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ONSET OF FAST CRACK PROPAGATION IN HIGH STRENGTH STEEL AND ALUMINUM ALLOYS

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ABSTRACT

The concept of driving force per unit crack front or the "force tendency" is discussed. This is a more general and more readily useful concept than the Griffith theory in that it does not require the assumption of the absence of plasticity and there is no complication involving fixed grips. It is shown that the force tendency "$\mathcal{G}$" exists for any stress condition and that it can be determined from strain measurements near the crack head. Various formulae for $\mathcal{G}$ are given and the case of hydrogen embrittlement is discussed.

PROBLEM STATUS

This is an interim report; work on this problem is continuing.

AUTHORIZATION

NRL Problem F01-03
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Onset of Fast Crack Propagation in High Strength Steel and Aluminum Alloys.

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In this discussion a crack extension "force tendency" concept will be used. Since applications of this concept to fracturing other than the Griffith theory are not widely known a brief description of it will be given first. Measured values of the crack extension force at onset of fast fracturing for a variety of materials and test conditions will be shown and discussed. Practical use of critical values of crack extension force will be illustrated. The paper will conclude with a few remarks on hydrogen embrittlement.

The force tendency is defined in the following way. Assume one has under consideration a solid object acted upon by external forces in such a way that it contains a growing crack. If, during a time, δt, the work input from these external forces is δW_E and the change in stored recoverable strain energy is δU we can write

$$\delta W_E = \delta U + \delta W_Q \tag{1}$$

where the difference, δW_Q, must consist primarily of a conversion of energy to heat within the object.

Putting

$$\delta W_E = \sum_i F_i \left(\delta \ell_i^{pl} + \delta \ell_i^{el} \right) \tag{2}$$

where $\delta \ell_i^{pl}$ and $\delta \ell_i^{el}$ are the non-recoverable and recoverable increments of motion of F_i, one may observe that

$$\delta W_Q = \sum_i F_i \, \delta \ell_i^{pl} + \sum_i F_i \, \delta \ell_i^{el} - \delta U \tag{3}$$

We define the crack extension force tendency as

$$\mathcal{G} = \frac{\sum_i F_i \, \delta \ell_i^{el} - \delta U}{\delta A} \tag{4}$$

in the limiting case as δA, the element of new fracture area, approaches zero.

The only reason for this expression to differ from zero is the configurational change of the object due to extension of the crack. Its value depends only upon the applied forces, the configuration, and elastic constants of the material.

Calculation of $\mathcal{G}$ for the central crack in a plate subjected to simple tension, σ, shows that the crack extension force is proportional to the square of σ times the length of the crack, as one would anticipate from the analysis by A. A. Griffith(1). Using any two dimensional situation in which stresses are applied to a plate containing a crack along the x axis and such that τ_{xy} along the x axis is zero (see, for example, the solutions given by Westergaard)(2) one can readily obtain a local stress interpretation of $\mathcal{G}$. If the Cartesian position coordinates of the stress functions are replaced by polar coordinates, r, θ, using one end of the crack as origin of coordinates, as r approaches zero, the separational stress, σ_y, approaches as nearly as we please to

$$\sigma_y = \left(\frac{E\mathcal{G}}{\pi}\right)^{\frac{1}{2}} \frac{\cos\ \theta/2}{(2r)^{\frac{1}{2}}} \left(1 + \sin\frac{\theta}{2}\ \sin\frac{3\theta}{2}\right) \qquad (5)$$

where E is Young's modulus.

Cracks extend by a process of development and joining up of new fracture origins near the crack tips. In connection with the comparative extension possibilities of cracks of different length, and in different fields of overall nominal stress, σ, the result of the preceding paragraph means that the development of new fracture origins proceeds under the influence of similar separational stress environments for two different cracks when the values of $\mathcal{G}$ for these cracks are the same. In general for cracks of various length and location in any given solid material the value of $\mathcal{G}$ measures the intensity of the crack tip stress field so long as the influence of plastic deformations accompanying fracture extension is limited to the close neighborhood of the crack.

The omission of $\sum_i F_i \ \delta \ell_i^{pl}$ from the expression for the force tendency is a matter of choice. For example, when we analyze the longitudinal yielding of a tensile bar with no crack present the term corresponding to $\mathcal{G}$ is of negligible importance compared to $F \delta \ell^{pl}$ and the significant component of the force tendency is simply

$$\frac{\delta W_Q}{\delta \ell^{pl}} = F \tag{6}$$

On a per unit area basis this is, of course, just the longitudinal stress. Fundamentally we are concerned with a rate problem. The time rate of conversion to thermal energy is, approximately, $F \dfrac{\delta \ell^{pl}}{\delta t}$. Given various values of the force, F, one wishes to know the corresponding rate of yielding. This must be found experimentally. The familiar result is that a plot of $\delta \ell^{pl} / \delta t$ as a function of F shows the yielding rate increases from slow to fast over a relatively small range of values of the force tendency. (We neglect work hardening effects since these appear to have no counterpart in fracturing.) The value of F for this range, per unit area, is the yield strength and is properly regarded as a significant physical strength characteristic of the material.

