Fracture toughness measurements on low strength structural steels

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Summary
Instrumented impact and double cantilever beam (DCB) tests were carried out on a low strength structural steel over a range of temperatures. In the DCB test, cracks arrested and re-initiated several times along the length of each piece. Initiation values of K_{1C} were compared favourably with previous results from slow bend tests, and arrest values (dynamic fracture toughness K_{1D}) with those deduced from instrumented impact tests on fatigue cracked and side grooved pieces. The impact test appeared to give results valid up to higher temperatures than the DCB test. These results lend support to the view that fracture toughness of mild steel is rate sensitive and give confirmation to an earlier suggestion that minimum dynamic fracture toughness values can be deduced from suitable instrumented impact tests. Impact K_{1D} values for several steels are shown as a function of temperature over the general range from $-196°$ to $+20°C$.

Introduction

Linear fracture mechanics was stimulated largely by fracture problems in materials of high strength/weight ratio. This work has been reviewed many times (for example Refs. 1 and 2) and needs no further discussion here. Brittle fracture of mild steel, met spasmodically since the late 19th century, became a major problem in the 1940's following a number of fractures in bridges and ships, but these two areas developed virtually independently up to and during the 1950's. Although Wells [3] early saw the relevance of fracture mechanics to the propagating crack, the link between brittle fracture of steel and linear fracture mechanics was not pin-pointed until in 1965, Eftis and Krafft [4] combined an analysis of the Illinois wide plate running crack tests with the data inferred from Krafft's co-relation between fracture toughness K_{1C} and the strain hardening exponent n. This presented K_{1C} for low strength steel* as a function of strain rate in a manner which, although not proven, offers a satisfying explanation of the diverse observations of the brittle fracture of steel at the macro-level. There now seems to be a general acceptance of toughness as a rate dependent property [5], the dependence being a function of temperature as shown schematically in Fig. 1. Thus valid values of K_{1C} for static macro initiation of a

* The different meanings attached to 'low' and 'high' strength in the structural steel and fracture mechanics fields should be noted. In the former, 30 ton/in² yield may be called high strength, in the latter 100 or 150 ton/in² yield is implied.

crack require low temperatures of large test pieces; valid propagation values can be inferred for a given size of test piece at temperatures well above that at which static initiation passes to above gross yield, typically up to ambient temperatures for common mild steels. Evidence consistent with this interpretation has been presented in several recent papers [6, 7, 8]. Recent design philosophy is moving correspondingly [9].

In a statically indeterminate situation, where unloading of the crack is possible (such as an unstress-relieved weldment perhaps situated in a redundant structural member), a crack initiating in a poor HAZ region could be arrested, or not, depending upon the crack length – K_{1D} relationship for the crack as it runs out into the parent plate. If arrest of a propagating crack is not feasible, because of the follow up nature of the load, then safety of the structure depends upon the avoidance of crack initiation on the macro scale. For many steels the initiation value (for undamaged material) is so high that, at least for sections less than some two or three or more ins thick a valid K_{1C} cannot be determined, except at very low temperatures, yielding fracture mechanics with its concept of a crack opening displacement [10, 11] being employed instead. This high value is however easily reduced by metallurgical damage of the type associated with welding. The complexities of these problems in real design circumstances [9] are beyond the scope of this paper, which seeks only to provide further evidence on the initiation K_{1C} for low strength steels and the usefulness of an instrumented impact test for determining minimum K_{1D} values up to about the 50% fracture appearance transition temperature (FATT). Since there is some doubt whether the dynamic values of K_{1C} apply to propagation at a particular crack speed, to arrest, or to dynamic initiation the symbol K_{1D} (dynamic) is used for any fracture toughness determination at above conventional static rates of testing or relating to the running crack*. If the dynamic value is thought to approximate to the minimum 'trough' value, Fig. 1, the symbol K_{1T} (trough) is used. It may be remarked in passing that there seems little direct evidence in the extensive literature on conventional fracture mechanics that for the high strength materials static initiation is more critical than high strain rate initiation or propagation, although this is implied by the concentration of effort on static testing for these materials.

