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MECHANISM OF LOW-STRESS BRITTLE FRACTURE IN NORMALLY DUCTILE MATERIALS

L. P. TRUDEAU

PHYSICAL METALLURGY DIVISION

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MECHANISM OF LOW-STRESS BRITTLE FRACTURE IN NORMALLY DUCTILE MATERIALS

by

L. P. Trudeau*

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ABSTRACT

A mechanism is suggested for brittle failures, in service, from small flaws in normally ductile materials which exhibit an impact energy transition behaviour. The theory suggests, and experiments confirm, that the minimum defect size required for fracture may be more than an order of magnitude smaller than the average size required. The necessary conditions for such a fracture are: (1) the material must exhibit dynamic nil-ductility behaviour under the conditions of stress, temperature and section size that apply in the structure; (2) there must be a defect of sufficient size to give rise to a stress field capable of sustaining a running fracture; and (3), at the base of this stress raiser, there must be a metallurgical imperfection or operative process capable of giving a fast increment of rupture over a microscopic distance comparable to the plastic zone size of a running fracture. The critical flaw size can be calculated from the dynamic K_{1c}. Impact loading is not necessary for initiation.

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Direction des mines

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MÉCANISME DE RUPTURE FRAGILE SOUS DE FAIBLES CONTRAINTES DANS LES MATÉRIAUX NORMALEMENT DUCTILES

par

L. P. Trudeau*

RÉSUMÉ

L'auteur propose une explication du mécanisme de rupture fragile, à l'usage, provenant de petites défectuosités dans des matériaux normalement ductiles qui présentent un comportement de transition énergétique sous l'impact. Selon sa théorie, confirmée par l'expérience, les dimensions minimales requises de la défectuosité pour provoquer une rupture peuvent être plus petites que les dimensions moyennes requises. Les conditions nécessaires à une telle rupture sont: (1) le matériau doit avoir un comportement dynamique de ductilité nulle sous les conditions normales de contrainte, de température et de dimensions de la section de la structure; (2) la paille doit être assez grande pour causer un camp de contraintes capable de faire progresser une rupture; enfin, (3) à l'origine de cette concentration de contraintes, il doit y avoir une imperfection métallurgique ou des conditions de chargement susceptibles d'accélérer la rupture sur une distance microscopique comparable aux dimensions de la zone de déformation plastique d'une rupture en progrès. Les dimensions critiques de la paille peuvent être calculées à partir du "K_{1c}" dynamique. La mise en charge par choc n'est pas nécessaire pour amorcer la rupture.

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1. THEORY

Brittle service failures in structures of ferritic steel have occurred sporadically. Correlations with steel behaviour in dynamic empirical tests have been established, but it has been difficult to reconcile the high resistance to fracture initiation shown by the steel in slow strain-rate tests, with its apparently brittle behaviour under similar strain-rate conditions in service. In this report a mechanism is suggested by which a stress raiser (or defect) much smaller than the average size required could lead to brittle fracture. The quantitative discussion centres on ferritic steel because the necessary data are available. However, the suggested concepts are considered to be quite general, and directly applicable to other materials with similar macroscopic fracture characteristics.

Using steel for the purpose of characterizing the class of materials, we wish to consider those of lower strength and higher ductility which exhibit an impact energy transition behaviour. The present ASTM fracture toughness testing recommendations(1) specifically exclude this type of material and note that the procedures apply to steels of small strain-rate sensitivity with a yield strength in excess of 200, 000 psi. The properties of the steels under present discussion are very strain-rate sensitive and the implications of this strain-rate sensitivity play a central role in the proposed failure mechanism.

An impact energy transition behaviour indicates that the energy absorption for fracture changes from low to high as the temperature increases over a certain temperature range under high strain-rate conditions. In the case of ferritic steel, this energy transition results from the presence or absence of crystallographic cleavage fracture. The applied strain rate, through its influence on the yield strength, affects the temperature at which this transition occurs. This particular change in energy absorption for fracture is a metallurgical effect, distinct from the normal decrease in energy absorption as the stress sytem changes from plane stress to plane strain.

At present, for protection against brittle fracture of critical components made of these lower strength ferritic steels, limiting service temperature criteria are established in terms of the nil-ductility drop-weight test in the U.S.A. and the Robertson crack-arrest temperature in Britain. These criteria are often applied through correlation with the more commonly used Charpy V notch impact test. The justification for these tests lies in empirical correlation with service failure experience.

To resolve the apparent paradox in dynamic tests correlating with service failures under "static" loading, the suggestion has been made that we should look for the possibility of something dynamic about "static" failure initiation[2]. Just because the material is strained slowly, the velocity of increments of rupture is not necessarily slow. This led to the analogy that the drop-weight test may correlate with service failures because it simulates the failure conditions, both with regard to the rate of loading imposed on adjacent metal by high-velocity increments of rupture and with regard to the low deflection possible in the elastically loaded structure during the early stage of fracture.

The stress levels in the type of structures being considered are only about one-quarter to one-third of the ultimate tensile strength. Since these materials have high nominal ductility, ductile rupture at the root of a notch with stresses at these levels would only occur, under average conditions, if the notch-type defect were several inches deep[3,4,5,6].

