# Studies of the Brittle Fracture of Steel during the past Twenty Years

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## Abstract

This paper looks back over the work on the brittle fracture of steel during the past twenty years, to identify the experiences that have been most effective in changing our ideas, and have led to the opinions that are now held. Three classes of investigator have contributed to the changes, the steelmakers who make the material, the design engineers who have to define the conditions of stress and temperature and must determine the tests by which it is ensured that the material is suitable for its purpose, and the physical metallurgists whose concern it is to understand the processes that lead to fracture in order that dangerous conditions may at all times be avoided. All have played characteristic parts and their mutual reactions make a fascinating study.

## Fracture theories prior to 1948

At the beginning of the period, the most generally held view of the fracture process was that of Ludwik(1) and Davidenkov(2), according to which each metallic material is characterised by a yield stress and a fracture stress. The yield stress was the stress at which plastic deformation began, and it was recognised that this was determined by the value of the shear stresses. The fracture stress was the stress at which parting occurred along a recognisable fracture path, and this was considered to occur when the greatest principal tensile stress exceeded a limiting value. Both properties were affected by plastic deformation. The rate of rise of yield stress with increase of total strain defined the work hardening capacity of the material. The fracture stress was generally thought also to rise with increase of strain, but at a lower rate. When the stress conditions were such that the fracture stress was reached before the yield stress the material broke without prior deformation, and was apparently brittle. When the yield stress was exceeded first, plastic deformation proceeded fracture, and continued until work hardening had increased the yield stress to the fracture stress (as modified by the plastic deformation that had taken place) at which point the material broke. The strain at fracture was thought to depend principally upon the original difference between the critical yield stress and the critical fracture stress, which might be large or small, and upon the work hardening capacity of the material (Figure 1).

This view accounted for the embrittling influence of notches, defects and sharp changes of section, for the fact that many normally brittle materials are capable of plastic deformation when sheared in compression, and, in combination with the observation that the critical shear stress is raised by increasing the rate of loading, and the assumption that the critical fracture

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stress is not, for the tendency of many otherwise ductile materials to exhibit brittle behaviour under shock. For brittle materials it was accepted that the fracture stress ($\sigma_f$) was determined according to Griffith's equation by the surface energy of the material ($\gamma$), its elastic constants (expressed in an isotropic material by E and $\nu$) and the length (C) of small defects assumed to be present

$$\sigma_f^2 = 2E\gamma/\pi(1-\nu^2)C$$

but Griffith himself recognised, and it was generally agreed, that the high strength of ductile metals containing quite long cracks was due to the energy absorbed in the plastic deformation around the tips of the cracks(3).

The Ludwik-Davidenkov theory was based on general observations rather than upon assemblies of precisely determined data, and although, as will appear later, it fell into disrepute, it is important to recognise that it expressed an experience familiar to all metallurgists who have tried to increase the strength of metals and alloys. For all metals there appears to be a limit beyond which the yield strength cannot be increased without danger of embrittlement. Whether the yield stress is raised by alloying, heat treatment, decrease of temperature or increase of rate of strain, the increase of yield point is eventually accompanied by decrease of the strain at fracture, ultimately to a very small figure. The art of the metallurgist is to obtain a useful increase in the yield strength with the least sacrifice of ductility: and much can be done. Nevertheless, after intensive research aluminium alloys with yield strengths above 45 kg/mm$^2$, and steels with yield strengths above 140 kg/mm$^2$ both of which were known forty years ago, can be used only with circumspection, in closely defined applications.

Between the first and second world wars, steels used in general structural engineering were accepted on the results of tensile tests and plain bend tests only. Largely in consequence of the pressure of the aircraft and automobile industries, steels subject to dynamic and shock loads were required in several countries also to meet a notched bar test, as evidence of their toughness. This led to much work on the heat treatment of medium carbon alloy steels, during which it was established that the best combination of high yield stress* and notch toughness is obtained by quenching the steel to a fully martensitic condition and then tempering it(4). The importance of the hardenability of the steel in relation to the cross section of the component was recognised, and led to extensive studies of the factors influencing hardenability(5). In this period grain refined steels were introduced, and it was established that for equal yield stress, grain refined steels had better notch toughness than equivalent coarse grained steels(6); but because their hardenability was less, and their properties were at first very variable, their adoption in practice was slow. In this period also it was recognised that the results of notched bar tests were very dependent upon the temperature of the sample, that for each sample there was a temperature range in which the energy absorbed would decrease rather suddenly with falling temperature, and that the temperature range in which this change occurred varied considerably in different steels, and could be systematically varied

* generally 0.2% proof stress





by modifying the heat treatment of the steel(7).

During the second world war, the tank battles in Northern France, North Africa and Russia were accompanied by a technical battle for superiority between tank armour and armour piercing shot. The influence of rate of strain was emphasized, since many samples of armour that appeared tough in static tests, behaved in a brittle manner in firing trials, and to increase the hardness that could be achieved without danger of embrittlement became a matter of great urgency. The principles that had been established for obtaining notch toughness in automobile steels proved also to be the principles for making tough armour: it was surprising how much plastic deformation a properly heat-treated alloy steel could endure without fracture even under the attack of high velocity projectiles, and poor performance was generally due to failure in wartime conditions to maintain normal standards of control. It was however learned that in fracture, the properties of the weakest direction of the plate are dominant, and that this is the direction across the thickness of the plate, and that the properties are strongly influenced by local segregates of impurities, and fine cracks. The behaviour of armour piercing shot underlined the influence of deformation upon the distribution of stress in the component. When a shot strikes a plate in the most lethal direction (normal to the surface of the plate) the conditions are entirely compressive until the shot yields. At this point large tensile strains appear at the surface of the shot, and the shot bursts under their influence. In this case the most efficient shot is the one with the highest yield point, and the stress under which the shot fractures is the stress required to make it yield(8).

The failures in 1943-48 of the American Liberty Ships were remarkable, considering the low yield point and apparently high ductility of ship plate steel, for the lack of deformation before fracture, the absence of any sign of unusually high stresses or shock, and the speed with which the cracks spread. Since novel methods of welding and construction had been used, the first reaction was to blame the welding and the design, which were thought to have produced a combination of high internal stresses and points of stress intensification, but although numerous examples of faulty welding and unnecessarily severe stress raisers could be found, it proved impossible to reproduce the characteristic features of the failures by reproducing these features alone. Even in the presence of stress raisers and intentionally produced residual welding stresses, no brittle fractures could be produced without applying stresses at least equal to the yield point of the steel, and far greater than the stresses that had been permitted in the design of the snips. All the failures had occurred in noticeably cold weather. Examination of the properties of the steels at the temperature at which the failures had occurred showed that all the plates that had failed would have passed their specification tests at that temperature, and that they were normal in composition and microstructure. Nevertheless, at the temperature of failure, the plates in which failure had started had energy absorptions in notched bar impact tests on the average distinctly lower than the average for plates through which cracks had not been transmitted, and this provided the evidence for connecting the failures with the fall of energy absorption in notched bar tests that occurs in all ferritic steels at sufficiently low temperature(9).

## Improvement of Steel Quality

This was the position that had been reached in 1945 when a conference on the brittle fracture of ship plate was held in Cambridge, and the subsequent development of steels resisting brittle fracture will be traced from this point (10). At this conference it was demonstrated that a mild steel plate, if suitably notched on the edge and tested in tension, would break by a slow extension of fracture from the root of the notch with appreciable plastic deformation of the plate as a whole and a fibrous fracture if tested above a given limiting temperature, but would break suddenly, with a cleavage fracture and no appreciable plastic deformation of the plate if tested below that temperature (Figure 2). The transition temperature differed appreciably in different steels. A similar transition took place in Izod impact test pieces, and a variety of other forms of notched bar, but at temperatures differing, in the same steel, according to the nature of the test and the sharpness of the notch.

The metallurgist was inclined to assume that all forms of notched bar test were essentially influenced by the same properties in the steel, and that changes which lowered the transition temperature would decrease the probability of brittle fracture, and represent an improvement of the material. Consequently he used his experience with notched bar impact tests to forecast the conditions which would lead to lowered notched bar impact transition temperatures, and gradually the use of a Charpy test piece with an Izod V-notch became customary. It was soon demonstrated in commercial steels that:

(1) Fully deoxidised steels have lower notched bar impact transition temperatures than rimming and semi-killed steels(11).

