AN EVALUATION
OF CURRENT KNOWLEDGE OF THE
MECHANICS OF BRITTLE FRACTURE

by

D. C. DRUCKER
Brown University

Including
Papers and discussions presented at
Conference on Brittle Fracture Mechanics
held on October 15 and 16, 1953
at Massachusetts Institute of Technology

Evaluation prepared for and conference sponsored by

NATIONAL RESEARCH COUNCIL'S
COMMITTEE ON SHIP STRUCTURAL DESIGN

Advisory to

SHIP STRUCTURE COMMITTEE

Division of Engineering and Industrial Research
National Academy of Sciences - National Research Council
Washington, D. C.

May 17, 1954

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ADDRESS CORRESPONDENCE TO:

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

17 May 1954

Dear Sir:

The attached report contains the papers presented at a special conference held at the Massachusetts Institute of Technology in October 1953, sponsored by the National Research Council's Committee on Ship Structural Design. In addition, the first paper in this report, "An Evaluation of Current Knowledge of the Mechanics of Brittle Fracture", prepared by Professor D. C. Drucker of Brown University at the request of the Committee on Ship Structural Design, is based on a critical analysis of the material presented at the conference as well as a general literature survey in this field.

The conference and the Drucker paper were for the purpose of assisting the Committee on Ship Structural Design in advising the Ship Structure Committee in the field of brittle fracture mechanics.

The M.I.T. conference, the Drucker analysis and the printing of this report were financed by the Ship Structure Committee as part of its program to improve the hull structures of ships through research. This report is distributed by the Ship Structure Committee at no cost to individuals and organizations concerned with the design, construction or operation of ships, and others who may find it of value.

Comments concerning this report are solicited and should be addressed to the Secretary, Ship Structure Committee.

Very truly yours,

K. K. Cowart

K. K. COWART
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

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SHIP STRUCTURE COMMITTEE

under

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National Academy of Sciences-National Research Council
Washington 25, D. C.

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FOREWORD

The Committee on Ship Structural Design of the National Academy of Sciences-National Research Council, in early 1953, assumed advisory guidance of a research program to be sponsored by the Ship Structure Committee and broadly entitled "Brittle Fracture Mechanics". A number of areas in the field had been explored, some of them intensively, while others have yet to be approached. In order to insure selection of the most promising avenues for further analytical and experimental study, it was necessary to evaluate the present status of our knowledge of the subject.

At the request of the Committee on Ship Structural Design, Professor D. C. Drucker of Brown University undertook a review of the literature on brittle fracture mechanics, and a Conference to survey the present status of the knowledge was held at Massachusetts Institute of Technology on October 15 and 16, 1953. At the Conference particular consideration was given to the brittle behavior of mild steel, with emphasis on the mechanics of fracture, considering the influence of size, geometry, stored elastic energy, and so forth.

There follow the papers presented at the Conference, written discussions submitted by persons attending the Conference, and in one case, a comment forwarded by an individual unable to be present. Professor Drucker's Evaluation, which precedes these papers and discussions, is based upon his survey of the literature ~~and on the papers~~ and on the papers and discussions at the Conference. At a meeting held on February 17, 1954, the Committee recommended the initiation of research in each of the three general areas suggested for study by Professor Drucker.

The Committee realizes that the material contained herein represents a number of diverse approaches to a confusing subject. Its value to the practicing engineer may therefore be questioned. As noted above, the primary purpose in assembling this information was to guide the Committee in establishing recommendations for future research in the field; consequently, this report may well be of limited value to those not actively concerned with brittle fracture mechanics research.

Persons attending the Conference on October 15 and 16, 1953, are listed below:

Conference Chairman:

Professor C. R. Soderberg, Massachusetts Institute of Technology

Contributors:

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Mr. F. J. Feely, Jr., Standard Oil Development Company
Professor R. A. Hechtman, University of Washington
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Professor N. M. Newmark, University of Illinois
Professor E. R. Parker, University of California, Berkeley
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Discussion Leaders:

Professor D. C. Drucker, Brown University
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Members, Committee on Ship Structural Design:

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Professor L. S. Beedle, Lehigh University
Dr. H. C. Boardman, Chicago Bridge and Iron Company
Vice Admiral E. L. Cochrane, USN (Ret.), Massachusetts Institute of Technology
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Mr. J. M. Crowley, Office of Naval Research
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Dr. E. Wenk, Jr., David Taylor Model Basin

AN EVALUATION OF CURRENT KNOWLEDGE OF THE MECHANICS OF BRITTLE FRACTURE

by

Professor D. C. Drucker
Brown University

ABSTRACT

An analysis of the brittle failure problem(1--10) and test program was undertaken at the request of the Committee on Ship Structural Design of the National Academy of Sciences-National Research Council. Much of the literature was read, and the papers and discussion presented at the Conference* of October 15 and 16, 1953, were studied carefully. (11--20)

An explanation is given of the trend of results found in the many static model tests based on the concepts and some solutions of the mathematical theory of plasticity. Transition temperature, size effect, initiation and propagation of cracks are discussed qualitatively and on dimensional grounds. The conclusion is reached that future model tests should be made at the operating temperatures of the prototype and at stress levels encountered in practice if fundamental information is to be obtained. Load, energy, or appearance criteria for static or impact tests appear to be interpretable only in terms of such basic tests or information from the prototype. Each new material or variant of an existing one should be subjected to the basic tests before an acceptance test standard can be relied upon.

INTRODUCTION

So much has been written on the brittle failure problem that one is now in the position of writing summaries of summaries(1--10). Nevertheless, large segments of the engineering population are still unaware of the menace. It is indeed difficult to accept the fact that a material like mild steel, which displays great ductility as a standard tensile test specimen, at room temperature, exhibits almost no ductility as a notched bar at sub-zero temperature. Even more incredible is the fact that at 40°F ships and other large structures

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*Conference to Evaluate Current Knowledge of Brittle Fracture Mechanics, Committee on Ship Structural Design, National Academy of Sciences-National Research Council, held at Massachusetts Institute of Technology.

have snapped apart at moderate and low loading.

Just a little more than ten years ago, so few believed the phenomenon to be real that much effort went into reproducing brittle fractures in the laboratory. These early tests served a most important purpose from the educational view point. They helped greatly in the redesigns which reduced the number of failures to tolerable proportions and also led to the requiring of more suitable steel for shipbuilding. Somewhat unfortunately, these tests were so interesting in themselves that many investigations developed, labeled in some cases as brittle fracture studies, which were actually not relevant to the main question. All were of extreme value for structural design and for an understanding of plastic action, but they tended to becloud the issue.

Structural failures are so spectacular and potentially so dangerous that every effort must be made to reduce their probability to zero under foreseeable conditions. Two related aspects are involved: design and material. Both have their limitations in terms of cost and practicability. The one extreme of a design which is so well proportioned everywhere that a poor material will still be free to flow plastically where and when needed is ruled out by the requirements of fabrication. The other extreme of a material so notch-insensitive that design and fabrication do not matter is ruled out at present on the basis of cost. A compromise is needed calling for attention to details of design and construction but also demanding the best material possible. Much remains to be done on the question of material to achieve a greater margin of safety than presently available with improved ship steel. As Dr. Hoyt stated at the Conference, metallurgists will eventually have to make a proper material.

A good case can be made for considering the problem purely one of developing a suitable material. Intersections of members, cut-outs of various shapes and sizes, arc strikes, imperfect welds, cracks--all conspire to produce local regions in which freedom to deform is restricted and from which brittle action may ensue. The best design practice may be nullified by a repair job done the easy way, perhaps the only possible way, with the materials on hand. Although such an argument has much merit, it is always true that the better the design, the less the probability of failure for any material. Good design requires thought but need not cost more.

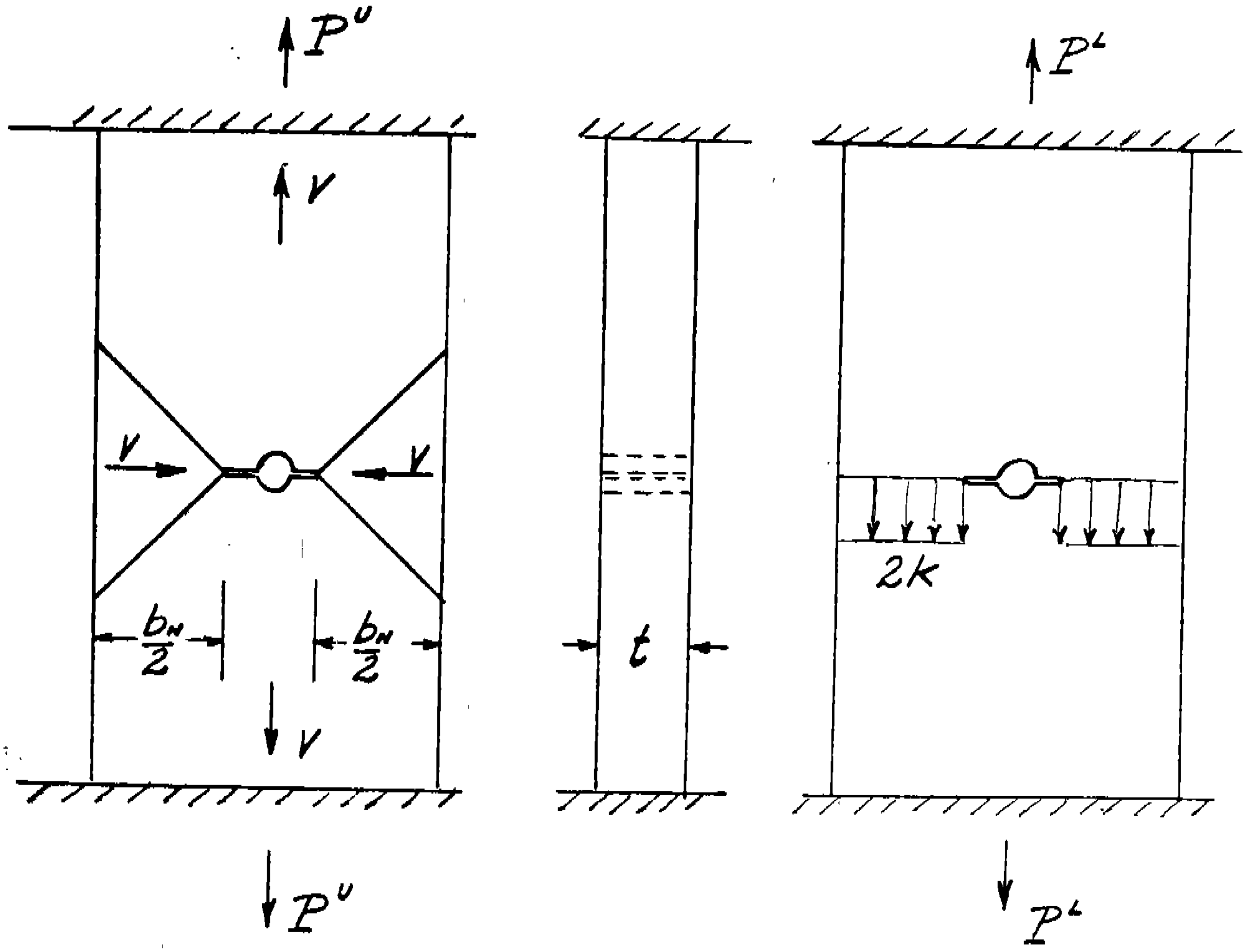
No matter which philosophy one espouses, the problem requires a thorough understanding of the mechanics of fracture. Such an understanding is still a long way off although many significant forward steps have been taken. The purpose of this paper is to offer a partial summary and evaluation of present knowledge from the vantage point and probable ignorance of an outsider and to suggest possible directions for future research.

EXPLANATION OF STATIC TEST RESULTS ON BASIS OF PLASTICITY THEORY

There is an extensive literature on tests of small and large specimens, including internally and externally notched bars and sheets, tubes under interior pressure and axial pull, and models of hatch corners. A number of conclusions have been drawn and checked repeatedly. One is that a concentric load increased slowly from zero does not cause failure of a notched specimen at a nominal stress below the yield strength of the material. The explanation is simple and has already been given clearly by Wells(21). Plasticity theory rather than elasticity governs. General yielding must take place to some extent to permit enough plastic deformation to occur at the root of the notch to cause failure.

As the application of plasticity theory to many of the problems is rather recent, not all the statements appearing in print are completely correct. An understanding of the differences between externally notched and internally notched sheets, notched bars, and notched beams is greatly facilitated by simple solutions based on ideal plasticity(22,23). The pertinent features, as far as loads are concerned, show up despite the neglect of work-hardening. In what follows, maximum shearing stress will be taken to control yield rather than distortional energy or another criterion because it gives reasonable results and is easy to use. Exact limit loads are often difficult to compute, but upper and lower bounds can be determined for many practical cases without too much trouble. Space does not permit more than a statement of the theorem that a kinematically admissible velocity pattern gives an upper bound P^u on the general flow or limit load and an equilibrium solution which doesn't violate yield gives a lower bound P_L. For proof, details and illustrative examples, see References 24--26.

The difference between externally and internally notched sheets or rectangular bars is of real interest and practical importance. As shown by the kinematic scheme of Figure 1(a),



(a) Relative velocity on surfaces of sliding is $V\sqrt{2}$

$$2P^U V = 4\frac{b_N\sqrt{2}\,t}{2}\cdot\sqrt{2}\,V\,k$$

$$P^U = 2kb_N t$$

(b)

$$P^L = 2k\;2\frac{b_N}{2}t$$

$$P^L = 2kb_N t$$

FIG. 1 INTERNAL NOTCH

which gives an upper bound, and the static field of Figure 1(b), which gives a lower bound, the limit load is the tensile yield stress multiplied by the net area

$$P = 2kb_N t \qquad [1]$$

where k is the yield stress in shear, 2k is the tensile yield stress, b_N is the total net width, and t is the thickness which may be large or small. It should be kept in mind that in this section of the paper, and in this section alone, an ideally plastic material of unlimited ductility is assumed.

The nominal tensile stress on the net section at the flow limit (general yield) is the tensile yield stress. Appreciable local deformation cannot take place until the yield stress is reached; and conversely, when the yield stress is applied, large local deformations must take place. It should be kept in mind that strain concentration at the edge of the notch is still large despite the evening out of the stress.

Externally deeply notched sheets or rectangular bars behave quite differently. Shearing action in-the-plane (plane strain) requires the much higher average stress of $(2 + \pi)k$, Figure 2(a). Shallower notches cannot take this stress, as shown by the kinematic pattern of Figure 2(b). Thin sheets, the usual test case, will shear out of the plane(27) and an upper bound on the limit load, Figure 2(c), is given by

$$P^u = 2kb_N t\left(1 + \frac{\sqrt{2}t}{4b_N}\right) \qquad [2]$$

An immediate conclusion from the analysis of Figures 1 and 2 is that the external notch in the sheet is more like the crack in a ship than the internal notch which resembles it pictorially. There is no free edge in the ship structure to which the yield zones may extend, so that the yield zone between the 45° lines of Figure 1 would be constrained in a ship by the surrounding material.

Figure 3 shows an externally notched round bar or equivalently a rectangular bar with notches all around. If the notches are very deep, the limit load is approximately

$$P = (2 + \pi)k\, A_N \qquad [3]$$

where A_N is the net area(28). This is the same value as for

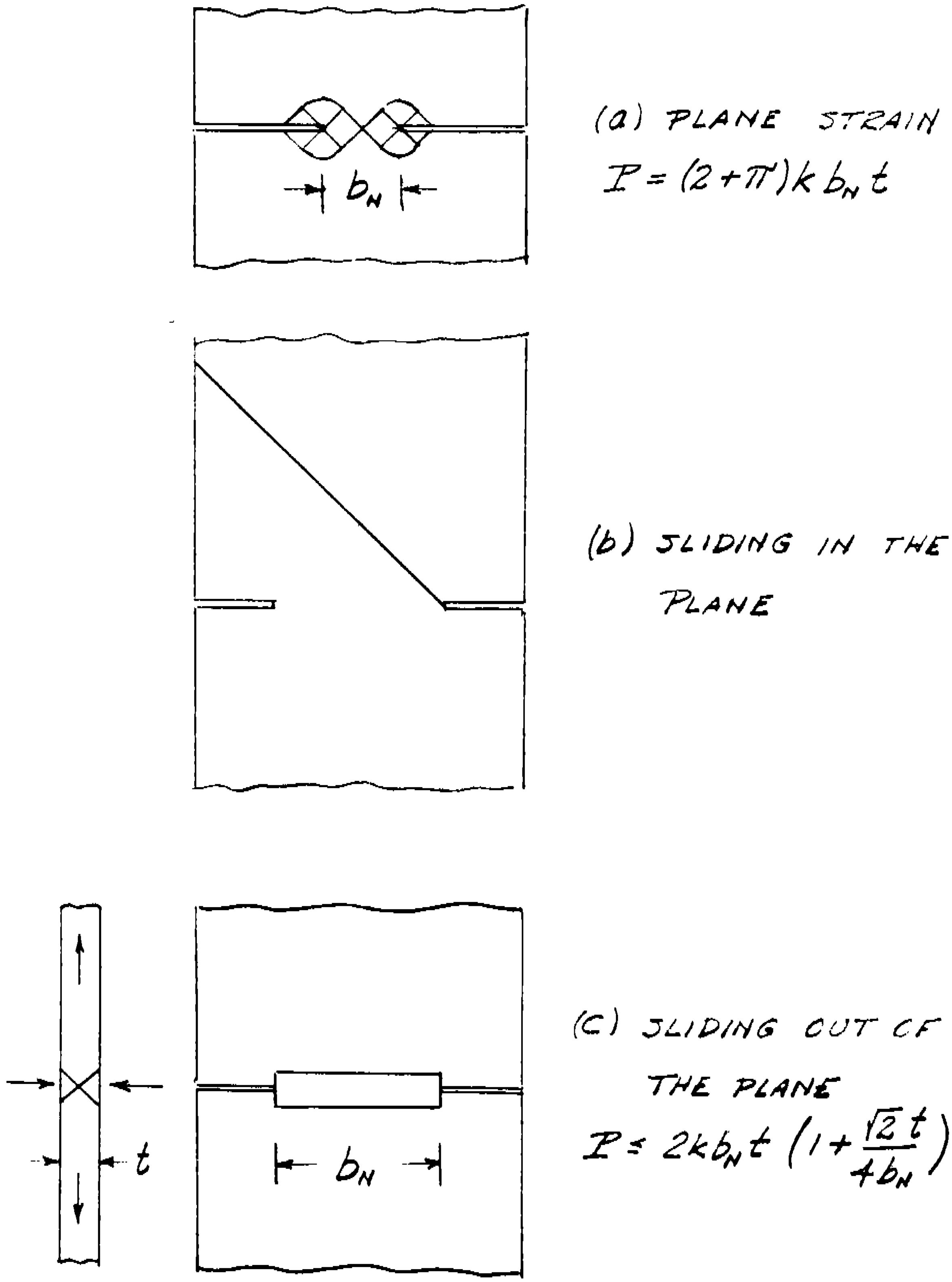


FIG. 2 EXTERNAL NOTCH

a deep notch in plane strain, Figure 2(a). Shallower notches will lead to lower limit loads, but general yielding requires appreciably higher loads than the notched sheet. Strain concentration at the very high stress levels makes the solution less applicable to a real material than the notched sheet solutions are.

Figure 4 shows a simple circular arc kinematic scheme(29) for a notched beam in bending and explains why the restraint factor is considerably lower in bending than tension. The ability to rotate brings the maximum stress for no strain-hardening to about

$$2(1.3)k$$

which is far below the $2(1 + \frac{\pi}{2})k$ for Figure 2(a) or Figure 3.

Replacing the central compression region by a rigid bar or pin cuts down enormously on the amount of plastic rotation possible before large strains occur on the tensile side over the entire middle cross section.

Perusal of the data obtained for notched specimens(15) shows that the strengths found do follow the simple theory when ductile cleavage fractures take place. Marked deviations from limit loads can be observed on the one hand, not for general yield, but for ductile fracture in which work-hardening permits much higher loads to be reached, and on the other in eccentrically loaded specimens so constructed that elastic deformations permit large plastic deformations below the limit load. Hatch corner specimens are apparently of this latter type, although not necessarily to a very marked degree(17).

