocations can move past these obstructions in the crystal lattice (Nicholson et al., 1959). Metallurgists have now made considerable strides in tailoring crystal structures for specific high-strength requirements (Davies, 1965), and it is to be hoped that comparable developments will be possible for improved fatigue properties.

3. ENGINEERING ASPECTS

An efficient structural design depends on the continuing integrity of the individual component parts of the assembly. Due to the progressive nature of fatigue, it is often desirable that cyclicly stressed designs should allow for the possible formation and partial or complete propagation of a crack. This requirement has led to the fail-safe design philosophy (or multiple load path approach) in aircraft, and the enforcement of rigid inspection and component-replacement schedules. Such practices are only possible when research has given the design engineer a reasonable understanding of the mechanics of crack propagation and of the effects of the more important parameters. In the following discussion, it should be borne in mind that the data have been established by laboratory tests on samples of controlled geometry. The relations derived cannot, in general, be used for service applications without considerable approximation.

The fatigue process, from an engineering viewpoint, can be loosely divided into the stages - crack initiation, crack propagation, and final fracture - assuming cycling is continued thus far. The initiation stage is sensitive to localized surface effects and environment, and is responsible for the majority of the scatter associated with fatigue testing. The propagation stage, from some arbitrary initial size, is generally much more consistent and has been shown to be amenable to mathematical analysis. When the crack reaches the critical size, determined by the fracture toughness of the material, failure occurs. In notched specimens, almost all of the observed fatigue life is occupied by the crack propagation stage. In smooth specimens an "engineering" crack of the order of 0.002 to 0.003 in. deep is normally not observed until quite late in life, particularly at the lower stresses, hence crack propagation represents a much smaller percentage. In passing, it should be mentioned that non-propagating cracks may occur in materials with very sharp notches in constant stress amplitude tests.

The crack propagation stage has been intensively studied during the past decade, mainly in sheet materials, and numerous laws have been proposed. It is not too surprising that most of these expressions for crack growth (dl/dN) can be reduced to the form

$$\frac{dl}{dN} = A\sigma^{m}l^{n} \quad , \qquad \ldots\ldots\ldots \quad (2)$$

where l = half crack length,

σ = applied stress,

m and n = numerical exponents,

A = material constant.

Frost and Dugdale (1958) adopted values of 3 and 1 for the exponents "m" and "n" respectively, but restricted the use of their expression to cracks not exceeding about one-eighth of the specimen width. Liu (1962) agreed with their value for "n", but argued that the stress exponent should be 2. The expressions proposed by McEvily and Illg (1958), and by Paris (1957), suggest that crack growth is proportional to some function of "K", Irwin's stress intensification factor used in fracture mechanics theory. For the case of a notched wide sheet in tension, Paris's expression can be reduced to that of Equation 2, with exponents of 4 and 2, respectively, for the stress and crack length (Paris and Erdogan, 1963). Each expression has been supported by experimental data, mainly obtained from tests on aluminum alloy sheet in view of the aircraft structural problem, but there appears to be increasing evidence in agreement with

Paris and Erdogan's equation as shown in Figure 6. More recently (Forman et al., 1966), the expression has been modified to allow for variations in the load ratio (minimum load/maximum load) and for instability when K approaches the fracture toughness of the material; very good agreement was found for two aluminum alloys.

The effects of several parameters on crack growth in sheet have been studied; in particular, specimen width, test frequency, temperature, and stressing conditions. In tests under constant net stress, the crack propagation rate appears to be proportional to specimen width, implying a constant time of propagation (Weibull, 1963). Investigations of the effect of test frequency have been made over the range 1/3 to 2200 cpm, again on aluminum alloys, at 20°C and at 150°C (Lachenaud, 1965; Schijve and Rijk, 1965). A significant decrease in crack propagation rate at 20°C was observed as the frequency was increased over this range, although the effect tended to disappear at longer crack lengths. A similar decrease was reported when the temperature was raised from 20°C to 150°C.

Service stressing conditions, unfortunately, are seldom those of constant, uniaxial loading, and the consequent effects on crack propagation become somewhat involved. For example, consider the simple case of a single change in load level (Figure 7) (Hardrath, 1963). When a low stress is changed to a high stress, the rate of crack propagation changes accordingly to the expected rate for that stress. However, if the order is reversed, the crack normally remains dormant for a period dependent on the higher stress and the stress difference. Once it re-commences to grow, the rate again is that which would be expected. Random loading is beyond the scope of this review, and reference should be made to Swanson (1967) for a 'state of the art' survey.

The area of high stress or low cycle fatigue, involving cyclic plastic strain ranges of the order of 1% and failure within about 10^4 cycles, has aroused considerable attention, particularly in the pressure-vessel field for which submersibles are a special case. Manson (1953) and Coffin (1954) independently derived the empirical relationship

$$N_f^k \Delta e_p = C \quad , \qquad \ldots\ldots\ldots \quad (3)$$

where N_f = number of cycles to failure,

Δe_p = plastic component of strain amplitude,

k and C = experimental constants.

This expression was subsequently given a theoretical basis by Grosskreutz (1964), whose analysis was based on the cyclic deformation of a 2-dimensional elliptical crack under plane strain, and by Gillis (1966) from considerations of dynamic dislocation processes. The Manson-Coffin law has been extremely well documented for a wide range of metals and alloys. Typical results are shown in the log-log plot of Figure 8 (Coffin, 1962). The slope of the lines, hence the value of the exponent "k", is approximately $\frac{1}{2}$. Since the fracture ductility in simple tension, "e_f", can be considered to represent Δe_p at $N = \frac{1}{4}$, and is defined as

$$\log_e \frac{A_o}{A_f} \quad \frac{\text{(initial cross-section area)}}{\text{fracture area}},$$

$$\text{hence,} \quad N_f^{\frac{1}{2}} \Delta e_p = \frac{e_f}{2} \qquad \ldots\ldots\ldots\ldots\ldots \quad (4)$$

Thus, it is possible in most cases to predict low cycle fatigue behaviour from a knowledge of the fracture ductility of a metal, provided the test temperature is sufficiently low that time-dependent effects do not occur to any appreciable extent. Higher values for the exponent of "N_f" have been reported, notably for nickel (Benham, 1961), and other values, $e_f/\sqrt{2}$ (Martin, 1961) and $0.8e_f^{.75}$ (Manson, 1962), have been proposed for "C".

In engineering practice, however, it is usually the total strain range that is known, rather than the plastic strain range. More recently, Manson (1965,1966) has utilized this parameter (Δ e) in an empirical expression developed to represent fatigue behaviour, based on his method of universal slopes (Figure 9):

$$\Delta e = 3.5 \frac{\sigma_n}{E} N_f^{-0.12} + D^{0.6} N_f^{-0.6}, \qquad \ldots \quad (5)$$

where σ_n = ultimate tensile strength, psi,

E = modulus of elasticity, psi,

N_f = number of cycles to fracture,

$D = e_f$.

It is assumed that the slopes of both the plastic and the elastic lines are the same on a log-log plot for all materials, and a comprehensive survey of 29 metals and alloys resulted in the values -0.6 and -0.12, respectively, given in the above equation. The intercept of the plastic line at $N_f = 1$ is $D^{0.6}$, and that of the elastic line is $3.5\sigma_n/E$, both of which can be determined from a tensile test. The validity of this approach was checked for a variety of materials, some of the comparisons being shown in Figure 10, from which it will be seen that the agreement is quite good. The discrepancy between the life exponent in Equation 5 and that in Equation 4 may be explained on the basis that the former refers to complete fracture and the latter to cracking.

The foregoing remarks apply essentially to small, smooth, laboratory-type specimens, normally used for evaluating the fatigue properties of materials. In practice, some form of stress-raiser or notch is generally present, and it is desirable to know the fatigue characteristics of the material under such conditions. One method, proposed recently by Topper and his co-workers (1967), is based on Neuber's rule (1961),

$$K_t = (K_\sigma K_\varepsilon)^{\frac{1}{2}} \quad , \qquad \ldots\ldots\ldots\ldots \quad (6)$$

where K_t = theoretical stress concentration factor,

K_σ = stress concentration factor $\Delta\sigma/\Delta S$,

K_ε = strain concentration factor $\Delta\varepsilon/\Delta e$,

σ, S = stress at notch root and nominal stress, respectively,

ε, e = strain at notch root and nominal strain, respectively,

and Peterson's (1959) expression for the fatigue strength reduction factor,

$$K_f = 1 + \frac{K_t - 1}{1 + \frac{a}{r}} \quad , \qquad \ldots\ldots \quad (7)$$

where r = notch root radius,

a = material constant.

The following relationship is then derived:

$$K_f(\Delta S \Delta eE)^{\frac{1}{2}} = (\Delta\sigma \Delta\varepsilon\, E)^{\frac{1}{2}}, \qquad \ldots\ldots \quad (8)$$

which can be interpreted as furnishing indices of equal fatigue damage in notched and unnotched specimens. In constant-amplitude, zero-mean-stress tests, if $K_f(\Delta S \Delta eE)^{\frac{1}{2}}$ for the notched specimen equals $(\Delta\sigma \Delta\varepsilon\, E)^{\frac{1}{2}}$ for the smooth specimen, they will form detectable cracks at the same life. Hence, a master life plot--such as Figure 11, obtained from smooth-specimen data--can be used to estimate the fatigue life of notched members. Illustrative results for 7075 aluminum alloy, using this procedure, are shown in Figure 12, and the agreement will be seen to be very good. The method, however, is limited to cases where the crack propagation stage is negligible and where there is no mean stress.

One further factor, to which passing reference has been made, is the non-constancy of the stress or strain in most engineering applications. The most widely known of the simpler procedures for evaluating the effect of fatigue behaviour is the Palmgren-Miner (1924, 1945) linear cumulative damage hypothesis:

$$\sum \frac{n_i}{N_i} = 1, \qquad \ldots\ldots\ldots\ldots \quad (9)$$

where n_i = number of cycles applied at load "i",

N_i = number of cycles to failure at load "i".

As a first approximation, the hypothesis has the merit of simplicity, but it does not take into account the effect of the order in which loads are applied, the effect of stresses below the initial fatigue limit of the material, or the "coaxing" effects present in some strain-ageing materials. Grover (1960) suggested that the reliability of the method might be improved by using separate linear damage rules for crack initiation and crack propagation, and Manson and his colleagues have attempted to do this (1964, 1966). Their approach was based on simple expressions derived from limited data for representing these two stages in terms of total life, and was intended to correct the deficiencies associated with the order of loading. The experimental data showed some improvements in life prediction as compared with the use of the conventional linear damage rule, but indicated that the crack propagation period is not uniquely related to total life, and that the cyclic hardening and softening characteristics of a material must be taken into account.

4. ENVIRONMENTAL ASPECTS

Many of the phenomena observed with fatigue are associated with the surface of the material, and it would, therefore, be very surprising if the environment did not have equally observable effects. It is now fifty years since first reference was made to a specific environmental effect (Haigh, 1917), and since this date many papers on the subject have been published (Gilbert, 1956).

Environmental effects can, with considerable simplification, be divided into two groups, according to the nature of the environment. The first is where the environment, liquid or gaseous, has a marked corrosive action on the subject material, while the second covers those instances where the environment is nominally non-reactive. To these groups should be added the phenomenon of fretting, or fretting corrosion. Fretting is not strictly an environmental effect in the same context as the two groups defined, but it should nevertheless be considered simultaneously as many of its ramifications are comparable to those of the more clearly defined environmental effects. Although distinction has been made between corrosive and non-corrosive environments, Gough and Sopwith showed, in 1932, that normal laboratory air itself was behaving as a mild corrosive, as fatigue tests carried out in vacuum showed significantly increased endurances. Hence the distinction between the two groups is one of degree only.

Fatigue in an aggressive environment, commonly termed corrosion fatigue, is most important as there are many engineering applications, particularly in marine environments, where cyclicly stressed components are exposed to corrosive conditions. Direct corrosion is a consequence of electrochemical differences between different metals or different microconstituents, and these electrochemical factors are equally applicable for corrosion concurrent with fatigue. Moreoever, areas of localized distortion in the crystal lattice have electrode potentials differing from those of the virgin lattice and are thus potential sites for corrosive attack. Hence, in many instances the corrosive attack is accelerated by stressing and cannot be considered as a simple additive mechanism.

A limiting fatigue stress is normally found for most ferrous materials, but under corrosive conditions the fatigue life curve continues to fall; see Figure 13 (Hara, 1956). Hence, any so-called corrosion fatigue limit is meaningless unless the endurance is specified. As corrosion is a time-dependent process, the elapsed time of any corrosion fatigue test is of significance, and concomitantly the endurance figures quoted are strongly dependent on the test frequency, in contrast to the case for normal non-corrosive fatigue tests.

Due to the very demanding aspects of the problem, there can be no simple approach in selecting material for a specific application. There has been a considerable amount of information documented in the literature (Gilbert, 1956), but due to the aforementioned frequency dependence, much of the data can only be related directly to very limited service conditions. However, certain general qualitative features are apparent.

One feature, which strikingly illustrates the serious nature of corrosion fatigue, is the drastic reduction in general strength levels, particularly for high-strength materials. Reference has been made (Gilbert, 1956) to many instances where the endurance limit has been decreased by a factor of two for low-strength steels to a factor of nearly five in the quenched-and-tempered condition. It is also a generalization that the nominally corrosion-resistant materials will likewise suffer some reduction in strength, although the proportionate reduction is not as great as with corrodible materials. With materials such as stainless steels, which obtain their corrosion resistance from a protective oxide film, the reduction of fatigue strength in corrosive environments is undoubtedly related to stress-induced damage to the protective oxide.

The prevention of corrosion fatigue is essentially that of preventing corrosive attack. If corrosion-resistant materials cannot be used, suitable protective measures should be employed. The simplest procedure is to prevent access of corrosive to the surface by suitable coatings. However, there are many applications where such coatings will be damaged in service and recourse must be made to one of two forms of electrochemical protection. The first of these is the use of coatings composed of or containing a sacrificial constituent less noble electrochemically than the metal to be protected, such as galvanizing on steel. The other method is direct cathodic protection by the application of a D.C. potential to the component, thereby ennobling the material in respect to a sacrificial anode, frequently magnesium.

Additionally, although tensile stressing tends to accelerate the corrosive effect, a compressive stress may, in fact, have the opposite effect (Harris, 1961) and thus any treatment to induce high residual compressive stress in the surface (e.g. shot peening) may be of substantial benefit.

In contrast to direct corrosive effects, the effects due to air or other non-corrosive environments are not as substantial, although they are of sufficient magnitude to be of concern (Gough and Sopwith, 1932; Eeles and Thurston, 1967).

For those materials which are affected, the active atmospheric constituents are oxygen and water vapour, either alone or in combination (Broom and Nicholson, 1961). The overall mechanism by which the environment affects the fatigue process has not been established. A detailed involvement with crack propagation has been documented (Hartman, 1965) and it is clear that crack propagation is influenced by the prevention of partial rebonding of the crack tip during the compressive part of the stress cycle. The crack initiation processes cannot be substantially different from those proposed and outlined previously. Hence, it is anticipated that the environmental involvement is an ancillary process. Broom and Nicholson (1961) proposed a mechanism, for aluminum alloys, by which moisture reacted with the bare metal, exposed by a fresh crack formed in the surface oxide film. They suggested that the hydrogen so formed diffused into the lattice, influencing in some way the dislocation processes taking place. Although the evolution of a gas, probably hydrogen, has been noted (Bennett, 1964), the environmental effect has been shown (Eeles and Thurston, 1967) to take place in the surface oxide film by a process not yet established.

