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SSC-120

WHERE WE STAND IN DESIGN WITH BRITTLE FRACTURE

by

B. L. Averbach

SHIP STRUCTURE COMMITTEE

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ADDRESS CORRESPONDENCE TO:

SECRETARY
SHIP STRUCTURE COMMITTEE
U. S. COAST GUARD HEADQUARTERS
WASHINGTON 25, D. C.

February 23, 1960

Dear Sir:

During the December, 1958 convention of the American Society of Civil Engineers, held in New York, Professor B. L. Averbach of the Massachusetts Institute of Technology, an investigator for Ship Structure Committee Project SR-136, "Metallurgical Structure," presented a review of some of the highlights of the engineering features of brittle fracture together with a summary of the current theory. An interpretive report, SSC-120, of the above project entitled "Where We Stand In Design With Brittle Fracture" contains the above presentation.

Project SR-136 is being conducted under the advisory guidance of the Committee on Ship Steel of the National Academy of Sciences-National Research Council.

This report is being distributed to individuals and groups associated with or interested in the work of the Ship Structure Committee. Please submit any comments that you may have to the Secretary, Ship Structure Committee.

Sincerely yours,

E. H. Thiele
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

Serial No. SSC-120

Interpretive Report
of
Project SR-136

to the

SHIP STRUCTURE COMMITTEE

on

WHERE WE STAND IN DESIGN WITH BRITTLE FRACTURE

by

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Massachusetts Institute of Technology
Cambridge, Massachusetts

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Washington, D. C.
National Academy of Sciences-National Research Council
February 23, 1960

ABSTRACT

This review paper presents some of the highlights of the engineering features of brittle fracture together with a summary of some of the current theories, including the recent Cottrell Theory. In an attempt to interpret recent data in the light of the Cottrell theory it is concluded that some modifications in the theory are required to explain recent experimental results. Nevertheless, the dislocation picture of brittle fracture has been very helpful in providing a theoretical framework for the fracture mechanisms, and it is expected that these concepts will be continually developed with more theoretical and experimental work.

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INTRODUCTION

A brittle fracture may be defined as one that absorbs a relatively small amount of energy in propagation. Failure by the cleavage of a large fraction of the individual grains along (100) planes is the form of brittle fracture that occurs in the common structural steels, and this problem must be taken into account in the design of large continuous structures. Brittle cracks propagate through steel plates with about one-third the velocity of sound and with very little plastic deformation on a macroscopic scale.

The source of the failure is almost always a stress concentration in the form of a sharp notch introduced during construction, a sharp corner inherent in the design, or a crack produced by metallurgical damage in construction or service. Brittle cracks stop when the residual strain energy is no longer sufficient for propagation or when a particularly tough plate or structural feature is encountered.

It has been shown in many cases that localized yielding is associated with the start of a running crack because there is evidence of plastic deformation at the crack interfaces. The structure as a whole does not have to yield prior to failure, and the danger of a catastrophic failure increases as the size of the structure increases and as the service temperature is lowered.

Several procedures have been used to minimize the danger of brittle fracture. For example, sharp corners and other notches in the design have been eliminated. An outstanding example of this approach was the case of World War II cargo ships where the incidence of brittle failure was greatly reduced by redesigning a troublesome hatch corner. Another approach involves careful inspection during construction in order to reduce the introduction of flaws. This procedure has been particularly effective in pipelines. These measures have not eliminated the problem, however, and it is recognized that the last safety factor must be built into the steel. Considerable progress has been made in the understanding of what can be done to improve steels in this respect, but the mechanism of fracture on an atomic scale is not well understood, and the present day precautions may be only temporary measures on the way to a better solution.

Descriptions of service failures in ships,[1, 2, 3] in tanks and other structures,[4] and in large generator rotors[5] are available and will not be discussed here. General design criteria have also been summarized recently.[6] The engineering tests which have been used to investigate brittle fracture in steels have also been summarized,[3, 7] but a few of these are discussed in the next section in order to indicate the approaches which have evolved in the specification of steels to resist brittle failure. Recent experimental work on the mechanisms of cleavage failure and some of the current theories of brittle fracture are presented in the final section.

It should be noted that brittle failures do not necessarily have to occur by cleavage. Steels and other alloys which have been heat treated to very high yield strengths are capable of only very limited plastic flow. High applied and residual stresses may combine to produce the proper conditions for the nucleation of microcracks which can propagate with very low energy absorption because of the low ductility. Even less is known about the mechanisms of these fractures, and this type of high-speed tearing phenomenon will not be discussed. However, materials with these properties are finding increased use in thin-skin construction, and it is evident that a concerted research effort in this area is required.