Test techniques

The instrumented impact test used has been described in detail, Ref. 12. In brief, semi-conductor strain gauges are affixed near the nose of a standard Charpy impact hammer, with a photocell on the frame, so placed that

* Where a generic symbol for toughness is required, be it static or dynamic, K_{1D} is still used.

the swinging pendulum cuts an incident light beam. Load on an displacement of the pendulum can thus be recorded on a CRO against each other, or both independently against time. A standard V notch Charpy piece, with a fatigue crack, some 0·010 in deep beyond the machined notch, and side grooves 0·020 in deep has been used. The load recorded is corrected for the gross effect of test piece inertia, [16]. This correction is only an approximation allowing for the acceleration of the test piece as a deflecting beam, ignoring vibrations. The corrected load is then inserted in the formula [13] for a centrally loaded beam to give K_{1D}.

The second test used is the double cantilever beam test (DCB.) generally as described by Hoagland [14] and Ripling [15]. The energy release rate of this type of piece can be varied by altering its shape to give an increasing, constant or decreasing characteristic with crack length, as shown by boundary collocation calculations [16] and tests [17]. In vesting mild steel in the brittle regime, the overriding characteristic is the marked effect of crack speed on toughness, so that a stable crack is not obtained even with a decreasing characteristic of test piece. The form and dimensions, Fig. 2, are a consequence of three factors; the maximum size of material available, 4 in × 18 in, cut from the 1 in thick plate already tested Ref. 8; the proportions most conducive of plain strain (w large and consequently w/b large requiring h large and L small to avoid fracturing the arms) and the proportions conducive of multiple crack arrests, so that several toughness values can be determined and averaged from each piece. Preliminary tests showed these latter requirements to be w and hence w/b small, h small and L large. At higher test temperatures (but still well below the FATT) the divergence between K_{1C} and K_{1D} gives long cracks between each arrest (e.g. two–three in per jump) so that, a useable middle portion of some ten-twelve in is desirable. The high values of K_{1C} at these temperatures also promote yielding of the arms so that larger values of h are beneficial. It is not claimed that the dimensions used here are a general optimum, particularly in view of the size of material available, but, preliminary tests on several other sizes suffered from the above drawbacks. In the standard piece a slit is cut by sawing to 1 in beyond the line of loading pins and side grooves cut with a standard Charpy cutter, giving a 0·010 in root radius, subsequently scored with a sharp pointed tool producing a final root radius of about 0·001 in. The slit was sharpened by a reduced thickness blade, and then fatigued, using a load to produce ¼ in crack extension in from 60-100 thousand cycles. Despite the prior side grooving, square fronted fatigue crack profiles were usually obtained.

The pieces were tested in an Instron testing machine, using a conventional coolant bath. Crosshead speeds were typically 0·005 in/min, although in some tests this speed was increased to 2 in/min.

Interpretation of test results

Typical diagrams from the instrumented impact tests are shown in Fig. 3 (a, b). Although the final fracture corresponding to a record such as Fig. 3 (a) is by cleavage, and corresponds to say 60% crystalline appearance (and hence judged brittle by an appearance transition criterion) the fracture occurs well beyond gross yield, and is thus not used for the calculation of fracture toughness described here. Only diagrams showing fracture on a rising load curve, such as Fig. 3 (b), are used. If these terminate before gross yield (as substantiated by observations of hardness, contraction and slip line patterns, [8] the maximum load, corrected for inertia [12] is used for the calculation of K_{1D} . If gross yielding has occurred, a dynamic COD is inferred from the displacements. Values so calculated [18] are likely to be less accurate than from a COD meter, but there is some evidence [19] that the differences are small.

