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PROPAGATING AND NON-PROPAGATING FATIGUE CRACKS IN METALS

by

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ABSTRACT

The development of the "fail-safe" theory for the design of aircraft aroused considerable interest in fatigue-crack propagation rates and in the effect of such cracks on the residual static strength. Methods of detecting crack initiation and of measuring subsequent growth are briefly discussed, and various analytical studies of fatigue-crack propagation are examined. The experimental results of numerous investigations of crack growth in laboratory specimens, and in simple and complex structures, of ferrous and non-ferrous. alloys are reviewed and compared with theoretical predictions and general conclusions are drawn. The particular case of so-called non-propagating cracks is considered, and conditions governing their initiation and development are outlined. Finally, the effect of fatigue cracks on static strength and on the subsequent response of the material to cyclic loading is discussed.

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Direction des mines

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PROPAGATION ET NON-PROPAGATION DE FATIGUE AU SEIN DES MÉTAUX

par

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RÉSUMÉ

L'application de la théorie "d'infaillibilité" (fail-safe) à la construction des avions a suscité un intérêt considérable pour les taux de propagation des fissures de fatigue et l'influence de telles fissures sur la résistance statique résiduelle. La présente circulaire étudie brièvement les procédés utilisés pour déceler les causes des fissures et mesurer leur croissance subséquente; elle traite également de certaines recherches analytiques dans la propagation des fissures de fatigue. Les résultats expérimentaux de nombreuses études de la croissance des fissures au sein d'échantillons de laboratoire et dans des structures simples et complexes faites d'alliages ferreux et non ferreux sont passés en revue et comparés avec les prédictions théoriques, et l'on en tire certaines conclusions générales. L'auteur traite du cas particulier de la non-propagation de certaines fissures, ainsi que des conditions qui en favorisent la naissance et le développement. Finalement, on examine l'influence des fissures de fatigue sur la résistance statique ainsi que sur le comportement subséquent de la substance soumise à l'action de charges cycliques.

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FIGURES

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1. INTRODUCTION

The study of the behaviour of metallic materials under cyclic loading conditions, which began with the early experiments of Wohler, has made considerable progress in the intervening years. The effects of various factors such as notches, size and shape, temperature, corrosion, surface treatments and variable-amplitude loading, to mention but a few, have been systematically investigated in numerous countries, and the state of knowledge regarding the fatigue phenomenon is now well advanced. Most of this information, however, particularly in the earlier years, related to those conditions under which the material either did or did not fracture, that is, to a determination of the fatigue limit or the fatigue strength at a given number of cycles; little attention was paid to the onset of crack initiation, or to the duration of the cracked stage.

The reasons for this lack of attention were probably two-fold. In the first place, the significance of the crack growth period was undoubtedly not appreciated at the time. Secondly, it was much simpler to design a fatigue machine to indicate complete fracture of the specimen, or at least an advanced crack, than to detect crack initiation. The advent of "fail-safe" design as applied to aircraft certainly changed the situation markedly in the former respect, with the result that interest in this aspect of the fatigue process was stimulated, and a large proportion of the literature in recent years has therefore been devoted to crack initiation and propagation.

The basic mechanism of fatigue failure is still not fully understood, although various stages have been proposed and different theories postulated.[1*] It is generally agreed that localized slip takes place in certain favourably oriented grains under cyclic loading, and that it concentrates in regularly spaced bands or striations. At stresses in the unsafe range, the bands tend to thicken, showing a fine structure of ragged slip lines, and one or more microscopic cracks appear in these highly deformed regions. (With the aid of the electron microscope, such cracks have been detected at less than 1% of the normal fatigue life.) These minute cracks then coalesce to form visible cracks which ultimately propagate to fracture.

Some difficulty arises as to the exact point at which a slip marking or a so-called persistent slip band becomes a micro-crack. By applying a tensile load, the surface at the slip band may be opened up to show a

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*Numbers in superscript correspond to references listed at the conclusion of the report.

micro-crack, but this is not positive proof of the pre-existence of the crack. An annealing treatment may remove a deformation texture, and leave a micro-crack; this method has revealed micro-cracks at an early stage,[2] as has the repeated electropolishing technique of Wadsworth and Thompson.[3] Bennett[4] has illustrated the difficulty very clearly, and concludes that it may be necessary to section and examine the specimen in order to distinguish between slip-bands and cracks. In the circumstances, therefore, and for the present purpose, it is convenient to divide the process into two stages, as suggested by Schijve[5]--the micro-stage and the macro-stage--the boundary being the point where the crack becomes visible to the naked eye with, possibly, the use of a low-power magnifying glass. The micro-stage may more loosely be designated the "pre-crack stage", and the other the "crack stage".

The factors which affect the duration of these two stages are necessarily different. As would be expected, local conditions near the origin will control crack initiation, whereas propagation will depend on conditions in those regions through which the crack passes. De Forest,[6] for example, showed that the degree of surface finish affected the pre-crack stage, but not the propagation. At a given stress level, cracks initiated earlier in specimens having a coarse circumferential finish than in those having a fine circumferential finish, while a coarse longitudinal finish gave the longest micro-stage.

Another example is afforded by cracks which initiate at a sharp notch and then fail to propagate; these are discussed more fully in Section 4.

A number of conclusions have been drawn[5] concerning the relative significance of the two stages, with particular reference to light alloys:

1. The relative duration of the crack stage is longer in notched specimens than in smooth specimens for the same total life; this seems reasonable, since a crack will produce a greater redistribution of stress in a smooth specimen, plus a stress-raising effect at the tip.

2. The relative duration of the crack stage is longer at high stresses than at low stresses; evidence in support of this conclusion can be found in the data of De Forest,[6] Bennett and Baker,[7] and Weibull.[8] Another feature of interest is that more cracks are formed at high stresses than at low stresses, an observation which has also been made on steel specimens. This feature is most apparent in the fracture faces of circumferentially-notched bars, and is responsible for the so-called "beach" marks around the periphery.

3. Greater scatter in endurance occurs in the micro-stage than in the macro-stage, and consequently the scatter normally observed in fatigue tests to fracture is due mainly to the variations in the number of cycles for crack initiation. At a given stress level it has been

found that the samples which crack the earliest will have the shortest lives. In the first stage, when the length of the micro-crack is of the order of the grain size it will grow slowly, and may reasonably be expected to be more susceptible to variations in slip resistance, grain orientation, inclusions, etc. Once the macro-stage is reached, the "visible" crack may be considered more of a bulk phenomenon less affected by local variations.

