Plane strain fracture toughness tests on two-inch-thick maraging steel plate at various strength levels

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Summary
Results of tests of 1·8-in-thick threepoint bend and compact (tension) specimens substantiate the current proposed ASTM recommended practice for plane strain fracture toughness (K_{Ic}) testing. Erroneously low results were obtained with 0·5-in-thick bend specimens when the ratio of K_{Ic} to yield strength exceeded 0·45 in$^{1/2}$ thus emphasizing the importance of requiring minimum specimen dimensions proportional to the square of this ratio. Shallow face grooves in 0·5-in-thick specimens had no useful effect. The maraging temperatures ranged from 700 to 1100° F, yield strength from 173 to 260 ksi, and K_{Ic} from 67 to 147 ksi-in$^{1/2}$. The K_{Ic} values were approximately proportional to the reciprocal of the square of yield strength, and also to a strain hardening index derived from the ratio of yield to ultimate tensile strength.

Introduction
The purpose of this paper is to discuss a systematic series of test results in relation to the recommended practice for plain strain fracture toughness measurement that has recently been proposed by Committee E-24 of the American Society for Testing and Materials (ASTM) [1]. The characterization of plane strain fracture toughness for engineering purposes in terms of the property K_{Ic} with dimensions (stress) $\times$ (length)$^{1/2}$was introduced by Irwin, Kies and Smith (1958) [2] as a rational development of the Griffith (1920) [3] concept of brittle fracture. Subsequent developments have been reviewedin some detail by Brown and Srawley [4]. It was found that K_{Ic}, like yield strength, was not always sharply indicated by a discontinuity in the test record, and it was necessary to define K_{Ic} in terms that were operationally precise, but somewhat arbitrary [5]. A similar situation exists with respect to engineering yield strength, for which there are several alternative operational definitions. These definitions are also somewhat arbitrary, but they are nevertheless useful.

It is inherent in the concept of K_{Ic}, and therefore unavoidable in the practice of K_{Ic} measurement, that the minimum necessary specimen dimensions increase in proportion to the square of the ratio of K_{Ic} to yield strength. This is simply a reflection of the fact that tough materials can tolerate larger cracks than more brittle ones, coupled with a linear elastic basis of test interpretation which admits only small plastic enclaves (as compared with specimen dimensions). In addition, there appears to be a strong inverse relation for materials in general between the ratio of yield strength to Young's modulus and the extent of the range of K_{Ic} toughness which can be encountered. Consequently, the size of specimen required for a K_{Ic} test of a comparatively tough material increases very rapidly as the yield strength level considered is reduced. The scope of the relevant published data is largely limited to materials that are either high in yield strength or low in K_{Ic} toughness, and much of the data was obtained with specimens one in thick or less. There is therefore a need for test data from larger specimens of tougher materials to further substantiate the current recommended test practice, or to find out where it should be modified.

The 18-Ni series of maraging steel compositions are particularly suitable for tests of large specimens covering a wide range of yield strengths. The strength (and toughness) of a given composition can be varied over a wide range by choice of maraging temperature and/or duration. The response to maraging is virtually uniform throughout very heavy sections, and distortion is negligible so that specimens can be fully machined prior to the marage treatment. In addition, the K_{Ic} toughness at a given yield strength is usually substantially higher than that of typical quenched and tempered steels.

Material and specimens

The material for the investigation was a single large plate of consumable-electrode-vacuum-melted steel of the following composition (weight percent): Ni 18·47, Co 7·40, Mo 4·81, Ti 0·40, Al 0·11, C 0·006, Si 0·01, Mn 0·06, S 0·007, P 0·006, Ca 0·05, Zr 0·010 B 0·002; nominally a 250-grade maraging steel composition. The plate had been rolled to nominal 2-in thickness from a press-forged slab of 36 by 7 in cross section, the rolling temperature starting at 2100°F and finishing at about 1750°F. After rolling the plate was solution annealed at 1500°F and air cooled. The Brinell hardness in this condition was 302 to 321.