Fretting, or fretting corrosion, is basically a fundamental result of the contact of two surfaces. It is a particular form of surface damage taking place when there is a slight oscillatory relative movement between the two surfaces. It is thus very common in structural applications involving vibration or other forms of cyclic loading. It is unfortunate that the effect of fretting on fatigue is comparable to that of either a sharp notch or an active corrosive and is thus a serious service problem. It occurs frequently in press-fitted parts and in many bolted or rivetted assemblies. In the latter case, it arises from the very slight relative movements possible between the head of the rivet and the component part, and fatigue cracks initiate from these fretted regions rather than from the rivet hole itself (Waterhouse, 1959). As an example of direct interest, fretting is one of the principal causes of marine propellor shaft failures (Heck and Baker, 1963).

The environment also has an added effect on fretting as an increased water content in the air, rather surprisingly, decreases the apparent severity of fretting (Waterhouse, 1959). Additionally, oxygen plays an important role, as illustrated in Figure 14 (Harris, 1961). Here, fretting in hydrogen is less detrimental than in air.

The mechanism by which fretting influences fatigue arises from the basic principles of friction. Contact takes place through a series of asperities which lock or weld at the interface. Small cyclic relative motion between the surfaces results in the specialized wear products characteristic of fretting, and increased contact through attrition of the original asperities. Fatigue cracks develop beneath these contact areas, resulting in either the dislodging of a massive particle from the surface or further propagation to failure (Field and Waters, 1967). This mechanism is substantiated by the observation (Cocks, 1966) that small relative tangential motion is accommodated by plastic shearing in the underlying material in a direction slightly inclined to the surface. The inclined cracks observed by Field and Waters (1967) have probably propagated along such lines of shear.

Much of the experimental work carried out has been on the basis of constant-amplitude fatigue tests. It should therefore be noted that the influence of fretting is apparently less serious with varying or programmed load spectra (Gassner, 1963). If this proves to be a generalization, extension of constant-amplitude fretting fatigue data to structural applications would result in favourable factors of safety.

The amelioration of fretting is by no means simple. If suitable compliant inserts can be used to accommodate the relative displacement between the two surfaces, then the basic mechanism for fretting is eliminated. If separate inserts are not feasible, various anti-fretting compounds have been used with varying degrees of success. These are usually based on molybdenum disulphide in a paint or grease-base carrier and essentially provide a low friction insert. It is important to note that interfacial inserts must be of a compliant nature, as fretting can still readily occur if rigid non-metallic inserts are employed. Hence the mere presence of a non-metallic insert is no guarantee that fretting, with its attendant danger of fatigue, cannot take place.

5. MATERIAL ASPECTS

The preceding Sections have given an admittedly brief outline of the fatigue problem, stressing its complexity, indicating the interaction of the numerous parameters, and drawing attention to the directions in which some of the more recent advances are being made. It is now pertinent to consider the more promising materials in the light of the theme of the Symposium, and to examine the effect on their fatigue properties of some of the more obvious operational conditions. It is inevitable that much of the discussion will be devoted to the deep submersible vessel, but some of the information will be equally applicable to such items as off-shore platforms, marine shafting, and oceanographic wire rope.

The materials of most interest are those with a high strength/density ratio, since payload will always be a consideration in hydrospace vehicles. An analysis by

Pellini (1964), on the basis of design knowledge, fabrication technology and strength/density ratio, placed potential materials in the order: steel, titanium, glass-reinforced plastics, and glass. A similar study by Park (1965), in terms of the effective ocean bottom area that can be covered, gave a measure of the advantages to be gained by increasing the strength level for steel and titanium alloys, and indicated the desirability of intensive development work in the glass-reinforced plastics field (Figure 15).

In view of these analyses, the materials selected for examination are classified into four main groups - ultra-high-strength steels, aluminum alloys, titanium alloys, and glass-reinforced plastics - and some attention is also given to recent progress in the field of metallic composites.

(a) Ultra-High-Strength Steels

Numerous ultra-high-strength steels, arbitrarily defined as having a yield strength greater than 150 ksi, have been developed in the last ten years, presumably under the stimulus of aerospace requirements. These steels range from the modified, low-alloy AISI 4340, through hot-work die steels, a variety of martensitic and austenitic stainless steels, to the high-alloy steels, such as 18 Ni maraging steel and 9Ni-4Co steel, and attain strength levels of up to 300 ksi. Unfortunately, as mentioned in Section 1, beyond a strength level of about 180 ksi the fatigue strength no longer increases at the same rate and may even decrease (Figure 16)(Swanson, 1964), and seldom does the value exceed 120 ksi. This is illustrated in Figure 17, taken from a recent survey carried out by Swanson (1964).

Further evidence has been provided by Tuffnell, Pasquine and Olson (1966) for 18 Ni maraging steel at three strength levels. Figure 18 shows a spread of less than 10 ksi in the fatigue strength for a spread of 67 ksi in the tensile strength. It is interesting to note, however, that in the low cycle region the higher tensile strength results in a marked increase in endurance, which is a matter of importance in submarine construction.

Various reasons have been advanced to explain the lack of improvement in fatigue strength above a certain point, and there is general agreement that a major factor is the presence of deleterious inclusions. Their size, number and distribution are all important characteristics, but evidence indicates that the effect of their size predominates. A change in number or distribution will alter the probability of an inclusion of critical size being in a region of high stress, but will have no effect if such inclusions are not present. Duckworth (1964) has estimated that the critical inclusion size for a fatigue strength of 200 ksi is about

10μ. The production of steels on a commercial basis with the inclusion size limited to 10μ presents a serious problem to the steelmaker, but recent developments in vacuum melting technology suggest that it is not beyond the realm of possibility.

Fatigue strength values in excess of 120 ksi have been reported (Borik et al., 1963), notably 170 ksi for an ausformed, 5% Cr die steel (H-11). The CEVM material was hot-rolled in the meta-stable austenite range to give a reduction of 90%, the resulting tensile strength being 360 ksi. The fatigue ratio, 0.47, is not far removed from that of 0.5, normally accepted for standard structural steels. It is known, however, that the finishing process induced compressive surface stresses of the order of 40 ksi, which undoubtedly contributed to some extent.

The ultra-high-strength steels, as one would expect, are quite sensitive to notches (Figure 18), and tend to show a high degree of scatter in endurance at the lower stress levels; values greater than 100 to 1, for example, have been reported for 18 Ni maraging steel (Cicci, 1964). Part of the scatter is due to variations in atmospheric environment (relative humidity), and a considerable amount of effort has been devoted to evaluating the effect of moisture on cyclic crack growth. According to research workers at the U.S.Steel Laboratories (Wei et al., 1966), several quenched-and-tempered and maraging alloys showed little difference in crack propagation rates when tested in an inert atmosphere. However, in a high humidity environment the rates tended to increase and to be inversely proportional to the fracture toughness of the specific alloy. Dahlberg (1965) and Van der Sluys (1966) have both confirmed that SAE 4340 is particularly susceptible to the humidity effect. In the low cycle region, the latter reported a 20-fold increase in crack growth rate in a moist environment. The significance of this effect is of special importance in the design of structures from a fracture mechanics viewpoint, since sub-critical flaw growth is frequently a controlling factor in determining the maximum permissible initial flaw size and the frequency of inspection.

Fatigue under normal atmospheric conditions, as mentioned earlier, may be regarded as a special case of corrosion fatigue, particularly for the ultra-high-strength steels. In an aggressive environment, such as sea water, it is presumed that they would be protected by some form of surface coating and/or impressed current, hence few corrosion fatigue data are available. Cathodic protection has proved to be quite satisfactory for lower strength steels in salt water, but for the high-strength steels there is the ever-present risk of hydrogen embrittlement. Brown (1964) has discussed the conditions under which the method may be effective, and Figure 19 illustrates some typical results for SAE 4330 steel in salt solutions. It is apparent that cathodic protection to the extent of -0.8 volt raises the endurance to about the "air" value, but over-protection will progressively reduce it.

Recent work by Carman and Katlin (1966) is noteworthy in connection with the Paris-Erdogan relationship proposed for crack propagation rates (Section 3). Tests were made on several ultra-high-strength steels in the 250-300 ksi strength range with some values of the cyclic stress intensity parameter (K) approaching the fracture toughness of the materials. At values of cyclic K up to 0.7 to 0.8 of the fracture toughness, the experimental data were in good agreement with the fourth-power relationship. Above this level, the crack growth rates were appreciably higher than the predicted values, with the result that under certain conditions the endurance could be 50% less than that predicted. Further studies in this region are obviously necessary before the relationship can be applied with confidence in structural designs involving repeated loading.

(b) Aluminum Alloys

Aluminum alloys offer the engineer a very wide range of properties for structural design. The favourable strength-to-weight ratio of these materials has encouraged their application in many fields of engineering, and their use for structural applications in marine engineering in particular is steadily increasing as the economic advantages become more apparent. Examples of direct interest are the hull construction of the Deep Quest submersible, for which it is understood 5083 aluminum alloy was selected, and that of the Aluminaut, for which the choice was the 7079 alloy.

When cyclic stressing is a major factor, however, conditions are less favourable than the simple weight advantage might suggest. The S/N curve for aluminum alloys is generally one of steadily decreasing slope with little levelling off before 5×10^8 cycles, and hence there is no fatigue limit. In the low cycle region, fatigue strength is approximately proportional to tensile strength, but at 5×10^8 cycles there is no such correlation (Stickley and Lyst, 1964)(Figure 20). While some of the medium-strength wrought alloys may give a ratio of fatigue strength (5×10^8 cycles) to tensile strength as high as 0.5, the high-strength alloys give much lower values, and at the present time the maximum quoted fatigue strength is about 22 ksi. Consequently, aluminum alloys should be most effective under conditions involving relatively few high stress cycles or when weight is the paramount consideration, as in aircraft and, possibly, deep submersibles.

Aluminum alloys obtain their strength by either strain-or age-hardening processes. The high-strength alloys are usually based on the latter mechanism and, broadly speaking, lack good corrosion resistance and favourable fatigue response. The strain-hardening alloys show better corrosion resistance, together with relatively good fatigue values, although the tensile strength will not normally exceed 50 ksi in any commercially useful condition. Their strengths are largely obtained by cold-working fully annealed materials and, with less ability to plastically relieve stress concentrations, this class is generally more sensitive to notch effect in fatigue. Detailed property differences exist between the various alloy types, which are based largely on Mg, Zn, Mn and Cr additions, and, as with the high-strength materials, many of the commercial alloys have been developed to meet specific property requirements (Anon.,1967).

The crack propagation rates of many of the aluminum alloys in sheet form have been investigated by Frost and his co-workers, and compared with those of other structural materials (Frost and Denton, 1966). The data were analyzed in terms of Equation 2 , using Frost and Dugdale's values of the exponents "m" and "n". In their expression, A is a material constant which may or may not depend on the mean stress, and the crack length does not exceed about 1/8 of the sheet width. A list of the growth rate coefficients (A) is given in Table 1, from which it will be apparent that only in the case of the aluminum alloys does the coefficient have a marked dependence on the mean stress. It will also be evident that, of the metals and alloys examined to date, the aluminum alloys have the highest crack growth rates and, of the two popular aircraft alloys, the Al-Cu type is superior to the Al-Zn-Mg type. A few tests were made on the Al-Cu type alloy in the clad condition, and it was observed that the cladding had no effect on the rate of crack growth. Somewhat similar tests on an alloy of the 2618 type also showed that the crack propagation rates were the same with or without cladding (Rooke et al., 1964).

A generally satisfactory resistance to saline corrosion has led to considerable marine usage of the 5000 series Al-Mg alloys. In most instances weld characteristics are favourable, and extensive use is now found in small boat hulls and various superstructural components. In the majority of these applications the cyclic stresses are not high and fatigue properties are not generally of great concern in material selection.

Where design requirements call for maximum strength levels, such as in deep sea vessels, the age-hardened aluminum alloys (e.g. 7079-T6) are at present the only available materials. The inferior corrosion resistance of these alloys in marine environments necessitates the use of both protective coatings and cathodic protection in order to maintain surface integrity, and welding should be avoided. One of the more promising high-strength alloys is the newly developed X7002 (Ailor, 1965), which has better corrosion resistance than other alloys in this series, but weldments are still a problem.

(c) Titanium Alloys

The development of titanium alloys is expanding so rapidly that reliable, detailed information on the fatigue characteristics of the newer alloys is not always available. In general, the fatigue strengths, relative to the tensile strengths, are as good as or better than those of steels, many showing a fatigue ratio greater than 0.5 (Forrest, 1962). The S/N curves for most of the alloys tend to become horizontal at long endurances and some show a definite fatigue limit. Their sensitivity to notches increases with the strength level and is similar to that of steel, but their resistance to corrosion is superior and their lower density places them in a strongly competitive position.

The most popular alloy is probably Ti-6Al-4V, which can be heat-treated to 170 ksi and which has been intensively studied. The fatigue strength of this material has been reported (Anon.,1957) as 85 ksi, or one-half of the tensile strength. The presence of a notch (K_t = 3.3) reduced the fatigue strength to about 37 ksi, corresponding to a strength reduction factor of 2.3. Data presented by Parker (1962) for titanium and three titanium alloys covering a strength range of 80 ksi to 150 ksi also gave an average fatigue ratio of 0.53 and comparable strength-reduction factors. Concurrent tests on two aluminum alloys (Al-Mg-Si and Al-Cu-Mg-Si) indicated some superiority for the titanium alloys in the presence of a notch. A similar comparison was reported by Hardrath (1967) for Ti-8Al-1Mo-1V and 7075-T6 aluminum alloy with a notch of K_t = 4. Figure 21 shows the S/N curves obtained with the stresses adjusted for density, and the better efficiency of the titanium alloy is evident, particularly in the low cycle range.

In comparison with ultra-high-strength steels, the superiority of titanium alloys on a strength-weight basis has also been demonstrated (Anon.,1960). The tests were made in direct stress with a minimum stress/maximum stress ratio of 0.25 on titanium alloys Ti-6Al-4V (UTS 165 ksi) and Ti-4Al-3Mo-1V (UTS 195 ksi), and stainless steels PH 15-7Mo (UTS 240 ksi) and AM 355 (UTS 210 ksi). The steels gave higher unnotched fatigue strength values than the titanium alloys, and were slightly better when notched (K_t = 3.5). When the data were plotted with a stress/density ordinate (Figure 22), however, the relationship was reversed.

Fatigue crack propagation studies have been carried out by Frost and Denton (1963) on titanium and two titanium alloys (Ti-5Al-2.5 Sn and Ti-15Mo) in the form of sheet. Tests were made with various mean and alternating stresses, and for short cracks (less than about 1/8 of the sheet width) the results were in good agreement with Frost and Dugdale's version of Equation 2.

Values of the constant A were determined and are given in Table 1. There was no significant effect of mean stress, and it is apparent that the crack growth characteristics of the three materials were similar, with one exception. Comparison with the data for an aluminum alloy (BS L71) shows that the titanium alloys were much the same at low mean stress (4 ksi), but were increasingly superior as the mean stress was raised; at 30 ksi, they were superior by a factor of about 10.