4. As a corollary of conclusions 2 and 3, it follows that smooth specimens will exhibit greater scatter than notched specimens, for the same mean endurance, since the pre-crack stage will be longer; this effect has been observed by Weibull.

5. A further corollary is that more scatter should be expected at low stresses and long endurances than at high stresses; again, this effect has been found experimentally.

Before leaving this section it is pertinent to mention some of the equipment and techniques that have been employed for the detection and measurement of fatigue cracks. During the micro-crack stage the most effective instruments have been the optical microscope and the electron microscope, but there are indications that refined ultrasonic methods may also prove helpful. For crack detection beyond this stage, most of the standard non-destructive testing methods have been used. These include fluorescent and dye-penetrant techniques for non-magnetic materials, magnetic particle methods for magnetic materials, radiography, and ultrasonic and eddy current methods.

The fracture-wire technique has also been used successfully for crack detection, and, to a lesser extent, for studying the rate of propagation. In this method a fine insulated copper wire, 0.001 or 0.002 in. in diameter, is cemented to the surface of the specimen or structure, at a point adjacent to the stress-raiser where the crack is expected to initiate. When the crack develops, the copper wire fractures; this point can be detected by a continuity check, or, more simply, by using the open circuit to trip a relay and stop the fatigue machine. The method has been employed to check crack growth in tension panels[9] by mounting a series of such wires at known distances apart.

Hult[10] used an ultrasonic pulse reflection system to measure crack propagation in square specimens of an aluminum alloy. The crack initiated from a longitudinal V-notch under cyclic torsional loading, and the depth was determined by means of a 6 Mc/s probe, previously calibrated against a machined crack. The accuracy claimed was about 0.1 mm.

Hunter and Fricke[11, 12, 13] have developed a simple and inexpensive replica technique, using Faxfilm, which has worked satisfactorily on smooth and notched cylindrical specimens. Other investigators[14] have utilized stop-motion photographic equipment synchronized to the peak of the stress cycle. Both of these methods give permanent records of the surface growth of cracks, and these can be analyzed subsequently.

More recently, Forsyth and co-workers[52] have used a replica method in which crack growth is determined from the markings on the fracture surfaces. In the case of certain light alloys, striations may be observed on these surfaces, and each striation is believed to correspond to one applied stress cycle. By measuring the distance between successive striations, or by counting the striations in a given interval, it is therefore possible to obtain crack growth curves. Successful results have been reported for DTD 546 clad sheet. In a programmed test on an L65 fork fitting, the load level changes were clearly seen with an optical microscope, but the striations could only be resolved with an electron microscope.

Crack growth has also been measured by an electrical resistance method with a double Kelvin bridge,[15] on R. R. Moore samples of a magnesium alloy. The correlation between the change in conductance and the percentage of area cracked was obtained by breaking specimens in tension, after cracking to various degrees, and measuring the fatigue-cracked area by means of a planimeter. Unfortunately, this method is not sensitive to the earlier stages of crack growth, since a 1% change in conductance corresponds to a

10% cracked area.

A more accurate method, the direct-current conduction method, has been developed and used in the Mines Branch.[16] The Crack Depth Indicator has an external four-electrode probe which can be designed to fit any particular application. The ratio of the voltage across the crack to the voltage remote from the crack, using constant current, is used to estimate the crack depth from an experimentally determined calibration chart. Further details of this and other methods will be found in reference 17.

2. CRACK PROPAGATION THEORIES

Analytical studies of the growth of fatigue cracks have not been very numerous--a fact which is not surprising in view of the many variables involved. The theoretical solutions that have been put forward differ essentially in the emphasis placed upon the different variables, and in the assumptions that of necessity must be made. Shanley[18] assumed that the growth rate for a given stress was proportional to the crack depth, and proposed the exponential relationship,

$$l = Ae^{\alpha n} \quad , \ldots\ldots\ldots\ldots\ldots \quad (1)$$

where l = depth of crack,

A = a constant,

α = a stress-dependent constant ($B\sigma^c$), and

n = number of cycles.

Hence, the derived rate of crack propagation is given by

$$\frac{dl}{dn} = A\alpha e^{\alpha n} = \alpha l \quad \ldots\ldots\ldots\ldots (2)$$

Weibull[8] considered that the peak stress at the tip of the crack was the main controlling factor in crack propagation. He assumed that the growth rate was proportional to some power of this peak stress, which in turn was proportional to some power of the nominal stress on the remaining cross-section. The expression he proposed for growth rate was

$$\frac{dl}{dn} = K\sigma^{\beta}, \quad \ldots\ldots\ldots\ldots\ldots (3)$$

where K and β are constants depending on the material, the test-piece dimensions, and, probably, the stress distribution.

A mathematical study of fatigue crack propagation has been made by Head,[19, 20] who considered that cycle-dependent work-hardening at the crack tip produced localized fracture (crack extension) when the local stress reached the true fracture strength. As the crack extended, the stress concentration factor at the new tip would be higher, the number of cycles required for sufficient work-hardening would be less, and the crack would therefore grow at an increasing rate. He assumed, amongst other things, that the medium was infinite, the applied stress constant, the mean stress zero, and the thickness of the plastic zone ahead of the crack ("a") constant and independent of the crack length. Head expressed his results in the form

$$\frac{1}{l^{\frac{1}{2}}} = \frac{D}{a^{\frac{1}{2}}}(c - n), \quad . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; . \quad (4)$$

where D is a factor depending on the stress, and c is a constant; whence

$$\frac{dl}{dn} = \frac{2Dl^{3/2}}{a^{\frac{1}{2}}} . \quad . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; . \; . \quad (5)$$

There is an obvious similarity between the expressions of Shanley, equation 2, and Head, equation 5. Both expressions imply that the growth rate is zero for zero crack length, and so they can only be valid from a finite length of crack. Weibull's equation 3 is of a different form and here the rate has a finite value for zero crack length. Head attempted to verify his theoretical relationship by means of the experimental data of Bennett (X4130 steel), De Forest (SAE 1020 steel) and Moore (car-axle steel). When plotted in the form $l^{-\frac{1}{2}}$ against n, the results gave reasonably straight lines, and appeared to justify his conclusions. Replotting, however, shows that they also conform with a variety of other expressions relating crack growth and length.[21]

Weibull[8, 22] carried out pulsating tension tests on certain light alloys in the form of wide sheets with small internal slits. When the applied loads were progressively reduced throughout the tests, so that the nominal stresses on the remaining cross-section were kept constant, he found that the crack length was directly proportional to the number of cycles, and that the growth rate versus stress curve followed a power relationship. In the initial stage of small crack lengths (his so-called transition period), the agreement, however, was poor, and there is some doubt as to the validity of the corrections made to the applied loads to keep the stress constant as the crack extended.[21]

Additional expressions have been put forward, more recently, by McEvily and Illg,[23] and by Hult.[10] The former authors based their suggestion on Orowan's work-hardening concepts and on the introduction of K_N, the stress concentration factor for a crack, corrected for size effect. From a series of pulsating tension tests on internally-notched sheet specimens of two light alloys, they derived the relationship

$$\log_{10}\left(\frac{dl}{dn}\right) = 0.00509\ K_N\sigma - 5.472 - \frac{34}{K_N\sigma - 34}, \quad . \ . \ . \ (6)$$

where σ is the net nominal stress.