If one follows an analogous procedure in the study of fracturing, various amounts of fracture extension force $\mathcal{G}$ may be applied to a test specimen containing a crack and the response measured in terms of time rate of fracture extension. It is found generally true that this extension rate changes from slow to fast for relatively small changes of the crack extension force. Therefore the critical $\mathcal{G}$ values, thus determined, have a significance relative to resistance to fracturing similar to that of yield strength as a measure of resistance to yielding.

Equations by means of which $\mathcal{G}$ may be calculated in certain crack extension situations are as follows:

1. For a central crack in a large plate subjected to tension, σ, normal to the crack direction

$$\mathcal{G} = \frac{\pi \sigma^2 a}{E} \tag{7}$$

where $2a$ is the crack length and E is Young's Modulus. This is the familiar Griffith expression. The same expression applies to an edge crack of length, a, normal to the tension, σ.

2. For an embedded "disc shaped" or "penny-shaped" crack in a medium in which the stress exerted normal to the plane of the crack is σ,

$$\mathcal{G} = k_1 \frac{\sigma^2 R}{E} \tag{8}$$

where R is the radius of the crack and k_1 is $4\,(1 - \nu^2)/\,\pi$ and ν is Poisson's ratio. This result was obtained by Sneddon(3).

3. For the embedded crack as above but with the addition of internal gas pressure, p, inside the crack

$$\mathcal{G} = \frac{k_1\,(\sigma + p)^2\,R}{E} \tag{9}$$

Table 1 shows values of $\mathcal{G}$ indicated by experiments to be typical for onset of fast fracturing in sheets and plates. There are undeniably large influences of material dimensions as indicated by the values for 24S-T4 in thin sheet as compared with thick plates. It is also interesting to note that a hard and somewhat brittle aluminum alloy 75S-T6 is tougher than mild steel when onset of fast fracture of the mild steel is accompanied by transition to predominately cleavage fracturing.

Table I

Typical values of the crack extension force, $\mathcal{G}$, necessary for onset of unstable fast fracturing*

Material	Plate of Sheet Thickness (in.)	Method	Temperature (Deg. Cent.)	$\mathcal{G}$ (in.lbs/in^2)
Steel, ship plate cleavage	0.75	a	- 20	100
Aluminum alloy 24ST4	1.00	b	+ 25	400
Aluminum alloy 24ST4	0.04	a	+ 25	600
Aluminum alloy 75ST6	0.04	a	+ 25	300
Aluminum alloy 78ST6	0.04	a	+ 25	150
Polymethylmethacrylate, as cast plates	0.125 to 0.50	a	+ 25	5
Polyesters, plates	0.125 to 0.50	a	+ 25	1
Polymethyl-alpha-chloracrylate plates	0.125 to 0.25	a	+ 25	2.8
Vulcanized natural rubber	0.035	c	+ 25	15 to 30
Glass, lantern slide covers, moist	0.02	a	+ 25	0.04
Glass, lantern slide in 2% R H air	0.02	a	+ 25	0.08
Cellulose acetate	0.025	a	+ 25	14
Cellulose acetate	0.002	a	+ 25	28

(a) Central crack type tension specimens in which the unstable crack length was about one quarter of the plate width.

(b) Slow bend tests of notched bars.

(c) Tear tests by Rivlin, Jnl. of Polymer Scie., vol. 10, no. 3, March 1953, pp. 291-318. Various compositions give values of $\mathcal{G}$ from 10^6 to 2×10^7 ergs/cm^2.

*Table prepared by J. A. Kies and H. L. Smith

Mild steel and plastics such as Plexiglas are strain rate sensitive materials. For such materials considerable variation of $\mathcal{G}$ can occur dependent upon the testing procedure. For example in the case of a typical sheet of Plexiglas for a given fixed magnitude of longitudinal stress the critical length of central crack for onset of fast fracturing can have three different values for the following situations: (a) the starting crack is represented by a slot made with a jeweler's hack-saw prior to application of the load; (b) the starting crack is tapped into the sheet prior to load application; and (c) the load is applied first, then the starting crack is formed suddenly by an impact. The values of $\mathcal{G}$ decrease from situation (a) through (c) by a total factor of about 2.