INITIATION OF CRACKS AND SIZE EFFECT

As is well known, no matter how sharp a crack or notch is introduced in a specimen, static loading ordinarily results in appreciable plastic deformation at the root of the notch. Even at temperatures far below 0°F, when the fracture is definitely of the cleavage type, a small but measurable thumbnail of plastic action must first be overcome. At all temperatures, of engineering interest certainly, ductile fracture precedes cleavage. For machined notches, saw-cuts, or the very sharpest cracks introduced at liquid nitrogen temperatures, the local ductility is large at temperatures encountered in practice, and the considerations of the

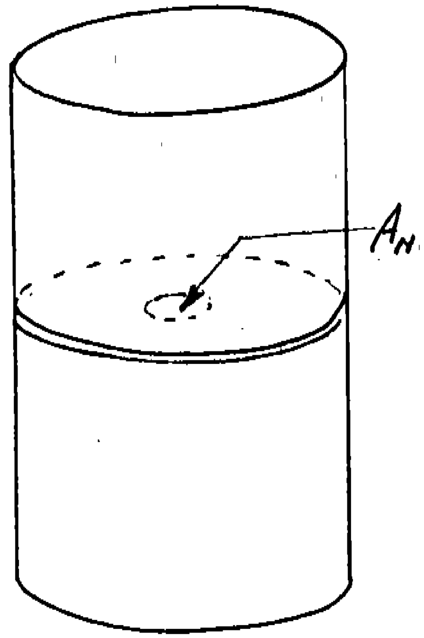


FIG. 3 DEEPLY AND SHARPLY NOTCHED BAR

$$P = (2+\pi)\, k\, A_N$$

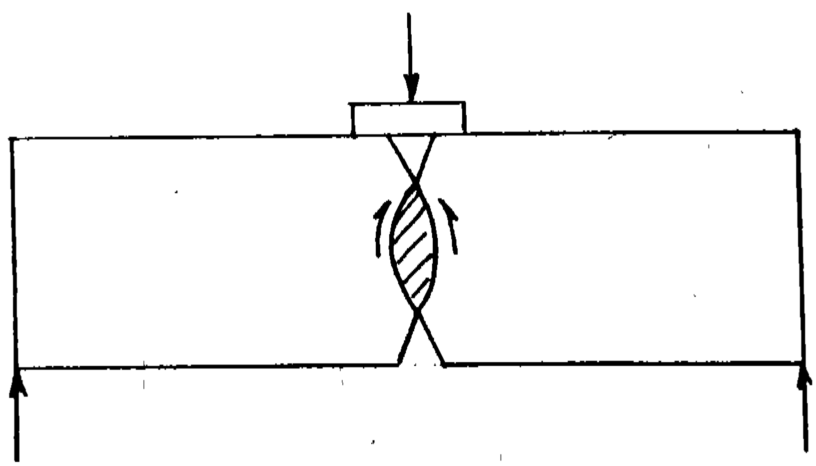


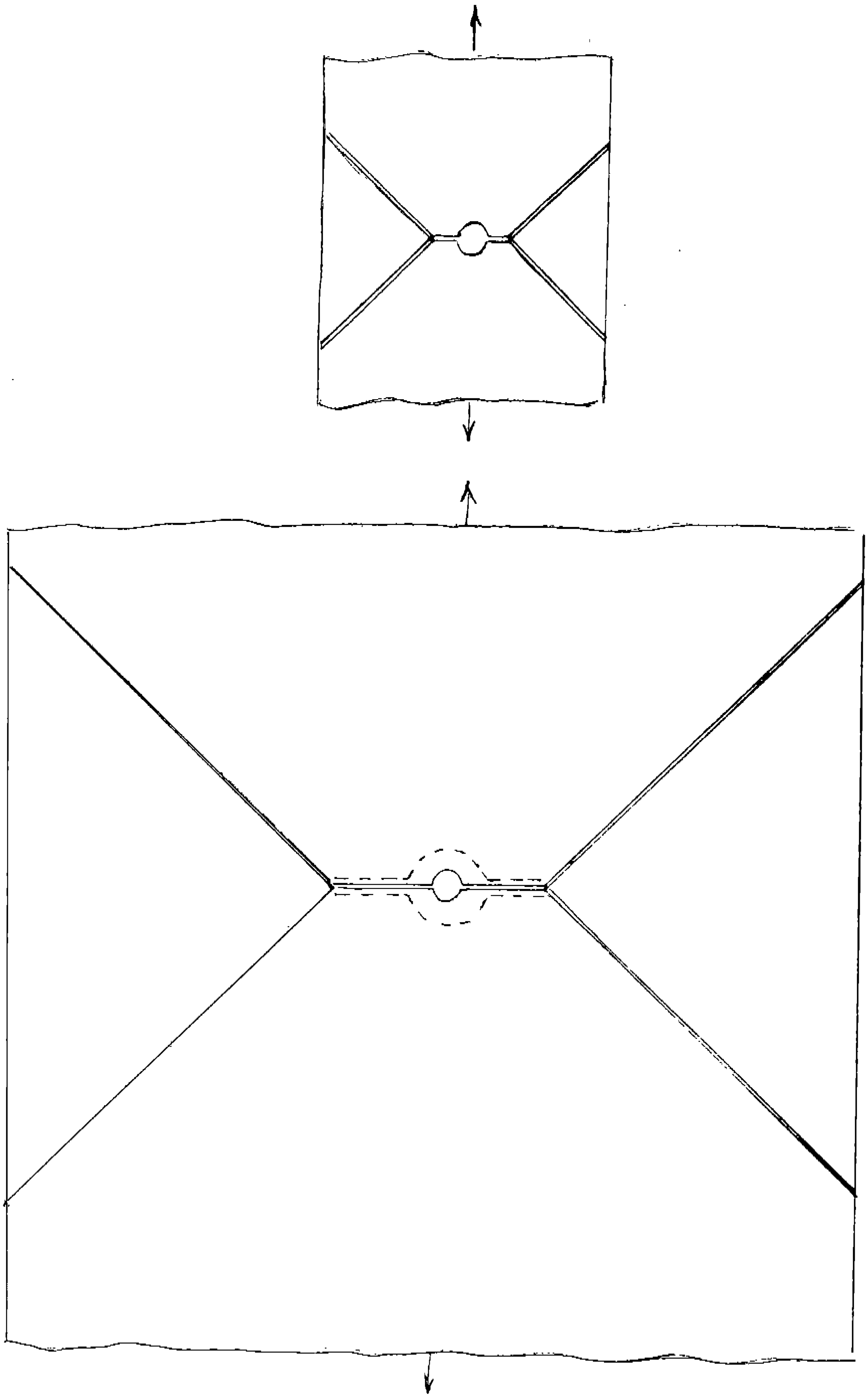
FIG. 4 BEAM IN BENDING

preceding section apply. Except when the constraint factors are extremely high, exhaustion of ductility locally requires general yielding, that is, reaching the flow limit for the geometric configuration present. If by some prior means the local ductility is exhausted or even partially used up, it seems likely that fracture can occur at appreciably lower average stress than the yield stress. Cold pressing of the notch, shot peening, fatigue or reverse loading all help to exhaust the ductility. As cracks do initiate in structures at low nominal stress, it seems probable that some such action does in fact take place.

Locked-in stresses on a large scale due to fabrication, thermal stresses and bending stresses over appreciable regions due to improper proportioning of rigidity of the structure certainly add in a significant way to the level of stress. Welding may often decrease the ability to deform locally, but residual welding stress on a small scale is probably wiped out by plastic action in most cases. Indications from actual ship fractures are, nevertheless, that brittle failure has resulted at total "nominal" stress below yield.

The concept of ductility exhaustion does account for some of the size effects observed. Internally notched specimens, in which the notch terminates in a jeweler's saw cut, show a definite increase in brittle behavior as the width of the sheet increases. Figure 5 provides a qualitative explanation. As shown in Figure 5(a), when the specimen is small, appreciable overall strain can be accommodated before strain hardening sets in at the notch apex because the number of available 45° slip planes is relatively large. When the specimen is large, Figure 5(b), the number of available planes remains the same and so is smaller relative to the dimensions of the triangle formed by the 45° lines. The local ductility is exhausted, and strain hardening sets in at the root of notch at a lower average axial strain in the triangle so that there is less stress equalization across the net section. Another way of looking at the effect is to enlarge the small specimen and superpose it on the large one as indicated by the dotted lines of Figure 5(b). The notch is then clearly much less severe in the small specimen. Perhaps a more important point is that all the little notches in the notch due to surface irregularities are the same for small and large notches.

The same reasoning applies to the welded hatch corner specimens in which the size effect is so spectacular; the nominal breaking stress for the 1/4 scale model was 48500 psi,



(a)

(b)

FIG 5 SIZE EFFECT

for the 1/2 scale 39000, and for the full scale 24200 psi. Welds in general and especially welding of plates at right angles tend to produce an absolute degree of discontinuity. These are relatively much sharper on a large than a small scale test specimen. The effect should, however, be a matter of plate thickness primarily and so can be reproduced on a laboratory scale.

The initiation phase as here defined covers the ductile phase and therefore includes the initial spreading of the crack. If progress is halted, the initiation phase must begin again. A ductile tearing failure in a sense never gets beyond the initiation phase because removal of the load stops the crack from spreading. The lower the temperature and the higher the constraint factor as given by ideal plasticity theory, the easier the transition from the initiation to the propagation of a crack.

CRACK PROPAGATION

Crack propagation, as here defined, refers to spreading of the crack partially or fully, independent of variation of the applied load. The propagation phase is rapid and primarily brittle. Small but measurable plastic deformation accompanies the fracture varying from a detectable shear lip, through an amount which is observable with X ray, to a fracture so brittle that an electron microscope is required to show the plastic action. Some plastic action must always occur because the cleavage planes are randomly oriented, and some planes must shear. Average measured change in plate thickness in ship fractures(14) is about 1.5%, most of which occurs near the surfaces of the plate.

Although a great deal of study has been devoted to the phenomenon of crack spreading, much remains to be learned. Irwin(30,31,18) extended the Griffith theory of fracture(32) by including the work required for plastic action. Orowan(20) likewise has discussed this approach which is essentially one of stability based upon energy considerations. Wells(21) has measured the heat generated and believes the minimum value of energy needed to form the surface is 4.5 ft-lb per sq. in. MacCutcheon(14) states that fast propagation is likely to occur if the figure is below 4000 in-lb per sq. in. MacCutcheon(14) has proposed a dimensionless ratio of energy needed to energy locally available.

The fundamental postulate of the Griffith type theory is that a crack will spread if the energy contained in the initially cracked material and the system of forces is as large or larger than the energy in the entire system with an extended crack plus the energy required to create the additional surface. The criterion is more easily stated for a "fixed grip" condition in which the external forces do not move and so do no work. Including the plastic action, the comparison is made between the strain energy stored in the initial system with a crack and the strain energy stored in the final system with an extended crack plus the energy dissipated and the energy used in the creation of additional surface.

Clearly, unless the energy is available in the above sense, the crack cannot spread. The converse does not necessarily follow. In other words, the Griffith type of theory gives a necessary condition for the extension of a crack, not a sufficient one. The omission of a kinetic energy term is not the primary difficulty. In general there is an auxiliary requirement to be met, perhaps maximum stress, which may be thought of as an energy barrier. Many problems in physics can be treated properly by the simple energy balance concept because thermal or other fluctuations are present in large enough amount to overcome the energy barrier. Most problems in mechanics of continua cannot be analyzed as easily. A reasonably detailed examination of loads and deformations and often of stress and strain is needed for proper understanding and for correct prediction of behavior.

The crack propagation problem in its most simplified form is one of dynamic elastic-plastic wave propagation and is far beyond our present ability to obtain solutions. Adding the unjustified assumption of perfect elasticity provides great simplification, but the problem is still far from trivial. A steady state plane strain elastic solution by Yoffe(33) for a moving system of loads on the surface of a half plane is indicative of what can be expected, but the relevance of quantitative or even qualitative conclusions is by no means obvious. The main point in the spreading of a crack is that the distribution of stress is quite different from the static and that fracture is governed by the time history of stress, by strain history and temperature. Almost nothing is known about the equivalent of the energy barrier either at low speeds or at the enormous rates of strain developed; that is, information is not available on the stresses developed and required. Nevertheless, the evidence does seem conclusive that in structural steel the barrier is enormous and that the energy balance is so highly unstable that a Griffith type theory can be completely misleading if applied to structures.

Instability is always troublesome to analyze and still more difficult to design against. Figure 6(a) is a guess as to the qualitative behavior of steel in a plot of work per unit area needed to create a crack against velocity of crack propagation. The basis for the initial portion of the curve is the known increase of ductility and strength with strain rate(34--36). The falling part of the curve results from the progressive localization of the plastic deformation with increase in velocity of loading. Irwin has suggested that the curve may later rise, and this possibility is indicated by the dotted line of Figure 6(a). As the energy available at higher speeds of propagation is the local strain energy, the ordinate may be replaced by the square of the stress in the region through which the crack travels multiplied by appropriate constants and by the available volume of material which can supply energy. Although the volume is a decreasing function of velocity, the curve for stress level against velocity for a given material at a given temperature should be approximately as shown in Figure 6(b). The faster the crack goes, especially in a material with a time delay for yielding, the closer the approach to elastic stress distribution, the higher the stress concentration. Data of Vreeland, Wood, and Clark(37) on time delay give 100 seconds at 38000 psi, 1 second at 40000, and 0.002 seconds at 50000 psi. With speed of propagation in the neighborhood of 5000 feet per second, very much smaller time delays are significant. Figure 6(b) does not imply that lower local stresses are needed for fracture. In fact it is likely that the true local stresses increase with velocity. Figure 7 is a guess on the shape of the stress distribution as the crack proceeds. If the pictures are qualitatively correct, the futility of applying a static energy criterion is apparent. The higher the speed of propagation, up to speeds appreciable compared with the speed of sound, the less the stress level needed. If the barrier of initiation could be overcome by random fluctuations which are always present, all structures loaded to what are presently considered as reasonable working stresses would fail by cleavage.

Length effects as observed by de Leiris(42), or equivalently the question of the influence of total energy stored in the structure, are important only in the sense that the necessary stress level for slower moving cracks can be maintained by the available reservoir of energy. This problem, although of great physical interest, is not relevant to brittle fracture at average stress levels appreciably below yield.

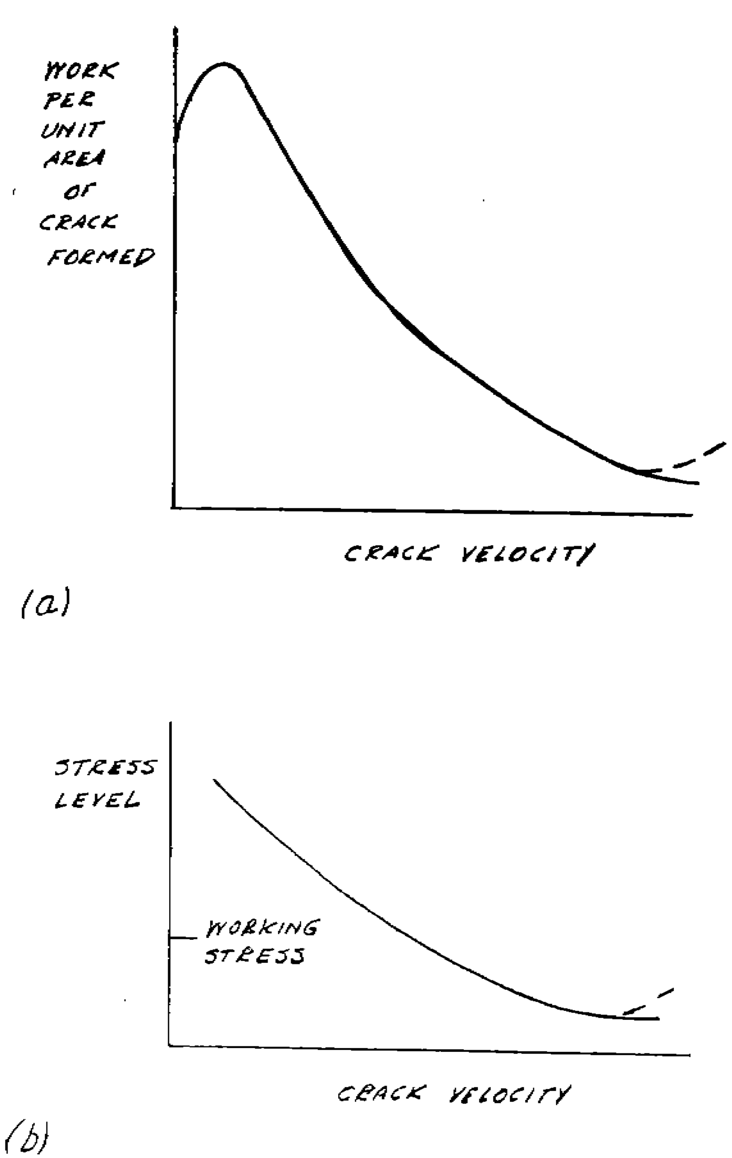


FIG. 6 PROPAGATION AND INSTABILITY

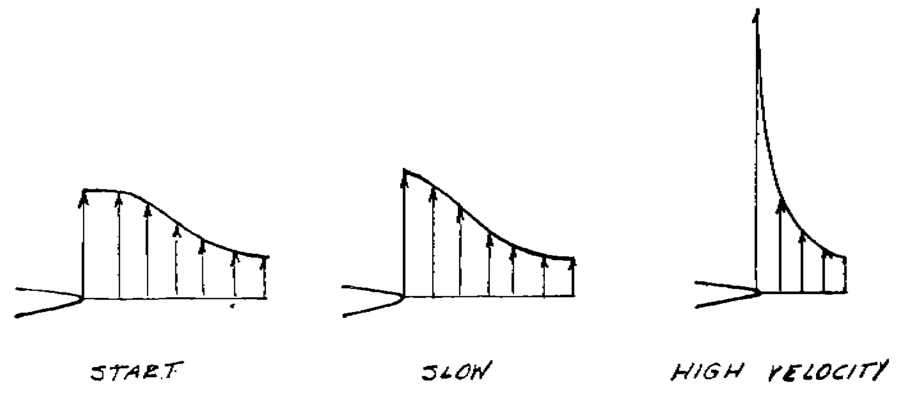


FIG. 7 STRESS DISTRIBUTION - PURE GUESS

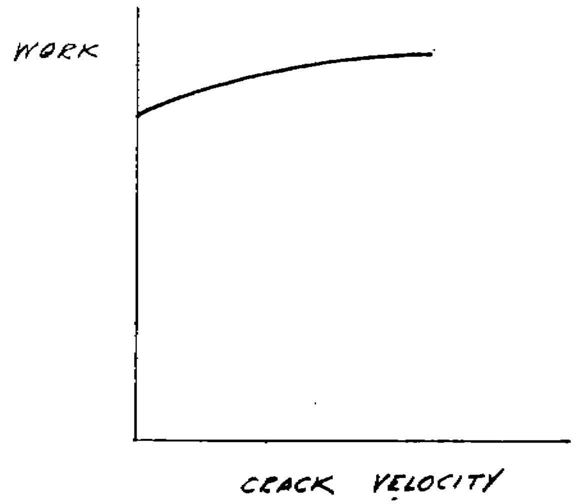


FIG. 8 STABILITY

A rising curve of local energy with velocity, Figure 8, as may apply to aluminum up to a quite high speed range corresponds to a stable instead of an unstable situation. An energy criterion is applicable, and as Orowan(20) has demonstrated, the order of magnitude of stress at the root of the advancing crack is correct. The advance of a crack through material with a rising instead of a falling dissipation curve is stable and not catastrophic as for structural steel.

ACCEPTANCE TESTS AND ANALYSIS OF VARIABLES

Taking all the various tests as a group, including the Kahn tear test, impact tests, Bagsar and Meriam specimens, etc., the functional relation can be written for a given material

$$p = p(\text{stress level, geometry, temperature, rate of loading, previous history})$$

where p is the probability of cleavage fracture. Similarly, the transition temperature T_o observed on an appearance or energy basis or even on ductility at the root of the notch(12) may be written as

$$T_o = T_o(\text{stress level, geometry, rate of loading, previous history})$$

Within wide limits, a given probability of cleavage fracture may be achieved by altering stress, geometry, temperature, or rate of loading. This does not mean that all are equivalent in some manner or that there is some dimensionless combination of all of them which governs fracture. Quite the contrary, as force, length, time, and temperature are independent units; each variable is probably a completely independent quantity, overlapping only in the sense that in many combinations they can be very effective in causing cleavage fracture while varied independently they are not generally able to do so. As is known, the temperature cannot be too high or the geometry too smooth or the stress level too low if the remaining variables are to take on reasonable values. Transition temperature similarly is affected by these variables except, of course, for temperature. Force, length, and time are again independent units; and stress level, geometry, and rate of loading are therefore likely to be independent variables.

An unlimited number of tests may be devised to obtain a given transition temperature. Achieving one that matches the temperature for an observed structural failure is in no way a guarantee that the acceptance test is a suitable one. Except for the extremely important quality control aspect, such a test may in fact serve no useful purpose. Unless all possible structural materials can be examined by a particular acceptance test procedure with reasonable assurance that the rating obtained will be a measure of successful application in the prototype, the test has no advantage over a Kinzel test, a Charpy impact, or a static tear test.

The question to be answered first is what a quantitative basic test must do. As the significant variables are truly independent, each must be controlled separately. Test temperature must be operating temperature; as Captain Roop has said, a steel at a different temperature is simply a different material. Strain rates should duplicate those of the prototype, significant geometric dimensionless ratios must be copied, and the stress level must be approximately the same as for the full scale structure. An immediate objection will certainly be raised. Models do not fail in a cleavage mode when the conditions are slavishly followed. Higher stresses or impact or suddenly applied loadings are employed to initiate cracks. Pellini and co-workers(38) use failure of a buttered over weld. Feely, Hrtko, Kleppe, and Northup(11) have shown that cracks will propagate in models at stress levels below those known to exist in full scale structures, but initiation is quite a different matter.

Two avenues of attack are open. One is to say that it is essential to play completely safe, to assume initiation cannot be prevented and to use design stresses below the level at which propagation is possible. The Standard Oil Development test(11) is based on this philosophy. The ductile phase is overcome by an impact transient which disappears in the propagation phase. There is, however, the unfortunate possibility that the safe stress will be too low to be usable economically. Some of the tests reported by Feely make this a very real possibility for some present day steels at least.

Almost all of the other tests so prominent in the literature do not attempt to duplicate service conditions and are of the initiation type. They rate steels on the basis of transition temperature or energy absorbed. It does seem logical that the lower the transition temperature, the better the steel's low temperature behavior will be, and in some cases this is known to be true. However, all structural steels in use operate below the transition temperature as determined by these tests. If the basic problem is one of initiation at working temperature and working stress, it is by no means obvious that such tests do provide the fundamental information which is needed so badly.

Until a test is devised which produces fractures equivalent in all respects to that of the prototype but on a convenient laboratory scale, a scientific study and evaluation of the brittle fracture problem cannot be achieved. It seems probable that the variable which must be investigated systematically is the history of the strain at the root of the notch. Again temperature will enter prominently, and a strain history at a higher temperature will not necessarily be the equivalent of one at a lower. Perhaps also welded specimens alone can give pertinent data.

ANALOGOUS DESIGN PROBLEMS

The brittle fracture of structural steel as it has been described here involves initiation at many different nominal stress levels with propagation possible at appreciable lower stress. A large number of structural engineering design problems are in essentially the same category. They are often called buckling problems, although in practice they actually involve combined normal force, bending and twisting moments acting upon bars, plates, or shells. The strongly unstable effect shows up in many perfectly elastic structures but is most pronounced and almost invariably appears when the elastic limit is exceeded(39).