Their data, irrespective of material, specimen width, and stress range, fitted this equation reasonably well. In addition, they found that Weibull's results showed good agreement with the equation. By combining equation 6 with Head's equations, they were then able to compute crack growth curves that also showed general agreement with the experimental data.

Hult directed his attention to the specific case of a notched bar in torsion, where the mathematical difficulties are less severe. With a slight modification of a suggestion made by McClintock,[24] Hult was able to derive expressions for crack initiation and the initial stages of crack growth, based on a critical-strain hypothesis. He concluded that the initial rate of crack growth was constant for a given applied strain, and varied as a certain function of the strain; and that neither the applied mean strain nor the notch depth was significant, provided the abscissa in the latter case was the number of cycles multiplied by the notch depth. In subsequent tests on square samples of a wrought aluminum alloy, many of his predictions were verified for the initial stages of crack propagation. The results of varying the notch depth, however, were not in accordance with the theory, a fact which Hult attributed to inhomogeneity in the material. It is important to note that the majority of his tests did not extend beyond about 1000 cycles, and that deceleration of the crack growth was apparent after the first few hundred cycles. In a few tests to 10,000 cycles, the growth was observed to continue in a step-like fashion. Hult concluded that this was not a strain-hardening effect--microhardness tests gave negative results--and attempted to explain it on the basis of friction between the crack surfaces. The discontinuous nature of crack growth will be discussed more fully in the next section.

3. CRACK PROPAGATION EXPERIMENTS

Crack propagation studies have been made on ferrous and non-ferrous materials by a number of investigators--on plate, sheet and round specimens, either plain or notched, and on simple structures. Most of the work has been carried out on non-ferrous materials, and in this field Hunter and Fricke have made notable contributions.[11, 12, 13] Their tests were made on smooth and notched specimens of various aluminum alloys in the form of sheet or rod, under plane bending or rotating bending conditions. They obtained smooth curves for crack initiation, similar in shape to the normal S/log N curves; for smooth specimens, these curves converged on the normal curves near the endurance limit, whereas for notched specimens the curves extended below the notched endurance limit (Figure 1). They found that propagation was not a steady continuous process, but took place in a series of steps, and that temporary immobilization often occurred when the crack reached a certain critical size (Figure 2). This hesitation was less apparent in smooth specimens at the higher stresses, and in notched specimens with the sharper notches. When growth was resumed, however, no change was apparent in the rate. The critical crack size in 6061-T6 aluminum alloy was about 0.3 mm, or approximately five grain diameters. Hunter and Fricke also observed that propagation took place by a number of modes; namely, by extension of the crack tip, by formation of a branch a short distance behind the tip, or by formation of a new crack ahead of the tip that united with the original crack. This last process was more important at the higher stresses, where interaction between major cracks was evident. Finally, they concluded from their data that the length of a fatigue crack was exponentially related to the number of cycles, and that crack growth rate bore a similar relationship to the applied stress level.

Hyler, Abraham and Grover[25] conducted somewhat similar tests in rotating bending on severely notched bars of extruded 2024-T4 aluminum alloy. The specimens were sectioned after predetermined lifetimes, and the depth of crack penetration was measured. Very early crack initiation was observed ($<$ 1% of lifetime), followed by slow propagation for the majority of the lifetime, and then by sudden acceleration towards the end. At a stress just above the notched endurance limit, no cracks between 0.003 in. and 0.025 in. in depth were found. Unfortunately, the established presence of residual stresses in the bar stock causes difficulty in the interpretation of some of their results.

Martin and Sinclair[26] tested 3 in. wide 2024-T3 sheet, with a central hole, under pulsating tension loading (zero minimum stress). They found that the cyclic rate of crack growth increased with increasing crack length until a final stage of rapid growth was reached. The cracks showed no long hesitation periods as reported by Hunter and

Fricke; this was attributed to the fact that no compression was applied in the stress cycle to close the crack, which would therefore act continuously as a stress raiser. Furthermore, the cracks were mainly perpendicular to the sheet surface, but shifted to a 45° shear plane in the final stage, with an accompanying change in velocity. The propagation in this stage was definitely observed to be in steps in tests made at $2\frac{1}{2}$ cycles per minute. Plotting the results in the form of Head's equation 4, they found a linear relationship over the centre portion, but deviation at either end, and concluded that the plastically deformed zone ahead of the crack was probably not constant.

Mention has previously been made of the work of McEvily and Illg.[23] Their tests were made on 2024-T3 and 7075-T6 sheet with a central notch under pulsating tension loading (minimum stress, 1000 psi). They reported that the cracks grew symmetrically from the notch in planes perpendicular to the sheet surface, but that after a distance which was inversely proportional to a power of the stress the cracks changed to a 45° shear plane; this change did not appear to affect the rate of growth. They also found, as have other observers, that the rate of crack propagation in 7075-T6 was always greater than in 2024-T3.

A very thorough investigation was recently carried out by Lipsitt, Forbes and Baird[14] on $1\frac{1}{4}$ in. wide, 1100-H18 aluminum sheet specimens under pulsating tension loading in the low cycle-high stress range (minimum stress, 1200 psi).

The specimens were notched on one side only with a semi-circular notch of 1/16 in. radius, and the synchronized camera technique was used for observation. Their measurements, also, showed that crack growth was not a continuous process; neither was there a linear relationship with the number of cycles, even on a semi-log plot. Smooth curves could only be obtained if the observations were made at sufficiently long intervals.

Initially, the cracks grew slowly to 0.005-0.010 in. and halted for 50-60% of the life. Then they shifted from the perpendicular-to-sheet direction to two 45° shear planes, and propagated at an intermediate rate to a length of 0.075 to 0.125 in. This length depended on the level of the applied stress. There was then a short pause, followed by rapid propagation on a single 45° plane up to failure. The rate of crack growth from any arbitrary length increased as the stress was increased, but the stress level appeared to have no effect on the final crack length.