On the other hand the aluminum alloys are much less strain rate sensitive and the difference in $\mathcal{G}$ determined by methods (b) and (c) is correspondingly small. G. I. Taylor and his associates showed that the difference in static and dynamic yield strengths of steels decreased with increasing hardness. Although this suggests that the strain rate sensitivity of high strength steels in fracture tests may not be large, some reservations to this attitude may be necessary for steels which are temper brittle and for embrittlement by hydrogen.

R. H. Raring and J. A. Rinebolt of the NRL Metallurgy Division have been investigating fracture strength of high strength alloy steels. Their ASM paper at the 1955 Metals Congress will give some of the results of their work. In this paper use will be made of several of their test results and of test results from Syracuse University in order to estimate values of $\mathcal{G}$ for cracks under conditions of biaxial constraint.

Figure (1) shows a fracture obtained with a tensile bar of SAE 4340 heat treated to a tensile strength of 260,000 p.s.i. There was a 43% reduction of area. The region of the origin of the fracture is clearly indicated. When a fracturing process changes from slow to fast the change normally occurs rather abruptly with the more slowly fractured region differing in appearance from the region of fast fracture. Although this appearance difference is hardly noticeable in materials such as the aluminum alloys, it is a common feature of fracture appearance in many materials. The particular fracture shown on Figure (1) shows a region of lighter appearance surrounding the fracture origin and ring-like markings near the periphery of this region which appear to be hesitation lines. It is believed the position of change from slow to fast fracturing in this instance was the outer edge of the zone of lighter appearance. The expression for $\mathcal{G}$ obtained from Sneddon's analysis of embedded "penny-shaped" cracks can be used to calculate the critical $\mathcal{G}$ value.

In this example the average longitudinal stress at fracture on the necked down region of the tensile bar was 350,000 p.s.i. and the measured value of R was 0.025 in. Thus

$$\mathcal{G} = \frac{1.16\ (3.5 \times 10^5)^2\ (0.025)}{30 \times 10^6} = 120 \text{ in. lbs/in}^2 \qquad (10)$$

Figure (2) shows a fracture of a tensile bar of the same steel composition heat treated to a greater hardness than would be customary in practical applications. The ultimate tensile strength was 320,000 p.s.i. and the average longitudinal stress on the slightly necked down region was somewhat less than 380,000 p.s.i. If one assumes the lighter toned region around the fracture origin is again the slow fracture zone, calculation as above gives

$$\mathcal{G} = 150 \text{ to } 200 \text{ in.lbs/in}^2 \qquad (11)$$

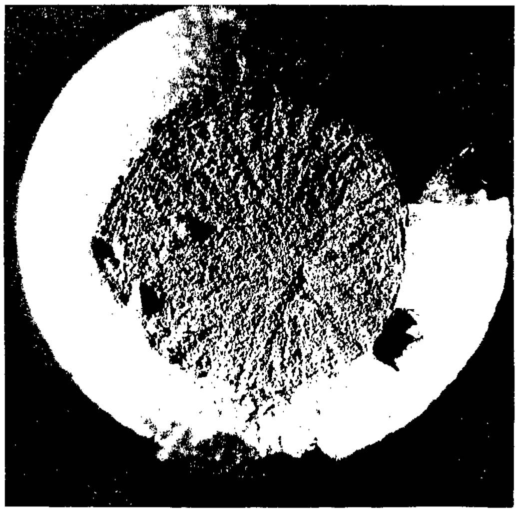


Fig. 1 - Fracture of SAE 4340 tensile bar, U.T.S. = 260,000 psi (see first example of Table II)

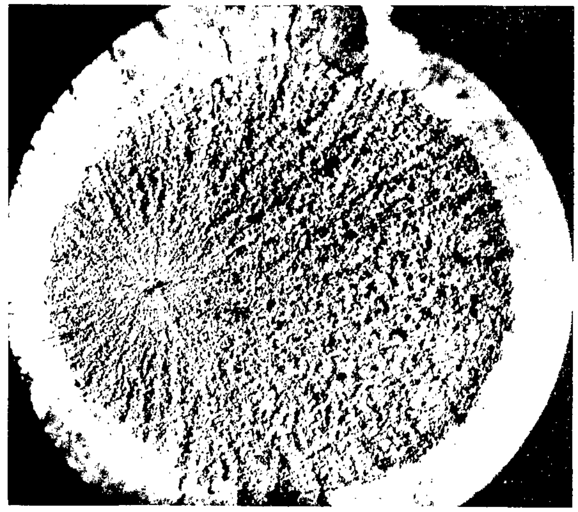


Fig. 2 - Fracture of SAE 4340 tensile bar, U.T.S. = 320,000 psi (see second example of Table II)