The effect of an aqueous environment on crack growth has also been examined, with somewhat surprising results. Tests at the Naval Research Laboratory (Judy et al., 1966) were made in the low cycle region on Ti-6Al-4V and Ti-7Al-2Cb-1Ta alloys in air and in 3.5% salt water. The Ti-6Al-4V alloy showed no effect of the change in environment, whereas the Ti-7Al-2Cb-1Ta alloy gave a marked increase in crack growth rate in the presence of salt water, particularly at the higher strain levels (Figure 23). Even distilled water resulted in a significant increase. A similar effect with the latter alloy was reported by Brown et al. (1965) in stress corrosion cracking studies under static load. Tests by Figge and Hudson (1967) on Ti-8Al-1Mo-1V alloy in air and in sea water, with or without intermittent thermal soaking, showed that this alloy, too, was susceptible. They reported that cracks grew faster by a factor of 2 to 3 in the aqueous environment. It is interesting to note that companion tests demonstrated about the same degree of sensitivity for 7075-T6 aluminum alloy, but less for 2024-T3 aluminum alloy. At the higher stresses, the latter alloy even gave some improvement in sea water.

Sufficient has been said to indicate that environmental effects can constitute a problem with titanium alloys. It is understood that a more recent alloy, Ti-6Al-2Cb-1Ta-0.8Mo, has been developed to overcome these difficulties, and that its fatigue properties in a salt water environment are significantly better than those of Ti-7Al-2Cb-1Ta alloy.

(d) Composite Materials

The term 'composite materials' is being used to differentiate between homogeneous or heterogeneous materials of natural origin, viz steels, etc., and those heterogeneous mixtures processed to form a unified composite structure, viz glass fibre-resin composites. This differentiation, based as it is on a processing factor, is arbitrary as the purist can argue that age-hardened aluminum alloys do not occur naturally. However, the "man-made" processing factor is convenient and emphasizes in part one of the reasons for such composite structures.

The strength of any material depends on its ability to resist plastic deformation, that is, its resistance to dislocation motion. One can do this on an atomic scale (solution hardening) or on a microscale (precipitation hardening). Available engineering materials can attain very high structural properties but these are limited by those bulk effects related in part to the surface-to-volume ratio. It has long been realized that filamentary materials often possess strengths far in excess of the bulk properties and, additionally, there exist a range of filamentary materials not normally available for usable engineering applications. Composite materials have arisen from attempts to use the desirable filamentary properties in dimensions and applications of general availability to the engineer. In very simple terms, composites consist of a higher strength component dispersed suitably in a softer matrix. For the purposes of this discussion, two groups will be considered. The grouping is based on the nature of the matrix material, either metallic or non-metallic.

There are certain basic features of composite materials which should be understood. Most composites use materials of differing elastic moduli. In addition, the reinforcing component frequently has virtually no plastic capability. When a composite is strained, considerable differences in stress level in each component will result, due to the equidistribution of strain. A shear stress will therefore exist across the bond interfaces within the composite. Further complications arise from plastic yielding of the matrix but will not be discussed here. The tensile strength of a composite depends thus on the respective strengths of the constituents, the volume proportion and the bond strength. It is possible to interrelate these mathematically (Kelly and Davies, 1965), and reasonable agreement with theory has been obtained. Fatigue is far less simple and such data for suitable engineering composites are not very common.

The composites most widely used, because of their ready commercial availability, are the non-metallic matrix group. Over the past ten years, the performance of these materials, notably those based on epoxy-resin-bonded glass fibres, has steadily improved. Laminates prepared from numerous resins (epoxides, polyesters, phenolics) with a variety of reinforcements (asbestos or glass mat, glass fabric, unwoven glass fibres) at different orientations have been studied (Boller, 1964). In general, the S/N curve for the laminates gives no fatigue limit and fatigue strengths are quoted at 10^7 cycles. The laminates show some sensitivity to notches, but are reported as being far less notch-sensitive than some high-strength steels (Davis et al., 1964). Resin content in the range 20% to 37% does not appear to be important, but the orientation of the ply or fibres can have a marked effect. Optimum results have been obtained with Scotch-ply, epoxy-resin laminates reinforced with unwoven glass fibres having alternate plies at $\pm 5°$ to the principal axis, and fatigue strengths as high as 40 ksi have been reported (Boller, 1966). On a fatigue strength/density basis, many of these materials are now superior to steels and may be considered as potential replacement materials in many applications where size requirements are not critical.

Metallic matrix composites are still in their infancy and a portion of the published data has been derived from model systems chosen for evaluation of theory rather than as potentially useful engineering materials. Matrix reinforcement has consisted of ultra-high-strength metal whiskers, non-metallic fibres, and high-strength wires. Although whiskers have generally the highest available ultimate tensile strengths and are therefore potentially most promising, their commercial production is not simple and successful incorporation into a composite is equally difficult. It is not surprising, then, that published data on fatigue properties are essentially limited to composites containing continuous or discontinuous fibres or wires.

Some of the earlier work (Forsyth et al., 1962) was carried out with 7075- and 2024-type alloys, using aluminum sheet and stainless steel wire mesh in a sandwich construction. In pulsating tension with a high mean stress, the maximum improvement in endurance was by a factor of about four for the 7075-type alloy and two for the 2024-type alloy. Subsequently, the investigators modified their sandwich-construction technique and obtained substantial reductions in crack propagation rate with the incorporation of only a small volume fraction of wire (Forsyth et al., 1964). There was also an increase in the unstable crack length, but the effect on the fatigue strength was negligible. This result was not unexpected since the principal benefit of the reinforcement, a reduction in crack growth rate, would be much less significant in the low stress region.

More recently, composites have been produced by hot-pressing stainless steel fibres pre-coated with aluminum (Baker, 1966). A decrease in fibre diameter from 5 mil to 2 mil was found to be beneficial, probably due to the increased area of interface. A similar improvement was observed with a change from short discontinuous fibres to continuous fibres, and was attributed to the elimination of the deleterious effects of the fibre ends. At high strain levels, however, the fibres were unable to prevent the direct propagation of cracks and the behaviour approached that of the unreinforced matrix.

Aluminum has also been reinforced with continuous silica (Baker et al., 1966) and boron fibres (Morris, 1967) by essentially the same process. In the latter case, Morris reported an increase in fatigue strength from 8 ksi for pure aluminum to 53 ksi for the boron-reinforced composite in pulsating tension. Similar experiments with a 2024 aluminum alloy matrix were rather disappointing since, although the tensile strength of the composite was increased, the fatigue strength was greatly reduced.

Continuous tungsten wire has been used to reinforce both copper (Ham and Place, 1966) and silver (Morris and Steigerwald, 1967) matrices, with conflicting results from the fatigue viewpoint. For the copper matrix, Ham and Place found a marked increase in tensile strength with increasing volume fractions of tungsten (up to 23%), but only a small increase in fatigue strength. They attributed the poor fatigue performance of the composite to cyclic hardening of the matrix at crack tips, leading to stress concentrations sufficient to fracture adjacent filaments. For the silver matrix, however, the increase in composite tensile strength with increasing amounts of tungsten was roughly paralleled by the increase in fatigue strength, again in pulsating tension. Discontinuous fibres were observed to be less effective, as noted previously. Another characteristic of the composites was their relative insensitivity to surface defects, which may provide an additional bonus for the designer.

While it is reasonable to conclude that the application of metal-matrix composites on a large scale to engineering structures is still many years away, their localized application to combat the fatigue phenomenon may not be so far removed. Work at the U.K. National Engineering Laboratory is being directed towards the insertion of small areas of composite material into larger metallic components in regions where fatigue cracks could be expected to initiate. The results of such studies could be of particular interest to the design engineer.

6. SUMMARY

As a broad and brief summary, it can be said that research into the phenomena of the fatigue problem has now reached a stage where there is a reasonable understanding of the processes taking place within cyclically stressed materials on the physical, metallurgical and engineering scales. With the progressive translation of this improved knowledge into currently available materials and into the design and development of more fatigue-resistant materials, it may be expected that superior alloys and composites will be provided for the design engineer. The clearer understanding of the effects of the numerous operational parameters should also facilitate the incorporation of more realistic factors of safety in cyclically loaded structures. However, it would be unwise to look forward to any abrupt technological breakthrough, as the pattern of fatigue research has been one of a steady evolution of knowledge of the subject rather than that of abrupt advances.

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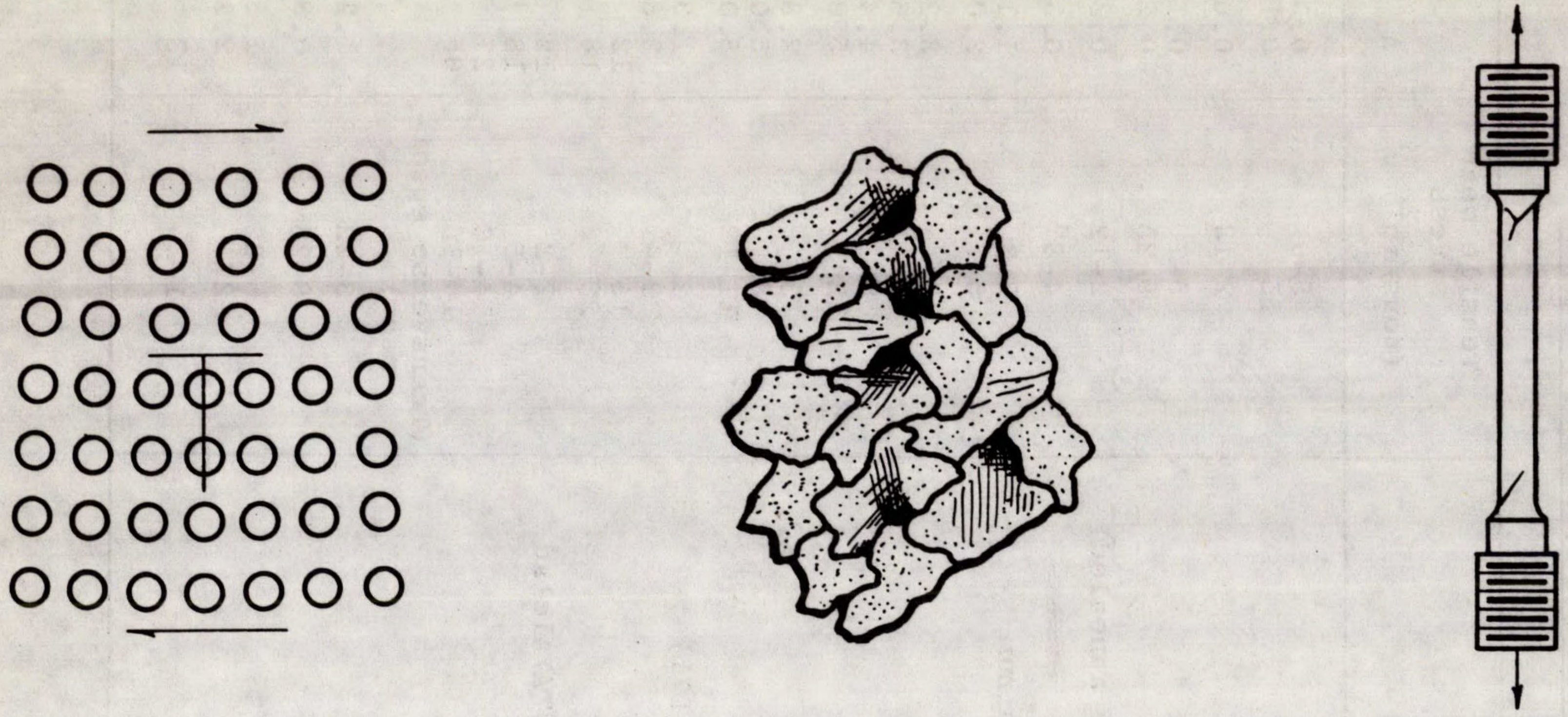
TABLE 1

Values of A (Growth Rate Coefficient)

Materials	Tensile mean stress (tons/in^2)	A
Austenitic steel	5-9 15	0.08 0.11
Mild steel	2-13	0.09
Cold-rolled mild steel	5-7 15-25	0.12 0.22
Copper (cold-rolled or annealed)	2-12	0.37
Brass	4-8	0.35
Commercially pure aluminum	2-3 5-7	1.1 2.0
Aluminum alloy HS30WP	3 5 8 12	3.2 3.6 4.4 5.0
5% Mg-Al alloy	3 8 11-12	2.8 5.0 8.0
4½% Cu-Al alloy (B.S. L71)	2 3-4 15 23	1.3 2.8 12 18
5½% Zn-Al alloy(DTD 687A)Alclad	2-3 4-5 14-16 14-15 (Transverse specimens)	12 18 27 65
Titanium	3-22	1.2
5% Al-Ti alloy	8-28	0.8
15% Mo-Ti alloy	8-22 27½	1.3 2.2
Zinc	1-3	13



1 PHYSICS OF SOLIDS

2 METALLURGY

3 SPECIMEN

Figure 1. Material models (after Dolan).

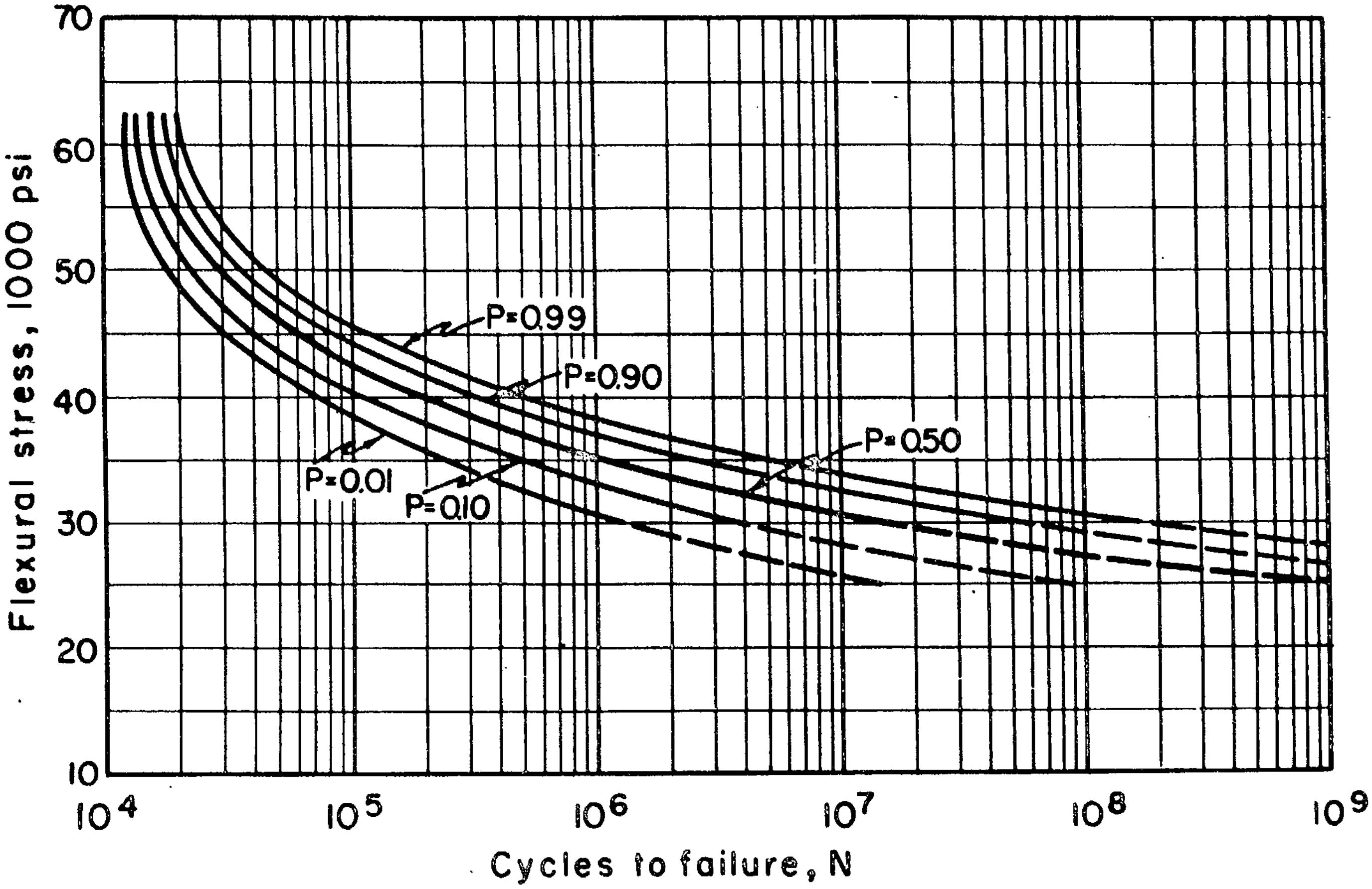


Figure 2. Composite S-N curves for 7075-T6 aluminum alloy (after Sinclair and Dolan).