A short study by Clapper and Watz[15] is of interest since it was made on a magnesium alloy, A263A, using the change-in-conductance method. Their tests were carried out on mildly-notched rotating beam specimens at two stresses near the notched endurance limit. They found that cracks propagated slowly until 10-20% of the area was cracked, and then propagated rapidly to failure. In an attempt to check Head's equation 4, they used the cracked area as a measure of "l^2", but the agreement obtained was only fair.

Tests have been conducted on clad aluminum alloys by a number of investigators.[11, 27, 28] As is well known, the low-strength cladding has an unfavourable effect on the fatigue properties of smooth specimens. Cracks form relatively easily in the cladding and, though they are retarded at the transition to the core material, they still reduce the pre-crack stage of the latter considerably. Notched specimens, as stated earlier, have a relatively shorter pre-crack stage, and the deleterious effect of the cladding is, therefore, less pronounced.

In the ferrous field, the greater part of the information available on crack propagation has been provided by investigations carried out at the National Engineering Laboratory, Scotland.[21, 29, 37] Apart from this work and that of De Forest mentioned earlier, Wilson and Burke[38] made a short study of fatigue crack growth in 12 in. x 3/4 in. structural steel plates. The specimens had a central slot, and were tested under reversed direct stress at $\pm$16,000 psi only. The data were plotted as average crack length versus number of cycles, and gave reasonably straight lines up to about 3/4 in., the limit of the observations (50,000 cycles). Temperature, in the range -40° to 120°F, had no significant effect, and crack velocity was found to be about twice as high in rimmed steel as in normalized, killed steel.

Lessells[39] has reported a series of comparison tests on cast and on forged nickel-chromium-molybdenum steel. The tests were made in rotating bending on crank sections consisting of two crank cheeks and a pin bearing. The comparison was based on the time to the detection of the first crack (1/8-3/16 in. long) and the time to failure. He concluded that the rate of crack propagation was higher in the forged material than in the cast material, although the load also was higher. Presumably, this higher rate was due to the more homogeneous structure of the forged steel. No direct measurements were apparently made of the growth of the cracks.

Of the work carried out at the N.E.L., the investigation most pertinent to this section is that of Frost and Dugdale.[21] The materials tested were two mild steels, aluminum alloy (L71), and commercially pure copper sheet. The specimens were 9 in. or 10 in. wide with a central slit, and were tested in pulsating tension. All crack lengths were measured from the centre of the slit. Two modes of cracking were apparent, the cracks starting at 90° to the sheet surface and then changing to single shear (45°) or double shear. No distinct change in the rate of growth was observed at the transition.

In the case of one mild steel in the annealed condition, it was found possible to observe strain markings in the plastic zone ahead of the crack. Measurements were made of the length of this zone, and less accurately of the width, and plotted against the corresponding length of the crack. It was found that the results were well represented by straight lines through the origin. Re-examining Head's equation 5 on the assumption that the parameter "a" is directly proportional to the crack length, leads to the expression

$$\frac{dl}{dn} = kl, \quad \ldots \ldots \ldots \ldots \quad (7)$$

or $$\log \left(\frac{l}{l_o}\right) = kn. \quad \ldots \ldots \ldots \ldots \quad (8)$$

When the data were plotted in the form log l versus n (Figure 3), straight lines were obtained for all materials tested, up to a crack length of about 1/8 of the sheet width, at which point stress increase due to the reduction in area began to have an effect. A slight deviation to a lower rate was noticeable in the early stages. Dormant periods were also observed, during which both cracks on any one specimen slowed down or stopped. This effect was more prevalent at the lower stresses, and affected the mild steels and copper more than the light alloy. After the dormant period the crack continued to grow at its original rate.

A straight-line relationship was also obtained when log k was plotted against log (alternating stress) for mild steel and the light alloy; insufficient results were available for the copper. From these plots, Frost and Dugdale derived the following expression:

$$k = \frac{\sigma^{p}}{N_s}, \quad \ldots \ldots \ldots \ldots \quad (9)$$

where N_s is a constant depending on material and mean stress, and $p = 3$ for the mild steel and the aluminum alloy.

The published literature on the propagation of fatigue cracks in simple structures is somewhat restricted. In this field, reference can only be made to the studies of Hardrath and his co-workers[40, 41] on aluminum alloy box beams and tension panels. The aluminum alloys used were 2024 and 7075. The beams were 20 in. wide x 8 ft long, and were tested under 4-point bending (13,000 $\pm$6,500 psi). The tension cover had a number of evenly spaced stiffeners or stringers, and a slot in the centre to initiate the crack. The crack growth was determined as the percentage of the tension area lost by fatigue cracking, and was plotted against the number of cycles after crack initiation. The cracks generally grew symmetrically, and with increasing velocity in the order of bonded, riveted and integral stringers (Figure 4). When the rivet pitch was changed, a 3/4 in. pitch was found to be superior to a $1\frac{1}{2}$ in. pitch, which in turn was superior to a 3 in. pitch. Closer spacing of rivets, however, did not give lower rates of crack growth if the crack avoided the rivet-holes. They also reported that a reduction of stringer area relative to skin area resulted in higher crack velocities, and that crack growth was appreciably slower in 2024 than in 7075 alloy, which was in agreement with findings on simple specimens. The harmful effect of reducing the relative stringer area was confirmed by tests on 30 in. wide tension panels at 14,000 $\pm$4,700 psi.

A systematic study of fatigue crack propagation in aircraft structures is obviously an expensive proposition, but a number of tests have been made on aircraft wings in various countries.[5] The wings consisted mainly of 2024 aluminum alloy and were conventional stressed-skin riveted constructions with discontinuous changes in cross-section and cut-outs at several places. The fatigue cracks were usually associated with rivet or bolt holes, but the mode of failure was often different in individual tests, and the first crack did not necessarily lead to final failure. A pre-crack stage, a stage of almost constant crack growth, and a final stage of accelerated growth were observed. The second stage occupied from 20 to 30% of the total life, but opinions differed as to the respective durations of the remaining stages. Some crack-growth curves were presented by McGuigan, Bryan and Whaley[53] for C-46 wings, from which it appeared that the crack stage was relatively longer at lower loads; this finding is in direct disagreement with the results on laboratory samples. It was also observed that stiffeners in the wings tended to act as crack-arresters at low loads, and produced discontinuities in the curves. At higher loads, the effect became less marked.

