BASIC CONCEPTS OF FATIGUE DAMAGE IN METALS

By THOMAS J. DOLAN

FOR the past century the problems that may be grouped under the general heading "Fatigue in metals" have grown at an ever increasing rate. This has been unavoidable as the speeds of equipment and driving engines increased and as the life of mobile conveyances was extended. Until the evolution of transportation equipment such as the steam locomotive, airplane, and the automobile (and the accompanying developments in high speed engines and turbines), engineers were little concerned with what we call "fatigue failure." As a logical step in the efficient design of all forms of mobile equipment, it became necessary to reduce the size and weight of all elements such as gears, shafts, connecting rods, etc.; this further tended to increase the peak stresses. In the early days of the steam engine the speeds were so slow and parts so cumbersome that the occurrence of an occasional failure was remedied by replacing the broken part with one made bigger and heavier; thus design for fatigue resistance was not a particularly acute problem. In contrast to this, fatigue strength is of paramount importance (particularly where emphasis is being placed on a high strength-weight ratio) in modern equipment running at high speeds with elements such as turbine blades vibrating at thousands of cycles per second.

In devising formulas for proportioning machine members, engineers commonly use equations based on "idealized" assumptions involving: (a) elastic material that is homogeneous and isotropic, (b) loads applied only a few times and for which the detailed reaction of a member during each loading is identical to that of the previous cycle and (c) a method of processing the member which does not materially alter the mechanical strength. The designer thinks in terms of nominal stresses computed from these idealized conditions though they are only approximations of the behavior of real metals. For loadings applied only a few times, the results are sufficiently

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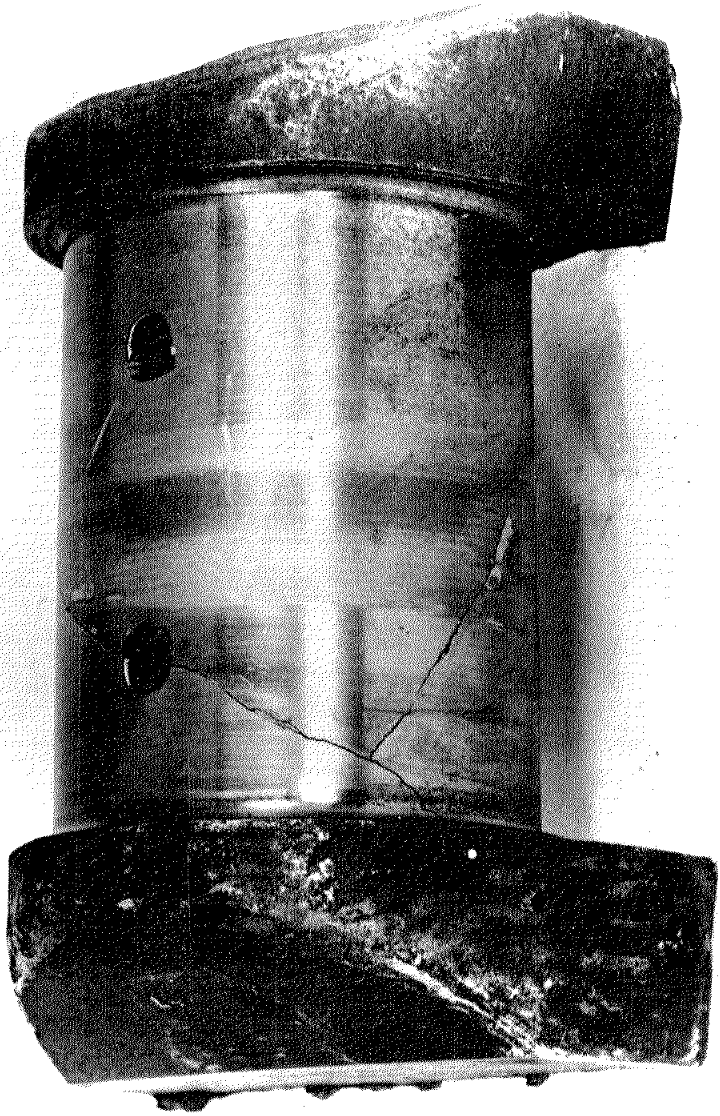

Fig. 1—Torsional Fatigue Failure of Crank Shaft. Failure started at oil hole in crank pin of internal combustion engine.

accurate from an engineering viewpoint; for members subjected to repeated loading, the behavior of commercial metals departs significantly from the idealized conditions.

Unfortunately the present day knowledge of fatigue is not sufficiently advanced to permit design for a specified life or to predict the life of an existing part within close limits. Hence it is necessary to look more closely at the basic nature of fatigue damage in actual metals in the hope that it may become possible in the future and to define quantitatively the conditions which lead to the initiation and propagation of a fatigue fracture.

PROGRESSIVE FRACTURE

When frequent stress fluctuations must be resisted by components such as springs, shafts, gear teeth and turbine blades, stresses far below the maximum static strength are sufficient to lead ultimately to rupture (1).[1] This apparent "weakness" under dynamic conditions as compared with static strength has been the subject of a great deal of study and research as well as of much controversy and speculation. The metal endures the repeated loading for a time (and by superficial observation appears unimpaired) but then suddenly microscopic cracks appear and spread during subsequent loadings as illustrated in Figs. 1 and 2. Finally after the crack has reached sufficient size, a sudden complete rupture may occur. Even in the most ductile of metals the fracture frequently resembles that of a brittle metal since there is no observable plastic distortion of the part as a whole. (See for example the fracture shown in Fig. 3.) About a century ago it was presumed that perhaps the metal had "crystallized" or undergone some subtle change and thus become embrittled (2).

Metallurgists have done much to clarify this mystery by showing that all solid metals are crystalline, and that no extensive recrystallization or grain growth occurs as the result of repeated stressing. The explanation for the apparent difference in behavior under repeated loading lies in the initial nature of fatigue damage and in the accumulative or progressive nature of the damage developing under each repetition of loading. Microscopic fatigue cracks of the types shown in Figs. 4 and 8 are the end result of an accumulation of prior damage on a submicroscopic scale. The whole process of "fatigue" is more aptly and correctly referred to as progressive fracture. The extensive studies of Gough (39) and of Moore (40) have done much to extend our knowledge of the fatigue phenomenon; their many publications in this field have been invaluable in analyzing the mechanism of failure and in ap-

[1] The figures appearing in parentheses pertain to the references appended to this paper.

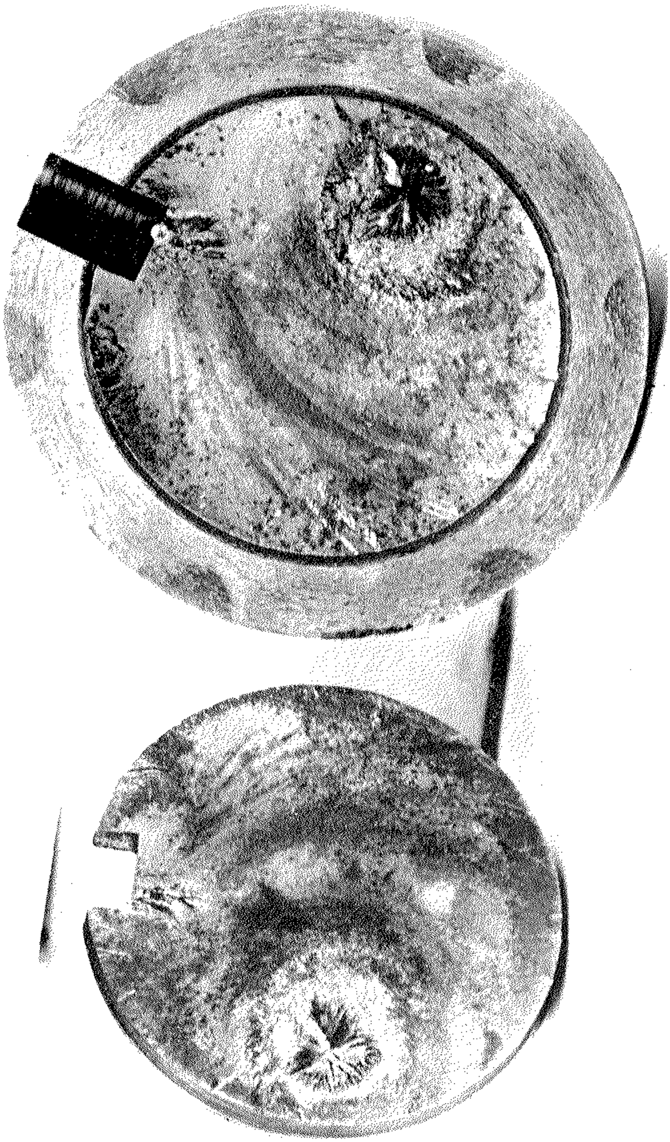

Fig. 2—Flexural Fatigue Failure of Shaft. Failure starting from sharp fillet and keyway developing slowly at relatively low stress levels.

praising the significance of the many factors influencing fatigue strength.

