SOME ASPECTS OF THE FLOW AND FRACTURE OF METALS IN TEST AND SERVICE

By

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If we leave on one side these service failures into which chemical factors, wear or accidental damage have entered, with consequent reduction of section and heightening of stressing, the circumstances leading to failure can fairly be summarised by the truism that any component which breaks in service does so because it has been subjected to unsuitable and excessive stressing. It is implied that components which fail are unsuitably dimensioned to resist the service loading. On this basis engineering design would appear to be primarily a matter of making satisfactory use of a selected material of construction in the light of the known mechanical properties.

It would, however, be an oversimplification of the position to represent that an adequate foundation for design would be provided by knowledge of the "specification properties" of the material (hardness, tensile strength and elongation, notch impact value, fatigue limit, etc.). If this were true, service failures would virtually be limited to defective components and to components made from materials purchased without recourse to engineering specifications.

Useful as knowledge of the specification properties undoubtedly is, it clearly does not provide a guarantee against failure in service of parts designed without the benefit of previous user experience. This state of affairs is sometimes ascribed to the complexity of stressing in service. It is argued that, while service fractures can be classified into failures understatic stress, impact and fatigue (with much the largest number in the last category), the service stress conditions are more complex than those in the tensile, impact and fatigue tests normally carried out in the the laboratory. There is much truth in this, but, on the other hand, examination of the conditions in "simple" laboratory tests shows the stressing at fracture to be complex. The standard tests are therefore capable of revealing more useful and significant information than is normally derived from them and a full appreciation of the mechanism of flow and fracture in these basic tests is of the utmost value to the design engineer.

This opens up a vast subject for review, which cannot be dealt with at all completely in the presentpaper. In what follows a summary is given of information on flow phenomena in general and results are presented from tensile, impact and creep tests on cubic metals, a division of the field with which the author has been in close contact.

FLOW OF CUBIC METALS.

In engineering calculations, metals are treated on the basis of elastic theory, which concerns itself not with metallic structures but with an elastic and isotropic continuum, in which a simple load distributes itself uniformly over the section normal to its direction of application, so that the value of force / area is constant whatever area is considered—including the infinitesimal limit —and that the value of the resulting elastic strains is independent of the direction of stressing. Metals, however, consist of crystals and within crystals atoms are arranged in crystallographic planes, and within the planes in rows. Since an applied load must be carried by the rows of atoms, the concept of stress as the infinitesimal limit of force over area is not strictly valid. In the absence of any workable alternative, we shall have, however, to continue to use the concept of stress, but with a realisation of its shortcomings.

The metal crystal has different elastic properties in different directions[1], even when it belongs to the cubic, so-called isotropic system. In the cubic metals aluminium, copper and

α-iron, the limiting values of Young's Modulus (stress for unit elastic strain) for different directions in the crystal vary between :—

	Maximum.	Minimum.
Aluminium	4900	4060 tons/sq. in
Copper	12400	4340
α—iron	18400	8550

the (intermediate) values for (intermediate) directions with Miller indices (hkl) being given by E in the formula :—

$$1/E = 1/E_{min} - 3(1/E_{min} - 1/E_{max}) \,.F$$

Where $F = (h^2 k^2 + k^2 l^2 + l^2 h^2) \div (h^2 + k^2 + l^2)^2$

Thus the different directions in a crystal behave elastically very differently and, according to the orientation of the crystals in a uniformly stressed test piece, individual regions reach given stages of deformation at quite different stress levels, and in particular they reach the elastic limit at which plastic deformation begins—at different stress levels.

The usual mechanism for plastic deformation of single crystals of cubic metals is by slip—although for body-centred types there is also the possibility of mechanical twinning. As an extreme case of slip there is shear displacement by cracking—shear fracture. There is possible also is some circumstances fracture by cleavage, about which more will be said later. In face-centred cubic metals (e.g. nickel, copper, aluminium,) slip occurs only along the face-diagonals of octahedral planes. There are 4 such (111) sets of planes and 3 (110) directions in each, given 12 possible orientations of slip. The actual direction along which a crystal will slip when its elastic limit is reached is determined by the directionof stressing, the slip direction being that one of the possible directions for which the resolved shear stress is a maximum. In body-centred cubic metals, (e.g., α-iron, tungsten) the direction of slip is always (111) but the slip plane is not always of one type. Tungsten, molybdenum and sodium slip on (112) planes up to a certain temperature and then on (110) planes, but for α-iron it is possible for slip at room temperature to take place on any of three sorts of planes, (110), (112) or (123), depending on which contains the direction which most nearly satisfies the resolved shear stress law. This multiplicity of sets of slip planes means that, in iron, slip or shear-cracking which starts in one crystal finds it much more easy to carry on into the next without significant change of direction than in the other body-centred cubic metals, or than in the face-centred cubic metals, which offer fewer available slip directions. As a consequence of this we may not be on sound ground if we try to apply to other metals (and particularly those which are not body-centred cubic in structure) impressions formed from iron.

In connection with metals, as we usually know them, as polycrystalline aggregates, the term "statistically isotropic" is often met, but this is correctly applied only to certain large-scale phenomena where the unit under consideration is sufficiently large in relation to the grain-size of the metal, and the grain orientation is sufficiently random, that the external strain is what would be expected in an isotropic continuum, similarly stressed. The concept is moreover, one which can often not profitably be applied. For instance, statistical isotropy is rarely present in wrought metals on whatever scale they are considered, because of the preferred orientations taken up by the crystals and of the directional elongation of non-metallic inclusions. Moreover, it can never be assumed when the scale of the phenomena is small relative to the grain-size.

Chalmers[2] has made experiments on tensile test pieces of tin, each of which consisted of two crystals separated by a boundary running as a plane along the axis. Conditions of growth had been so controlled that in each specimen the two crystals made equal angles with the boundary, but that the angles varied from 0 to 90 degrees from specimen to specimen. The contribution of the grain boundary as such was clearly equal in all the specimens, but when Chalmers measured the Proof Stress of his test pieces this was found nevertheless to vary widely, from 400 to 650 gm/mm^2 as the angle changed from 0 to 90 degrees, showing that an important contribution was being made by the reaction of the one crystal on the other in an effort to preserve conformity at the grain-boundary.