Engineering Tests and Specifications

The appearance of the fracture surfaces in service cases of brittle failure has been well documented.[3, 7] A chevron pattern which points back to the origin of failure is frequently observed in failures occurring in mild steel, but this characteristic is not unique to brittle steel failures; it has also been observed in glass and plastics. The chevron pattern appears to be associated with a crack that proceeds in a discontinuous fashion, i.e. in a sequence of initiations ahead of the main crack followed by a breakdown of the intervening material to form a union with the principal crack. The crack front also is not straight, with the main fracture front extending in the center well in advance of the surface trace of the crack. In mild steel, it appears that cleavage microcracks are initiated ahead of the main crack, and it is this repeated initiation of microcracks which provides the discontinuous feature of the traveling crack. The problems of initiation and propagation of cleavage cracks thus tend to merge, since conditions for the initiation of a cleavage microcrack may be very close to those required for the discontinuous propagation of a gross crack.

One of the engineering approaches to this problem has been concerned with the search for tests that produce brittle failure, and the correlation of data from these tests with service behavior. Some of these tests appear to emphasize the initiation of cleavage cracks, others the stopping of a propagating brittle crack, but almost all of the tests involve the introduction of a notch and the observation of the onset of brittle behavior as the test temperature is lowered. Many tests also try to reproduce the dynamic aspects of a traveling crack, and this feature may be quite important in correlations with service behavior, since the yield point of steel is very much dependent on the strain rate. Each of these tests emphasizes different features of the brittle-fracture phenomenon, and it is not surprising that they frequently evaluate differently the ability of a material to resist brittle failure. The tests almost always define a transition temperature below which brittle fracture occurs under the test conditions, and these critical temperatures are used in two ways: (1) in direct comparison with the temperature for service failure of the same material and (2) in an evaluation of the effectiveness of various metallurgical variables in lowering the transition temperature. It is now generally agreed that a decrease in transition temperature in any of the tests probably reflects an improvement in the material, even though all of the tests may not indicate an equal improvement.

The engineering tests have been summarized recently,[3, 6, 7] but the V-notch Charpy impact test deserves special mention because of its widespread use. A typical set of impact data is shown in Fig. 1 for specimens taken from a 3/4-in. thick plate of rimmed steel. The energy absorption changes considerably over a rather small temperature interval, and the fracture appearance changes correspondingly from the fibrous appearance associated with the high-energy region to the specular appearance which results from the cleavage of many grains along (100) crystallographic planes. The temperature at which 15 ft-lb is absorbed is frequently defined as a ductility transition temperature T_d and is associated with the temperature region where