Fatigue failures almost invariably initiate at irregularities in the surface of the metal; these may be due to surface blem-

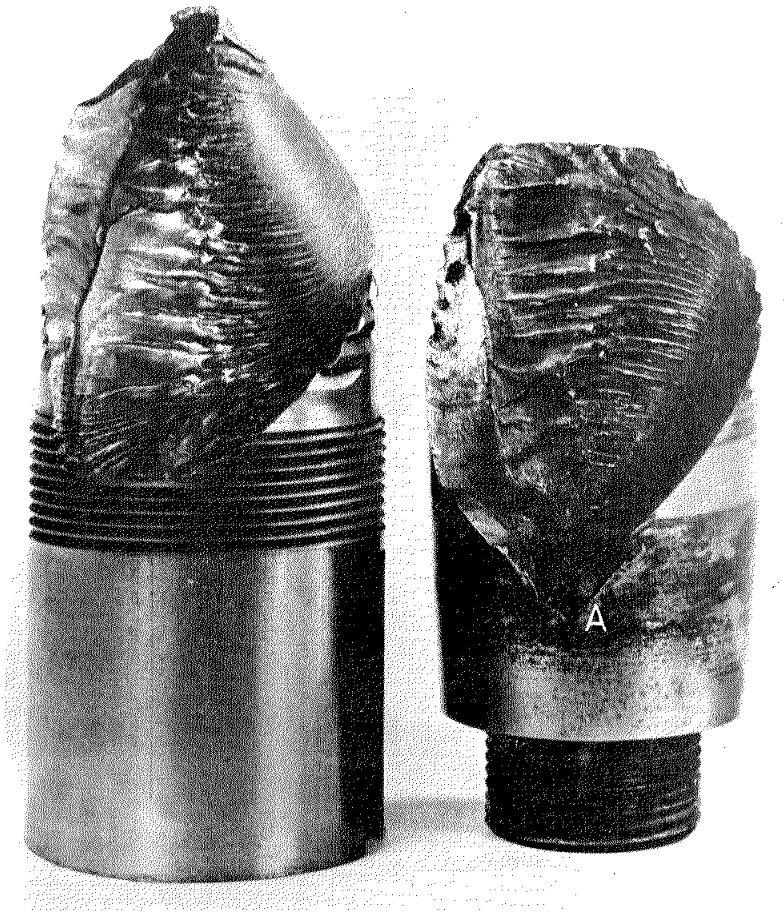

Fig. 3—Torsional Fatigue Failure in Shaft. Fracture initiated in area of fretting corrosion at point A under a hub assembled at this point.

ishes arising from processing, service induced nicks, or the irregular contours of the part itself that are frequently imposed by the designer to fulfill some functional purpose. The localized stresses reach exceedingly high values in the regions of surface recesses and, hence, are usually several times the value of the nominal stress calculated by the designer. When subjected to a single static load, the ductility characteristics of the material are brought into use and inelastic deformation may redistribute the stress over a larger area of the member without resulting in immediate fracture (unless the material is exceedingly brittle). However under repeated stressing a

Fig. 4—Photomicrograph of Fatigue Cracks Initiating from Small Corrosion Pit in Low Carbon Steel.
Note: Fine cracks C propagate across grains on preferred planes which differ in orientation for the various crystals. G indicates grain boundaries × 2200

steady progression of damage develops crystal by crystal throughout the heterogeneous aggregate of grains which make up the member until a large portion of the critical cross section has its structure sufficiently disrupted to initiate microscopic cracks. These progress and upon subsequent cycling, will grow, join, and finally result in a visible crack.

Levels of Observation

What is "basic" in concepts of fatigue damage depends upon the viewpoint of the individual. There are various levels of association of the elementary units contributing to the mechanical properties of a metal and the reactions and conclusion of an observer are prejudiced by the phenomena he records at a particular level of observation. For convenience, these might be grouped in a general manner as follows:

1. Large-scale or phenomenological[2] behavior, characterized by visual observation;
2. Microscopic and submicroscopic phenomena, apparent only by use of special equipment and techniques (such as microscopes and X-ray diffraction patterns);
3. Atomic interactions, which cannot be observed individually and, hence, must depend upon imaginative thinking for hypothetical behavior patterns to explain mechanisms of the structure changes observed at the higher levels of association 1 and 2.

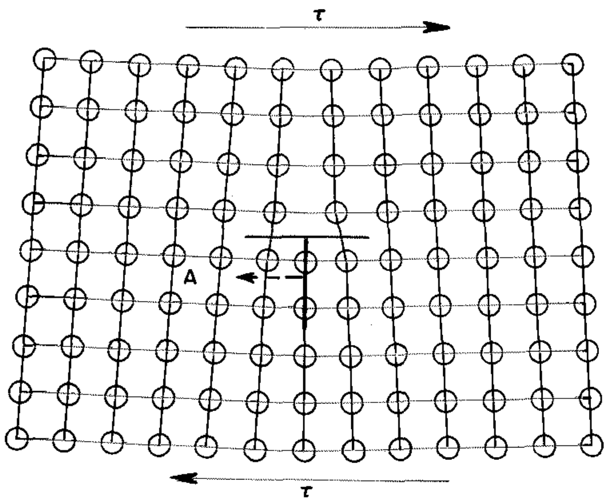

Fig. 5—Hypothetical Diagram of a Dislocation in the Atomic Pattern of the Crystalline Lattice. Slip caused by shearing stress τ is visualized as occurring by a step-by-step movement of the atoms in the direction "A" (to slide the mis-match in arrangement along the slip plane). More complex types of dislocations have been conceived to explain other strength (or weakness) characteristics of single crystals.

Most of the basic concepts of mechanical strength at the atomic level have been built upon dislocation theory from idealized pictures such as that in Fig. 5; movement by slip

[2] Phenomenological refers to characteristic behavior in terms of phenomena that can be measured by large-scale or visible observations. In contrast, events on the microscopic or submicroscopic scale occur in a different manner at an earlier stage in the process and are essential in determining the specific effect of altering important variables.

within a crystalline lattice is visualized as a step-by-step displacement of a dislocation or mismatch in the lattice. Thus far the theory is not sufficiently advanced to deal successfully with polycrystalline metals and the multitude of foreign atoms present in real materials, nor with the complexities introduced by reversal and repetition of stress. Therefore, for the purposes of this discussion, "basic concepts" will refer mainly to recent observations of microscopic structural changes and of accompanying phenomenological behavior.

Why Are Metals Weak in Fatigue?

The heterogeneous nature of polycrystalline metals is easily recognized in micrographs such as those shown in Figs. 4 and 6. That is, metals composed of an aggregate of randomly oriented and shaped crystalline grains are far from homogeneous. When loads are applied to a part, the elastic strength of individual crystals may be readily exceeded even at nominal stress levels far below those usually thought of as being the elastic strength of the polycrystalline aggregate. Localized deformations referred to as "slip bands" develop within crystals or parts of crystals as shown in Fig. 6. In the early stages these are sometimes seen only in a few scattered grains. As the stresses are gradually increased in a static test, the slip bands multiply until practically all grains are greatly fragmented and deformed.

In contrast to this, however, repeated loading at stresses below the nominal yield point of the material localizes these effects in a few crystals in regions or zones of high stress concentration. Thus the significant actions that lead to fatigue fracture originate in a more localized region than can be accurately treated by means of concepts based on homogeneous elastic material. Under repetitions of stressing the fragmentation and crystal break-up continues until eventually submicroscopic cracks form; in the early stages these are usually closely associated with the slip bands as shown in Figs. 7 and 8. As these disruptions develop, they accelerate by jumping together in small groups until a microscopic crack is formed. Thus the mechanism includes an accumulation of chance effects in nucleation and growth and is quite different from that encountered under a single static load. The microscopic readjustments that initiate in weaker crystals are multiplied in

Fig. 6—Slip Bands in Crystals of High Purity Aluminum After Repeated Stressing. Lines L are Slip Bands; G is a Grain Boundary. × 300.

their severity by the repetitions of loading until fractures are developed at relatively low nominal stresses. Hence the metal appears to be weak as compared with the usual concepts of static strength. Because of the heterogeneity of the polycrystalline metal, the response of a part to an external load will necessarily be that of a model built up statistically of ele-