More recent experiments by Boas and Hargreaves[3] have given an even more striking demonstration of this sort of effect. The material under examination was coarse-grained pure aluminium. Before pulling the test piece, each grain had a uniform hardness. The hardness and elongation resulting from deformation varied widely from crystal to crystal and within individual crystals. A grain for which the average deformation or hardening was less than that of its neighbour deformed more at the edge than the centre, and vice versa. It is easy from such results to see that, when a polycrystalline metal is deformed, the overall strain on a large gauge length may not be a very accurate indications of the strain on a microscopic scale and that areas will exist within the piece where elastic deformation is still proceeding while other area have yielded plastically. Consequently when a tensile test piece of a ductile metal is taken beyond its elastic limit and unloaded, the highly elastically deformed regions pull against the remainder of the structure and place the weaker areas under a stress of reversed sign. Well-known effects arising from this are a slow change of dimensions on standing unloaded-Elastic after-working-and a reduced yield point in compression—the Bauschinger effect.

FLOW AND FRACTURE IN THE TENSILE TEST.

Tensile load/extension diagrams take on a typical form for all ductile metals. First there is the straight line defining the elastic response, with a slope equal to Young's Modulus, then there is a curving portion leading from the elastic limit to the maximum load, with or without a sharply defined yield point, and finally in a ductile metal there is a falling portion leading from the maximum load to the point of fracture and defining the stage in which "necking" of the test piece takes place. By setting down instead of a load scale a scale of load divided by original area, this diagram is converted into a "nominal stress"/elongation diagram. The diagram plotted in this way serves its purpose as a testing house record, but would only be directly applicable to the design of components which were to be subjected in service to steady longitudinal loading.

Stress/strain diagrams plotted in this way thus convey only limited information. Alternative methods of reporting the results of tensile tests have been widely recommended, but less widely used. In these, "true stress' is substituted for the nominal stress, "true stress" being the average tensile load per unit acting area, as distinct from load per unit original area. To asses this figure, area measurements are necessary at each loading stage of the test. Beyond the point of necking, the minimum area of the neck is taken. It is better to plot the "true stress" against reduction of area than against elongation, since exception can be taken to the elongation figures on the ground that an ever-changing length of material contributes to the extension of the test piece. A refinement of this method seeks to associate the "true stress" with "true strain", obtained by plotting the reduction of area on a logarithmic scale, to take out the "compound interest effect". When plotted in this way stress-strain diagrams take the form of a gently rising and slightly curved line." [4]

Figure 1 shows typical nominal stress/strain and "true stress-strain" diagrams for two rather different non-ferrous materials, pure nickel in a soft condition and Nimonic 80, which is an age-hardened 80-20 nickel chromium alloy containing approxiately $2\frac{1}{4}\%$ titanium and 1% aluminium. Reading from the left hand curves, the pure nickel has an elastic limit of about 5 tons/sq. in. and a tensile strength of 29 tons/sq. in. with a high uniform elongation of 39% and an elongation of 60%at fracture. According to micrometer measurements on the necked test-piece the reduction of area at fracture was 81% but the load supported immediately before fracture was extremely small, corresponding to a nominal stress of a fraction of 1 ton/sq. in. The Nimonic 80 has an elastic limit of about 40 tons-sq. in. and a tensile strength of 71 tons-sq. in., again with a high uniform elongation, of 35%, but as would be expected for a much harder material the elongation at fracture is down, at 42%. The nominal stress at fracture is, however, 63 tons/sq. in.

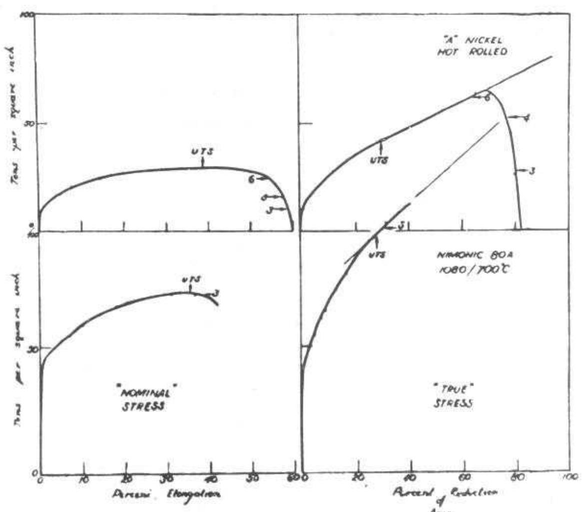

Fig. 1. Nominal Stress/Strain and "True Strees"/ Strain Diagrams for nickel and Nimonic 80.

When plotted on the "true stress" reduction of area basis as shown on the right of the Figure, the stress/strain curves naturally are considerably altered in shape, though the points already made can in the main still be appreciated. An additional point which is brought out is that the curves progressively decrease in gradient until the point corresponding to the tensile strength, but from then remain constant in slope over an extended range (for the nickel) or until fracture(for the Nimonic 80). This is of course no new observation and for some years it has been believed that the slope of this straight part of the curve is an index of the work-hardening of the metal under test.[5] The Nimonic 80 appears to work-harden almost twice as fast as pure nickel, a result which can readily be accepted.

Taking the curves at their face value, it would seem that Nimonic 80 continues to work-harden up to the point of fracture but that when necking in nickel has proceeded so far, hardening ceases and the stress supportable by the material decreases rapidly, i.e. the nickel softens rapidly. This does not appear to be a very probable construction. The situation was explored further by making additional tests as indicated by the numbers against the arrows on the curves, and stopping the straining at various stages between necking and fracture, afterwards examining sections through the specimens so treated to look for internal cracking, as had been done earlier by a variety of workers on other materials, including Timoshenko[6], Parker, Davis and Flanigan[7] and Pumphrey's[8] .

For the Nimonic 80 there was no sign of internal cracking in the specimen stopped at "Stage 3". With the pure nickel, on the other hand, very extensive cracking was present at "Stage 3", and extensive cracking at "Stage 4", though no cracking had occurred at "Stage 6". Figure 2 shows the general appearance of the fractured test-piece of Nimonic 80 and Figure 3 roughly sectioned nickel bars from "Stages 3" and "4". It is noted that the internal crack at "Stage 3" occupies so much of the area of the neck that the shaping tool cut into it when it had gone down only 0.015 in. from the surface. Assuming the crack to be symmetrically disposed within the bar there would be at "Stage 3" a cross sectional area of about 0.016 sq. in., carrying a load of 2.5 tons, that is to say an average tensile stress of approximately 150 tons/sq. in., compared with nominal stress of less than 1 ton/sq. in. Figure 4 shows the "Stage 3" fissure on on a polished section, and Figure 5 the appearance on the section of the end of the fissure, which is near the surface of the neck. The end of the crack is extremely sharp.