The foregoing investigations have been concerned essentially with fatigue crack propagation under constant-amplitude cyclic loading. Little is known about the rate at which a crack progresses under random loading, as in aircraft service. According to Christensen,[45] when stepwise loading is decreasing there is a considerable delay before the crack recommences to propagate at a lower rate. This effect was apparent at each decreasing load-level change in the data presented When the load is increased in steps, cracks propagate immediately and at a higher rate.

One investigation that has some bearing on this aspect was carried out on clad 2024 aluminum alloy sheet by Schijve and Jacobs.[28] Interval tests at two stress levels with a cycle ratio of 5% were made on both smooth and notched samples under fluctuating tension. The fractures of the notched samples showed the well known sea-shell markings, indicating the progress of the cracks at the two levels. The number of these markings was fairly constant and independent of the result of the test, which again shows that the crack stage is less susceptible to scatter than is the pre-crack stage. When the log of the crack area, expressed as a percentage of the total area, was plotted against the number of intervals before failure, a straight line could be drawn through the points, suggesting a continuous steady process. Obviously, this suggestion is not quite correct, but the results indicate that for these particular conditions the effects referred to above are not significant. With longer intervals, or larger differences between the stresses, this is unlikely to be true.

Before leaving this section, it is convenient to summarize below, briefly, the general conclusions that can be drawn on the basis of the experimental evidence presented:

1. Fatigue crack propagation is a discontinuous, step-wise process; dormant periods may occur, particularly at low stresses, but when growth is resumed the rate is essentially unchanged.
2. The average rate of crack propagation is proportional to the crack length and to some power of the applied stress.
3. In sheet material the crack initiates at 90° to the surface, and after a certain period it shifts to single or double shear at 45° to the surface. Opinion is evenly divided as to whether some change in crack velocity accompanies the shift.
4. The relative duration of the crack stage is greater for notched than for smooth specimens, and for high stresses than for low stresses: this may not apply to complex structures.
5. More cracks are present at high stresses.
6. There is greater scatter in the pre-crack stage than in the crack stage, in smooth than in notched specimens, and at low stresses than at high stresses.

So far as materials are concerned, crack propagation is more rapid in 7075 than in 2024 aluminum alloy, in rimmed than in normalized, killed 0.2% carbon steel, and in forged than in cast nickel-chromium-molybdenum steel. In the case of simple structures, bonded stringers are superior to riveted stringers, though the latter may be improved by reducing the rivet pitch. Riveted stringers, in turn, are superior to an integral construction.

4. NON-PROPAGATING CRACKS

Probably the first investigators to demonstrate the existence of large non-propagating fatigue cracks in areas of high stress concentration were Horger and Buckwalter.[42] In a series of full-scale bending fatigue tests on railway axles, they observed that cracks could exist in the wheel seats for substantial numbers of cycles. From the results of the tests they determined a "breaking-off limit" and a "crack limit", the latter being of the order of half the former. Between the two limits, the cracks did not progress to failure.

Some years later Lessells and Jacques[43] carried out rotating beam tests on notched samples of ship plate material, and found that cracks were developed at nominal stresses far below the notched endurance limit ($\pm$25,800 psi). The lowest stress for which cracks were reported was $\pm$10,000 psi, and at $\pm$21,500 psi they were still spreading after 83×10^6 cycles.

This apparent phenomenon of cracks which formed under cyclic stressing but did not propagate to failure was then studied intensively by Phillips, Frost, and co-workers. Such cracks were first reported[44] in round, mild steel specimens, having a V notch of 0.002 in. root radius, tested in direct stress. Similar tests made on an aluminum alloy (L65)[33] also proved the existence of non-propagating cracks in this material. It was found that there was a minimum nominal stress below which no cracks would form. Between this stress and the notched endurance limit, cracks would initiate at a relatively early stage but would only propagate to a limited extent. It appeared that the cracks were fully formed after about 100,000 cycles. At this point it was tentatively concluded that the extent of the crack penetration depended on the stress field generated by the notch, and was equal to the depth of material over which the alternating maximum principal stress was greater than the fatigue strength of the virgin material.

Using the theoretical treatment of Neuber, the stress distribution below the notch was plotted in terms of an arbitrary nominal stress. Since the fatigue strength of smooth specimens in direct stress was known, it was then possible to determine the length of material over which the stress exceeded this value for different nominal stresses. When the experimental points were plotted on the theoretical curve derived in this manner, the agreement was excellent. The points obtained from the previous work on mild steel also showed good agreement, hence the theory seemed to be substantiated.

Subsequently, Frost and Phillips made a more extensive series of tests[31] on the same two materials under both direct stress and rotating bending conditions. The specimens were notched with a 55° V-notch, and the effect of varying the root radius was examined. It was found, for L65 alloy, that the fatigue limit decreased with decreasing root radius to a minimum, corresponding to a certain critical value, and then increased again (Figure 5). Below this critical value, non-propagating cracks were found, but above it, all cracks formed propagated to failure. This critical value of the root radius, and hence of K_t, the theoretical stress concentration factor, coincided with the maximum strength reduction factor that could be obtained with the material. Similar results were obtained for mild steel, except that the fatigue limit remained constant below the critical root radius.

Careful examination of the material at the tip of propagating and non-propagating cracks failed to reveal any significant difference. In general, non-propagating cracks were transverse to the direction of loading, but the shortest cracks (0.001 in.) were invariably found at about 45° to the axis, and appeared to be propagating under the influence of shear stress.[46] No correlation was apparent between the direction of propagation of cracks and the characteristics of the individual grains of the material. The terminal point of a non-propagating crack was randomly situated either within a grain or on a grain boundary.

Further tests were subsequently made, by Frost and Dugdale,[35] on edge-notched mild steel plate, 0.3 in. thick, under alternating direct stress to enable the material at the notch root to be continuously examined by means of a microscope. Measurements of crack length were made up to 50×10^6 cycles in some cases, but all non-propagating cracks stopped before 10×10^6 cycles. A significant conclusion from this work was that the minimum amplitude of alternating stress required to cause initial cracking at the notch root was consistently given by the value of the smooth specimen fatigue limit divided by the theoretical stress concentration factor.

