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FOREWORD

This report was prepared by Sidney O. Davis, Materials Information Branch, Materials Applications Division, Air Force Materials Laboratory. The work was initiated under Project No. 7381 "Materials Applications," Task No. 738106 "Design Information Development," and was administered by the Air Force Materials Laboratory, Research and Technology Division, Air Force Systems Command, Wright-Patterson Air Force Base, Ohio. The report covers work conducted during the period September 1964 to January 1965. The manuscript was released by the author in October 1965 for publication as an RTD Technical Report.

This technical report has been reviewed and is approved.

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ABSTRACT

This study summarizes the accumulated information and knowledge on precracked Charpy testing methods obtained as a result of an intensive literature survey. The precracked Charpy impact and slow bend test methods are defined and analyzed and their advantages, disadvantages, and limitations are critically appraised. The information will be especially useful to laboratories and engineers applying structural materials to aeronautical and aerospace structures.

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SECTION I

INTRODUCTION

It is generally recognized that one of the primary reasons for failure in high-strength sheet materials is that structural components (welded or unwelded) contain flaws and discontinuities which act to trigger crack propagation; the bigger the flaw the more dangerous the situation. The size of the flaw that will cause catastrophic failure depends on the fracture toughness of the sheet. Unfortunately, sheet (and welds) may contain flaws too small to be detected by practical inspection techniques, and yet such flaws can cause catastrophic failure in high-strength materials of low fracture toughness. If materials are used with inherently high fracture toughness or if heat treatments are used that improve toughness, it is sometimes possible to subject a given design to higher service stresses, depending on the combined effects of stress concentration and the resistance of the material to crack growth.

The Materials Advisory Board (MAB) Subpanel on Alloy and Stainless Steels in its report MAB-131-M, April 1958, listed notch sensitivity of sheet materials, its importance and how to measure it, as an "area of ignorance" requiring further research and an area where Government programs might well be sponsored. This was but one example of a growing awareness of an acute problem that was then brought into focus because of the premature failure of solid propellant motor cases in the Polaris, Minuteman, and Zeus missile systems. In particular, experience with solid propellant rocket motor casing failures has pointed out the danger of catastrophic failure in sheet materials when heat-treated to yield strengths above 200,000 psi (Report MAB-150-M).

In February 1959, the MAB conducted a symposium on "The Testing and Evaluation of Materials for Solid Propellant Rocket Motor Casings" (1). Among the many important facts brought out was that inspection techniques were not sufficiently sensitive to detect flaws of the size that cause premature failure in chambers heat-treated to high strength levels. It was also noted that there was wide scatter in notched tensile results (G_c) and that transition-temperature behavior may have contributed to the scatter in the room temperature G_c test results (1). Srawley (2) pointed out that the transition temperature may lie above room temperature when yield strengths are above 200,000 psi, thus emphasizing the need for testing over a range of temperatures. Espey, Jones, and Brown (3) showed that notch sensitivity is modified by mill variables and chemistry; Klier, in summing up the findings of the symposium, pointed out that "... although it has not as yet been treated, a factor of major importance unquestionably is metallurgical structure."

SECTION II

EFFECT OF MICROSTRUCTURE ON FRACTURE TOUGHNESS

Although the effects on Charpy transition temperature produced by changes in microstructural features of mild steels have been studied extensively at the Naval Research Laboratory, Battelle Memorial Institute, Massachusetts Institute of Technology (MIT) and elsewhere, there has been very little research aimed specifically at determining the effect of microstructural details on fracture toughness in high-strength sheet. The most recent work was performed by the Crucible Steel Company of America during the period 1962 to 1964 under the sponsorship of the Air Force Materials Laboratory (4). A cursory review of the literature on the effects of microstructure on toughness in ingot iron, plain carbon and structural steels reveals some information. A study of pearlite spacing on Charpy V-notch transition temperature in 0.20 and 0.40 percent carbon steels showed that as the pearlite spacing was changed from coarse to fine, the transition temperature was lowered (5). Whereas Rinebolt reported a decrease in transition temperature with decreasing interlamellar spacing, the data of Gross and Stout (6) indicated an opposite trend. In "Part III" of their study, Gross and Stout reported that the ferrite phase of a structural steel was the controlling microconstituent influencing the notch toughness of this material. Heat treatment designed to vary the pearlite hardness and spacing but maintain a constant ferrite grain size resulted in specimens with quite similar transition temperatures. Small amounts of untempered martensite were found to markedly reduce the notch toughness. In "Part IV" of their study, they showed that the notch toughness of ferrite, which is the predominant phase in structural steels, was controlled primarily by its grain size. Mn and Ni increased the toughness of ferrite largely by refining the ferrite grain size. The effect of Cr was generally low in this respect and Mo was ineffective. Grossmann (7) showed that small amounts of ferrite in the prior austenite grain boundaries have a very significant effect on the notch toughness of hardened steels.

At MIT, Owen et al. (8) described the relationships between the significant microstructural parameters, as observed with the light microscope, and Charpy V-notch transition temperature. For annealed structures in ship steels, it was observed that the transition temperature increased with ferrite grain size, pearlite patch size, and the related parameter of austenite grain size. The correlation was found to be strongest with ferrite grain size. In normalized structures, the variations of transition temperature could not be explained in terms of the above metallographic parameters. It was concluded that cooling rate, and probably compositional differences, can produce effects which influence the transition temperature but cannot be measured by light microscopy.

In general, little emphasis has been placed on the effect of microstructure on the initiation and propagation of fracture. Bruckner (9) has investigated the micromechanism of fracture in low-carbon steels and ingot iron in the tension impact test. Baeyertz, et al. (10) conducted a similar metallographic investigation of fracture in impact specimens of a structural steel. However, the studies by Bruckner and Baeyertz were confined to a limited variety of microstructures. Danko and Stout (11) at Lehigh University made a study of the effect of microstructure on the initiation and propagation of fracture in ingot iron and carbon steels of several carbon contents. The ultimate objective was to establish, if possible, a parameter relating microstructure to notch toughness. For a course-grained ferrite-pearlite microstructure in 1025 steel, initiation of cleavage occurred at the ferrite grain boundaries, at the ferrite-pearlite interfaces, and possibly at mechanical twin boundaries within the ferrite itself.

At the International Conference on the Atomic Mechanisms of Fracture (12) quantitative data were reported on the incidence and morphology of microcrack formation as a function of grain size, temperature, strain and strain rate (13). The study showed that quasibrittle fracture below the lower yield stress merely reflects a change in the dominant mechanisms of deformation from slip to twining. The fact that many microcracks larger than the critical size predicted by the dislocation models are formed without inducing fracture, demonstrates that present theoretical treatments do not adequately define the conditions for cleavage propagation of fracture. A likely explanation is that grain boundaries act as effective barriers and impede microcrack growth from one grain to the next.

The effort sponsored by the Air Force Materials Laboratory and performed by Crucible Steel Company of America characterized the structural factors that govern the heterogeneous micromechanics of flow and fracture in high-strength materials. A variety of high-strength steels and a high-strength titanium alloy were investigated in detail. Among the alloys studied were AISI 4340, a low-alloy, high-strength steel; Type H-11, a secondary-hardening, hot-work steel; Types 410 and 422, martensitic stainless steels; Type 301, a deformation-strengthened steel; Type PH15-7Mo, a precipitation hardened semiaustenitic steel; 300 grade 18 Ni marage steel; Cr-Co stainless marage AFC-77; and B-120VCA, an age-hardened beta titanium alloy. In these materials, the detailed fine structures are characterized and related to strength and fracture properties. Thus, a better understanding of the interrelation between strength and structural factors was achieved (4).

SECTION III

EVALUATION OF PLANE STRESS FRACTURE TOUGHNESS

CURRENT TESTS

Fracture toughness refers to the ability of a material to resist fracture propagation. The choice of tests for evaluating such behavior is of paramount importance. In view of the many variables that have an effect on fracture toughness (heat treatment, material composition, temperature, strain rate and many others), much work remains to determine which materials offer the most promise and which processing techniques develop optimum combinations of strength, ductility, and fracture toughness. Thus, it is desirable that the test chosen for evaluating fracture toughness be highly sensitive and reproducible, simple to perform and evaluate, economical in specimen material and preparation time, and capable of evaluating fracture toughness as a function of the many variables involved.

There are a number of tests currently available for assessing notch sensitivity in high-strength sheet materials. Two committees, one sponsored by the American Society of Testing of Materials (ASTM) and the other by the Aerospace Industries Association are actively engaged in seeking the most suitable tests for evaluating notch sensitivity in high-strength sheet materials, and their recommendations for a screening type of test have been published (14).

Currently, the most widely accepted criterion for fracture toughness in sheet is "G_c" (or K_c which is directly related to G_c; i.e., $K^2 = EG$). The G_c criterion, which can be directly related to critical crack length under plane stress at any desired stress level, stems from the work of Dr. Irwin and his associates at the Naval Research Laboratory (15) and is based on the original hypothesis proposed by Griffith about 42 years ago. Griffith postulated that a crack becomes self-propagating (catastrophic) when the elastic energy, released by relaxation of the stresses in the vicinity of the crack as a result of the crack's extension, is greater than the energy absorbed in creating the new fracture surfaces. This hypothesis was formulated for brittle materials but its successful application to cracking in ductile materials has been demonstrated independently by both Orowan and Irwin, by substituting a term representing the plastic deformation work at the fracture surface crack tip for the surface tension energy term in the Griffith relationship. Thus, for example, at the onset of fast fracture in a center-notched tensile specimen, the elastic energy release rate equals the rate at which energy is being absorbed in propagating fracture. Since the elastic energy release rate may be calculated from Irwin's formulae, the energy to propagate fracture per unit of crack area may be found.

PRECRACKED CHARPY TEST

Extensive testing of sheet materials for fracture toughness has been conducted at the Watertown Arsenal Laboratories (16). At the beginning of the study, a notched-tensile-impact test was evaluated but later abandoned when it became apparent that the results were seriously affected by kinetic and elastic energy losses. Most of the subsequent work was done using a modified Charpy test. It was demonstrated that in moderately tough materials with sheet thickness down to about 40 mils, significant data could easily and economically be obtained at impact rates of strain or in slow bend. These results, although not quantitative, showed fair qualitative correlation with G_c. However, on brittle materials the test failed, showing fictitiously high values of absorbed energy. It was believed that the anomalously high values in Charpy tests of brittle materials were due to elastic energy losses and, subsequently, it was demonstrated that elastic energy losses could be eliminated by precracking the specimens prior to testing (17). A machine for fatigue precracking V-notched Charpy specimens has been developed and is now used routinely. Since the precracking of sheet Charpy specimens eliminates not only elastic energy losses but also the energy required to initiate cracking, the test provides a direct measure of energy to propagate cracking. Expressed in units of in.-lb/in.^{2} this value is directly comparable although not necessarily identical with G_c.

An attempt to correlate precracked Charpy results with G_c (17) was made using a variety of materials in the form of broken G_c tensile specimens obtained from several other laboratories by Manlabs, Incorporated. The resulting correlation showed a direct proportionality but with considerable scatter. When both the G_c and Charpy tests were made from a variety of aluminum alloys supplied by Alcoa, with the tests made in one laboratory under carefully controlled conditions, an excellent correlation between precracked Charpy energy and G_c was obtained.

An important but much overlooked variable affecting fracture toughness is strain rate. It is generally felt that most high-strength materials are relatively insensitive to strain rate and that the effect of strain rate can therefore be neglected in fracture toughness evaluations of these materials. Attempts to investigate the effect of strain rate on fracture toughness in high-strength sheet materials have generally met with little success due to the mechanical difficulties involved in measuring fracture toughness at sufficiently high rates of strain in most fracture-toughness tests currently in use. However, the precracked Charpy test is ideally suited to an investigation of this type because fracture toughness may be evaluated at either impact or slow bend rates of strain. Impact and slow bend tests of a large variety of steel, aluminum, and titanium alloys have indicated that strain rate is generally a small but by no means insignificant variable. Contrary to expectation, it was found that impact rates of loading increased the fracture toughness in many of the materials tested, although a marked reversal of this trend was noted in the case of solution-treated B120 VCA titanium (17). The difference in fracture toughness as measured by slow bend and impact rates of loading were substantiated by lateral expansion* measurements in the aluminum alloys and the fracture appearance of the steels investigated. The rather surprising observation that slow rates of loading tend to provide lower fracture toughness values has also been reported by the Southern Research Institute (18) based on their shear-cracked tensile tests. They also observed an increase in fracture toughness in some materials with increased strain rate and suggested that the increase in fracture toughness was due to a lowering of the transition temperature. They stated that "... the effects of strain rate on transition temperature may depend upon the type of alloy, increasing strain rate effecting a decrease in transition temperature in some materials, and an increase in other alloys." With the precracked Charpy test, the effect of slow bend versus impact rates of loading can be easily investigated.

CENTER-NOTCHED TEST SPECIMEN

The standard 3-inch wide by 12-inch long center-notched tensile specimens are normally employed to evaluate the Irwin-Kies G_c parameter. The specimen thickness is the same as the sheet thickness. Two 5/8-inch holes are drilled and reamed at each end of the tensile specimens for connecting to the tensile machine platens by means of a special self-aligning fixture. The center notches are normally ground using a 1-inch diameter by 40-mil thick grinding wheel. The slot ends are squared off with a sabre saw and hand sharpened with a file before fatigue cracking. Fatigue cracking is accomplished using a tension-tension fatigue machine. "Through-the-thickness" cracks are formed as shown in Figure 1. The fatigue cracks should be made deep enough so that, in all cases, the crack is through the thickness. The specimens should be heat-treated before being fatigue cracked. See Reference 14 for additional information on center-notched test specimens.

The center-notched tensile specimens are pulled on a standard tensile machine. In the

*Lateral expansion on compression side, lateral contraction on tension side.

early tests, India ink was used to delineate the extent of slow cracking; however, in later tests a compliance gauge was employed to obtain a more accurate measure of the critical crack length and to reduce the possibility of corrosive attack on the crack tip by the ink.