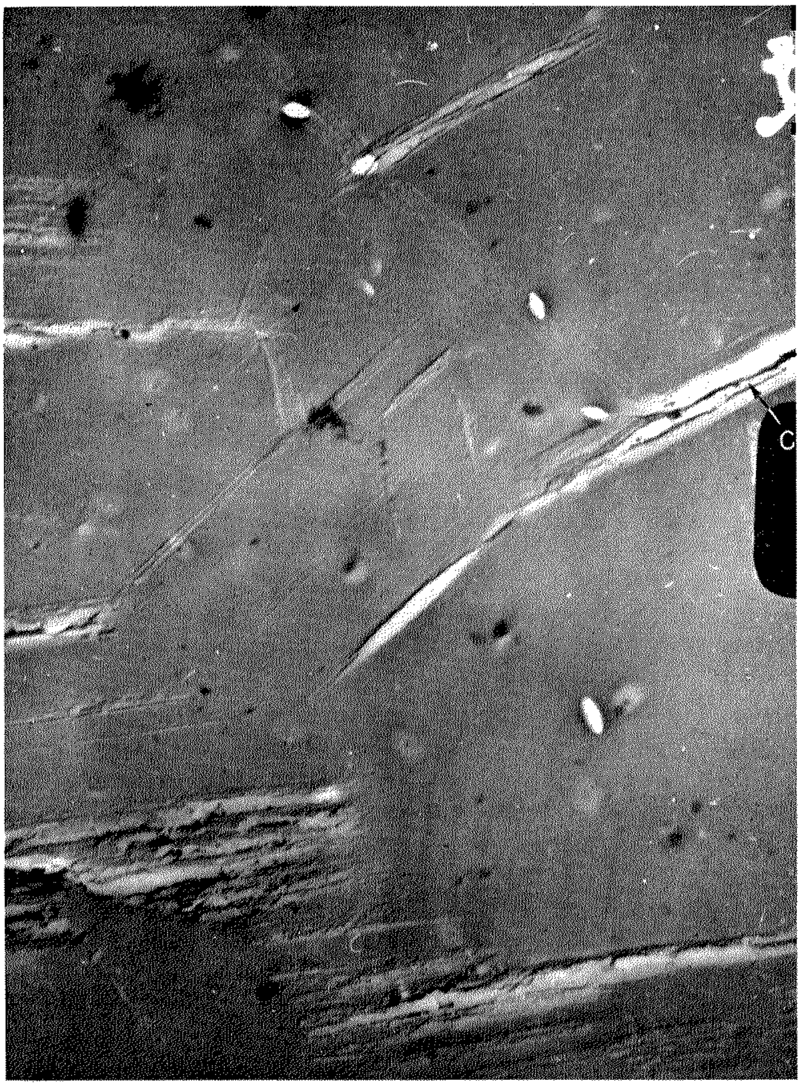

Fig. 7—Electron Micrograph of Alpha Brass. After reversed bending at 23,000 psi for 1,023,000 cycles. Deformation bands and fine cracks show up in the unetched electropolished surface. Note the random localized appearance of the deformation. Dark lines within the zone "C" are initial microscopic cracks. Craig (3). × 11,000.

ments that have a range of values in size and whose strength varies as a function of size and orientation of the elements. The engineering concepts of a "fatigue limit" and a "fatigue life" therefore need to be stated in terms of a probability to recognize the fact that the development of damage is a statis-

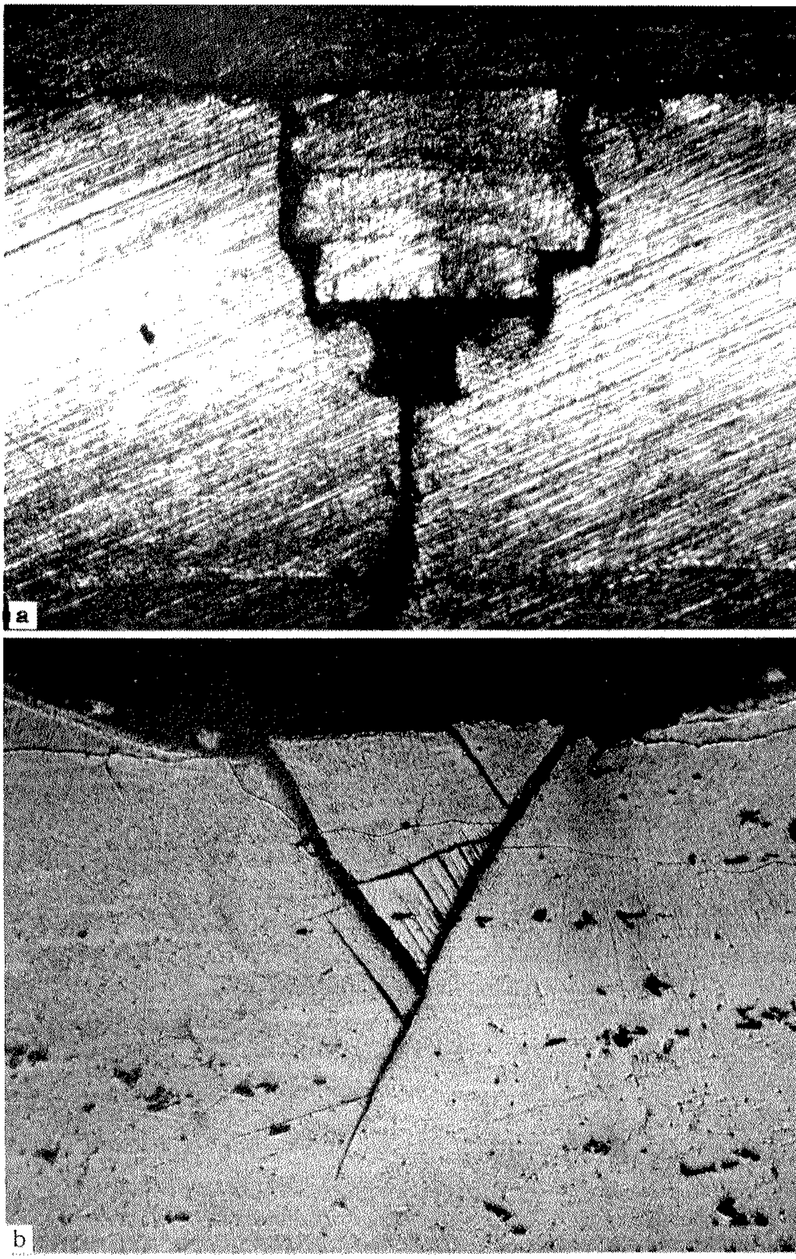

Fig. 8—Fatigue Cracks Developing on Planes of Maximum Shearing Stress. (a) Longitudinal and circumferential cracks in cylindrical specimen of 75S-T aluminum after reversed torsional stressing. × 100, unetched (b) Cracks at root of small notch in beam after 41 million cycles of a nominal stress cycle from zero to 19,000 psi *compression*. Cracks developed and material was displaced along shear planes in the compressively loaded region. × 600, Keller's etch.

tical process depending largely on chance and on the distribution of stresses in the weakest crystalline zones.

The weakness of metals under repeated loading is not surprising when consideration is given to the fact that metals contain the following inhomogeneities that locally lower the resistance to mechanical stressing:

(a) Random agglomeration of anisotropic crystals of wide variation in shape and size.

(b) Mixed phases of different elastic characteristics. The "mixed phases" may consist of inclusions differing physically from the matrix and crystal imperfections in lattice or atomic pattern. Any discontinuities or segregation streaks may also be visualized as large-scale evidence of mixed phases. Textural stresses on a micro scale are undoubtedly developed at interfaces between the various constituents because of their differences in physical and mechanical characteristics.

In general, fatigue cracks initiate in the exposed surface of a metal part unless unusual circumstances or processing treatments (such as nitriding) develop surface layers of substantially higher strength than the inner zones of metal. Apparently the exposed surfaces are inherently weaker than interior grains that are confined by bonds with adjoining grains. Analyses of the ideal crystal usually consider the lattice to extend indefinitely in all directions, whereas they may possess a free surface at which the interatomic equilibrium must be altered. Imperfections of atomic arrangement must form the surface layer on all bodies due to the existence of the free surface. A marked increase in fatigue strength has been reported after shot peening members subjected to axial loading. Since all fibers are subjected to essentially the same imposed stresses, the interior grains must have been much stronger than the surface to account for the increased strength. The shot peening strengthens only the surface layers; any residual stresses induced in the interior could only be tensile stresses that would actually tend to lower the strength.