(2) The phenomena associated with strain age hardening, particularly those due to high nitrogen content, are harmful(12).

(3) "As-rolled" steels have generally higher transition temperatures than the same steels in the normalised condition, but steels with carefully controlled low finishing temperatures can give very good results(13).

(4) High carbon contents raise the transition temperature. Steels with a given tensile strength have lower transition temperatures when they have low carbon content and high manganese content, than when they have high carbon content and low manganese content. The improvement is largely associated with the finer pearlitic structure of the former steel(14).

(5) Further improvement can be effected by refinement of the grain by aluminium and other additions(15).

(6) In a given steel, thick plates have higher transition temperatures than thin: the difference is associated with the finer structure produced by the generally quicker cooling of thin plates through the critical temperature*(16).

(7) In low carbon alloy steels, very low transition temperatures can be attained by quenching and tempering(17).

* This difference is observed when test pieces of equal dimensions are tested. When test pieces of full plate thickness are tested, there is an additional rise of transition temperature due to the effect of large plate thickness as such.





These conclusions have been confirmed by experiments on carefully controlled laboratory made materials(18). Pure iron in a coarse grained condition has a remarkably sharply defined transition temperature at -15°C, and this is appreciably raised by quite small additions of oxygen, nitrogen and phosphorus, which produce an intercrystalline weakness (Figure 3). Carbon raises the transition temperature provided it is not in solid solution, and its effect is greater when the rate of cooling is slow so that coarse carbide particles are formed (Figure 4). When a small amount of carbon is held in solution by quenching, the transition temperature is remarkably lowered and the yield stress is raised, so that moderately strong material of great toughness can be produced by quenching iron containing a few hundredths of one per cent of carbon (Figure 5). When manganese is added to iron containing 0.05% of carbon, the transition temperature is progressively lowered, and at the same time the structure is refined (Figure 6). When the grain size of ferritic materials or low carbon steels consisting substantially of ferrite is progressively refined, the notched bar impact transition temperature is also lowered(19).

Unfortunately most devices that lower the transition temperature entail some increase in cost, and as the cost of steel is a substantial part of the cost of a ship, much attention has been given to the cheapest way of producing a useful lowering of transition temperature. Undoubtedly the development of oxygen steel-making has owed much to the desire to eliminate nitrogen from steels made in converters. The degree of lowering of the transition temperature that is necessary to ensure freedom from brittle fracture in any particular case is unknown, and the question how much it is worth to attain a given value of the transition temperature has not been answered by the steelmaker. Steelmakers have however agreed to prepare series of steels with guaranteed values of impact value at stated temperatures, as determined by a standard form of notch test, for example the Charpy V-notch test(20). In Great Britain the series described in Table I is offered for steels for pressure vessels, stress relieved. They are provided in a series of grades, with different carbon contents and tensile strengths, and for simplicity the requirements for the "Grade 28" only with tensile strength between 44.1 and 52.7 kg/mm$^2$ are given. The required yield stress decreases with increasing thickness of plate, but the required impact values are held constant, except for plates over 50 mm thick, which are subject to agreement between manufacturer and purchaser. Most other countries will offer similar series, different in details of composition, in the way in which the impact test requirements are stated, and in the degree to which the effects of prior treatment and plate thickness are allowed for.

All these steels have yield points little higher than those of conventional mild steels, and are weldable. But there is a great desire for weldable steels of substantially higher yield point, equally resistant to brittle fracture. At present a second stage of development to provide steels of this kind is rapidly gaining impetus. The first steel of this kind is a semi-killed carbon manganese steel of normal carbon content with a small addition of niobium(21). It is normalised in such a manner as to bring a proportion of the niobium carbide to a finely dispersed condition. In this form the niobium carbide serves partly as a grain refiner of the austenite and partly as a dispersion hardening element, and gives a product of improved yield point and lowered transition temperature, in which the balance between yield point and transition temperature is determined largely by the amount of niobium retained in solid solution. The second is an experimental product called pearlite-free steel, which similarly contains manganese and a little niobium, titanium or vanadium but has a substantially lower carbon content.(22) After getting all the niobium into solid solution and allowing some of it to precipitate as fine niobium carbide, the steel is rolled at a controlled, rather low finishing temperature, which results in a structure with very fine ferrite grains and a fine dispersion of niobium carbide formed at the same time. A third variety consists of low carbon low alloy steels, subjected to a quenching and tempering treatment. They are particularly suitable for thick plates in which, on account of the limited cooling rates, the desired structure cannot be obtained without substantial additions of alloying elements.

Table II gives typical properties of some of these steels. They range from well established to highly experimental. A tendency towards lower carbon content, in the interests of weldability, and more elaborate rolling and heat treatment is apparent. The steels have much in common, in that the aim is to obtain a fine ferritic structure with a fine dispersion of carbide particles similar to that obtained in heat treated alloy steels. Ultimately control of pretreatment, carbon content, alloy content and final rate of cooling will all be employed to obtain the desired result in the most convenient way.

The new steels demand a more sophisticated control in their production. They offer a combination of high yield stress, exceptionally low transition temperature, ease of working both hot and cold, and freedom from cracking in the heat affected zones during welding. Much detailed metallographic and physicochemical study will be necessary before the most economical methods of production are established, but when this has been done, their production may well initiate a revolution both in steel making and in structural engineering.

Tests of Steel Quality: Fracture Toughness

Whilst the steelmaker can produce a wide range of steels of progressively lower transition temperature, the engineer is faced with the problem of choosing which steel to use and what stress to apply to it. Experience with the Liberty Ships and many other cases of brittle fracture shows that mild steel plate, as it has normally been made for many years is occasionally used in dangerous conditions. Yet the majority of the Liberty Ships and the great bulk of mild steel structures of all kinds have a perfectly satisfactory record, and only a slight improvement seems necessary to ensure that the dangerous condition cannot arise in service.

The first reaction to the problem was to produce a great variety of notch tests, each of which was claimed in various ways to reproduce "the conditions of service", and to assume that if the steel in service was invariably above its transition temperature in the appropriate test, the danger of brittle fracture would not arise. Much time was spent in comparing different types of test, and in discussing whether the energy absorption or the type of fracture was the better criterion of performance(23). The early results led to a degree of confusion, since the various tests frequently placed the same steels in different orders of merit. But more exhaustive tests generally showed that this result was largely due to local variations





in the supply of steel of nominally uniform quality, and that if sufficient tests were done the differences in the orders of merit obtained with different types of notch tests would be much reduced. Figure 7, taken from ASM Metals Handbook 1961 Vol.I.p.242 illustrates the general situation very fairly. When applied to one sample of steel, different types of notch test gave characteristically different transition temperatures, according to the form of notch, rate of strain and size of test piece employed, and it was impossible to be sure which test corresponded to the "conditions of service". Indeed conditions of service vary so widely that no single test can be expected to represent them all.(24)

Two further difficulties arose. In all these tests it was at first observed that when, on lowering the temperature, brittle behaviour appeared at the transition temperature, the nominal stress on the notched section was never below the yield stress of the material. Yet in many cases of failure it seemed impossible that stresses of this magnitude could have been applied. Moreover it is impossible to say that structures are unsafe whenever they are at a temperature below the transition temperature, for it is well known, for example, that bridges have been made of cast iron, which is brittle at room temperature, and have remained in service for hundreds of years. There is some nominal stress that can safely be applied to a notch brittle material, and the evidence was that this is probably lower than the yield stress; but there was no clear guide to its true value.

Four leading experiments pointed the way out of this dilemma. Robertson showed that if a brittle cleavage crack were initiated in a steel plate, for example by making a notch in one edge, cooling the material around the notch and striking a violent blow, the crack would spread rapidly across the plate, provided a tensile stress, above a measurable critical value, was present across the direction of the crack(25). For mild steel plate at low temperature this critical stress was small, around 6-9 $kg/mm^2$ and well below the design stresses permitted in ships, but on raising the temperature a point was reached at which the critical stress rose sharply to a value well above the yield point of the plate. At the same temperature the crack would change from a flat cleavage crack showing little deformation to a slow fibrous tear accompanied by local reduction of section. This experiment demonstrated the importance of crack initiation, for at low temperature the plate would clearly not sustain a stress after crack initiation that it would easily have carried before crack initiation; it defined a stress that could be carried in the presence of the crack, and a temperature limit above which that stress would be sufficiently high to permit the design of useful structures. This temperature has been termed the "Robertson transition temperature". Variations of the Robertson test were devised and used in the assessment of plate steels(26)(27)(28). The Japanese "double tensile test", in which the crack is initiated under one stress, and then propagated under a chosen lower stress, is outstandingly adaptable, and gives very clear indications of the relative resistances of different steels to the propagation of a crack (Figure 8)(28). Steels defined by the requirement that the Robertson transition temperature should be below the minimum service temperature were certainly very tough, but the test was too severe, for it excluded many types of steel that have proved reliable in service, and it became necessary to look more carefully into the conditions of crack nucleation.