The following equation was derived (14) to take into account the occurrence of plastic deformation ahead of the crack in a centrally-notched tensile specimen:

$$G_c^* = \frac{\sigma_g^2 w}{E} \tan\left[\left(\frac{\pi a}{w}\right)+\left(\frac{E G_c^*}{2 w \sigma_y^2}\right)\right] \qquad (1)$$

where

G_c^* = critical crack extension force or strain energy release rate where a plastic zone

$$\left(\frac{EG_c^*}{2w\sigma_y^2}\right)$$

correction is required

σ_g = nominal gross stress at maximum load in the test section away from the center notch

σ_y = yield strength of the material obtained from the unnotched tensile test

w = specimen width

E = Young's modulus

2a = total length of the center notch including the slow crack extension as measured by the ink-stain method; a = 1/2 the total length.

Values of G_c^* are obtained by graphical solution of Equation 1.

Using a compliance gauge to measure critical crack length, the value obtained is the effective crack length since it includes the effect of plastic deformation ahead of the open crack. Therefore, no correction for the plastic zone has to be made and the following equation may be used to obtain G_c:

$$G_c = \frac{\sigma_g^2 w}{E} \tan\left(\frac{\pi a}{w}\right) \qquad (2)$$

where 2a = critical crack length as measured with a compliance gauge. See Figure 2 for a typical compliance curve obtained with a compliance gauge.

PRECRACKED CHARPY TEST SPECIMENS

Impact and slow bend tests are performed on subsize V-notch Charpy specimens that have been precracked in fatigue (Figure 3). Specimens are prepared either from blanks prior to heat treatment (Figure 4) or from broken halves of center-notched tensile specimens if center-notched specimens are used to determine G_c. The standard Charpy specimen dimensions are used except that the thickness is that of the sheet material. The fatigue cracks are extended approximately 0.030 inch beyond the root of the V-notch.

IMPACT TEST

The energy needed to propagate a crack per unit area of fracture surface (W/A) is determined for impact rates of loading. W/A values are calculated by dividing the impact energy by the cross-sectional fracture area and therefore correspond to twice the effective surface energy used in the Griffith-Orowan equation (19 and 20). Tests are carried out on an impact testing machine which has a maximum capacity of 24 ft-lbs and an accuracy of ±0.01 ft-lb in the low energy range.

SLOW BEND TEST

Although the impact test is the simplest and most rapid means of evaluating plane stress fracture toughness, slow bend tests are also performed to study the effect of strain rate on the energy required for crack propagation. It is believed that a better correlation exists between the notched tensile and slow bend tests in view of the fact that the rates of straining are more comparable than between the notched tensile and impact tests. The specimens as tested in slow bend, two or three per condition, are subjected to three-point loading in a tensile machine fitted with a tup and anvils of the same geometry as used for the impact machine. The energy absorbed in propagating a crack is determined from the area under the measured load-deflection curve. Recent work performed by Hanna and Steigerwald (21) show that 4-point loading gives more accurate results than 3-point loading.

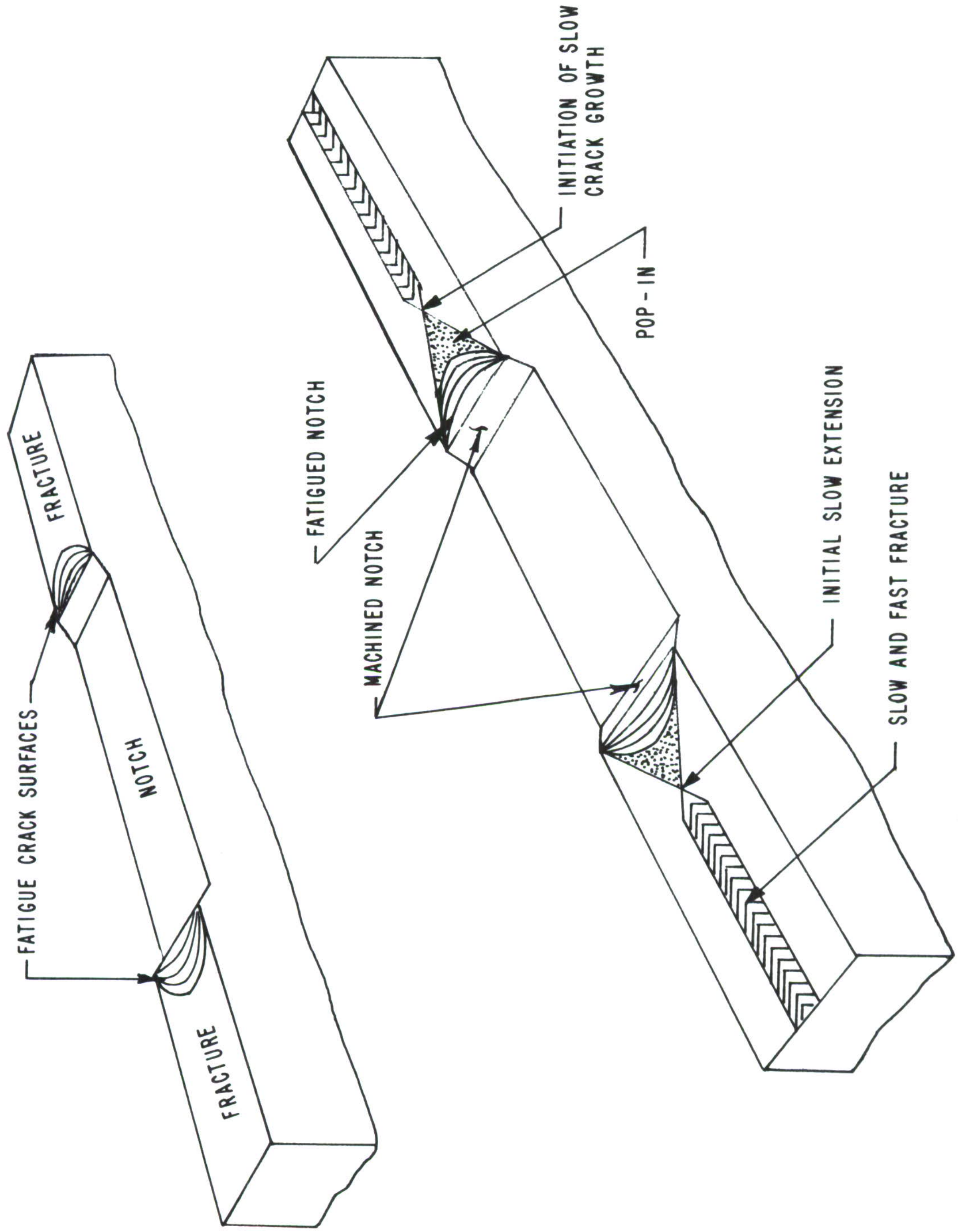

Figure 1. Fracture Surface of Fatigue Cracked Center-Notched Tensile Specimen. Note "through-the-thickness" fatigue crack surfaces resulting from fatiguing in tension-tension.

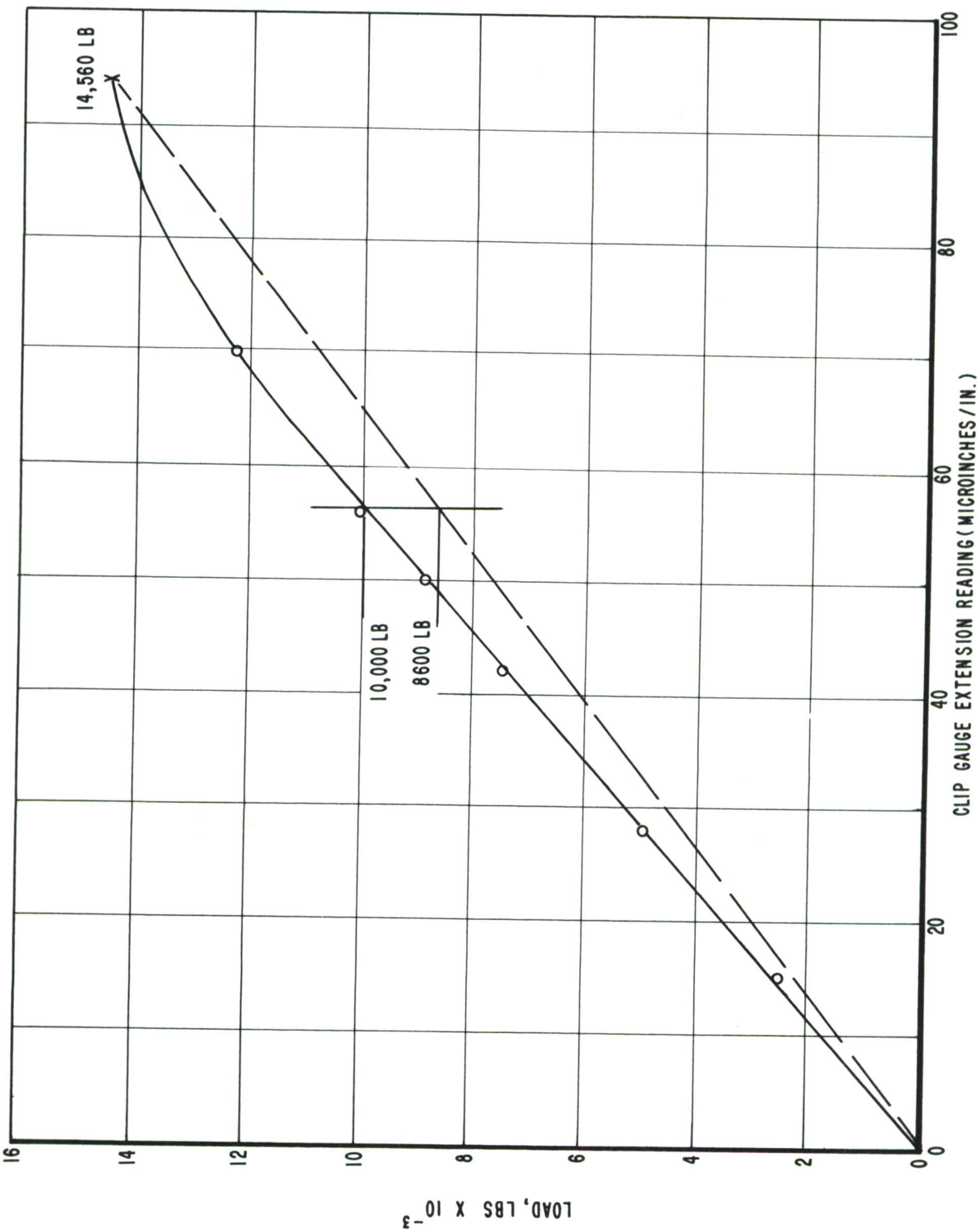

Figure 2. Typical Load-Compliance Curve for Calculation of G_c

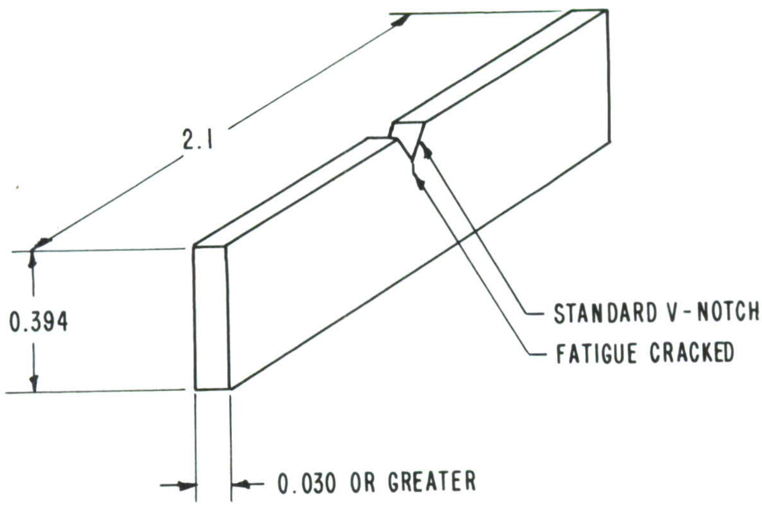

Figure 3. Fatigue Cracked Sheet Charpy Specimen (19)