Notched bend specimens as shown in Fig. 1 and standard 0·5-in diameter ASTM tension test specimens were machined from the solution annealed plate. The crack orientation in the bend specimens was *RW*, which means that the crack plane was normal to the rolling direction *R* and the crack front was normal to the plate width direction *W*. The tension specimens were longitudinally oriented to correspond with the bend specimens. Compact specimens for K_{Ic} tests (Fig. 1), of the same thickness as the primary bend specimens, and smaller bend specimens, 0·5 in thick, were machined from the pieces of primary bend specimens after these had been tested.

Specimens were aged for 6 hours at a series of temperatures (Table 1) from 700 to 1100 °F, controlled within plus or minus 5°F. It was considered that aging for 6 hours would give greater assurance of uniformity of heat

treatment than the usual 3 hours. The series provided a wide range of yield strengths for both underaged (less than 900 °F) and overaged (greater than 900 °F) conditions, as shown in Table 1. Additional specimens were aged for 24 hours at 800, 900 and 1000 °F to get an idea of the effect of extended aging in the vicinity of the normal 900 °F aging temperature.

After the specimens had been maraged they were fatigue cracked at 3600 cycles per minute. The maximum stress intensity during the final stage of crack growth beyond the chevron notch tips did not exceed 25 ksi-in$^{1/2}$, and the ratio of minimum to maximum stress intensity was about 0·1. The average rates of crack extension during the final stage were of the order of 0·001 in per kilocycle as required by the ASTM Proposed Recommended Practice. It should be mentioned, however, that measurement of rates of crack propagation by observation of the traces of the cracks on the surfaces of thick specimens is unlikely to be very accurate. In fact it is now generally agreed that the requirement regarding fatigue cracking should be stated differently. The fatigue crack surfaces were flat, smooth and without shear lips, and the crack fronts, while mostly slightly curved, were well within the ASTM proposed requirements for straightness and squareness.

Test procedure and data treatment

All tests were conducted at 70 to 75 °F in an airconditioned laboratory atmosphere with relative humidity about 50 percent or less. The published test procedure for bend specimens [1] includes detailed descriptions of the clip-in displacement gage and the bend test fixture. The only difference in procedure for testing the compact tension specimens was that they were loaded in tension through clevises of 280 ksi yield strength maraging steel with slightly loose-fitting pins of the same material. The relation used to calculate stress intensity factors for compact tension specimens is given in reference [4], Fig. 8, H/W equal to 0·6, and has since been independently confirmed by Wilson [6].

The test procedure is simple, but the calculation of K_{Ic} involves several steps which are necessary to ensure uniformity of interpretation of the test and consistency with linear elastic fracture mechanics [5]. During a test an autographic record is obtained of load (as ordinate) versus displacement across the crack at the specimen edge (sensed by the clip-in gage which fits in the knife edges shown in Fig. 1). The two examples shown in Fig. 2 are typical of the records obtained in the present series. To determine a provisional value of K_{Ic}, called K_Q, a secant is drawn with slope 5 percent less than the initial slope of the record.* This secant intercepts the record at load P_5 which corresponds to an effective crack length 2 percent greater

* For bend specimens with relative crack length a/W between 0·45 and 0·55. For compact tension specimens the equivalent secant slope is 4 percent less than the initial slope.

than the initial length. The actual crack extension at P_s is somewhat less than 2 percent because crack tip plastic deformation also contributes to the deviation of the test record from linearity. There is therefore a precautionary step to determine whether the actual crack extension at P_s is at least 1 percent. If not, the test result is invalid. The criterion is that the deviation of the test record from linearity at a load 0·8 P_s should not exceed one-fourth of that at P_s [5]. The load P_Q from which K_Q is calculated is either the same as P_s or is the maximum greater load which precedes P_s on the test record (the so-called pop-in load, if there is one). To determine whether K_Q is a valid K_{Ic} result, the corresponding value of the plastic enclave size factor is calculated: $F_Q = (K_Q/\sigma_{YS})^2$, where σ_{YS} is the 0·2 percent offset tensile yield strength. This size factor is a length, but should *not* be construed as an estimate of the *absolute* size of the plastic enclave. If the ratios of this size factor to initial crack length, and to thickness, F_Q/a_0 and F_Q/B, are both less than 0·4 the result is valid according to the proposed test practice. If not, it is necessary to test sufficiently larger specimens to satisfy these criteria.