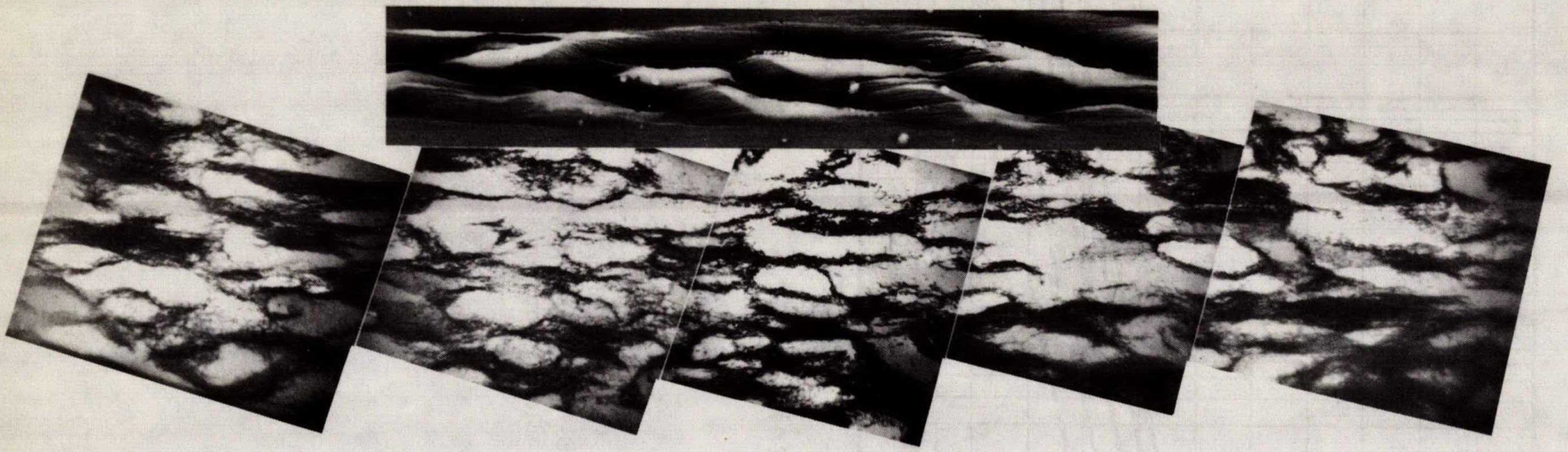
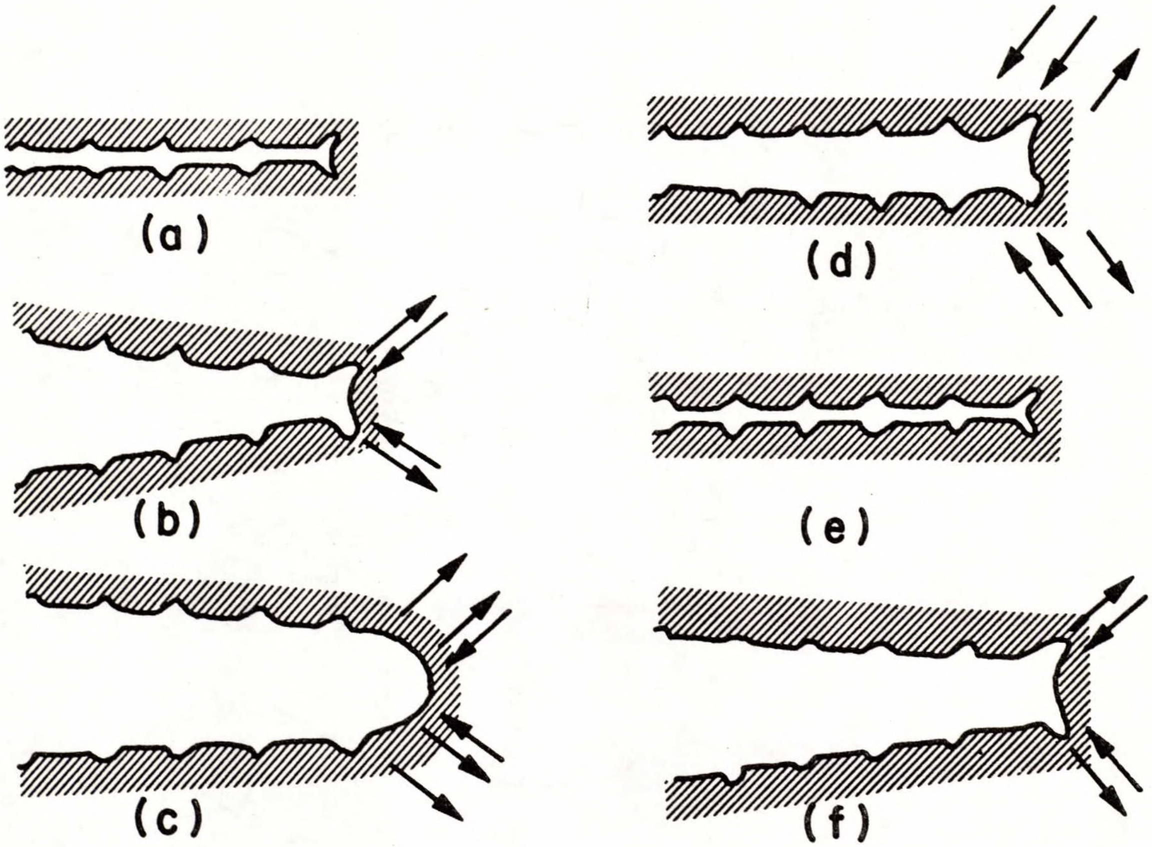


Figure 3. Dislocation substructure in copper (Laufer and Roberts).



a - ZERO LOAD
b - INCREASING TENSION
c - MAX. TENSILE STRESS

d - COMPRESSIVE CYCLE
e - FORMATION OF NEW NOTCH
f - REPEAT OF b

Figure 4. Crack propagation mechanism (Laird).

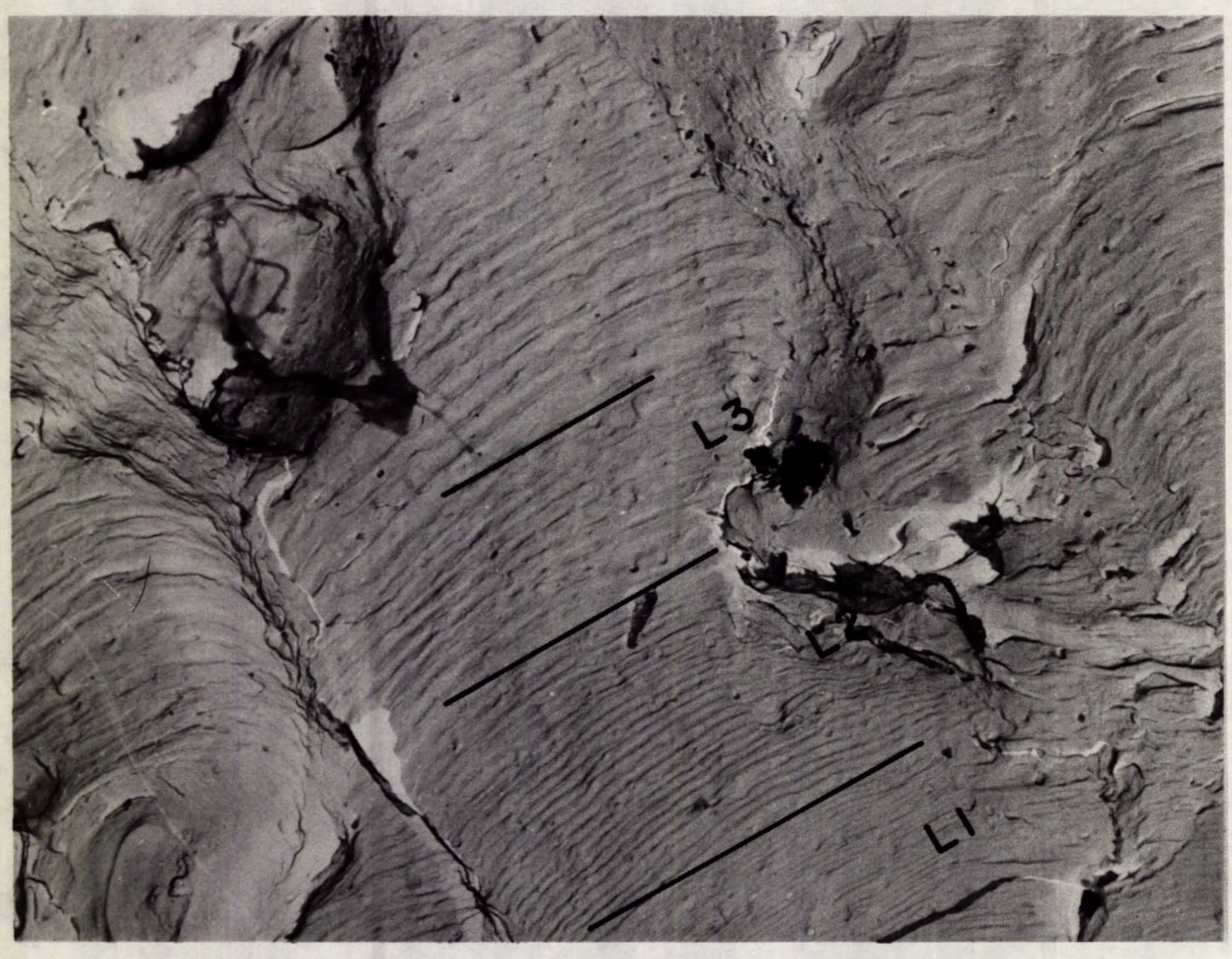


Figure 5. Fracture surface of aluminum alloy aircraft wing component (Wiebe).

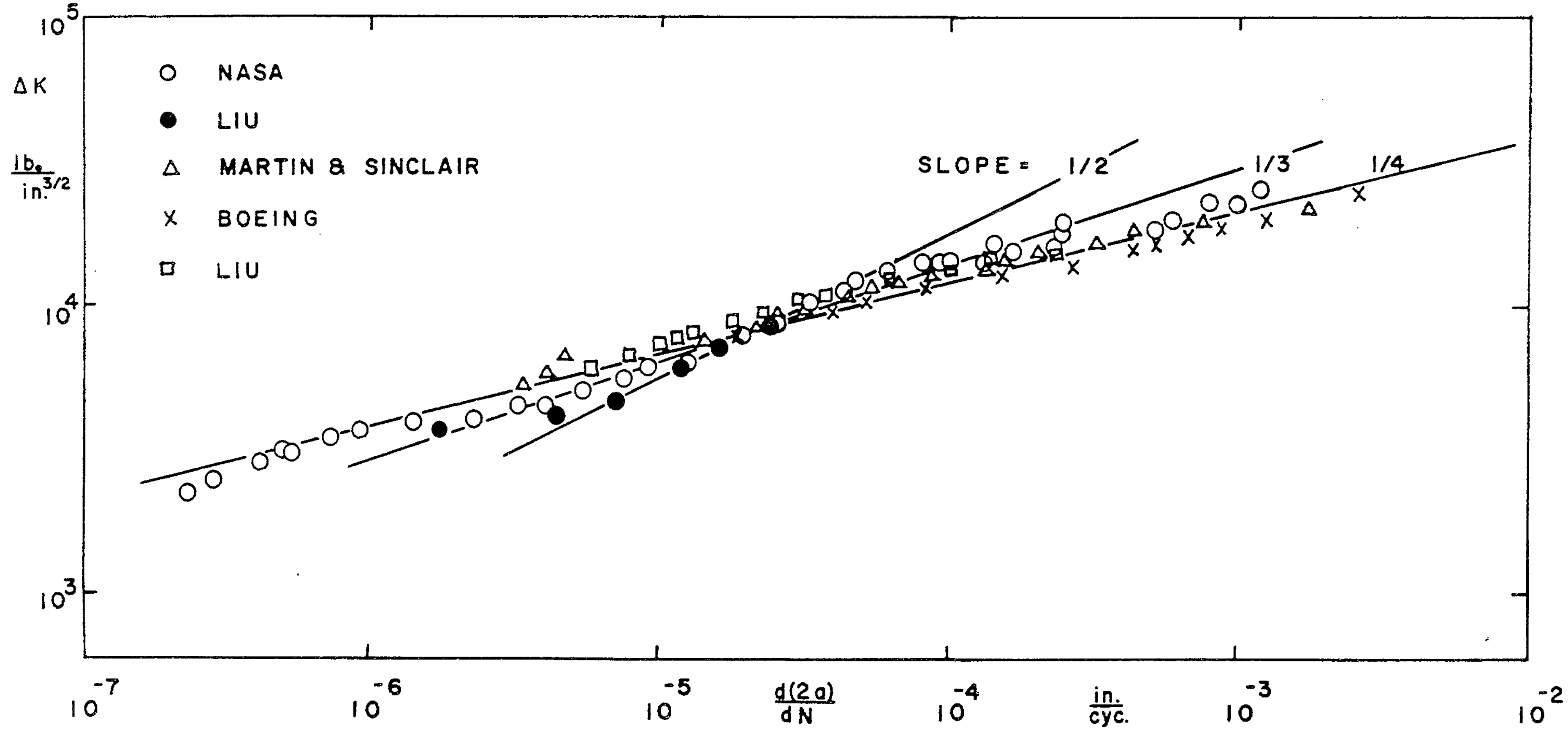


Figure 6. Crack growth data for 2024-T3 aluminum alloy (Paris and Erdogan).