It is evident that the fractures of both the Nimonic 80 and the nickel specimens are variants on the commonly occurring "cup and cone" type, in which the fracture surface is composed of a central, rugged crack which runs into a smooth conical surface of macroscopic shear running either downward or upward from the periphery. In the Nimonic 80 specimen the shear surface was "eared", i. e. contained parts of both the upward and downward cones. The mode of failure of the specimens of nickel differed from that of normal cup and cone fractures only in as much as the central fissure opened out widely and continued to extend sideways until the wall of the necked-portion was extremely thin before final fracture occurred. Evidently the stress concentration at the end of the internal fissure became such that it was easier for the rugged crack to continue than for macroscopic shear to occur in the wall of the test piece. The section of a fully fractured soft nickel bar is shown in Figure 6, the last of the series. From this it appears that there were in fact two rings, or a spiral, of almost infinitesimal section carrying the final breaking load, which as stated earlier was only a small fraction of 1 ton. Similar fractures have recently been described by Pumphrey [8] but his paper gives the impression that they are found only in exceptional circumstances, usually at temperatures near the melting point of the metal under test.

In connection with this fractre, it is of interest to note that a careful micro-hardness survey on a longitudinal section through the fracture did not indicate any significant difference in the extent of work hardening from centre to outside. This was felt to emphasise the impression that cracking started at a very late stage of test. If a crack had been present in the centre of the section from an early stage, e.g. when necking started, the area adjacent to that crack would clearly have a hardness value little over half that near the final crack, since hardness is proportional to strength.

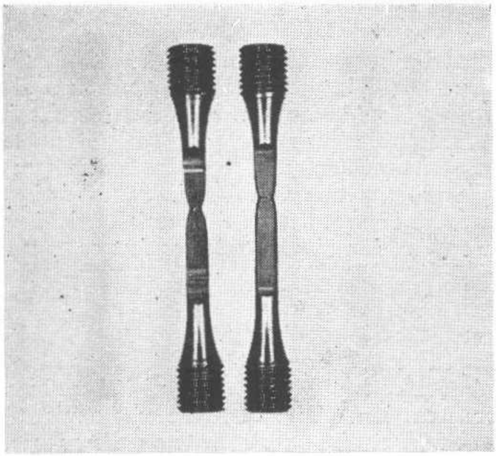

FIG. 2. Fractured tensile test-piece of imonic 80.

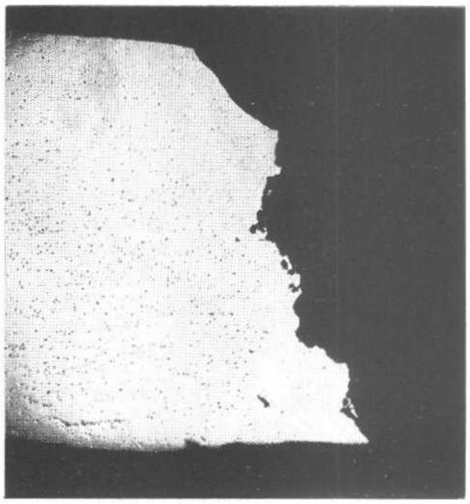

FIG. 3. Roughly sectioned bars of nickel from stages 3 and 4.
(See Figure 2).

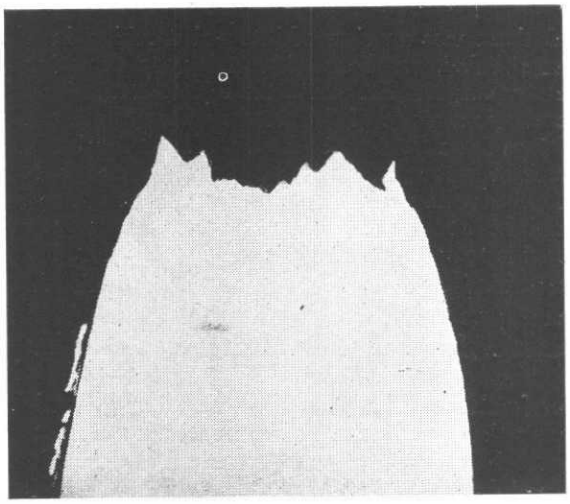

FIG. 6. Fully fractured nickel bar.

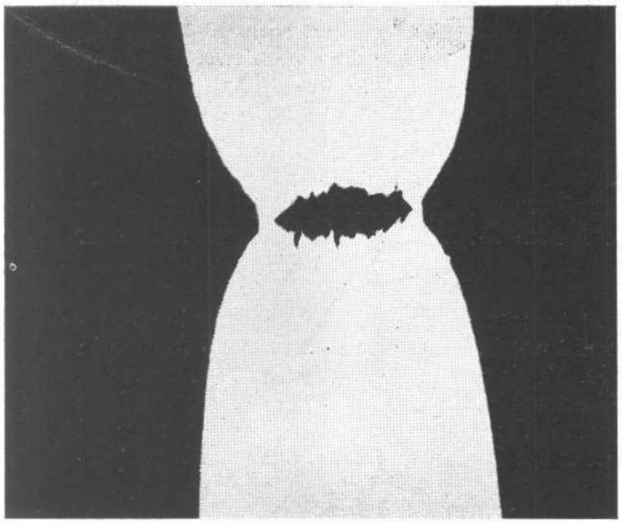

Fig. 4. Polished Section of nickel bar from stage 3.

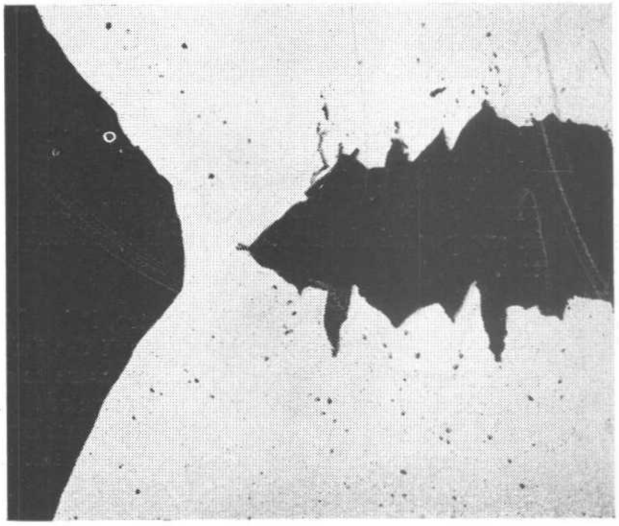

Fig. 5. End of fissure shown in Figure.