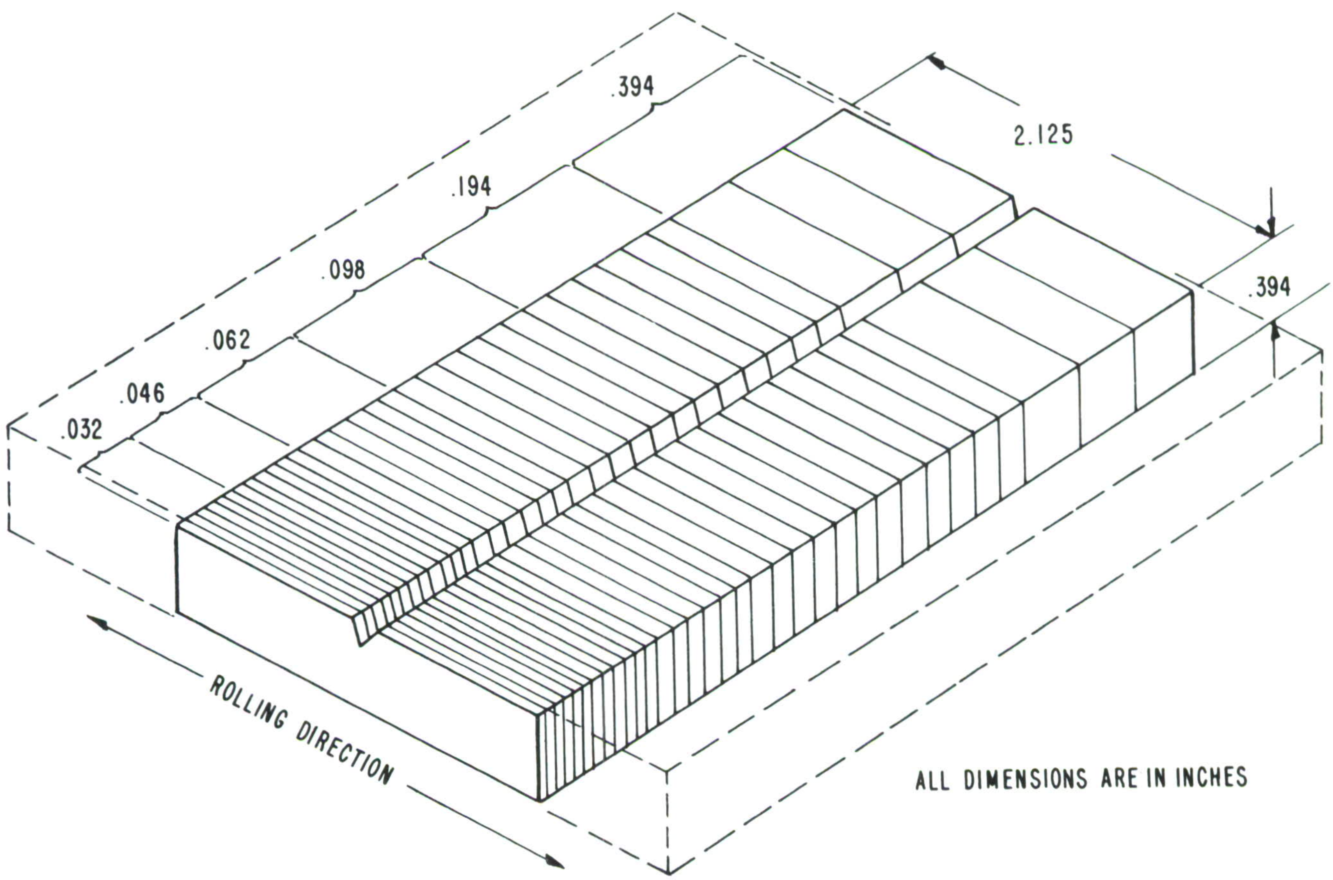

Figure 4. Method Used to Obtain Charpy Specimens (19)

SECTION IV

EFFECT OF TEST VARIABLES ON FRACTURE TOUGHNESS AS DETERMINED BY PRECRACKED CHARPY TESTS

EFFECT OF PRECRACK DEPTH

Experiments have been carried out by Manlabs, Inc. (19) to determine the effect of fatigue-crack depth on the Charpy W/A value in both a tough and a brittle material condition. Charpy specimens were cut from as-received 4340 sheet, V-notched, normalized, fatigue precracked to depths ranging from 0.005 to 0.25 inch, hardened and tempered at 410°F, and finally tested at room temperature and at -40°C.

In Figure 5, W/A is plotted as a function of crack depth. Note that at the higher toughness level (room-temperature tests), fracture toughness decreases approximately linearly with increasing crack depth. The percentage of oblique shear in the fracture surfaces (measured at mid-depth) decreases, in a like manner, from approximately 95 percent in a specimen containing the smallest precrack to approximately 50 percent in a specimen containing the deepest crack. At the lower toughness level corresponding to the -40°C test, the data show little effect of crack depth on W/A, although a slight upward trend in W/A for increasing crack depth may be noted. This upward trend in W/A, in spite of a decreasing percentage of shear lip (from 55% to 35%), is probably the result of the kinetic energy expended in accelerating the specimen halves (in the same direction as the pendulum moves) to some velocity below that of the tup. The absolute value of the kinetic energy lost is essentially independent of crack depth but, percentagewise, assumes much greater proportions as the total energy to fracture a specimen decreases with increasing crack depth. For a depth of fatigue crack below about 50 mils, the calculated kinetic energy loss is approximately 20 in.-lb/in.^{2} and, therefore, is not significant (19).

Note in Figure 5 that the data on the ordinate representing zero fatigue-crack depth (i.e., data from V-notched specimens not fatigue cycled) show much higher W/A values than the fatigue-cracked specimens. This increase in energy represents not only the plastic energy necessary to initiate cracking but also in the case of the specimens broken at -40°C an elastic energy loss. This was evidenced by the considerable velocity at which the specimen halves were ejected from the anvils in a direction opposite to the motion of the pendulum (19). It has been demonstrated previously (17) that in V-notch Charpy tests of high-strength brittle materials without a natural crack, the elastic energy loss can overshadow all other sources of energy absorption combined. In an extreme case (unreported work at Watertown Arsenal Laboratory) where 0.394-inch square Charpy specimens of 1095 steel, heat-treated to R_c62, were tested in slow bend, an energy difference of nearly 20 to 1 was observed between standard V-notched and fatigue-cracked specimens.

EFFECT OF SHEET THICKNESS

The fracture toughness transition curves obtained for the specimens in the various thicknesses tested are illustrated in Figures 6 and 7. Note that in some cases considerable scatter is evident at the high energy levels. Much of the scatter is belived due to variations in the mode of fracture. Beachem (22) has reported two types of oblique shear fracture (sometimes occurring in a single center-cracked tensile specimen) described as "orthogonal" shear and "single" shear. These types of shear are illustrated schematically in Figure 8. In Charpy testing, it was found that when specimens fracture completely or nearly completely in shear, the energy absorbed is generally higher when the shear is of the orthogonal rather than the single type.

Figures 6 and 7 show that the transition temperature curve is lowered with decreasing specimen thickness for both the 800° and 450°F temperatures. The variation of fracture toughness with sheet thickness as measured at a series of constant test temperatures is shown in Figures 9 and 10. For the 800°F temperature, the fracture toughness as measured at 100°C increases to a maximum value which occurs at a thickness value between 0.1 and 0.2 inch (Figure 9). This initial increase in fracture toughness is believed to reflect an increase in the volume of plastically deformed metal that is associated with the occurrence of lateral contraction preceding the tip of the advancing crack. This is indicated by the results of measuring the extent of lateral contraction (adjacent to the fracture) in a direction parallel to the specimen length. Table I shows that the linear extent of lateral contraction in this direction is approximately equal to the specimen thickness up to a thickness value of about 0.1 inch, and does not change by more than about 25 percent with increase in thickness up to 0.394 inch. Considering the fact that for specimen thickness up to at least 0.1 inch the mode of fracture is entirely shear, it appears that the volume of plastic deformation is approximately equal to the specimen thickness (19). Consequently, the W/A value, which is impact energy divided by the product of width and thickness, should be approximately proportional to the specimen thickness; and therefore, should increase with thickness as observed. With increase in thickness beyond that corresponding to the toughness maximum, the observed decrease in fracture toughness is believed to reflect the occurrence of flat fracture to an increasing extent in the center portion of the specimens.

Figure 11 presents percentage of oblique shear (not to be confused with percent fibrosity) measured at mid-depth below the notch as a function of test temperatures, with specimen thickness as a parameter. At a given test temperature, the percentage of shear was found to increase with decreasing specimen thickness. In the thicker specimens, 100 percent oblique shear did not occur at any of the test temperatures studied by Hartbower and Orner (19).

EFFECT OF TYPE OF TEST ENVIRONMENT

Because of the possibility that environmental effects might affect fracture toughness,

TABLE I

VARIATION OF EXTENT OF LATERAL CONTRACTION WITH THICKNESS OF 4340 STEEL PRECRACKED CHARPY SPECIMENS TEMPERED AT 800°F AND TESTED AT 100°C (19)

Specimen Thickness inch	Linear Dimension* of Lateral Contraction inch
0.032	0.030-0.035
0.045	0.039-0.047
0.062	0.062-0.072
0.100	0.090-0.107
0.197	0.106-0.150
0.394	0.125-0.154

* Measured in direction parallel to specimen length

a limited number of experiments was performed using precracked Charpy specimens of X200 sheet tempered at 700°F (19). Slow bend tests were conducted in three environments: air, India ink, and dye penetrant. The results of these experiments are illustrated in Figure 12. The numbers above each bar represent the maximum loads developed in the slow bend test. Note that both fracture toughness and maximum load are lowered by the use of ink as compared with air; whereas, with dye penetrant, both the maximum load and the toughness are higher than the values obtained in air.

To confirm these results, one G_c tensile specimen was tested in each of the environments (19). The load-elongation curves from these tests are superimposed in Figure 13. Consistent with the Charpy results, the maximum load is highest with dye penetrant, intermediate with air, and lowest with ink. Because of the load differences, the extent of slow cracking, as indicated by the deviation of the curve from linearity at the point of fast fracture, is increased with ink and decreased with dye penetrant. Thus, the effect of ink is to lower the stress at which the slow crack growth occurs and, therefore, the crack attains a greater length before the onset of fast fracture than it would without ink. This increased critical crack length may, in some cases, progress to the point where the boundary conditions for the G_c analysis are no longer satisfied. On the other hand, the higher loads obtained with dye penetrant may be the result of the dye protecting the advancing crack front from atmospheric moisture (19). The fact that a small amount of unstable slow crack propagation occurs even in the presence of the dye suggests that even it may adversely affect the load required for crack propagation. Still higher loads might be obtained if deleterious environmental elements could be entirely eliminated during the test.

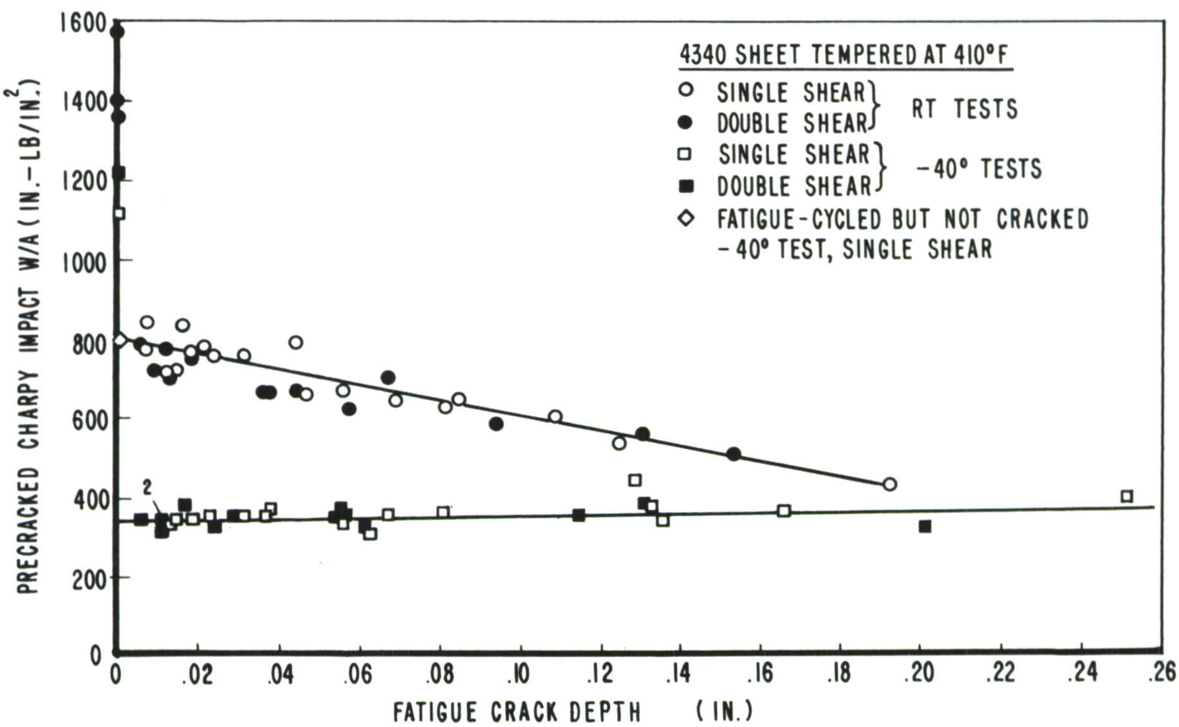

Figure 5. Effect of Fatigued Precrack Depth on Charpy W/A Impact Values (19)

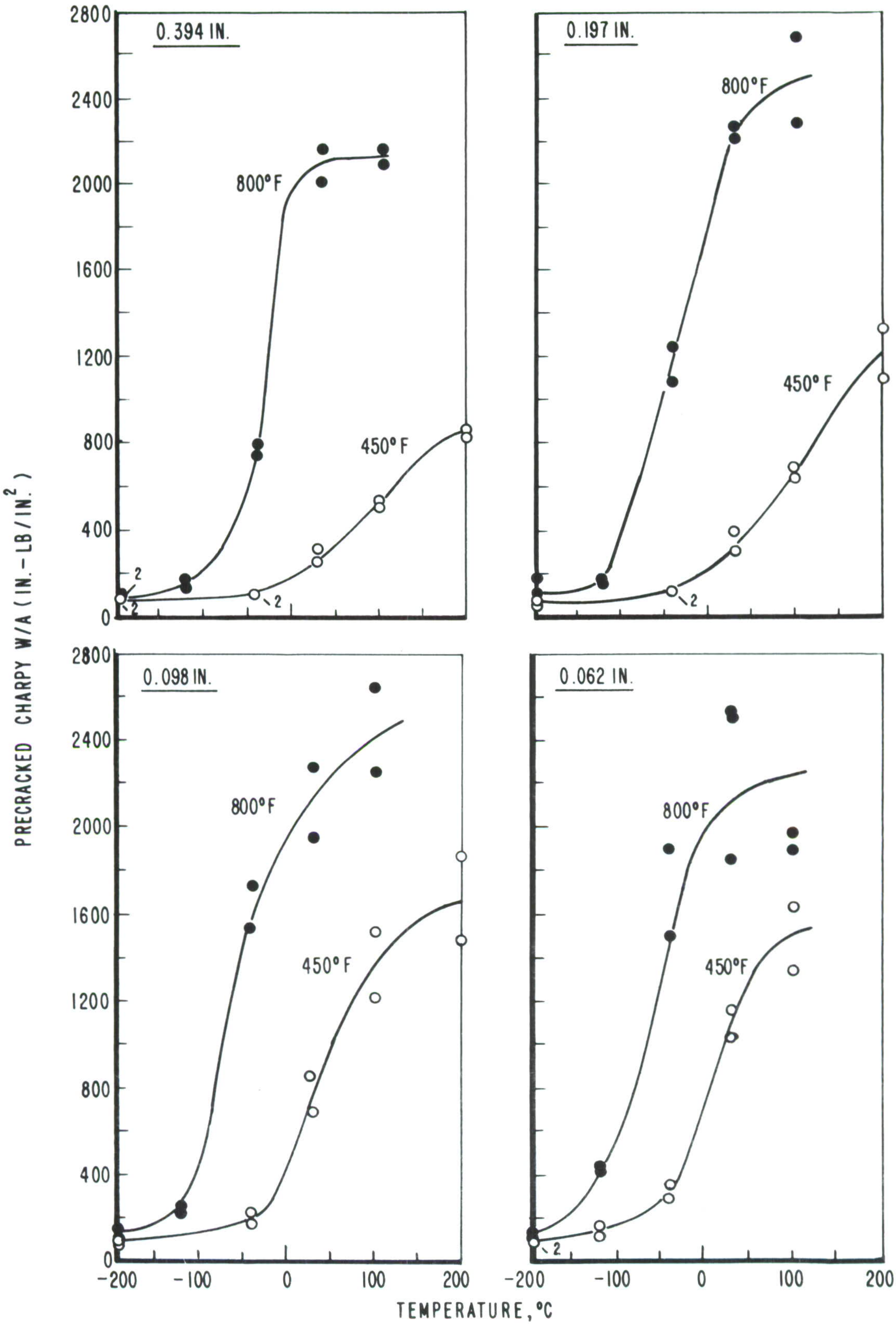

Figure 6. Effect of Sheet Thickness on Impact Fracture Toughness of 4340 Tempered at 450° and 800°F. (0.394 in. to 0.062 in.)