Micro-Mechanisms of Progressive Fracture

Electron micrograph and X-ray studies in recent years have done much to clarify the mechanism of fatigue damage in its early stages and to differentiate between the effects of static and dynamic stressing (3,5,6,7). Cyclic stressing develops localized slip bands that are grouped or congregated into a striation (as illustrated in Figs. 6 and 7) which tends to increase in width by development of added slip on adjacent planes with continuous cyclic stressing. Forsyth (5) indicates that the spacing between striations is that of the set of slip bands produced by the first stress cycle and that the number of deformed zones appearing on the face of a crystal is proportional to the stress level in the particular region involved. Since the striations grow wider at approximately the same rate, a few highly stressed grains may be eventually covered by deformation bands because the original slip spacing was dense. He finds that each stress cycle produces some additional slip but the number of visible bands is not directly proportional to the number of stress reversals. During the first few cycles a large number of slip bands form by a sort of avalanche process. However between the deformation bands or striations the crystal sometimes exhibits slight surface corrugations running in the direction of the slip bands; these may be due to a more homogeneous form of slip which leaves no microscopically visible bands. At high frequencies of stressing the slip bands apparently group to a very marked degree in isolated striations.

In separate studies Love (7) and Craig (3) also found that fatigue and static strain effects were dissimilar in that the structure changes caused by repeated stressing were more localized in nature and tended to concentrate in only a portion of the available grains. The visible slip produced tended to occur in localized bands which frequently affected only a small portion of the crystal. With the aid of the electron microscope, minute fatigue cracks could be detected at an extremely early stage of the fatigue process (as low as one-tenth of 1% of the normal fatigue life) and these were associated with slip striations of the types in Fig. 7.

Most of the early observations of deformation and work hardening were based on behavior of statically stressed crystals; it is not surprising to find that cyclic stressing may alter the appearance or extent of slip that occurs. Upon reversal of load one might presume that either reversed slip could take place on the original planes or that slip may occur on neighboring planes. Forsyth (8) offers conclusive evidence that the slip progression develops alternately on closely spaced neighboring planes. Wood and Head (6) showed by using X-ray methods that a large amount of the lattice distortion produced on the positive half of a stress cycle may be recoverable upon completion of the negative half of the cycle. The higher the frequency of cyclic stressing, the less the initial deformation and the more complete the recovery.

Later X-ray studies by Wood and co-workers (9) confirmed the localization of deformation leaving large parts of the grain undisturbed. They found the total strains developed by a rapidly applied stress to be comparable with that produced by the same stress applied slowly, but concentrated into fewer regions of the grain. Localization of the strain occurred only when the rate of cyclic stressing exceeded a critical value, this evidence of a "delay period" may explain the localization of slip in striations observed by Forsyth and by Craig; rapid cyclic stressing serves to excite only the more easily activated slip regions of the weaker crystals. Cyclic stressing develops continued movements in the local zones but these are not necessarily accompanied by local strain hardening.

Deformation by a simple glide process on a crystallographic plane as illustrated in Fig. 5 may cause no major structural change since only a translation of lattice position occurs; this could take place without work hardening. That is, slip could proceed as a movement of dislocations under the action of shear with the dislocations being dissipated at the ends of the slip bands. Apparently it is this type of deformation that may occur between the striations of intense deformation observed by Craig and by Forsyth; however, the concept of simple slip by movement of dislocations in the lattice may be over-simplified when one considers the polycrystalline aggregate. Rotations and grain boundary movements may materially alter the picture as well as the complexity introduced by reversed slip on closely spaced and interleaved layers of the lattice within the crystal. Particularly when tested at high frequencies, the microscopic appearance of the slip bands suggests that they may have experienced a considerable rise in temperature because of the presence of what appears to be stains or granules of oxide in the striations (8). If these localized zones attain temperatures considerably above room temperature, it would readily explain the fact that slip in fatigue is confined to the striations where considerable thermal softening has occurred. The heating would also have the effect of lowering the cohesive strength of the metal in these regions. Thus the usual concepts of work hardening and embrittlement may not be the primary cause of fatigue cracking; instead a progressive piling up of an avalanche of disruptions accompanied by localized heating sets up a chain reaction which leads to local fragmentation without embrittlement of neighboring zones.

The above observations present a rather consistent picture of the progressive nature of the damage occurring during repeated stressing long before the formation of microscopic cracks, and emphasize the localized nature of the breakdown as contrasted to the general yielding that occurs under excessive static loading.

Statistical Nature of Fatigue Damage

The foregoing evidence indicates that metals are not stable but undergo re-adjustments on a microscopic or submicroscopic scale throughout the process of repeated stressing. These instabilities and chance changes in structure lead to highly diverging results in phenomenological tests. When one observes in a microscope the marked heterogeneous appearance of a metal involving irregularly shaped grains, the presence of complex alloying constituents, oxides, nitrides, etc., it becomes obvious why it is futile to prescribe a definite fatigue limit for a material. Each of the elementary crystals and its cross links with neighbors has a different characteristic strength behavior and, hence, when loaded may vary widely in its reaction to the imposed loading. As soon as the cumulative damage is sufficient to nucleate a small microscopic crack, it may rapidly propagate into neighboring crystals. Appraising progressive fracture from this viewpoint, it is not difficult to see why metals are weak in fatigue. The real problem, however, is to predict the most probable strength of a part and to take into consideration the many factors affecting this strength (both those imposed by the service condition and those inherent in the metal).

Recent studies indicate that most of the scatter observed in finite fatigue life testing of metals is an inherent characteristic of the material and does not necessarily indicate poorly adjusted machines or improper testing techniques (4,10,11). The distribution of the fatigue life of groups of similar specimens is quite varied but has been found in several instances

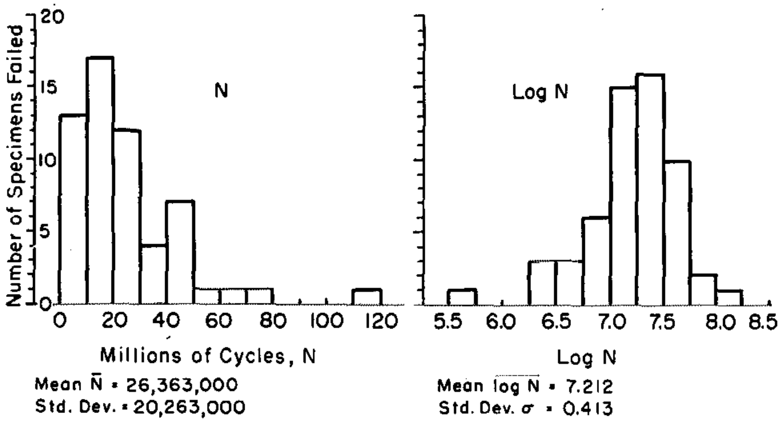

Fig. 9—Distribution of Finite Fatigue Life for 57 Small Specimens of Aluminum Alloy 75S-T Tested in Reversed Bending at 30,000 psi. Sinclair and Dolan (10)

to closely approximate a logarithmic normal distribution as indicated in Fig. 9. The amount of scatter, which is usually measured by means of the standard deviation in life, is related to the stress amplitude (being smallest at high stresses and largest at stresses just above the fatigue limit). Therefore one cannot legitimately speak of the fatigue limit in terms of a single value (11) but must regard the S-N curve more as a functional relation of a family of curves, each of which indicates a definite probability of failure, P, as shown in Fig. 10. In this figure it is obvious that even under carefully controlled laboratory conditions using duplicate test specimens from the same bar, it is not unusual to encounter ratios of ten to one in cycles of a given stress to fracture. Thus designing for precise specified life is not a practical procedure.

The conventional S-N curve as normally obtained with 10 to 15 specimens must be regarded as a crude attempt to determine the 50% probability of failure line; for given conditions of overload, it gives little or no information about the life that may be expected for a very low probability of failure. However it is seldom economically feasible to run enough experiments to obtain the statistical information necessary to design a part for limited life with a predictable degree of certainty. Furthermore a consideration of the many variables and uncertainties that occur in practical service conditions makes it impossible to predict a precise fatigue life for a member such as an airplane wing, a turbine blade, or a gun tube.