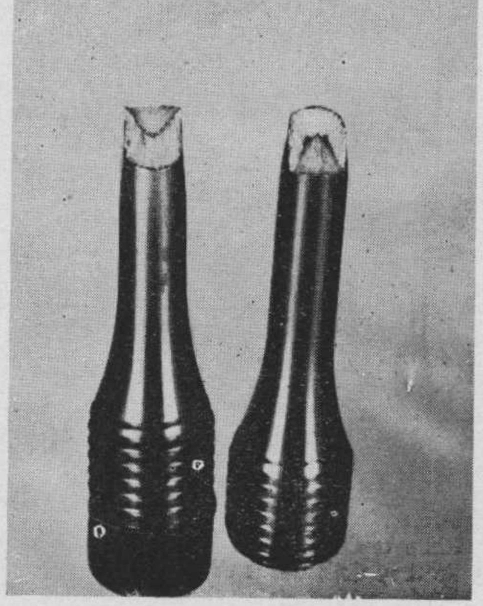

Fig. 7. Fractured tensile test-piece of En 25 steel. U.T.S. 124.7 tons/sq. in.

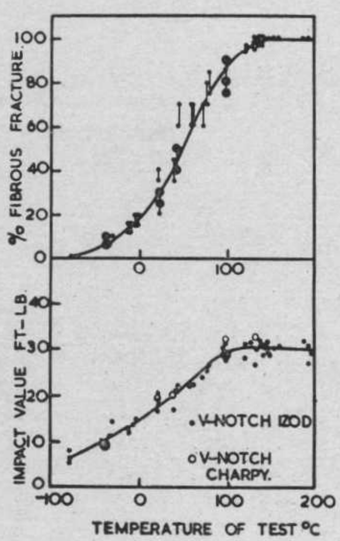

Fig. 8. Typical impact transition curves for low-alloy steel[10]

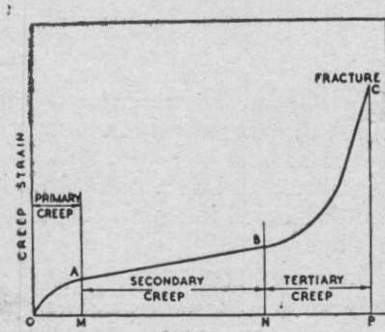

Fig. 9. Schematic creep curve.

Figure 7 shows the fractured surface of a further tensile test piece, this time of an En 25 ($2\frac{1}{2}$ percent nickel-chromium-molybdenum) steel, in an oil-hardened and lightly tempered condition, giving a tensile strength of 124.7 tons/sq. in., with 12% elongation and 28% reduction of area. It is evident that in this hard steel the central crack of the "cup and cone" had only to advance a very short distance before the conditions for macroscopic shear arose. Here, as in the case of the Nimonic 80 specimen, there was no detectable drop in the "true stress"/strain curve before final fracture.

All three results can be regarded as giving considerable support to the validity of the "true stress"/strain diagram as a means of representing the course of events leading to fracture. Such a diagram can, however, indicate only the mean value across the acting section of the tensile stress; we know that once necking has taken place the tensile stress is not evenly distributed and that significant stresses occur in radial and circumferential directions.

An important contribution to our knowledge of tensile fracture was made a few years ago by Parker, Davis and Flanigan[7]. These authors set out to determine the stress distribution in the necks of tensile test-pieces which fail with cup and cone fractures and to confirm whether internal fissures were cleavage cracks, as had been suggested by Timoshenko[6], or are composed of crystallographic shear cracks. Both points were examined by stopping off tensile tests on mild steel specimens between necking and fracture. To arrive at the stresses in the neck they first made calculations of the stress distribution due to the neck form (along the lines of Neuber in "Kerbspannungslehre") and then added on the residual stresses indicated by strain gauges attached to the test-piece while they drilled out the centres progressively. In their computation they ignored the Bauschinger effect, which makes their calculated stress gradients lower than the actual ones. They found that the longitudinal stress remains uniform across the section and there are no radial or circumferential stresses until necking begins. Beyond this point, the variation in longitudinal stress and the magnitude and variation of the two other stresses increase progressively until fracture occurs. At the stage of fracture, the average longitudinal stress the "true stress"—was 60 tons/sq. in., but the true maximum longitudinal stress was almost half as high again; 85 tons sq. in. This high level of longitudinal stress was found at the axis of the bar and did not change appreciably in the first fifth of the radius. In the next two fifths the longitudinal stress was shown to fall to the average value, in the next it falls to a minimum a little below the average, a value to which it returned at the surface. The radial and circumferential stresses varied from 0 at the surface to 23 tons/sq. in. at the centre. Somewhat similar results have been obtained by Davidenkov and Spiridonova[9] for Armco iron.

Parker, Davis and Flanigan[7] checked the fracture habit in their mild steel by comparing the direction of the cracking with the orientation of etch pits. They found that at room temperature the fractures even in the interior of the bar were by-shear cracking, but at sub-zero testing temperatures the habit changed to cleavage cracking. At liquid air temperature the whole of the fracture was of the cleavage type.

Impact Fracture.

For research purposes, tensile tests are often made on (notched) specimens, where the speed of relative movement of the two ends of the test piece is sufficiently high for the tests to be described as 'impact tensile tests". Such tests have supplied valuable information but most of our experience and data come from impact-bending tests. In British practice, where impact tests are demanded in standard specifications for hardened and tempered steels, the testing is carried out by means of the Izod machine. In the Izod test a square bar 75 mm. long, 10 mm. across the flats, is provided with a 45° notch 2 mm. deep, root radius 0.25 mm., 28 mm, from one end is positioned as a cantilever in the vise of the machine, where it is struck near the free end by a knife edge on a pendulum with a striking energy of 120 ft. lbs. Experience has shown that a round bar may be substituted for a sqauare bar and the mode of support and position of striking may be varied without significant alteration of the absorbed energy. The test results are, however, very liable to alteration if the notch-form, the width of the test piece, the testing temperature or the striking velocity are varied. The energy absorbed in facturing an Izod test piece of a steel is a clear indication of whether it is "ductile" or "brittle", but an elaborate programme of testing under different conditions (varying one or more of the factors referred to in the last sentence) is needed to show whether the distinction is essentially a matter of composition, heat-treatment or of the testing conditions themselves.