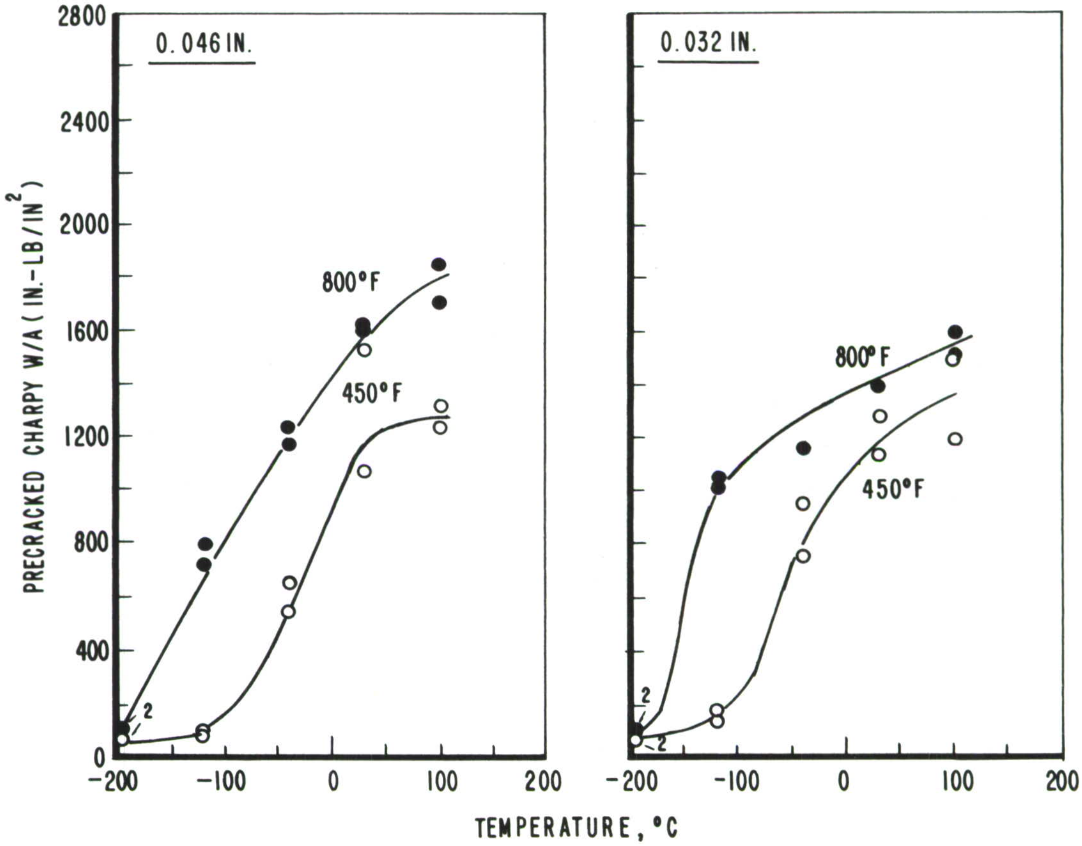

Figure 7. Effect of Sheet Thickness on Fracture Toughness of 4340 Tempered at 450° and 800° F (0.046 in. and 0.032 in.)

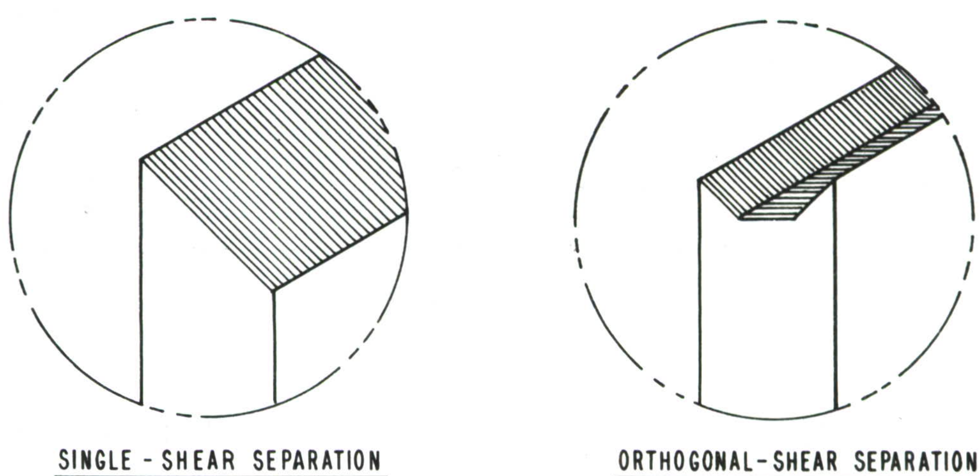

Figure 8. Schematic Illustrations of Single-Shear and Orthogonal-Shear Separations

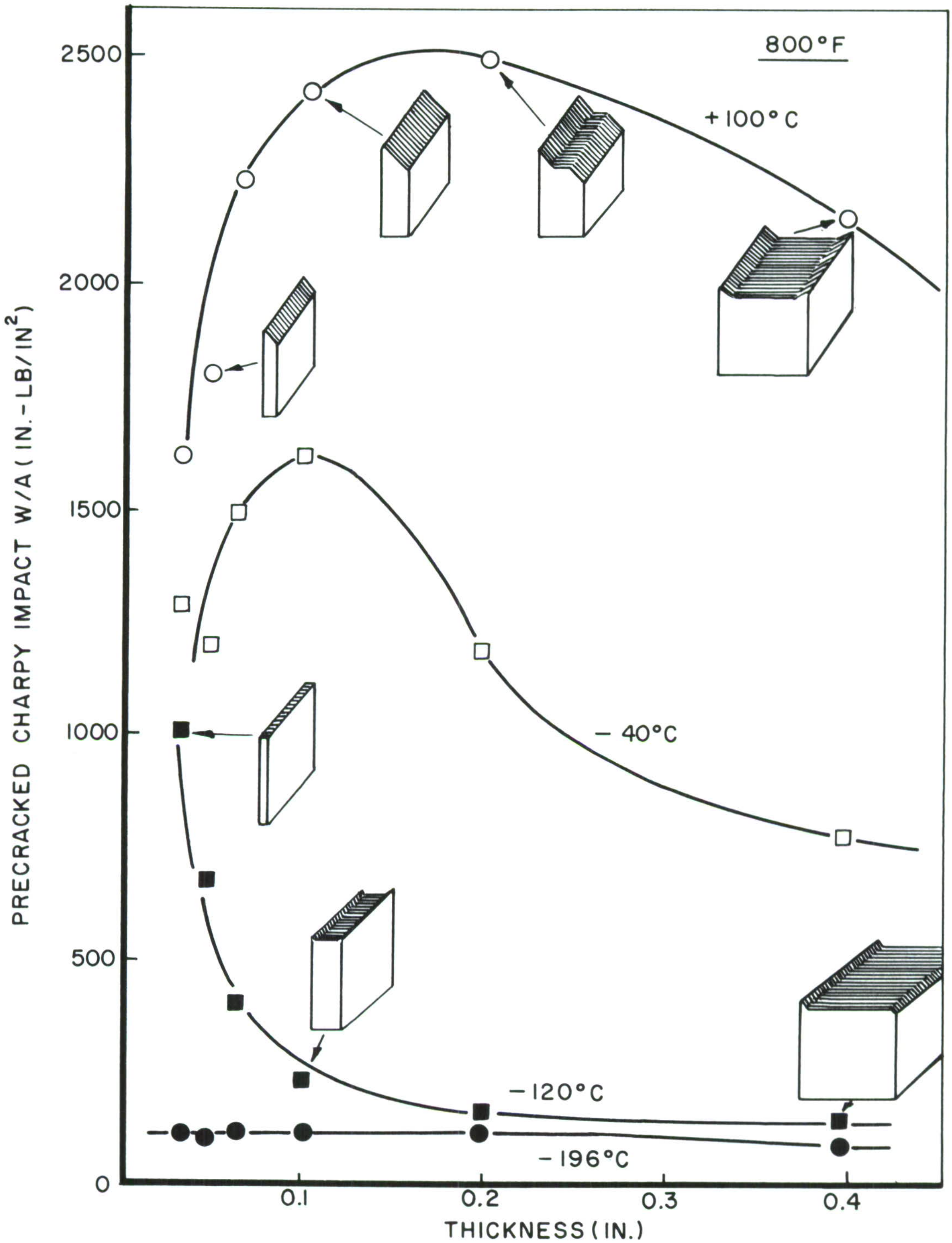

Figure 9. Fracture Toughness at Constant Test Temperatures vs Specimen Thickness for 4340 Tempered at 800° F

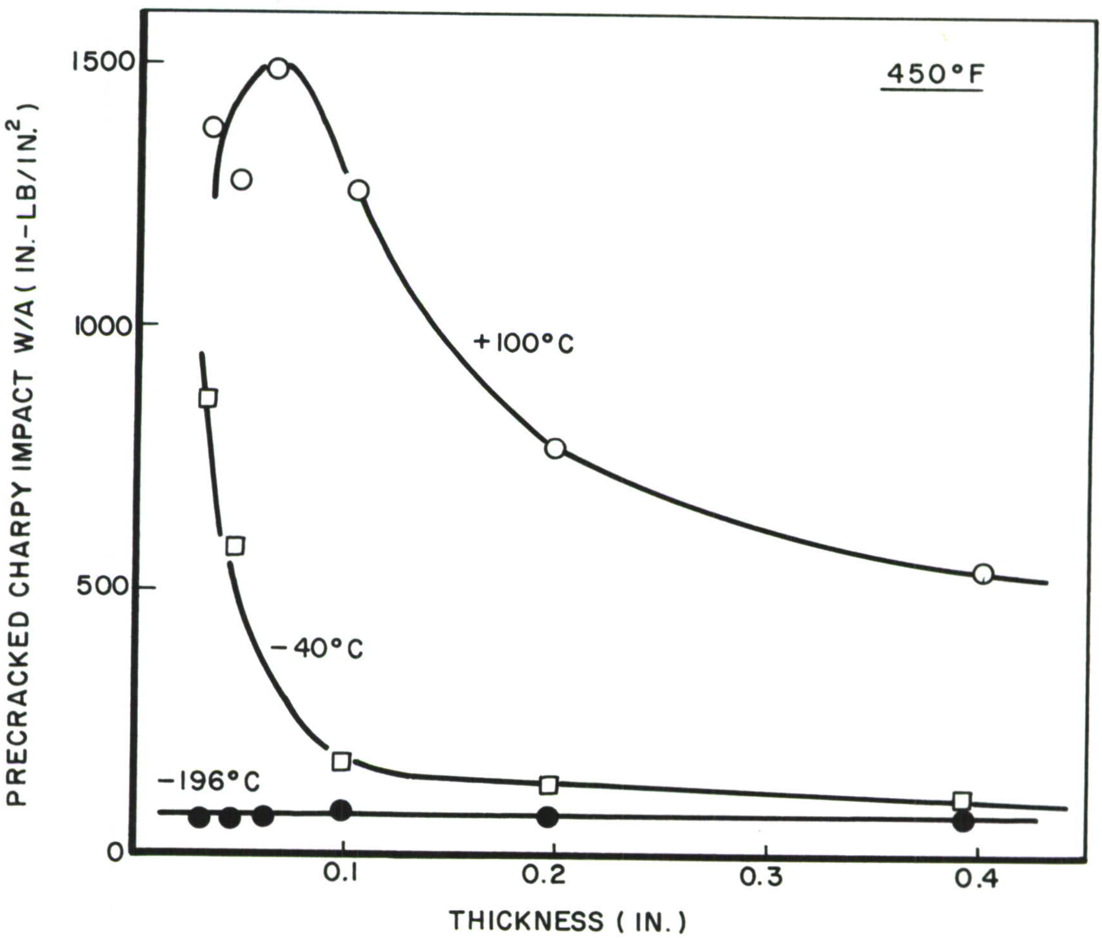

Figure 10. Fracture Toughness at Constant Test Temperature vs Specimen Thickness for 4340 Tempered at 450° F