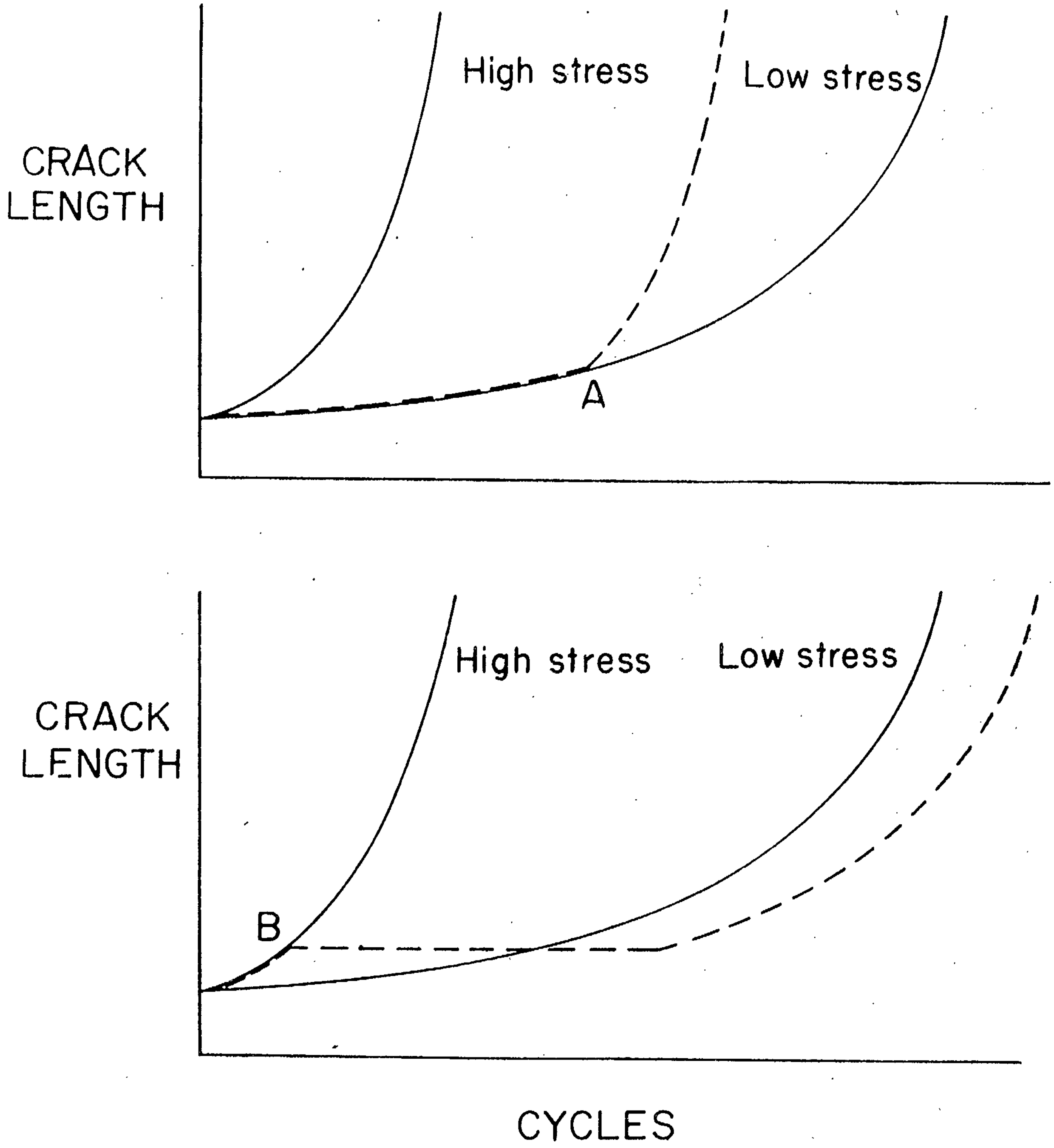


Figure 7. Crack growth curves for two-level tests (Hardrath).

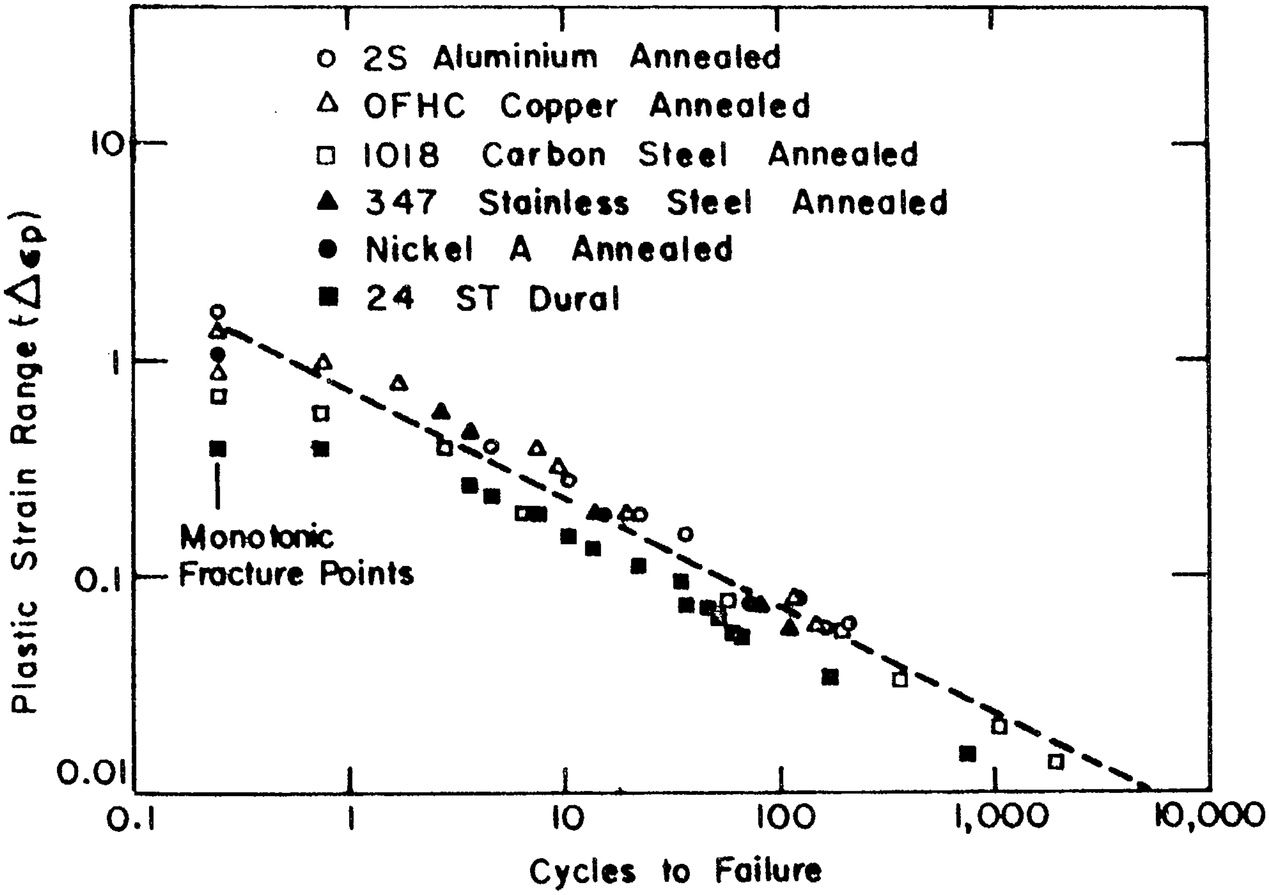


Figure 8. Plastic strain range versus endurance (after Coffin).

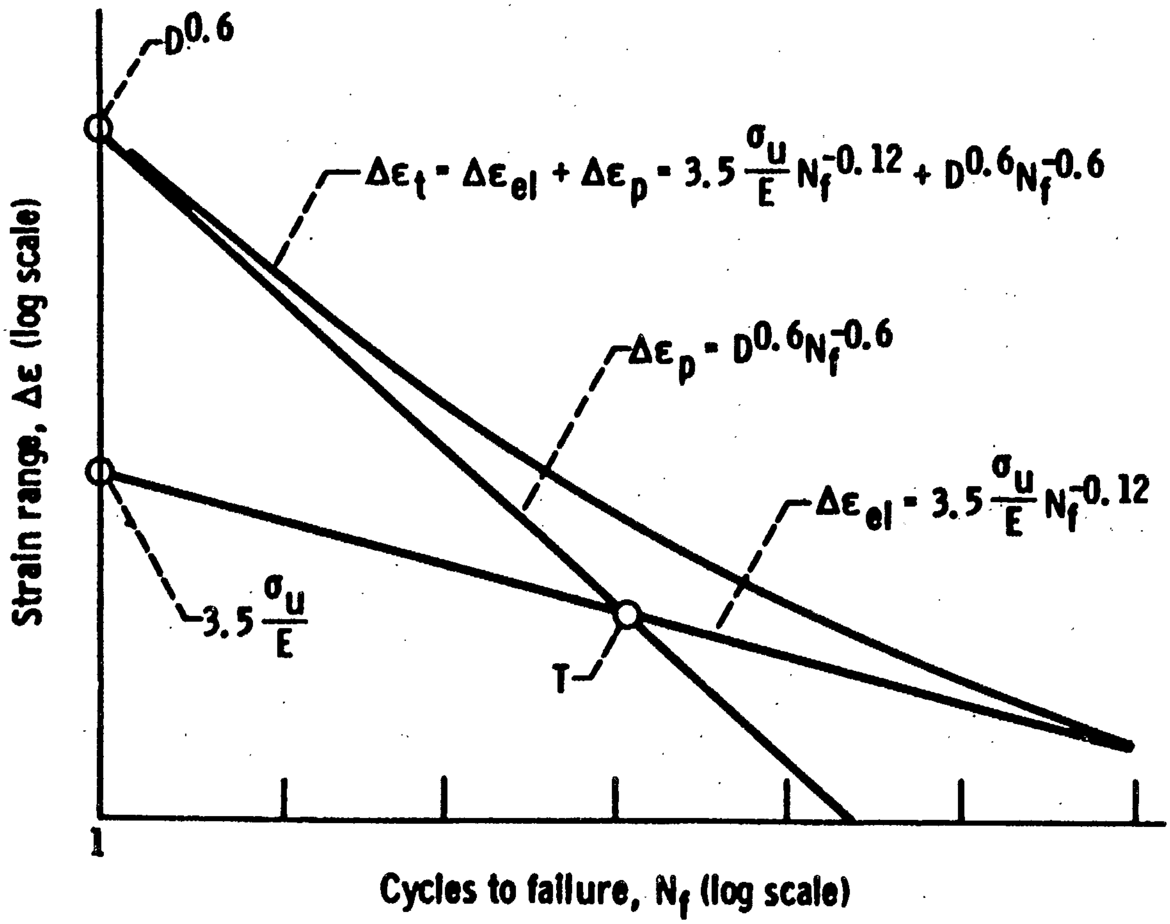


Figure 9. Universal slope method (Manson).

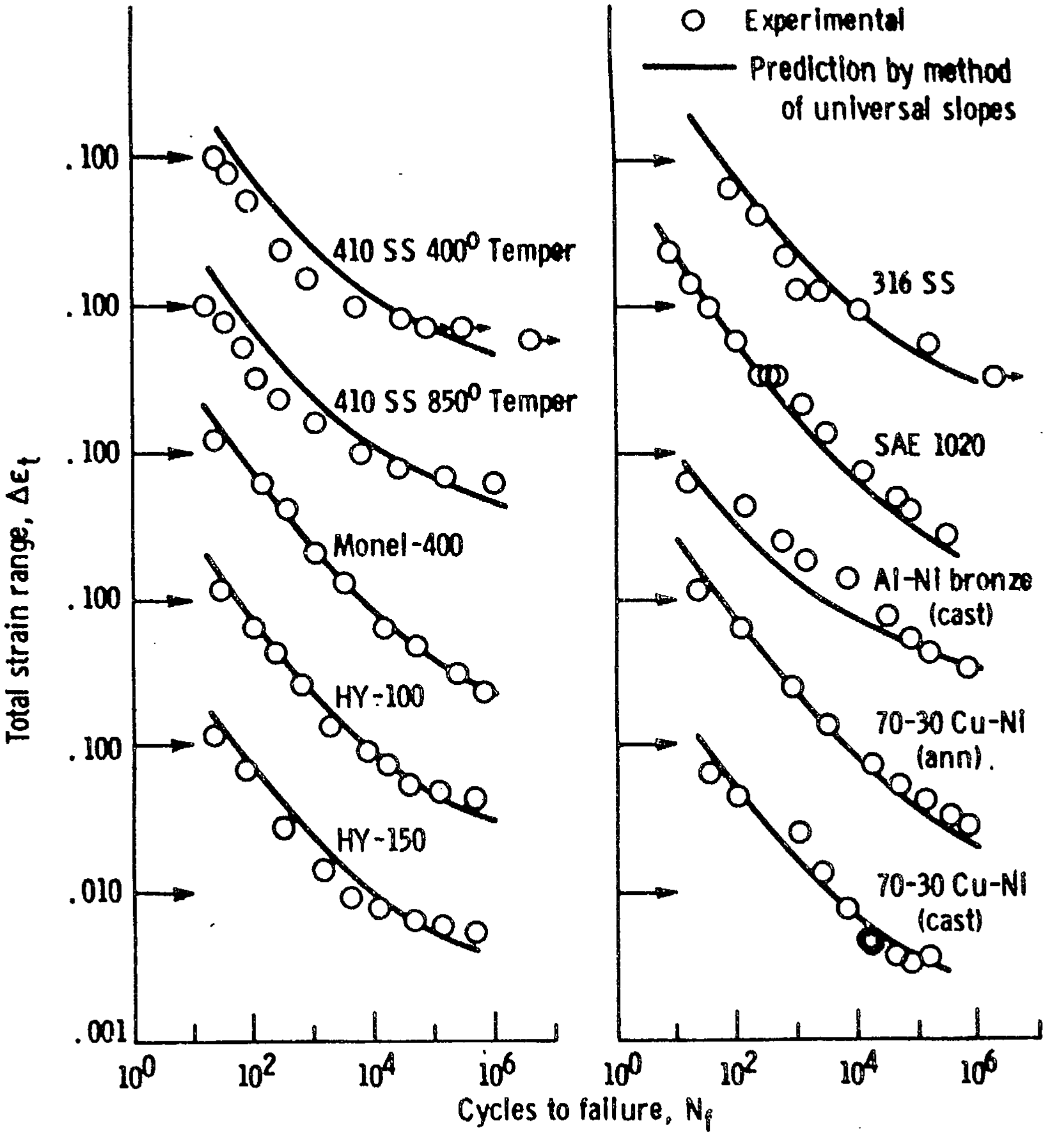


Figure 10. Predicted and experimental fatigue behaviour (Manson).