If an increment of rupture is to occur at the base of a defect that is an order of magnitude smaller than this average size, then the material at the base of this defect must not exhaust its average ductility before an increment of rupture occurs. Such a special increment of low ductility fracture could result from an inclusion at the base of the stress raiser, corrosion or other causes. If this increment of rupture, which will be assumed to be fast, is to result in further increments of rupture, the stress field must be high enough to sustain a fast-growing fracture. At the same time, the material must exhibit nil-ductility behaviour under these fast-fracture conditions. For ferritic steel, we might define nil-ductility behaviour by saying that the radius of the plastic zone accompanying the fracture (or the shear lip width near the start of the fracture) must be of order of magnitude 0.001 in. or less. In the drop-weight test the plastic zone size at the empirical nil-ductility temperature is somewhat larger than this, because yield point stress is applied and some plastic bending of the specimen is allowed to occur. Nil-ductility behaviour is necessary because, when the fracture is propagating from the small defect size up to the several-inch size required for initiation under average conditions, the fracture must grow with a small crack-opening displacement. The fast fracture must be sufficiently brittle that a deflection of elastic magnitude is enough to accommodate it.

In considering the minimum necessary length of the initial increment of fast fracture, it seems reasonable to postulate that it must occur over a distance comparable to the plastic zone size of a running fracture. If the material is at or below the nil-ductility temperature, this plastic zone is of microscopic dimensions.

This fast increment of rupture will cause the stress field to move ahead locally in much the same way as does the stress field in a propagating fracture; that is, a rapid rate of loading will be imposed on the adjacent metal. At or below the impact energy transition temperature, a low energy mode of fracture is possible under these high strain-rate conditions. The applicable fracture toughness, then, is that for a propagating fracture at the service temperature, since the proposed initiation mechanism essentially eliminates the need to consider initiation energy.

Recapitulating, this hypothesis or model for fracture hinges on the lack of ideality of the material in the structure. The theory suggests that the minimum defect size required for fracture may be more than an order of magnitude smaller than the average size required. This type of fracture initiation is rarely found in laboratory tests because, with uncorroded, machined specimens, the average ductility of the metal is usually exhausted before an increment of rupture occurs. We have postulated a minimum length for the necessary increment of fast fracture. It may be, of course, that in service failures the initiating increment of fracture was often longer than this minimum. Also, residual tensile stress in a structure will help to propagate the fracture and will decrease the necessity for nil-ductility behaviour. However, the presence of such stresses is not a requirement for the applicability of the proposed mechanism.

A method is now needed for calculating the defect size required to sustain a fracture initiated under these conditions. This defect may have resulted directly from fabrication, or a smaller defect may have grown to critical size as a result of fatigue or other processes. Since the structure as a whole is behaving elastically and the plastic zone size for the dynamic crack is of microscopic dimensions, the elastic stress intensity concept (K_{1c}) developed by G. R. Irwin can be employed with good accuracy. This approach has been extensively used for high strength materials[1], so a background of information on the effect of flaw configuration and other factors is available. For present purposes, a simple edge crack model with crack depth "a" will be used (Figure 1). This is the form of the specimen used for the experiments reported in Section 2.

Using polar coordinates (r, θ) from the tip of the crack in the usual way[7, 8], the opening mode stress field perpendicular to the crack plane for values of "r" that are small relative to "a" is

$$\sigma_y = \frac{K_1}{\sqrt{2\pi r}} \cos\frac{\theta}{2}\left(1 + \sin\frac{\theta}{2}\sin\frac{3\theta}{2}\right). \quad \ldots\ldots\ldots\ldots \quad (1)$$

Since the terms involving θ are dimensionless, K_1 must be, from dimensional considerations alone, of the form (stress $\sqrt{\text{distance}}$)[8]. For the crack considered[8],

$$K_1 \approx \sigma\sqrt{\pi a}\,, \qquad \ldots\ldots\ldots\ldots\ldots \qquad (2)$$

where σ is the nominal applied stress remote from the crack, and "a" is the crack depth.

For the purpose of setting up an experiment, an approximate value for the critical $K_1(K_{1c})$ is needed. The exact value is expected to be a function of temperature and metallurgical variables such as composition and grain size. By explosively loading notched specimens, J. M. Krafft and A. M. Sullivan[9] have determined that the dynamic K_{1c} for ferritic steel (evidently fracturing by crystallographic cleavage) is about 20, 000 psi$\sqrt{\text{inch}}$. Now, if we assume that the steel is under a nominal stress of about 20, 000 psi, then from Equation 2,

$$K_{1c} \approx \sigma\sqrt{\pi a}$$

$$20,000 \approx 20,000\sqrt{\pi a}$$

$$a \approx 1/3 \text{ in.}$$

This defect size is more than an order of magnitude smaller than that which would be expected on the basis of the average slow strain rate ductility of the steel.

The dynamic K_{1c} can be estimated from a Robertson test as well. For the notch depth, take the starting notch plus the distance ahead that the stress field is distorted by the initiating impact (say 1/4 in. and 1 in. respectively[10]). Below the arrest temperature, cracks have run at 10, 000 psi:

$$K_{1c} \approx 10,000\sqrt{\pi(1.25)} \approx 20,000 \text{ psi}\sqrt{\text{inch}}$$

This figure is similar to that obtained by Krafft and Sullivan. Since cracks have run in a Robertson test at stresses below 10, 000 psi, it would seem that K_{1c} can be lower than 20, 000 psi$\sqrt{\text{inch}}$ and the critical defect size correspondingly smaller.