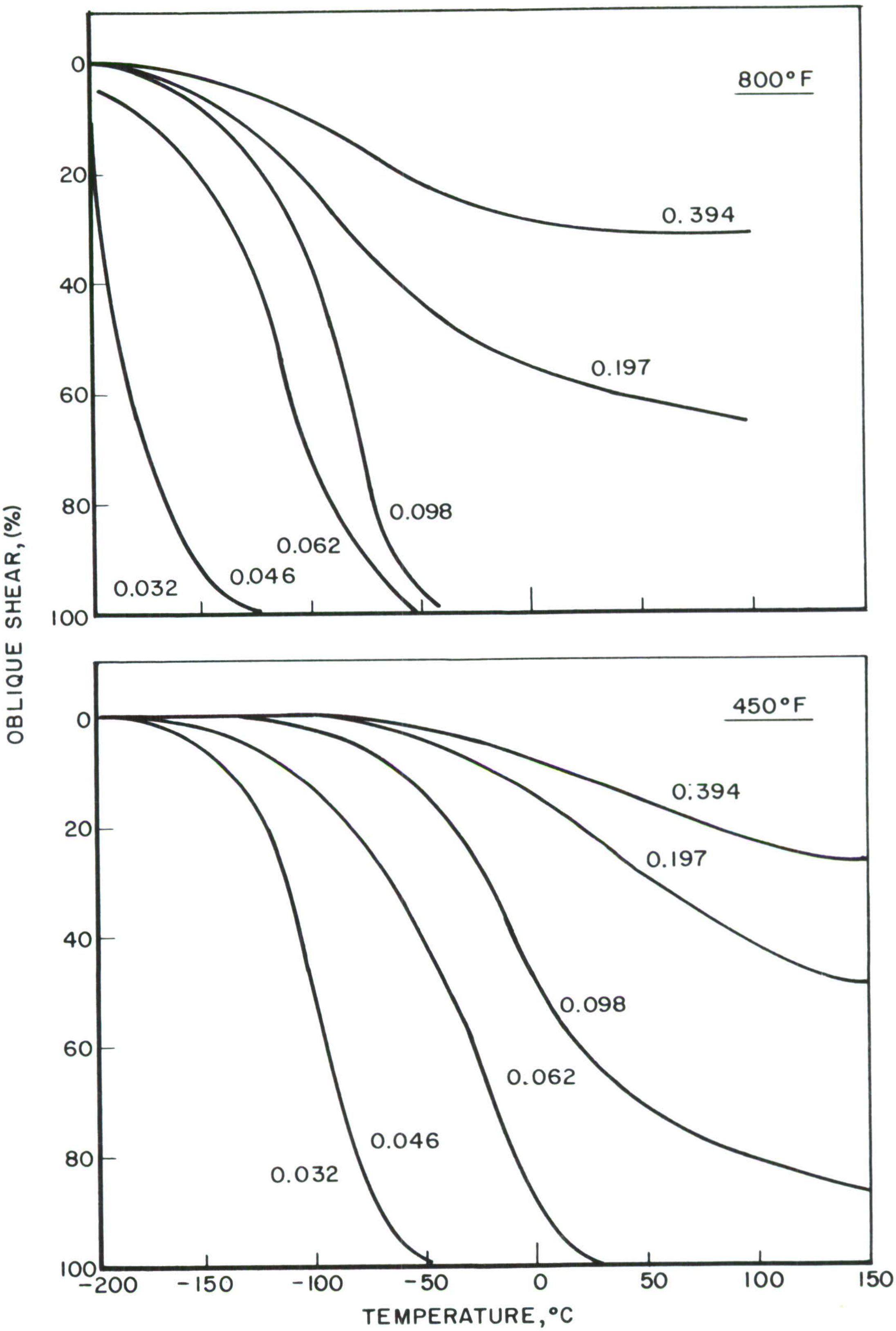

Figure 11. Effect of Thickness on the Percentage of Oblique Shear in the Charpy Fracture Surface for 4340 Tempered at 800° and 450° F

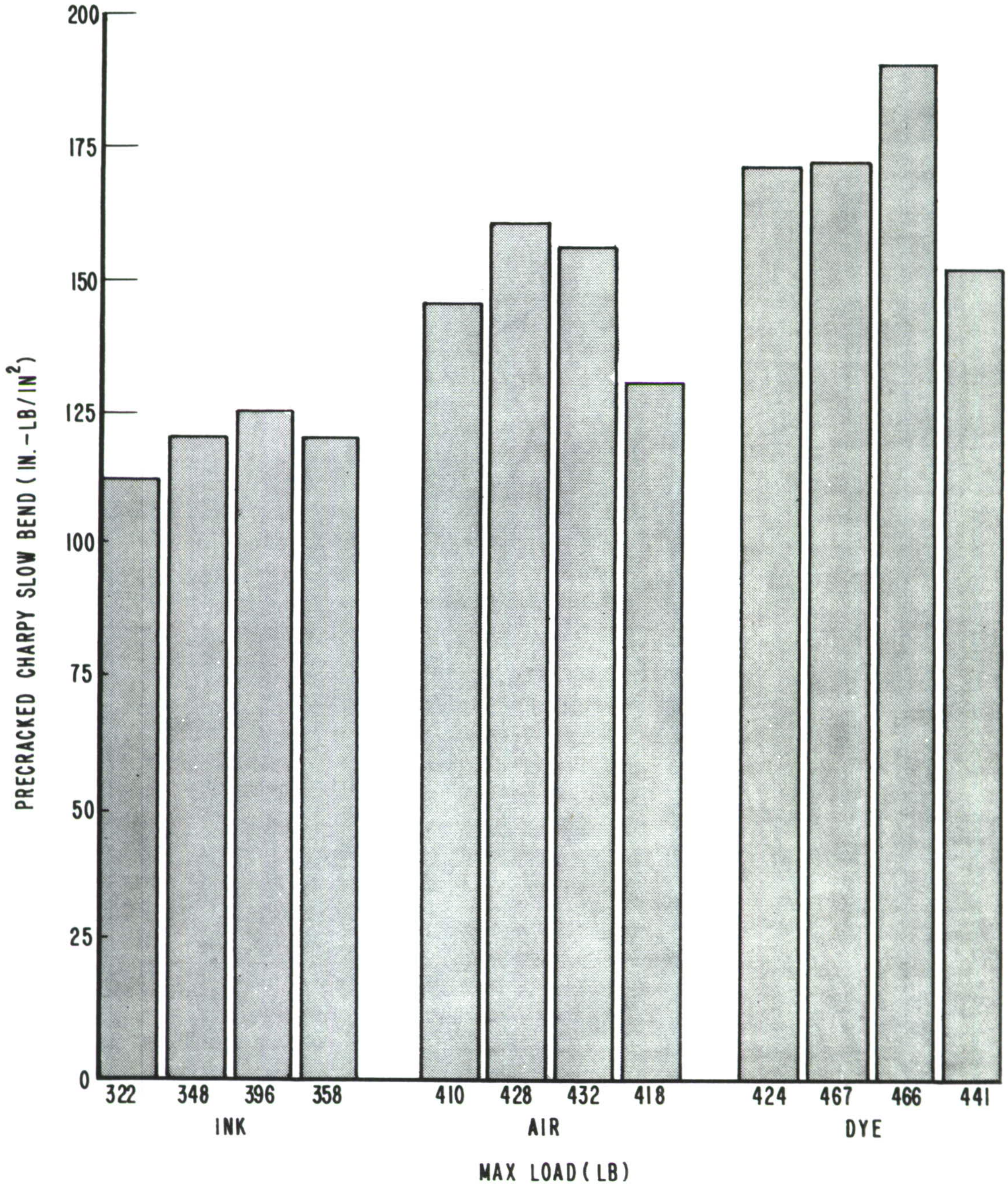

Figure 12. Environmental Effects of Air, Ink and Dye Penetrant Determined by Charpy Slow Bend Tests of X200 (0.40% C) Tempered at 700°F (19)

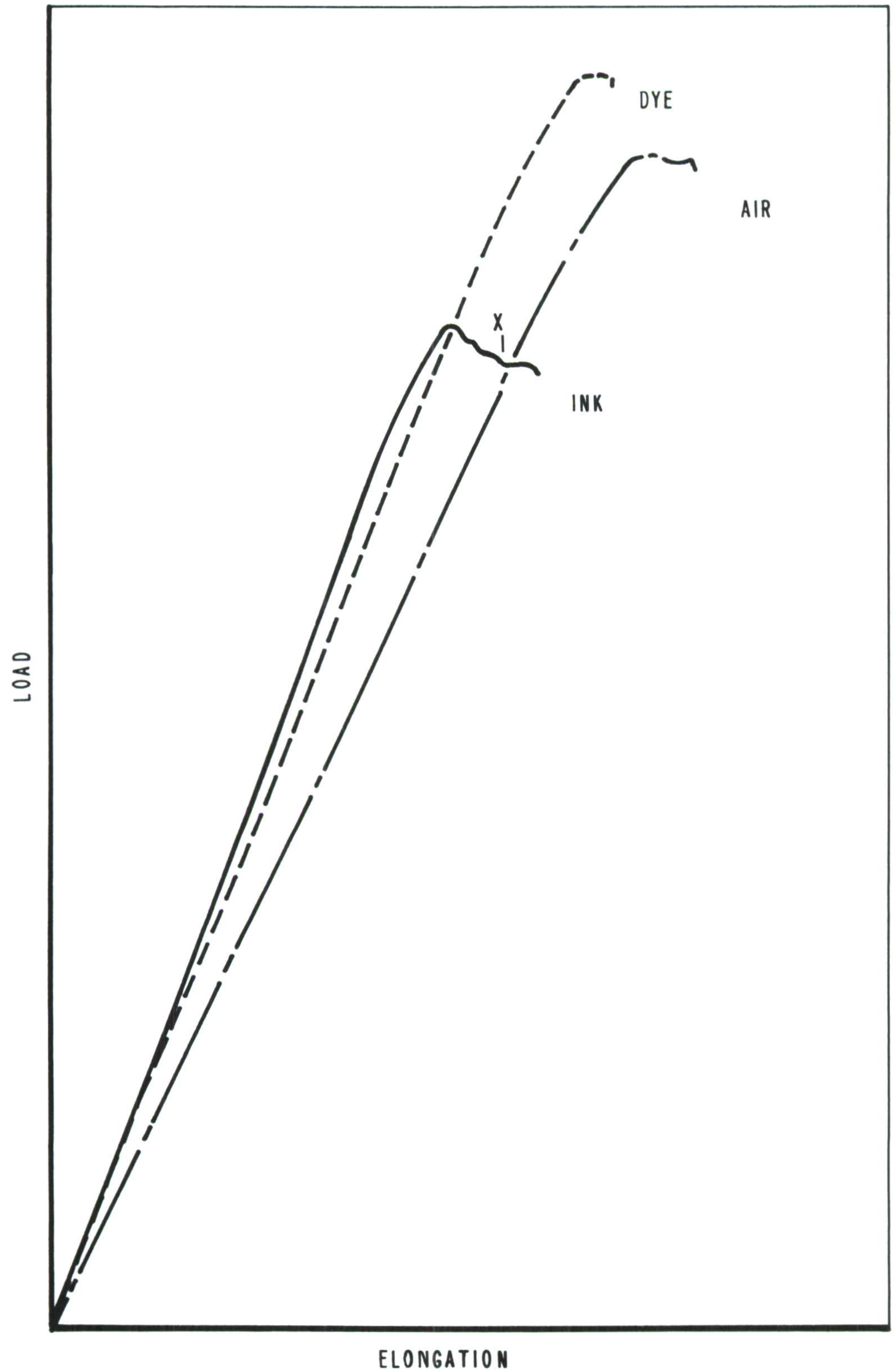

Figure 13. Load-Deflection Diagrams for X200 (0.40% C) Tempered at 700°F as Determined by Center-Notched Tensile Tests Conducted in Air, Ink, and Dye Penetrant Environments (19)

SECTION V

DISCUSSION OF RELATIONSHIP BETWEEN G_c AND W/A

BASIS OF PREDICTION OF RESIDUAL STRENGTH

Relatively good agreement (within 15%) was obtained between (a) measured values of residual strength of four medium-carbon alloy steels in a total of nine tempered conditions based on center-notched tensile tests and (b) predicted values based on slow bend (W/A) and use of a modified Griffith-Orowan relation (20). The modified Griffith-Orowan relation used by Hartbower is as follows:

$$\sigma = \left[\frac{(W/A)\,E}{\pi a_o}\right]^{1/2} \tag{3}$$

where a_o is the critical initial semicrack length. Equation 3 is an empirical relation which has as its main justification the relatively good agreement obtained in predicting residual strength in the cases studied. However, there is additional justification based on a comparison with the more exact relation given by Irwin for a wide plate:

$$\sigma = \left[\frac{G_c E\left[1 - \frac{1}{2}\left(\frac{\sigma}{\sigma_{ys}}\right)^2\right]}{\pi a_c}\right]^{1/2} \tag{4}$$

An alternate form of this relation is as follows:

$$a_c = \frac{G_c E}{\pi\left[1 - \frac{1}{2}\left(\frac{\sigma}{\sigma_{ys}}\right)^2\right]^{-1}\sigma^2} \tag{5}$$

where a_c is the critical semicrack length. Equating Equations 3, 4, and 5 gives the following condition for Equation 3 to be valid:

$$\frac{(W/A)}{G_c} = \left(\frac{a_o}{a_c}\right)\left[1 - \frac{1}{2}\left(\frac{\sigma}{\sigma_{ys}}\right)^2\right] \tag{6}$$

Since $a_o = a_c$, it appears that Equation 3 is valid only if $(W/A)/G_c$ happens to equal the bracket term, which can vary from 0.5 (for $\sigma/\sigma_{ys} = 1.0$) to 1.0 (for $\sigma/\sigma_{ys} = 0$). For materials with as low a $(W/A)/G_c$ ratio as 0.4, the error involved in calculating σ by Equation 3 varies from -10 percent for $\sigma/\sigma_{ys} = 1.0$ to about -35 percent for $\sigma/\sigma_{ys} = 0$; and with as high a $(W/A)/G_c$ ratio as 1.0, from about +40 percent for $\sigma/\sigma_{ys} = 1.0$ to zero for $\sigma/\sigma_{ys} = 0$. Since the error involved in predicting σ from Equation 3 can vary between the limits of about -35 percent to +40 percent, it appears that the error is likely to be within about ±20 percent in the majority of cases.