In the DCB test, diagrams of the form Fig. (a, b, c) are obtained, each step corresponding to a cycle of macro crack initiation, propagation and arrest. With testing conditions where yielding of the arms does not occur, the steps can be extrapolated back to a common origin (dotted lines Fig. 4 (a, c)). The slopes of such lines are used for calculating the test piece compliance for the particular crack length. Since the extension recorded is that of the machine crosshead, the diagram is first corrected for the extension of the machine and shackles. A standard specimen fitted with a dial gauge to measure its own extension, showed the correction to be linear and repeatable at 0·0035 in per 1,000 lb load. From each step of each test record the compliance ϕ is calculated from the corrected extension Y_C and the load F_i at initiation of a jump in the crack or F_a at the arrest. From each test piece, the thickness w of the actual fracture surface is measured. The length, a, of the crack is determined in two ways. On a typical fracture surface, Fig. 5, a narrow band of differing texture is observed between each propagation region, and the length to this band is measured as discussed later. The second determination of crack length is by use of a calibration specimen, the compliance of which was measured for successively longer sawn slits. This calibration includes the effects of the side grooves and is thus preferred to the calculated compliances [16]. The calibration curve so obtained, is shown Fig. 6. The value of compliance ϕ derived for the test piece record, as above, allows a value of crack length, a, and rate of change $d\phi/da$, to be read off the calibration curve. The fracture toughness is calculated from the formulae for energy release rate G,

$$G = \frac{F^2}{2w}\frac{d\phi}{da}$$

$$\text{and } K^2 = EG/(1-\nu^2)$$

using values appropriate to initiation or arrest as required.

This method gives results in good agreement from both determinations of crack length for fractures with no yielding of the arms. In determining the slopes, $d\phi/da$, tangents can be drawn to Fig. 6 in the conventional way, or a plot made of log ϕ – log a. For the standard test piece size used here it was found, for mild steel,

$$\phi = 0{\cdot}201 \times 10^{-5}\, a^{2\cdot 67}$$

for $2{\cdot}5 \leqslant a \leqslant 12{\cdot}5$ ins, with the load expressed in pounds. At the higher test temperatures or with increased values of w, as in the preliminary tests, yielding of the arms precludes the direct measurement of compliance, ϕ, from each step of the test record. The crack length, a, is then measured only from the fracture surface and the calibration curve used to infer the effective compliance, and hence $d\phi/da$, in the absence of yielding. For this case of yielding in the arms Hoagland [14] obtained the compliance by unloading the test piece just prior to fracture, and correcting the test record accordingly. It is, of course, difficult to know just when to unload the test piece, but a few attempts to use that method gave results consistent with the method used here.

Sample values K_{1C} and K_{1D} are shown Fig. 7 against crack length, for no yielding in the arms. Tests at –196 °C gave as many as 27 crack jumps. At –100°C, 7 jumps was more typical. In the following discussion a quoted value of K_{1C} or K_{1D} from a *single* DCB test is the mean of the values for each jump in that particular test. The first (shortest crack) and last (longest crack) values to be high because of end effects, so that these first and last values (some 20 and 10% higher) should perhaps be omitted from the averaging.

Results of DCB and instrumented impact tests

The major series of tests after satisfactory development of the DCB testing technique, was on a relatively tough notch type steel (50% Charpy FATT = –20 °C), low strength in terms of linear mechanics but medium to high strength in terms of structural steels. This material, Table 1, had been used for previous tests [8, 20] so that K_{1C} from large slow bend tests and K_{1D} from instrumented impact and running crack tests were known. Static and Dynamic COD values were also available [8]. Because of doubts over the interpretations of an impact test with dynamic initiation as reasonable estimates of minimum K_{1T} values, interest centred on comparison of K_{1D} arrest values from the DCB test with the instrumented impact results. It was accepted intuitively that a valid arrest K_{1D} value would be a good approximation of the minimum K_{1T}, although Hoagland [14] has expressed an opinion to the contrary. Unfortunately the material was already cut to the size of the beams used in Ref. 8, so that the direction of cracking in the DCB pieces was perforce parallel to the long edge of the piece (parallel to the rolling direction) whereas the beam pieces of Ref. 8 suffered cracks across the beam width. Pieces cut parallel to the rolling direction, with crack plane perpendicular to the rolling direction, had been used for all previous impact tests on this material and for small slow bend tests, which, because of the small size, gave valid K_{1C} results only below $-160\,°C$. A second series of impact and small slow bend tests was therefore conducted with the cracking parallel to the rolling direction. Material was not available for a repeat of the large slow bend tests with crack parallel to the rolling direction. The original and present results are plotted against temperature Fig. 8. Taking the two sets of impact and small slow bend results as a guide, there seems to be a fairly small degree of anisotropy, the toughness with crack parallel to rolling being about the same as in the perpendicular direction at low temperature but falling some 20% below, according to the impact results, at higher temperatures.