Using 20,000 psi$\sqrt{\text{inch}}$ as an approximate value for K_{1c}, a check on the model can now be made. J. A. Hendrickson, D. S. Wood and D. S. Clark[11] found in tests on a carbon steel that a tensile stress of 210,000 psi was necessary at the plastic-elastic boundary if the crystallographic mode of failure was to operate. Krafft and Sullivan in their tests found evidence for a "process zone" of radius about 0.001 in. Considering, also, the appearance of Robertson test specimens with regard to shear lip width, it seems reasonable to take 0.001 in. as the radius (the width on the crack plane) of the plastic zone for ferritic steel near but still below the NDT temperature. Then, using Equation 1 with K_1 = 20,000 psi$\sqrt{\text{inch}}$ and r = 0.001 in.:

$$\sigma_y = \frac{20,000}{\sqrt{2\pi(0.001)}} = 250,000 \text{ psi.}$$

This stress at the location where the plastic-elastic boundary is expected to be is consistent with the value Hendrickson, Wood and Clark found necessary for their particular steel. When the initial increment of fast fracture occurs, the stress field can be expected to move ahead and adjust to its new value before any deformation process associated with the next increment of rupture has had time to occur.

The dynamic crack is very brittle and the tensile stress required for cleavage is very high. However, the low-strain-rate strength of the steel is low and the original straining is at a low strain rate, so the yield zone at the defect is larger than one might at first suppose. Estimated from the usual relation with K_1 = 20,000 psi$\sqrt{\text{inch}}$ and a yield strength of 40,000 psi, the yield zone for plane strain conditions is approximately:

$$\text{Yield zone width} = \frac{K_1^{\,2}}{6\pi(\text{Yield Strength})^2}$$

$$= \frac{(20,000)^2}{6\pi(40,000)^2} = 0.013 \text{ in.}$$

Under plane stress conditions the yield zone is about 3 times as wide. So, even in the absence of residual stress, there is yield strength loading for some distance ahead of the defect.

An elastic-type analysis has been used throughout this discussion. Using a photoelastic coating, J. R. Dixon and W. Visser[12] found that in an aluminum alloy the strain distribution tended to remain similar to an elastic one even after some plastic flow at the tip of a crack. A mild steel sheet behaved similarly until the development of a visible plastic zone, at which point the strain distribution ceased to be of the elastic type. However, on reloading, the strain distribution was of the elastic type up to the previous maximum load. Depending on the loading history, this possible qualification may be more relevant to test conditions than to service conditions.

As the temperature increases into the notch impact transition temperature range, the stress required for fracture will increase because the plastic work and the deflection, required for fracture, increase. The gradualness of this change in fracture behaviour can be expected to vary with the material. There may even be marked differences in behaviour among different heats of one grade of steel. An illustration of these differences may be found in the variation of the correlation between the Robertson test and the drop-weight test. G. D. Fearnehough and H. G. Vaughan[13] found that the Robertson gradient arrest temperature could only be predicted from the drop-weight NDT temperature to within plus or minus 32°C and the Robertson isothermal arrest temperature to within plus or minus 25°C.

An implication for the Griffith theory, resulting from inertial effects associated with fast increments of rupture, is discussed in the Appendix.

2. EXPERIMENTS

The theory presented in the previous section suggests an approach to an experimental check. Suppose we had a tension test specimen containing a notch and at the base of this notch we had a thin layer with a static crack toughness equal to the dynamic crack toughness of the rest of the material. Then, when this specimen is pulled in an ordinary tension test, the brittle layer would crack at the applicable stress and present a short increment of fast rupture to the adjacent material. In the presence of an approximately plane strain, elastic-type stress field, the crack should continue in a brittle manner if the test is done in a temperature range where the material shows dynamic nil-ductility behaviour. The concepts of fracture outlined in Section 1 are not limited to steel, but general interest, material cost and the availability of related technical data all favoured the use of steel for the initial tests.

Specimen dimensions were arrived at by quantitative theoretical considerations plus the additional requirement that the plate be thick enough for a drop-weight test to be run on it for comparison. The test specimen was a saw-cut rectangle of plate 3/4 in. thick, 4 in. wide and 22 in. long, with a single edge notch 1/2 in. deep or less. Some tests were also run on 1/4 in. by $1\frac{1}{2}$ in. plate samples with a similar notch. At the base of the notch there is a brittle layer the function of which is to provide a starting increment of rupture at a stress below the yield stress. The specimen is cooled to the desired temperature and pulled in a tension test machine, using wedge grips at an ordinary tension test rate of loading. The alcohol-and-dry-ice cooling bath is clamped to the specimen and remains in place during the test. To avoid ambiguity, no auxiliary devices such as wedges are used to start the fracture. The main experimental difficulty hinges on this point, because the characteristics of the brittle layer, combined with the notch radius, must be such that the layer cracks in unaided simple tension in the stress range of interest. The objectives of the tests were to demonstrate that low stress failures could be obtained with small notches at temperatures fairly close to the impact energy transition temperature range, and to study the cracking behaviour, rather than to measure the dynamic K_{1c} of the steel, although some information on this was obtained. An acoustic pick-up was used for some tests but usually the starting increment of fracture was audible if a complete break did not occur. Three different lots of plate were used.