A column under direct load is a well known example. The load deflection curve is as shown in Figure 9. The larger the initial eccentricity of loading, the smaller the maximum load the column can take. "Playing safe" in this case cannot mean designing for zero load but rather for a reasonable value of eccentricity. This same type of decision on safety must be made explicitly or implicitly in all column design, as well as in the design of beams able to buckle laterally or to twist(40, 41) and in plates and shells under compressive stress in the surface. In some cases as the deflection grows, the altered geometry finally strengthens the structure, and a minimum load will be reached. Staying below this load gives a safe but often uneconomical design.

Perhaps the eccentricity of load on a column corresponds in effect to the strain history at a notch in a ship structure. If so, the very difficult problem of a measure of the effect must be met and evaluated in terms of allowable stress. Field

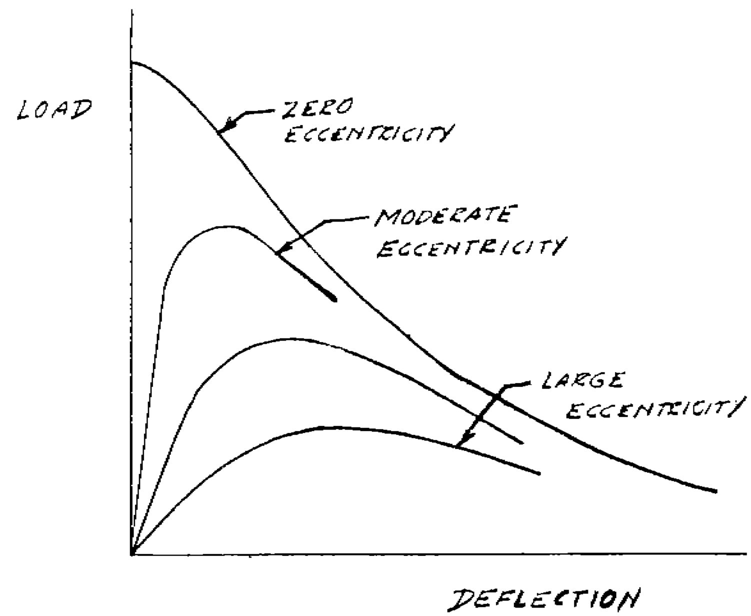


FIG. 9 LOAD-DEFLECTION CURVES FOR ECCENTRICALLY LOADED COLUMNS IN THE PLASTIC RANGE

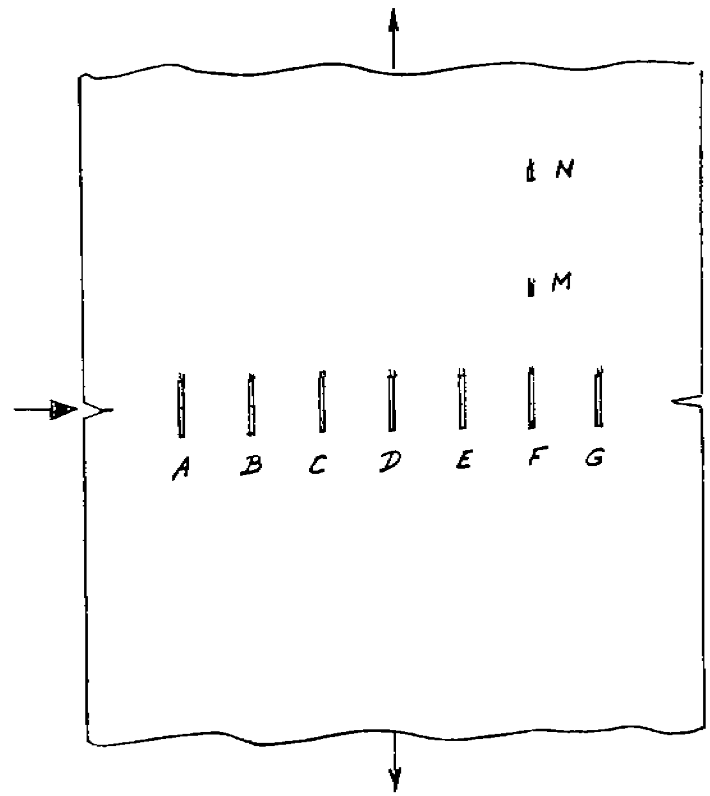


FIG. 10 INSTRUMENTED FOR INSTANTANEOUS STRAIN RECORDING

experience is the best means of arriving accurately at such a measure, but the process seems a little too costly. It hardly pays to develop a better steel and to try it out by making a ship or a welded Vierendeel truss from it.

FUTURE TEST PROGRAMS

In the interest of economy it appears desirable to perform experiments to determine whether or not the initiation barrier to crack propagation can be depended upon in design. At the same time in the interests of safety and basic understanding, every encouragement should be given to tests which evaluate crack propagation stress levels. The Standard Oil Development type of test, which evolved from the Robertson specimen(43), appears ideally suited for the propagation study, but additional information would be most helpful. In particular it would be of extreme interest to find out whether the stress field is altered appreciably by the advancing cleavage crack, whether, in fact, the change is sufficient to modify the concept of brittle failure proceeding through the known original stress field. Strain gauges feeding into instrumentation having very high frequency response as well as static stability might be placed as shown in Figure 10. The longer gauges A-G along the crack path would serve a dual purpose, indicating both stress field in the conventional way and velocity of propagation by the time of break. Short gauges MN further away would indicate transients if any are set up by the advancing crack. A nucleation process should produce appreciable transients, while a steady advance should not.

None of the commonly used tests will determine the initiation condition as it is encountered in actual structures insofar as stress level is concerned. Every effort must be made to duplicate cleavage fractures on concentrically loaded, edge notched, symmetric models of reasonable size at working temperatures and working stress levels. As Parker points out(12), the so-called brittle failures observed in the laboratory show 10% ductility at base of notch while service fractures show 1 or 2%. Eventually failures should be producible at approximately 35°F and 18000 psi with present ship steels of 3/4-inch thickness. Prior strain history must provide the key perhaps aided by low level impacts comparable to dropping loads on ship deck. Welding in itself is very likely to be an important part of the history. Many carefully controlled experiments on welded and unwelded notched specimens will be required to establish a useful test procedure.

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REPORT ON BRITTLE FRACTURE STUDIES

by

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I. INTRODUCTION

This paper covers progress to date on brittle fracture studies being conducted in the Standard Oil Development Company, Esso Engineering Department Laboratories. These studies were undertaken to investigate the cause of failure of two large oil storage tanks in England in the spring of 1952. The object was to determine what can be done to insure against similar tank failures in the future.

The investigation has covered the broad field of possible causes of tank failures and definite steps have been taken to improve the quality of welding through the use of radiographic examination. At the same time, a laboratory program was begun to determine whether the use of better steel was the ultimate solution to the brittle fracture problem.

From the beginning of the laboratory program, emphasis has been placed on developing a test to simulate conditions at the time of the tank failure. This paper outlines the steps taken in developing this test, and, in particular, studies of the effect of changing specimen size, geometry, notch acuity, impact, and other test variables. A comparison is made showing the close agreement between test results and actual conditions at the time of failure, using steel from the two tanks which failed in England. Finally, a correlation is made between the test results and fundamental physical properties of the materials tested.

II. SUMMARY

The Standard Oil Development Company investigation of the brittle fracture in tank steel makes use of a test procedure which was originated in the company laboratory. This procedure was developed with a view to duplicating the conditions in a tank which fails in a brittle fashion. Certain features of other studies have helped in developing the test, but it is believed to be unique in most respects. The object of the test is to relate both stress and temperature to the propagation of brittle fractures.

The test developed utilizes notched tensile specimens varying from 6 inches to 40 inches wide and 3 feet to 16 feet long. The notch is an extremely fine brittle crack which is introduced at liquid nitrogen temperature prior to the test. Under constant stress and temperature conditions, a hardened steel wedge is driven into the notch using an impact gun. If the stress is below a certain critical value, the notch is only slightly extended and no failure occurs. If the stress is above this critical value, however, the specimen fails completely with a brittle fracture.

Using this type of test, approximately 120 specimens of steel plate have been tested thus far. Several different types of steel have been tested, with the majority of work being concentrated on a single large ASTM A-285 plate in order to investigate all test variables. The data obtained indicates that below a certain temperature called the SOD transition temperature, the brittle breaking stress is constant while above this temperature the brittle breaking stress increases markedly.

Considering the usual variation in physical properties found in testing steel, remarkable agreement exists between the brittle breaking stress of the Fawley steel as determined by the Esso Engineering test and the calculated stress in the tanks at the time of failure. The stress required for brittle failure to occur in the laboratory for steel from one tank was found to be between 12,000 and 14,000 psi at 40°F. Failure of the tank occurred at a calculated stress of 15,800 psi at 40°F. This is felt to be a good indication that the test is truly indicative of conditions in a tank at the time of failure.

One of the objectives of the laboratory program is to determine whether or not conventional ASTM A-7 tank steel should be replaced with a higher quality steel which is less subject to brittle fracture. Tests on American A-7 tank steels show a wide range of notch-resistant properties for this material. From a preliminary standpoint, results indicate that American Bureau of Shipping Class C steel corresponds to an A-7 plate which has high notch-resistant properties. English steel from the broken tanks at Fawley, on the other hand, correspond to an A-7 plate which has low notch-resistant properties. Considerably more work is necessary, however, before relative quality can be established on a satisfactory statistical basis.

There appears to be a fairly good correlation between the transition temperature determined by the SOD test and that found with the conventional Charpy V-notch impact test. Preliminary data also indicate that a correlation may also exist between the brittle breaking stress and the modulus of rupture at liquid nitrogen temperature.

III. DEVELOPMENT OF A TEST

Background

Early in the laboratory program an effort was made to determine what causes a brittle fracture to propagate in some instances and stop in others. It is assumed that conditions which cause a brittle fracture to start initially are localized and are probably more severe than the average conditions. In the case of tank failures, there is evidence to indicate this is true. For example, on two occasions, short brittle cracks have occurred in the same tanks which later failed completely under more severe stress conditions.

A review of the literature on brittle fracture studies prior to 1952 indicates that most of the emphasis was placed on determining the temperature where a given steel is subject to brittle fracture. These investigations used a number of different tests to indicate transition temperature range and correlated the results with actual failure experience. Notched tensile tests indicate brittle failures occur when the stress is in the range of the yield stress or higher. It seems, therefore, that such tests are indicative only of conditions that lead to the start of a brittle fracture and do not shed light on propagation.

It was interesting to note, therefore, the work of Robertson in England who studied the effect of stress on the propagation of a brittle crack in specimens with a superimposed temperature gradient. Particularly interesting was the fact that he was unable to get a brittle crack to propagate when the stress was below 10,000 psi. This raised the question of whether there is a minimum stress below which brittle cracks will not propagate. It was decided to investigate this question further and an effort was made to duplicate Robertson's work with certain modifications as described hereafter.

Robertson Test

At this point, it is perhaps desirable to review briefly the Robertson test to provide background for the deviations employed. As reported in the British publication "Engineering," dated October 5, 1951, Robertson utilized a specimen of the type shown in Figure 1. This specimen contained a stress concentrator composed of a nub on the side of the plate with a

FIGURE I

ROBERTSON TYPE SPECIMEN

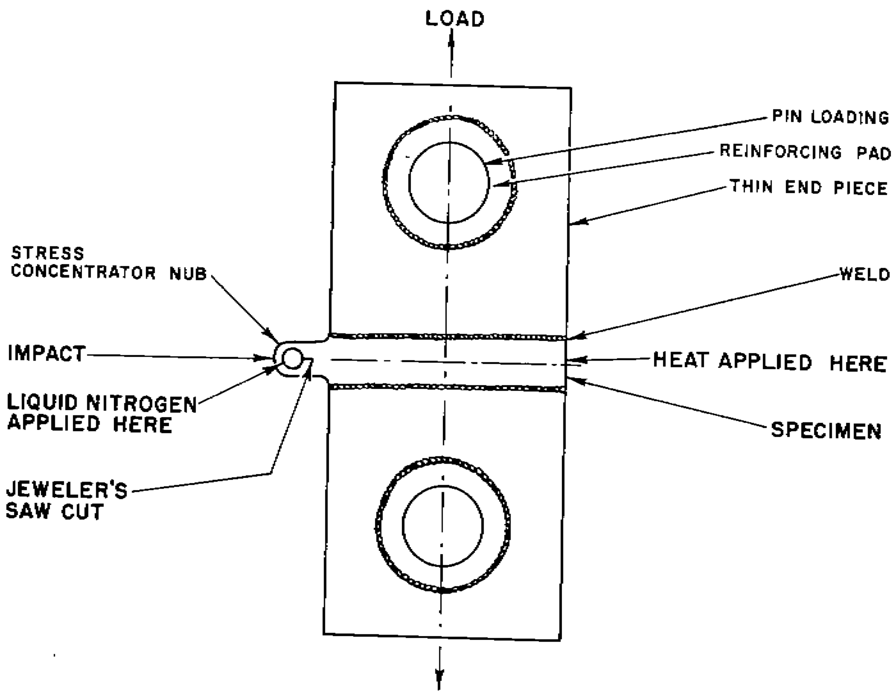
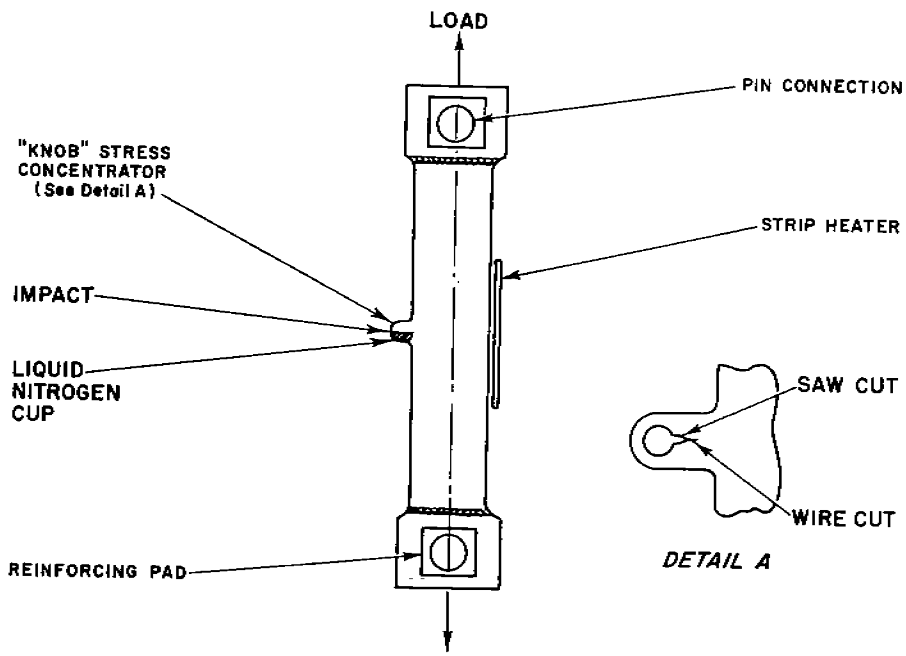


FIGURE 2

TEMPERATURE GRADIENT SPECIMEN



1-inch diameter hole where liquid nitrogen was used as a coolant. Heat was applied on the opposite side of the plate to establish the desired gradient across the specimen. On the inside of the hole, Robertson made a jeweler's saw cut. The plate was loaded through end connections which were of thinner plate so that they were stretched beyond the yield point during the test. This helped to correct any misalignment or bending conditions in the specimen. When the desired temperature gradient and stress were applied to the plate, an impacting device was directed at the outer surface of the stress concentrator nub. This resulted in a brittle crack starting at the jeweler's saw cut and propagating across the plate. The length of the crack thus produced depended upon the temperature gradient and the average stress level. Robertson determined the determined the temperature at the end of the brittle crack by measuring the position of the tip of the thumbnail at the end of the brittle crack with respect to the temperature gradient on the specimen. He related this temperature to the stress in the specimen. Using this test, he found that different steels exhibited wide variations in properties. It was felt that perhaps the most significant finding was that below a stress of 10,000 psi Robertson could not get the brittle crack to propagate.

Several questions come to mind in considering this test, and its interpretation. Briefly, these are:

1. Was it possible to get uniform stress distribution with the specimen employed, particularly with a temperature gradient present?
2. Was the stress measured prior to impact significant and how was it related to the stress at the time the crack was arrested?
3. Were the results affected by the use of impact to start the crack?
4. Were the results affected by the size or geometry of the specimens employed?
5. Was the tip of the thumbnail the significant point from which to measure temperature?

Modified Robertson Test

The modified Robertson test which was run in the S.O.D. laboratory employed a specimen as shown in Figure 2. A stress concentrator of the same design as Robertson's was used. In place of the plastic end connections used by Robertson, a longer specimen was substituted since it was felt that the transient stress from the impact would not be reflected back to disturb the stress distribution until the crack was well under way. In view of the temperature gradient, it was necessary to do considerable experimental work on temperature compensation on strain gauge readings. The stress gradient across the plate is shown on Figure 3. It can be seen that the tensile stress in the direction of loading dropped off considerably near the notch due to the reinforcing effect of the stress concentrator nub. Also, it can be seen that a compressive stress in the transverse direction existed in the vicinity of the notch.

Five specimens were tested using this type test. All were made from a single ASTM A-285 rimmed steel plate from Lukens. It was found that the brittle crack started by impact would not propagate farther at an average stress of 15,000 psi than it would at no stress. At 20,000 psi average stress, the brittle crack traveled completely across most of the specimens due to the fact that the need for as high a temperature as was required to arrest the crack was not anticipated. The crack was finally stopped by increasing the temperature on the far side of the plate to approximately 140°F. With an average stress of 20,000 psi, the crack propagated across the plate and stopped at a point where the temperature was at 120°F. At this stage of the program, an evaluation of the modified Robertson tests was made and it was concluded that:

1. Poor stress distribution existed, with the axial tension low near the notch. An appreciable transverse compressive stress was also present. Both of these appeared to retard the start of a brittle crack and hinder propagation in the early stages.

2. There was an indication of a shifting stress pattern, and it was thought that the difficulty encountered in stopping the crack might be related to this stress realignment.

3. The combined effects of items 1 and 2 raised a serious question as to the significance of any relationship between average stress prior to impact and the temperature where the crack stopped.

4. There was, however, a definite indication of a minimum stress below which brittle fractures would not propagate.

FIGURE 3

STRESS AND TEMPERATURE GRADIENT

MODIFIED ROBERTSON SPECIMEN

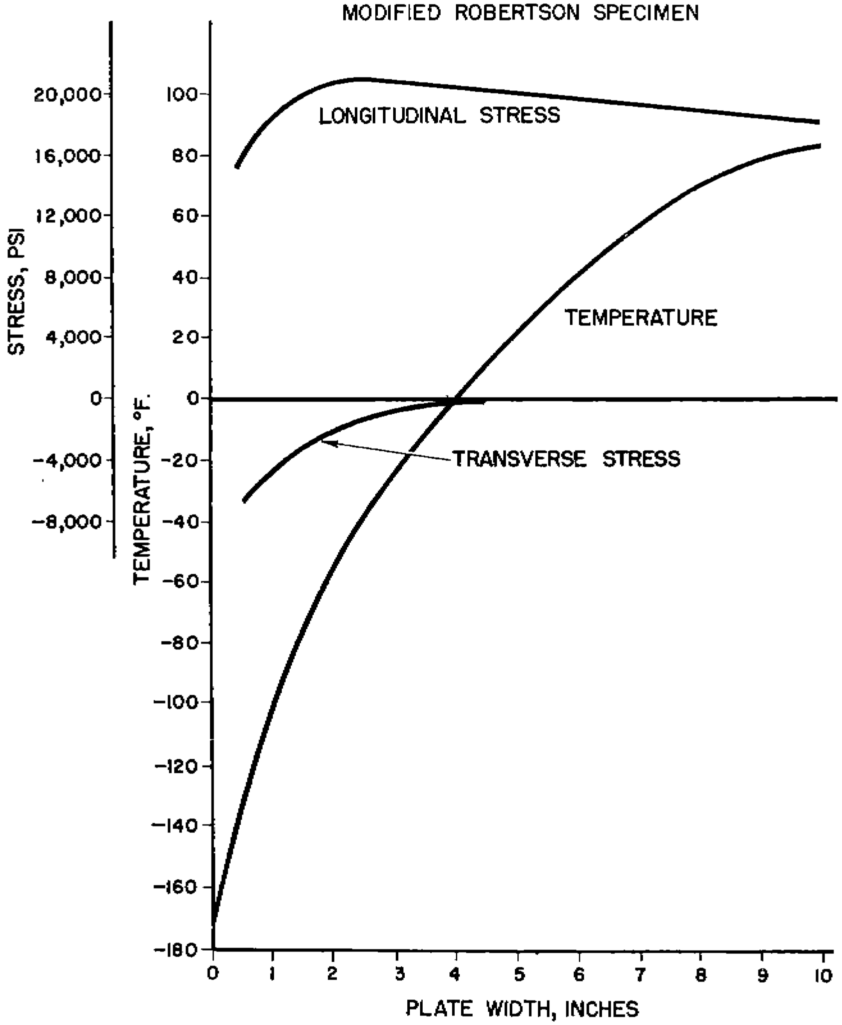
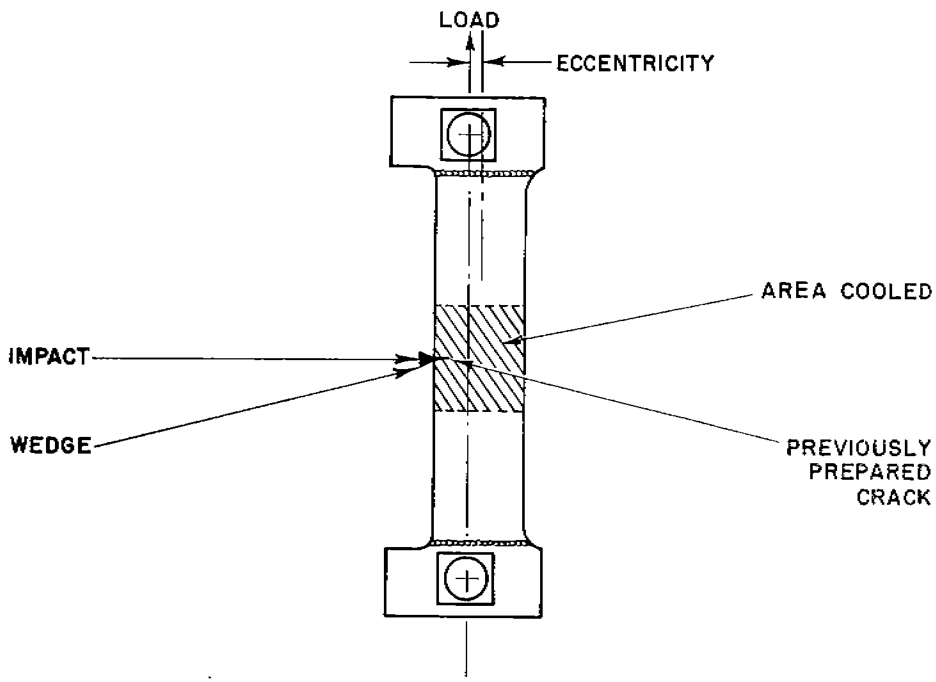


FIGURE 4

STRESS GRADIENT SPECIMEN



Stress Gradient Test

In view of the difficulties encountered in the modified Robertson work, it was decided to see if a crack could be stopped by having it run from a zone of higher stress into one of lower stress at constant temperature. It was reasoned that such a test might indicate the extent to which stresses shift as a crack propagates. For this test, an entirely new type of specimen was employed as shown in Figure 4. This specimen was loaded eccentrically and contained a notch on one side. The notch was a brittle crack previously introduced into the plate at dry ice temperatures. The proportions of the specimen were such that the stress gradient existed across the plate. When this specimen was loaded in the machine to the point where the nominal tensile stress at the notch was approximately 20,000 psi, a compressive stress of 2,000 psi existed on the opposite side. The temperature of the specimen was reduced to 0°F by circulating a cool nitrogen stream inside an insulating box. With the specimen at the desired stress and temperature, a hardened steel wedge was driven into the notch using an impact device. With a tensile stress on the leading edge of 15,000 psi, no break occurred; however, with the stress increased to 20,000 psi, complete failure occurred. The fact that complete failure occurred means that the crack either progressed through a compression zone, which seems highly unlikely, or that the stress pattern shifted so that tension was always present. With crack velocity being in the order of one-third the velocity of elastic strain waves in steel, such a shift in stress pattern might have been expected. This experiment, however, indicated that a shift definitely did occur and that it would be extremely difficult to predict the stress existing at the time a crack was arrested with the Robertson type test or any other test involving a running crack.