Similar tests carried out on mild steel, aluminum alloy, and copper sheet, and mentioned previously in Section 3, also demonstrated the existence of non-propagating cracks in both steel and copper specimens, though none was reported for the aluminum alloy. It is interesting to note that in these sheet specimens the non-propagating cracks almost invariably ended in a fork. In discussions, it has been suggested that non-propagating cracks may possibly be due to a residual compressive stress barrier below the notch induced by the notch-preparation process. This idea has been discounted by tests on mild steel in which the specimens were stress-relieved in vacuo after machining the notch; non-propagating cracks were found as before.

Confirmation of some of the above findings has been given by the studies of Hunter and Fricke[13] on rotating beam specimens of 6061-T6 aluminum alloy. In tests of notched specimens having a range of values of root radius, they reported that cracks would form provided the theoretical stress at the notch root exceeded a critical value roughly identifiable with the endurance limit of smooth specimens. If the applied stress was below the notched endurance limit, the cracks did not individually grow to a large size.

It will be apparent from the foregoing that fatigue cracks are less effective than sharp machined notches as stress-raisers, from which it may be concluded that the stress gradient at the tip is so steep and acts over such a short distance that its effect on the original stress field is negligible; or, that the effective radius at the tip of the crack is larger than that of the notch.[54] Harris[47] has attempted to explain non-propagating cracks on the basis of the existence of characteristic flaw patterns in the crystal structure; and Coffin[48] has suggested that closure of the crack may occur in compression after it has reached a certain depth, thereby reducing the stress concentration effect. Neither of these explanations is entirely satisfactory; in particular, non-propagating cracks have been found when no compression was applied in the cycle.

To summarize the studies made on so-called non-propagating cracks, it can be tentatively concluded that cracks will initiate in notched material when the theoretical stress at the notch root is greater than the endurance limit of the unnotched material. The cracks will then apparently propagate to the depth at which the theoretical stress, using Neuber's treatment, is equal to the unnotched endurance limit. In an aluminum alloy this process is normally completed at 100,000 cycles, but in mild steel it may take 10×10^6 cycles. No significant difference has been observed in the material just ahead of propagating or non-propagating cracks.

5. RESIDUAL STRENGTH OF CRACKED MATERIAL

During the past few years the "fail-safe" design concept has become increasingly popular in the aeronautical industry, and specific requirements have been included in certain aircraft codes. For example, the U.S. Civil Air Regulations establish the minimum fail-safe load requirements for civil airplanes at approximately 80% of design limit flight loads plus normal maximum fuselage pressure. The principle of fail-safe design is to make the structure tolerant of fatigue cracks; hence, it is desirable that the progress of a fatigue crack should be reasonably slow, preferably in a readily inspectable location, and that the static strength should not be markedly reduced.

A systematic study of this subject has been conducted by Hardrath and co-workers[49, 9, 40, 51] on 2024 and 7075 aluminum alloys. Notched samples of sheet in various widths and of 3/4 in. thick bar were subjected to repeated axial loading until cracks of different lengths were formed. The cracked specimens were then tested in static tension to determine the residual strength. The residual strength, based on the original net area and expressed as a percentage of the ultimate tensile strength, was plotted against the percentage of area cracked (Figure 6). The data showed that small cracks caused rather large decreases in static strength, that 7075 alloy was significantly more sensitive to cracks than 2024 alloy, and that on a percentage basis wide specimens were more adversely affected than narrow specimens. In addition, the authors predicted loss-of-strength curves based on an analysis of the elastic stress concentration factor corrected for size and plasticity. The derived curves were in fair agreement with the observations.

A similar procedure was followed with simple box beam structures and stiffened panels of the same alloys, tested respectively in bending and tension. Here the agreement with theory was poorer, probably due to the redistribution of loads as the stringers failed.

Schijve and Jacobs[28] carried out tests on 2024 aluminum alloy sheet with a more severe edge notch. Their results were generally similar to those above, but the loss in strength was less. This suggests that the strength loss decreases as the notch severity increases, a conclusion which should be regarded with some suspicion at this time in view of the scarcity of evidence.

Another series of tests was made by Nordmark and Eaton[50] on 1/4 in. thick flat samples with a central hole. The materials were 2014-T6, 2024-T4, 6061-T6 and 7075-T6 aluminum alloys, and the fatigue cracks were initiated under cyclic tension loading. Static load-deformation curves were presented and showed that the cracks caused a large reduction in ductility (total deformation), as might be expected, as well as a reduction in ultimate load. Only a few crack lengths were investigated for each alloy, but the results confirmed the previous finding that the reduction in strength in all cases was greater than the corresponding reduction in net area. The susceptibility of the alloys for a 1 in. crack increased in the order 6061, 2024, 2014 and 7075; which is also the order of increasing tensile strength. It was found that stop-holes drilled at the ends of the cracks increased the residual strength, but not to a level commensurate with the residual area, except for the alloy 6061.

Published data for complex structures such as airplane wings are scarce, but some encouraging results have been reported by Illg and Hardrath.[49] From tests on a C-46 wing, it was concluded that, up to 30% of failed tension material, the percentage loss in ultimate load was about equal to the percentage of failed material. In comparison with the behaviour of the simple structures previously mentioned, this result is rather gratifying, but it should only be considered as applying to the particular configuration for which it was determined.

In addition to the effect of fatigue cracks on static strength, it is also of interest to examine their effect on the subsequent response of a material to cyclic loads, and to discuss the conditions under which they will re-propagate. For this purpose, reference is again made to the researches of the National Engineering Laboratory, Scotland. Over a period of about ten years, an impressive amount of data was collected regarding the conditions under which pre-formed fatigue cracks would propagate in mild steel under direct stress or rotating bending loads. A thorough survey of this work, together with a considerable amount of additional data, has recently been presented by Frost.[30] The tests were made in general on specimens, stress-relieved in vacuo, of flat or round material covering a range of sizes, notch dimensions, and crack lengths. The survey also included the results of certain other investigators on mild steel.

The suggestion was put forward, initially, that the effective crack length should include the actual depth of the notch from which the crack was formed. The tests were therefore designed to determine the critical alternating stress governing the re-propagation of cracks of effective lengths from about 0.1 mm to 25 mm; both propagating and non-propagating cracks were included in the study. Circumferential notches, edge notches and central slits were employed, and in some cases the specimens were re-machined after the formation of the crack in order to reduce its effective length. The number of tests on any particular crack length was somewhat small, but the results were surprisingly consistent. In the case of the larger cracks, of length greater than 1/8 of the specimen width, the simple assumption was made that the nominal stress (σ_o) was the mean of values derived from the gross area and the net area. When the data were plotted in the form of effective crack length against nominal limiting propagation stress, they were found to lie around a smooth curve, indicating that some functional relationship existed between the two variables (Figure 7).