For initiation, the original large slow bend results are shown together with an estimated curve for cracking in the rolling direction based on the degree of anisotropy described above. These initiation values in both slow bend and DCB tests ceased to be valid because of excessive yielding at the crack tip above $-80\,°C$, for this material. The DCB K_{1C} values are marginally below the original slow bend results and rather above the estimated curve for cracking parallel to rolling. The DCB K_{1D} values fall between the two sets of impact results. Because of the yielding associated with initiation, it is difficult to assure validity of DCB arrest measurements above perhaps $-50\,°C$ whereas the impact K_{1D} values are judged adequate up to about $-10\,°C$ for this material.

In a further series of DCB tests a cross head speed of 2 in/min (i.e. 400 times faster) was used. The results are shown in Fig. 9, falling between the K_{1C} values at the slower rate of testing and the K_{1D} values from the impact test. At the higher test temperatures (above about $-80\,°C$) the 2 in/min test results do not follow this pattern but coincide with the 0·005 in/min results. This coincidence is probably fortuitous. Yielding occurred at the crack tip at these higher temperatures, and gross yielding of the arms reduced the effective crack tip strain rate for the given crosshead speed. Further development of test piece geometry may be necessary here and the present results at 2 in/min are not judged valid above about $-80\,°C$.

At low test temperatures, the band between successive crack jumps, Fig. 5, is so narrow that no ambiguity on crack length arises. At higher temperatures, the band broadens and the validity of measuring to the near edge, for arrest, and indeed the significance of the zone itself, is debatable. Ultrasonic measurements after arrest, not described here*, show that

* The authors are grateful to A. Pollak, Physics Dept., Imperial College, for his work on this problem.

crack lengths so determined agree to $\pm\frac{1}{16}$ in with the length to the near edge of the zone, but the technique has not been developed by the present users to the state where it can be used in regular tests in low temperature baths. Agreement with the crack lengths is also found between the compliance of the saw cut calibration piece described above and direct measurements on actual test pieces, if the length is measured to the near edge of the zone, rather than to its far edge, for arrest. The zone is taken to be a small region of slower speed crack initiation, still cleavage at the lower temperatures but of mixed micro mode at higher temperatures. Micro fractographic studies have not been made. Further examination of the implication of the zone on measured initiation values of K_{1C}, the possible effects of local yielding, crack blunting or residual stresses are also required.

These points should be resolved to assist understanding of the DCB test and crack propagation in general, but for practical purposes the agreement found here is taken to confirm the impression already gained from other test techniques (summarized Ref. 12) namely that the instrumented impact test provides a simple means of determining K_{1T} for low strength steels.

Although not directly relevant to the present paper, it was thought desirable to support the general interpretation of high initiation K_{1C}, crack jumping in the DCB test, low K_{1D} and the corresponding well known increase of severity of the impact test over a slow bend test, for mild steel, by contrasting tests on a material not highly strain rate sensitive. These tests, on a four per cent copper aluminium alloy, will be described elsewhere when completed. Suffice it to say here that DCB tests over a range of temperatures gave load records showing a ripple, rather than the steps of Figs. 4 (a),(b),(c). On the fracture surfaces, no jumps were apparent. The K_{1D} values from the instrumented impact test are somewhat *greater* than K_{1C} values from the slow bend tests. This complete reversal of trends in *both* types of tests is taken as general corroboration of the links between the two tests on which the present interpretation depends.

Dynamic fracture toughness values for several structural steels

On the evidence of crack arrest results (Ref. 7) running cracks (Ref. 20) and the DCB tests described above, the impact K_{1D} values for mild steel, are taken to be approximations to the minimum K_{1T} value. This statement may not be supportable for other steels with different strain rate characteristics. It may be noted that K_{1D} values derived from impact energy values using $K^2 = EG/1 - \nu^2$; $G = W/A$ where W is absorbed energy and A cross sectional area of the test piece sometimes do and sometimes do not agree with the K_{1D} values deduced as here from load measurements (Refs. 12 and 21), so that no such energy derived values are quoted here.