The second important experiment was performed in the laboratories of the General Electric Company in America, in the investigation of a turbine rotor failure(29). The prior existence of severe stress concentrators had been established, but small scale tests had made it difficult to believe that these could possibly have extended under the quite accurately calculable stresses present in the rotor. But a turbine rotor is a very massive piece of steel, and notched bar tests were therefore made on bars of progressively increasing size. It was found that the nominal stress at which failure occurred decreased progressively as the dimensions of the test bars increased, and that whilst the nominal failure stress was well above the yield stress when the test bars were small, it was well below the yield stress when the test bars were large(Figure 9)(52).

The third experiment was that of Irwin(30)(31), who studied the stresses necessary to cause the extension of cracks of various lengths, artificially produced in plates. He found that the extension, at first very slow, would become rapid and uncontrollable when the nominal stress at the tip of the crack exceeded a given value, and that this value decreased as the length of the crack increased, according to equations of the general type

$$\sigma_f^2 = \mathcal{K} G_c / \ell$$

where $\sigma_f$ is the nominal stress for rapid crack extension, $\ell$ is the length of the crack, and $\mathcal{K}$ is a numerical constant dependent on the elastic constants of the material and geometrical factors such as the position of the crack in the plate and the plate thickness. Under plane strain conditions $G_c$ is a material constant, representing the strain energy released per unit of severed area of fracture and is distinguished by the suffix $G_{Ic}$. Irwin's view is that all materials contain crack origins of some equivalent length $\ell$, and in this respect it is similar to that of Griffith. The value of $G_c$ determines the stress necessary to cause the uncontrolled growth of these origins, and the establishment of this uncontrolled growth is the process of crack nucleation that is necessary before the rapid crack propagation studied in the Robertson test can be realised. The maximum permissible stress is determined by the ratio $G_c/\ell$ and when the working stress, required by the design and function of the component is defined, the maximum size of the cracks or other defects that may be permitted in the component is directly proportional to the $G_c$ value of the material. This viewpoint has been very successful in systematising the behaviour of high tensile steels. In a given steel, the $G_c$ value is found to fall as the yield stress rises, so that, assuming the defect size determining $\ell$ to remain constant, a yield stress level is reached at which the stress for crack propagation becomes less than the yield stress, and at this point the steel begins to break in a brittle manner(32). Thus the inability to increase the yield stress of a metal indefinitely without danger of brittle fracture, which was the basis of the Ludwik and Davidenkov theory was satisfactorily explained. $G_c$ is also found to be dependent on the temperature of the steel and the rate of strain, both being factors that influence the effective yield stress.

The fourth experiment relates to the nature of the crack origins that determine the length $\ell$ and was made by Mylonas(33). He took an ordinary mild steel in which notched tensile test pieces had a nominal breaking stress well above the yield point of the material, and subjected the test pieces to a preliminary compression, sufficiently large to produce substantial deformation at the root of the notch. When the test piece was subsequently





tested in tension, it broke in a brittle manner at a low stress, in some cases below half the yield stress. Mylonas attributed this result to "exhaustion of ductility" meaning by this that highly deformed regions were formed and that these, being incapable of further deformation, served as fracture nuclei. Recently some work published from the British National Physical Laboratory has shown that highly compressed pure iron has still ample ductility if subjected to tensile stress transversely to the direction of compression, and in the direction of compression if the temperature is sufficiently high, but that when tested below room temperature in the direction of compression the compressed iron breaks at a low stress in a completely brittle manner (Figure 10)(34). The brittleness is thought to be due to lines of weakness developed along the boundaries of the flattened grains, rather than to any inherent inability of the material to deform further (Figure 11)(34).

These experiments show how in the presence of notches and defects that serve as fracture nuclei, brittle fractures may occur at definable stresses well below the yield point of the material, and that these notches and defects may be present in the material before the stress is applied. The combination of circumstances that in practice may lead to brittle fractures has been illustrated in a series of wide plate tests on welded mild and low alloy steels by Wells(35). These tests showed that the unwelded steels, even when deeply notched, will not break at nominal stresses below the yield point at any reasonable temperature, though below a certain temperature fast fractures sensibly free from plastic deformation are produced. Nor will they break in similar circumstances when welded if the welded plates have been heat treated in such a way as to relieve the welding stresses and the associated effects of the welding strains. But if before welding a saw cut of suitable length has been introduced, so that welding stresses and strains might have been concentrated at its end, and the plates are not subsequently stress relieved, a crack may extend from the end of the saw cut when only a quite small stress is applied. This crack may stop at the limit of the region of high residual tensile stresses, and require a further increase of applied stress, generally to the yield point, to cause the plate to fail completely, or it may spread immediately through the whole plate, according to the temperature of the plate and the range and severity of the welding stresses. Figure 12(35a), replotted from one of Wells's sets of results, shows how the stress at failure was related to the yield stress of the material and the temperature. The relations between the lengths of the observed cracks, and the stresses required to make them spread rapidly through the plate, are consistent with Irwin's equation, and no failures under low stress occurred in plates that were above their Robertson crack arrest temperature.

Thus the stresses that may safely be applied to steels below their transition temperatures are determined by the factors that control $G_c$ and $\ell$ and the change from brittle to ductile fracture is a consequence of the change of $G_c$ with rising temperature. This causes the fracture stress to rise above the yield point when a certain temperature is exceeded.

## Theories of Fracture

The factors affecting $G_c$ and $\ell$ are determined by the physical nature of the fracture process. The Ludwik-Davidenkov theory of fracture came to be widely criticised shortly after the second world war, largely because attempts to measure the stipulated fracture stress, and its assumed strain dependence, by adopting various devices to suppress the intervention of plastic deformation during the test, failed to give any satisfying impression of its value. Improved understanding of interatomic forces led to the conclusion that the forces between atoms in metallic lattices were far higher than the experimentally observed fracture stresses, and the correctness of the conclusion was confirmed by the extraordinarily high strengths occasionally displayed by sufficiently thin single crystal "whiskers". Moreover, it was noticed that instead of the fracture stress and the yield stress being mutually independent, an increase of the yield stress, for example by rapid straining or by strain ageing, was frequently accompanied by an equal increase of the experimentally observed fracture stress. This led to the replacement of the Ludwik-Davidenkov theory by the suggestion, originally put forward by Zener(8a), that the difference between the observed breaking stresses of metals and the theoretically calculated breaking stresses was due to the presence in the metal of defects produced in the process of plastic deformation by the running together of dislocations, particularly at points where the motion of the dislocations was halted by the presence of some obstacle.

In support of this view, it was pointed out that Lüders bands in mild steels could, under favourably chosen circumstances, be shown to contain small cracks(36), whereas unstrained parts of the same specimen, which had nevertheless been subjected to the same stress, contained no cracks; and that when steels were embrittled at low temperatures, the fracture stresses in the brittle condition were frequently very close to the yield stresses that would have been expected at the same temperature from the course of the yield stress-temperature curve, although no yield might actually have been observed(37). Moreover, an attractive explanation of the well-known effect of small grain size on the impact transition temperature and general resistance to fracture could be arrived at on the basis of the assumption that the size of the groups of dislocations arrested at the grain boundary would be proportional to the square root of the grain diameter, and that these groups would produce the stress concentrations necessary for the initiation of fracture(38).