A suggestion as to the type of relationship was obtained from Frost's work mentioned in Section 3, in which it was shown that the length of the plastic zone ahead of the crack was directly proportional to the crack length for a given stress. Since yielding is dependent on the maximum tensile stress in the cycle (σ_t), the constant of proportionality (C) was plotted against values of σ_t and was found to give a smooth curve represented by

$$C = 1 + 0.0013\sigma_t^3, \quad \ldots \ldots \ldots \ldots \quad (10)$$

whence

$$l_p - l = 0.0013\sigma_t^{3} l, \quad . \; . \; . \; . \; . \; . \; . \; . \; . \quad (11)$$

where $l_p - l$ is the length of the plastic zone ahead of the crack, measured from the crack tip. If the ability of a material to resist breakdown at the tip of a crack is dependent on some finite quantum of material ahead of the crack, then it might be expected that the critical propagation stress and crack length would be related by a parameter of the form $\sigma^3 l$ = constant. The constant was evaluated as 140, when lengths were expressed in mm, and the corresponding curve was shown to fit extremely closely to the experimental points. Hence, bearing in mind the results in the previous section, it now appears possible, for mild steel at least, to determine the conditions for crack initiation with a given notch, and the conditions under which it will subsequently propagate.

One unresolved discrepancy which conflicts with the above findings exists in some earlier work reported from the same laboratory.[29, 34] Unnotched specimens of mild steel, nickel-chromium steel and aluminum alloy L65 were pre-cracked in rotating bending at stresses above the smooth fatigue limit, and then re-loaded in rotating bending or direct stress. The results obtained were expressed in stresses based on the net area, with an allowance for the non-axiality of loading of the cracked samples. The cracks were of various depths, but the data gave reasonably consistent S/N curves, and it was concluded at that time that definite limits existed below which the crack would not grow, and, furthermore, that crack length was not a significant parameter. Even when consideration is given to the effect of the crack depth/specimen size ratio on the stress calculation, the propagation stresses obtained are higher than would be expected from the previous results. One tentative explanation advanced by McEvily and Illg[54] is based on the high level of the stresses applied initially to form the fatigue cracks. It is claimed that the high stress-low stress sequence would introduce high residual compressive stresses with a resultant increase in the effective tip radius and consequently a higher re-propagation stress. Frost has made a similar suggestion, and hopes to resolve the discrepancy by work now in progress at N.E.L. on the static and dynamic pre-loading of notched specimens.[30]

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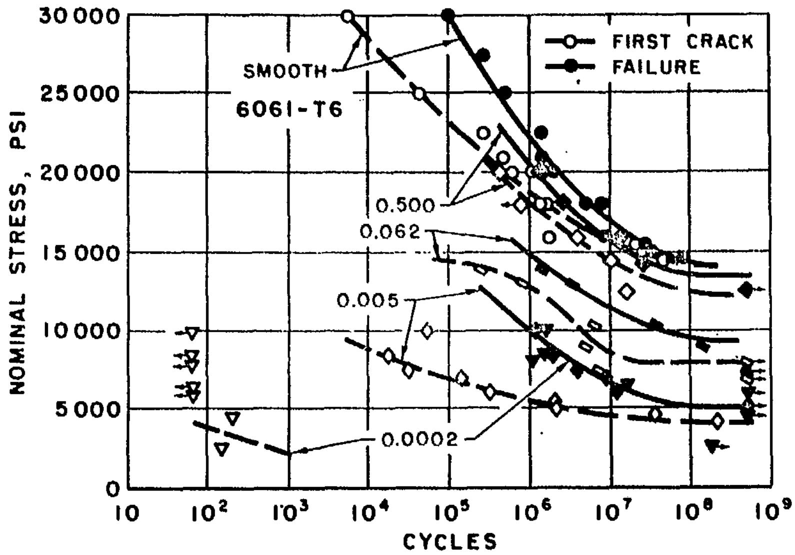


Figure 1 - First crack and failure curves of smooth and notched specimens of 6061-T6 aluminum alloy; notch radii are shown in inches. (Ref. 13)

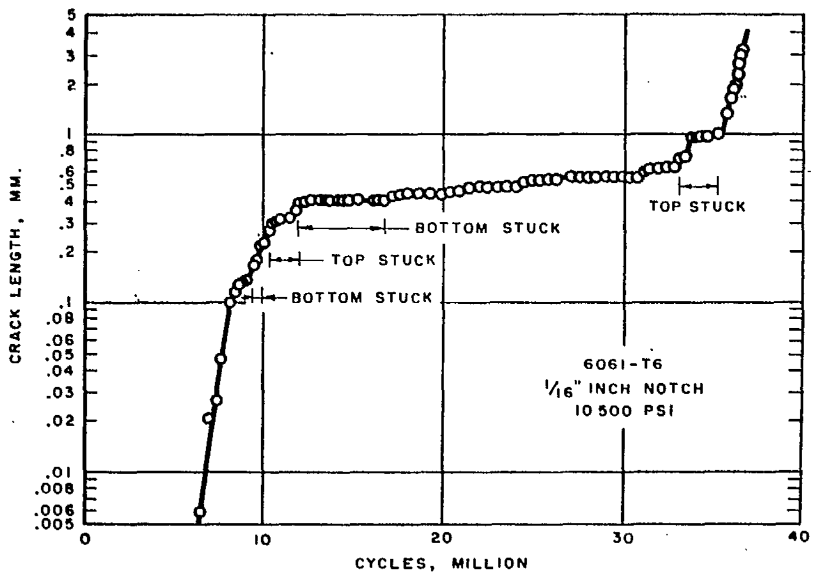


Figure 2 - Typical crack growth of 6061-T6 aluminum alloy, showing hesitation period. (Ref. 13)

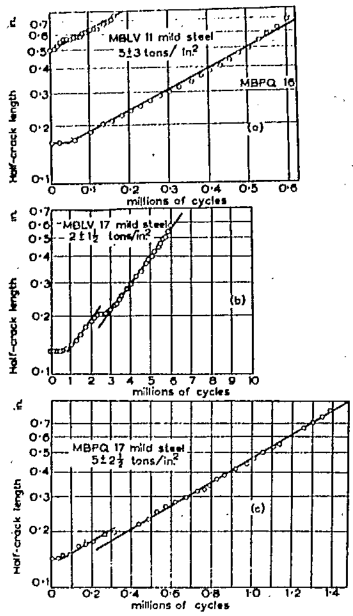


Figure 3 - Crack growth curves for centrally-notched mild steel sheet.