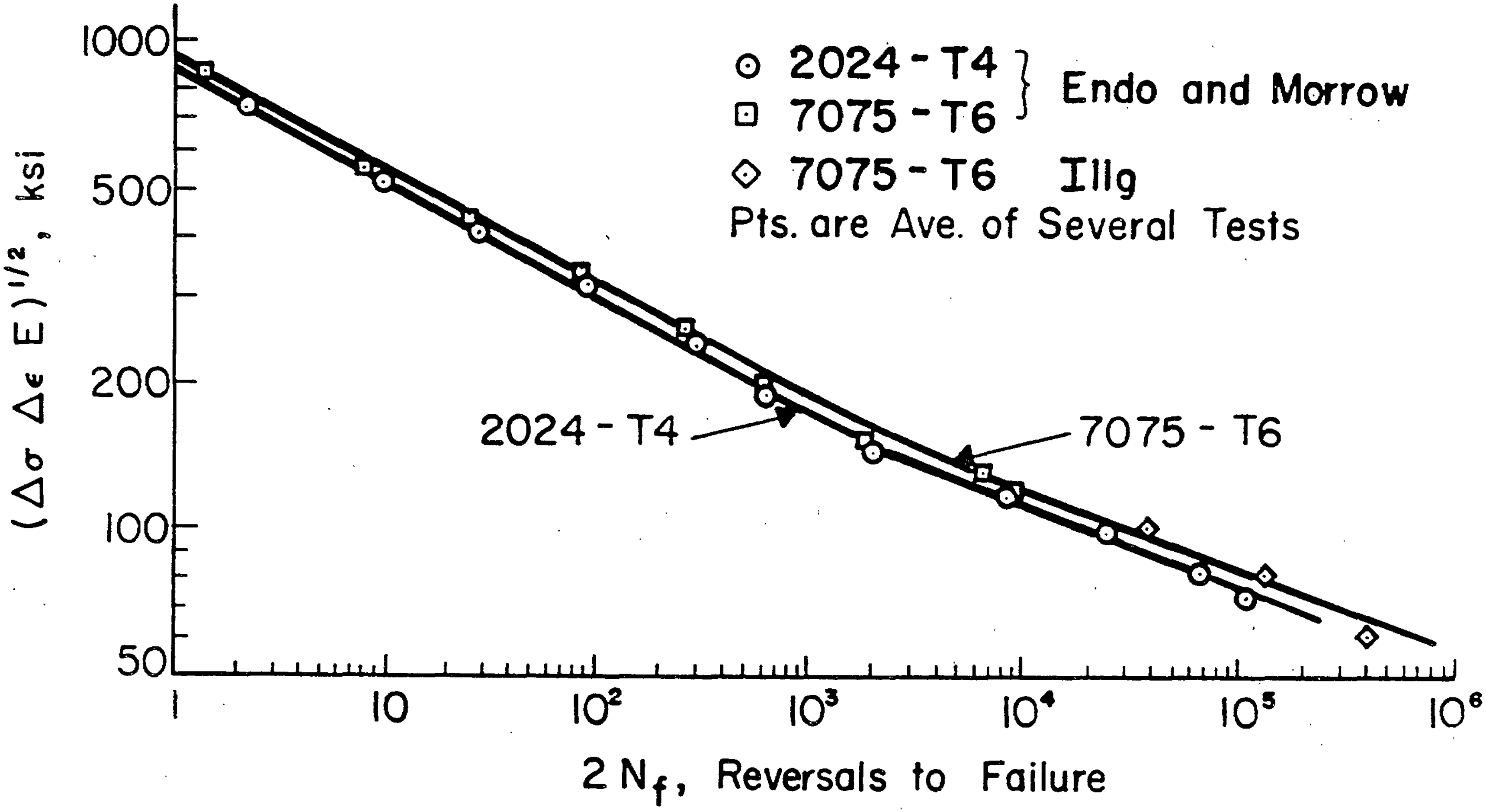


Figure 11. Smooth specimen data (Topper et al).

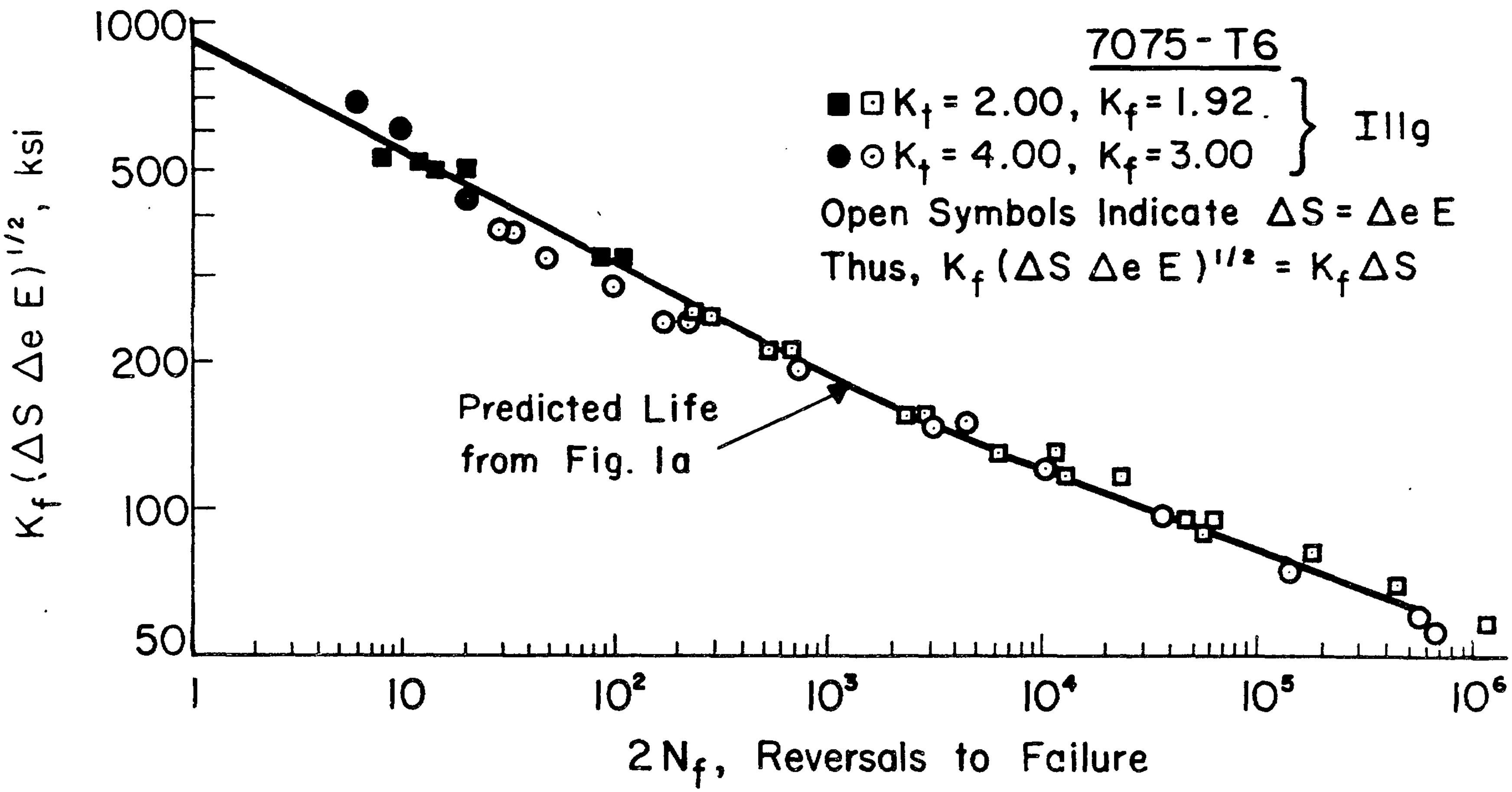
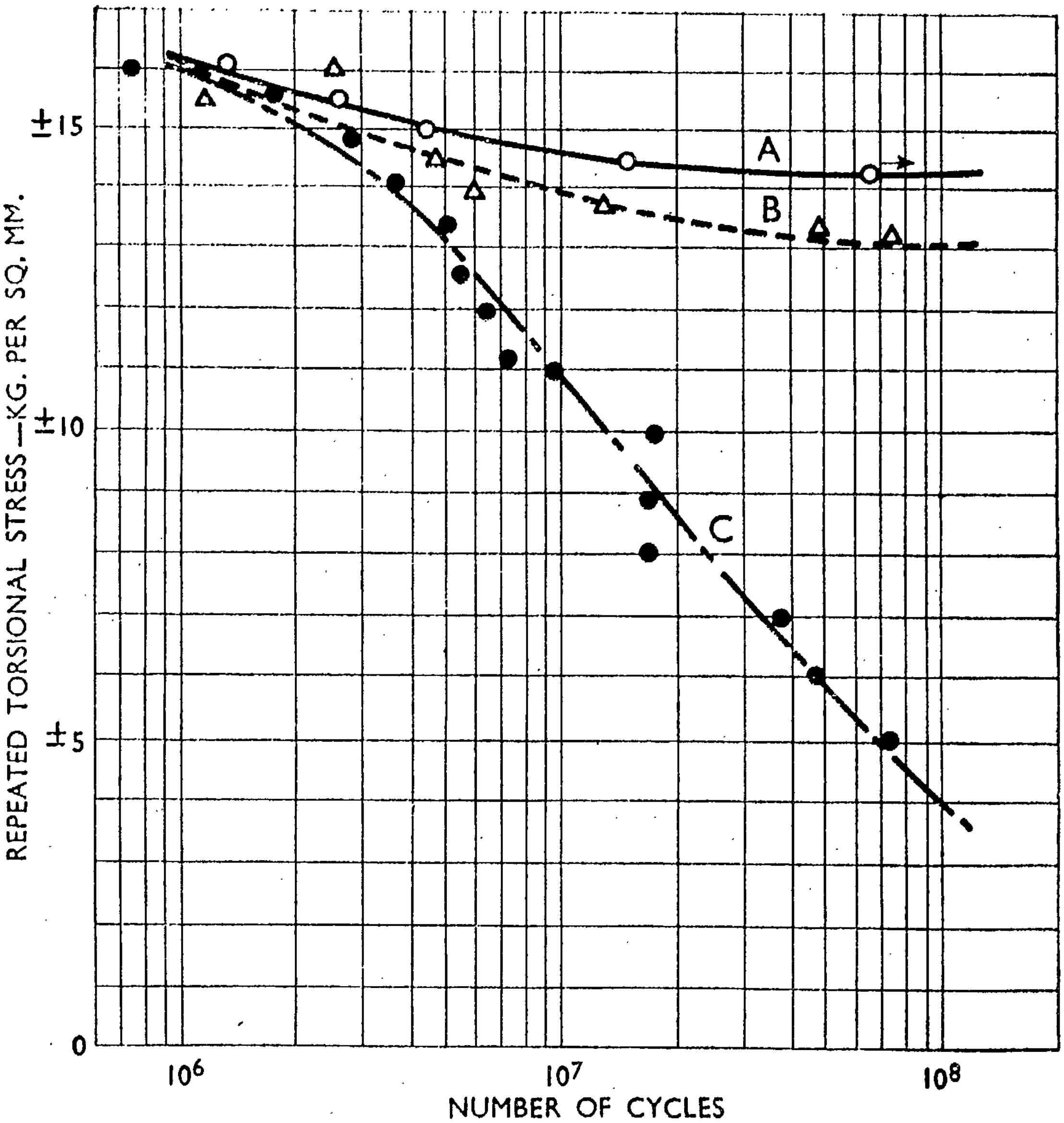


Figure 12. Notched data compared with prediction from smooth data (Topper et al.).



A In air.
B With fresh-water corrosion.
C With sea-water corrosion.

Figure 13. S-N curves in air, fresh and sea water (after Hara).

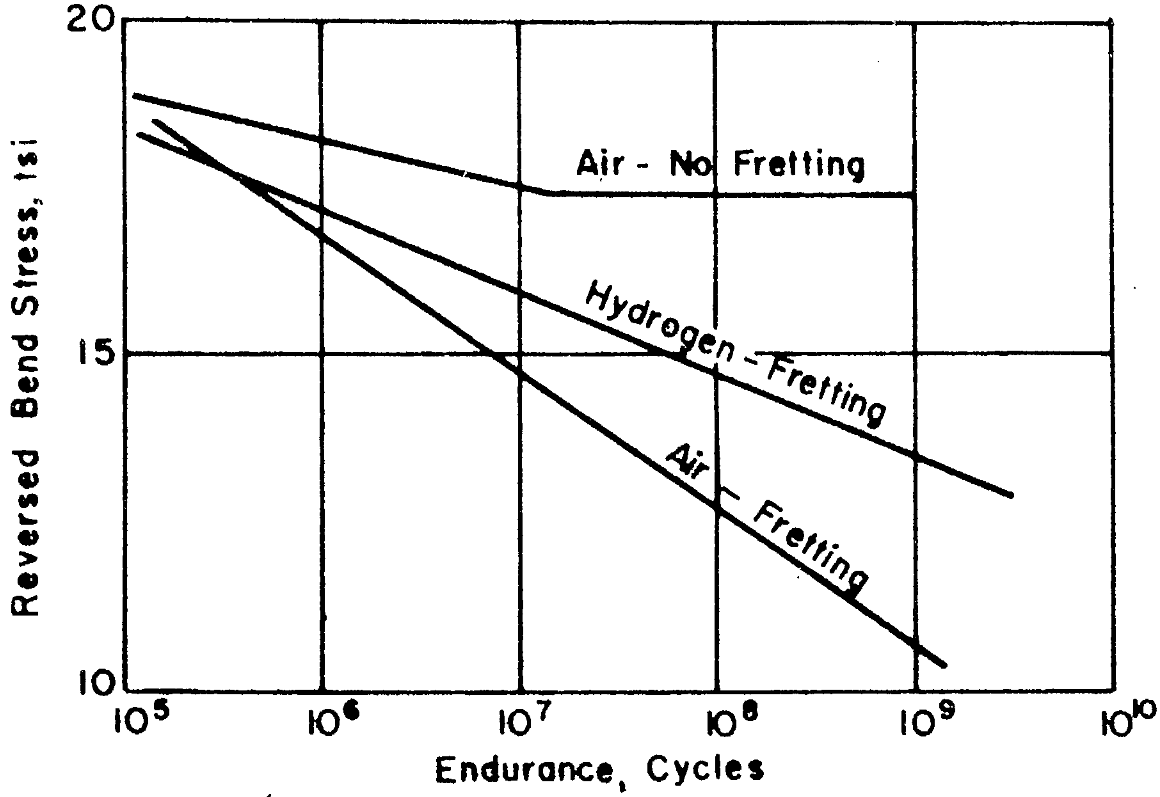


Figure 14. Fretting fatigue properties of Ni-Cr steel (after Harris).

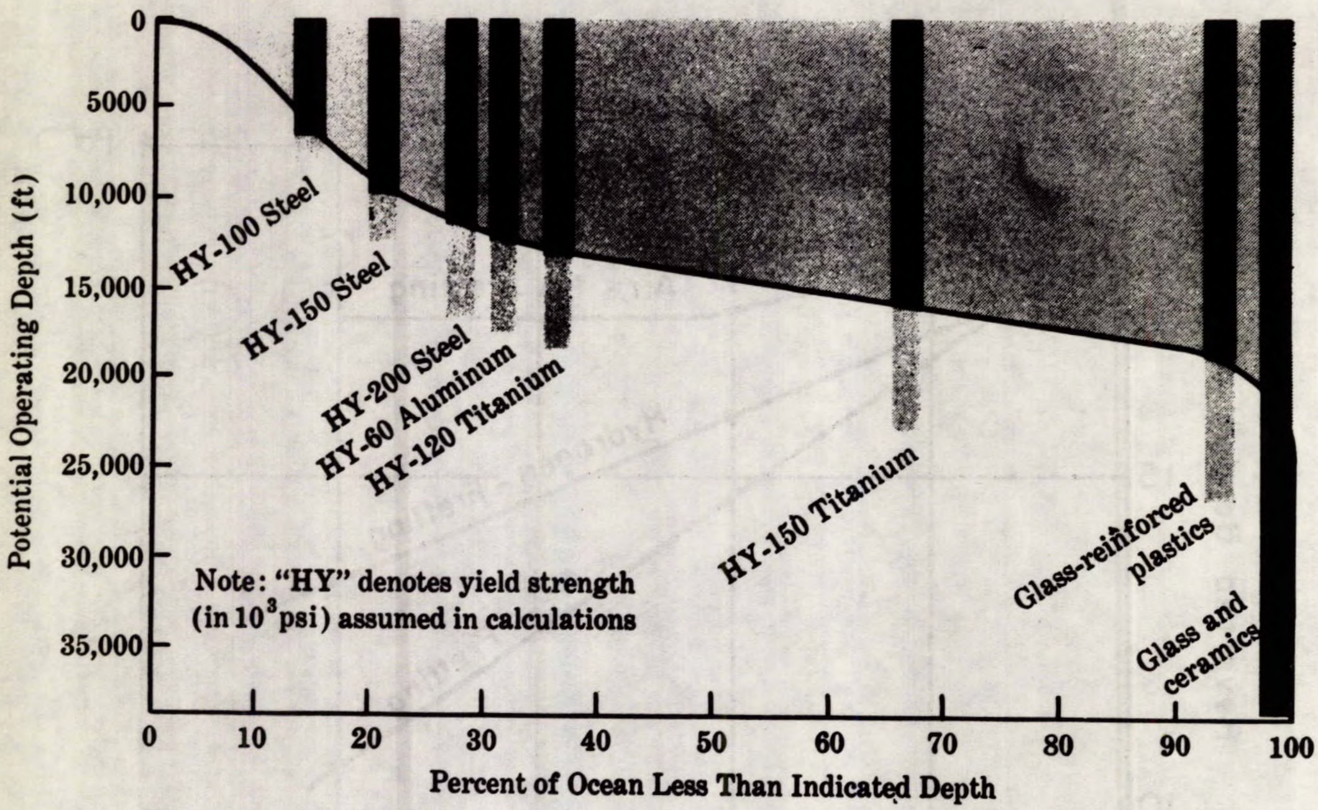


Figure 15. Potential materials for deep diving vessels (after Park).