In the first trials, a hard-surfacing welding electrode was used for the brittle layer with the notch cut by a diamond-studded wheel. Troubles were experienced with natural cracks in the deposits, but one trial was successful with the specimen breaking in two in a brittle manner with a nominal stress on the gross section of 26,000 psi, which was well below the yield strength. Examination showed, however, that the technique was inadequate because carbon diffused into the steel and a comparatively tough layer was formed in the heat-affected zone. There was also a question about the possible role of residual stresses. The deposit softened too much on stress-relieving to crack in the desired stress range. Various other techniques were tried; the two most successful were (a) nitriding, (b) electron beam welding using a zirconium wire to form a surface layer at the base of the notch.

First Lot of Welded Plate

This 3/4 in. plate was commercial, carbon structural steel, hot-rolled and pickled, with a brittle layer formed from 1/16 in. diameter zirconium wire or a Nicrobraz alloy applied by electron beam welding. All of the deposits were cracked from the welding process, with the cracks extending to the steel interface in most cases.

Brittle fractures were obtained in tests at -73°C, -47°C, -25°C and -14°C. Figure 2 shows photographs of the fractures at the lowest and highest temperatures. A micrograph of the profile at the start of the fracture at -73°C is shown in Figure 3, and it may be seen that the whole heat-affected zone is less than 0.010 in. deep. There is a martensite layer about 0.002 in. deep at the bottom of the notch at the interface between the deposit and the steel. The consistency of breaking load among the specimens suggests that, under tension, the natural crack in the deposit caused cracking in this martensite layer and thus initiated the fractures. The nominal stress for this fracture was 33,000 psi, with the average stress on the net section 38,000 psi. Tensile test bars (0.505 in.) were machined from the fractured specimen and pulled at -73°C. These were very ductile with an upper yield strength of 48,500 psi, lower yield 47,100 psi, ultimate tensile strength 69,300 psi, 46% elongation in 2 in., and 66% reduction in area. This shows that brittle fracture occurred well below the yield strength. The rough appearance of the fracture suggests that initiation at a lower stress would still have resulted in failure.

The specimen broken at -14°C, unlike the ones at lower temperatures, had a noticeable shear lip, which is not clearly shown in Figure 2. This specimen broke at a nominal stress on the gross section of 34,000 psi and an average stress on the net section of 39,000 psi. A duplicate test at -15°C, using a welded Nicrobraz deposit, gave an audible click at 33,000 psi but the fracture increment did not propagate. Tensile bars machined from the broken specimen and tested at -14°C were again very ductile, with an upper yield strength of 35,700 psi and a lower yield strength of 34,700 psi. These results suggested that the temperature limit for propagation with yield point stress and a 1/2 in. notch was about -14°C and also suggested that the empirical drop-weight nil-ductility temperature should be fairly close to -14°C. Exact agreement would not be expected, because some plastic bending occurs in the drop-weight specimen and a somewhat more ductile crack can presumably be accommodated. Seven drop-weight test bars of

2 in. x 5 in. size were prepared from the broken specimens. Because of circumstances at the time, a non-standard machine with a higher impact energy and larger deflection was used. This gave what are defined as breaks in the ASTM Standard up to -6°C, but no break at -1.5°C. This is in agreement,, within about 10°C, with the tensile test results.

From the test at -15°C in which an increment of fast rupture occurred but the fracture did not propagate, a lower limit for the dynamic K_{1c} can be estimated. The presence of plastic deformation stress will increase the strain energy release rate above that calculated from an elastic approximation. But using an elastic approximation, K_{1c} is greater than $1.2(33,000)\sqrt{\pi(0.5)} = 50,000 \text{ psi}\sqrt{\text{inch}}$ for this 3/4 in. thick material. This indicates that this drop-weight test applied a higher driving force than that represented by $50,000 \text{ psi}\sqrt{\text{inch}}$.

Second Lot of Welded Plate

This was a 3/4 in. thick plate of hot-rolled 0.22 carbon structural steel. Most of the electron beam welds were made with 1/16 in. diameter zirconium wire. In four cases titanium was used, but this material seemed less satisfactory for the purpose.

One notable occurrence in this series of tests was that one specimen broke at -67°C with a nominal stress of 14,200 psi on the gross section, which is an average stress of 16,400 psi on the net section. The K_{1c} for this fracture was approximately:

$$K_{1c} = 1.2(14,200)\sqrt{\pi(0.5)} = 21,000 \text{ psi}\sqrt{\text{inch}}.$$

This value is thus very close to the one estimated before the tests began. The fracture was quite smooth. It is possible, of course, that initiation at a somewhat lower load would still have resulted in failure.

Charpy V-notch impact test results for this steel are given in Figure 4. Drop-weight test results on 2 in. x 5 in. specimens, using ASTM standard test conditions, showed variable behaviour with an NDT of approximately -18°C, but one specimen at -24°C was a "non-break". The Charpy V-notch energy absorption at -18°C was approximately 19 ft-lb.

Tensile test fractures were obtained up to -15°C, but above about -35°C they were not nil-ductility fractures, since the shear lips were quite evident near the start of the fracture. This may be seen in Figure 5, which is a photograph of the fracture at -15°C. This fracture occurred at an average stress on the net section of 41,300 psi, which is a stress between the upper and lower yield points of 42,500 and 37,500 psi at this temperature. Initiating increments of fracture in the yield-point stress range were necessary for fracture at temperatures above -35°C. An initiating increment of fracture occurred at an average stress of 39,600 psi on the net section at -35°C, but was arrested. The upper and lower yield stresses at this temperature are 45,000 and 40,000 psi, respectively. This arrested fracture suggests that the K_{1c} value for this 3/4 in. thick material at -35°C was greater than 50,000 psi$\sqrt{\text{inch}}$. The Charpy V-notch impact energy absorption was variable at this temperature, but averaged about 10 ft-lb.