With this background, it was decided that a test of the "go, no-go" type at uniform stress and temperature was the most desirable from the standpoint of interpreting the results, and steps were taken to develop such a test.

S.O.D. Test

The test which was finally developed employs a specimen of the type shown in Figure 5. This specimen contains a notch composed of a brittle crack on one edge and a saw cut of equal depth directly opposite. The brittle crack is introduced at liquid nitrogen temperatures. The exact technique used in preparing the specimen is as follows.

FIGURE 5
TYPICAL TEST SET UP

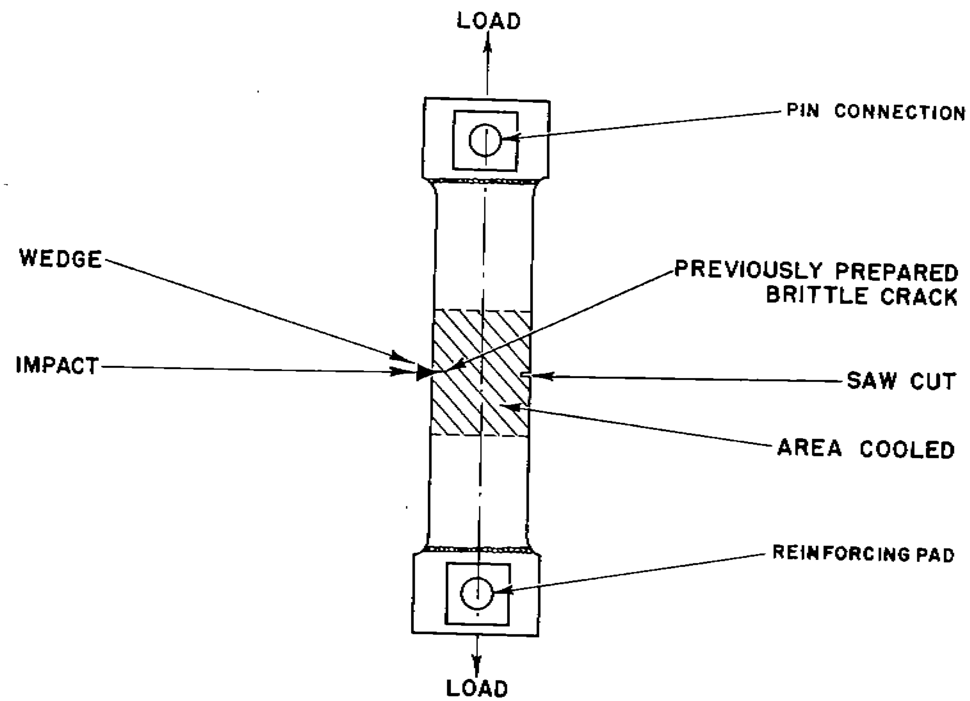
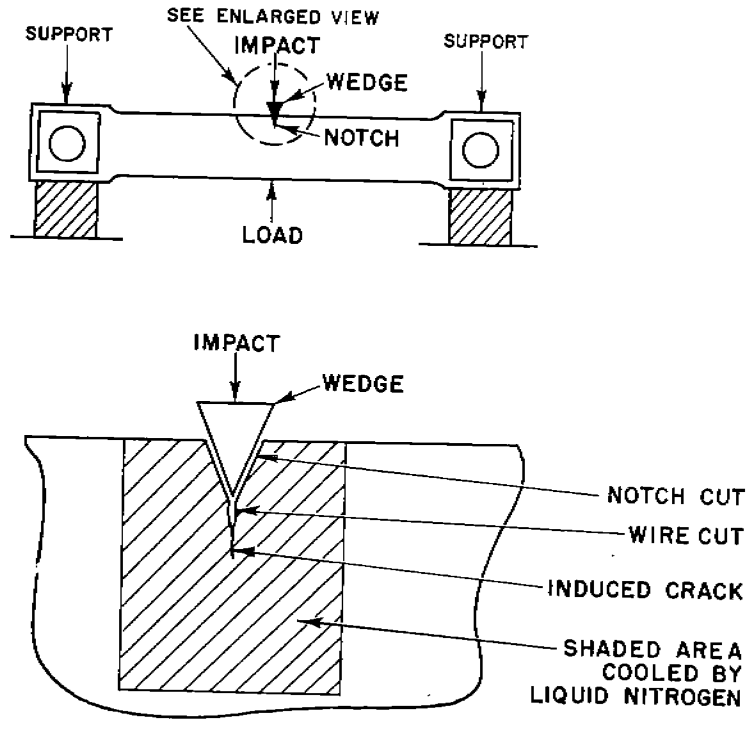


FIGURE 6
PREPARATION OF SPECIMENS



ENLARGED VIEW

Steel which is to be tested is fabricated into a specimen which consists of a slab of the material with the necessary end connections for use in an ordinary tensile testing machine. A typical test piece is shown on Figure 6, together with the method employed in preparing the specimen. A saw cut followed by a subsequent wire cut using a fine wire and grinding compound is made at the middle of the plate on one side. The specimen is then loaded as a simple beam to create a tensile stress in the outer fiber at the cut. The area in the vicinity of the cut is then cooled by means of liquid nitrogen. When the area is at about the temperature of liquid nitrogen, a wedge is driven into the saw cut by impact from a small slug shot at high velocity from a tool commercially available. At the base of the cut a fine crack is induced which is believed to be as severe a stress concentrator as any encountered from welding defects or other causes. A saw cut of the same depth as the crack is made on the opposite side of the specimen to improve stress distribution.

The test specimen is placed in a tensile testing machine and loaded axially through pin connections, as shown in Figure 5. An insulated box is placed around the central portion of the plate and a cooling system placed in operation. Low temperatures are obtained by the use of dry nitrogen which is cooled in coils immersed in liquid nitrogen and is injected into the box in controlled quantities on both sides of the plate. The arrangement of equipment used in the laboratory is shown in Figure 7.

Thermocouple readings taken during the test period permit accurate control of plate temperature. When the desired temperature is reached, a predetermined tensile load is applied to the specimen. At this point, an impact device is used to drive a hardened steel wedge into the previously prepared notch. The wedge is driven by the impact from a small slug shot at high velocity from the tool.

If complete brittle failure occurs, the material is considered to be subject to brittle fracture at the prevailing conditions. If failure does not occur, the tensile load is increased and the plate is subjected to impact at succeedingly higher stress levels until brittle fracture occurs. If the temperature is below a certain critical value, identical results are obtained regardless of whether the specimen is subjected to a number of impacts at progressively higher stresses until the brittle strength is reached or whether this final stress is applied at the first impact. At higher temperatures, this situation does not exist and it is necessary to determine

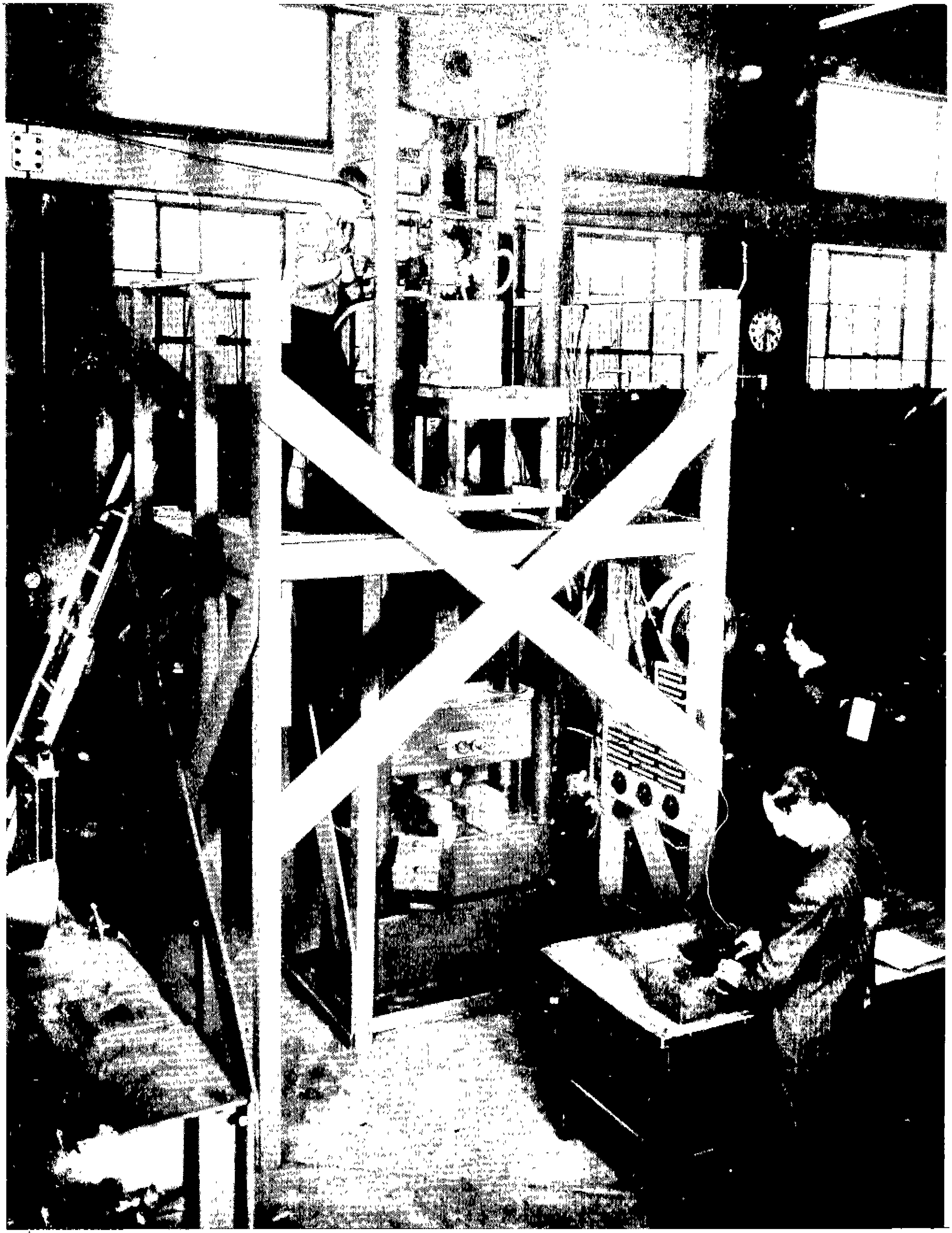


Figure 7

through the use of additional specimens from the same plate the lowest stress at which the material will fail on first application of impact.

IV. STUDY OF TEST VARIABLES

Using the S.O.D. test, approximately 120 specimens have been tested to date. The major portion of these have been from a single Lukens rimmed steel plate of ASTM A-285 specification. The properties of this plate, henceforth called Plate "A", are shown in Table I.

The principal purpose of this phase of the program was to determine what test conditions should be employed to simulate conditions existing at the time of tank failure. Accordingly, the effect of each test variable on the brittle breaking stress was studied individually by holding all other variables constant. The following variables were studied in this manner:

1. Temperature
2. Notch sharpness and length
3. Impact
4. Geometry and size
5. Material

It will be noted that plate thickness is not mentioned. Thus far, no work has been done on this variable. This work is scheduled, however, in the near future.

In general, it can be said that the results with the S.O.D. test are quite reproducible. Deviations are usually less than 1,000 psi where the brittle breaking stress is 10,000 to 15,000 psi. Using the "go, no-go" testing technique, it is almost always possible to narrow the range between failure and no failure to 1,000 psi by testing two or possibly three specimens at a given condition.

Temperature Vs. Stress

When all other test conditions are maintained constant, the relation between brittle breaking stress and temperature is shown in Figure 8. The standardized conditions for all these tests are noted on the curve. Below 0°F, the brittle breaking stress is shown to be essentially constant. Above this temperature, the breaking stress increases markedly. This temperature is called the S.O.D. transition temperature, and it has been found to exist for all steels tested. It

TABLE I
PROPERTIES OF TEST PLATE "A"

SPECIFICATION: ASTM A-285 GRADE C (RIMMED STEEL).

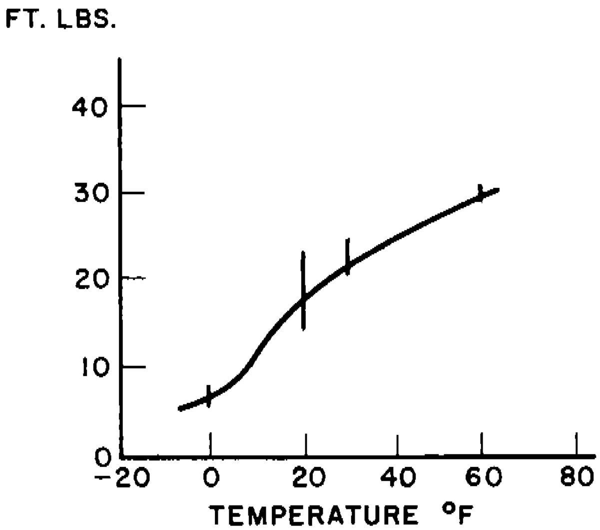
SOURCE : LUKENS

CHEMICAL ANALYSIS	
CARBON	.25
MANGANESE	.49
SILICON	–
PHOSPHOROUS	.016
SULFUR	.040

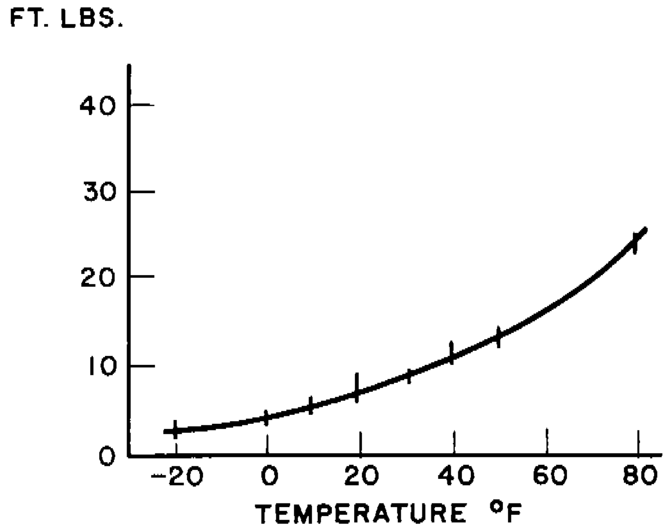
PHYSICAL PROPERTIES	
YIELD STRENGTH	38,300 PSI
ULTIMATE STRENGTH	59,500 PSI

IMPACT DATA

CHARPY U NOTCH



CHARPY V NOTCH



varies for different steels as will be discussed later. The exact shape of the curve between 0°F and +40°F, while not definitely established by the data shown on Figure 8, is known from the mass of additional data where some non-critical variables have been changed.

Notch Sharpness and Length

The effect of notch sharpness was studied over a considerable range and it was found that the brittle breaking stress is reduced as the notch is made sharper and sharper. The following tabulation presents the results of this work:

Type Notch(1)	Radius at Root	Brittle Breaking Stress(2)
Saw Cut	0.020 inch	26,000--30,000 psi
Wire Cut	0.003 inch	21,000--26,000 psi
Dry Ice Crack	Unknown	16,000--17,000 psi
Liquid Nitrogen Crack	Unknown	14,000--15,000 psi

Notes:

1. All notches were essentially the same length, although this has been shown to have no effect.

2. Tests were conducted on steel from Plate "A" at 0°F using specimens 6 feet long by 10 inches wide by 1 inch thick, and medium impact from the small gun tool.

The effect of notch length was also investigated. Liquid nitrogen cracks varying from 3/4 inch to 2 inches in length were tested in pieces varying in width from 6 inches to 16 inches, and no effect due to crack length was observed at any temperature. A few longer cracks were also tested and with these the breaking stress was found to be higher. This is believed to be due to the fact that the wedge was not as effective with the longer crack. It would appear, therefore, that the crack length effect indicated by the Griffith theory does not apply to this test.

Impact

Considerable work was carried out to determine the effect of varying the impact energy used to start the brittle fracture. Figure 9 shows the brittle breaking stress as a function of impact energy with all other variables held constant.

FIGURE 8
STRESS VS TEMPERATURE
ASTM A-285 GR.-C
SPECIMENS 6'x10"x1"
LARGE GUN, HEAVY CHARGE

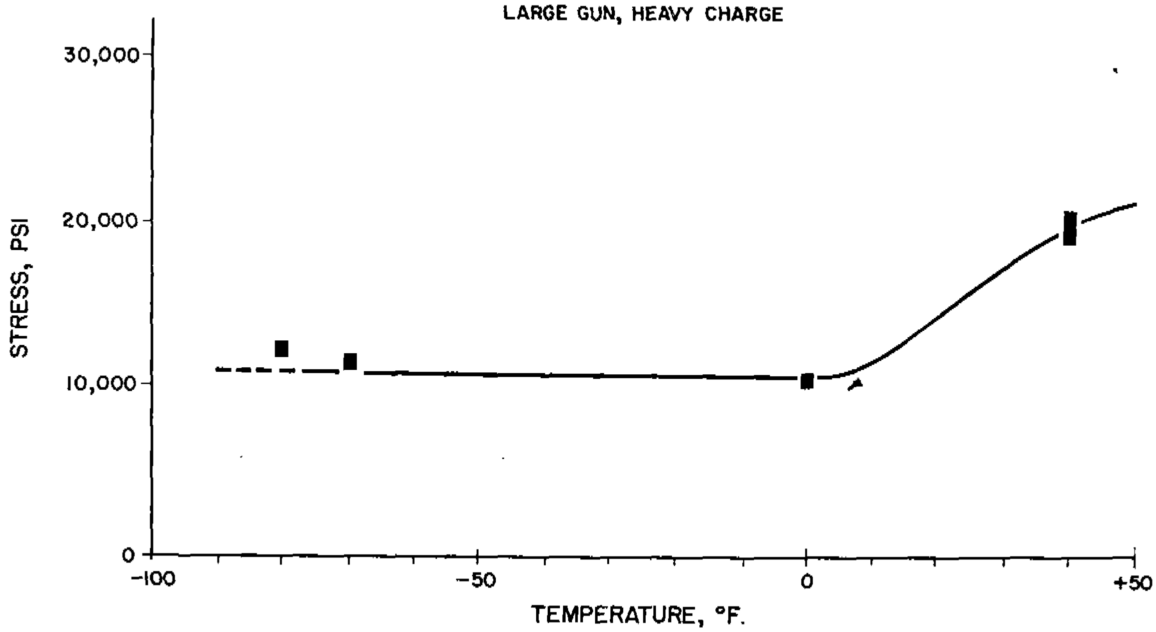
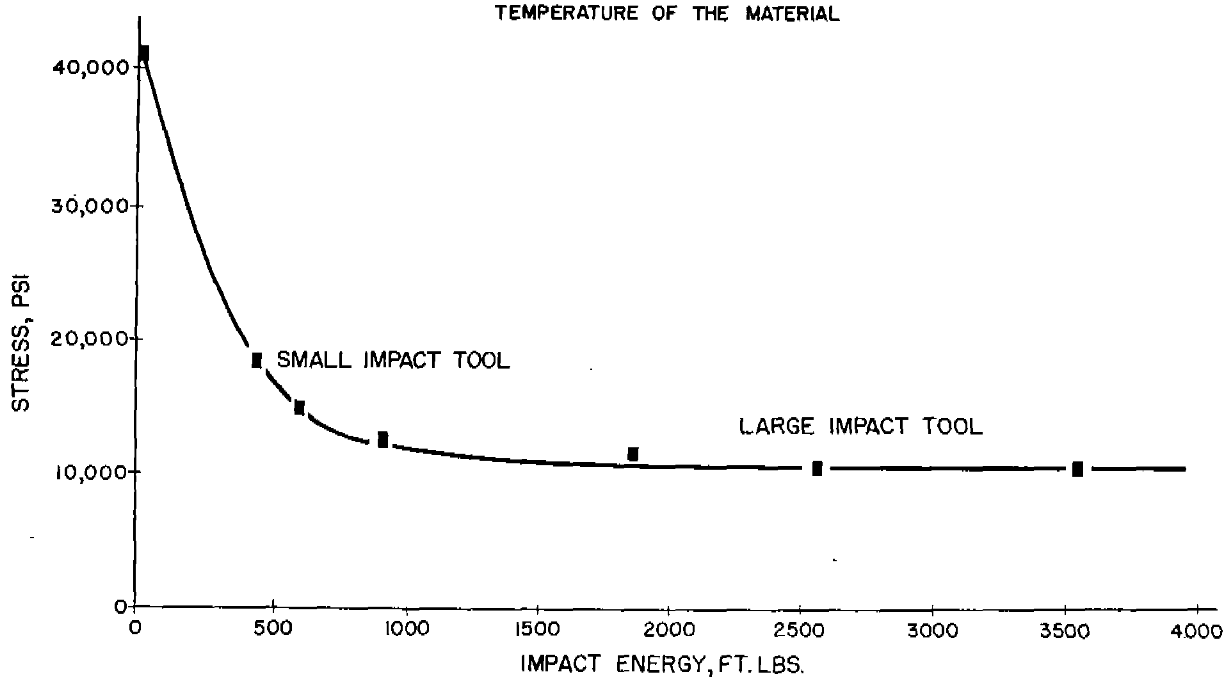


FIGURE 9
IMPACT EFFECT
ASTM A-285-C
60" LONG, 10" WIDE, 1" THICK SPECIMENS
FOR TEMPERATURES BELOW TRANSITION
TEMPERATURE OF THE MATERIAL



In obtaining data for this curve, two types of impact guns were used, each having a range of muzzle velocities. The slug shot from the small gun weighed 8.75 grams and the slug from the large gun weighed 49.75 grams. The velocities were determined through tests conducted at the Remington-Arms Laboratory. The energy in each case was calculated from the above data.