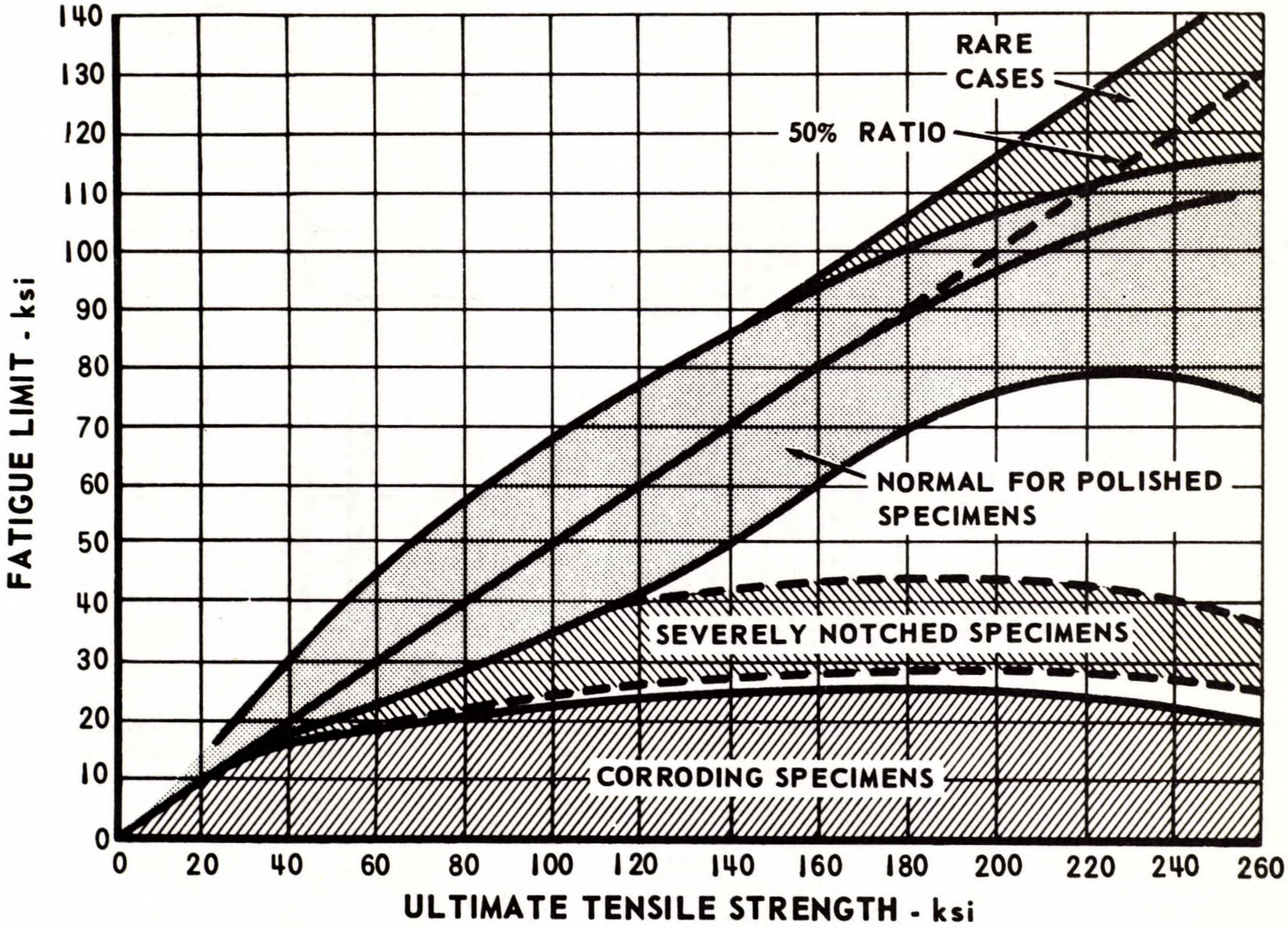


Figure 16. Fatigue properties of wrought steel (after Bullens).

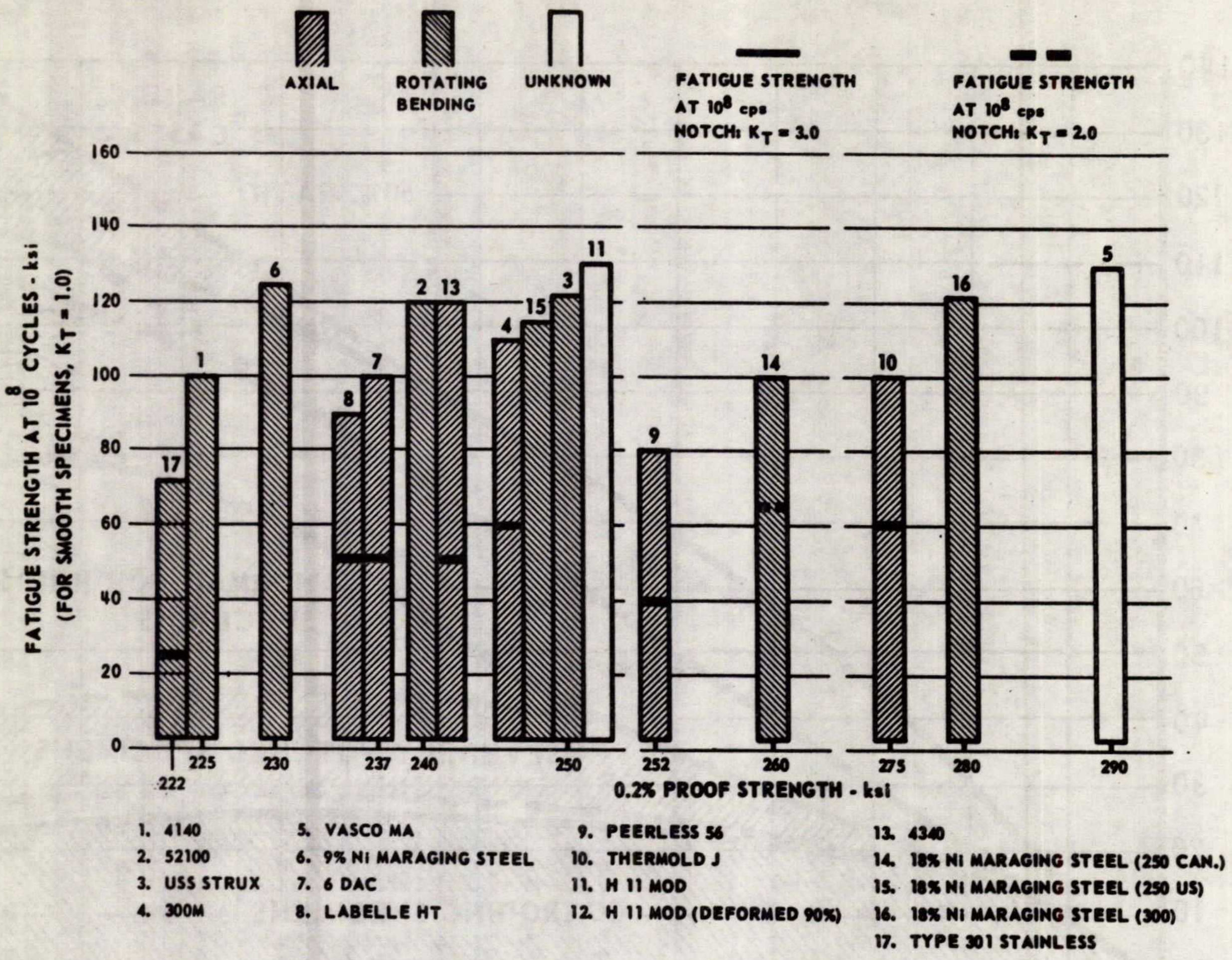


Figure 17. Fatigue properties of high-strength steels (Swanson).

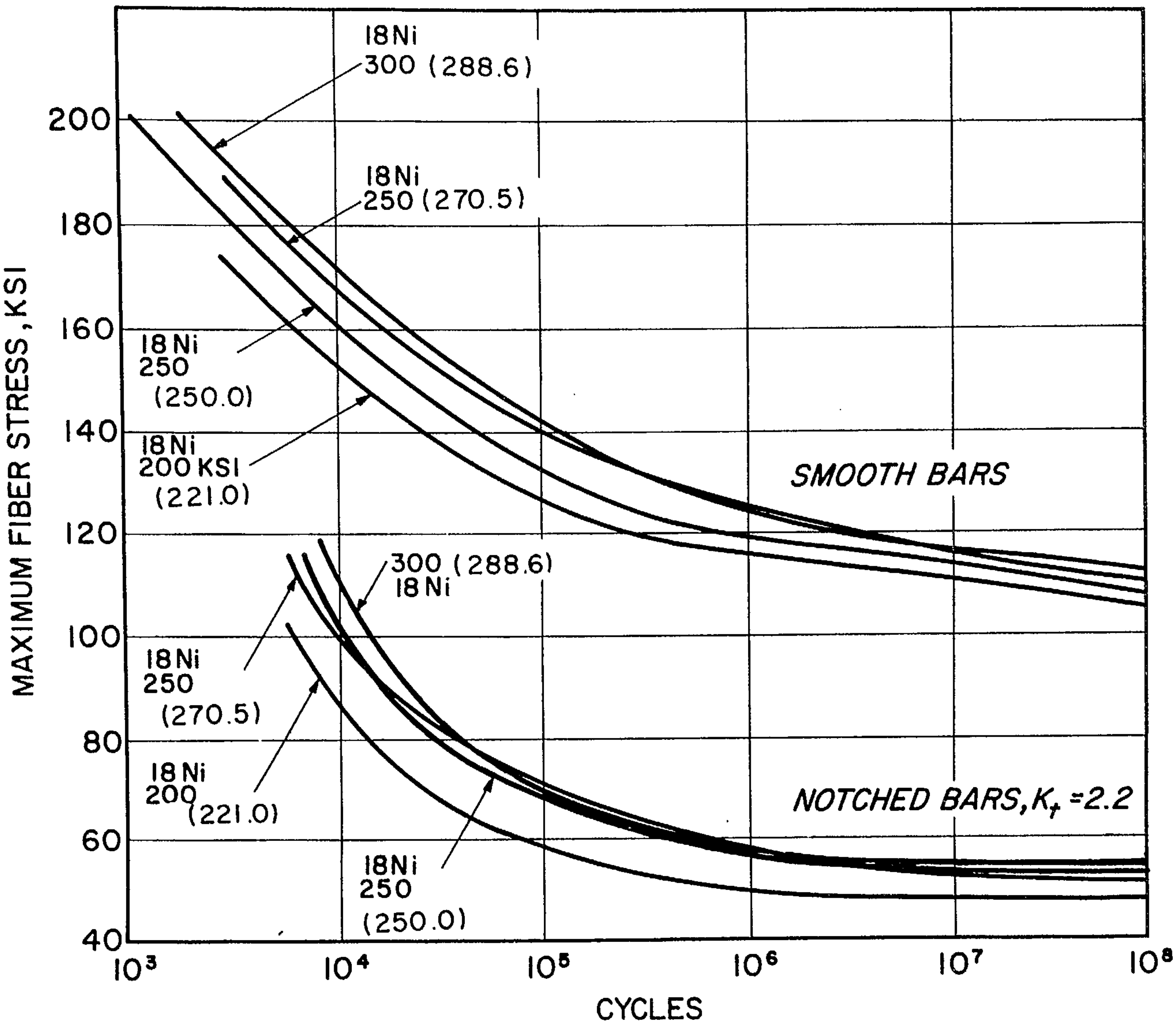


Figure 18. S-N curves for 18% Ni maraging steel (Tuffnell et al.).

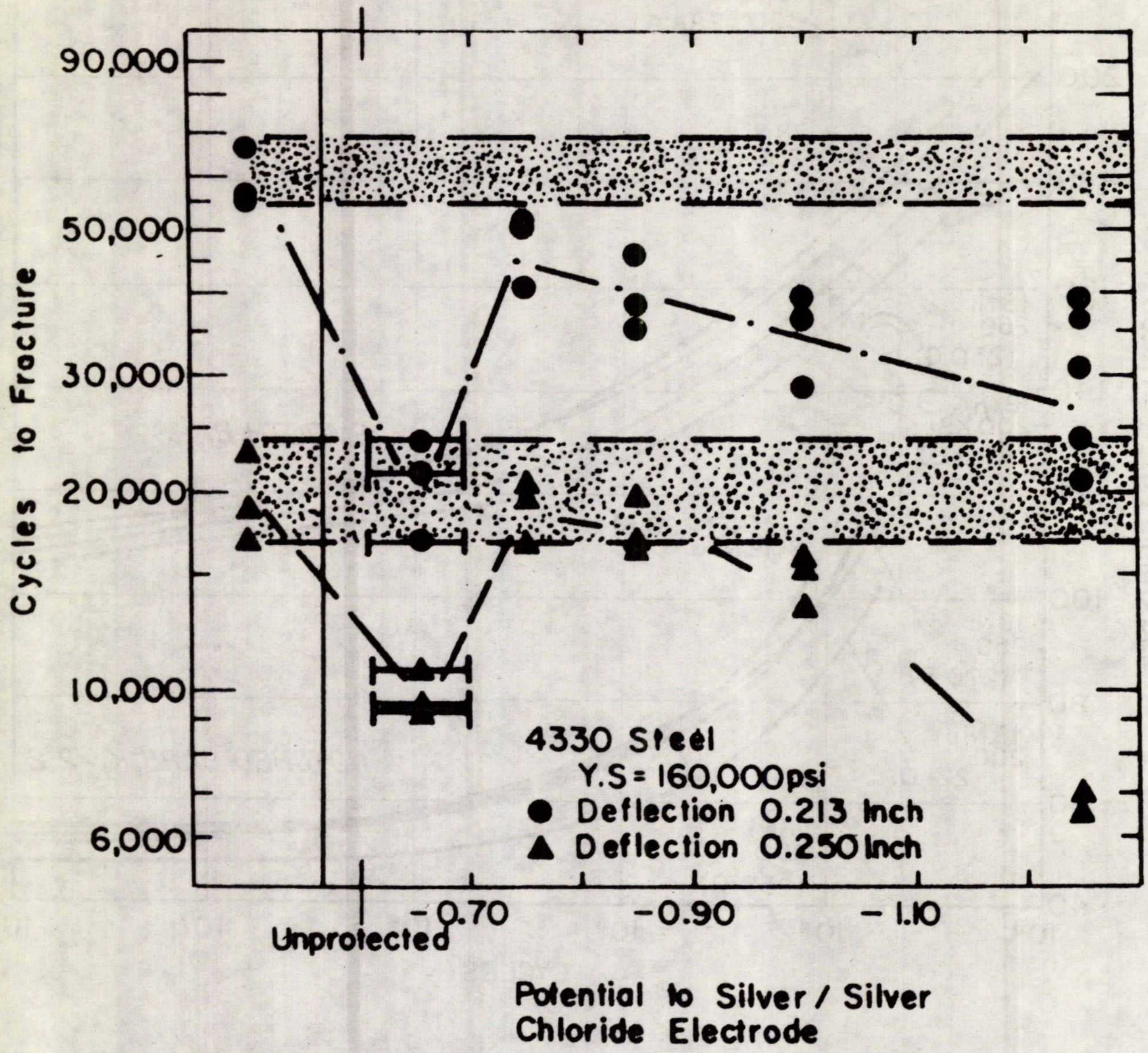


Figure 19. D.C. voltage protection of medium-strength steel (after Brown).