Some of the tests provided an opportunity to obtain supplementary experimental data on the relation of crack extension to test record secant slope. These were tests of 1·8-in-thick specimens maraged at either 700 or 900 °F, or above 900 °F, in which the crack extension was gradual and controllable to well beyond 2 percent of the initial crack length (test record type I in Fig. 2). For the other marage temperatures the crack extension was not controllable (test record type II). Most of the controllable tests were interrupted by unloading at various estimated crack extensions so that the arrested crack fronts could be marked by fatigue cycling before the specimens were completely fractured. Fig. 3 shows the estimates of crack extension (as fractions of initial crack length) corrected for plastic strain [5], plotted against measured crack extension (averaged over the specimen thickness). The degree of correlation is considered adequate to confirm the suitability of the secant intercept procedure for standard interpretation of K_{Ic} test records. The accuracy of any method of monitoring crack extension is subject to the limitations imposed by the changing shape of the growing crack front.

The four results nearest the origin on Fig. 3 have plastic strain corrections which exceed the actual crack extensions. These are from tests of specimens maraged at 700 °F which were invalid because the deviation of the test record from linearity at the load 0·8 P_s exceeded one-fourth of the deviation at P_s. It happened in these cases that the ratios F_Q/a_0 and F_Q/B exceeded 0·4, so that the tests were invalid on these grounds also. But if the specimen size is marginal it can happen that F_Q does not exceed 0·4 a_0 or 0·4 B, *even* though P_Q is predominantly a result of plastic deformation rather than crack extension. This possibility could be avoided by

making the specimen size requirements even more severe than currently proposed. The precautionary step in procedure involving the load 0·8 P_s is considered a sufficient and more practical safeguard.

Results from 1·8-in-thick specimens

The individual results of all 1·8-in-thick bend and compact tension tests are listed in Table 2, and the corresponding tensile properties are given in Table 1. As already noted, the toughness test results for the 700 °F/6h treatment are not vaild K_{Ic} measurements according to the proposed recommended practice. There is no reason to suppose that K_{Ic} does not continue to increase as the marage temperature is decreased below 725 °F, and extrapolation of the valid data suggests that K_{Ic} would be about 180 ksi-in$^{1/2}$ for the 700 °F/6h treatment. This would require a crack length and specimen thickness of about 2·7 in for valid K_{Ic} measurement at the yield strength of 173 ksi. The compact tension test results (but not the bend test results) for the 1100 °F/6h condition are also invalid because F_Q/a_0 exceeds 0·4, although F_Q/B is slightly less than 0·4. This is so because the crack lengths in the compact specimens (obtained from the broken bend specimens) were only about 1·5 in whereas the thickness was 1·8 in. The width and crack length of a compact specimen are measured from the plane of loading, not the specimen edge which is arbitrary (Fig. 1).

The proposed recommended practice includes a table of suggested thickness and crack length related to the ratio of yield strength to Young's modulus of the material to be tested. This was provided because it may be necessary to decide on specimen dimensions for materials for which there is no relevant data other than tensile properties. It is based on a variety of K_{Ic} data, from which was inferred a probable upper bound to F_{Ic} (that is, $(K_{Ic}/\sigma_{YS})^2$) which decreases as the ratio of yield strength to modulus increases. These suggested specimen dimensions are usually adequate, and would have been adequate for the present material in the marage range 800 to 1000 °F, but barely sufficient for the tougher conditions.

Comparison between bend and compact tension results

The average K_{Ic} values for bend and compact tension tests are compared in Fig. 4. The agreement is generally good, but there is a 10 percent difference for the 725 °F/6h treatment, and a 5 percent difference for the 750 °F/6h treatment. The good agreement supports the view that linear elastic fracture mechanics is a satisfactory working hypothesis when the ratio of the plastic enclave size factor F_Q to crack length a_0 is sufficiently small. The 10 percent discrepancy when the proposed upper limit of 0·4 for F_Q/a_0 is approached indicates that this requirement is not unduly restrictive. Greater allowed values of F_Q/a_0 could result in still greater discrepancies, not only between different specimen types, but also in application of test results to fracture analysis of structural components. For sufficiently small values of F_Q/a_0 it is assumed that crack extension is governed by the idealized singular elastic stress field of the crack tip, which is independent of specimen form. As F_Q/a_0 increases, the governing elastic stress field becomes more remote from the crack tip, and increasingly influenced by specimen (or component) form and load distribution. In turn, crack extension behavior is increasingly affected by these individual circumstances.