Residual Stress Measurements

Since these samples had a deposit applied by welding, some consideration was given to the possible influence of residual stress. It was not thought that residual stress could be playing an important role, because the deposits were cracked and the heat-affected zone was only about 0.010 in. This heat-affected zone was smaller than the yield zone (see Section 1), so the local residual stress should be eliminated by plastic deformation. Nevertheless, residual stress determinations were made on two samples, using a layer of photostress plastic. The residual stress was found to be confined to a layer within about 0.015 in. from the deposit, and the peak shear stress was about 5,000 psi.

Some residual stress on a microscopic scale can also be expected in connection with the nitrided specimens to be described. It may be noted, though, that residual stress on a microscopic scale can be expected even in stress-relieved material, as a result of difference in expansion of oxide scale and metal or as a result of increased volume of corrosion product in a defect. Residual stress, in the ordinary macroscopic sense of the term, was not a factor in these tests.

Nitrided Specimens

Some tests were done with specimens nitrided at the notch. Before notching, the specimens were electroplated with a layer of nitriding stop-off, which consisted of more than 0.002 in. thick copper except for one specimen which had a 0.001 in. nickel layer. After an initial saw cut, the notch was deepened by electrodischarge machining, using 0.002 in. thick brass shim stock for a cutting tool. The base of this notch was then sharpened with a serrated razor blade. Nitriding was for either 8 hr or 24 hr at 524°C (975°F). Metallography and microhardness surveys showed that the plating acted as an effective nitrogen barrier.

Two of the specimens were from 3/4 in. C.S.A. G40.8 "as hot rolled" plate, which was aluminum-killed with 0.19% carbon and 1.38% manganese. The specimen nitrided 8 hr broke at -75°C at an average stress on the net section of 32,000 psi, while that nitrided 24 hr broke at an average net section stress of 27,300 psi. These stresses are well below the yield stress but the driving force was high enough to completely sever the nickel plate in one case and the copper plate in the other. The fracture face of the specimen broken at 27,300 psi (nominal stress 23,800 psi) is shown in Figure 6. The fracture is practically a perfect plane strain one. A microhardness survey at the notch showed a peak hardness of 46 R_c which extended to a depth of 0.010 in.; the hardness then decreased to 21 R_c at a depth of 0.030 in. The hardness was approximately that of the base plate at a depth of 0.050 in. The Charpy V-notch impact energy absorption of this steel was about 10.3 ft-lb at -40°C.

Some smaller samples, 1/4 in. thick, $1\frac{1}{2}$ in. wide, and approximately $10\frac{1}{2}$ in. long, were machined from the second lot of plate, for which full thickness tests have already been described. After the copper plating, a single-edge notch was cut to a depth of about 7/16 in., using, in the way just mentioned, a saw-cut, 0.002 in. brass shim stock and a serrated razor blade. The specimens were then nitrided. Two of these specimens broke at -75°C, far below the yield, with an average stress on the net section of 29,000 psi (nominal stress 21,000 psi). One of the fractured faces is shown in Figure 6 and, as may be seen, is an example of essentially ideal plane strain fracture. A microhardness survey at the notch showed a peak hardness of 27 R_c and this extended to a depth of 0.007 in. Beyond a depth of 0.010 in., the hardness had decreased to that of the base plate.

In some further cases of tests at this temperature, a click was heard but the fracture did not propagate. In these cases, a yield zone was seen which indicated a strain distribution far from that of the elastic type. An example is shown in Figure 7, for a sample in which the initial increment of fracture occurred at an average net section stress of about 36,000 psi. A noticeable thickness reduction accompanied these yield zones, and this reflects the closer approach of these thinner specimens to a plane stress condition under low strain rate loading. If the load was increased until the specimen broke, the running fracture was a very brittle one. It is not yet known whether the non-propagation of some of these starting increments of fracture is connected with the formation of these yield zones, or whether the low hardness developed on nitriding caused the starting increments of rupture to be rather ductile and not as close to a sharp crack-like increment. Further tests are planned.

Pre-Cracking in Fatigue

The usual technique for measuring fracture toughness in high strength materials is to test notched specimens pre-cracked in fatigue. With this method of preparation on low strength steels, some variation in breaking strength would be expected in the brittle temperature range on the basis of the mechanism of fracture suggested herein. In order to check that the material and test conditions described would give approximately yield point strength in the presence of a fatigue crack, two specimens of 3/4 in. thick material from the second lot of welded plate (p. 11) were pre-cracked by fatigue in bending. These 4 in. wide specimens were the same as those used for the tests described except that a fatigue crack approximately 1/8 in. deep was grown from the base of the zirconium weld deposit. The total notch depth was then about 5/8 in.

One specimen broke at -76°C with an average stress on the net section of 39,000 psi, while the other broke at -72°C at an average stress on the net section of 43,000 psi. These breaking stresses may be contrasted with the value of 16,400 psi on the net section at which one specimen from this plate broke at -67°C. The lower yield strength was 46,000 psi at -75°C and 36,000 psi at room temperature.