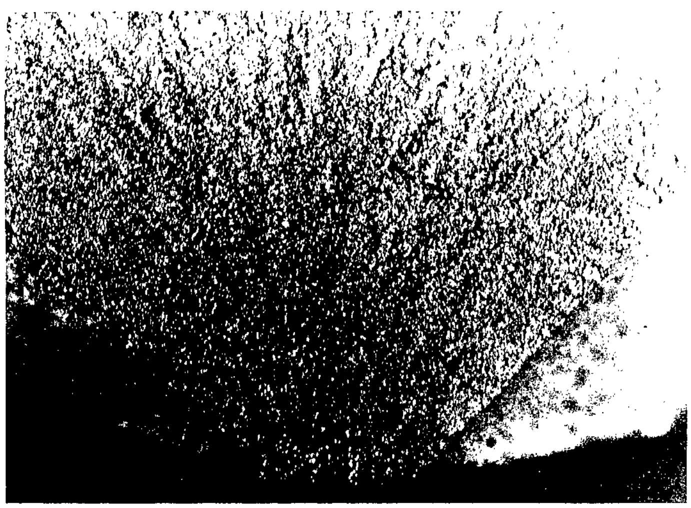


Fig. 3 - Fracture of SAE 4340 tensile bar, U.T.S. = 300,000 psi (see third example of Table II)

depending both upon judgment of the value of R and upon the uncertain magnitude of the load drop before fracture occurred.

Figure (3) shows another tensile bar fracture of the same steel again at a high hardness level. The ultimate tensile strength was expected to be about 300,000 p.s.i. Fracture occurred at a true stress of 293,000 p.s.i. at a surface flaw after about 2% reduction of area had occurred. The fracture appearance shows that the tendency of the material to develop localized plastic shearing near the bar circumference halted the spread of the flat central tensile fracture. Separations and weakening incidental to localized plastic shear then induced a change in the direction of the fracture onto surfaces of maximum shear stress. A value of $\mathcal{G}$ was computed using equation (7) and a value of a taken from measurements on the photograph to be 0.013 inches. The result was a $\mathcal{G}$ value of 117 in. lbs/in^2.

The results of Raring and Rinebolt included tests in which sharply notched tensile bars were used. A larger group of results was also available from notched bar tests at Syracuse University. In both places the depth of the notch was chosen so that the holding area was reduced by a factor of two.

A fair approximation of $\mathcal{G}$ for inward extension of a circumferential crack can be made using the following expression

$$\mathcal{G} = \frac{\pi \sigma^2 a}{E} \left\{ 1 + \frac{a}{R_o - a} \left(1 - \frac{2}{\pi} \right) \right\} \qquad (12)$$

Here the notch depth plus the average depth of the starting crack at instability is a, R_o is the radius of the bar, and σ is the nominal stress in regions of the bar remote from the crack.

Using the work at Syracuse University, examination of the notched bar fractures for various size bars of an SAE 4340 steel and for various size bars of 75S-T6 aluminum alloy was made with the purpose of selecting a situation most nearly corresponding to inward spreading of the crack from the notch root all around the specimen. The graph of results for notch strength as a function of size showed in each case a plateau region for the small sizes followed by a drop off in notch strength with increasing test bar diameter. In both materials the fractures most nearly approximating to the desired "inward spread of the crack" situation appeared on test bars at the beginning of the notch strength drop-off region of bar diameter. Thus calculations were made for a 0.35 inch steel bar and for a 0.8 inch aluminum alloy bar.

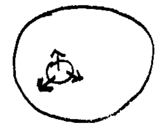
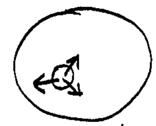
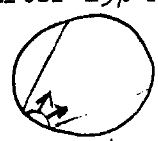
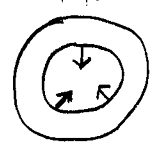
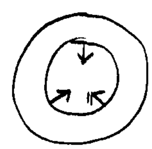
For the SAE 4340 steel bar, using the value of notch depth for "a" in the equation, the $\mathcal{G}$ value obtained was 180 in.lbs/in^2. For the 0.8 inch 75S-T6 test bar the calculation on the same basis gave 100 in.lbs/in^2.