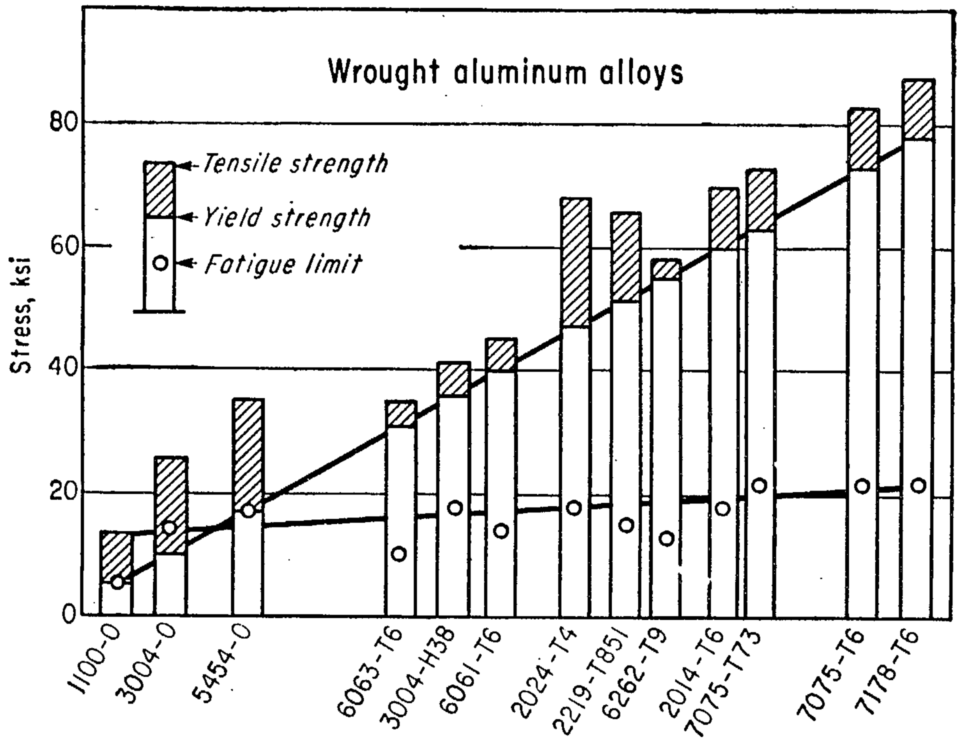


Figure 20. Fatigue properties of aluminum alloys (Stickley and Lyst).

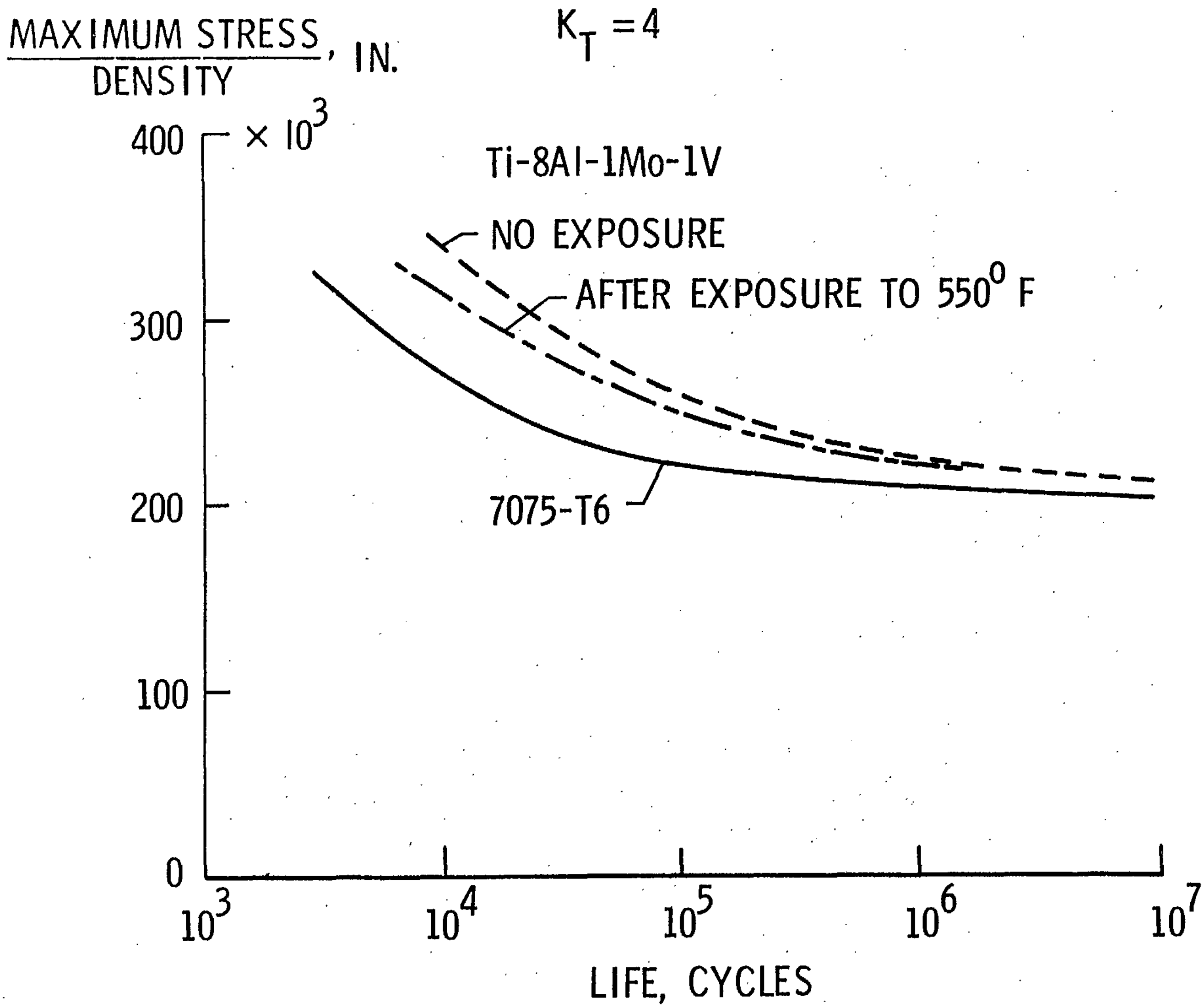


Figure 21. **Fatigue of Ti-8Al-1Mo-1V alloy (Hardrath).**

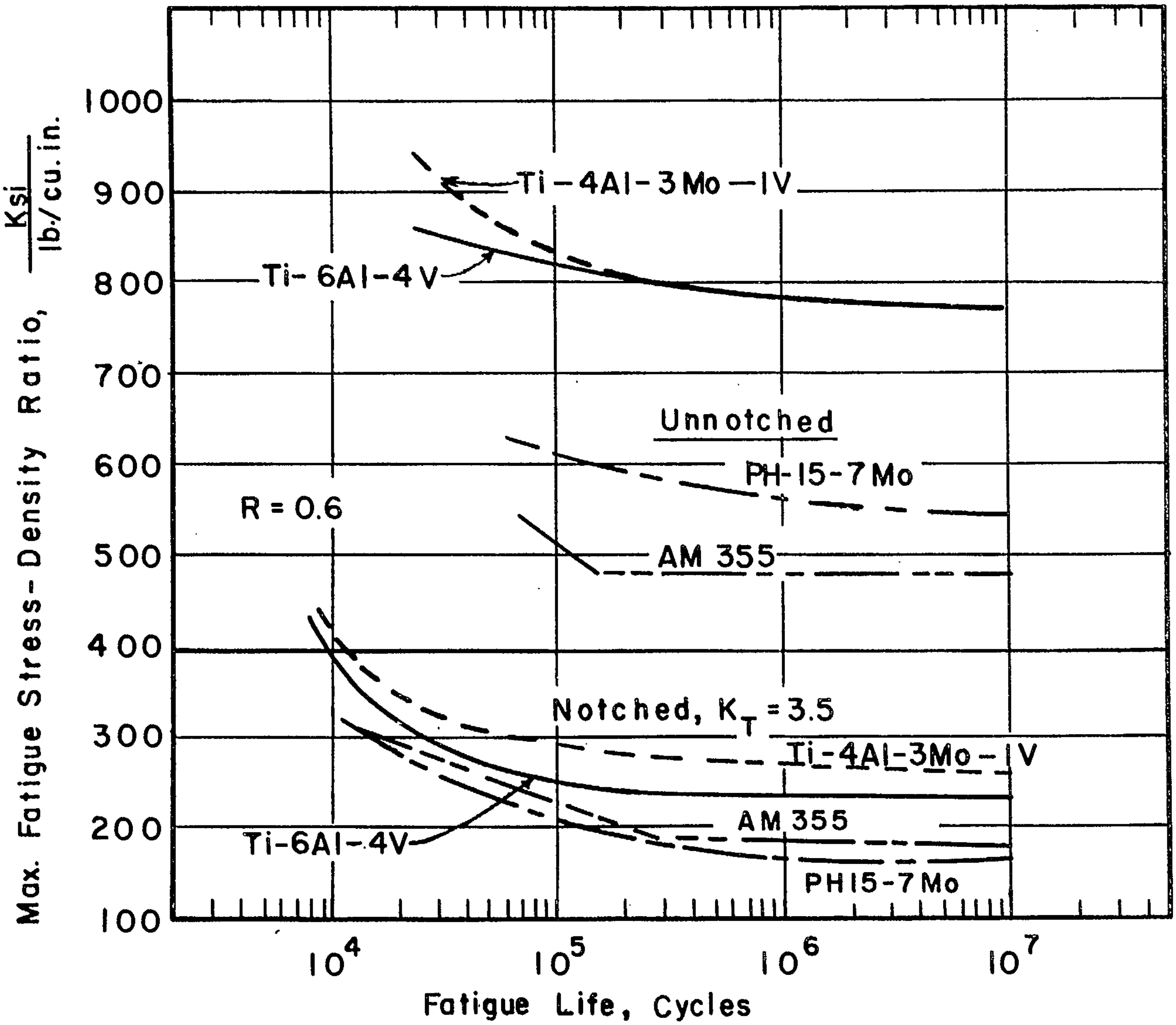


Figure 22. Fatigue of various Ti alloys.

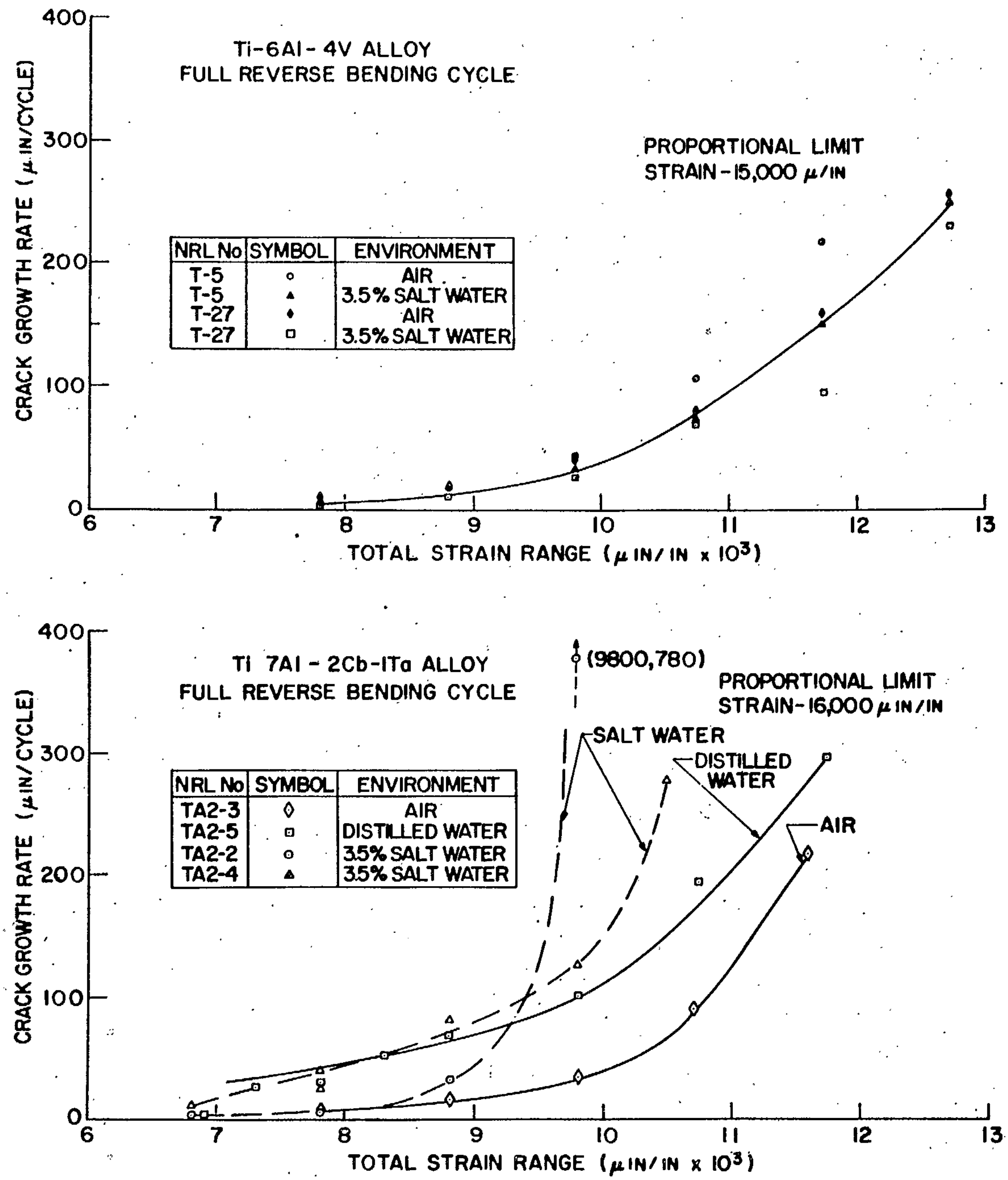


Figure 23. Crack growth rates for 2 Ti alloys (Judy et al.).

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