(a) Growth at the same stress with different notch depths.

(b) and (c) Growth at different stresses; note dormant periods. (Ref. 21)

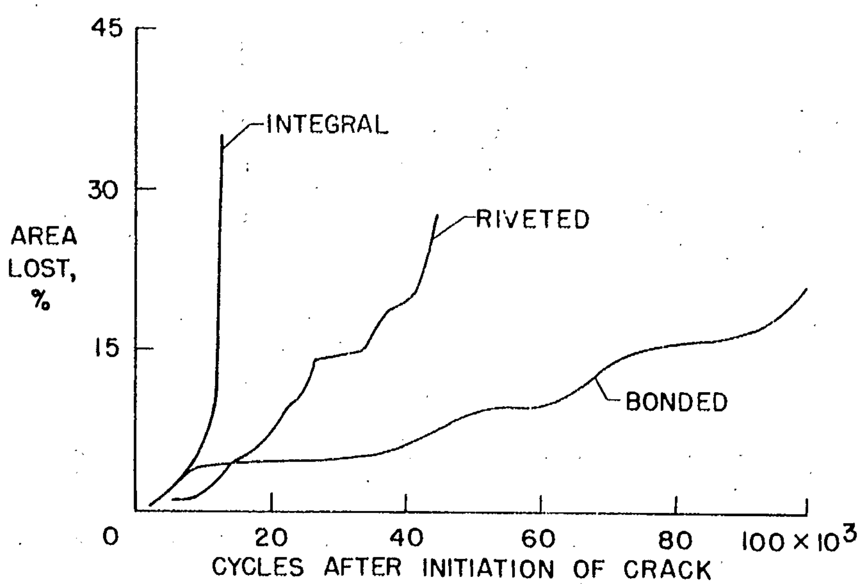


Figure 4 - Crack growth curves for box beams of 7075 aluminum alloy of different constructions. (Ref. 40)

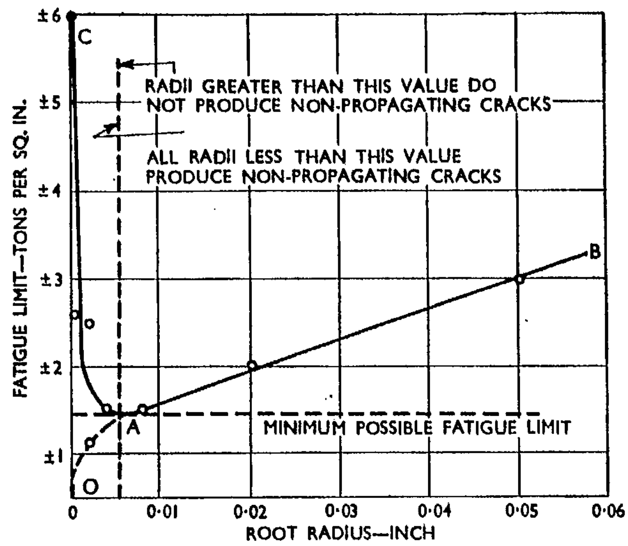


Figure 5 - Fatigue limit of V-notched specimens of aluminum alloy plotted against the root radius; •=value for pre-cracked specimens. (Ref. 31)

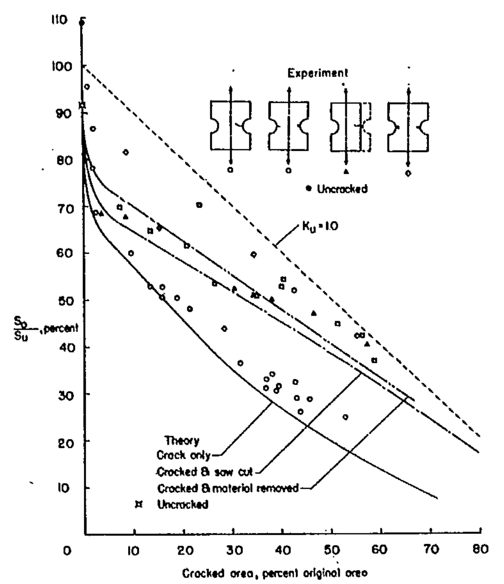
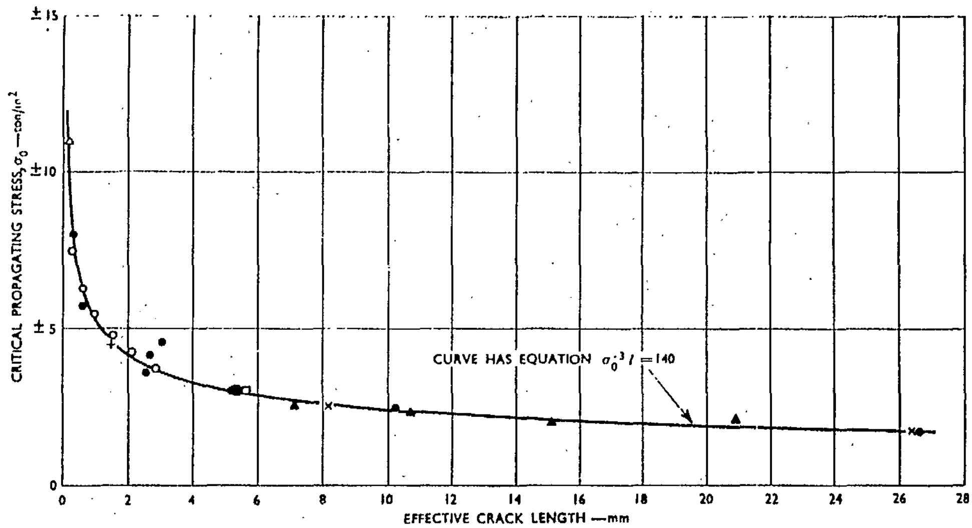


Figure 6 - Effect of fatigue cracks on the static strength of 7075-T6 aluminum alloy sheet. (Ref. 9)



△ 0·4 in. diameter round bar with circumferential crack.
⊙ Plate specimens with small edge cracks.
▲ Plate specimens with notches or slits greater than 0·2 in. deep.
□ Plate specimen with 0·2 in. deep notch.
■ Round bar specimen with 0·2 in. deep notch.
+ 0·5 in. diameter rotating bending specimens with 0·05 in. deep notches.
× 10-in. wide sheet specimens with central slit.
● Results from the literature.

Figure 7 - Critical propagating stress plotted against effective crack length for mild steel. (Ref. 36)

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