It may be mentioned also that eccentricity introduced by specimen misalignments does not account for the low stress fractures. One further 3/4 in. thick specimen from the same plate was pre-cracked in fatigue so that the total notch depth was 5/8 in. The specimen was then misaligned in the tension grips so that the upper grip was about 5/8 in. off centre with the force on the notched side of the specimen. The combined eccentricity was several times as much as that which could be expected in the other tests. The average stress on the net section at failure at -74°C was 36,000 psi.

3. CONCLUSION

The experimental results have been consistent with the theory and have shown, with a high degree of freedom from ambiguity, that fractures under low stress and low strain rate conditions can occur, under exceptional circumstances, from small defects if the material shows dynamic nil-ductility behaviour.

4. ACKNOWLEDGEMENTS

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LPT:(PES):KW

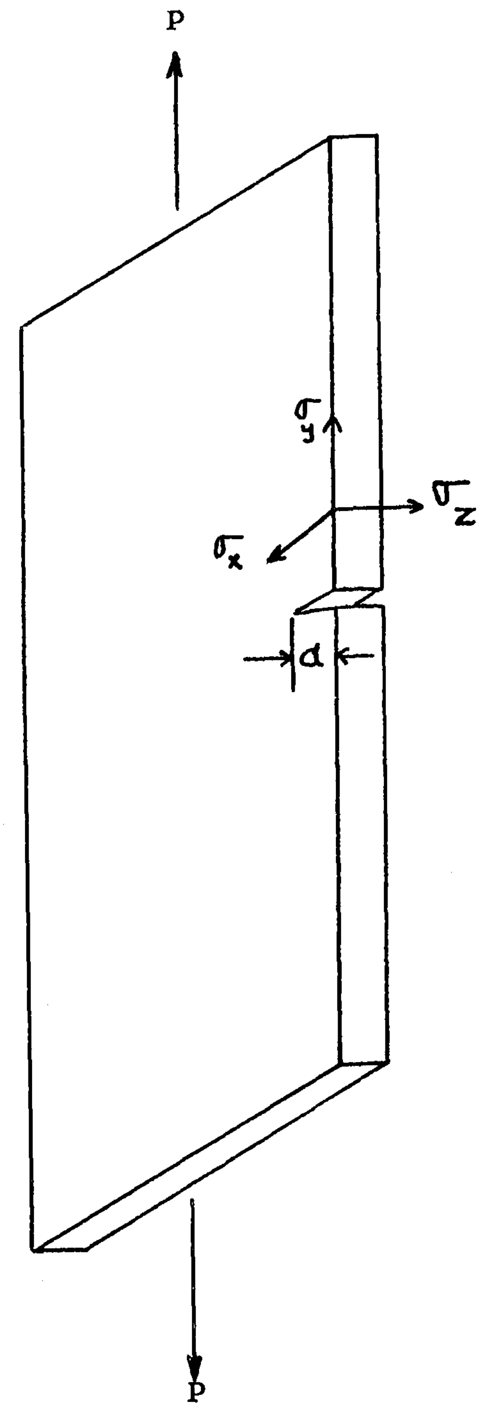
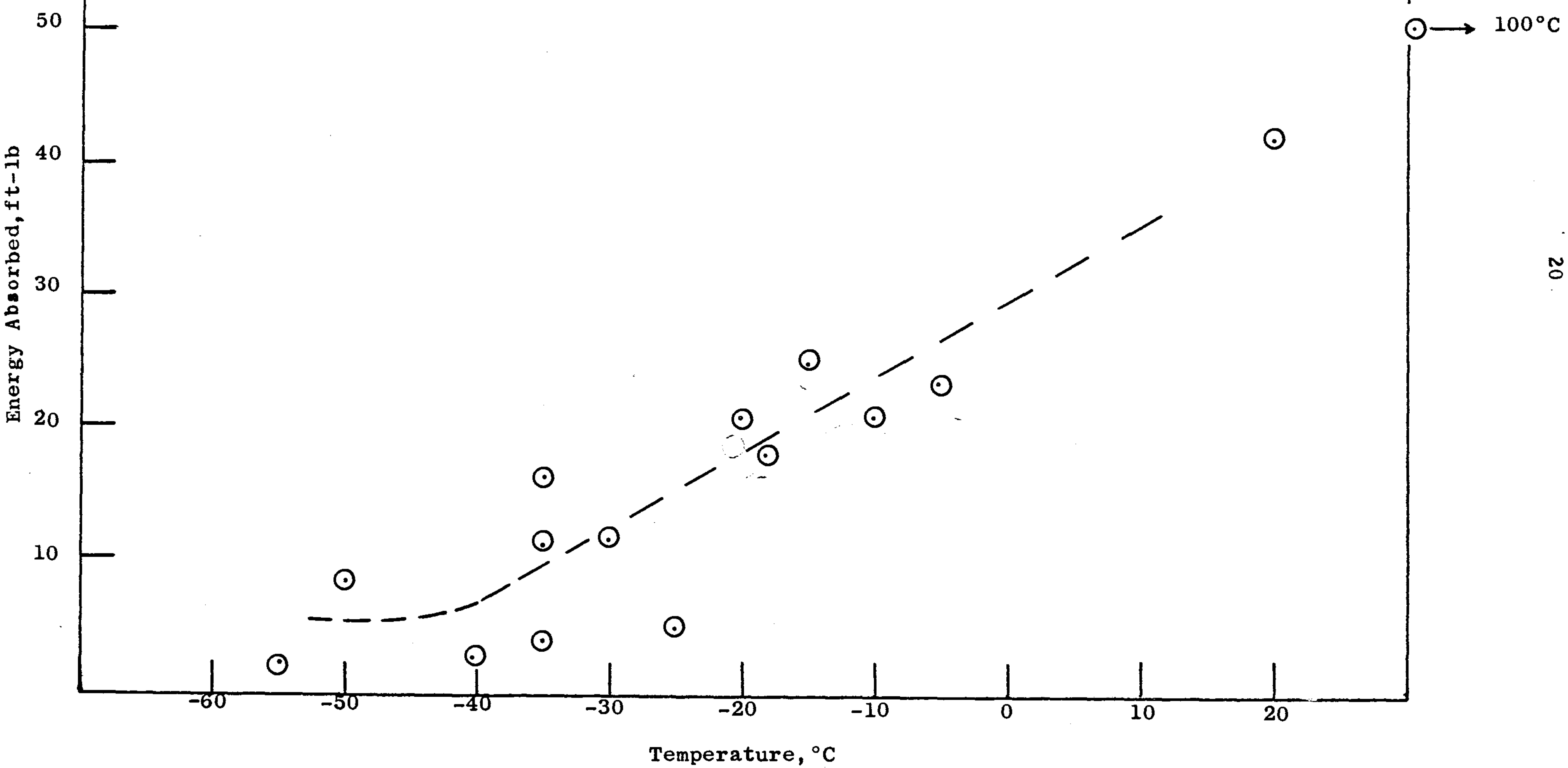