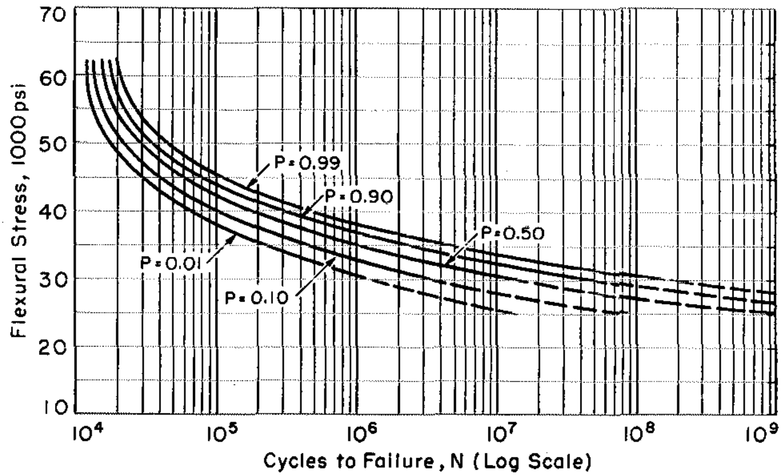

Fig. 10—Composite S-N Curves for Various Probabilities of Failure, P. Derived from data for small unnotched specimens of 75S-T aluminum alloy. Sinclair and Dolan (10)

PHENOMENOLOGICAL EVIDENCE OF FATIGUE DAMAGE

In recent studies (12,13) it has been postulated that the fatigue process might be characterized by three stages; namely, Stage 1 during which crystal slip and fragmentation take place and are accompanied by work hardening until the so-called "limit" of cold work is reached; Stage 2 during which a statistical disruption of the crystalline lattice takes place (breaking high-energy atomic bonds in the severely cold-worked striations of slip within the crystals) thus resulting in the formation of submicroscopic cracks; and Stage 3 during which the submicroscopic cracks coalesce to form visible spreading cracks which result ultimately in fracture. In this sequence it appears that a true fatigue damage occurs only during Stages 2 and 3 since work-hardened metal is not necessarily "damaged" metal. In fact, rolling or drawing operations normally improve the fatigue strength of a metal. Many observations have indicated that once visible cracks appear, they spread very rapidly, and thus Stage 3 is of relatively short duration in comparison with the total number of cycles required to form a crack in the member.

Some years ago Haigh (14) investigated the mechanical hysteresis in metal specimens subjected to repeated loading and discussed these same three characteristic stages in the fatigue process. He found the first stage to be of comparatively short duration. Microscopic plastic flow was accompanied by a relatively high rate of hardening of the metal, and the evolution of heat was detected by means of thermocouples. The temperature rise was gradual at low stress levels but jumped up abruptly at high stress levels. The second stage,

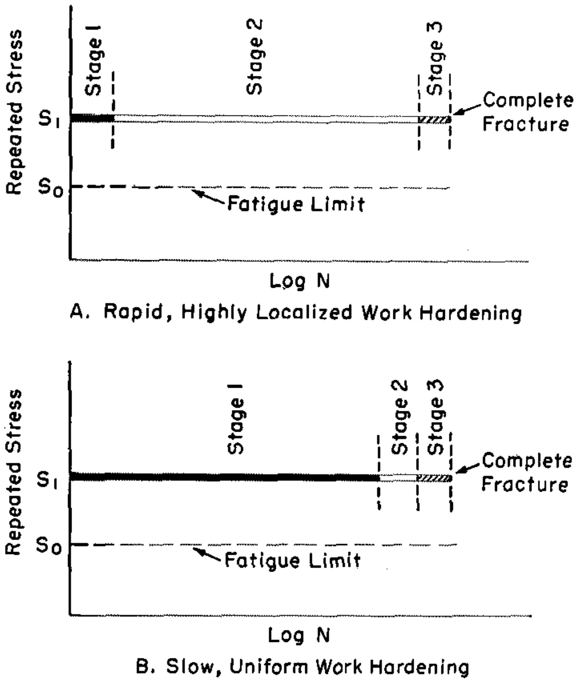

Fig. 11—Relative Duration of Work Hardening Stage in Fatigue According to Possible Mechanisms A and B (Schematic)
Stage 1—Work hardening
Stage 2—Formation and growth of sub-microscopic cracks
Stage 3—Growth of visible cracks
Sinclair and Dolan (12)

extending over the greater part of the test, was characterized by relatively constant thermal energy dissipation that was not accompanied by pronounced hardness changes. The third stage was again relatively short directly preceding final fracture, and was evident by a rapid increase in heating accompanying the appearance of visible cracks. Slip bands form copiously during the first stage but not appreciably during the second and third stage.

It is of practical importance to examine in more detail the relative duration of Stage 1 (the work hardening stage) as cartooned in Fig. 11. The fact that the effects of work hardening from rolling or drawing operations may be removed by annealing would also be important in removing fatigue damage if Stage 1 was comparatively long as shown in Fig. 11b. However if the work hardening stage was short as indicated in Fig. 11a, it would be difficult to re-anneal the member frequently enough to remove the damage before entering Stage 2.

Data obtained to check these hypotheses are plotted in

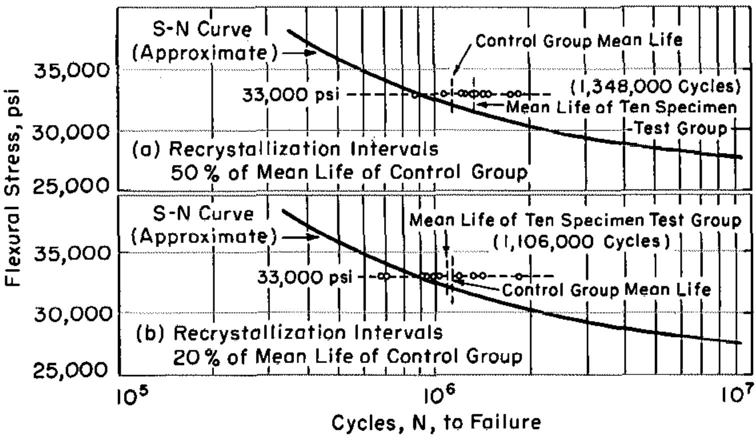

Fig. 12—Distribution of Fatigue Life for Test Groups of Alpha Brass Given Recrystallization Heat Treatment. Sinclair and Dolan (12)

Fig. 12. A control group of 15 specimens of alpha brass were run at a single stress level to determine the mean life and standard deviation in life for the series. Two other groups of 10 specimens each were reheated to the recrystallization temperature at intervals of 50% or of 20% of the mean life of the control group. For the group reheated at intervals corresponding to 20% of the mean life, the average life was only slightly less than that of the control group. The test group reheated at intervals of 50% of the control group mean life had a slight increase in average life. However a statistical analysis of the data indicates that these differences in the mean life could have occurred by chance sampling one out of every ten times. Accordingly, there was little reason to believe that the periodical heat treatment produced a significant change in the fatigue life of either test group. It was therefore concluded (12) that a true fatigue damage corresponding to actual disruption of atomic bonds and formation of submicroscopic cracks is initiated at a relatively early stage of the fatigue

life. Any stage of fatigue during which only work hardening occurs appears to be of very short duration.

The exact nature of "fatigue" damage prior to the appearance of a visible crack and the actual prediction of fatigue life of a member have long been problems of great interest. At the present time estimations of the probable fatigue life of a member that is subjected to repeated stressing at several different amplitudes are usually based on the assumption that the damage at each stress level is directly proportional to the number of cycles of that stress (38). That is, the total cumulative damage is assumed to be proportional to the sum of the cycle ratios for all stress levels encountered.[3]

The observations of Haigh and others would lead one to doubt the accuracy of the assumption that the fatigue damage at any one stress level is proportional to the number of cycles of stress since the rate of damage may be quite different in each of the three stages described above. Later experimental evidence (15-17) also indicates that this assumption is probably inaccurate since the cumulative cycle ratio for failure of a member can be changed merely by reversing the sequence in which two different amplitudes of repeated stress are applied.

For example, Bennett (15) found that if a specimen is stressed first to 60% cycle ratio at a stress S_1 and then run to failure at a higher stress level S_2, the remaining life at the stress S_2 is usually greater than the 40% cycle ratio predicted on the basis of the cumulative cycle ratios. If the specimen is stressed first at the level S_2 and then run to failure at a lower stress S_1, the remaining life is found to be less than that predicted. Other studies of the type shown in Fig. 13 have also indicated a wide divergence in cumulative cycle ratios for failure; they serve to illustrate the scatter due to the statistical nature of the fatigue life of individual samples of a controlled group. There is little likelihood that any accurate method for predicting the fatigue life of any individual part can ever be evolved even for simple conditions of overload.