As the energy of the slug striking the wedge increases, the breaking stress decreases. This stress reaches a minimum of 10,000 to 11,000 psi for Plate "A", at approximately 2,500 ft-lb energy. Further increases in energy do not lower this stress.

The following is offered as one possible explanation for this behavior. The assumption is made that the running brittle crack is started by the combined effect of tensile stress due to load on the specimen and transient stress due to impact. At the lower impact levels, the transient stress available to initiate a running crack is insufficient and therefore additional stress in the form of stress due to load on the specimen is needed. Once the crack is started, there is enough energy in the stress field to sustain its continuation. As more and more impact is provided, however, less and less additional stress is needed from the specimen stress field to start the crack. Eventually a situation is reached in which no stress is required from the specimen stress field for crack initiation. In this case, the only stress needed in the test specimen is that required to propagate the crack. If the stress level in the specimen is lower than this minimum value, then the crack will not propagate.

Perhaps the feature of the S.O.D. test which is most difficult for the average person to accept is the use of impact to start a brittle fracture. The question "Where does impact come from in a tank failure?" has been raised by most of those to whom the test has been shown. It is believed that impact causes the wedge to spread the original brittle crack and this results in a very slight extension of the crack which usually is not measurable. At higher temperatures, when the stress is just below the failure stress, fairly large drops in load of the tensile testing machine have been noted, indicating appreciable extension of the notch. Therefore, it is believed that impact in itself is not important. It is merely a means of obtaining a moving crack in a specimen where a general stress level exists. As mentioned earlier, the length of specimens tested has been purposely kept long so that secondary effects from the transient stress due to impact will not affect the results.

An interesting experiment was conducted to attempt to demonstrate how impact might set off a tank failure. In this test a conventional specimen was prepared with the liquid nitrogen crack about 2-7/8 inches long. This crack was then "buttered over" with a thin weld deposit on both sides of the plate. The resulting cross section is shown in Figure 10. This specimen was cooled to -40°F and then gradually the load was increased until the brittle fracture occurred. The average stress at failure was 27,400 psi. This is 8,000 psi below the yield point for this material at the temperature of the test. It is believed that if the amount of the weld deposit were less or the deposit less ductile, failure could occur at a lower stress, perhaps even approaching the stress found in the S.O.D. test.

Geometry and Size

The effect of specimen geometry and size has been studied quite extensively below the S.O.D. transition temperature and it has been found that variations in width, length, shape, and size have no effect on the brittle breaking stress. Some work has been done at higher temperatures and still more is required to fully explain variations found. The present extent of knowledge is outlined as follows.

Width Effect

Specimens from Plate "A" of varying width have been tested at 0°F, the S.O.D. transition temperature, and +40°F which is above this transition. The results are shown in Figure 11. This work was carried out using the small impact gun with a medium charge and, under these conditions, it was found that at 0°F the brittle breaking stress was constant at 14,000--15,000 psi. At 40°F it was found that an increase in width resulted in an increase in breaking stress. This effect has been checked for different length specimens and it has been found that a family of curves of varying slopes exist, two of which are also shown in Figure 11.

No obvious explanation exists for this behavior, but it may well be due to the resistance offered by the specimen to the wedging action which was employed in starting the crack. Fortunately, this effect is not extreme.

Length Effect

Specimens of varying length were also tested under similar conditions and the results are shown on Figure 11. As mentioned

FIGURE 10

SPECIAL TEST SPECIMEN

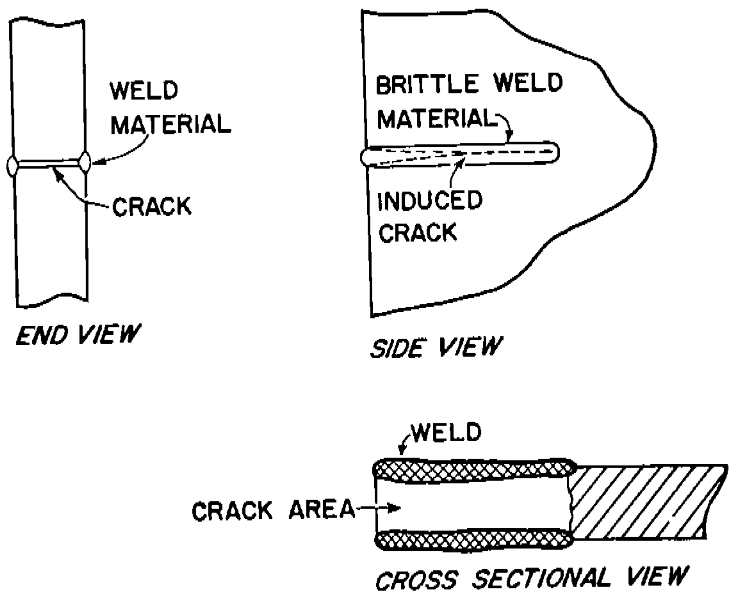
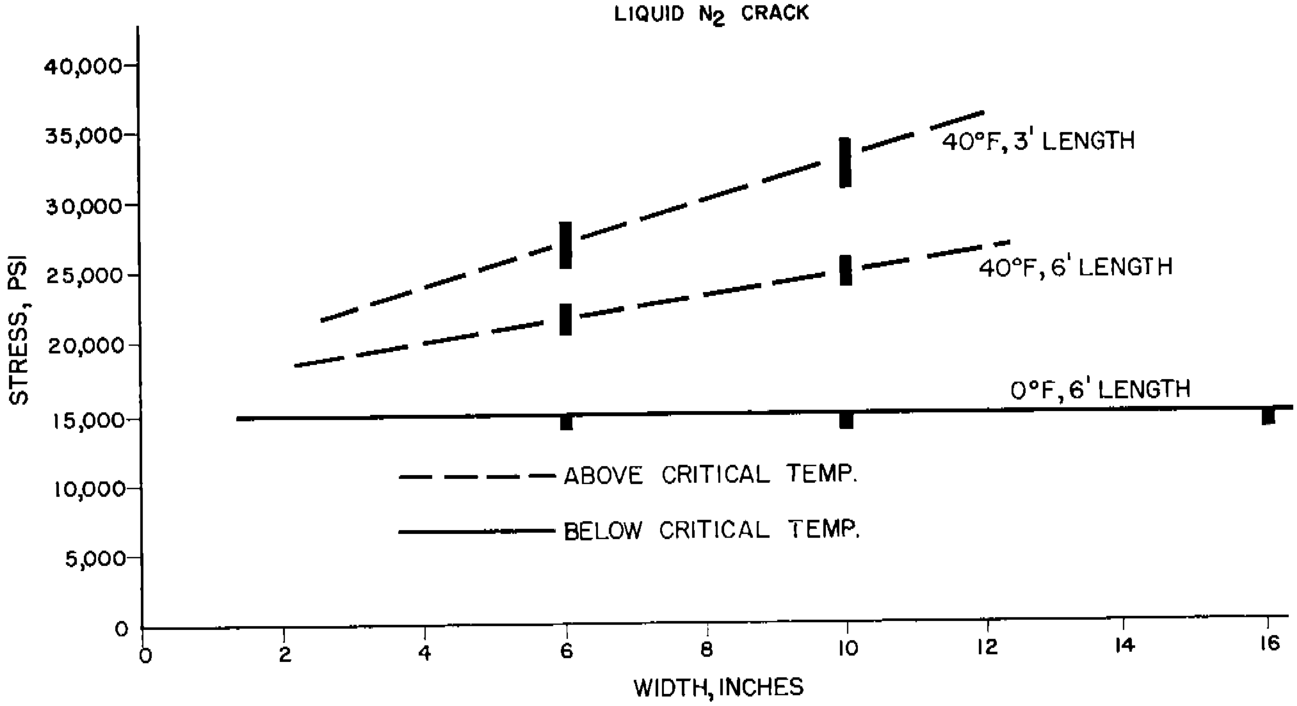


FIGURE 11

WIDTH EFFECT

PLATE "A" ASTM-A-285
SMALL GUN, MED. CHRG.
LIQUID N_2 CRACK



previously, no effect was found where the temperature was 0°F, which is the S.O.D. transition temperature, for Plate "A". At +40°F, the length affected the brittle breaking stress, being lower for longer specimens. It will be noted that as the specimen length increased the reduction in stress became progressevely less and approached a constant value. Data on length effect, using specimens of two different widths (6-inch and 10-inch) and testing at +40°F are also shown in Figure 12. As might be expected from data in Figure 11, the wider plates had higher breaking stresses. It has been found, however, that if the length of the wider plate is increased sufficiently, this difference disappears.

Size Effect and Shape Effect

Using the data accumulated for the studies of width and length effects, it is possible to reach certain conclusions regarding size and shape effects.

It is apparent that at or below the S.O.D. transition temperature no size or shape effect exists since changes in both width and length have no effect.

Above the S.O.D. transition temperature, in the tests conducted on Plate "A" at +40°F, no effect of size was noted on two geometrically similar specimens.

It is planned to broaden the range of this work in the future with larger-scale tests at one of the universities.

On the basis that geometrically similar pieces of different size will give the same results, all the test data have been plotted on a single curve for shape effect. This is shown in Figure 13. Here the brittle breaking stress vs. the length-to-width ratio (L/W) has been plotted. For tests at the S.O.D. transition temperature (0°F), the stress is, of course, constant. At a higher temperature (+40°F), the brittle breaking stress falls off as the L/W ratio is increased and approaches a limiting stress asymptotically. This relationship provides a simple means of making adjustments to data obtained on one size specimen for comparison with another. It also indicates that it is desirable to use a specimen of L/W ratio of about 10 to insure that results close to the minimum brittle breaking stress are obtained when tests are run above the S.O.D. transition temperature.

Miscellaneous Variables

Although all of the principal test variables have been covered, there are a few miscellaneous items which should be mentioned.

FIGURE 12
LENGTH EFFECT
PLATE "A" ASTM-A-285
SMALL GUN, MED. CHRG.
LIQUID N_2 CRACK

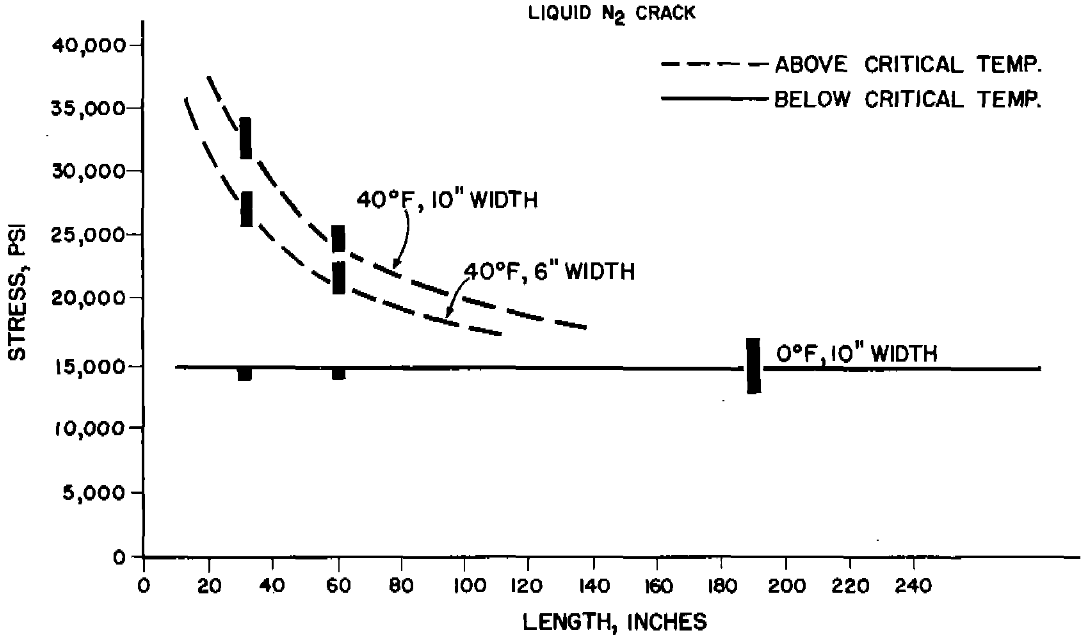
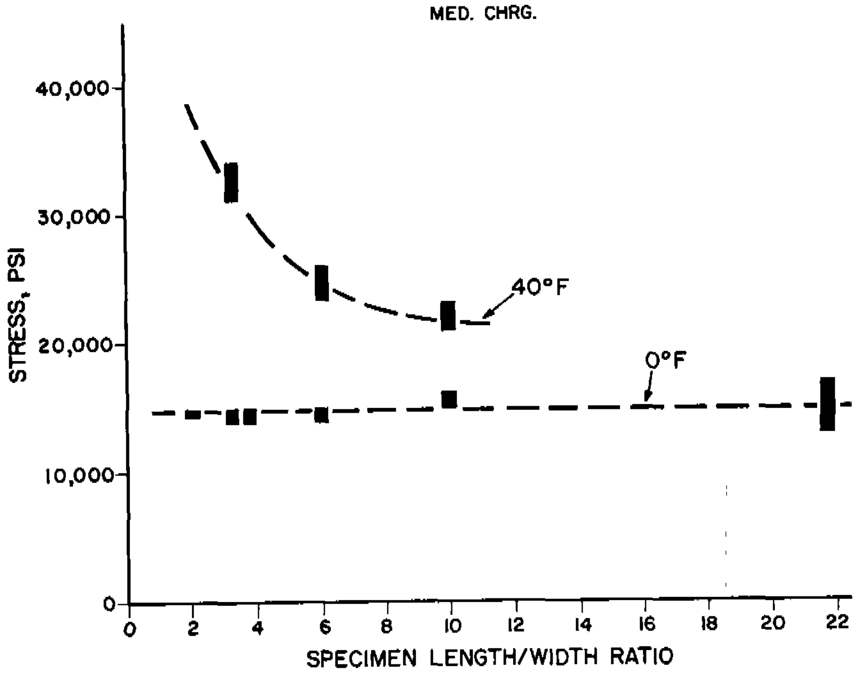


FIGURE 13
SHAPE EFFECT
ASTM A-285-GR-C
1" THICK SPECIMENS
SMALL GUN
MED. CHRG.



Stress distribution in the S.O.D. test specimen has been investigated using strain gauges and found to be fairly good. Stresses in the immediate vicinity of the notch and saw cut are undoubtedly high. However, they drop down to within 20% of the average stress a short distance away where the strain gauges were located. Earlier it was mentioned that a saw cut equal in length to the crack is made on the opposite edge of the plate to insure good stress distribution. Actually, a few early tests were run without this saw cut and the results are identical with later tests employing the cut.

Mention was also made of the effect of repetitive impacts earlier in the paper. More details on this situation are outlined below:

1. Below the S.O.D. Transition Temperature:

Repetitive impacting has little effect. Tests have been made at four or five increasing stress levels and the results were identical with first or second impact tests. At stress levels just below the brittle breaking stress, say within 1,000 psi, repetitive testing will result in failure. Some tests indicate an increase in brittle breaking stress when the number of impacts is excessive, say 5 to 10. This makes it desirable to make a rough survey for brittle breaking stress first and close in on the range in later tests.

2. Above the S.O.D. Transition Temperature:

If no failure occurs on the first impact when the temperature is above the S.O.D. critical point, continued testing is usually of little significance. It appears that the first impact results in dulling of the notch through deformation at the root of the crack and the final breaking stress is usually considerably higher than would be found on first impact. This means that additional tests must be run to see what is the lowest stress at which failure occurs on first impact.

Another interesting effect noted during the laboratory work was the wedge width. When a wedge narrower than the plate thickness was used, the brittle breaking stress was higher than when the wedge was equal to or greater than the thickness.

V. COMPARISON OF TEST RESULTS WITH FAILURES

The next important step in the laboratory program involved testing of steel from the English tanks which failed. The purpose of this work was to determine whether the S.O.D. test would indicate the steel to be subject to brittle fracture at the conditions existing when the tanks ruptured.

Both of these tanks were built of steel purchased to British Specification BS 13. The failure in the first tank started at the junction of the first and second courses in a flaw at the root of the replacement weld made to fill up a weld probe. The failure in the second tank started in a poorly repaired crack at the tee junction between the vertical weld in the first course and the horizontal weld between the first and second courses. The failure started, however, in the second-course plate where the crack extended up into this plate.

The properties of the steel in the first two courses of both tanks are shown in Table II. All plates fell within the rather broad specifications of BS 13 which is the British equivalent of ASTM A-7. Charpy V and U values for all plates are also plotted on this sheet.

At the time of both failures, the temperature was about +40°F. Both tanks were being water-tested when failure occurred, and the calculated stresses at the time are shown below.

First Tank Failure	Calculated Stress
1st Course (top)	14,000 psi
2nd Course (bottom)	15,000 psi
Second Tank Failure	
1st Course (top)	12,500 psi
2nd Course (bottom)	14,200 psi

The S.O.D. tests were made on plates from both the first and second courses of the first tank which failed, but only from the second course of the second tank. All specimens were 6 feet long by 10 inches wide. The large impacting gun was used and tests were made at varying temperatures from -150°F to +40°F. The results of these tests are shown on Figure 14.