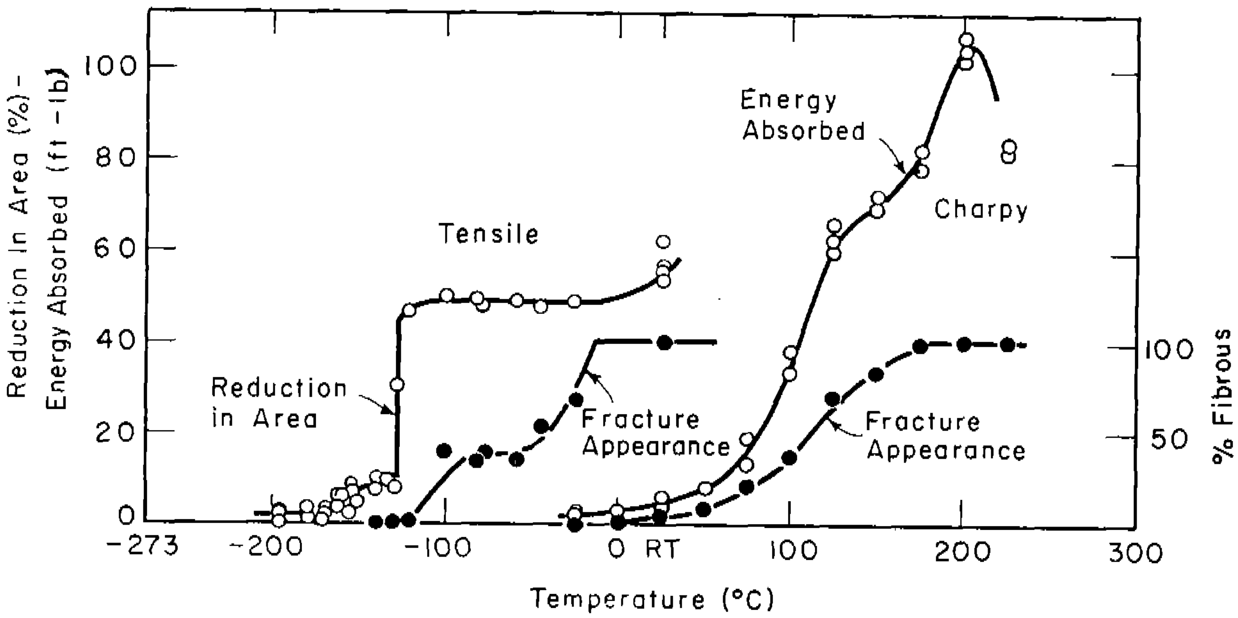

Fig. 1. Tensile and V-Notch Charpy Impact Transition of Large Grain (ASTM GS No. 4) Project Steel E: 0.22 C, 0.36 Mn, 0.002 Si

cleavage cracks initiate the brittle fracture.

The 15 ft-lb transition temperature is based on an investigation of the plates in World War II cargo ships that exhibited brittle fracture. It was shown that all of the plates wherein brittle fractures started absorbed less than 10 ft-lb at the failure temperature,[1,2] whereas the end plates consistently absorbed more energy. It should be emphasized that this correlation refers to the semikilled and rimmed steels involved in that particular ship construction; subsequent data have indicated that the energy level for the ductility transition may have to be raised to about 20 ft-lb for killed steels and some alloy grades.

Another approach to the V-notch Charpy test defines a fracture transition temperature T_f for 50 per cent fibrous appearance, which satisfies a condition where brittle fracture may be initiated by a ductile crack. This situation might occur in a structure that already contains a crack that was extended because of high local stresses at the tip. As long as the crack proceeds by shear, a large amount of local deformation will be required (in mild steel), considerable energy will be needed, and the crack will move slowly. When the crack reaches a critical size, which depends on the temperature, the stress field in front of the crack can become large enough to initiate cleavage microcracks which could join the original shear crack by cleaving and tearing the intermediate material. Conversely, at a given temperature, if the material were above the fracture transition temperature and were presented with a running brittle crack, it could stop the crack because of the high energy required to shear a large portion of the material. It is apparent that the fracture transition T_f is higher than the ductility transition T_d.

The anticipated service conditions are quite important in deciding which transition temperature to consider. For example, in the case of a submarine hull, it might be expected that severe local deformation could result in a shear crack. It would be important to prevent this shear crack from turning into a rapidly propagating cleavage crack and thus fracture transition criteria would be of prime importance. Similarly, a large generator rotor might contain internal flaws which remain undetected because of the massive size of the forging. One of these cracks could reach the critical size during service or overspeed tests and eventually propagate as a brittle crack. On the other hand, a storage vessel might be constructed under conditions where there was sufficient inspection to avoid the introduction of major flaws. At low ambient temperatures, however, it would be necessary to avoid the initiation of a cleavage crack in the vicinity of a stress concentration, and the ductility transition temperature could be used as a basis for choosing the material.

Both philosophies are used in merchant ship building. The specifications of Lloyd's Register of Shipping[8] now include a class of plate, which is specified for key regions in the ship, requiring 35 ft-lb at 0°C. In time, the requirements probably will also include a 30 per cent fibrous appearance.* This specification was chosen on the basis of service data for plates that successfully stopped running brittle cracks. American practice specifies the composition and steelmaking practice of all of the material in the ship according to the thickness of the plate. The V-notch Charpy characteristics of current ship plate have been monitored, and it has been shown that the ductility transition temperatures of the present day material are considerably lower than those of the World War II plates. The composition and practice are designed to produce material with a low ductility transition in the heavier plates, which are presumably

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*At present the requirement of 30 per cent fibrous appearance is being waived, but the data on the fibrous appearance at 0°C are required.

at the regions of higher stress. There has been a continuing effort to improve the quality of ship plate, as may be seen from a recent review of the work carried out under the guidance of the Committee on Ship Steel.[9]

The engineering tests are valuable for correlation with service behavior, and they have formed the basis of the investigative procedures used to select improved steels. Each type of service requires a new correlation, however, and the multitude of correlations using tests of various kinds for a variety of steels has made it desirable to pursue this problem on a fundamental basis. The principal objective of the basic research is an understanding of the atomic mechanisms of fracture so that the resultant concepts can be applied to a broad spectrum of materials and problems. Several recent theoretical and experimental approaches are discussed in the last section of this report.