Figure 1. Structure under a load "P" with edge crack of depth "a".

Figure 2. Brittle fractures obtained in mild steel plate, using weld-filler starter. Tensile test temperatures, -14°C and -73°C respectively.
Mag. 1.5 X

Figure 3. Profile of start of fracture for specimen in Figure 1. Tested at -73°C.
Mag. 100 X

Figure 4. Charpy V-Notch Results for Hot-Rolled 3/4 Inch Thick Plate of 0.22 Carbon Structural Steel. The fracture plane of the specimens was the same as that for the notched tensile tests.



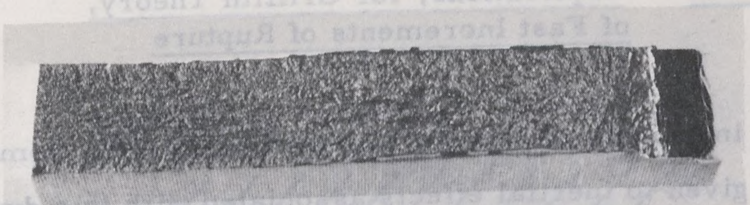


Figure 5. Fracture face of specimen tested at -15°C. Shear lips are evident. Mag. 0.9X

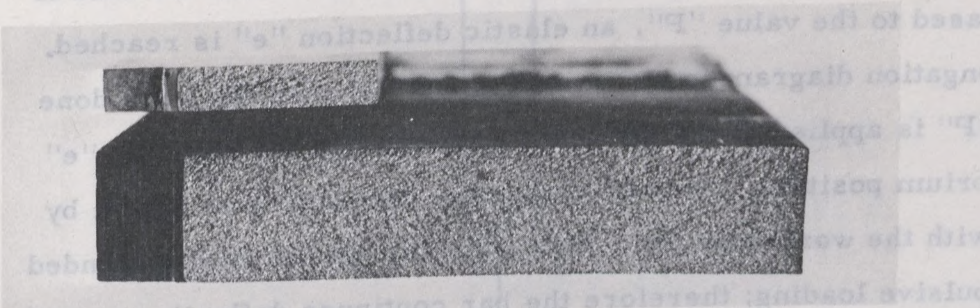


Figure 6. Fracture faces of nitrided specimens broken at -75°C. Mag. 0.94X

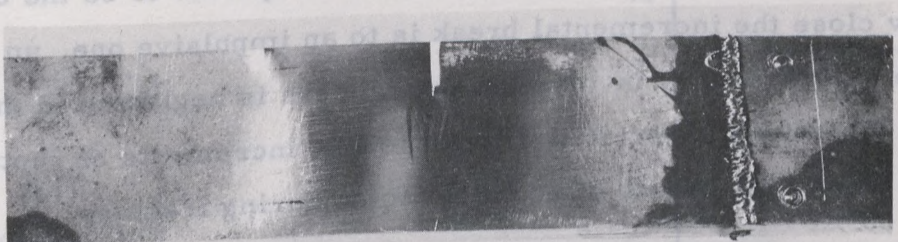


Figure 7. Irregular yield zone at root of notch of 1/4 in. thick specimen tested at -75°C.