TABLE II
PROPERTIES OF FAWLEY TANK STEEL

SPECIFICATION: BS-13, (SEMI-KILLED)
SOURCE WESSOE, LTD

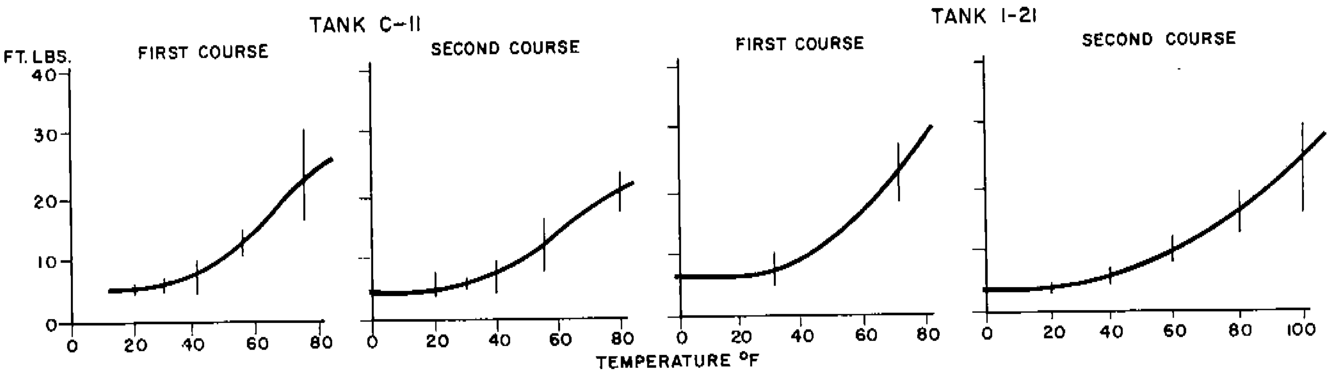
CHEMICAL ANALYSIS

	TANK C-11		TANK I-21	
	FIRST COURSE	SECOND COURSE	FIRST COURSE	SECOND COURSE
CARBON	.165	.21	.245	.22
MANGANESE	.54	.56	.62	.54
PHOSPHORUS	.024	.024	.031	.031
SULFUR	.036	.041	.027	.03
SILICON	.015	.02	.04	.05

PHYSICAL PROPERTIES

	TANK C-11		TANK I-21	
	FIRST COURSE	SECOND COURSE	FIRST COURSE	SECOND COURSE
YIELD STRENGTH, PSI	31,400	33,400	30,400	37,000
ULTIMATE STRENGTH, PSI	59,600	65,800	68,100	66,100
% ELONGATION	31	26.5	38	34
REDUCTION OF AREA	60.4	56.0	57	56

IMPACT DATA CHARPY V NOTCH



IMPACT DATA CHARPY U NOTCH

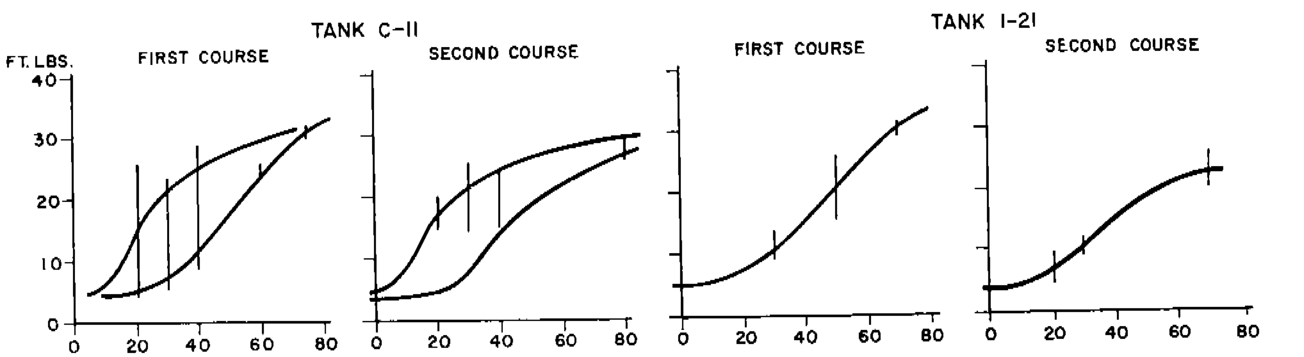
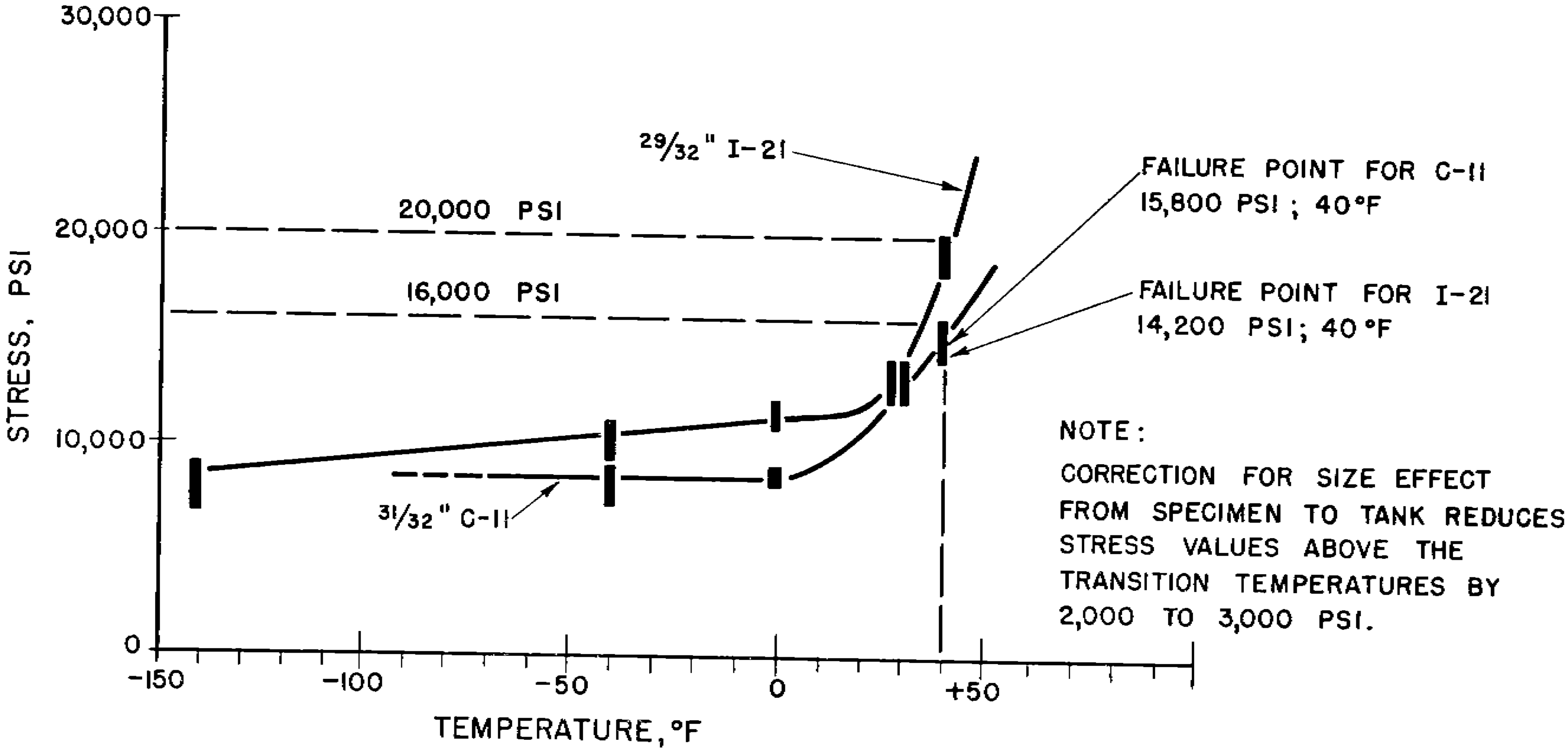


FIGURE 14
STRESS vs TEMPERATURE

FAWLEY TANK STEEL
SPECIMENS 6' x 10" x 1"

LARGE GUN
SECOND COURSE PLATE



These data indicate that at +40°F the brittle breaking stress of the second course of the first tank to fail was 14,000--16,000 psi. The second-course plate for the other tank has an indicated brittle stress of 18,000--20,000 psi. These stresses compare fairly well with the calculated stresses as shown below:

	Calculated Stress	S.O.D. Tests
First tank failure, 2nd Course	15,800	14,000--16,000
Second tank failure, 2nd Course	14,200	18,000--20,000

It is believed that the test results would show lower stresses for both plates if longer specimens were tested. This can be seen from Figure 12, where length effect is illustrated above the S.O.D. transition temperature. Unfortunately, the testing machine in the S.O.D. laboratory cannot take larger specimens; however, it is planned to test a larger specimen in one of the universities. A further point in this regard must also be brought out, and that is that the laboratory tests were conducted on specimens having a length-width ratio of 6.0 while the tank proper has a ratio of 10.0. If a correction for this size effect is made, which has been determined by experiments to be 2,000 to 3,000 psi, the comparison of laboratory results and actual field experience for both tanks is then in even closer agreement. It must be pointed out, however, that any minor change in the notch-brittleness properties of a section of the steel, which would result in a slight shift of the curves, would also affect the correlation between laboratory and field results.

It is felt that satisfactory correspondence between the S.O.D. test results and calculated stress conditions when the actual tank failures occurred has been obtained. These results indicate that the S.O.D. test closely simulates conditions at time of failure and therefore provides a reliable means of evaluating steels with respect to their susceptibility to brittle fracture.

VI. EVALUATION OF STEELS

The following different steels have been tested during the S.O.D. tests:

ASTM A-285	1	
ASTM A-7	1	
BS 13	3	(tanks that failed)
ABS Class C	1	
	6	

The properties of these steels are summarized in Table III, where Chemical Analysis, Physical Properties, and Impact Properties of all are tabulated.

S.O.D. Test Results

The comparative properties of all 6 are shown in Figure 15, where results from the S.O.D. tests are plotted. Several interesting observations can be made on the basis of these data.

1. Wide variations exist in the brittle breaking stress below the S.O.D. transition temperature.

2. Wide variations exist in the S.O.D. transition temperatures for these steels.

3. The susceptibility of any steel to brittle fracture depends upon both factors mentioned above, that is, brittle breaking stress plateau and transition temperature.

4. One steel appears to have two transition temperatures, while another appears to have a gradually falling brittle breaking stress at lower temperatures.

5. No conclusions on relative merits of different types of steel can be reached on the basis of these few tests.

Correlation of S.O.D. Results with Properties of Steels

It is fairly obvious that it would be impractical to test every plate used in oil tanks or any other major structures by the S.O.D. test to determine whether it would be safe from brittle fracture at proposed operating conditions. Throughout the investigation, therefore, an attempt has been made to correlate S.O.D. test results with common physical properties as determined by simple conventional tests. After reviewing results from the steels tested, a fairly good correlation between the S.O.D. transition temperature and Charpy V notch impact results was found. This correlation is shown in Table IV. A Charpy V notch value of 5--6 ft. lbs. corresponds to the point where the slope of the energy vs. temperature curve starts to increase markedly from the very low energy values. From the tabulation it can be seen that the temperatures are in fairly close agreement considering the spread in data possible from Charpy tests.

FIGURE 15
STRESS VS TEMPERATURE
COMPARISON OF VARIOUS STEELS
TESTED IN EED LABORATORY

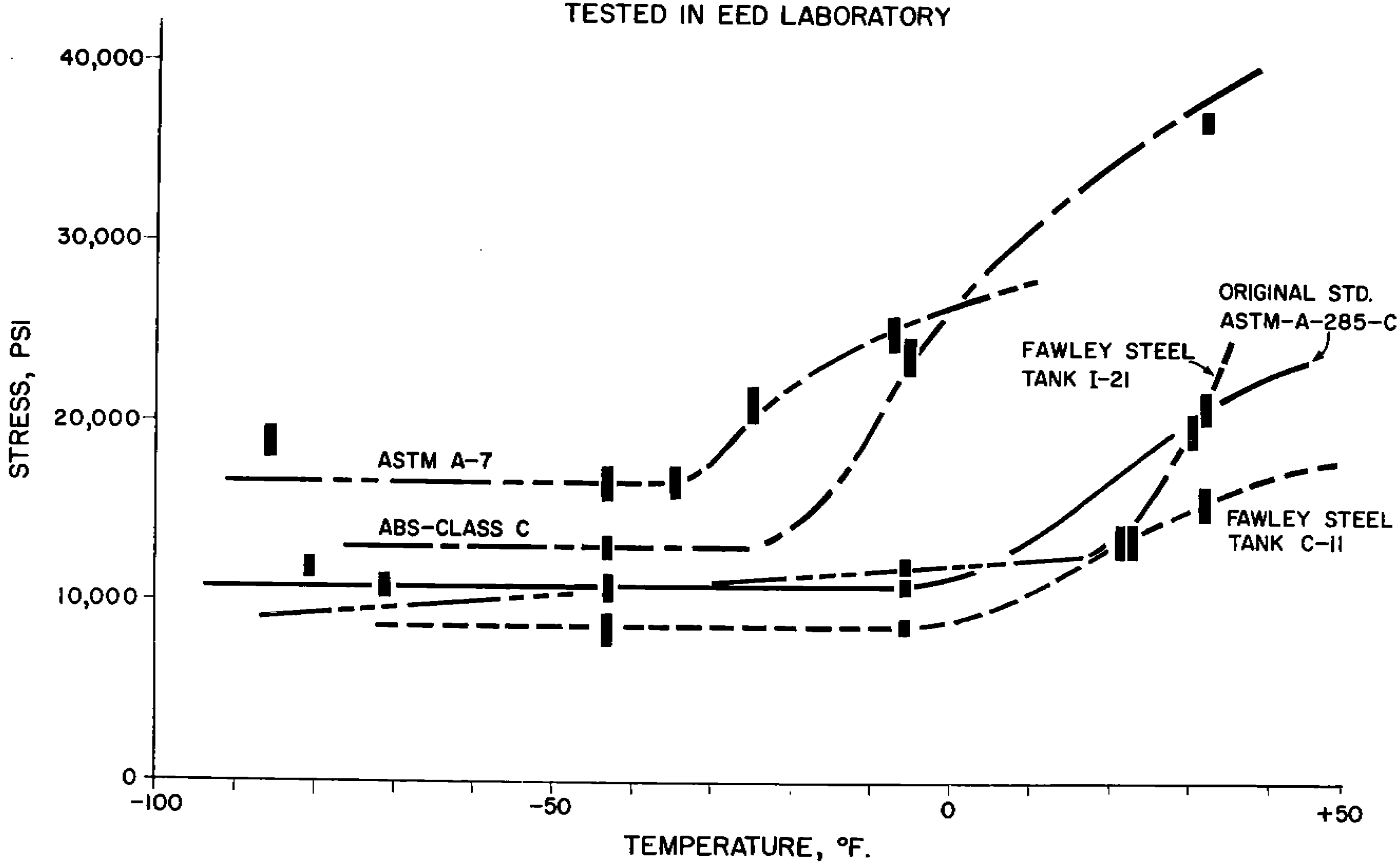


TABLE III

PROPERTIES OF ALL STEELS TESTED

Steel Plate	A	B	C	D	E	F
Source:	Lukens (Rimmed)	Inland Steel	First Course Fawley Tank C-11	Second Course Fawley Tank C-11	Second Course Fawley Tank I-21	Bethlehem Steel
Specification:	ASTM-A-285 GR C	ABS Class C	BS-13	BS-13	BS-13	ASTM A-7
Chemical Properties:						
Carbon	.25	.20	.165	.21	.22	.20
Manganese	.49	.71	.54	.56	.54	.47
Phosphorus	.016	.014	.024	.024	.031	.008
Sulphur	.040	.030	.036	.041	.03	.033
Silicon	--	.25	.015	.02	.05	.04
Physical Properties:						
Yield Strength	38,300	40,200	31,400	33,400	37,000	33,400
Ultimate Strength	59,500	68,800	59,600	65,800	66,100	61,200
% Elongation	--	25	31	26.5	34	--
Reduction of area	--	--	60.4	56	56	49.9

Impact Data:

	Temp.	Ft. Lbs.	Temp.	Ft. Lbs.	Temp.	Ft. Lbs.	Temp.	Ft. Lbs.	Temp.	Ft. Lbs.	Temp.	Ft. Lbs
U-Notch:	62	30,29,28	72	38,34,30	76	32,29,28	80	26,28,29	72	20,26,25	72	34,34,36
	32	24,24,20	20	32,29,29	58	24,23,25	40	15,22,23	32	8,10,11	32	10,28,30
	20	23,20,13	-20	6,22,23	30	4,17,23	30	6,6,25	20	5,4,9	20	7,12,30
	0	8,6,6	-50	16,2,2	20	4,6,26	20	4,6,20	0	3,6,3	0	3,5,5
			-65	2,3,2			0	3,3,3			-20	3,4
V-Notch:	88	23,23,21	68	60,59,47	76	32,19,17	80	18,22,23	125	42,43	72	31,31,38
	52	13,10	32	28,20,24	56	15,12,13	54	10,15	100	15,23,20	32	8,10,10
	40	9,9,12	20	18,18,18	40	6,8,9	40	4,6,8	60	8,10,12	20	4,6,8
	30	7,8	10	15,14,14	30	6,6,7	30	5,5,6	40	6,7,6	0	4,4,5
	20	6,6,8	0	7,10,7	20	4,6,4	20	6,7,4	20	5,3,5	-20	2,4
	10	4,6	-25	4,6,7			0	4,4	0	2,3,4		
	0	3,3,4										

TABLE IV
CORRELATION OF S.O.D. TRANSITION TEMPERATURE
WITH CHARPY V-NOTCH

Material	5--6 Ft. Lb. Charpy Value	S.O.D. Brittle Fracture Test
A - ASTM A-285	0 -- +10 F	0 -- +20 F
B - ABS Class C	-20 -- 0 F	-20 F
Fawley Steel		
C - 1st Course C-11	0 -- +20 F	10 F
D - 2nd Course C-11	0 -- +20 F	10 F
E - 2nd Course I-21	+20 -- +40 F	30 F
F - ASTM A-7	-20 -- 0 F	-30 -- -20 F

Much greater difficulty was experienced in finding some property which correlated with the brittle breaking stress plateau below the S.O.D. transition temperature. Yield and ultimate strength seem to bear no relation, and impact energy values from Charpy tests do not seem related in any simple manner.

Very recently, an effort was made to relate the modulus of rupture in bending at liquid nitrogen temperature to the brittle breaking stress from the S.O.D. test and the preliminary results of this work are shown in Table V. This correlation shows promise but it is still too early to be certain that it is valid.

VII. FUTURE WORK

The present program planned by the Standard Oil Development Company will include:

1. Further tests to compare ASTM A-7 with ABS Class C.
2. An investigation of the brittle properties of thinner plates.
3. Additional work on size effect and impact effect, using a larger testing machine.
4. Further work on correlations between S.O.D. test results and physical properties.

TABLE V
CORRELATION OF BRITTLE BREAKING STRESS PLATEAU WITH MODULUS OF RUPTURE IN BENDING

(Liquid Nitrogen Temperature)

Material	S.O.D. Brittle Fracture Stress Plateau, psi	Modulus of Rupture in Bending, psi
A - ASTM A-285	10,000--11,000	171,600 ; 179,500
E - Fawley Steel Second Course Tank I-21	8,000--9,000	160,000 ; 186,000
B - ABS Class C	12,000--13,000	222,000

VIII. ACKNOWLEDGMENTS

The authors wish to acknowledge with gratefulness their indebtedness to their associates and the management of the Standard Oil Development Company who made possible the rapid progress in these studies. The assistance of Professor Gensamer of Columbia University, who acted in a consulting capacity throughout the program, was invaluable. The tests themselves were capably run in the laboratory by Messrs. R. A. Haisch and S. F. Herbster.

DISCUSSION

by

Dr. M. Gensamer
Columbia University

It seems difficult to explain the results presented by Mr. Feely and his associates, just as it has been difficult to explain the earlier results of large plate tests in which the crack was initiated simply by increasing the tensile load until a crack started, by any adaptation of the Griffith concept. Such treatments have been advanced, first by Irwin(1), then by Orowan(2) in a simple and elegant form based on Inglis' solution of the elastic problem of an internal notch, and recently by Wells(3) for an external notch by using the Neuber solution for such notches. It seems that while the Griffith criterion, that the energy consumed in advancing the crack be less than that released from the store of elastic strain energy, must almost surely be a necessary condition, it cannot be the only one.

In the Griffith-type calculations of Irwin and of Orowan, of the stress required to keep a crack moving, the stress varies as the square root of the reciprocal of the crack length. In these tests, as in the older ones, failure should occur then at any stress, however low, if a sufficiently long crack is provided. In these tests crack lengths were varied by a factor of more than four, so that the stress should have varied by a factor of two; no such variation of stress with crack length was observed. In the older tests no complete failures were observed below the yield point, although the yield point was approached as a limiting value by very wide plates. In Wells' formulation, based on Neuber's treatment of external notches, the stress should vary as the square root of the reciprocal of the net width of the plate. This solution predicts less variation with crack length, but still too much, and again provides for no limiting lower value of the necessary stress.

Wells derived the formula

$$\sigma = \sqrt{\frac{\pi Es}{4a}}$$

for deep external notches and the formula

$$\sigma = \sqrt{\frac{3Es}{2a}}$$

for shallow external notches, where σ is the stress, $\underline{E}$ is Young's modulus, $\underline{S}$ is the specific energy requirement per unit area of plate that is to be cracked, and a is half the net width of the plate or the distance from the root of the crack to the centerline. Applying these to the data under discussion, and choosing 16-inch wide plates as the more favorable case, with crack lengths of 3/4 inches and 2 inches so that 2a has values of 14.5 inches and 12 inches, the ratio of the stresses necessary by Wells' formulas should be 1.10. This is near the limit of error in any individual test, but should have been observed had the stress varied with crack length in this manner. For narrower plates the stress variation by the Griffith-Wells treatment should have been greater. The absence of any size effects in the data reported by Feely et al. for all tests below their transition temperature is inconsistent with any of the Griffith-type treatments.

Gensamer(4) treated the problem of the length of crack that would be developed in a plate of limited size, deriving the expression

$$x = \frac{\ell b}{2E} \cdot \frac{S^2 - S_o^2}{q}$$

where x is the crack length that develops, ℓ is the effective plate length, b is the plate width, E is Young's modulus, S is the applied stress, S_o is a limiting stress level below which the crack cannot develop, and q is the specific energy requirement (q in this formula is the same as S in Wells' formula above, and the same as p in Orowan's formula presented at the same conference, for the infinite plate,

$$\sigma = \sqrt{\frac{Ep}{b}}).$$

In this formulation the concept is introduced that some critical value of stress is required, below which the crack cannot propagate however much energy may be available. Gensamer took ℓ to be six times b, on the basis that cracks travel at about one-third the speed of sound so that elastic strain-energy stored at points more remote than three times the plate width was not available in time to be used, and giving q the same value as in a V-notch Charpy test at the 10 ft-lb level (1000 in-lb per sq. in.), calculated with good agreement the slope of a plot of the square of the stress causing failure against the reciprocal of the plate width at temperatures just below the fracture-appearance transition temperature for the wide plate tests conducted under the Ship Plate program. Such a plot extrapolates to the yield point for very wide plates. At such high temperatures fully developed plastic areas would be expected at the notch root, and an average stress higher than the yield point would be required to raise the stress at the root of the notch to the value required for fracture, even though the notch be quite sharp.