Metallurgical Factors in Brittle Behavior

Many of the steelmaking variables have been evaluated for the weldable mild structural steels in terms of impact transition temperatures. It has been shown that the transition temperature is lowered as the carbon content is lowered and the manganese content is increased in structural steels,[3] and thus several low-carbon high-manganese steels have been introduced because of their improved resistance to brittle failure. It is significant to note that alloy additions other than manganese either raise or have no effect on the transition temperatures of these steels. Aluminum-treated steels made to fine-grain practice are superior to rimming or semikilled grades, and their advantage may be further enhanced by normalizing. Investigations have shown that the transition temperature is raised as the ferrite grain size is increased, and there is evidence to indicate that the transition temperature for a steel cooled slowly from the annealing temperature is higher than for a steel which has been rapidly cooled.[10]

There is a limit, however, to the improvement that may be attained in ferritic steels by these means, and even then, the basic mechanisms by which these improvements are attained are not well understood. Furthermore, these empirical concepts do not extrapolate readily to higher alloy steels or to other types of microstructure, such as bainite or martensite. Experience with the coarse bainitic microstructures which have been used in some large generator rotors has indicated that they are susceptible to brittle cleavage failure despite the high alloy content of the steel. On the other hand, fine bainitic or martensitic microstructures of the same composition may have rather low transition temperatures, and it is probable that the improvement may be associated with the small size of the ferrite regions in the latter microstructures. The metallurgical factors at play in the brittle behavior of the high strength steels have not been investigated in the same detail as the weldable ferritic grades, and it seems as if data on the former will now be required in increasing quantity for some of the new structural applications. One of the characteristic features of brittle cleavage failure is that it does not occur in metals with face-centered cubic crystal structures. Thus, the austenitic stainless steels do not exhibit brittle failure, and this material is frequently used for structures which must operate at temperatures below -80 C.

Atomic Mechanisms of Cleavage Fracture

Dislocation concepts are used in most of the current theories of brittle fracture. These theories, which have been reviewed recently,[11, 12, 13] will be discussed here along with some recent pertinent experimental work. Most of the fundamental experimental work uses unnotched bend or tensile specimens at low temperatures in order to simplify the stress pattern. Figure 1 shows that the tensile transition temperatures are considerably lower than the Charpy V-notch impact transitions but that the principal features of a loss in ductility and a change in the fracture appearance are common to both tests. The influence of notch and impact loading on the transition temperature is quite evident.

The relationship of the ductility and fracture parameters to the tensile properties is shown in Fig. 2. Several modes of fracture are observed at temperatures below 20 C. At temperatures down to about 0°C, the fracture is entirely ductile, with considerable necking prior to fracture. Between 0° and -120 C, a ductile crack

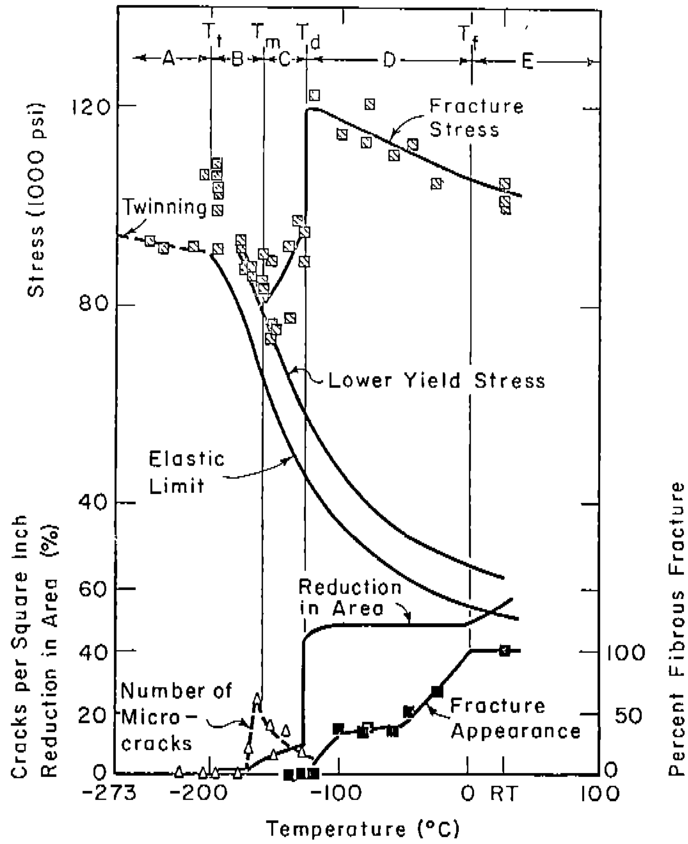

Fig. 2. Relation of Tensile Properties to Fracture Appearance and Microcracks for Coarse Grain (ASTM GS No. 4) Project Steel E