It is clear that in this connection "brittleness" is by no means the converse of "strength :" "brittleness" means a small absorption of energy to fracture, i.e. a small extent of plastic deformation of the test-piece-a situation which is often associated with high strength. We can have both "inherent" and "derived" brittleness, the latter being due to either (a) the effect of stress concentrations and constraints in limiting the "effective gauge length" or volume of metal which suffers the significant deformation or to (b) the effects of second or third principal stresses which, having the same sign as the main tension, reduce the possibility of slip by reducing the shear stresses. As a limit of this, there is the positon where there are three equal tensions (perfect triaxility or hydrostatic tension) in which theoretically failure is impossible and deformation cannot take place.

The notch in an Izod testpiece applies a pronounced degree of triaxility, but one which cannot be defined independent of the material of the test piece. The Charpy "key-hole" notch is a less severe source of triaxility. In view of this situation and the fact that existing notch impact test specifications do not define testing temperatures, it appears a matter of considerable importance to examine the implications of impact testing in the broadest possible way.

It has already been pointed out that the mode of fracture of tensile test-pieces of steel is liable to change from shear at ordinary temperatures to cleavage at low temperatures. Qualitatively, the same is generally true of notch impact fractures though here we have to be prepared for the alternative of the low temperature fracture being intercrystalline, in severe cases of temper brittleness. Figure 8 summarises a typical set of impact test results on a hardened and tempered low-alloy steel [10], actually a steel with 0.29% carbon, 1.25% nickel and 0.67% chromium oil-quenched from 870°C and tempered at 425°C to a hardness of 399 D.P.N. It will be seen that Vee-notch tests were made on both Izod and Charpy machines at temperatures from —80° C to + 200°C and that, in addition to recording the energy required to cause fracture, observation was made of the characteristics of the fracture surfaces. At temperatures from 100°C upwards, where the impact value was around 30 ft. lbs. the fracture was entirely of the "fine-grained" or "fibrous" shear type. Below this temperature the proportion of the fracture which was of this type steadily decreased until at —80° C the whole fracture was of the "cleavage" type, showing coarsely developed bright facets, and the impact value had dropped to about 7 ft. lbs. The range of temperatures in which the steel is in course of "transition" from a ductile to a brittle condition is clearly defined. In the case illustrated, it is an extended range; the transition would be much steeper for plain carbon steels, and for softer steels (whether produced by heavy tempering or by normalising, and especially the latter) and would also be steepened by increasing the sharpness of the notch. Using the mean point, or point of inflexion of the transition range, to define the "transition temperature" we may say that all these changes, except increased tempering, would also have the effect of raising the transition temperature.

In general, a lower transition temperature so defined would be expected from the steel the lower the hardness, but it has to be noted (a) that most hardened steels have a more or less clearly defined "Izod-trough" where, in a high hardness range, the transition temperature passes through a maximum as the tempering temperature increases and the hardness falls (b) that a low hardness achieved by slack quenching usually means a high transition temperature and, (c) that many steels are liable to "temper-brittleness," slow-cooling from high tempering temperatures raising the transition temperature.

It is the main function of impact testing to determine the position of the service temperature with respect to a transition curve. If the transition curve can be determined for conditions closely resembling those of service (or such a curve can be inferred by bracketing methods) the results will clearly have their optimum significance. Experience of unexpected service failures, particularly those in the plating of welded ships, has drawn much attention to the need for development of more searching tests than the Izod test[11], i.e. for notch conditions which are more severe, and atleast one attempt has been made, by Schnadt, to devise a system of testing which would expose the response of a given steel to all possible conditions of stress complexity[12]. In this connection it is important to note that chromium[13] and molybdenum[14] exhibit similar transition phenomena to ferritic steels (and it can be assumed that all body-centred cubic materials share this feature) but no evidence of a brittle transition has ever been found in face-centred cubic metals or alloys. While stress complexity is important, it is thus not quite as essential a factor in impact brittleness as the existence of a mechanism of deformation and fracture alternative to shear. Schnadt's development of his system seems to ignore this fact ; his testing procedure may nevertheless be informative. In the author's opinion, however, the best index at present available of the suitability of a material for service in conditions more severe than those of he Izod test may well be the margin of temperature by which the service temperature exceeds the transition temperature of the material in that test.

An aspect of service failure to which little attention has yet been given is the final stage of of fracture is a component subjected to fatigue. In such a componet, the creeping fatigue crack, with its characteristic conchoidal markings, proceeds part may through the material and then sudden fracture of the weakened section occurs. In steels this "snatch fracture" is often coarsely faceted : hence the old view of fatigue—that the material had "crystallised." Assuming this fianal crack to be an impact fracture, it seems logical to speculate that the extent to which the creeping crack has to penetrate the section before final fracture occurs is a function of the impact properties of the steel. Where the steel is operating well above its transition temperature the service life should thus be much longer than when the service temperature is below the transition temperature, and final fracture of fatiguing shafts in vehicles should be a more frequent occurrence in cold weather than hot. Practical data on these points are, unfortunately, lacking.

CREEP FAILURES.

Possibilities of flow and fracture along other lines than those so far discussed arise in normal metals at elevated temperatures and in soft metals at room temperature. Bengough[15] said in 1912; "It has always appeared a remarkable fact that the facture of metals and alloys tends to pass through the body of the metallic crystals rather than between the crystalline junctions, especially in the case of pure metals and solid solutions. It would seem that at the ordinary temperature the coheson between the faces of separate crystals is greater than that between different parts of the same crystal." Bengough drew attention to a change in fracture habit from transcrytalline to intercrystalline which in ordinary tensile tests commonally takes place when the temperature reaches two thirds of the melting point, which he refered to as the temperature of recuperation and others since have called the equicohesive temperature. This temperature had the value of 700° C for a copper-nickel alloy studied by Bengough. Less rapid tensile tests or constant load creep tests would have shown intercrystalline fracture starting at considerably lower temperatures.

Andrade[16] was the first to make a systematic investigation of creep in a variety of alloys over a wide range of stresses and temperatures. As is well-known, he found hat flow at constant stress consists of two parts, a transient part progressing as the cube root of the time and a quasi-viscous part progressing exponentially with time, the change in specimen length being covered by the formula :—

$$l_t = l_0(1+\beta t^{\frac{1}{3}}) \text{ exp. } kt.$$

The same law applied to several metals and gave the impression that there was little inherent difference between what we normally regard as soft metals and hard metals except in the temperature at which they exist relative to their internal temperature scale. However, there are two constants in Andrade's formula and is difficult to secure a match at the same times in both the transient and the quasi-viscous parts of creep curve.