At lower temperatures, and with sharper notches such as real and running cracks, one would expect the critical stress to be less than the yield point, but not zero. There is some plastic action associated with the progression of a crack, even at temperatures below the transition temperatures reported by Feely et al. Any plastic deformation at the root of an advancing crack decreases its effectiveness, by relaxing the stress at the crack root. Neuber(5) developed the formula

$$\frac{S}{p} = \frac{4}{\pi}\sqrt{\frac{a}{r}}$$

where S is the stress at the crack root, p is the average stress on the net section, a is the distance from the crack root to the centerline, and r is the radius of the cylindrical zone at the advancing crack edge, in which the stress is relaxed by plastic deformation. It is just as though a hole of radius r had been drilled at the end of the crack, for no load is being carried there because of plastic yielding. This makes the effectiveness of the crack very much less than would be the case if the action were wholly elastic. It means that some stress is required to keep the crack going, to produce the plastic action that seems to be a necessary prelude to fracture. Gensamer has observed, but not yet published, that some deformation (about two per cent) precedes fracture in well polished tensile test specimens even at 25°K, just above the hydrogen boiling point.

It is interesting to calculate what might be the radius of the stress relaxed zone from the data of Feely et al. They observed that unnotched, polished bend specimens failed at 79°K (boiling point of nitrogen) at stresses of the order of 200,000 psi. Such a value might be expected for the yield stress at the temperatures and high rates of straining prevailing in the plastic zone at the edge of an advancing crack. Their observed stress to keep a crack running, after presenting the stressed plate with a running crack, was of the order of 10,000 psi. This makes the stress concentration factor about 20. Putting this equal to $\frac{S}{p}$ in Neuber's formula, $\frac{a}{r}$ takes the value of 250. With a having a value of 6.25 inches, as in their relatively deeply notched 16-inch plates, r is seen to be of the order of 0.025 inch; with narrower plates it would be proportionately less. This is not an unreasonable value, keeping in mind the roughness of the fracture surface and how the crack progresses by the nucleation and growth of new fractures in advance of the moving front, these joining up at different levels to produce the characteristic "chevron" markings.

It seems, though, considering the constancy of the fracture stress level in the work under discussion, both with changes in geometry and in temperature (below their transition temperature, which I venture to name the crack-brittle temperature), that Neuber's formula is not really applicable to such wide plates, and that the stress concentration factor is in fact constant when we are dealing with deep cracks in wide plates. The data suggest that this stress concentration factor might be about 20, and that further increase in the plate width would not cause any change in this value. By testing narrower plates, until a size effect sets in, one might evaluate the effective radius at the root of a crack.

From a practical point of view, this limitation on the stress concentration factor is of very great interest, for it holds out the hope that we can design safely for even quite brittle materials, by keeping the average stress below a critical value. For something as brittle as glass, this critical stress would be quite low, for there is certainly little deformation at the advancing crack edge to limit the effectiveness of the crack as a stress-raiser. But steel down to the temperatures used in these tests does not seem to be glass-brittle, and the stress concentration at the root of a running crack is limited. It would be extremely interesting to extend these tests to still lower temperatures to see if at some very low temperature higher stress concentrations and lower critical stresses might be obtained.

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DISCUSSION

by

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The discussions on the mechanism of brittle cracking have been most illuminating. For the most part they have been confined to fundamental considerations largely concerned with crack initiation. It seems noteworthy that, after about a decade of study in a number of laboratories by experts in the field, many questions are still unanswered; in fact, it would seem that on some points there is disagreement. The fundamental studies are continuing, and there is no doubt that they should continue. The more we know of the fundamentals relating to a problem the better can we cope with that problem.

However, while this theoretical work is going on, structures must be built with reasonable assurance as to their suitability for service. In our case we must build better and perhaps bigger oil storage tanks. In order to do this the designing engineers must be provided with numbers rather than ideas or theories, some of which may be controversial.

After studying several tank failures and having a general knowledge of the art of welding, we came to the conclusion that perfection in workmanship was impossible of attainment and that the plate itself might contain cracks or other metal discontinuities. Therefore, crack starters would be present and cracks of varying lengths could be expected. These cracks could be tolerated if they did not continue through the structure, but stopped after a relatively short travel. For this reason Standard Oil Development Company has devoted its major effort in brittle fracture studies to crack propagation rather than crack initiation.

It is anticipated that from this work will come the numbers needed by the engineers, information pointing the way to structural steels more resistant to crack propagation, and a better understanding of brittle fracture phenomenon.

The quenched and drawn steel was better than the hot rolled stock. It has been conclusively demonstrated since the time the plate tests were made that quenching and drawing lowers the transition temperature. This effect is probably due to two changes which occur; namely, that the ferrite grain size is reduced, and secondly that the carbides are uniformly distributed as spheroids rather than concentrated in pearlite colonies.

The 3 1/2 per cent nickel alloy was found to be far supeprior to the carbon steels, having a much lower transition temperature. The beneficial effect of nickel on notch toughness had long been known and the plate tests merely confirmed with large specimens what had been known to be true for small specimens for a long time.

The nominal strength of plate specimens was found to decrease as the width was increased. For 3-inch wide specimens, strengths approaching ultimate strengths of the standard 0.505-inch diameter tensile specimen were found; the tensile strength decreased with increasing width, approaching a lower stress level approximately equal to the yield point.

The effect of specimen width on the fracture transition temperature was generally small. As an example, for hot rolled semi-killed steel "C" the transition was 100°F for both the 12- and 72-inch widths. The same situation prevailed for the other steels and for transition temperatures based on energy absorption.

Plates from one of the heats had been rolled to different thickness ranging from 1/2 to 1 1/8 inches. Three-inch wide edge notched specimens were made from plates of each thickness. Fracture transition temperatures ranged from 30°F for the 1/2-inch stock to 125°F for the 1 1/8-inch thick specimens. This difference was attributed to two influential factors; one was the finer grain structure of the thinner plates; the second was the effect of specimen thickness on the severity of the notch (higher degree of triaxial tension stress in thicker plates with identical notches).

The effect of microstructure was eliminated in a second series which was machined from the thickest plate stock. These specimens varied in thickness from 1/2 to 1 1/8 inches and had hacksawed edge notches like the previous set. The 1/2-inch specimens had a fracture transition temperature of 85°F as compared with 125°F for the thickest specimens. Above one inch, increasing the thickness seemed to have no effect.

UNIVERSITY OF CALIFORNIA TUBE AND FLAT PLATE TESTS

by

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ABSTRACT

Summary of Results

The materials used in this investigation were three lots of hot rolled semi-killed medium carbon structural steel, one lot of nickel alloy steel, one lot of fully-killed steel and one lot of fully-killed quenched and drawn steel.

The plate specimens used were all 3/4-inch thick and contained a narrow transverse slot having a length equal to one-fourth of the specimen width; the slots were terminated with jeweler's hacksaw cuts. The plates were tested in tension in widths ranging from 12 to 108 inches. Tests were made on each size of specimen at a number of temperatures in order to determine the temperature at which the mode of failure changed from shear to cleavage.

In the tests, observations were made of the following: the maximum load, load at development of cracks, fracture load, energy absorbed to maximum load, mode of fracture, strain distribution over the faces of plates, and thickness reductions near the fracture surface.

Results from the tests show that the fracture mode transition temperatures of semi-killed steels may vary from freezing to well above room temperature. Tests of two lots of steel of substantially the same chemical composition, except for the nitrogen content, revealed that the steel with the higher nitrogen content had a considerably higher fracture transition temperature. However, the ferrite grain size was larger in the steel with the higher nitrogen steel, and this undoubtedly was partially responsible for the difference in behavior. No appreciable difference was found when one lot of steel was tested in the as-rolled and in the normalized conditions. Subsequent tests at Battelle and elsewhere have shown that the effect of normalizing in primarily due to the change in ferrite grain size. If the ferrite grain size is unaltered, there is no appreciable effect (as in the case of the plates); an increase in grain size raises the transition temperature and a reduction lowers it.

Tests were also made at a series of temperatures on three sizes of geometrically similar specimens all machined from the same large plate. Each had a square hole with sides at 45° to the specimen axis; the hole occupied 1/3 of the specimen width. Thicknesses of 0.180 inch, 0.360 inch and 0.720 inch were used; the widths were 3 inch, 6 inch, and 12 inch; notch radii were 0.008 inch, 0.0016 inch, and 0.032 inch. Similar behavior was expected in the plastic range prior to the formation of the first crack. However, this was not found to be the case. At a nominal stress of 28,000 psi, the longitudinal strain at the base of the notch varied from 4 per cent in the 3-inch specimen to 12 per cent in the 6-inch and 20 per cent in the 12-inch widths. Similarity of behavior was not expected and not found for the transition temperatures, which ranged from below 0°F for the 3-inch width to above 100°F for the 12-inch size. The main reason for this difference was attributed to the fact that cracks formed at the base of the notch prior to the final fracture and thus the specimens were no longer geometrically similar at the time of the final failure.

In addition to the work on plates, a number of large welded tubes were tested. They were 10 feet long, 20 inches in diameter and had a wall thickness of 3/4 inch. The tubes were fabricated from hot rolled plate of medium carbon structural steel. Plates were rolled into two half cylinders which were subsequently joined by two double-V longitudinal welds. One tube had no heat treatment, eight were stress-relief annealed prior to welding and three were stress relieved both before and after welding. Tests were made at both 70°F and -40°F; various ratios of axial to circumferential stress were employed. Tests on the tubes revealed that the steel in this form had considerably less strength and ductility than did standard tensile test bars machined from the same steel, e.g., stresses as low as 45,000 psi caused failure in a welded tube at -44°F; the corresponding elongation was only 2 per cent. The test bar ultimate strength was 59,000 psi and the elongation was 40 per cent. All tubes tested at 70°F exhibited considerable ductility prior to fracture. All tubes tested at -40°F, with the exception of the one which was heat treated after welding, were relatively brittle having thickness reductions ranging from 2.0 to 10.8 per cent. The reduction in thickness for the heat treated tube was 31 per cent.

The ratio of principal stresses was considered to play a minor role. Fractures originated as transverse cracks in the welds and thus the longitudinal stress was the controlling one. When the longitudinal stress was too low to initiate a failure, the circumferential stress caused failure to originate in the plate several inches from the weld.

Residual welding stresses played no part in the failures. The minimum longitudinal elongation was 1.6 per cent, which was sufficient to mechanically stress relieve the tubes.

Cracks originating in welds did not form at gross weld defects, such as hot cracks, but may have originated at microscopic defects such as slag inclusions.

DISCUSSION

The transition temperatures were based on fracture appearance and energy absorption. These criteria yield high values of transition temperature. Had the tests been conducted at even lower temperatures and the criterion chosen had been the ductility at the base of the notch, the story would have been considerably different.

As an illustration, the 108-inch wide plate of steel "C" which was tested at 27°F broke with a 100 per cent cleavage fracture and hence was below the transition temperature. An examination of the ductility of this plate revealed, however, that there had been <u>over 10 per cent elongation at the base of the notch</u>. For over a foot along the crack away from the notch the elongation exceeded 3 per cent. Clearly the behavior of this plate, which incidentally was typical, did not correspond to the behavior of ship steel plates in service, where the maximum elongation near the crack origin rarely exceeded one or two per cent.

To clarify this point, perhaps a brief discussion of the meaning of transition temperature would be in order. When notched specimens are tested at various temperatures, a range is found in which an appreciable change takes place in some property such as the energy absorption, the ductility, or the appearance of the fracture. This change sometimes takes place rather abruptly but frequently is spread over an appreciable temperature range. In any particular test specimen, the transition range can usually be determined from any one of a number of different measurements having to do with energy absorption, ductility and fracture appearance. All of these measurements, however, do not necessarily give the same transition range or temperature, even with a single type of specimen.

Since ranges are difficult to work with it is common practice to select some temperature within this range for rating purposes. When the fracture appearance is used as a criterion, the transition temperature is almost invariably higher than when the ductility at the base of the notch is selected. Thus the "ductility transition," i.e. the temperature below which the metal is so brittle that the strain at the base of the notch is small (e.g. one per cent), may be expected to correlate with service performance much better than the fracture transition. Ductility transition temperatures were not determined on the large plates. Hence it must be concluded that a good deal of the value of the tests was lost. It is fair to say, however, that at the time the plates were tested the significance of the "ductility transition" was not appreciated.

You will see a striking example of the fallacy in using the fracture appearance as a criterion of service performance when Professor DeGarmo discusses the results of the hatch corner tests. The Kennedy hatch corner failed 100 per cent by cleavage, yet it had a nominal tensile strength almost equal to the tensile test bar ultimate strength and it deformed a great deal prior to the cleavage failure. This specimen was tested below its fracture transition temperature but well above its ductility transition temperature.

MECHANICS OF BRITTLE FRACTURE IN NOTCHED PLATE SPECIMENS

by

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This is a subject touched by the Swarthmore tests only incidentally.

Width and length effects might be considered with reference to the key question whether, in a very long plate, notched at mid-width and mid-length, all fracture would be brittle. The Swarthmore data suggest no answer. In such a test, as in a short specimen under dead-weight load, fracture would be rapid even though ductile; most ordinary tensile tests of ship steel are like this. But none of the wide plate tests gives reason to suppose that the fracture surface in a very long specimen would tend to be square rather than oblique.

When the ratio of width to thickness in a wide plate specimen is reduced from 96 to 4, the temperature of fracture transition drops by about 45 degrees (Figure 1). This is the result of analysis of all the wide plate data by a uniform procedure(1). It reveals no disparities between results from different laboratories except those attributable to moderate differences in materials.

Gauge-length was always 3/4 width, so that wide specimens were also long specimens, taken in ratio to thickness. Length beyond the gauge lines was not great but was not specially noted. Since the rigidity of the loading system was constant, it contributed an effective additional length also proportioned to width.

Thus it might be supposed that the rise of transition temperature with width was caused by length. However, such a supposition would be incorrect in the cases of the unknown number of small specimens tested in a machine of 1/10 the capacity of the big one. Since data from these small specimens fall in consistently with the rest in the width analysis, they work adversely to the idea of length effect.

The choice between square and oblique fracture is made chiefly by the material. From the point of view of mechanics, a steel at a different temperature is simply a different material. It is hard to explain why either width or length should cause a specimen to act as though made of different

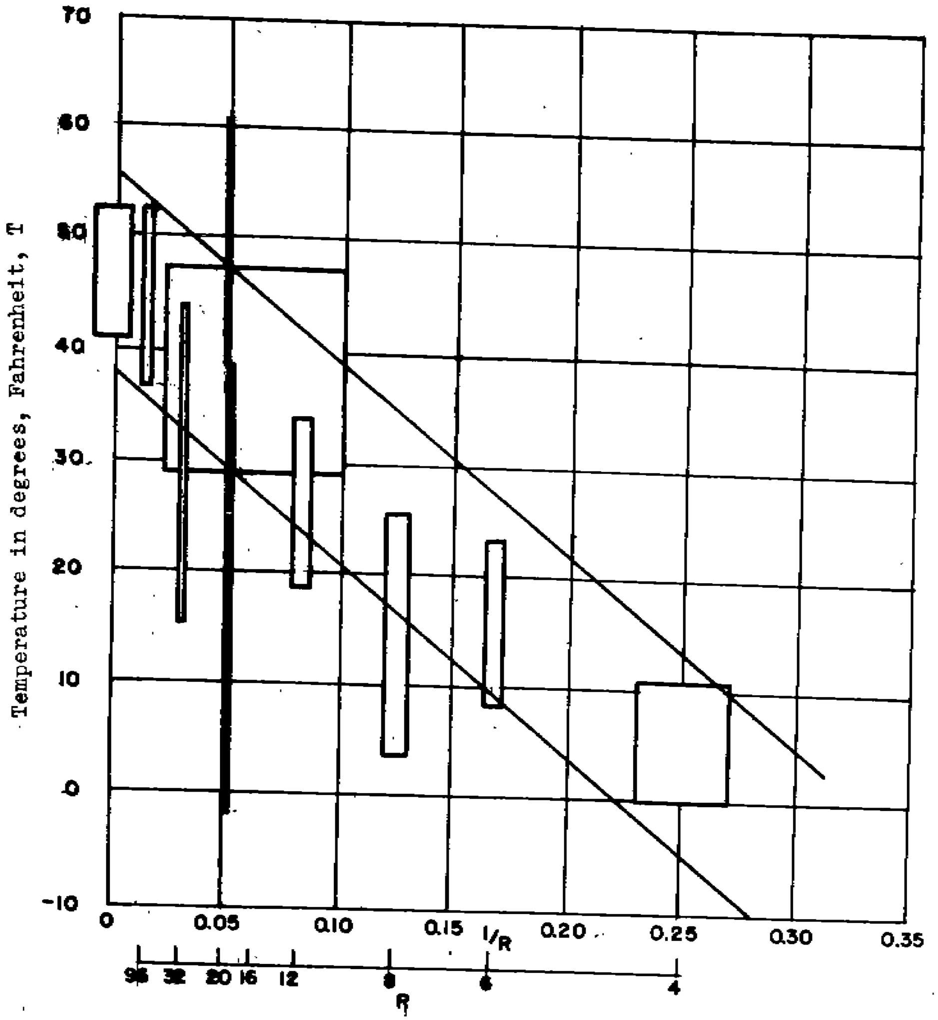

Figure 1 - Effect on Fracture Transition Temperature, T, of Ratio of Width to Thickness, R, in Wide-Plate Specimens

Height of each block gives double standard deviation at each ratio value.

Area of block gives number of specimens. Each square 0.01 x 1 F represents four specimens. Total number, 504.

Straight diagonal band represents relation between temperature and ratio drawn from all specimens.

$$T = 46.4 - \frac{168}{R} \pm 9 \text{ F.}$$

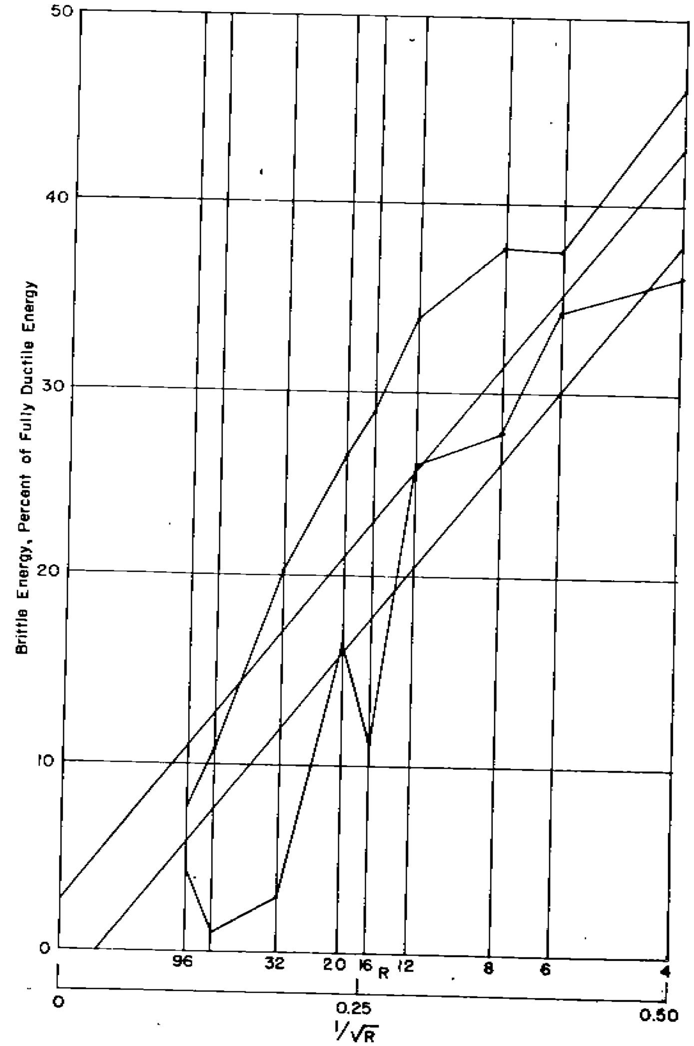

Figure 2 - Total Energy Absorption in the Fully Brittle Condition as Affected by Aspect Ratio

Spread at each aspect ratio is the approximate double standard deviation.

The linear band shows the variation of weighted averages by subgroups fitted by the expression

$$\frac{81.2}{\sqrt{R}} \pm 2.8 \text{ percent}$$

material. Nevertheless it is a fact that the temperature of shift from ductile to brittle action is affected by purely geometrical changes.

For the width effect on ductile energy absorption a simple explanation is found in terms of localization, but the same cannot be said for an effect on a value of temperature, a quantity which cannot be brought into a mechanical dimensional analysis.

<u>Analysis of brittle action</u> in the wide plate tests is conveniently made by evaluating energy absorbed as a fraction of that in the fully ductile condition. When this is done it is found that brittle energy, that absorbed at temperatures below the fracture transition, also varies with width of specimen. Two cases are considered separately.

A. When fracture is all brittle, at temperatures within and just below the fracture transition, energy is near zero in very wide specimens, but rises, roughly in the proportion of the reciprocal square root of the width-thickness ratio, to 40 per cent of the fully ductile value when the ratio has a value of 4 (Figure 2).

B. When fracture is partly brittle and partly ductile, a good correlation exists between the fractional extent of ductile fracture surface and energy absorbed as a fraction of that in full ductility. By using this relation, energy absorbed in mixed fractures can be extrapolated to a value as for completely brittle fracture (Figure 3). The result is quite different from that in Case A. From 30 per cent of the fully ductile energy in a very wide specimen, the fully brittle energy rises toward a value of 100 per cent of the ductile value in a specimen about two thicknesses wide (Figure 4).

The difference between cases A and B indicates alternative modes of behavior at low temperatures: either (A) a drop all the way to a stable reduced energy value with a frequency increasing as temperature is reduced; or (B) a gradual encroachment of partially brittle action with corresponding gradual loss in energy absorbing capacity. Whether the difference in the stable reduced energy values at great widths, as between cases A and B, is significant may, perhaps, be doubted, but it can hardly be doubted that the result in both cases varies with width.