starts at the center of the specimen and propagates to a fraction of the diameter, with the remainder of the fracture occurring by cleavage. The size of the ductile crack required to initiate the cleavage failure becomes smaller as the temperature is lowered (see Fig. 3), but the overall ductility is still quite high since necking is required to initiate the ductile crack. This temperature range corresponds to the fracture transition. At temperatures below -120 C, the mode of fracture changes. The first observable crack is a cleavage microcrack of the order of one grain diameter in size, and the final failure has the gross appearance of 100 per cent cleavage. It is important to note that the mechanism of initiation changes completely in this region and that -120 C corresponds to our previous definition of a ductility transition.

It should be noted from Fig. 2 that the lower yield stress is reached prior to fractures that are classed as entirely brittle. In addition, a well-defined elastic limit, corresponding to an observable plastic strain of 1×10^{-6}, is observed in all cases prior to the discontinuous yield. A reduction in area is observed at temperatures as low as about -170 C, even though the fracture appearance is entirely cleavage, and an examination of the crack surfaces and the microcracks shows that considerable local plastic deformation is associated with the ends of microcracks and with the crack interfaces. A typical micro-

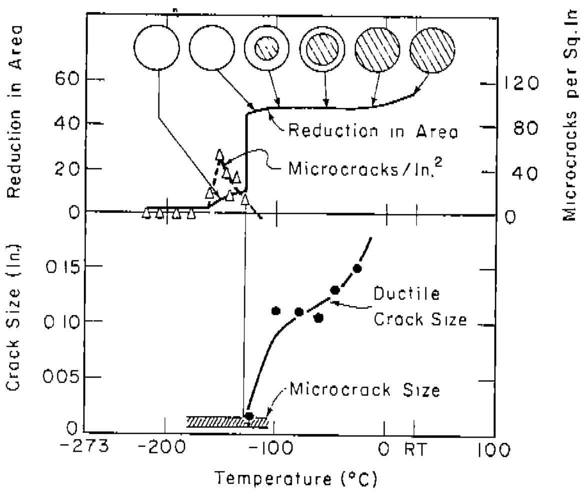

Fig. 3. Fracture of Coarse Grain (ASTM GS No. 4) Project Steel E.

Fig. 4. Microcrack at surface of unfractured E steel specimen. 250X

crack can be seen in Fig. 4, which shows the surface of a cylindrical specimen that was electropolished prior to testing and unloaded prior to fracture after stressing to the yield point. The cleavage microcrack and the attendant distortion are quite evident.

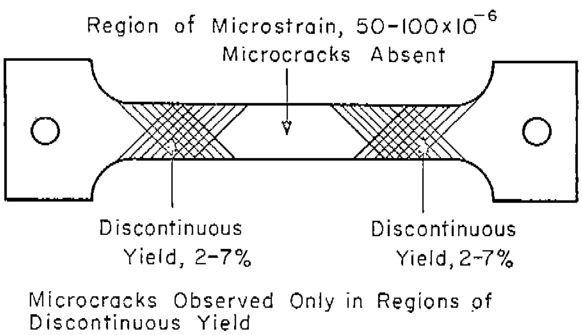

Fig. 5. Location of Microcracks in Flat Tensile Specimen.

The relationship between cleavage microcracks and discontinuous yielding was demonstrated by Owen et al[14] in an experiment where Lüders bands were passed along thin, flat, prepolished steel specimens at liquid nitrogen temperature. The Lüders bands were passed only part way along the specimen, and then the specimen was unloaded and examined for microcracks. The microcracks were very similar in appearance to the one shown in Fig. 4; the location of the cracks is shown schematically in Fig. 5. The experiment was repeated recently with an additional variation: The specimen was heated to room temperature after unloading and given an additional tensile elongation in order to reveal microcracks that might have closed partially on unloading. The results were the same in both cases--microcracks were observed only in the region that had yielded discontinuously. They were not observed in the region between the Lüders bands, even though this region had suffered a plastic strain of $50-100 \times 10^{-6}$. It is thus evident that the cleavage microcracks require a local deformation of the order of the discontinuous yield prior to initiation but that the overall extension of the specimen may be very small if the Lüders band does not travel far before the specimen fractures. Cleavage fracture is thus closely associated with yielding, and the same factors that influence the yield point would be expected to influence the cleavage fracture stress. This has been generally observed.