Andrade's experiments were carried out at constant average stress, but creep tests are more usually made at constant load. This means that if the specimens elongate considerably, the later part of the curve is tilted upwards. Normally creep curves take on the general form shown in

Figure 9. This curve consists of an initial immediate extension (partly elastic) followed by the primary stage of decreasing creep rate, a secondary stage where the creep rate is constant and a tertiary creep stage of increasing creep rate[17]. At one time, there used to be argument as to whether there were really three stages of creep or only two, the minimum rate being regarded merely as a point of inflexion[18]. There have however now been published several curves with constant rate of extension for periods of 1000 hours or more[19].

Onset of tertiary creep is recognised as the point beyond which service should never be intentionally prolonged, since throughout the tertiary stage creep-cracks are liable to appear and spread. In applications where fatigue failure is possible as a result of fluctuating stress superimposed on the steady creep stress—as in a gas turbine blade—such cracks are properly looked upon as extremely serious. The diagram in Figure 8 may give the impression that imminence of failure is always clearly signalled by a stage of rapid extension and that therefore creep failure should be easy to avoid in practice. This may not, however, always be the case. Sometimes a tertiary creep stage may occur but no opportunity exists of it being noticed; again, there are materials in which extension in tertiary creep is limited and they appear to fail in the secondary stage.

The view expressed above with regard to the necessity of not entering tertiary creep in service, means that the relative merits of two alloys should never be judged by their capacity for extension in tertiary creep, since this stage should never enter into service. A much more real index of the ductility of materials in the creep range is the elongation at the onset of tertiary creep. This, like total extension at fracture, tends to diminish with decrease of load and therefore with increase of life to onset of tertiary creep. The result is that while high extensions in secondary creep may be shown by some soft metals when failing relatively rapidly, the extension which can be expected in materials giving useful lengths of test life under suitable stresses for service application is usually small, and often between 0.1 and 1 per cent.

The situation in his respect for a typical creep-resistant alloy (Nimonic 80A at 650°C) may be inferred from Figure 10[17]. It will be seen that stresses for 0.5% extension in a given period differ little from those causing fracture in the same time, but that a design based on the times to creep 0.1% would have a large time factor of safety, e.g. 15 tons sq. in. could be chosen for a component with a service requirement of 2000 hours life with real certainty that the component would remain unbroken, since the fracture time for that stress is nearly 20,000 hours. Doubts about service temperatures are the biggest bugbear in design at creep temperatures, for the rate of creep is extremely sensitive to temperature change, often doubling or trebling for an increase of 10°C, when the load is kept constant.

Efforts have been made recently to put creep on to a theoretical basis [20], [21], but so far only transient creep has shown any prospect of yielding to the attack, and even here the treatment seems to raise more problems than it explains. In consequence it is more profitable to base ones impressions of the mechanism of creep directly on experimental evidence. Hanson and Wheeler[22] traced the changes in single crystals and polycrystalline specimens of aluminium subjected to creep by making direct observations of the changes taking place at the polished surfaces of the test-pieces. A typical series of results was given by a specimen of medium grain-size which was stressed at 1.4 tons/sq. in. at 250°C. The recorded creep curve was very similar to that shown schematically in Figure 9. During the rapid primary stage slip bands were produced, but no more were generated in the secondary stage, despite considerable extension. When tertiary creep was entered the grain boundaries of the crystals became prominent and broader, while much distortion took place in their neighbourhood leading to inter-crystalline cracking. No slip bands appeared, however, until the section was much reduced by cracking and deformation had become too rapid properly to be regarded as creep. The fracture also changed at this stage to a transcrystalline course.

Figure 11, due to Moore, Betty and Dollins [23], shows some of these effects and, brings out a further mode of deformation in creep, namely rotation of the grains. These workers prepared their specimens, which were of lead, by a method which left a series of parallel longitudinal scratches on

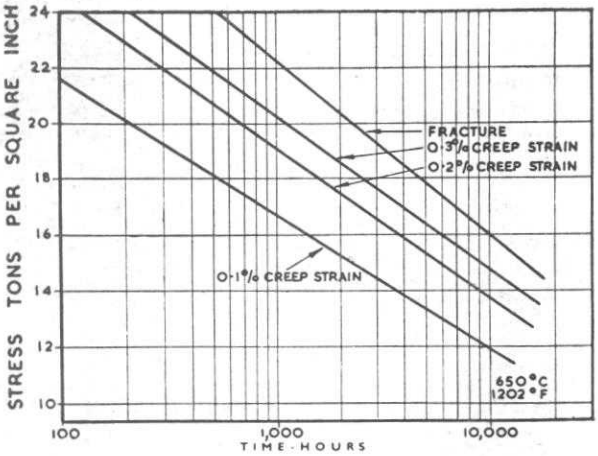

FIG. 10. Creep design data chart for Nimonic 80A.

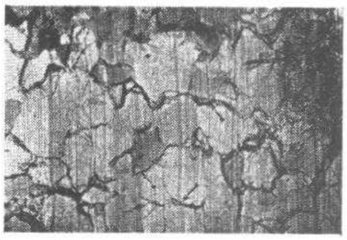

FIG. 11. Creep deformation in lead, after Moore, Betty and Dollins[23]

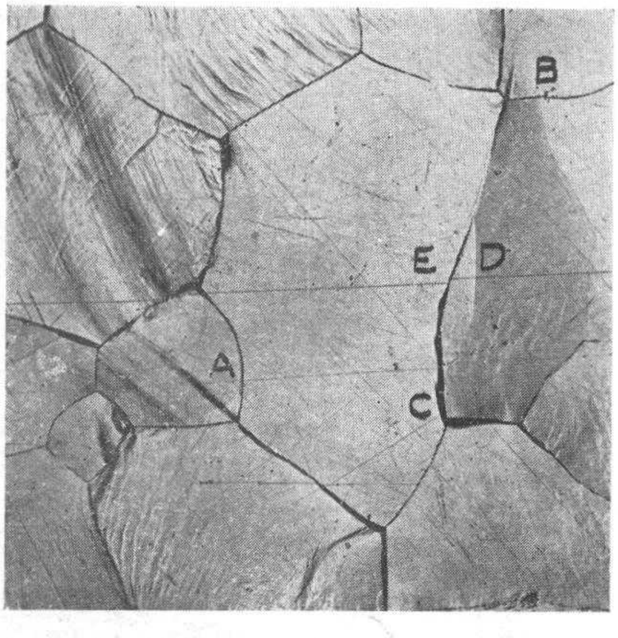

FIG. 12. Creep deformation in tin-antimony, after Betteridge and Franklin[27].