ANALYSIS OF DISCREPANCIES BETWEEN PREDICTED AND MEASURED RESIDUAL STRENGTH VALUES

Values of the critical plane stress intensity parameter (K_c) can be calculated on the basis of the following equations (23).

$$K_c = \sigma\left[W \tan\left(\frac{\pi a_c^*}{W}\right)\right]^{1/2} \tag{7}$$

$$a_c^* = a_c + r_p = a_c + \frac{1}{2\pi}\left(\frac{K_c}{\sigma_{ys}}\right)^2 \tag{8}$$

where

- w = width of plate
- a_c^* = critical crack semilength corrected for plastic zone at tip
- a_c = measured critical crack semilength
- r_p = radius of plastic zone at crack tip

When the width (w) of the test panels is about eight inches, the Irwin (24) relation, Equation 4, which is applicable to wide plates can be used as a good approximation for the more exact Equation 7. The ratio of predicted residual strength (σ_H) using the Hartbower relation, Equation 3, to predicted residual strength (σ_I) using the Irwin relation, Equation 4, is as follows:

$$\frac{\sigma_H}{\sigma_I} = \left[\frac{(W/A)}{G_c\left[1 - \frac{1}{2}\left(\frac{\sigma_I}{\sigma_{ys}}\right)^2\right]} \right]^{1/2} \qquad (9)$$

In summary, the modified Griffith-Orowan relation based on precracked Charpy slow bend tests was found to give predicted values of residual strength that agreed within about 15 percent with the measured values for 19 out of 20 Douglas panels representing two heats each of five materials. Analysis has shown that this relatively good agreement is due to particular combinations of $(W/A)/G_c$ and σ/σ_{ys} ratios for a given material and condition, and that agreement within about ±20 percent can be expected in the majority of cases. However, since the agreement may conceivably be as poor as about ±40 percent, a direct G_c determination and the use of the Irwin relations, Equations 4 or 7, is preferable particularly if the application is a critical one.

If, in addition to the slow bend (W/A) fracture toughness and initial through-the-thickness semilength (a_o), the σ_{ys} of a wide plate of a given material is known, a more accurate prediction can be made of the residual strength by using an approximate form of the Irwin relation, Equation 4, instead of the modified Griffith-Orowan relation Equation 3. Since the ratio of slow bend $(W/A)/G_c$ is generally found to be in the range of 0.4 to 1.0, this ratio can be taken as equal to 0.7, with an uncertainty of 40 percent. Accordingly, the Irwin relation becomes:

$$\sigma = \left[\frac{1.4(W/A)\, E \left[1 - \frac{1}{2}\left(\frac{\sigma}{\sigma_{ys}}\right)^2\right]}{\pi a_o} \right]^{1/2} \qquad (10)$$

and the maximum error in predicted σ values should be about ±20 percent. For 4335-V steel, $(W/A)/G_c$ was taken as equal to 0.85 rather than 0.7 in using an alternate form of Equation 4 for calculating a_c values. However, this was based on the range of $(W/A)/G_c$ ratios for similar medium-carbon low-alloy steels as determined by Hartbower and Orner (25).

CORRELATION OF PRECRACKED CHARPY AND CENTER-NOTCHED TENSILE TEST

It was found that residual strength (σ_c) as determined by a center-notched G_c^* tensile test can be predicted quite accurately by using values of precracked Charpy slow bend W/A and initial crack length ($2a_o$) in conjunction with the Griffith-Orowan equation:

$$\sigma_c = \sqrt{\frac{(W/A)E}{\pi a_o}} \qquad (11)$$

The (W/A) value in Equation 11 is considered equal to twice the effective energy for crack propagation (P), which appears in the Griffith-Orowan equation. This is due to the fact that W/A is determined by dividing the total energy to fracture by the cross sectional area, whereas P refers to the energy associated with each fractured surface.

As shown in Figure 14, good agreement was obtained between observed and calculated values of residual strength for the five steels in a total of nine heat-treated conditions. W/A slow bend values are given in Figure 15. It should be noted that the observed residual strength values were determined in the G_c tests conducted with a compliance gauge, i.e., in the absence of ink staining. For the G_c^* tests conducted using ink staining (refer to Figure 16), the agreement between observed and calculated residual strengths is only fair as shown by Figure 17. In all cases, the calculated residual strength is 10 to 30 percent higher than the corresponding observed value.

Figure 18 represents the results obtained in a previous investigation of H-11 by Fopiano (26). This steel was austenitized at 1850°F for 30 minutes, air-cooled to room temperature and double-tempered at the indicated temperatures for 2 plus 2 hours. The center-notched tensile specimens were 2 inches wide and an initial crack length of about 0.76 inch was used. The agreement between observed and calculated values of residual strength is better than 5 percent in six cases, and from 15 to 25 percent in three cases.

Similar comparisons were made for 2024-T81, 2024-T86, 2014-T6 and 7075-T6 aluminum alloys in the form of 80 mil sheet obtained from alcoa Research Laboratories. The G_c^* tensile tests were conducted on 6-inch wide center-notched tensile tests. Figures 19 and 20 show the calculated residual strength as a function of initial crack length based on the slow bend W/A values. The plotted data represent the experimentally determined values. Note the excellent agreement between predicted and actual residual stress values. The fact that a number of known variables have been neglected in this comparison (slow crack growth: plastic extension factor, etc.) suggests that the high accuracy of these predictions is fortuitous. Nevertheless, the fact that residual strength could be predicted with a high degree of accuracy should be of considerable importance from an engineering point of view.

Center-notch tensile results previously obtained at the Watertown Arsenal Laboratory (16) were re-evaluated in a similar manner. These tests were made with 1/8 inch high-strength aluminum sheet alloy using 3- and 6-inch wide center-notched tensile specimens. The center notches were 1/3 of the specimen width in all cases, but mechanically sharpened instead of fatigue-cracked. Figure 21 shows that the predicted values are consistently high for the 3-inch wide specimens and consistently low for the 6-inch wide specimens. The agreement between observed and calculated residual strength values is mainly within 15 percent although it ranges up to about 30 percent (19).

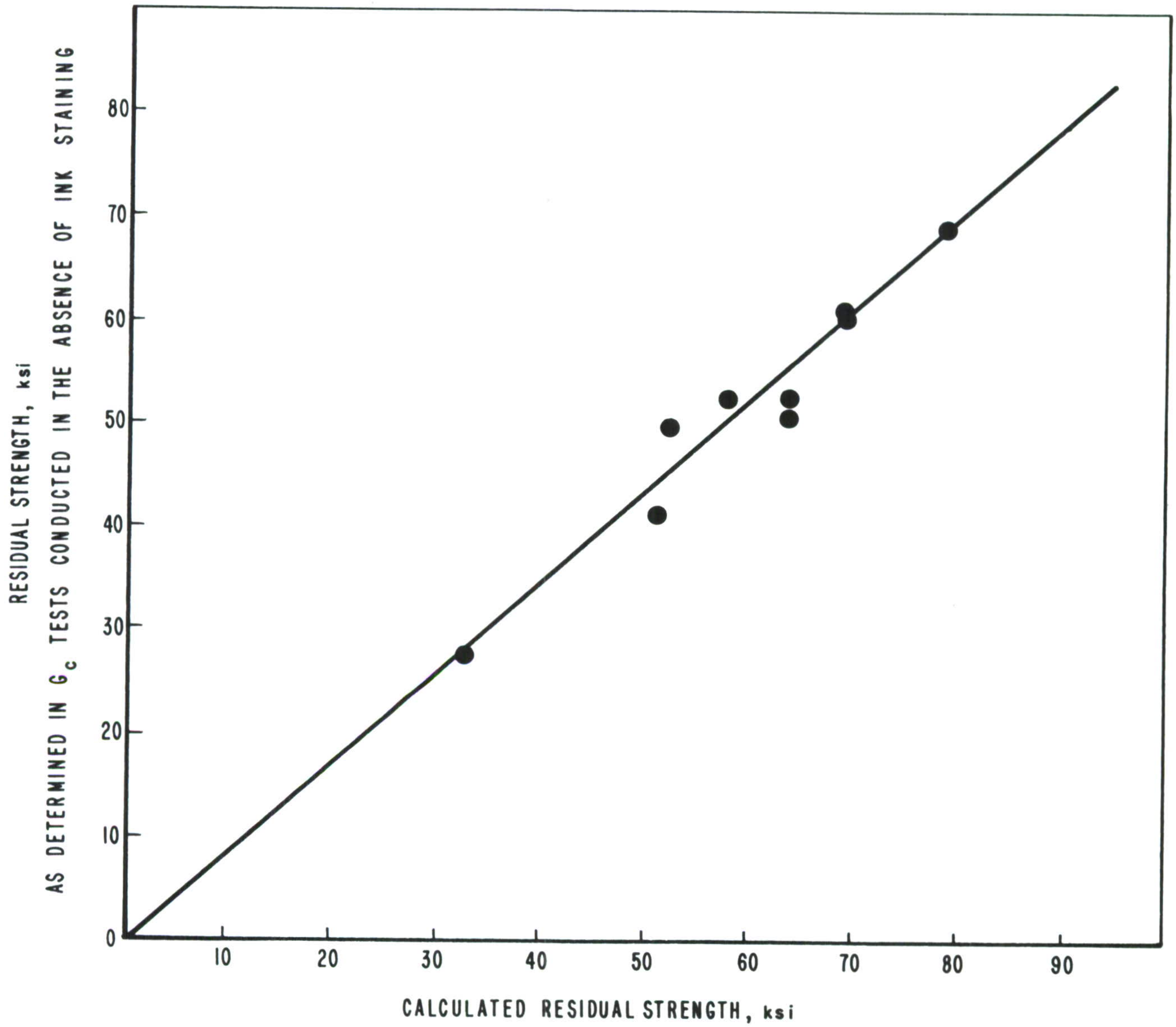

Figure 14. Comparison of Observed Residual Strength vs Residual Strength Calculated on the Basis of Average Slow Bend W/A Values for Five Steels in a Total of Nine Heat-Treated Conditions (19)

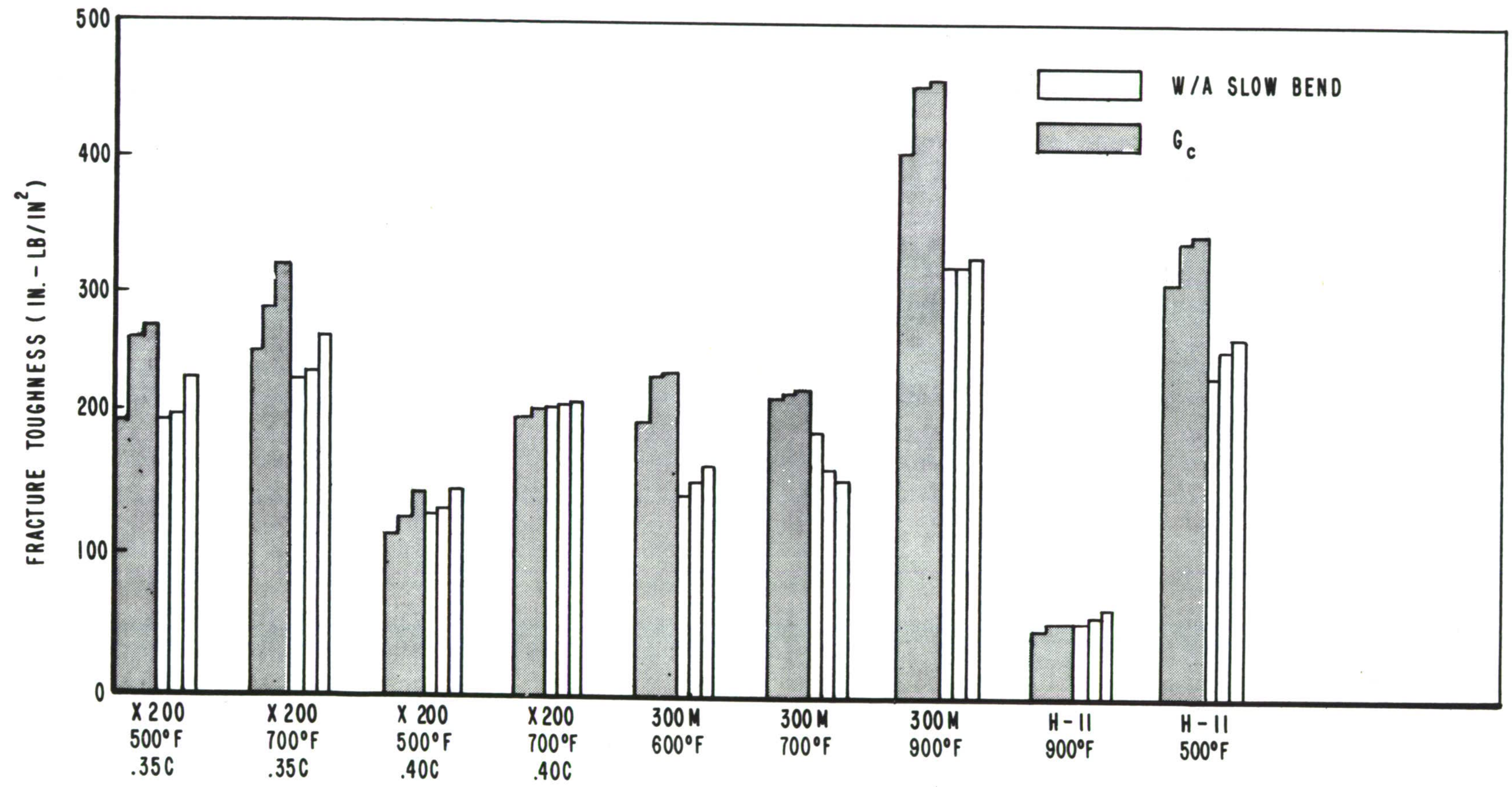

Figure 15. W/A vs G_c for Five Steels. In determining G_c values, measurements of critical crack length were made using a compliance gauge technique. (19)