Several general experimental features must be considered by the theoretical treatments of brittle cleavage failure.

1. Cleavage failure occurs in body-centered cubic and some hexagonal crystal structures but not in face-centered cubic materials. The tendency for cleavage failure is greatly influenced by the presence of interstitial elements such as carbon and nitrogen.

2. Plastic flow precedes the cleavage crack, and the surfaces of the interface exhibit considerable local deformation.

3. The tendency toward brittle fracture is accentuated at low temperatures. The yield strength of iron and steel rises sharply with decreasing temperature below 20 C, and the brittle behavior is undoubtedly associated with this rise in yield strength. Face-centered cubic metals show only a small rise in yield with decreasing temperature and are not subject to cleavage failure.

4. Cleavage failure is favored by high strain rates. Recent work[15] has indicated that the discontinuous yield in steel does not occur instantaneously on application of the load but that there is a delay time which depends strongly on the temperature, varying from about 10^{-4} sec at room temperature to about 10^{3} sec at liquid nitrogen temperature. A high strain rate can thus raise the yield stress in the same way as does a decrease in temperature.

One of the early theoretical approaches to the problem of brittle failure was made by Griffith who considered the stress required to propagate a sharp crack in an elastic medium. For a two dimensional crack of length 2C, the tensile stress σ required to propagate the crack is given by:

$$\sigma = \left[\frac{2E\gamma}{\pi(1-\nu^2)C}\right]^{1/2} \tag{1}$$

where E is Young's modulus, ν is Poisson's ratio, and γ is the specific surface energy (ergs/cm^2) of the new crack surface. Cracks of the order of 10^{-4} cm are required to account for the observed brittle fracture strength of steel.

The data shown in Fig. 2 indicate that one of the primary assumptions associated with Eq. 1 is not fulfilled. It is evident that plastic flow occurs prior to cleavage failure and that therefore the elastic stress concentration factor used in deriving the equation is no longer valid. Orowan modified the Griffith equation by recognizing that considerable plastic work is associated with the passage of a cleavage crack in ferritic steel. This consideration leads to the equation:

$$\sigma = \left[\frac{2Ep}{\pi\ C}\right]^{1/2} \tag{2}$$

where p is now the effective surface energy, including the plastic work. This effective surface energy has been estimated to be about 10^6 ergs/cm^2. This concept leads to the assumption of the presence of cleavage microcracks of length 2C of the order of the grain size d (Fig. 4); the presence of such microcracks prior to fracture has been demonstrated by Low[16] and by Owen et al.[14] The principal difficulty in this equation is in accounting for the temperature dependence, and it must be assumed that the term p varies strongly with temperature. In addition, there is little indication in this picture for the need for plastic flow prior to cleavage. It would ap-

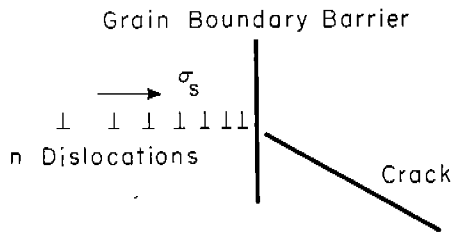

Fig. 6. Dislocation Pile-up Relieved by a Crack (after Stroh).

pear that this approach accounts for some of the macroscopic features of the phenomenon and thus represents only a part of the picture.

Stroh[11] introduced the concept that a dislocation pile-up in a slip band contributes to the conditions necessary for the initiation of a cleavage crack. Zener[17] had considered earlier that a blocked slip line would resemble a freely slipping crack under a shear stress, and Stroh considered the dislocation pile-up as producing a stress concentration equivalent to that of a Griffith crack. If such a dislocation pile-up along a slip line against a grain boundary barrier (Fig. 6) produces a crack along a cleavage plane, the condition for two dimensional crack formation can be derived simply by considering that all of the strain energy associated with the dislocations is relieved by the formation of the crack. This gives directly:

$$\sigma_s nb = 2\gamma \tag{3}$$

where σ_s is the applied shear stress, n the number of dislocations in the pile-up, b the Burgers vector of the dislocation, and γ the specific surface energy of the crack. This equation is similar to Stroh's with the exception of the numerical factor 2, which is $\frac{3\pi^2}{8}$ in the more exact calculation.