APPENDIX - Implications, for Griffith Theory, of Fast Increments of Rupture

Since a fast increment of rupture has been postulated, some consideration should be given to inertial effects associated with this dynamic process. It is a standard result in first order vibration theory that a suddenly applied load can cause twice the deflection of the same load applied slowly. A simple interpretation of this result is helpful. Consider an elastic bar loaded at the bottom as shown in Figure 8. When the load is slowly increased to the value "P", an elastic deflection "e" is reached. The load-elongation diagram is shown in Figure 8(a) with the work done 1/2 Pe. If "P" is applied at the start and acts through the distance "e" to the equilibrium position, the load-elongation diagram is as shown by Figure 8(b) with the work done Pe. Twice the energy has been expended with this impulsive loading; therefore the bar continues deflecting past the equilibrium position and vibrates up and down about this position.

Consider this process, now, in relation to a fast increment of rupture. G. R. Irwin[7] has shown that, in the Griffith theory, the energy balance corresponds to that for a process in equilibrium at all times. When a fast increment of rupture occurs, this would not be expected to be the case. Depending on how close the incremental break is to an impulsive one, up to one-half of the elastic energy involved in the process is neglected in the Griffith theory. Vibrations associated with these fast increments of rupture presumably give rise to the bursts of sound detected during fracture[14].

Since the crack opening displacement is linearly proportional to K, the stress on the adjacent metal will be expected to increase above its equilibrium value and decrease below it as the faces of the crack opening oscillate through their equilibrium positions. If the stress field results from a defect of the order of 1/4 inch deep, the perturbation of this field from a fast increment of rupture over a distance of the order of $\frac{1}{1000}$ inch is presumably small but may be important in some cases.

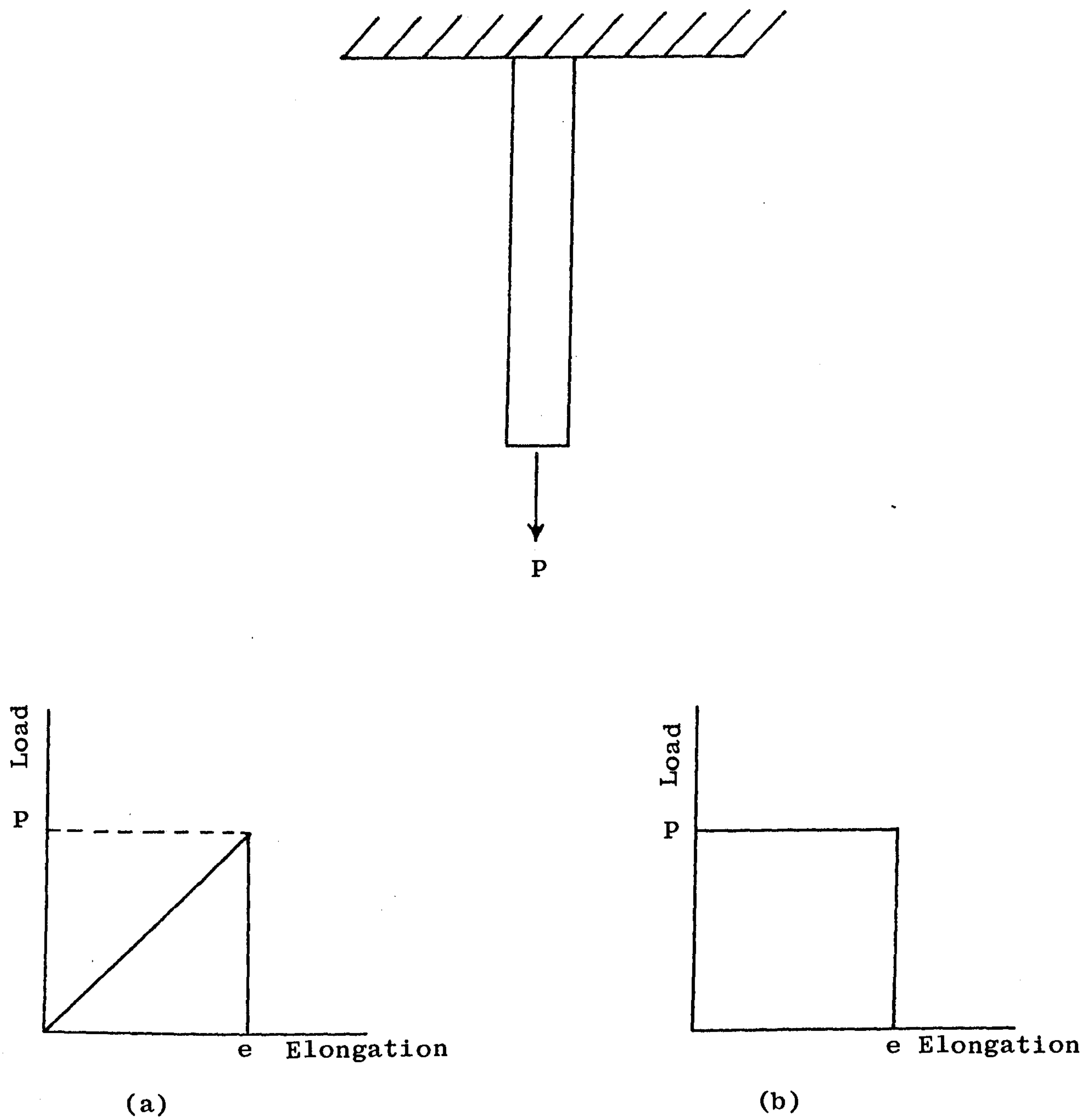


Figure 8. Load-elongation diagrams for elastic bar loaded slowly (a), and impulsively (b), to the load "P" corresponding to an equilibrium elongation "e".