Between 1948, when the theory was first advanced, and the time of the Swampscott conference in 1959, favourable evidence accumulated, and the theory was widely accepted(39); but with increasing knowledge of the effect of temperature on the yield stress of iron, and with the development of the ability to observe the dislocations in iron, certain difficulties became apparent. When the properties of steels at temperatures well below the brittleness transition temperature were examined, it was found that the coincidence of the fracture stress with the yield stress extended for only a short range of temperature. Figure 13(53) illustrates a case in which the fracture stress decreased after having been close to the yield stress during a narrow transition range. Below this range the fracture stress either increased with falling temperature at a rate much less than the rate of increase of the yield point, or it remained substantially constant, or decreased. The decreases of fracture stress were particularly pronounced in alloys to which embrittling impurities such as oxygen, nitrogen, phosphorus or silicon had been added(18)(40). The shape of the fracture-stress temperature curve was in these cases very similar to that found for a notched bar, so that it was reasonable to think that these materials contained some internal feature that served the purpose of a notch, and that the fracture





origins supposed by the Zener theory to be produced by plastic deformation were already present in the unstrained material(41). The dislocation structure of iron stressed just to the yield point did not show the groups of arrested dislocations required by the theory; such groups were found only after substantial strain and then were formed at space intervals far smaller than grain size (Figure 14)(42)(42b). Moreover the behaviour of Charpy V-notched bars of polycrystalline pure iron was not really in accordance with the theory. Such bars show a very sharply defined brittleness transition, the temperature of which depends on the sharpness of the notch. Just above the transition temperature the bars bend through a large angle, showing very great plastic deformation at the surface representing the root of the original notch, but no cleavage cracks. At a temperature a few degrees lower they break completely with a predominantly cleavage fracture, and practically no deformation, as is illustrated in Figure 15(54). Both specimens in Figure 15 have twinned, but in the ductile specimen the twins have deformed substantially, without giving rise to cracks. By varying the sharpness of the notch it would be possible to get two test pieces of the same iron tested at the same temperature, one of which has cleaved with little deformation, whilst the other has deformed substantially, and not cleaved. If the production of cracks by a small amount of deformation had been the essential condition for the cleavage of the first, it is most unlikely that no cleavage should be found in the second. It is, of course, the difference between the stress systems at the root of the notch that is the essential difference between the two test pieces, the ratio of tensile stress to shear stress being higher in the test piece with the sharper notch, so that higher tensile stresses can build up in it in spite of the fact that the critical shear stress of the material is the same as in the other test piece.

The importance of the tensile component of the stress is well brought out in studies of the fracture properties of single crystals of pure, and slightly impure iron (43)(46). At -196°C these either cleave or break in a ductile fashion according to the angle which the most favourably oriented cube plate normal makes with the direction of tension, the critical angle being for pure iron about 25°. The stress at which either cleavage or deformation by slip occurs suggests that cleavage is preferred if the resolved tensile stress across the cube plane exceeds about 45 kg/mm$^2$ before the resolved shear stress along the most favourably oriented slip plane exceeds 28 kg/mm$^2$. If the critical shear stress of the crystal is raised by addition of impurity, such as carbon, nitrogen or phosphorus, or by radiation hardening, the effect is to increase the critical angle by such an amount as suggests that whilst the critical resolved shear stress is raised the critical resolved cleavage stress has remained substantially, though not accurately, constant(47). The effect of lowering the temperature is to increase the critical angle, in accordance with the expected change of critical shear stress, and that of raising the temperature to decrease the critical angle. Dr. J. Harding at the National Physical Laboratory, England, has recently completed and is about to publish a rather thorough study of the effect of very rapid strain on pure iron crystals(48). At -196°C the yield stress is raised substantially, and the critical angle is accordingly increased, but the resolved fracture stress remains obstinately at about 42 kg/mm$^2$. In accordance with the increased resolved shear stress, cleavage fractures are obtained in rapid tests at higher temperatures than are possible with slow rates of strain. Figure 16(48) illustrates the effects of addition of phosphorus and of increase of the rate of strain at -196°C.

In all these fractures, twinning processes intervene at about the same stresses as are necessary to cause cleavage, and the extent to which the stresses resulting from twinning contribute to the cleavage stress has been much discussed. Nevertheless the results suggest that whatever the reason, iron single crystals of the type we have employed have a characteristic critical resolved cleavage stress of about 45 kg/mm$^2$, and that the mode of failure, whether by cleavage or slip, depends almost entirely on the shear stresses and the tensile stresses in the crystal in relation to the critical cleavage and shear stresses. The effect of raising the resistance to shear is much as would be expected on the Ludwik-Davidenkov theory, and the picture of shear as a process that intervenes to prevent the fracture stress being reached seems to be valid in this case.

However, assuming that a crack origin is present in the metal, either existing from the beginning, or produced in the course of plastic deformation, two views are expressed about the conditions in which the origin may spread to produce a fracture. One is that the crack is narrow, but has at its tip a region of shear stress, of size dependent on the critical tensile stress required to spread the crack. In this region plastic deformation occurs, and the work per unit area required to spread the crack is the surface energy, $2\gamma$, plus the plastic work $W$ involved in the forward movement of this plastically strained region (Figure 17a). A modified Griffith equation in the form

$$\sigma_f^2 = E\,(2\gamma + W)/\pi\,(1-\nu^2)C$$

is used, where $\sigma_f$ is the fracture stress, $C$ is the crack length, and $E$ and $\nu$ are elastic constants.

The other view is that the crack is wide and that there is a critical lateral displacement $\phi_c$ of the surfaces of the crack (Figure 17b)(49). The energy required for fracture is that required to produce this displacement at the end of the crack and is of the order of magnitude $\phi_c\sigma_y$ per unit increase of the length of the crack, where $\sigma_y$ is the yield stress of the material. $\phi_c$ is related to the yield stress $\sigma_y$ and the applied stress $\sigma$ by the relation

$$\phi_c/C = (8\sigma_y.(1-\nu)/\pi E)\ \ell n \sec(\pi\sigma/2\sigma_y)$$

in which $C$, again, is the crack length.

Fracture occurs when

$$\sigma_f^2 = E\,\phi_c\,\sigma_y/\pi\,(1-\nu^2)C \qquad (1)$$

The essential difference between the assumptions is seen by considering the behaviour of a notched bar, sufficiently wide to give plane strain conditions at the centre (Figure 18)(32). When the loads are so low that the yield stress is nowhere exceeded, the maximum tensile stress and the maximum strain are both at the surface of the root of the notch. When the yield stress is exceeded at the root of the notch, the maximum strain, now partly elastic





and partly plastic, is still at the root of the notch, but the maximum stress, on account of the influence of the second and third principal stresses, is somewhat below the surface, at the extreme limit of the plastically deformed zone. Up to a point, this limit lies more deeply in the specimen, and corresponds to a higher stress, the greater the applied load. On the first view, failure should begin from the point of maximum stress, below the surface: on the second it should begin at the surface. It is not always easy to discover where a rapid fracture has begun, but the evidence is that well above the notched bar transition temperature, fracture begins at the surface, and below the transition temperature, beneath the surface.

On the second view it is difficult to account for the existence of the sharp notched bar transition temperature, for on lowering the temperature $\sigma_y$ rises, and there is no obvious reason why $\phi_c$ should suddenly fall (50). Yet the amount by which the root of the notch distorts before failure certainly decreases sharply at the transition temperature. The phenomena at the transition temperature on the other hand are readily and naturally explained on the first view, in terms of the changes of stress in the bar as the load is increased.