The influence of grain size may be seen in the following way. The dislocation pile-up is considered to occur over a region d/4, and thus the local shear strain is given by $\epsilon_s = 4nb/d$. The shear strain is given approximately by σ_s/G, where G is the shear modulus. Introducing these into Eq. 3, we see:

$$\sigma_f \approx 2\sigma_s = K\,d^{-1/2} \tag{4}$$

where σ_f is the fracture stress and $K = (32\gamma G)^{1/2}$. (Stroh's more exact calculation gives $K = [6\pi\gamma G/(1-\nu)]^{1/2}$.) In order to correlate the theoretical calculations with the observed dependence of the fracture stress on the grain size, a frictional stress σ_o is introduced to give the final equation:

$$\sigma_f = Kd^{-1/2} + \sigma_o \tag{4}$$

This picture has several difficulties. The cleavage crack is pictured as being produced by shear stress alone, but it is evident from other data that the presence of normal tensile stresses is very effective in raising the cleavage transition. In order for the dislocation pile-up to act as a Griffith crack, it is necessary that the surrounding grains do not flow. This implies that all of the dislocation generators near the pile-up must be pinned by interstitial atoms so that slip is prevented and cleavage occurs. The dislocation pinning mechanism is strongly temperature dependent, and this would probably provide the proper temperature dependence required to fulfill the equation except that it is difficult to see why slip does not occur rather than cleavage in the surrounding grains. In fact, an expression of the form of Eq. 4 describes the influence of grain size on yield stress σ_y with greater accuracy than it does the influence of grain size on the fracture stress. Figure 2 complicates Eq. 4 further by showing that the yield and fracture stresses in the 100 per cent cleavage region have very similar values, as well as by showing a plot of the frequency that finite cleavage microcracks of the order of a grain diameter are formed at the yield stress. The observed fracture stress is thus for a material containing numerous microcracks, and it would appear to be of doubtful fundamental significance. In addition, the Lüders band experiments of Owen et al[14] show that the microcracks occur only in the region of discontinuous deformation where the plastic flow is very heterogeneous and much more complex than the simple pile-up shown in Fig. 7.

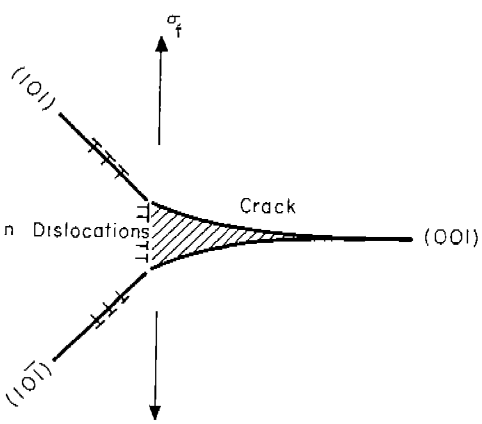

Fig. 7. Cleavage Crack Formed by Coalescence of Dislocations on Intersecting Slip planes. (after Cottrell)

Cottrell[12] has recently reconsidered the problems associated with dislocation pile-ups. He proposes that dislocations traveling on two intersecting (110) slip bands could produce a dislocation pile-up in the form of a giant dislocation with a Burgers vector nb.

This could act as an atomic knife along (100) cleavage planes in b.c.c. material. Such a pile-up is shown in Fig. 7. In this case, energy considerations similar to those used in Eq. 3 give:

$$\sigma_f nb = 2p \tag{5}$$

It should be noted that the normal fracture stress σ_f is considered here rather than the shear stress. The effective surface energy term p includes the irreversible plastic work required to form the cleavage crack.

The grain size dependence is obtained as follows. It is assumed that the fracture stress σ_f is approximately equal to the yield stress σ_y and that the yield stress has a temperature dependence of the form

$$\sigma_y = K_y d^{-1/2} + \sigma_{yo}$$

where σ_{yo} is the frictional component of the yield stress. The shear stress at yield is $\sigma_s = \sigma_y/2$. The dislocation displacement nb is then given approximately by

$$nb = (\sigma_s - \sigma_i)\, d/G$$

where σ_s is the applied shear stress and σ_i is the frictional component of the shear stress. Introducing these relationships into Eq. 5, we get:

$$\sigma_s K_y d^{1/2} = \beta G p \tag{6}$$

The factor β is approximately unity in the tension test. A comparison with experiment gives a value of 18,000 ergs/cm^2 for p; this is about 10 times the value of the surface free energy γ.