Table II gives a summary of critical crack extension force estimates for the examples discussed above. It will be noted that the value for 75S-T6 in Table II is one third of the value shown for this material in Table I. The difference is due to the additional work which enters into the fracturing process for a crack running across a plate in regions of the crack close to free surfaces of the plate. Although no values for $\mathcal{G}$ pertaining to running of cracks in plates of high strength steel have been measured it may be expected that these values when obtained will be greater than 200 in.lbs/in^2., probably in the neighborhood of 500 in.lbs/in^2. In all the $\mathcal{G}$ results for metals presented above, the plane of the fracture is perpendicular to the direction of greatest rolling or forging extension. If the expected weakest fracture orientations had been used the values of $\mathcal{G}$ would have been greatly reduced.

As an example of use of measurements of critical $\mathcal{G}$ values consider the case of fracture propagation which appears to approximate the unfortunate Comet planes. The material used here is based upon work by Kies, Smith and Brossman(4)

Table II

Critical Crack Extension Forces for
Cracks under Biaxial Constraint

Material	Un-notched Tensile Strength (p.s.i.)	Assumed Conditions of Crack Extension	$\mathcal{G}$ (in. lbs/in^2)
SAE 4340	260,000	 (After 43% RA)	120
SAE 4340	320,000	 (After 29% RA)	150 to 200
SAE 4340	300,000	 (After 2% RA)	117
SAE 4340	260,000		180
75S-T6	78,000		100

at the Naval Research Laboratory. If a pressurized fuselage of 75S-T6 aluminum alloy is made to bear service hoop stress loads of 20,000 p.s.i. a simple calculation using 300 in.lbs/in^2 as the critical $\mathcal{G}$ value shows that a crack longer than 5 inches will be unstable. Now it may be shown analytically and has also be demonstrated in NRL laboratory tests that a crack starting from the edge of an opening in a sheet in tension is subjected to a crack extension force approximately equivalent to that of a crack equal in length to the full size of the opening. Thus a crack started by fatigue or otherwise at a window opening in such a pressurized fuselage would propagate unless the window dimensions were less than 5 inches. On the other hand a more cautious practice in plane design might result in a fuselage of 24S-T4 aluminum alloy carrying hoop tension stresses no larger than 15,000 p.s.i. Using 600 in.lbs/in^2 for the critical value of crack extension force one finds that in such a plane cracks developing out from a window opening would not become unstable until the combined length of crack plus window opening exceeded 18 inches.

Numerous practical utilizations of the crack extension force concept cannot be expected until supported by a more extensive set of experimental measurements than is currently available. The limited number of applications which have so far developed pertain to the "crack running in a plate" type of situation.

Values were given in preceding discussion for onset of fast fracture in situations where plastic distortions accompanying crack extension are restricted as for an embedded crack or for a crack at the root of a sharp notch. These critical crack extension force values were rough approximations only. When additional experiments provide values that can be used with greater confidence it will be possible to calculate the depth of surface cracks and diameter of internal cracks which, at given stress levels, may be expected to cause onset of fast fractures.

We turn next to the influence of hydrogen molecular gas pressure within an embedded crack. Carney, Chipman, and Grant(5) calculated that pressures ranging from 105,000 to 220,000 p.s.i. would develop at room temperature in cavities within steels containing 0.0001 to 0.001 per cent by weight of hydrogen. The lower figure is not an uncommon hydrogen content. Selected regions of hydrogen embrittled steel might contain 0.001 per cent hydrogen. From equation (9) it would appear that an internal crack, capable of extension by 300,000 p.s.i. tensile load under conditions such that hydrogen pressure has no effect, would extend equally fast at 200,000 p.s.i. when assisted by 100,000 p.s.i. of hydrogen pressure. However, the extension rate must be very slow, allowing time for diffusion of hydrogen to maintain the high internal pressure, in order for hydrogen to have its maximum influence.

A multiplicity of well developed cracks only one or two of which are included in the final fracture surface is a typical appearance feature of hydrogen embrittled steel under stress rupture test conditions. To explain this appearance feature one needs a self-arresting crack extension process. Referring to equation (9) again, one may note that, when a quick increment of crack extension occurs, the value of $\mathcal{G}$ would be increased by the increase of crack radius, "a", but would be decreased by decrease of the internal pressure, p. Whether $\mathcal{G}$ decreases as the crack extends quickly depends upon the ratio, p/σ, and upon the initial non-elastic volume of the crack. Calculations based upon these considerations lead one to conclude that minimum internal pressures ranging from $\frac{1}{4}\sigma$ (for negligible non-recoverable crack volume) to $(5/4)\sigma$ (for relatively large non-recoverable crack volume) are required for the crack extension to be self-arresting. It appears therefore that the mechanical internal pressure influence of hydrogen in steel provides a plausible explanation of the extensive growth of cracks as a self-arresting process as well as of other major features which characterize hydrogen embrittlement.

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