Since machine members in service often encounter wide fluctuations in load, studies have been conducted to measure the cumulative fatigue damage produced by several simple types of stress history in which the amplitude of stress is changed at

[3] Cycle ratio is defined as the ratio of number of cycles at a given stress amplitude, to the number necessary to cause failure (as represented by the life to the ordinary S-N curve) at that amplitude.

regular intervals. In Fig. 14 some results of tests are shown in which 1000 cycles of a major stress S_A were alternated with 9000 cycles of a lower stress S_a. These indicated many

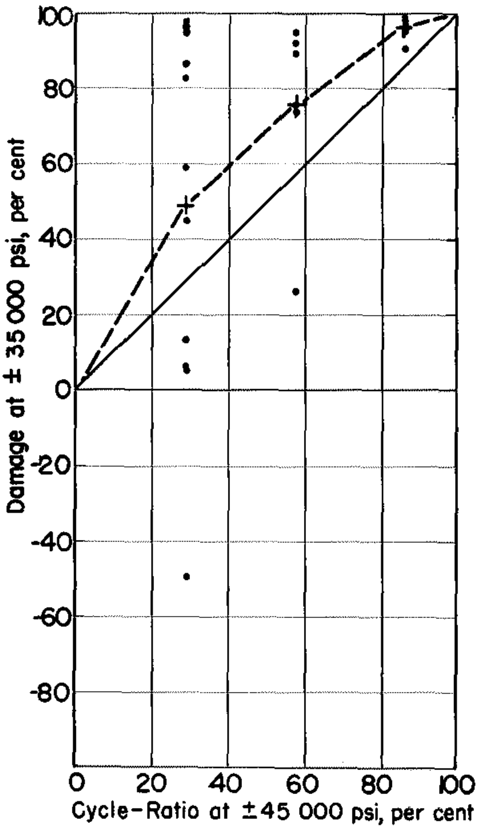

Fig. 13—Damage at Test Stress Due to Prestressing at 45,000 psi for 75S-T Aluminum Alloy. The "damage" is expressed as the percentage reduction in life (or cycle ratio) at the test stress after prestressing at 45,000 psi. Note wide scatter for individual tests; crosses indicate average values of ordinate. If damage were proportional to cycle ratio the points would all fall along the diagonal solid line. (17)

puzzling trends which require clarification by further study. For example, it was found that the fatigue limit was not definitely a unique property of the steel; it may be altered by a previous stress history. Moreover, as Fig. 14 indicates, an increase in the lower stress level from 60,000 to 78,000 psi resulted in decreased life, even though these lower stresses were below the original "fatigue limit" of 87,000 psi, and normally would not be expected to cause damage. For extremely high values of the major stress, S_A, the total fatigue life was approximately that predicted by assuming failure to occur when the cumulative number of stress cycles of only this major stress reached the normal S-N curve. The dashed line labeled "10 x S-N" indicates the expected life if only the cycles of the major stress caused failure. In general, the amount of damage caused by repeated stressing is not a simple linear function of the number of cycles of stress.

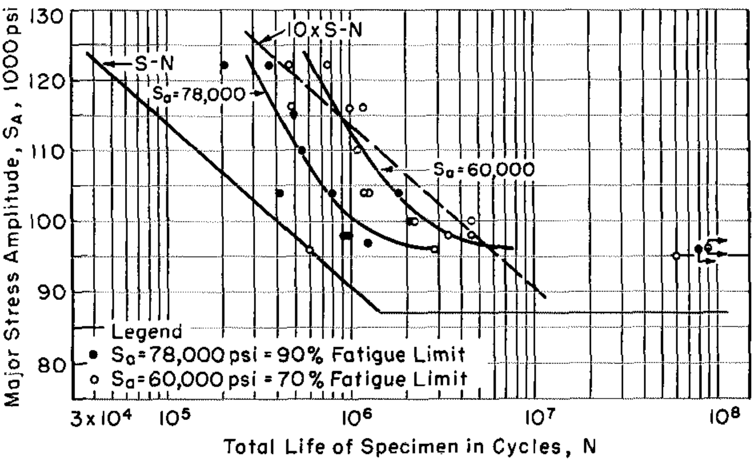

Fig. 14—Fatigue Life of SAE 4340 Steel Subjected to 1000 Cycles of Stress S_A after Each 9000 Cycles of Stress S_a. Points are plotted at values of the major stress in the test, and for the total life including all cycles at both stress levels. Arrows indicate specimens that did not fracture. The solid line labelled "S-N" represents the normal mean S-N curve for these specimens.

It has long been known that the fatigue limit of some metals may be improved by "understressing" followed by a process of gradually increasing the amplitude of the alternating stress in small increments as indicated in Fig. 15, a procedure ordinarily called "coaxing."[4]

Sinclair (18) has suggested that this coaxing effect is governed by the time-dependent localized strengthening that occurs through strain aging and is not related to the ability of the metal to be strengthened by cold work. In general, any process which will prevent or quickly arrest the continuation of slip in the metal will also increase its fatigue strength. Strain aging may be observed in mild steel, for example, when it is deformed plastically in tension by a small amount

[4] Understressing is usually referred to as the repetition of stresses at a level below that of the ordinary fatigue limit of the material.

and then allowed to rest or age at room temperature for several days. On retesting the metal, it is found that the tensile elastic limit of the metal has been increased to a value greater than that of the unstrained material. If strain aging ocurs, the increase in strength is larger than that accounted for by work hardening alone. Those metals which have a matrix of ferrite capable of strain aging will have their fatigue resistance improved by a coaxing procedure. Those which are work hardenable but which have little or no

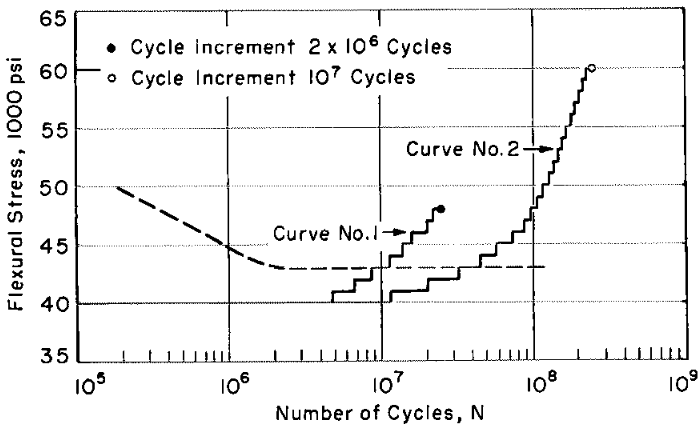

Fig. 15—Influence of Cycle Increment on Coaxing Failure Stress of SAE 1045 Steel. Curve No. 2 indicates an increased fracture stress from the greater coaxing occurring with longer time of cycling at each stress level.

capacity for strain aging show no coaxing effect. Either decreasing the stress increment or increasing the number of cycles at each stress level (as in Fig. 15) will increase the dosage of the coaxing effect by providing the material with more time during which the strain aging process can continue. Slight plastic deformations that produce slip in the weaker crystals at low stress levels may stimulate the microscopic precipitations or changes in metallurgical constituents that result in "aging" and thus gradually improve their fatigue resistance by strengthening the weakest elements of the structure.

INFLUENCE OF MICROSTRUCTURE ON FATIGUE STRENGTH

In view of the large variety of constituents and the complexity of structure of most polycrystalline metals, it is not surprising that quantitative predictions of strength properties based on microstructure alone are frequently inaccurate. However for simple single-phase metals the resistance to slip (and hence the resistance to initiation of fatigue damage) may be related to the sizes of the polycrystalline grains (19).

In general the bonds between atoms at the interface between two crystals appear to be stronger than within the crystal grains (except at high temperatures). The intercrystalline boundaries thus tend to slow down or inhibit either slip

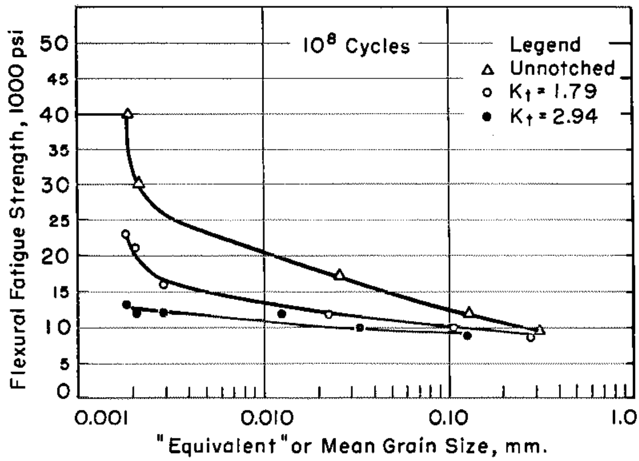

Fig. 16—Influence of "Equivalent" or Mean Grain Size on the Fatigue Strength of Alpha Brass at 10^8 Cycles. Specimens plotted with circular symbols had a sharp notch for which the value of Kt is the theoretical stress concentration factor. Karry and Dolan (22)

or crack propagation at the microscopic level. Hence one might expect that the larger the number of "road blocks" (due to fineness of grain structure) the larger the number of fatigue cycles required to produce failure. Localized precipitation of constituents within the grain has a similar effect and is of importance in the study of the more complex alloys. For the single-phase metals it thus seems logical to assume that the strength properties are proportional to the reciprocal of the mean grain diameter as suggested by the fragmentation theory of Bragg (20,21). Confirmation of this hypothesis is indicated by the fact that for an alpha brass the fatigue strength is dependent upon the largest (or weakest) crystals present and is improved as shown in Fig. 16, by processing treatments that reduce the grain size. As is indicated in Fig. 17 the fatigue notch-sensitivity q is also increased for fine-grained metal

(22). This is perhaps evidence that the fine-grained poly-crystalline metal behaves more nearly as a homogeneous unit (and approaches the "idealized" conditions assumed for purposes of stress computation).