Equation 6 is applied in an interesting way. If the left hand side is smaller than the right hand side, there is insufficient energy to propagate a cleavage crack beyond a certain length. In this case, a cleavage microcrack may form but it will not grow. When the left hand side is larger than the right, the yield stress is high enough to make the cleavage crack grow to complete failure. A transition point is thus defined, and Cottrell has assumed that this is the ductility transition T_d. The largest stable microcrack may be calculated from the Cottrell equations, and this turns out to be:

$$C = \frac{d}{2\pi(1-\nu)} \cdot \frac{(S-1)}{S} \qquad (7)$$

where $S = \sigma_s/\sigma_i$. The factor S varies with grain size, but a typical value for steel with $d^{-1/2} = 6\ mm^{-1/2}$ is S = 3. Taking $\nu = 1/3$, this gives C = d/6, or $C \approx d$. Figure 4 shows an example of such a microcrack.

Figure 2, which shows the temperature regions wherein different fracture mechanisms operate indicates, however, that the transition temperature described by Eqs. 5 and 6 may refer to a temperature somewhat lower than the ductility transition temperature T_d. In region E, above the fracture transition T_f, the fracture is entirely by shear. In region D, between T_f and T_d, the fracture is started by a shear crack and is completed by cleavage. In this region the size of the shear crack required to initiate a cleavage failure decreases as the temperature decreases (see Fig. 3). In region C, which is between T_d and T_m, we propose that the failure is initiated by cleavage microcracks and that these microcracks join together to form the final fracture. The microcrack curve shows the number of stable microcracks prior to fracture in this region, and it is evident that a sharp maximum is reached at T_m, which we shall call the microcrack transition temperature. T_m corresponds also to the sharp valley in the fracture stress where the yield stress σ_y and the fracture stress σ_f are about equivalent. In region C, the yield stress rises as the test temperature decreases whereas the fracture stress falls. The falling fracture stress is apparently associated with the increasing number of stable microcracks that are formed at the yield stress, and the fracture would appear to occur by the linking of these microcracks. Although the fracture appearance is 100 per cent cleavage on a gross scale, microscopic observations have shown that there is considerable local plastic flow associated with the linking of the microcracks.

Region B is defined by the limits T_m and T_t. Below T_t another mechanism of brittle failure is observed. The microstrain elastic limit is not observed, and the first indication of deformation is mechanical twinning which leads immediately to cleavage failure. While large local deformations are associated with these twins, the details of the fracture mechanism have not been observed. It is significant to note, however, that stable microcracks have not been found below the microcrack transition T_m. It appears, therefore, that the transition calculated by Cottrell applies to T_m and not to T_d, since the two conditions required in the calculation are satisfied: (1) $\sigma_f \approx \sigma_y$ and (2) stable microcracks are observed above T_m and are not observed below. These conditions are only met in region B and are not satisfied in region C. In addition, it is unlikely that the fracture stress in region C has the assumed grain-size dependence. The Cottrell theory appears to fit the microcrack transition temperature T_m quite well, and Eqs. 5 and 6 provide a convenient way of summarizing the influence of steel variables on the brittle fracture problem.

The overall picture is still probably incomplete. The role of mechanical twinning in region A is not understood, and the mechanism of failure may be further complicated in materials with more complex microstructures. The energy factor p is also troublesome since it is now obtained from empirical correlations with the data; its grain-size and temperature dependence have not been explored. The cleavage facets also show peculiar "river" patterns which have been interpreted as showing the location of intersecting dislocations, but the role of the localized plastic deformation in linking up the cleavage facets is not well understood.

The atomic mechanisms of fracture were discussed at an international conference held in Swampscott, Massachusetts in April 1959, and the published proceedings[18] indicate the current theoretical and experimental approaches. The role of microcracks in the final macroscopic fracture is not clear and the nature of the fundamental materials constants that govern the tendency to exhibit easy cleavage fracture is still obscure. The Griffith flaw theory modified by dislocation concepts is still basic to most of the thinking, but there seem to be many gaps in understanding the mechanisms that operate between the initial postulated dislocation pile-ups and the final fracture. Nevertheless, the dislocation picture of brittle failure has been very helpful in providing a theoretical framework for the fracture mechanisms, and it is expected that these concepts will be continually developed with more theoretical and experimental work.

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