Dynamic fracture toughness values, interpreted as K_{1T} are shown, Fig. 10, for several mild and other steels. The materials are summarised, Table

1. It is seen that the materials fall into three groups. The steels P, P-type, DL and 5624 show fairly constant or but slowly rising values of K_{1D} until some 60 °C below their respective 50% Charpy FATT, above which temperature there is a fairly rapid upswing in K_{1D} values. At about the 50% Charpy FATT the tests pass to an above gross yield behaviour, so that the derived K values are no longer valid.

More tests, with statistical interpretation, might show that there were indeed differences, significant to the user, between the various low strength steels but at the broad sweep with which Fig. 10 is drawn, it is doubtful whether there is any marked difference in K_{1D} values along the low temperature plateau, although the values at the higher temperatures (still below the 50% Charpy FATT) are of course greatly affected by the temperature at which the upswing occurs, the slope of the rising portion of the curves being roughly the same.

The second group comprises the two directions of test on the UXW steel and the EN40C results for a fairly soft temper. Both of these materials are in the medium strength range, though not at all similar (Table 1). The trend differs from that for the other steels, but rising much more rapidly over the whole temperature range. Initiation K_{1C} values for these two steels are also shown for comparison.

The Battelle steel is an exception, in that K_{1D} at first rises rapidly with temperature, as for the second group, but flattens off at higher temperatures at K_{1D} values similar to the first group. This material is a pipeline steel, tested extensively (Ref. 22) and is again a medium strength material, but in a partly cold worked condition. The different patterns may not reflect a complete difference in behaviour since, if the curves for other steels are extended to yet higher temperatures the apparent K values obtained level off, or even reduce slightly. (Corresponding to tests showing low percentage crystalline appearance and clearly not valid K determinations and most certainly not plane strain.) The whole plot of nominal K versus temperature, invalid at its upper end, is thus an ogee-shaped curve, not unlike the well known impact energy–temperature curve, with different proportions covering the whole range of data in Fig. 10.

Conclusions

The results presented here are consistent with the general trend of the Eftis and Krafft strain rate dependence of fracture toughness of low strength steels. The Double Cantilever Beam test gives results for fracture initiation consistent with those from slow bend tests, and for crack arrest, consistent with those for instrumented impact tests. The instrumented impact test of a fatigue cracked and side grooved test piece is a convenient way for determining what are believed to be minimum (trough) dynamic K_{1T} values for low strength steels up to about the 50% Charpy FATT. Values

of K_{1T} for several low strength steels are presented. At very low temperatures, there is little difference in the values for the several mild steel types although some medium strength steels show rather higher K_{1T} values. Between the 50% Charpy FATT and about 60 °C below this there are larger differences in the K_{1T} values for all the steels, in part consequent upon the different absolute values of the Charpy FATT.

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Table 1
Material characteristics composition

Steel	*Composition %*						
	C	*Si*	*S*	*P*	*Mn*	*Ni*	*Cr*
P	0·18	0·05	0·039	0·028	0·54	0·16	0·010
P type	0·175	0·045	0·036	0·022	0·66		
UXW	0·19	0·30	0·04	0·05	1·60	0·5	0·30
En 40c	0·52	0·23	0·006	0·005	0·68	0·03	3·04
DL	0·155	0·24	0·021	0·024	1·35	0·22	0·67
5624	0·17	0·18	0·041	0·026	0·83		
Batt-elle	0·21–0·33	0·03–0·10	0·025	0·015	0·83–1·93		

Table 1
Material characteristics (contd.)
Mechanical properties

Steel	*YS tonf/in²*	*UTS tonf/in²*	*Charpy 50% FATT*	*Remarks*	*Refs.*
P	147	27·4	44°C	semi-killed as rolled	26, 27
P type	172	28·3	25°C	basic	12
UXW	36	45	−18°C	normal-ised & tempered	13, 24
En 40c	76	86	−63°C	tempered 465°	B.S. 970
DL	33·6	41·8	+20+35°C	normal-ised & tempered	
5624	13·2	28	+40°C	as rolled	
Batt-elle	28	40	+10°C	semi-killed	28