Results from small bend specimens

For each maraged condition nine small bend specimens were obtained from one of the pieces of a large bend specimen. These small bend specimens were proportioned like the large specimens but had thickness and nominal crack length equal to 0·5 in. Three out of each set of nine specimens were from the center of the thickness of the large specimen, and six from immediately adjacent to the surface. Of the six surface specimens four had vee grooves machined in the faces along the plane of extension of the crack. The groove depths were such that the ratio of net to gross thickness B_N/B_G was 0·95 for two specimens and 0·90 for the other two. The flank angle of the vee grooves was 90° and the root radius 0·01 in.

The test results for specimens without grooves are shown in Fig. 5. Each plotted point represents the average of results for three center specimens or two surface specimens. There is very good agreement between surface and center specimens, indicating that the material was uniform through the thickness of the plate.

The ordinate in Fig. 5 is K_Q, the provisional value of K_{Ic} in the proposed recommended practice, and the points for which F_Q is greater than 0·4 B (and 0·4 a_0) are clearly distinguished. All of the other points except those for the 1050 °F/6h marage are valid K_{Ic} results. The exceptions failed to satisfy the check procedure for at least 1 percent crack extension at the load P_s, though they did satisfy the size requirements. They are a good example of the need for this check procedure since the K_Q values are about 20 percent lower than the K_{Ic} value from the thick bend specimens.

All the valid K_{Ic} values from the small bend specimens agree quite well with the results from the large specimens, but the invalid results from small specimens are seriously discrepant. Indeed, if the small bend test results were to be taken at face value it would appear that there was not much variation of plane strain fracture toughness over the range of marage temperature from 700 to 1100 °F, and that the maximum toughness resulted from maraging in the range 750 to 800 °F. Both these conclusions would be highly misleading in the light of the results from the 1;8 in thick specimens. It

is true that specimens that are too small usually (but not invariably) give results that are lower than the K_{Ic} values of the materials, and thus might seem to err on the side of safety. But this is not necessarily true because comparison of invalid K_Q results for alternative materials could lead to a false conclusion about which was toughest.

Face-grooved specimens

The face grooved specimens were tested because it is sometimes claimed that such grooves will reduce the thickness necessary for a valid K_{Ic} result. The ratio F_Q/B is associated with the three-dimensional shape of the plastic enclave around the crack front, which can be visualized as a cylinder with flared ends. In the cylindrical region the state of stress is plane strain, but the flared ends are transition regions where the state of stress changes progressively to plane stress on the specimen faces. If F_Q/B is sufficiently small the end regions of the plastic enclave will be small compared with the specimen thickness and their effect will be negligible. If F_Q/B is too large, however, the end regions will be dominant, and the state of stress will be nonuniform along most of crack front. The idea of face grooves is to introduce constraint to plastic flow at the ends of the crack front, but apparently it is not always sufficiently appreciated that the grooves do not eliminate the plane stress transition regions, but merely change their shape and complexity of stress field. One kind of end effect is replaced by another which is not necessarily weaker.

While visualization of the physical significance of F_Q/B is helpful in appreciating the importance of setting a useful limit on it, present theoretical knowledge is not sufficient to set the limit more closely than within the range 0·2 to 1. The current limit of 0·4 for specimens without face grooves is a practical compromise based on available experimental data. Face grooved specimens are more complex, and a thorough experimental exploration would involve the groove variables (relative depth, sharpness and flank angle) as well as specimen thickness and material properties. The present results are only a limited contribution, restricted to shallow, moderately sharp grooves.

Average results from duplicate tests of the face-grooved specimens are shown in Fig. 6. The ordinate in this figure is $K_Q(B_G/B_N)^{1/2}$, where K_Q is calculated as though there were no grooves (according to the expression in the proposed recommended practice) and B_G/B_N is the ratio of gross to net thickness. This ordinate expression is due to Irwin *et al.* [7] and is based on the assumption that the crack extension force (proportional to K^2) rather than the stress intensity factor K, should be averaged over the net thickness. The results in Fig. 6 are consistent for the two depths of face grooves, and agree, in general, with those in Fig. 5 for specimens without face grooves. The conclusion from these limited results is thus that specimens with *shallow* face grooves can be expected to give satisfactory results when they meet the requirements of the recommended practice in all other respects. The face grooves, however, do not reduce the gross thickness necessary for a satisfactory result. Essentially the same conclusion was reached by Freed and Krafft [8].