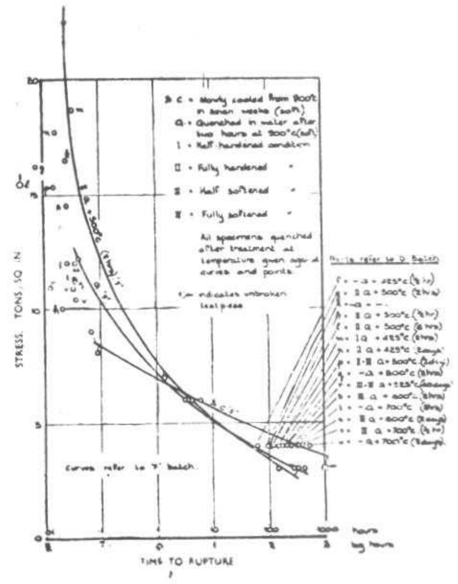

FIG. 13. Creep results on 3% Ni-Si-Cu alloy, after Jenkins, Bucknall and Jenkinson[32].

the polished surface. After creep, with a large degree of extension, these were seen to have turned bodily within various grains and to step sharply at some of the grain boundaries, particularly at the points which have been indicated by arrows. No effort to maintain conformity at the grain boundaries appeared to be in operation. Extension of another specimen rapidly by the same amount at the same temperature caused no grain rotation, so that it is evident that in the high temperature field a given degree of overall deformation can be arrived at by different mechanisms, and no successful generalisations could be based on the idea that a given overall strain means that the same state has been reached by the metal.

Recent work on the mechanism of creep has emphasised the occurence of deformations within the crystals as well as those at the crystal boundaries. Wood and Rachinger [24], Cottrell, and Aytekin [25], and Greenough and Smith [26], amongst others, have drawn attention to the sub-division of crystals into smaller " cells " during creep. In a recent paper, Betteridge and Franklin [27], have dealt with the structural changes in a tin-antimony alloy during creep at room temperature. They observed several different effects, several of which are illustrated in Figure 12. *First* there was rumpling of the surface of grains, varying from relatively sharp parallel lines resembling slip bands to irregular marking. These are seen in the top grain on the slide. *Second*, severe localised strain markings within the grains, usually running from points of intersection of grain boundaries and often continuing the line of one of these, e.g. at A and B. *Third*, thickened grain boundaries, E. *Fourth*, displacement of scratches at grain boundaries, and *Fifth*, subsidence of parts of grains, D. Betteridge and Franklin show that the third, fourth and fifth observations are all indications of grain movements with a component perpendicular to the surface. They go on to point out the special importance of the areas of localised strain in relation to the mechanism of creep, since energy stored in these areas must oppose the quasi-viscous flow of the boundaries. This leads to the same idea as developed by Cottrell and Aytekin [25] from their experiments in creep in zinc single crystals, that creep is a running balance between strain and recovery and that one of the dominant factor must be the rate of thermal softening of which the material is capable at the service-or test-temperature.

Where intercrystalline creep cracks occur, cracking is invariably multiple and the cracks are generally disposed more or less normally to the direction of stressing. Sectioning shows that such cracks usually make their first appearance at the periphery of the test-piece [28] and extend inwards with passage of time. There has been some controversy as to whether initial creep cracks are true stress cracks or are caused by stress-corrosion, the surrounding atmosphere being the corroding medium. Against this it has been found that creep cracking occurs in noble metals or in common metals in vacuo just as freely as in air and the extent of creep cracking does not increase regularly with rise of temperature. It is in fact at a maximum at temperatures just within the range where creep is an engineering problem and where ductility is at a minimum. As Sully [29] has pointed out in his book " Metallic Creep " much more elongation occurs at higher temperatures. Specimens then pull out to narrow pointed necks to give what Orowan [21, 30] has appropriately called " treacle fractures ". The position seems to be that stress conditions are readily set up in grain boundaries of a creeping metal which can only be relieved by fracture or by recrystallisation, which may perhaps be pictured as deceiving the material by moving the grain boundaries from the places where it had decided to crack. Some creep resisting materials turn out to be alloys in which there is a large temperature time dependence as regards cold-working, recorvery and recrystallisation. In Nimonic 80, for instance, slow deformation at 650°C does not cause the metal to work-harden, though fast deformation does at temperatures up to and above 1100°C.

Pure metals, including those of high melting point, are not usually outstanding in their creep resistance. In solid solutions creep resistance increases up to the phase boundry but, in practice it often proves advantageous to add rather more of the second component than will give saturation at the service temperature [17]. This may in some cases be no more than a device to ensure that local impoverishment at the grain boundaries does not occur and reduce the creep resistance where it matters most. In other cases, however, there is a definite merit in having an alloy which is capable of heat-treatment. For instance, where the required service life is relatively short—say a matter of a few hundred hours—any treatment which improves the short-time tensile properties at temperature (such as age-hardening) must be of benefit to the service life. Again there may be direct advantages to be gained from the presence of dispersed particles of a second phase.

Bennek and Bendel [31] have published results which demonstrate that the creep properties of ferritic carbon and alloy steels are at a maximum not when the steel has been properly hardened and tempered or when it has been normalised and tempered, but when it has been cooled before tempering at some critical intermediate rate which results in transformation in the upper part of the Bainite range. This result clearly means that a determinative factor in the creep performance of steel is the spacing of the dispersed carbide phase.

A result of the same kind was obtained earlier by Jenkins, Bucknall and Jenkinson [32] for an age-hardenable copper alloy containing 3 per cent of nickel silicon. As Allen [33] pointed out recently, " Jenkins, Bucknall, and Jenkinson, showed that if the time and temperature of test were below a certain limit, the precipitation—hardened condition had great advantages. Above this limit other conditions might be better, but the properties would in any case be poor. At the laboratory in which the author was interested, and probably at other laboratories as well, investigations began with a search for precipitation-hardening systems in which this limit was at the highest possible temperture."

Figure 13 shows typical test results on this alloy in a variety of conditions of heat-treatment, while Figure 14 and 15 show fracutred creep bars in fully hardened and overaged (softened) conditions both broken at 350°C. It will be noted that intercrystalline cracking is less severe in the latter and this gives the greater impression of ductility.