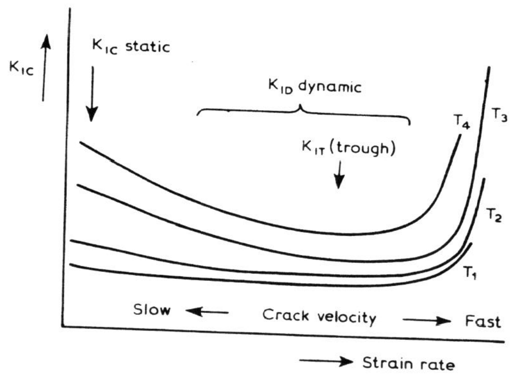

Fig. 1. General relationship K_{1C} vs strain rate $T_4 > T_1$. schematic only.

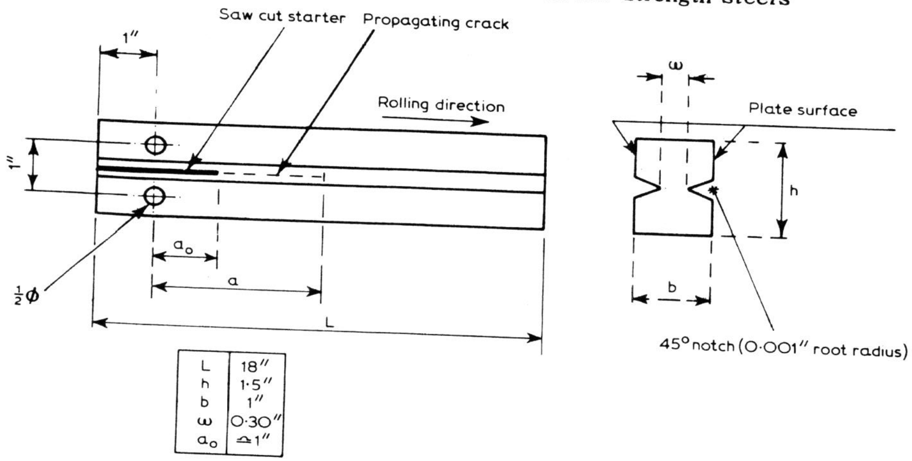

Fig. 2. DCB test piece.