Relation of K_{Ic} toughness to tensile properties

It is obvious from Fig. 4 that there is an inverse relation between K_{Ic} and yield strength (or, alternatively, tensile strength). Such an inverse relation is usually found when these properties are varied in a material of a given composition by changing some parameter such as heat treatment, test temperature or test speed. For example, Krafft and Sullivan [9] found that for mild steel K_{Ic} could be correlated empirically with the upper yield point strength to the power $-1{\cdot}5$ when the test speed and temperature were varied for a given composition. A similar empirical correlation for the present results is shown in the upper part of Fig. 7 where the product of K_{Ic} with the square of yield strength is plotted against marage temperature. The K_{Ic} values used in this plot are the combined averages for 1·8-in-thick bend and compact tension tests; the individual averages show essentially the same pattern.

There is no fundamental basis for this particular form of correlation, and the variation in the plot of $K_{Ic}\,\sigma_{YS}^2$ versus marage temperature is almost certainly not random. The heat treatments involve at least two distinct precipitation processes and, at temperatures above 900 °F there is also some reversion of the martensitic matrix to austenite [10, 11]. There appears to be some embrittlement associated with the range of marage temperature between about 800 and 900 °F, indicated by the results for 850 °F/6h and 800 °F/24h. treatments compared with 900 °F6/h and 900 °F/24h. Similar evidence of embrittlement in this range of marage temperature has been found in other heats of 18-Ni marage steel, and it appears that the range 900 to 900 °F is preferable for maraging to maximum strength levels.

Substantial improvement in K_{Ic} toughness can be obtained at the expense of somewhat lower yield and tensile strengths by maraging either below 800 °F or above 1050 °F. There are many applications in which this greater margin of safety from fracture should be exploited. Comparison of the present 725 °F/6h results with those for 1100 °F/6h, and 750 °F/6h shows that for equal yield strength the K_{Ic} toughness is appreciably better for the lower temperature treatments. There is some doubt as to whether this is true of 18-Ni maraging steels in general, and the choice of lower or higher marage temperature might well be decided by other factors such as the circumstances of fabrication and service.

The relation of toughness to strength is of direct engineering interest, but in seeking more fundamental understanding in terms of critical strain Krafft [12, 13] and Hahn and Rosenfield [14] have suggested reasons why K_{Ic} should be proportional to the so-called strain hardening exponent n. One difficulty with this idea is that most materials do not conform to the implied linear relation between flow stress and plastic strain to some constant power n. The slope of the plot of the logarithm of flow stress versus the logarithm of plastic strain is usually not constant, but varies with increasing strain. If the concept of a strain hardening exponent is emphasized (rather than some other index of strain hardening) then n should be regarded as a function of plastic strain. In this case one has to choose a point value of n, or an average over a restricted range of plastic strain, to correlate with K_{Ic}. There is one choice which is particularly convenient because it requires only a knowledge of the engineering yield and tensile strengths, σ_{YS} and σ_U. For the 0·2 percent offset yield strength the function m of the yield ratio is given by $(0{\cdot}002\ e/m)^m = \sigma_{YS}/\sigma_U$, where $e = 2{\cdot}718\ \ldots$ is the base of natural logarithms [15]. For a material which strain hardens according to the constant power relation, m is identical to the constant exponent n. For other materials it is an average measure of strain hardening capability over the range of plastic strain from 0·002 to the tensile instability strain. Obviously, a similar quantity can be defined for any particular yield strain.

The lower part of Fig. 7 shows that the ratio K_{Ic}/m is almost as insensitive to marage temperature as the product $K_{Ic}\,\sigma_{YS}^2$. Since the results represent a series of different, but closely related materials, the variation of K_{Ic}/m could be interpreted as reflecting the variations in what Krafft [12, 13] calls the 'process zone size' with the marage treatment. This is quite plausible since the size and spatial distributions of precipitate particles, and the proportion of austenite in the matrix will certainly vary with marage treatment.