As the load is increased, the maximum tensile stress along the centre line through the root of the notch rises almost linearly at first, but tends to reach a constant value when the point of maximum stress reaches the region in which the triaxiality ratio of the stress system becomes constant. At this point the plastic zone, which previously was increasing its size whilst maintaining roughly the same shape, develops a forked shape, deformation being concentrated in those regions where the triaxiality ratio is least, and little further deformation occurs in the region between the prongs of the fork (51). The exact form of the curve relating the maximum tensile stress to the load is influenced by the work hardening within the plastic zone, and has not been accurately calculated, but a much simplified approximation suggests for the Izod V notch that the maximum stress reached is about 1.93 times the uniaxial yield stress ($\sigma_y$), and is reached when the nominal stress at the root of the notch is about 0.8 times the yield stress. If the fracture stress of the steel is less than 1.93 $\sigma_y$ fracture begins at the point where the fracture stress is first exceeded and the energy absorbed by the specimen at this point is predominantly the energy of plastic deformation within the small plastic zone. If the fracture stress is greater than 1.93 $\sigma_y$ fracture is delayed and general plastic deformation continues, until at some place in the test piece, the stress as determined by the local yield conditions, is equal to the local fracture stress, as modified by the plastic deformation that has taken place at that point. If the fracture stress is only slightly higher than 1.93 $\sigma_y$ this point is likely to be just within the outer edge of the plastic zone, where the yield stress has been appreciably raised by the rapid initial stages of work hardening, but if it is considerably higher than this it is likely to move towards the point where the plastic strain is greatest, which is at the root of the notch, and fracture will begin at a strain comparable with the fracture strain in a simple tensile test. The energy absorbed by the specimen when fracture begins is the energy of plastic deformation of the large plastic zone at that point. If the material has a sufficiently high fracture stress in relation to the yield stress, fracture may proceed as a slow plastic tear, absorbing considerable further energy, but in a less ductile material the increased stress concentration at the tip of the newly formed crack may cause initiation of new ruptures beneath the tip of the crack, particularly as the effective yield stress is increased by the increase of rate of strain that begins as soon as the test piece begins to fail. In this way the frequently observed fractures changing from ductile at the notch root to brittle at some distance below the surface are formed, and in this case the subsequent progress of the crack absorbs little energy (32).

When fracture is initiated below the surface, the maximum load on the test piece at fracture becomes a smaller proportion of the yield load as the yield stress rises in a given steel, the maximum load is the same function of the yield stress whether the change of yield stress is brought about by change of temperature or (within limits) by change of heat treatment. There is a limiting value of the yield stress above which the maximum load at failure falls below the nominal yield load of the bar and this approximately defines the conditions in which the change from a probably ductile to a completely brittle type of failure occurs (Figure 19) (32).

It is possible to have steels of entirely similar yield properties and different fracture behaviour. For example, a heat treated manganese molybdenum steel free from phosphorus has for equal yield stress a fracture load consistently greater than that of a similar steel to which an addition of 0.05% of phosphorus has been made. The tensile stress that is reached in the region of triaxial stress is consistently higher. Consequently at equal yield stress, the notched bar transition temperature is appreciably lower in the phosphorus-free steel, and the level of yield stress to which it can be heat treated without the risk of brittle behaviour at room temperature is appreciably higher.

This kind of phenomenon is quite common in steel and is consistent with the view that steel has a fracture stress that is independent of the yield stress. It cannot be so easily dealt with on the view that the displacement at the root of the notch is the decisive factor, since the equations ( 1 ) require that steels of equal yield stress shall behave alike. However, it is to be noted that the two steels just described do behave in exactly the same way when they are both well above their notched bar transition temperature, and it may well be decided ultimately that the critical displacement hypothesis is the one to be adopted for steels that are sufficiently above their notched bar transition temperature (and this comprises the conditions in which the great majority of structural steels should be used) but that the critical stress hypothesis is valid for those that are used below their notched bar transition temperature. But it also needs to be emphasised that it is the critical stress (in relation to the yield stress of the steel) that decides when the transition occurs.

The critical displacement hypothesis provides a feasible explanation for the effect of size on brittle fracture stress, as for example, it was determined by Winne and Wundt (29), and by Lubahn and Yukawa (52) for the critical displacement is a constant characteristic of the material and decides the extent of the plastic zones. The intersection of the plastic zones with the free surfaces of the test piece brings about the greater freedom of plastic deformation that is characteristic of a small bar, and in turn is responsible for its greater ability to absorb energy, and consequently, its greater strength. The critical stress hypothesis leads to the conclusion that, in those conditions in which a simple tensile rupture originates the fracture process, the nominal





breaking stresses of geometrically similar bars should be independent of size. In practice this is not found to be so. Small bars have substantially higher nominal breaking stress than large bars. The law that governs the variation with size has not yet been sufficiently studied. Again, the extent of the plastic zones when the critical fracture stress is being approached in the region of triaxial stress would be expected to be important, for this determines the rate at which dislocations accumulate in the plastic zones, and therefore the rate of work hardening. The smaller the rate of work hardening, the greater is the energy of plastic deformation that has to be expended before the critical fracture stress is reached, and therefore the greater the apparent strength of the bar. But the very complex calculations of the extent of the plastic zones, taking into account the influence of all the free surfaces of the test piece and of the work hardening capacity of the material, have not yet been satisfactorily performed, and it is most important to engineers and metallurgists that applied mathematicians should steel themselves to face the difficulties.

To the metallurgist, the important subject of study is the nature of the factors that determine the critical fracture stress. Some of the estimates that can be made from the results of experiments reported in the literature, or from unpublished work of the National Physical Laboratory are given in Table IV.

From this table it is to be seen that the property can be varied over a great range, the extreme values being 30 and 430 kg/mm$^2$. Grain size, cold work, and the directional nature of the structure produced by cold work, purity, and quantity of a finely dispersed second phase can be identified as important variables. In quenched and tempered steels, provided the normal quenched and tempered structure is obtained, the nature of the alloying element used does not seem to be very important. In mild steels and 1.5% manganese steels, the rate of strain does not at present appear to be very important. But clearly much more work needs to be done on this property, for a systematic understanding of the ways in which it can be influenced is essential if the strengths of steels in conditions in which they may be brittle are to be fully understood.

Acknowledgment

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TABLE I

British Steels with Controlled Notched Bar Impact Values

(Steels for Fired and Unfired Pressure Vessels, Stress Relieved
Grade 28 : 44.1-52.7 $kg/mm^2$ tensile strength)

| Description | Composition % | | | | Condition | Yield Stress (Min) | | Energy Absorption | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | C (max) | Si | Mn | Others | | Plate Thickness | $kg/mm^2$ | Temp. °C | kgfm (min) | Test |
| BS 1501-211 Carbon Manganese, semi-killed | 0.19 | 0.10 max | 0.90/<br>1.50 | - | Normalised and Stress Relieved | <16 mm<br>16-32 mm<br>32-64 mm | 26.0<br>25.2<br>24.4 | 20<br>0<br>-15 | 4.8<br>3.5<br>2.8 | Charpy V-notch |
| | | | | | As Rolled and Stress Relieved | | | 20<br>0 | 4.1<br>2.8 | Charpy V-notch |
| BS 1501-221 Carbon Manganese Silicon-killed | 0.19 | 0.10/<br>0.55 | 0.90/<br>1.50 | - | Normalised and Stress Relieved | <16 mm<br>16-32 mm<br>32-64 mm | 26.0<br>25.2<br>24.4 | 20<br>0<br>-15 | 5.5<br>4.1<br>3.5 | Charpy V-notch |
| | | | | | As Rolled and Stress Relieved | | | 20<br>0 | 4.8<br>3.5 | Charpy V-notch |
| BS 1501-224 Carbon Manganese Silicon-killed, Aluminium Treated | 0.17 | 0.10/<br>0.55 | 0.90/<br>1.50 | Sufficient Al to give McQuaid-Ehn grain size of 5 or finer | Normalised and Stress Relieved | <16 mm<br>16-32 mm<br>32-64 mm | 29.1<br>27.6<br>26.0 | 20<br>0<br>-15<br>-30<br>-50 | 6.9<br>6.9<br>5.5<br>4.8<br>2.8 | Charpy V-notch |





TABLE II

More Advanced Notch Tough Steels

| Description | Composition % | | | | Condition | Tensile Strength | Yield Stress | | Energy Absorption | | |
|---|---|---|---|---|---|---|---|---|---|---|---|
| | C | Si | Mn | Others | | $kg/mm^2$ | Plate Thickness | $kg/mm^2$ | Temp. °C | kgfm (min) | Test |
| BS 1501-213 Carbon Manganese Semi-killed, Niobium Treated Grade 28. | 0.17 max | 0.10 max | 0.90/<br>1.50 | 0.01/0.06 Nb | Normalised and Stress Relieved | 44.1/<br>52.7 | <16 mm<br>16-32 mm<br>32-64 mm | 30.7<br>29.9<br>29.1 | 20<br>0<br>-15<br>-30 | 6.9<br>5.5<br>4.1<br>2.8 | Charpy V-notch |
| ditto Grade 32. | 0.22 max | 0.10 max | 0.90/<br>1.60 | 0.01/0.06 Nb | Normalised and Stress Relieved | 50.5/<br>60.6 | <16 mm<br>16-32 mm<br>32-64 mm | 34.6<br>33.9<br>33.1 | 20<br>0<br>-15<br>-30 | 6.9<br>5.5<br>4.1<br>2.8 | Charpy V-notch |
| U.S. Steel T1 Steel | 0.10/<br>0.20 | 0.15/<br>0.35 | 0.60/<br>1.00 | 0.70/1.00 Ni<br>0.40/0.80 Cr<br>0.40/0.60 Mo<br>0.03/0.10 V<br>0.15/0.50 Cu<br>0.002/0.006B | Quenched and tempered | min<br>81/95<br>74/95 | < 50 mm<br>50-63 mm | 70†<br>63† | -12<br>-40<br>-70 | 4.1*<br>4.1<br>2.8 | Charpy V-notch |
| BISRA Pearlite Free Steel | 0.05 | - | 1.9 | 0.06 Nb<br>0.011 N | As-rolled, finishing Temp.690°C | - | - | 64 | -60 | ~ 2.8 | Charpy V-notch |
| | | | | | As-rolled, finishing Temp.950°C | - | - | 45 | -20 | ~ 2.8 | Charpy V-notch |