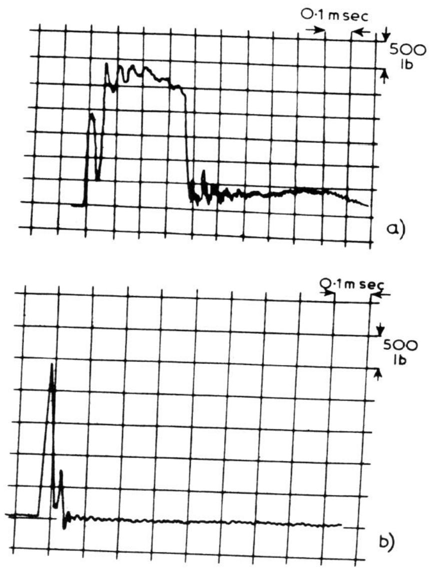

Fig. 3. Load-time diagrams of an instrumented impact test

Steel UXW, fatigued and side notched

(a) UX7FS at −2°C, 60% crystallinity

(b) UXX7FS at −75°C 100% crystallinity

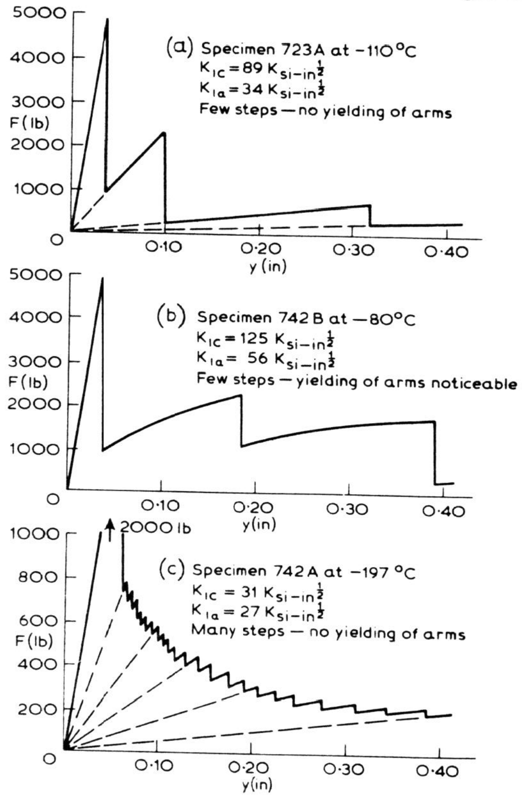

Fig. 4. Records load vs extension DCB specimen, steel UXW at ·005 in/min.

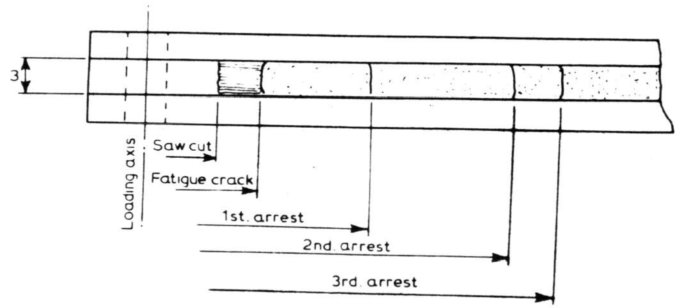

Fig. 5 (a). Fracture appearance of DCB specimen, steel UXW, schematic.

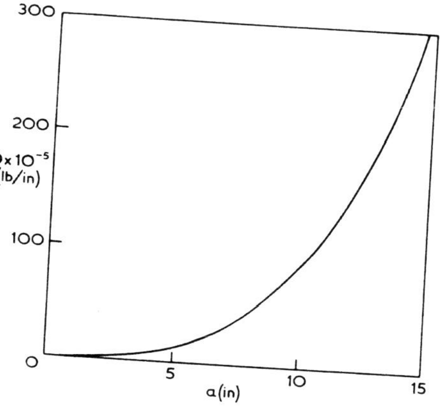

Fig. 6. DCB calibration for standard specimen.