Krafft's hypothesis is relatively simple and involves only the strain hardening exponent and the Young's modulus (which can be assumed invariant in the present series of results). Hahn and Rosenfield [14] suggest that K_{Ic} should vary also in proportion to the square roots of the yield strength and the tensile fracture strain. Although these two factors tend to offset one another, the net effect of including them is to emphasize somewhat the variation in the present results rather than to decrease it. This does not necessarily mean that Krafft's simpler hypothesis is better, but it does suggest that Hahn and Rosenfield have included either too many factors or too few. The merits of the respective hypotheses can only be properly judged on the basis of results for a wide variety of materials.

Several alternative indexes of strain hardening were investigated. Many fitting functions for stress-strain curves have been proposed, and the three-parameter function of Palm [16] and Voce [17], for example, usually provides an excellent fit. However, these functions have no particular interpretive advantage for the present purpose over direct methods of treating the curves of true stress versus plastic strain. The normalized rate of strain hardening and the logarithmic rate of strain hardening are easily obtained as functions of plastic strain. The normalized rate is the derivative of the natural logarithm of true stress with respect to strain, and the logarithmic rate is the derivative of the logarithm of stress with respect to the logarithm of strain (the slope of a log-log plot). The logarithmic rate is, of course, the generalization of the constant strain hardening exponent in the simple power function. An alternative generalization of the strain hardening exponent has been proposed by Rhee and McClintock [18] and also by Halford [19]. This index is obtained for a given plastic strain by taking the ratio of the corresponding stress to the average stress over the stress-strain curve up to that point, then subtracting unity. This procedure involves integration rather than differentiation of the stress-strain curve.

For the present maraging steels these several indexes were all found to be decreasing functions of plastic strain. Various particular values and averages of the indexes were tried, but none correlated any better with K_{Ic} than the function m of the yield ratio.

Conclusions

The agreement between K_{Ic} plane strain fracture toughness test results from 1·8-in-thick bend and compact tension specimens is consistent with the view that K_{Ic} is a material property which can be used for strength calculations involving cracks. The operational definition of K_{Ic} embodied in the current proposed ASTM recommended practice is further substantiated by the present results. The importance of the specified dimensional requirements for K_{Ic} test specimens is emphasized by the erroneous and misleading results obtained with 0·5-in-thick bend specimens for yield strengths less than 250 ksi (for the particular plate of 18-Ni maraging steel used). Shallow face grooves in 0·5-in-thick bend specimens had no useful effect.

Over the range of marage temperature from 725 to 1100 °F, K_{Ic} was approximately proportional to the reciprocal of the square of yield strength (which ranged from 180 to 260 ksi), and also to an index of strain hardening based on the ratio of yield to ultimate tensile strength. There were fluctuations from proportionality which probably reflect the complexity of the microstructural changes which occur over the wide range of maraging temperature investigated.

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Table 1.
Tensile properties. Average of two tests of standard 0·500-in-diameter specimens

Marage Temperature, °F	Marage Time, hrs	Tensile strength, ksi	Yield strength, (0·2 % offset) ksi	Total elongation %	Reduction of area, %	Rockwell hardness number R_c
700	6	192	173	17	62	38·7-40·6
725	6	206	190	16	58	43·8-44·5
750	6	218	203	14	57	45·9-46·3
775	6	227	213	14	55	46·7-48·0
800	6	238	227	13	54	48·4-48·9
850	6	263	253	12	46	51·3-51·9
900	6	266	259	11	51	51·0-52·1
950	6	261	252	12	49	50·8-51·2
1000	6	242	232	14	47	49·3-50·3
1050	6	218	204	15	50	45·6-45·8
1100	6	198	180	17	60	43·7-44·0
800	24	271	260	10	42	51·1-52·3
900	24	268	259	11	49	51·9-52·0
1000	24	225	209	14	44	47·0-47·2

Table 2
Plane strain fracture toughness test results from 1·8-in-thick bend and compact tension specimens