Jenkins, Bucknall, and Jenkinson did not define precisely the limit to which Dr. Allen refers, although it seems possible to do so on the basis of the published results obtained by them. This has been attempted in Figure 16 which summarizes the results of creep tests in terms of life to-fracture on ordinates of temperature and time in hours, on a logarithmic scale. The heavy curve divides the conditions in which the age-hardened alloy is at advantage (to the left of the curve) from those in which the slow-cooled equilibrium condition is more resistent (to the right). For example, at 300°C, the stresses supported by the age-hardened alloy are higher for lives up to 40 hours, but beyond 40 hours they are lower.

In addition to the limiting curve, contours of equal stress have been drawn in the two fields. Those for the age-hardened state are much steeper than those for the equilibrium state, which slope at about the normal rate for a pure metal. Thus, age-hardening seems to confer short-time benefits that are impermanent at most testing temperatures. In the early fighter-aircraft jet engines the expected life was short enough for very real benefits to be conferred by age-hardening; but it does not follow that advantage can be taken of age-hardening for other high-temperature applications, where long service lives are required.

In the author's opinion the rapid loss of "strength" shown by the age hardened copper alloy under creep conditions can readily be explained. It is thought that local lattice strain and recrystallisation during test, along the lines envisaged by Cottrell and Aytekin[25] and by Betterdige and Franklin[27], is more damaging to an age-hardened alloy than a metal or softened alloy because recrystallisation implies rejection of the dissolved solute.

It is evident that in the field where failure by creep is liable to occur even greater discretion is needed in the selection of materials of construction and in engineering design than in the more everyday application of metals. Pfeil, Allen and Conway [34] have emphasised this by stating "Experience with high-temperature materials shows that no one alloy in a single conditions of heat treatment gives the best properties for all types of service."

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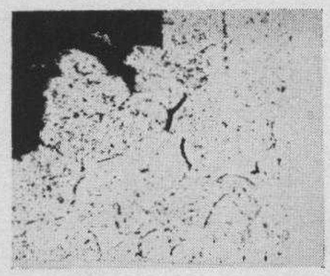

FIG. 14. Creep fracture at 350°C in age-hardened 3% Ni-Si-Cu alloy[32]

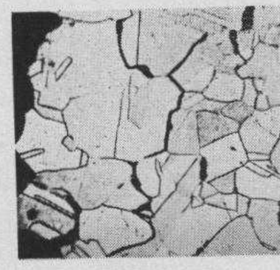

FIG. 15. Creep fracture at 350°C in over-aged 3% Ni-Si-Cu alloy[32].

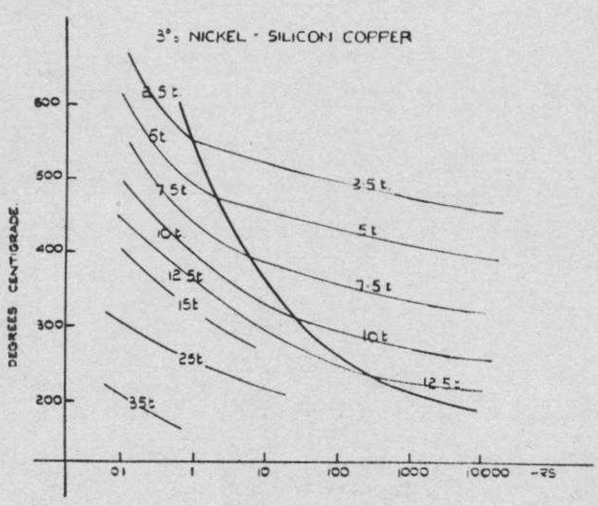

FIG. 16. Summary of creep test results on 3% Ni-Si-Cu alloy in age-hardened and over-aged conditions.

dent of the Birmingham Laboratory, and other colleagues for stimulating discussions and to Mr. W.G. Tallis, for making the tests summarised in Figure 2.

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DISCUSSION

Mr. S. Viswanathan:

What explanation the author had for the variation from brittle to tough fracture often seen within a notched impact specimen as the fracture ran inwards from the notch ?

Dr. A.K. Chatterjee:

Orowan by his classical work on mica proved by using grips smaller than the test piece, that the observed mechanical strength (tensile could be tremendously increased by eliminating the surface cracks. Bucknall has proved by his exceellent micrographs that the failure in a normal tensile fracture started from the centre and propogated towards the surface. Obviously these two observations needed co-relation, as both of them had been confirmed. Would the author kindly explain the reason for initiation of fracture from the centre of a tensile test piece ?

To Mr. Viswanathan's question about the variation in the nature of fracture in an impact test, Dr. Chatterjee added.

"The impact specimen contains a notch and when the pendulum strikes the test piece, the stress induced at the root of the notch is probably tri-axial tensional in nature. And tri-axial tension, as is well known, causes brittle fracture. Now when the region of tri-axial tension traversed, the remainder of the test piece fails in a ductile manner. Obviously one expects two types of fractures—brittle at the end of the notch and ductile over the remainder, in an impact test piece."

Mr. E.H. Bucknall:

"I do not altogether agree with Dr. Chatterjee's answer to Mr. S. Viswanathan's question. In my experience, cracking always starts *at the notch*, where we are told that the stresssing is essentially uniaxial (but raised to a high value by notch-effect) and not *behind the notch*, where tri-axial stressing occurs. Again, the stress distribution in a steel specimen above its brittle-transition temperature must, be, microscopically at least, closely similar to that of a steel specimen below is transition temperature, i.e. both contain triaxial stressing zones, but only the latter fails by brittle fracture. Thus the triaxial stress zone is neither the seat nor the direct cause of brittle-fracture, though undoubtedly it plays an important indirect part, by providing 'dead metal' which cannot shear freely and so limiting the volume of metal bing deformed at any time, i. e. raising the strain rate."

"Whilst it is impossible in the present state of knowledge to give a complete answer to Dr. Chatterjee's question the following can be said :".

"In order that a tensile test piece shall exhibit internal cracking it is necessary that the material shall have a high ductility—which involves both deformation and fracture in being by transcrystalline shear. In such a material local necking occurs and produces complex stress conditions, together with intensive work-hardening, in the layers immediately underlying the neck. The axial material is left more free to deform and presumably fractures as soon as the resolved shear stress (or the shear-strain energy) reaches a critical value. Since the axial stressing is simple, criterion of fracture is fulfilled at a value of tensile load/acting area which is unable to produce critical shear conditions near the surface in face of the other principal tensions."

"In non-ductile materials, including metals in certain special circumstances, e. g. in creep conditions and at low temperatures (for b.c.—cubic metals), where necking does not take place, cracking invariably starts from the surface, and these materials are presumably sensitive to surface defects in the same way as are glass and mica."