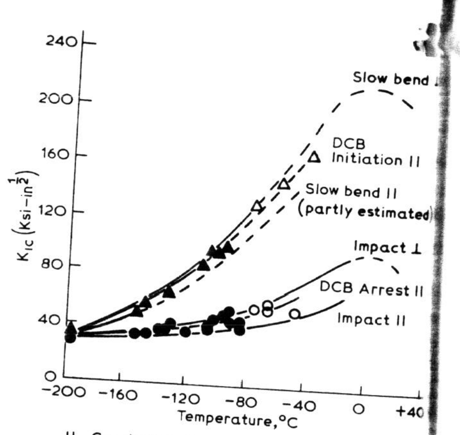

Fig. 8. K_{1C} **vs temperature, steel UXW.**

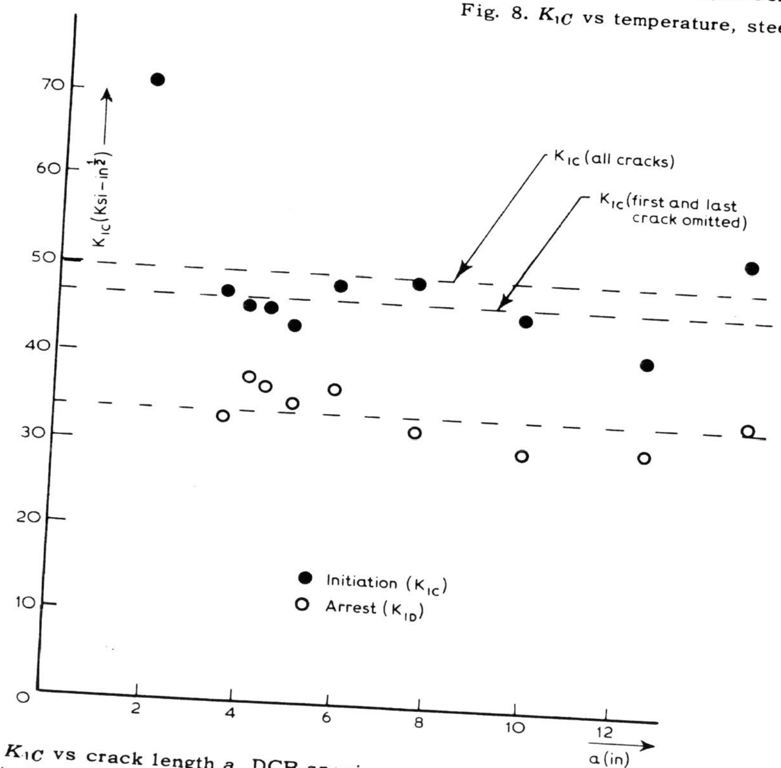

Fig. 7. K_{1C} **vs crack length** a**, DCB specimen, steel UXW. No. 110 at −150°C, at = ·005 in/min, total no. of cracks = 10.**

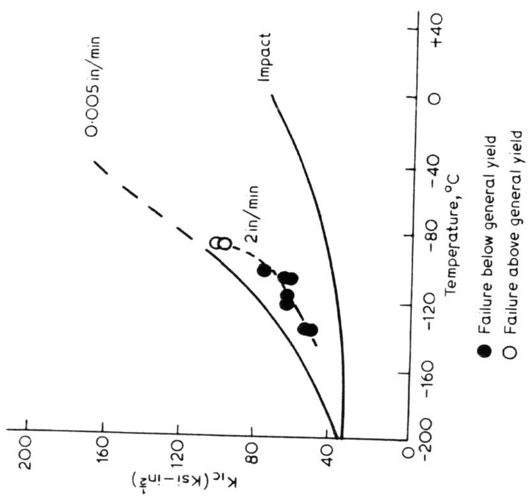

Fig. 9. K_{1C} **at different testing rates, steel UXW crack plane II R.D.**

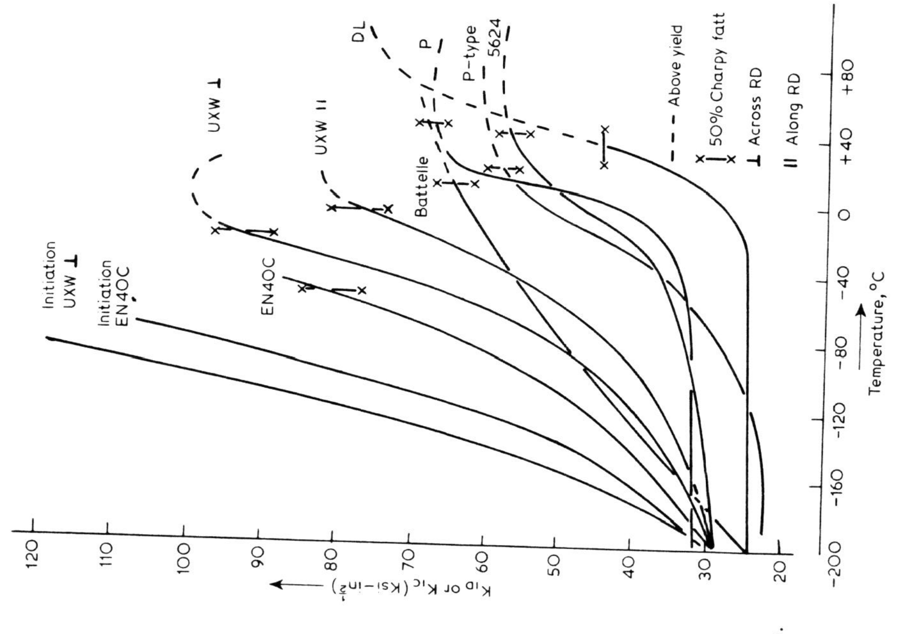

Fig. 10. K_{1C} **vs temperature, comparison of various steels.**