Marage Temperature, F	Marage Time, hrs	Bend K_{Ic} or K_Q( )*, ksi-in$^{1/2}$	Bend	Bend	Compact tension	Compact tension	Compact tension
700	6	(150)	(147)	(146)	(153)	(150)	(148)
725	6	158	158	147	147	138	136
750	6	143	136	134	133	127	126
775	6	124	116	111	121	112	110
800	6	102	96	93	97	93	90
850	6	76	73	—	78	77	76
900	6	88	84	81	81	81	79
950	6	89	84	82	84	83	82
1000	6	88	87	83	88	88	87
1050	6	108	105	103	107	106	105
1100	6	148	145	—	(147)	(136)	(126)
800	24	71	66	64	70	69	59
900	24	82	80	79	82	81	80
1000	24	99	95	94	98	95	95

* K_Q values in parentheses are not valid K_{Ic} values, as explained in text.

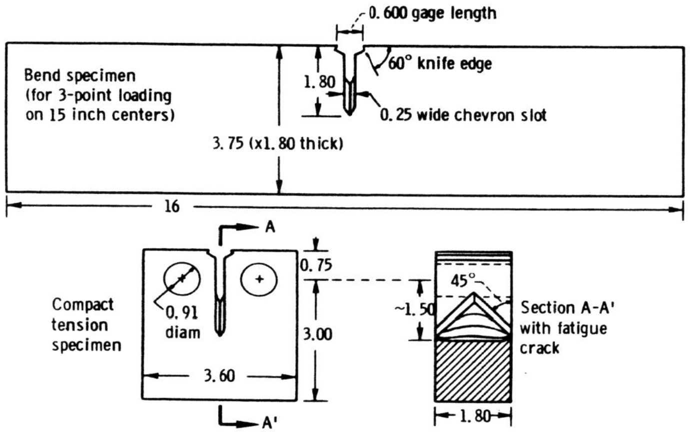

Fig. 1. Plane strain fracture toughness specimens with dimensions in in.

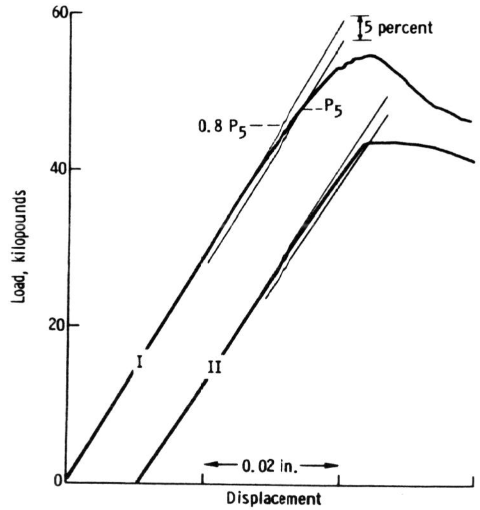

Fig. 2. Typical load-displacement records from 1·8-in-thick specimens.

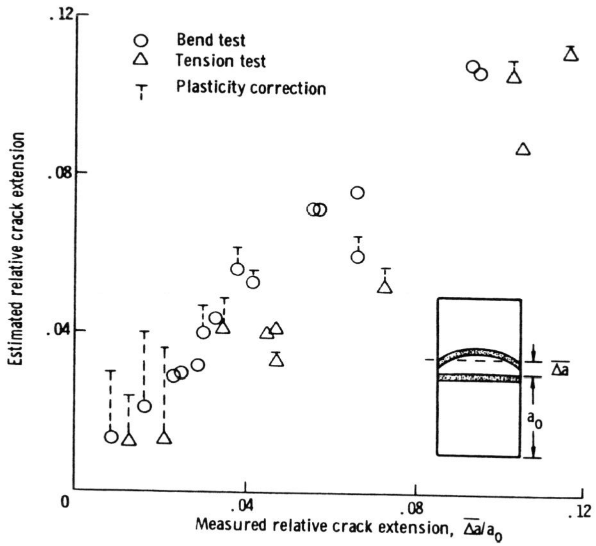

Fig. 3. Results of interrupted tests to compare estimates of crack extension from test records with measured crack extensions marked by fatigue cycling.