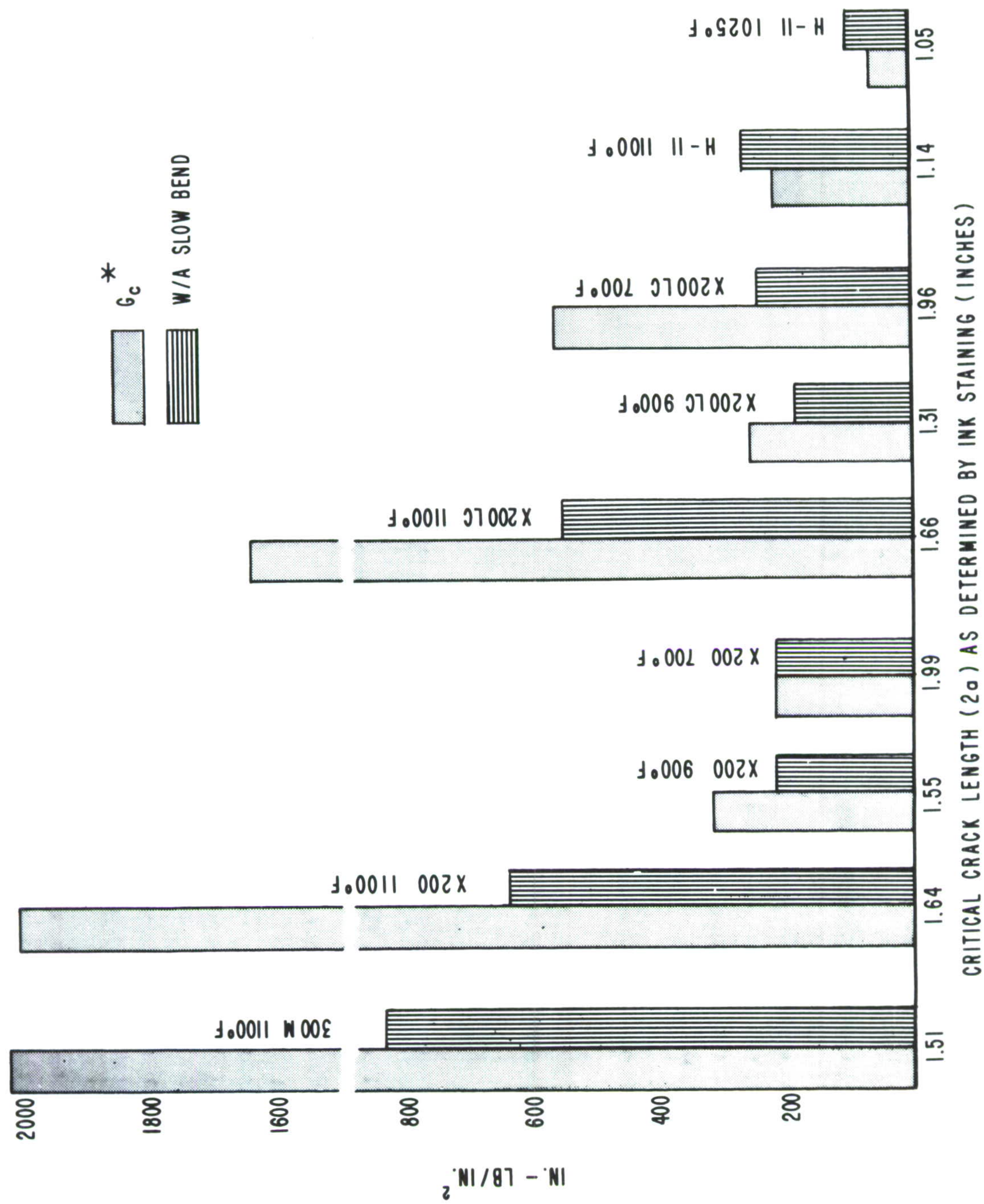

Figure 16. Comparison of Fatigue-Cracked, Center-Notched, 3-Inch Wide Tensile G_c^* With Fatigue-Precracked, Slow-Bend Charpy W/A Test Results. (19)

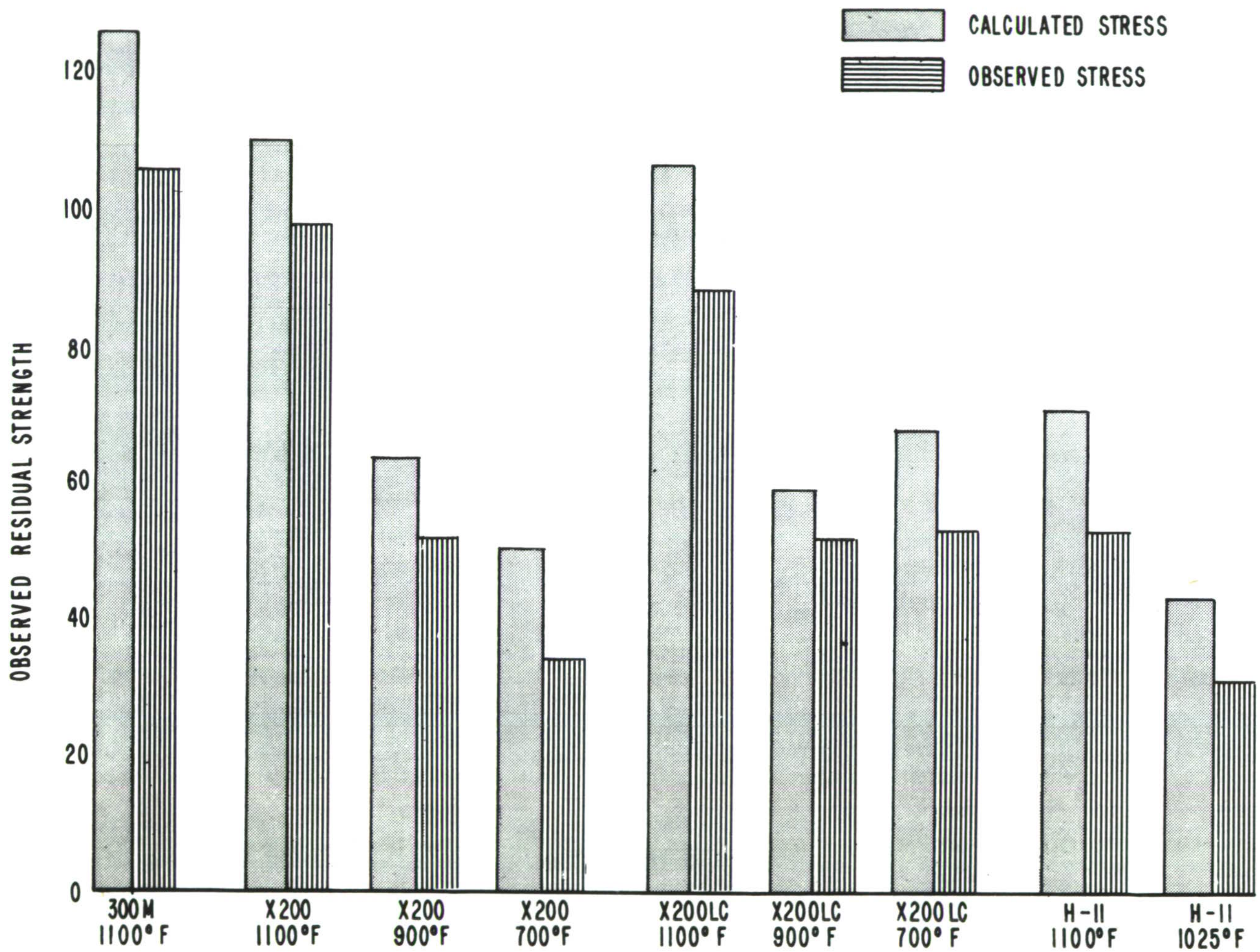

Figure 17. Calculated and Observed Residual Fracture Strength in 3-Inch Wide Steel Tensiles With Fatigue-Cracked Center Notches. The tempering temperatures follow the alloy designations; the length of notch ($2a_o$) was approximately 1/3 the specimen width. (19)

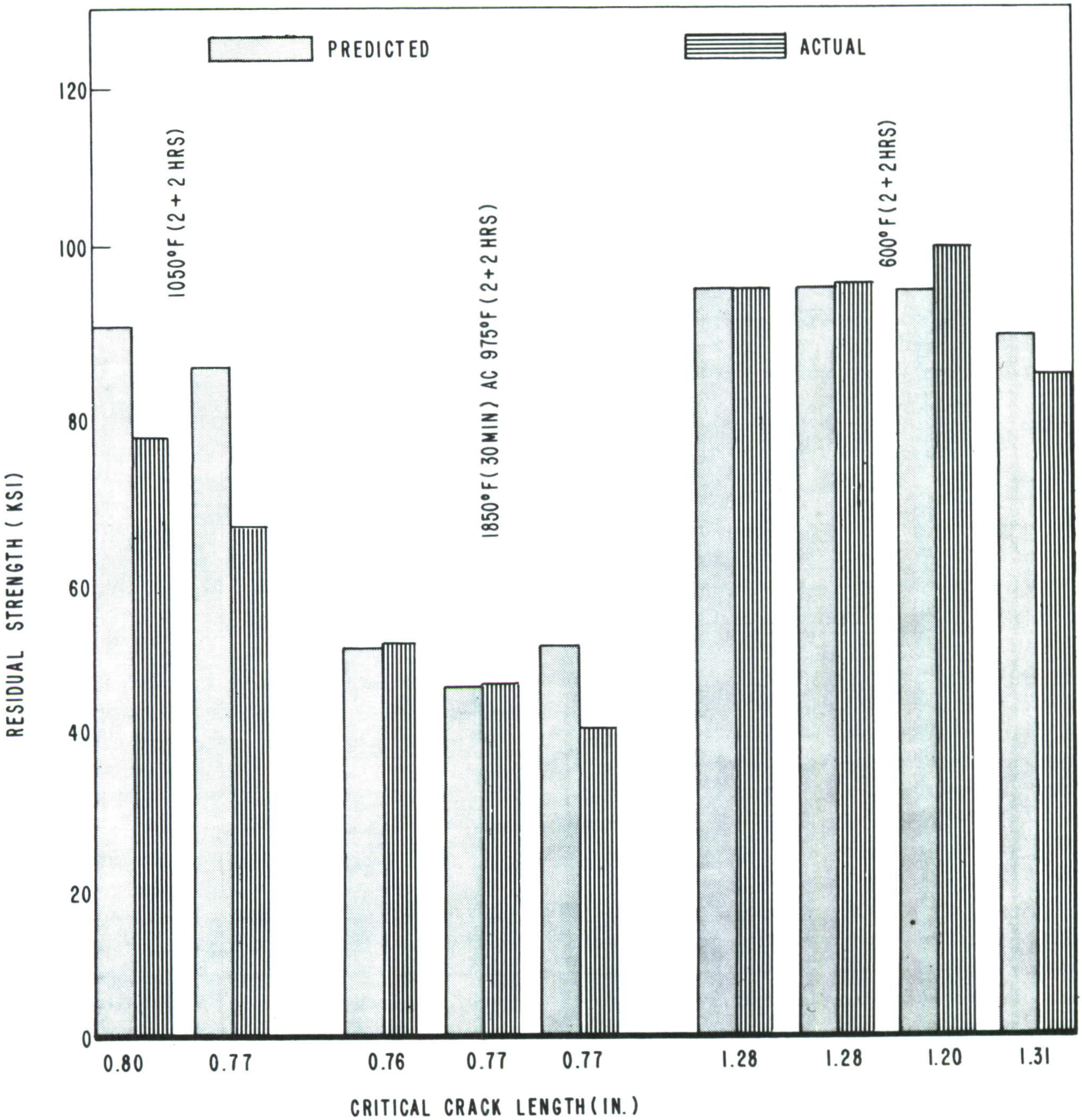

Figure 18. Comparison of Observed and Calculated Residual Strength of Steel Tempered as Indicated. The initial crack length was 0.75 to 0.77 inch. (19)

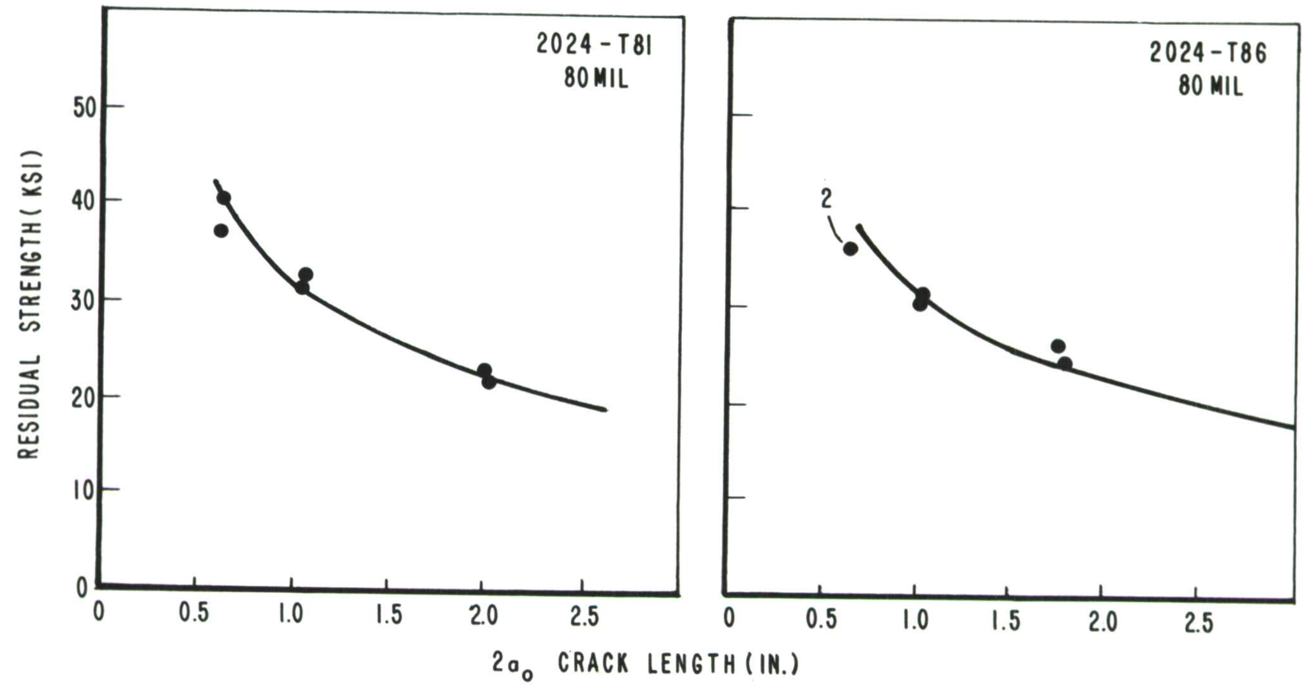

Figure 19. Prediction of Residual Strength of Fatigue-Cracked, Center-Notched, 6-Inch Wide Aluminum Tensile Specimens Based on Charpy Slow-Bend W/A and the Griffith-Orowan Relationship. The curves represent calculated maximum stress values, while the plotted data were obtained from center-notch tensile tests. (19)