TABLE II (continued)

| Description | Composition % C | Si | Mn | Others | Condition | Tensile Strength kg/mm² | Yield Stress Plate Thickness | Yield Stress kg/mm² | Energy Absorption Temp. °C | kgfm (min) | Test |
|---|---|---|---|---|---|---|---|---|---|---|---|
| Japanese Steel for ## Gas Storage Tanks | 0.09 | 0.22 | 1.03 | 0.014 Al | Rolled, quickly cooled in press. Tempered | 47/55 | 12 mm<br>25 mm<br>36 mm | 33/47 | -136<br>-125<br>-125 | 2.0<br>2.0<br>2.0 | Charpy V-notch |
| QT 35 | 0.15<br>0.17<br>0.18 | 0.3<br>0.3<br>0.3 | 1.2<br>1.2<br>1.3 | 1 Cr max<br>1.2 Ni<br>0.5 Mo<br>0.12 V | Quenched and Tempered | -<br>-<br>- | 9-44 mm<br>44-57 mm<br>57-76 mm | 57-69<br>60-71<br>60-71 | -40°C | 8.3 | |

† ASME Boiler and Pressure Vessel Case 1204-8, May 1961.

* L.C. Bibber Welding Journal Research Supplement 1955.

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TABLE III

Resolved Mean Fracture Stress of Iron Crystals

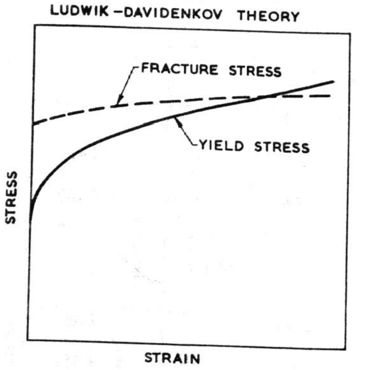
| Material | Temperature | Stress kg/mm² | No. of Tests |
|---|---|---|---|
| STATIC TESTS | | | |
| Pure Iron as grown (3 batches) | -196°C | 46<br>41<br>47 | 4<br>8<br>9 |
| Pure Iron, aged | " | 41 | 2 |
| " " irradiated | " | 45 | 6 |
| " " carburised | " | 44 | 7 |
| " " nitrided | " | 52 | 17 |
| " " nitrided and aged | " | 52-59 | 25 |
| " " +0.15% P | " | 46 | 10 |
| Pure Iron as grown | -253°C | 46 | 11 |
| " " " " | -269°C | 41 | 11 |
| DYNAMIC TESTS | | | |
| Pure Iron as grown | -78°C<br>-160°C<br>-196°C | 38<br>41<br>42 | 2<br>3<br>24 |
| Pure Iron + 0.15% P | -196°C | 47 | 9 |





TABLE IV

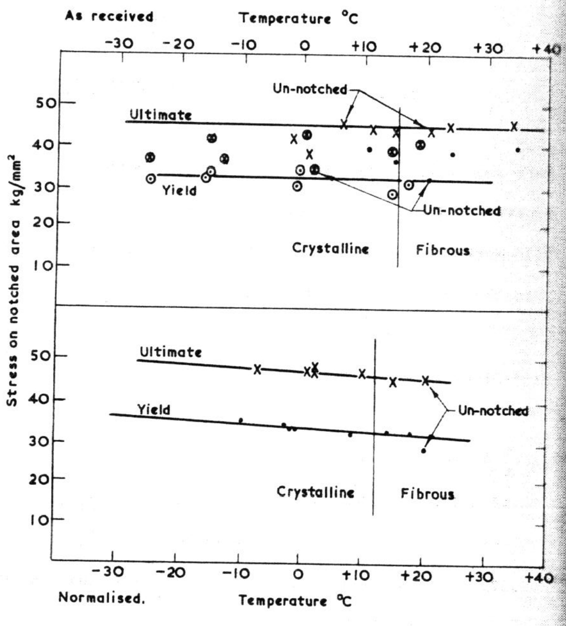
| Material | Temperature | Estimated Critical Fracture Stress kg/mm² | Reference |
|---|---|---|---|
| Pure iron single crystal | -196°C | 45 | Table III |
| Annealed polycrystalline pure iron | -196°C | not less than 94 | (34) |
| Cold Worked polycrystalline pure iron | -196°C | | |
| in the longitudinal direction | | ~ 125 | |
| " transverse " | | ~ 30 | |
| Mild Steel - Static Test | - 78°C | 123 | (41) |
| " " " " | -196°C | 124 | (32) |
| Steel Plate | | | |
| 0.12-0.18% C, 1.2-1.5% Mn | | | |
| Dynamic tests on 5 different steels | - 70°C | 90-124 | (56) |
| 0.17%C 0.4% Mn Static test | -78°C and -129°C | 148 | (51) |
| 0.4% C, 1.5% Mn, 0.3% Mo steel | | | |
| quenched and tempered | | | |
| 0.001% P | - 78°C | 330 | (32) |
| 0.054% P | -78°C and 20°C | 270 | |
| 3% Mo 0.5% V steel | | | |
| O.Q.1050°C, T.500°C 1 hour | -196°C | | |
| 0.16% C | | 278 | (57) |
| 0.30% C | | 336 | |
| 0.46% C | | 363 | |
| Maraging steel (18% Ni) | 20°C | not less than 430 | (32) |




GRIFFITH THEORY (BRITTLE MATERIALS)

$$\sigma_f^{\,2} = 2\gamma E / \pi (1 - \nu^2) c$$

Fig 1




As received Width 19 mm Notches 3 mm, 45°
⊙ and ● indicates width 76 mm

Normalised Width 25 mm 2 notches 3 mm

FIG. 2. NOTCHED TENSILE TESTS (NOMINAL THICKNESS 8·6 mm)

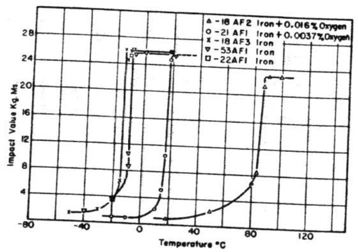



Figure 3. Impact Values at Various Temperatures of Five Irons Normalised at 950°C.

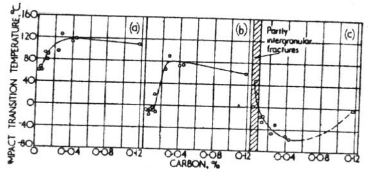
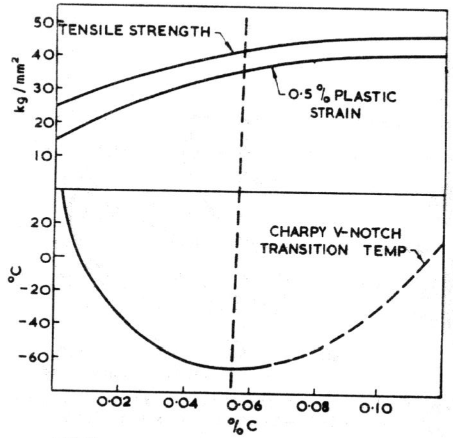



Figure 4. Charpy impact transition temperatures of iron-carbon alloys (a) furnace cooled (b) air cooled (c) water quenched from 950°C.