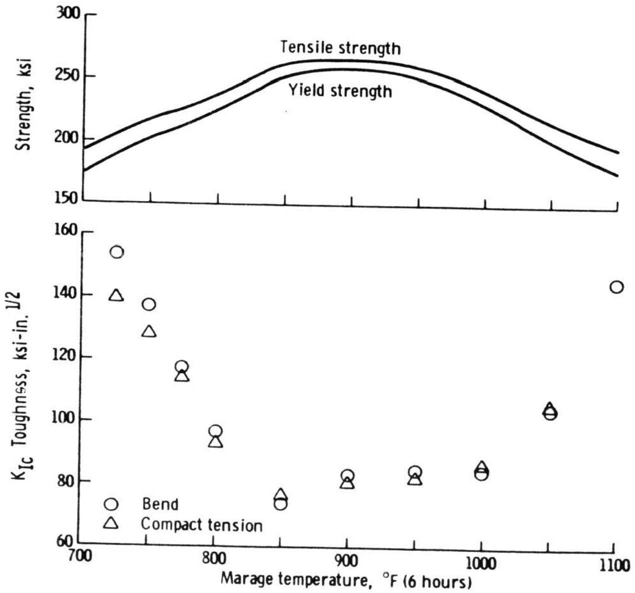

Fig. 4. Average K_{Ic} toughness and tensile properties of specimens maraged 6 hours.

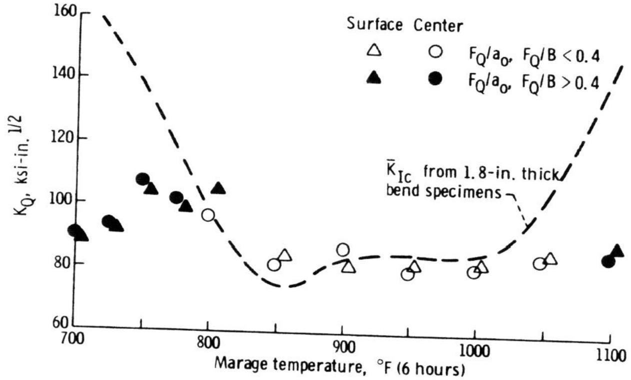

Fig. 5. Average results from small bend specimens without face grooves (obtained from surface or center of thickness of large bend specimens).

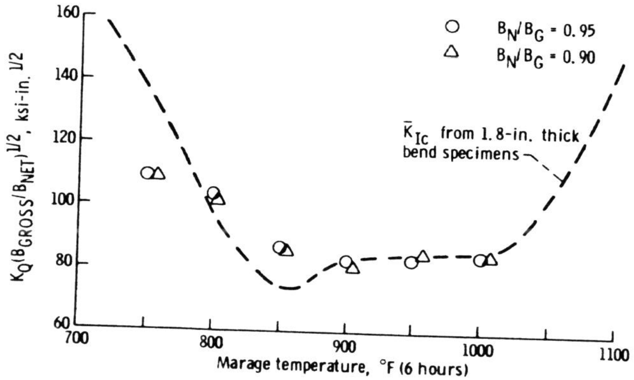

Fig. 6. Average results from small bend specimens with shallow face grooves.

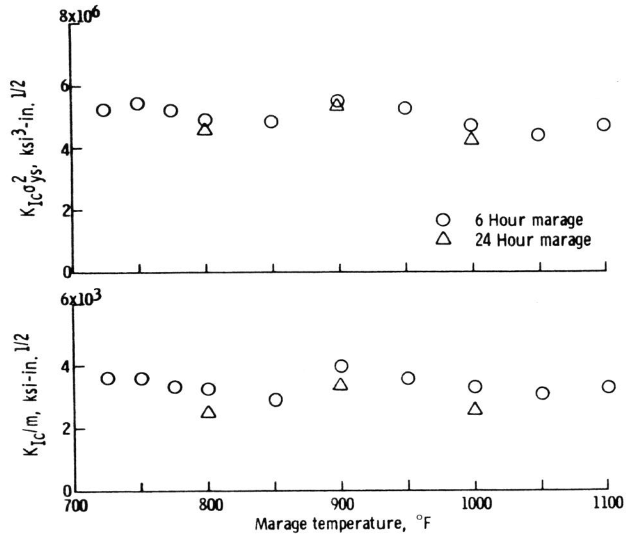

Fig. 7. Correlation of K_{Ic} with reciprocal of square of yield strength (top), and with strain hardening index m calculated from yield ratio (bottom).