In the region of a sharp notch there is a steep stress gradient such that the average stress on a large grain may be considerably smaller than the peak stress at the surface. In

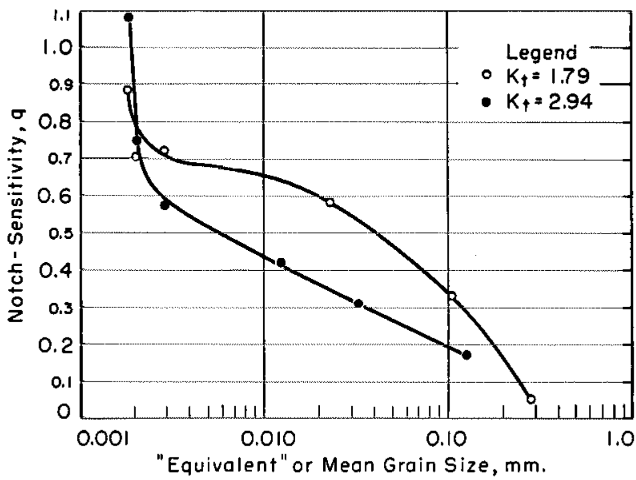

Fig. 17—Influence of "Equivalent" or Mean Grain Size on Fatigue Notch-Sensitivity of Alpha Brass at 10^8 Cycles.

$q = \frac{K_f - 1}{K_t - 1}$ in which K_t is the theoretical stress concentration factor, and K_f is the experimentally determined strength reduction due to the notch. The full theoretical strength reduction ($q=1.0$) is approached for the finest grain sizes. Karry and Dolan (22)

Fig. 18 a crude attempt has been made to calculate the approximate average stresses on grains at the surface of the brass specimens for the data plotted in Figs. 16 and 17. The close grouping of these curves seems to indicate a fair agreement between the average stresses on the most critically stressed grains in the unnotched specimens and those in the notched specimens; they also illustrate the general trend to lower fatigue strength as the grain size is increased.

Cold working of a single-phase metal has the effect of reducing the crystallite size and thus of improving the fatigue strength of the metal. There exists a considerable difference of opinion as to whether the increases in fatigue strength observed after localized cold working operations such as shot peening or surface rolling are due to the accompanying favorable macro residual stresses induced or whether the improvement is due primarily to increased mechanical strength (that is, due to work hardening of the surface layers). Experiments that have been planned on a phenomenological scale have never been able to separate entirely the individual effects of inelastic deformation and of macro residual stresses. It appears probable that both effects are important though not to the same degree.

When one considers factors of grain size in the ferrous metals or the precipitation hardening alloys, the conditions are

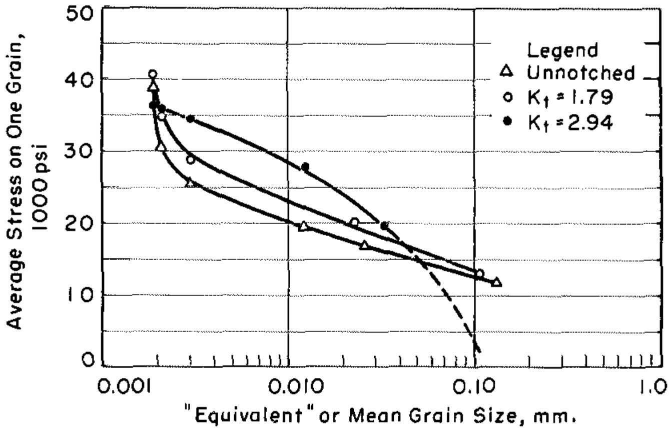

Fig. 18—Average Stress on One Grain Located in the Critically Stressed Surface of a Notched or Unnotched Specimen (based on 10^8 Cycles of stress) For the same Data as that Plotted in Figs. 16 and 17.

so drastically altered by the complex constituents present that the grain sizes as conventionally defined have little influence on fatigue properties. The majority of the commonly used structural metals and alloys are more logically regarded as metallic aggregates consisting of a hard phase dispersed in a softer one. For these it has been suggested that the resistance to inelastic deformation is proportional to the logarithm of the mean straight path through the continuous phase (23); that is, the amount of hard particles is of less consequence than the mean free path for slip to occur from one hard particle to another in the soft matrix. Thus the prior austenitic grain size or ferrite grain size alone may have little influence on the fatigue strength of steels.

The fatigue limit of polished laboratory specimens of steels ranges from about 0.4 to 0.6 of their static tensile strength as indicated in Fig. 19. It would be exceedingly advantageous to be able to obtain the upper limit of 0.6 in all cases since this is a 50% increase in fatigue limit over the lower nominal value of 0.4. Several studies (24,25) have indicated that the fatigue strengths were improved for a variety of steels when the heat treatments involved relatively rapid cooling to produce finely dispersed hardening constituents. Fig. 20 shows that the fatigue strengths of both notched and unnotched specimens were raised when the microstructure consisted mainly of

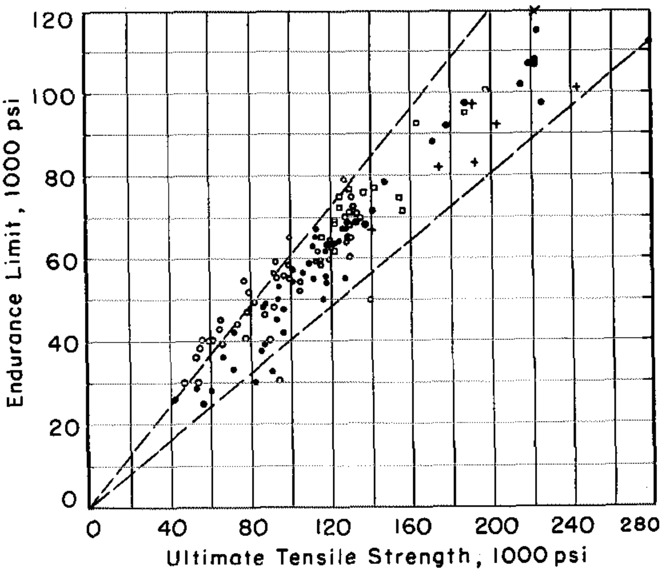

Fig. 19—Relation Between Endurance Limit and Tensile Strength for Ferrous Metals (24). Dashed Lines represent ordinates of 0.4 and 0.6 of the tensile strength.

tempered martensite rather than of more coarsely dispersed carbides.

The existence of "weak" areas of coarse pearlite or of free ferrite results in fatigue limits that are inferior to those obtained for a tempered martensitic structure. Microscopic cracks initiating in the weaker zones would have little trouble progressively developing through the remainder of the specimen during repeated stressing (25).

The presence of slag, oxides, sulphides, or other inclusions in steel is frequently suspected as being harmful and often directly blamed for premature fatigue failures (28). While the presence of this "dirt" is theoretically undesirable, the presence of some minimum amount is practically unavoidable, and it is debatable whether the amounts usually found in steels are sufficient to cause a significant fatigue strength reduction. Certainly the overall quantity of inclusions is not a direct evidence of fatigue weakness. For example, several studies indicate that the fatigue limit of wrought iron is about the same in the direction of rolling as in the transverse direction in spite of the fact that it contains up to 2.5% of silicous material in the form of stringers elongated in the

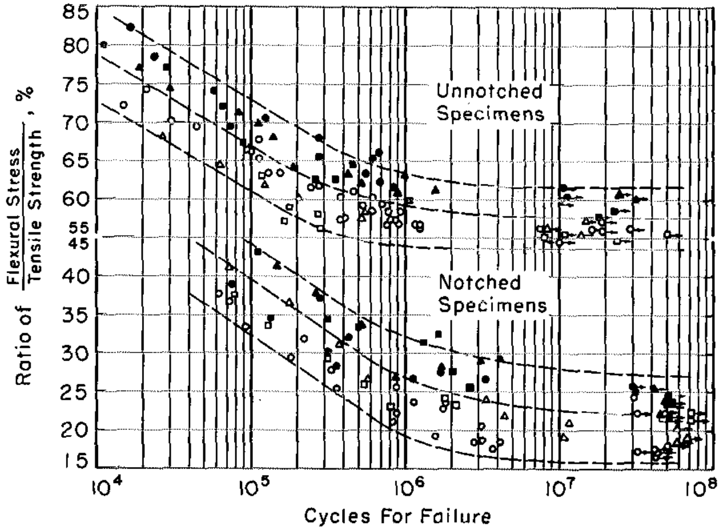

Fig. 20—S-N Curves for Three Steels in Terms of Static Tensile Strength. Open symbols represent specimens slowly cooled in quenching whereas the solid symbols show data for specimens rapidly cooled in quenching. The latter group, showing increased fatigue life and fatigue strength, had structures with *finely dispersed* hardening constituents (consisting mainly of tempered martensite).

direction of rolling. In extensive studies of a variety of aluminum alloys Templin and co-workers (43) found that the fatigue strengths in a direction transverse to the rolling fiber were not significantly different from those in the longitudinal direction. Small quantities of lead particles dispersed throughout the grains in leaded steels have no deleterious influence on the fatigue strength even though their physical properties differ markedly from those of the matrix (35,41).