TENSILE AND IMPACT PROPERTIES OF WATER QUENCHED IRON CARBON ALLOYS

Figure 5. Tensile and impact properties of water quenched iron-carbon alloys.

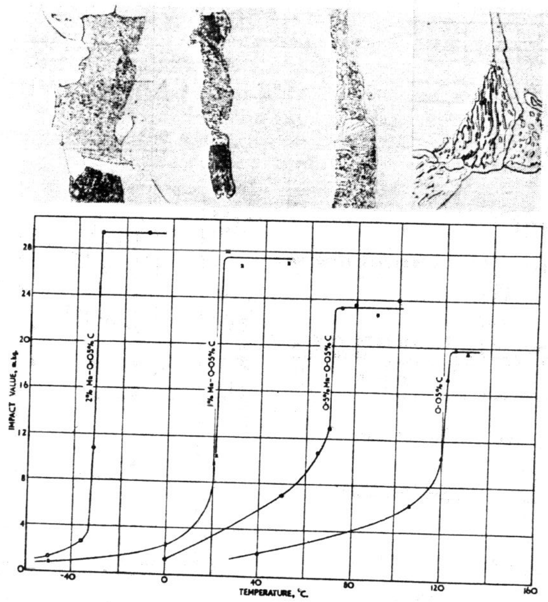



Figure 6. Impact values at various temperatures of iron -0.05% carbon alloys with manganese additions up to 2% furnace cooled from 950°C showing microstructures (807X)

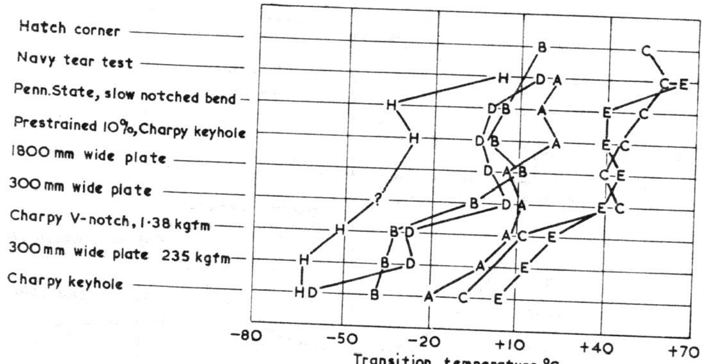



FIG.7. COMPARISON OF TRANSITION TEMPERATURES DETERMINED BY NINE DIFFERENT METHODS ON SIX STEELS.

| Chemical composition % | | | | |
|---|---|---|---|---|
| C | Mn | Si | P | S |
| ·18 | ·41 | – | ·023 | ·021 |

| Mechanical Properties | | |
|---|---|---|
| Yield point | Ultimate strength | Elongation |
| 29 kg/mm² | 44 kg/mm² | 28% |

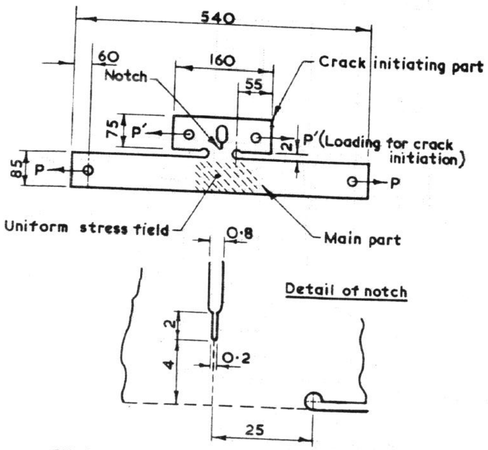



FIG.8a. DOUBLE TENSION SPECIMEN (Units: mm)

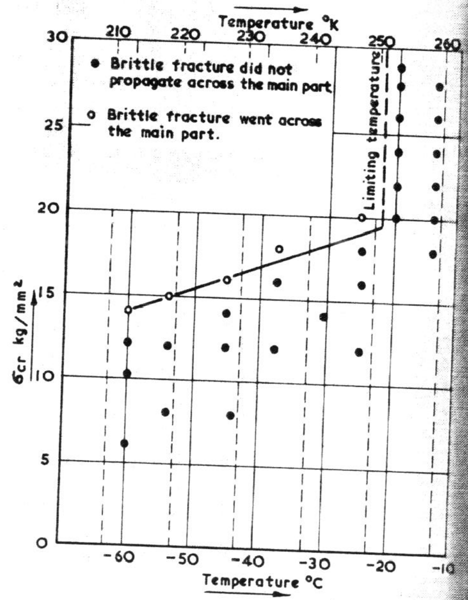



FIG.8b. RELATION BETWEEN CRITICAL STRESS FOR THE PROPAGATION OF BRITTLE CRACK AND TEMPERATURE





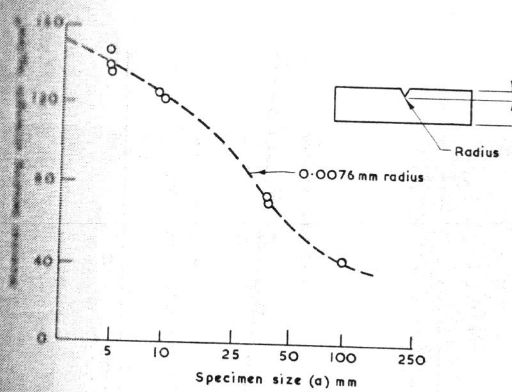



FIG.9. SLOW NOTCH-BEND STRENGTH VERSUS SIZE

Figure 11. Thin Foil Electron Micrograph of large crack approximately 10 mm long in iron compressed 75%.

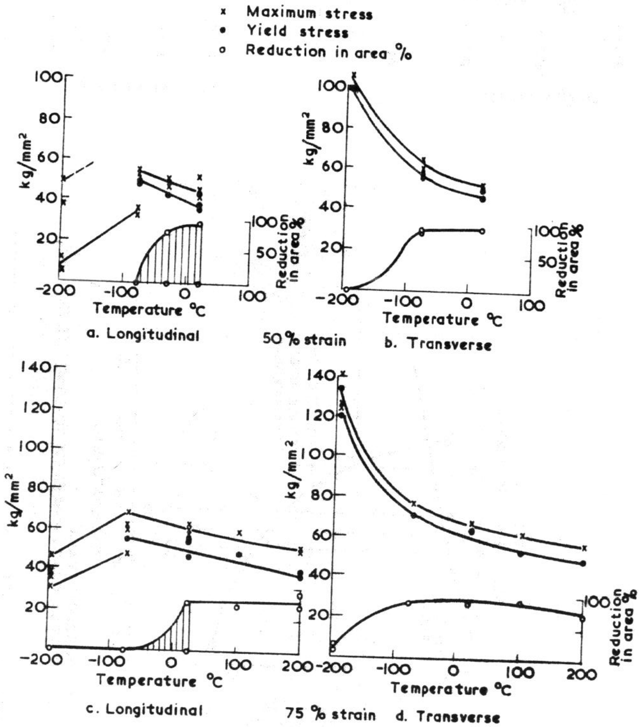



FIG.10. NPL PURE IRON STRAINED BY COMPRESSION VARIATION OF YIELD AND MAXIMUM STRESSES WITH TEMPERATURE.