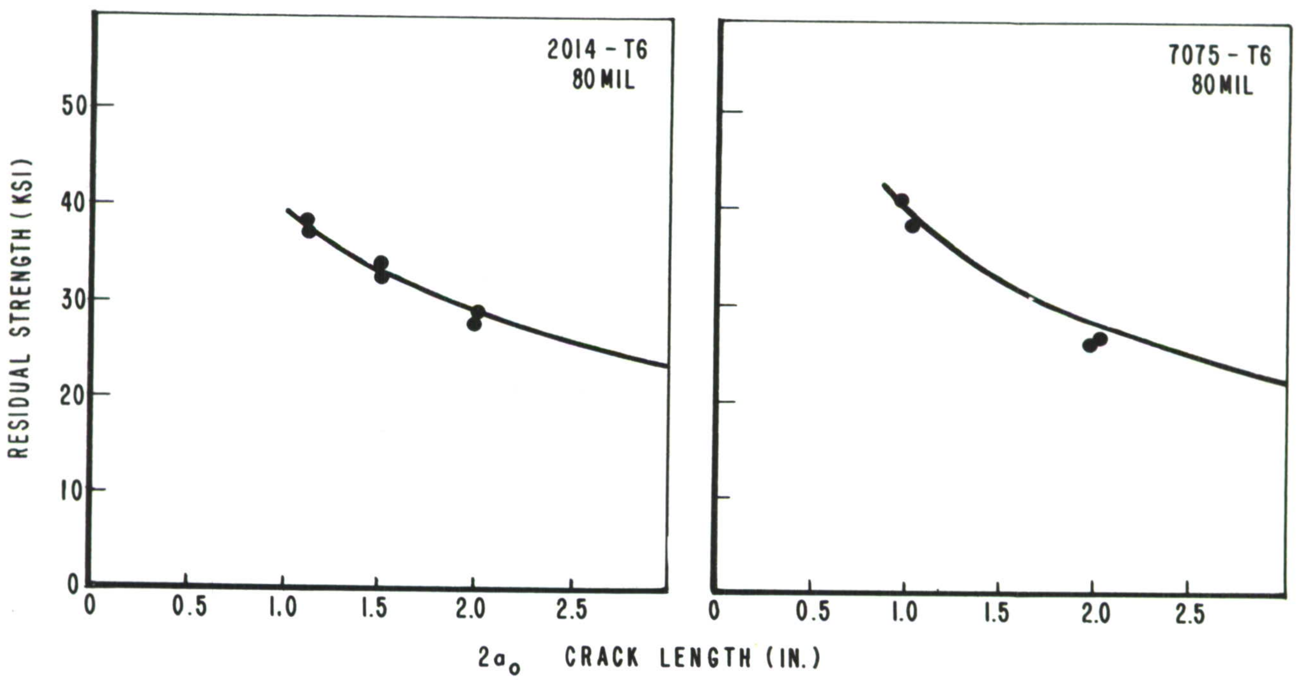

Figure 20. Prediction of Residual Strength of Fatigue-Cracked, Center-Notched 6-Inch Wide Aluminum Tensile Specimens Based on Charpy Slow-Bend W/A and the Griffith-Orowan Relationship. The curves represent calculated maximum stress values, while the plotted data were obtained from center-notch tensile tests. (19)

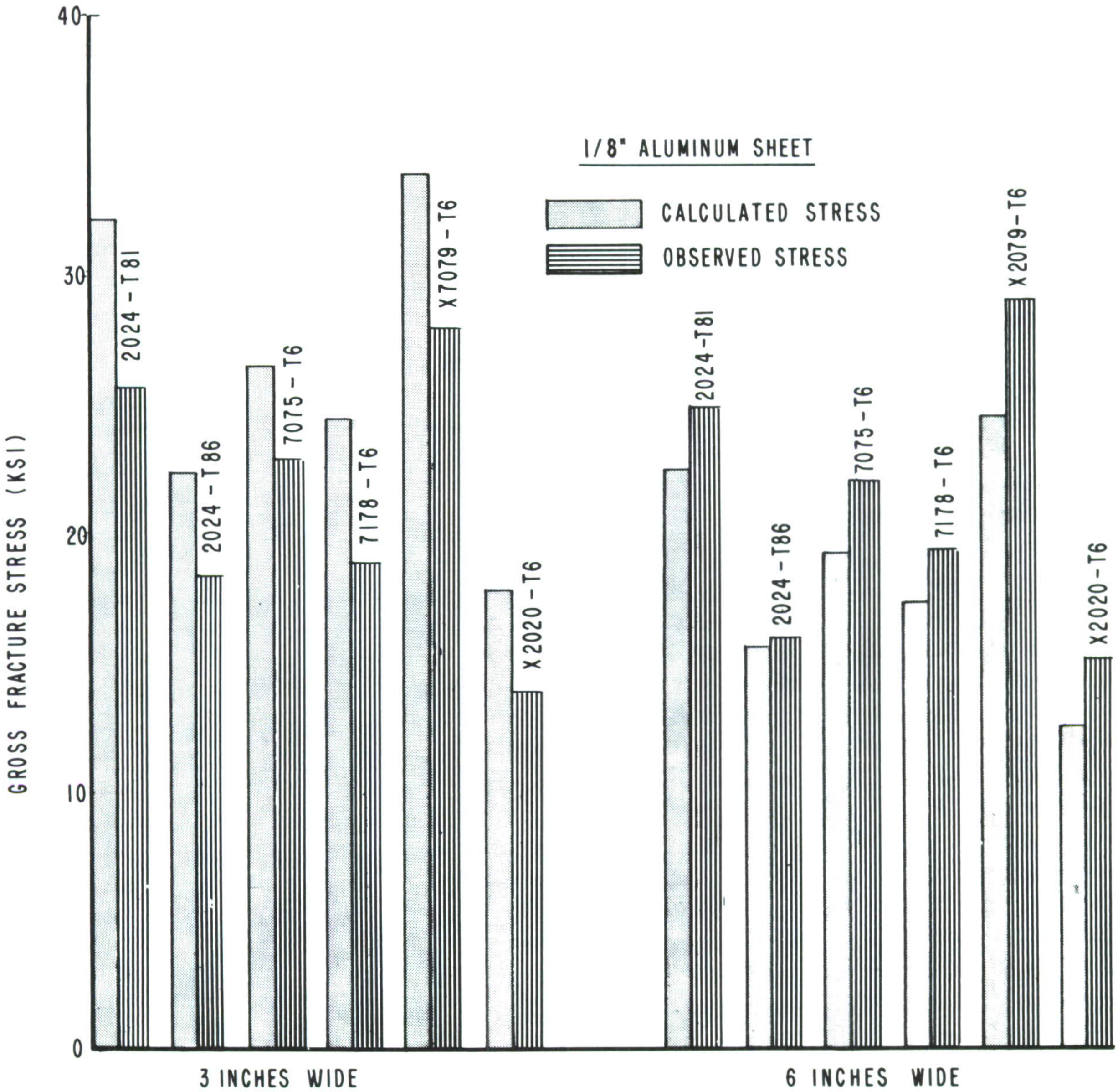

Figure 21. Comparison of Calculated and Observed Residual Strength Values for 3- and 6-Inch Wide Aluminum Alloy Tensile Specimens Containing Mechanically Sharpened Center Notches. The length of notch ($2a_o$) was approximately 1/3 the specimen width. (19)

SECTION VI

EVALUATION OF PRECRACKED CHARPY TESTS

1. Studies have been made of the effects of a number of test variables on fracture toughness of sheet as measured by the precracked Charpy test.

2. For the variation of fatigue precrack depth usually employed, the effect of initial crack depth on fracture toughness is not appreciable.

3. The effect of sheet thickness on fracture toughness depends on the toughness level. Starting with a thin size that exhibits a 100 percent shear fracture, the toughness measured as a function of increasing specimen thickness at constant test temperature may first increase to a maximum and then decrease. This initial rise in toughness is attributed to an increase in the volume of plastically deformed metal associated with the occurrence of lateral contraction; and the subsequent fall is attributed to an increase in the amount of the flat fracture at the center relative to the shear fracture at the surface. For a relatively low toughness level, only the fall in toughness associated with increasing flat fracture may occur; and at a very low toughness level (100% flat fracture), thickness has essentially no effect.

4. It appears that test environment can appreciably affect fracture toughness, with India ink being detrimental and dye penetrant being beneficial as compared to air.

5. Two methods have been utilized for measuring plane strain fracture toughness with the precracked Charpy specimen. In the specimen difference method, toughness is calculated by dividing the difference in fracture energy by the difference in cross-sectional area of specimens of two different thicknesses. In the brittle boundary method, carburizing is combined with austenitizing to accomplish case hardening and thereby avoid the shear mode of fracture at the surface. A combination of both methods yields the most accurate results.

6. Comparisons have been made between fracture toughness values as determined by center-notched tensile tests and by slow bend precracked Charpy tests (W/A); and between residual strength values as determined by actual measurements and by calculations based on use of the slow bend W/A values.

7. A relatively poor correlation was found between slow bend W/A values and G_c^* values as determined by conventionally center-notched tensile tests involving the ink staining technique. It appears that India ink can lower the stress at which slow crack growth occurs, with the result that the critical crack length becomes too large to satisfy the conditions for a valid G_c^* analysis.

8. A relatively good correlation was found between G_c and slow bend W/A values by using a compliance gauge to determine the effective critical crack length in the G_c tensile tests and thereby avoiding the use of ink staining.

9. Based on the use of slow bend W/A and initial crack length values in conjunction with the Griffith-Orowan equation, quite accurate predictions were made of residual strength values measured in center-notched tensile tests of four steels in a total of 12 heat-treated conditions and of 6 aluminum alloys in a total of 10 conditions (25).

10. Both a specimen difference method and a brittle-boundary method involving carburizing have been utilized to determine plane strain fracture toughness involving the precracked Charpy test. This is useful for test temperatures at which a combination of shear and flat fractures are normally encountered. The specimen difference method involves testing specimens of two different thicknesses, and subtracting out the contribution

of shear lips to fracture toughness. The brittle-boundary method involves combining liquid carburizing with the hardening treatment, which produces a brittle case and thus avoids formation of shear lips. For thin sheet specimens, a combination of the specimen difference and brittle-boundary methods appears to give the most accurate determination of plane strain fracture toughness (19). However, other test methods listed in Table II should be used rather than the precracked Charpy test in order to obtain more accurate results under plane strain test conditions.

TABLE II

TEST TECHNIQUES USED TO CALCULATE K_{Ic} (21)

Test Method	Conditions for Accurate Measurement of K_{Ic}	Method of Calculating* K_{Ic}
Tensile Test on Pre-cracked Round Specimen	$\sigma_N < 1.1\ F_{TY}$	$K_{Ic} = 0.414\ \sigma_N \sqrt{D}$ (formula applies for notch specimens where d/D 0.707) or Bruckner Analysis
Tensile Test on Center-Notched, Precracked Sheet Specimen	$\sigma_{ig} < 0.8\ F_{TY}$ $r_p < 1/2t$	$K_{Ic} = \sigma_{ig} \left[W \tan \frac{\pi a_o}{W} \right]^{1/2}$
Tensile Test on Edge-Notched, Precracked Sheet Specimen	$\sigma_{ig} < 0.8\ F_{TY}$ $r_p < 1/2t$	$K_{Ic} = \sigma_{ig} \left[W \tan \frac{\pi a_o}{W} + 0.1 \sin \frac{2\pi a_o}{W} \right]^{1/2}$
Tensile Test on Surface-Cracked Specimen	$\sigma_f < F_{TY}$	$K_{Ic}^2 = \dfrac{3.77\ \sigma^2 b}{\phi^2 - .212 \left(\frac{\sigma}{F_{TY}}\right)^2}$
Single-Notched Specimen	$\sigma_f < F_{TY}$	Experimental Calibration or Srawley Analysis
Bend Test Brittle-Boundary Test	$\sigma_f < F_{TY}$ $\sigma_f < F_{TY}$	Experimental Calibration or Kies or Wundt Analysis

* All K_{Ic} values are corrected for plasticity by adding a plastic zone size

$$r_p = \frac{K_{Ic}^2}{2\pi \sigma_y^2}$$

σ_N = net notch tensile strength

F_{TY} = 0.2% yield strength

σ_{ig} = gross stress at which slow crack growth is initiated

b = crack depth

W = specimen width

ϕ = elliptic integral

$2a_o$ = initial crack size

t = thickness

σ = gross failure stress

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13. ABSTRACT

This study summarizes the accumulated information and knowledge on precracked Charpy testing methods obtained as a result of an intensive literature survey. The precracked Charpy impact and slow bend test methods are defined and analyzed and their advantages, disadvantages, and limitations are critically appraised. The information will be especially useful to laboratories and engineers applying structural materials to aeronautical and aerospace structures.

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