The shape and distribution of the inclusions (or of the hard carbides in steel and the graphite in cast iron) are fully as important as their size, and must be further considered in relation to their effect on the behavior of the surrounding matrix. High strength hard steels (such as ball and roller bearings, case-hardened gears, etc.) may be more adversely affected by inclusions than a low strength ductile steel (29). Grant (36) found that specially treated cast iron with spheroidal graphite exhibited a fatigue strength definitely superior to that of the same cast iron (with the same tensile strength) but with the characteristic stringy dispersion of graphite flakes. Apparently large closely-spaced segregations or elongated stringy types of inclusions are less desirable than finely dispersed spheroidal types.

Interfacial stresses will always exist on a microscopic scale between the different constituents in a metal. Little is known of their magnitude, but their incompatibility due to different physical constants[5] could lead to severe conditions when repeated external loads are applied. For example, localized precipitation within grains may develop volume changes and a complex interfacial residual stress pattern. These micro-stresses are superimposed on the macro-stresses developed by the external loading and, hence, the significant stresses that initiate cracking may be greatly different from the computed nominal stresses. The fatigue properties of the aluminum-copper precipitation hardening alloys are practically independent of the degree of dispersion, and the alloys of highest static strength exhibit only a small increase in fatigue strength over those of low static strength. The explanation for some of these disappointingly low fatigue strengths may lie in development of unfavorable textural stresses on a micro scale.

Effect of Temperature

Materials which normally show a definite fatigue limit at room temperature are found to continually decrease in fatigue strength at elevated temperature as a function of the number of cycles (or of the total time of testing). Since creep and fatigue damage occur simultaneously, the elapsed time affects the behavior as well as the number of stress cycles (30,31). High temperature increases the mobility or accelerates the place change of atoms; this would facilitate greater slip and deformation prior to fracture. However the fragmentation of grains is opposed by the tendency to re-form into a minimum

[5] For example, differences in modulus of elasticity and coefficient of thermal expansion could lead to high stresses at the interface between two dissimilar constituents when loads are applied or during cooling after heat treatment.

stable grain size, so that recrystalization may result. Other metallurgical changes such as precipitation or aging probably are accelerated by repeated stressing.

The fatigue severity of a stress raiser is not alleviated by the creep that may occur at high temperature. Jones and Wilkes (37) in examining data of the type shown in Fig. 21 found that at 1200 °F (650 °C) the notch-sensitivity of high temperature alloys in reversed bending was practically the same as that at room temperature. The influence of a given

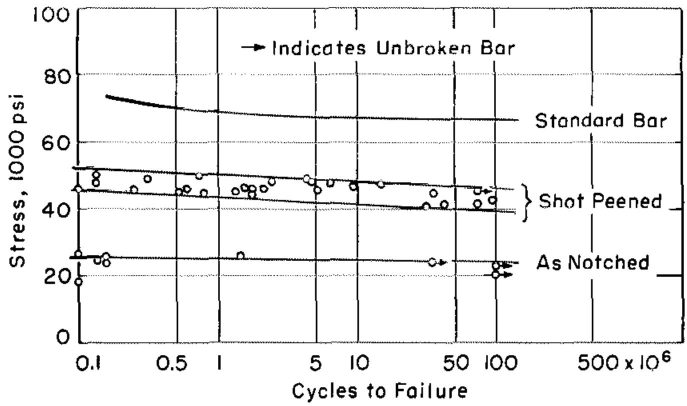

Fig. 21—Flexural Fatigue Curves for S-816 at 1200 °F. Note the pronounced strength reduction due to a notch even at this high temperature. Shot peening the notch improved the strength but did not approach that of the unnotched standard bar. Jones and Wilkes (37)

notch depended somewhat on how it was produced. Severely ground notches reduced the fatigue strength at room temperature by a factor of 5.0 whereas at 1200 °F (650 °C), the strength was reduced by a factor of only 2.4. Nevertheless when the same shape notch was cut by a lathe tool, the strength was reduced by factors of 1.8 at room temperature and 2.0 at 1200 °F (650 °C). The authors attributed these differences to residual stresses caused by the method of manufacture and their partial relief at elevated temperatures. It is important to note that the reduction in fatigue strength due to a stress raiser at high temperature was comparable to that obtained for ordinary steels at room temperature. The superposition of a high mean stress may develop creep to redistribute the stresses at the notch but may not be helpful in reducing the alternating component of stress superimposed by the fluctuation of loading.

Failure due to creep at high temperatures usually starts with grain boundary movements and separation whereas fatigue fractures are normally transcrystalline. Thus the mixture of time-dependent and cycle-dependent phenomena whose terminal effects are different lead to much complication in prediction of structural behavior. At room temperature slip is a discontinuous process governed by the resolved shear stress but the relaxation at elevated temperatures allows further slip to proceed at the same stress level. However the per cent elongation to fracture is pronouncedly reduced by the superposition of alternating stress on a static preload, even though small amplitudes of alternating stress do not appreciably reduce the mean stress to produce creep rupture in a given life (42). Environmental effects of erosion, oxidation and chemical attack at elevated temperatures all serve to develop potential nuclei for progressive fracture. The fatigue strengths of materials at elevated temperatures are, therefore, greatly lowered by this imposing combination of decreased crystal strength, instability of structure, accumulation of creep damage and unfavorable environmental factors.

Fortunately these unfavorable trends are all reversed by depressing the temperature; all metals show improved fatigue properties at low temperatures (32,33). Because of the widespread publicity given to problems of brittle fracture of carbon steels at low temperatures, many engineers feel that metals must also become brittle or notch sensitive in fatigue as the temperature is lowered. However there is no relation between notch sensitivity concepts in fatigue with those obtained from notched bar impact tests, and the fatigue strength of notched specimens is also improved by lowering the temperature.

DESIGN CONSIDERATIONS

It is fortunate that small decreases in the peak stresses developed in a member under a given set of conditions will result in large improvements in the fatigue life. Thus an effective engineering solution to the problem is not impossible. Many of the chronic cases of fatigue failure in service are traceable to poor design details or to processing mistakes or service induced defects that result in conditions well known to be poor from the fatigue viewpoint. Even though exact procedures of a mathematical nature for proportioning parts are not available, the elimination of severe stress raisers and care in processing operations will do much to prolong fatigue life and give satisfactory service.

Many factors affect the fatigue life of a metal part. The rate at which damage is accumulated either in initiating or in propagating visible cracks is, in general, influenced by: 1—the mechanical strength properties of the fabricated member (these include not only the chemical composition and metallurgical structure, but also the influence of surface finish, plastic deformations in processing, and shape and size factors inherent in the design) ; 2—the state and range of stresses that must be resisted in the most critically stressed zones of the member; 3—the environmental conditions which include the presence of corrosive atmosphere, erosion, elevated temperatures or fretting corrosion (34). Of these, the most acute troubles arise from the severity of localized stress due to shape factors, or from excessive range of imposed stress or corrosive service conditions.

As long as the metal is sound and free from inherent defects, selection of material within wide limits is not as important in achieving satisfactory fatigue life as is the care in design, fabrication and maintenance. For members containing severe stress raisers, it is sometimes found that only minor improvements can be expected by changing to a "higher strength" metal except under unique conditions such as those involved in operation at extremely high temperatures. Quite often the elimination of harmful conditions is found to be a more economical and effective means of improving fatigue strength:

(a) Eliminate sharp recesses and severe stress raisers by use of generous fillets with general streamlining of the part to provide smooth gradual transitions of stress;

(b) Avoid sharp surface tears from rough machining, punching, stamping, shearing, etc.;

(c) Prevent surface decarburization during processing or heat treatment;

(d) Control (or afford protection against) corrosion, erosion, or chemical attack in service.

(e) Alter design to eliminate press fits, dove-tails or other mating parts that develop a microscopic chafing action in transferring the load that seriously reduces fatigue strength (see Fig. 3).

ACKNOWLEDGMENT

The author wishes to acknowledge the support given to this work by the Office of Naval Research through their sponsorship of studies being conducted in the Department of Theoretical and Applied Mechanics at the University of Illinois as Project NR031-005. Many of the ideas and experimental data presented in this paper are those developed by my colleagues, in particular by H. T. Corten, W. J. Craig, A. M. Freudenthal, W. J. Love, G. M. Sinclair, and C. E. Work, during their participation in the research on this cooperative project to study the effect of repeated stressing on the properties of metals.

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