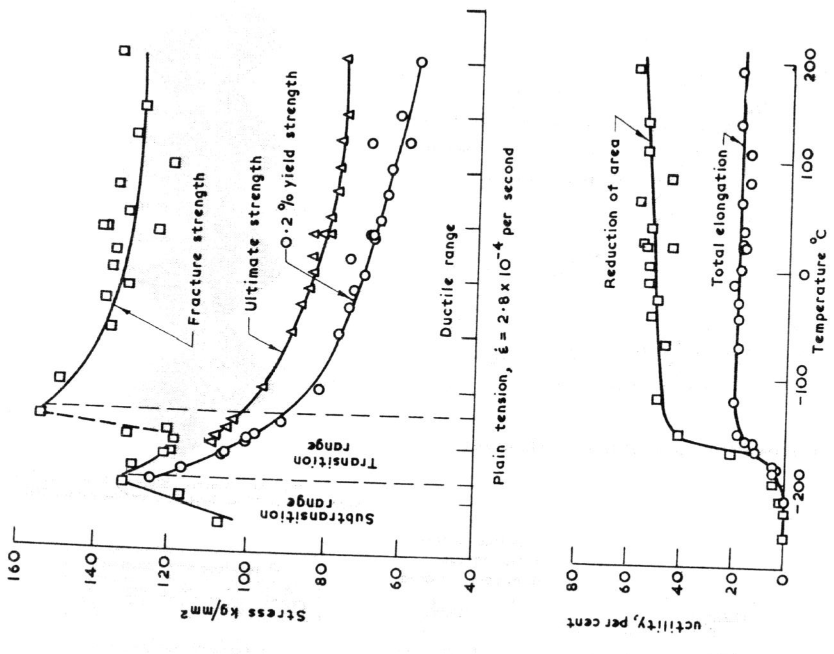



FIG. 13. TEMPERATURE-DEPENDENCE OF THE UNIAXIAL TENSILE PROPERTIES OF A Ni–Mo–V FORGING

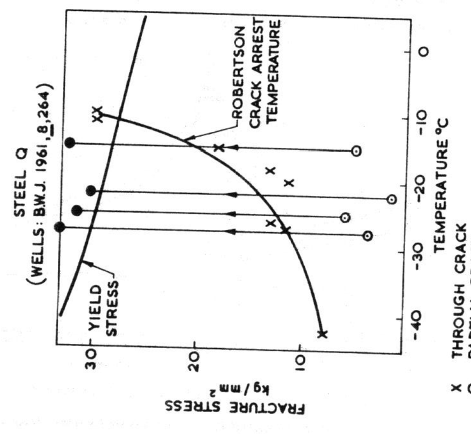
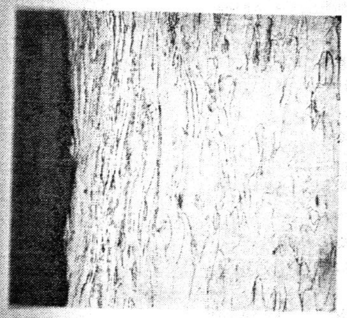



Figure 12. B.W.R.A. wide plate tests - typical results.



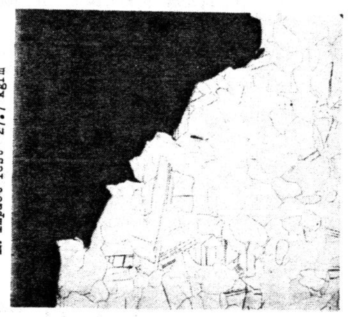





Temperature of Test -10°C

Energy Absorbed in Impact Test 27.7 kgfm

notch root




Temperature of Test -15°C

Energy Absorbed in Impact Test 10.8 kgfm

Iron -0.2% Manganese alloys. Microstructure at notch root of Charpy V-notch impact test pieces

Figure 15

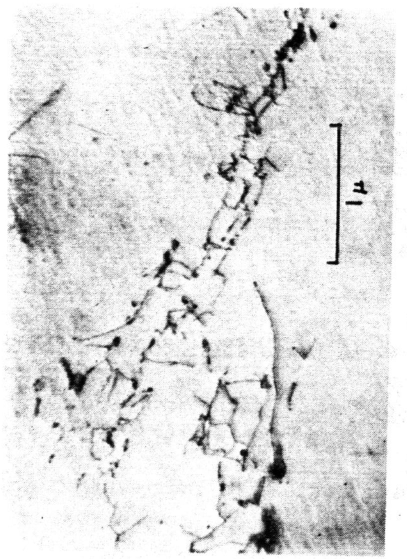



Figure 14

Dislocation structure in pure iron at completion of Lüders Extensions

Strain = 0.75% Grain Size = 0.2 mm dia.

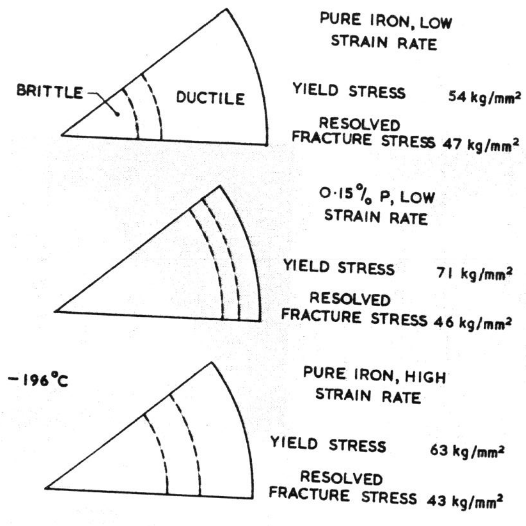
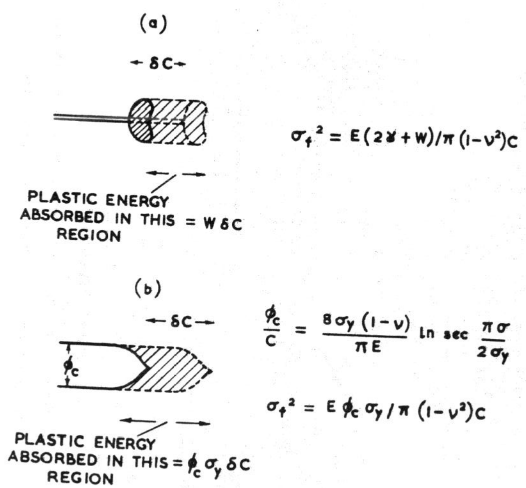



FIG. 16. TRANSITION ZONES BETWEEN SLIP AND CLEAVAGE FRACTURE IN IRON SINGLE CRYSTALS AT −196°C.




$$\sigma_f^2 = E(2\gamma + W)/\pi(1-\nu^2)C$$

$$\frac{\phi_c}{C} = \frac{8\sigma_y(1-\nu)}{\pi E} \ln \sec \frac{\pi\sigma}{2\sigma_y}$$

$$\sigma_f^2 = E\phi_c\sigma_y/\pi(1-\nu^2)C$$

Figure 17. Energy absorbed in crack propagation.





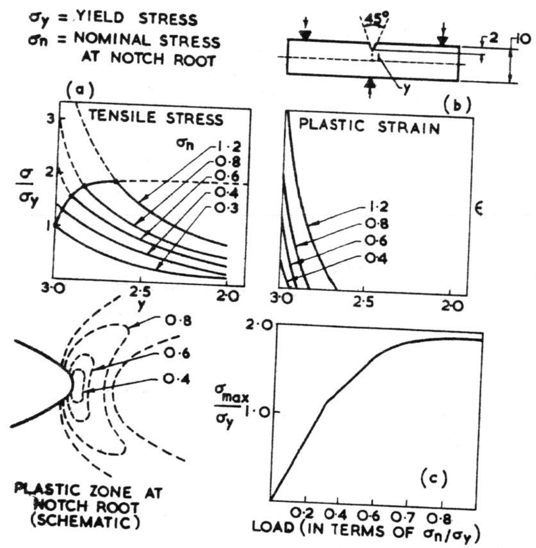



Figure 18. Stress and strains at notches stressed in 3 point bending.

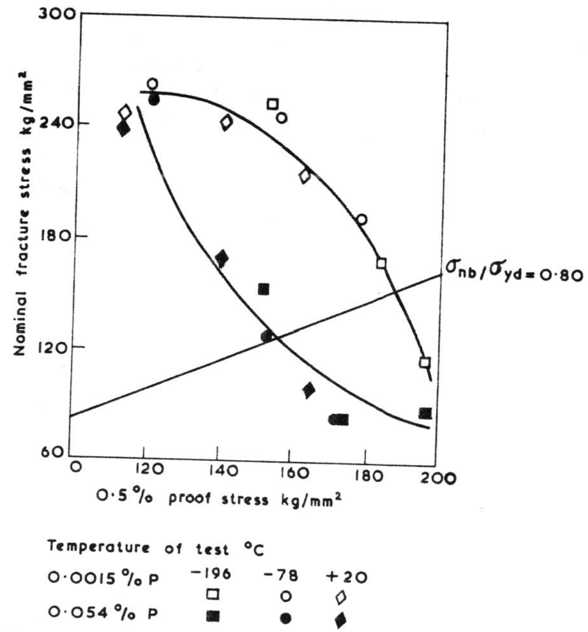



FIG. 19 NOMINAL FRACTURE STRESS PLOTTED AGAINST 0·5 % PROOF STRESS