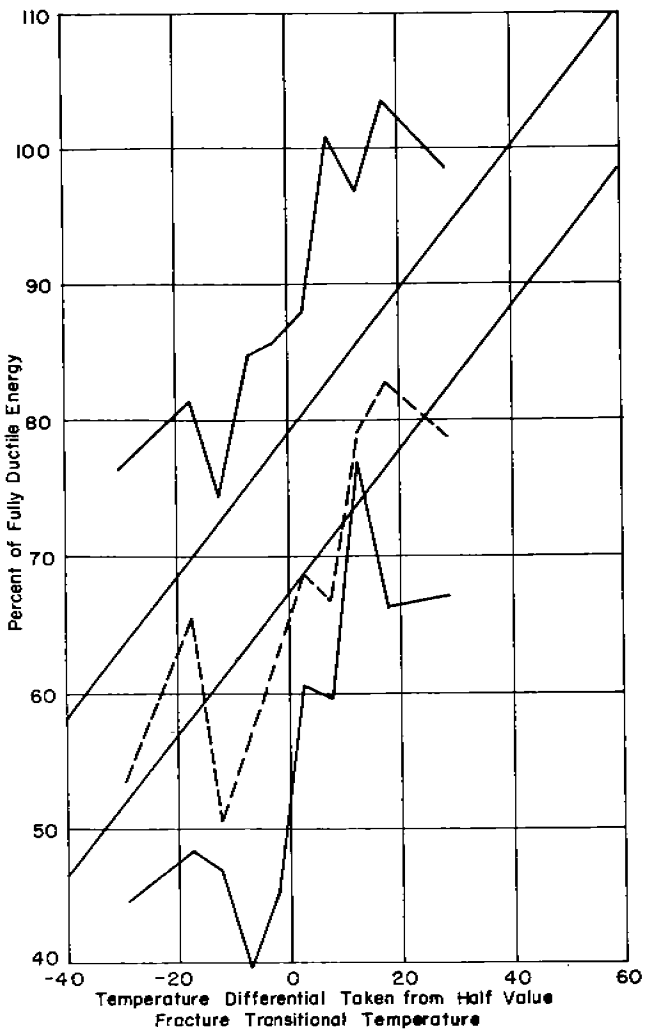

Figure 3 - Effect of Temperature on Total Energy in Specimens with Mixed Fractures as for Median Steel of all Aspect Ratios

In fitted band, energy = 72.54 + 0.517 T ± 5.8 percent of fully ductile value.

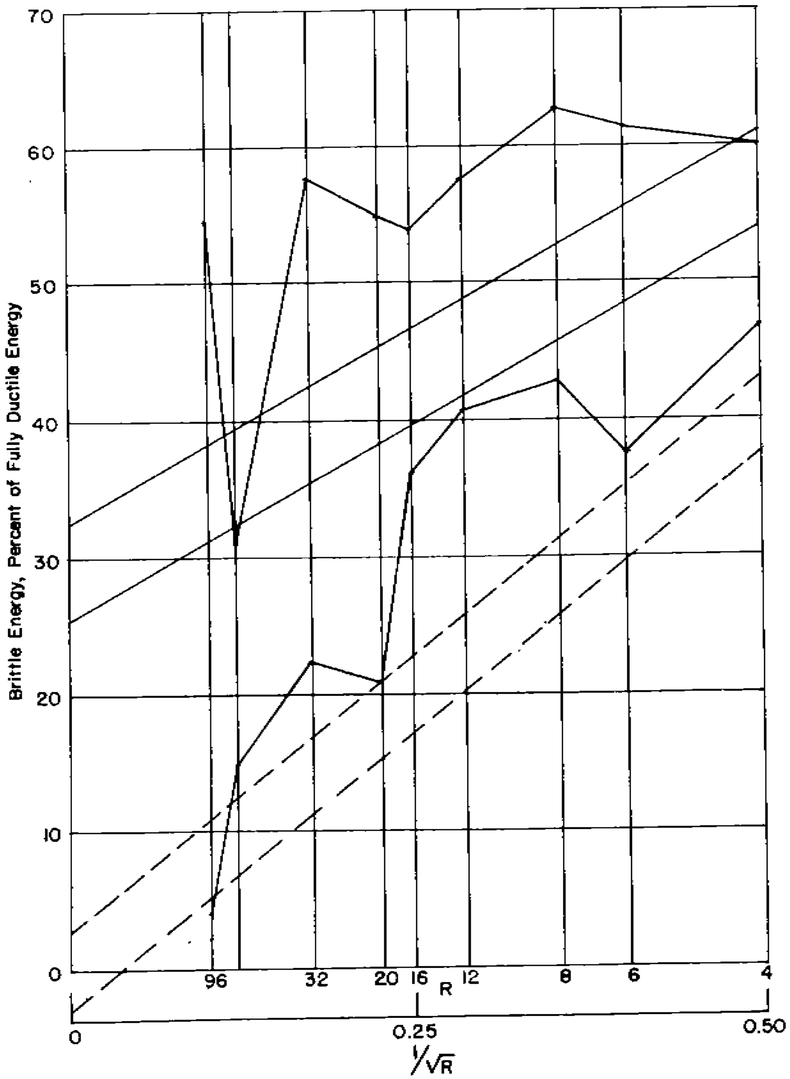

Figure 4 - Total Energy Absorption in the Completely Brittle Condition (Inferred from Specimens in Mixed Fracture) as Affected by Aspect Ratio

Full band shows weighted averages by subgroups fitted by the expression

$$28.8 + \frac{57.8}{\sqrt{R}} \pm 3.4$$

Dotted band taken from Figure 2 based on tests in observed zero shear.

Since width is thus a mechanical determinant of brittle behavior, these results are perhaps suitable for consideration here.

The energy transition also has only an indirect connection with the subject of the conference. The basic feature of brittleness is taken to be low capacity for absorbing energy. Appearance of fracture is then only a symptom and not a very dependable one. The energy transition is therefore more significant than the fracture transition. But there is no evidence in the wide plate data for an energy transition differing in temperature from the fracture transition(1). Energy beyond the point of zero slope, as in the Navy tear test, correlates perfectly with appearance of fracture. There is nothing in the data to suggest a second drop in energy absorption at a temperature below that of the fracture transition. Even if such an effect exists, it can have no direct significance for service behavior, since the range of service temperatures has been thoroughly covered in the wide plate tests.

Summarizing, we note that the fracture transition temperature rises with increasing width; probably the same is true for the energy transition temperature. The loss in energy absorbing power at temperatures below the transition which we call brittleness is more pronounced as width increases.

The effect on transition temperature is greatest at narrow widths, and an azymptotic limit is reached (within the limits of precision of the data) at a width-thickness ratio of 20. Width effect on brittle energy is also azymptotic, and while the limiting value at great widths is not far from zero, it cannot be said that at great enough width all fracture would be brittle.

To make this point clear, imagine two sets of tests, each of a large number of specimens identical in every respect, all specimens in both sets all tested at the same temperature within the range of the fracture transition. Set (b) differs from set (a) only in having a greater width. By this is meant that the thickness and all notch details are the same in both sets, but width and the gauge-length (3/4 width) are greater in (b) than in (a).

The shift in width will have two relevant consequences:

1. In tests with completely brittle fracture, the energy, expressed as a fraction of the energy to zero slope in ductile fracture, will be less in (b) than in (a).

2. The <u>frequency</u> of brittle fracture will be greater in (b) than in (a).

These are average effects, subject to scatter, purely empirical in nature. No connection between the different effects has been demonstrated; in particular, no reason is seen why low brittle energy, as in (1), should cause high frequency of brittleness, as in (2).

<u>Ductile Energy Absorption</u>.

A third consequence of the difference between (a) and (b) is the chief result drawn from the wide plate tests. It may be stated as follows: The average specific energy in tests with <u>ductile</u> fracture will be less in (b) than (a). This is irrelevant to the mechanics of brittle fracture since it relates to the case in which final rupture is by shear, not by cleavage.

If brittle mechanics were to refer only to absolute brittleness (fracture within elastic limits), none of the wide plate data would be relevant. In these tests brittleness in this sense did not occur. Even the most positively brittle fracture was preceded by absorption of large amounts of energy by plastic flow. Since brittleness consists, in the wide plate tests, in only a partial loss of energy absorbing power, the degree of that loss is significant. It can be estimated only with reference to energy absorbed in full ductility. Brittleness is only lack of ductility.

The loss of specific power to absorb energy at a given temperature as width is increased is explained by reference to localization in the pattern of strain distribution(2). As specimens are wider, the relative localization of energy absorption near the apex of the notch is greater. The point of zero slope on the load-elongation curve is reached sooner and the specific energy absorbed is less.

These are effects of relative width, taken in its ratio to thickness. Energy in similar specimens differing only in scale is taken to be proportionate to the cube of the scale factor. However, this is found to be strictly true only after moderate allowances are made for material differences and when similitude extends to all details of the strain pattern. Data on effects of width on specific values of ductile energy are given in Figures 5 and 6.

In general, tests relating to plastic behavior have been of two different types, with different aims. In one of these

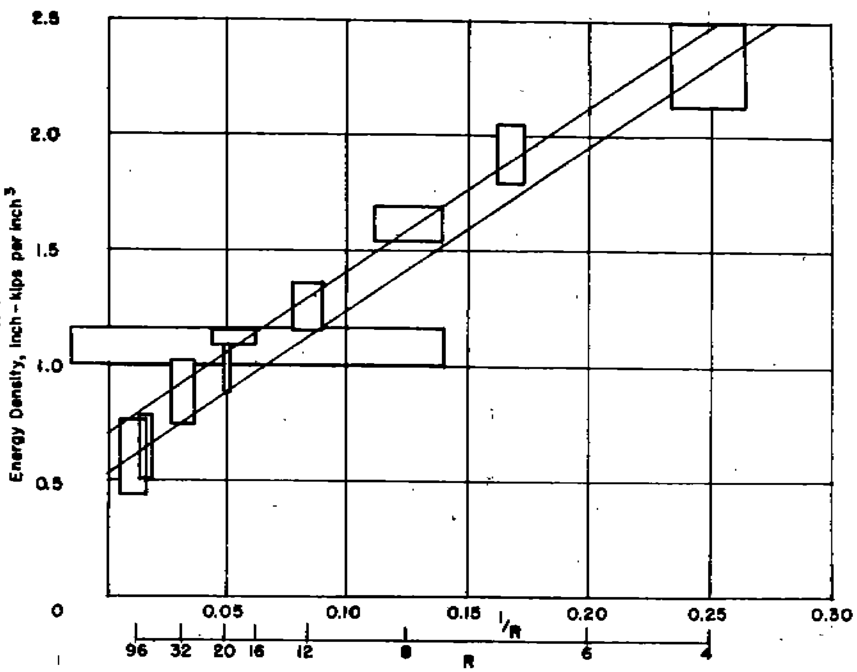

Figure 5 - Effect on Ductile Energy to Maximum Load of Ratio, R, of Width to Thickness, t, in Equivalent 3/4-Inch Plates of Median Steel

Height of each block gives double standard deviation at each ratio value.

Area of block gives number of specimens. Each square 0.01 x 0.1 represents four specimens. Total. 224.

Straight diagonal band represents relation between ratio and energy drawn from all specimens.

$$\frac{E}{R^2 t^3} = 0.619 + \frac{7.09}{R} \pm 0.081$$

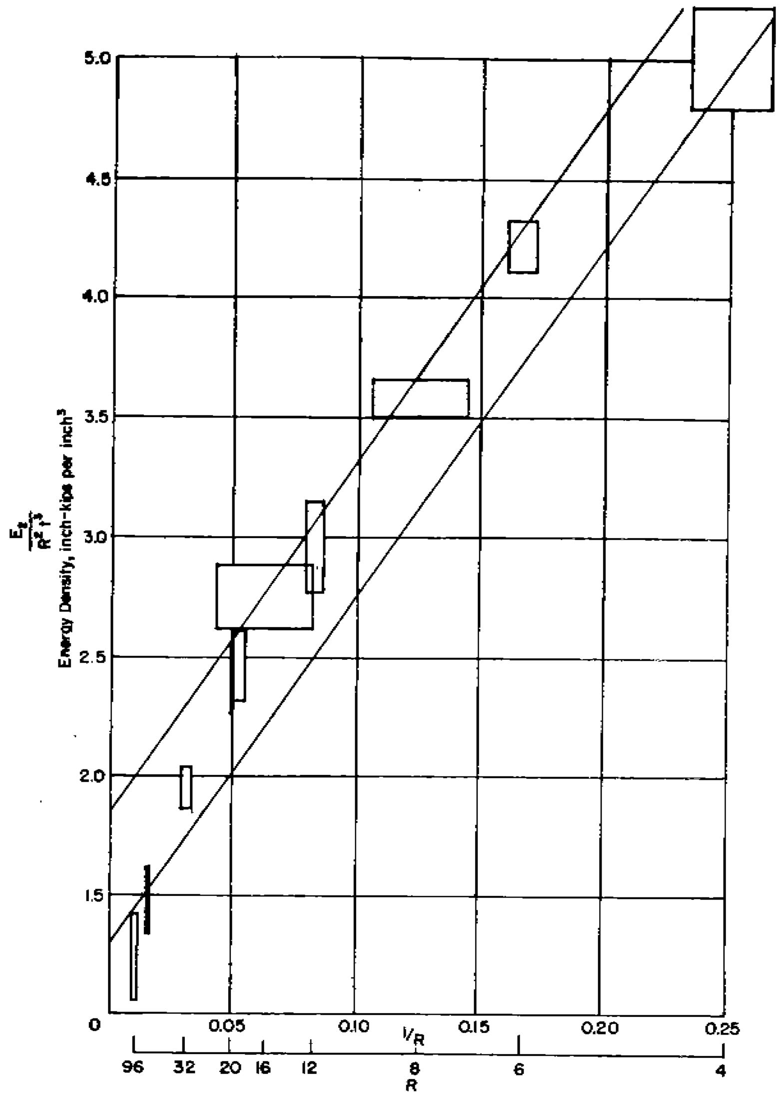

Figure 6 - Effect on Total Ductile Energy of Ratio, R, of Width to Thickness, t, in Wide Plates of Median Steel

Height of each block gives double standard deviation at each ratio value.

Area of block gives number of specimens. Each square 0.01 x 0.1 represents four specimens. Total, 162.

Diagonal band shows relation between ratio energy in all specimens.

$$\frac{E_2}{R^2 t^3} = 1.583 + \frac{14.75}{R} \pm 0.275$$

Energy Field Distributions.

The key to understanding the behavior of complex configurations of loaded metal lies in knowledge of the field distribution within the metal.

For many purposes the field quantity to be studied is stress; design for service within elastic limits has within a generation been revolutionized by increased knowledge of the elastic stress field.

In design involving plastic flow, with reliance on immunity to brittleness, it is more revealing to deal with strain as the field tensor.

If the material has a unique stress-strain curve, unique values of stress, strain, and energy absorption are associated with each other for each point on the curve. Of these three quantities, however, energy has an advantage in being a scalar.

Even if the stress-strain curve is not unique, but varies somewhat with triaxiality, yet these departures from the simple conditions of deformation theory are still conveniently described in terms of energy as the primary variable.

In most engineering calculations the effects of triaxiality still are ignored. When triaxiality is very small the deviation of effective stress from the greatest principal stress is small. Error in working with the greatest principal stress is accepted.

But it is now widely known that triaxiality cannot with safety always be taken to be zero. Values up to 30 per cent have been measured in simple configurations. In this situation reference to specific energy as the field quantity is useful. It may offer a measure of the margin remaining, after a given local strain, before fracture there is to be expected.

No doubt the precision of this practice is limited, but it will carry us so far beyond the older practice of regarding only the greatest principal stress that its use is recommended.

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3. Data on similitude; presented by Earl Parker in the present conference.

the aim has been to establish a basic law, elementary as is the inverse square law; reliance is then placed on calculation, as in celestial mechanics, for prediction of behavior in given configurations.

The basic law of plasticity is expressed in the stress-strain curve, generalized so as to give a relation between the complete stress and strain tensors. This relation varies with material. For purposes of application it has been necessary to make use of fictitious materials with idealized properties.

And even so, calculation has been found possible only in certain idealized configurations (e.g. plane strain). The results of such calculations, therefore, even where mathematical techniques are available, are subject to two causes of deviation from observed data: the deviation of actual from postulated material properties, and the deviation of actual from postulated boundary conditions.

A second quite different method has also been used, leaning strongly on similitude, and using models at reduced scale for prediction of behavior of large structures. Many years ago the question was asked whether considerations of similitude can be applied to plastic flow. Kick's Law, in effect, asserts that it would be applicable under the right conditions, but that ordinarily these conditions have not been realized(3).

In the Swarthmore tests the conditions of similitude were satisfied, with the single exception of material differences caused by the rolling operation.

The wide plates were designed to serve, in small widths, as specimens for material tests, and in great widths as models of a ship's deck. A chain of correlation, based partly on similitude, was constructed to link these two extremes. In principle, the gap between material and structure has thus been bridged.

Success depends, however, on reproduction of patterns of distribution on the two scales, and much work must be done to get reassurance on this point. The method, as actually available for use, is subject to two primary limitations. With respect to material, it is not possible to get complete identity in different thicknesses. Error of this kind can be estimated but not eliminated.

The other limitation is more basic. When we speak of the pattern of strain distribution, we must have regard to all the elements of the strain tensor, and to the succession of states, at each point of the field. It is a tacit assumption in this work hitherto that the strain tensor is sufficiently characterized by the energy value and that the pattern goes through the same transformations in progressive deformation in the model and in the prototype.

A further limitation is imposed in the assumption that the continuity of the metal is unbroken, and it is at this point that the study of ductile behavior comes into contact with brittle mechanics. A hypothesis widely (but often tacitly) accepted is that fracture occurs at a given point in the field when the specific energy absorbed there reaches a limit which is characteristic of the given metal at the given temperature. Mechanical effects on brittleness are exerted through effects on the strain pattern.

Future Work Needed.

1. Triaxiality and Localization. A step in advance of the simple energy hypothesis just stated has been proposed. The idea that brittleness is favored by high triaxiality has been widely accepted although evaluation in numbers has seldom been made. The experiments of G. Welter, with cubes equally loaded on all faces, offer a good demonstration of the need for considering both triaxiality and intensity of stress. The system of testing of H. Schnadt is based on a hypothetical segregation of these two quantities.

In the wide plate tests a segregation was achieved as between metallurgical and geometrical influences on ductile behavior. A similarly quantitative segregation as between triaxiality and intensity of stress is needed; perhaps it might be achieved by a modification of Welter's method.

Until this is done, our unguarded talk about constraint as a cause of brittleness should be considered to be speculative.

2. New Wide Plate Tests. In the wide plate data there is repeated evidence that results in the greatest widths deviate from those which would be obtained by extrapolation from the greater body of tests at moderate widths. Before placing reliance on numerical values as extrapolated to infinite width by use of data at all widths, this anomaly should be resolved. Either the work at great widths should be checked, or we should accept a more intricate formula for curve fitting than that used in Taylor Model Basin Report 853.

3. <u>Error in Use of Models</u>. Some indication of the material effect of rolling is given in the wide plate tests. At equal average strains, specific energy is increased in the ratio of 92 to 76 by rolling down from 3/2- to 1/2-inch thickness. A standardized technique for similar determinations in all scale model tests at reduced scale is necessary.

The alternative of testing assemblies of limited extent in full scale thickness can also be used to bring loads within available limits. The matching of model with service structure in this case is not concerned with similitude since the full scale is established by the thickness. The special difficulty lies rather in matching boundary conditions in test with those in the part as embedded in the more extended structure.

In comparing similar wide plates at different scale values, even a very rough matching of the continuous distribution by a zonal pattern in the model is enough to reproduce the main features of behavior.

In order to place the model method for study of plastically deformed structures on a basis as sound as that for dynamic phenomena like that of ship propulsion, we must follow up this demonstration in wide plates by similar work on structures a little (but only a little) more elaborate, with attention to patterns of distribution and with work on at least two different scales.

Testing of fully developed structural details cannot be expected to give dependable quantitative results until the conditions under which full-scale behavior can be predicted from laboratory tests have been more fully explored. One such condition will be concerned with centering of load, as in the tests of ship bottom structure made at the National Bureau of Standards, but not represented in the present agenda.

An approach to the testing of eccentrically loaded parts of a ship structure might be made by a study of eccentric tear test specimens on different scales, comparable with that made on the symmetrical wide plates. Until some such exploration has been carried out, the qualitative results, which have in some cases been obtained by use of very elaborate but eccentric specimens, might perhaps better be had by use of wide plates, say of non-uniform thickness.

SUMMARY OF DAVID TAYLOR MODEL BASIN AND ALCOA NOTCH-TENSILE TESTS AND OBSERVATIONS ON ENGINEERING APPLICATIONS OF FRACTURE MECHANICS

by

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The presentation is in two parts; the first of which is a summary of four projects:

1. "A Study of Slotted Tensile Specimens for Evaluating the Toughness of Structural Steel"(1).

2. "Transition Temperature of Ship Plate in Notch-Tensile Tests"(2,2a).

3. "Effect of Temperature on Notch Sensitivity of 61S-T6 Plate"(3).

4. "Transition Characteristics of Prestrained, Notched Steel Specimens in Tension"(4).

Projects 1, 2, and 4 were carried out at the David Taylor Model Basin; and Project 3 at the New Kensington Laboratory of the Aluminum Company of America.

The summary of the projects is included to make the background for the conference complete. The second part of the presentation consists of partially related and somewhat incomplete developments and hypotheses bearing on the engineering applications of fracture mechanics.

The first two projects at the David Taylor Model Basin were a part of the larger Ship Structure Committee program employing 12-inch wide slotted steel plates. The ALCOA study employed identical specimens for the evaluation of an aluminum alloy suitable for ship hull construction. The specimens used in these tests are shown in Figures 1(a) and 1(b).

The first series of studies at the Model Basin employed the slot shown in the lower illustration in Figure 1(b). Two parameters were varied in the study; (a) radius at the end of the notch (by varying the diameter of the drill employed). (b) the length of the slot (see reference 1). In addition, a series of 6-inch wide specimens were tested, and a series of 12-inch wide specimens were tested with special notches at an angle of 45° with the plate surface. Fracture appearance was recorded, and energy

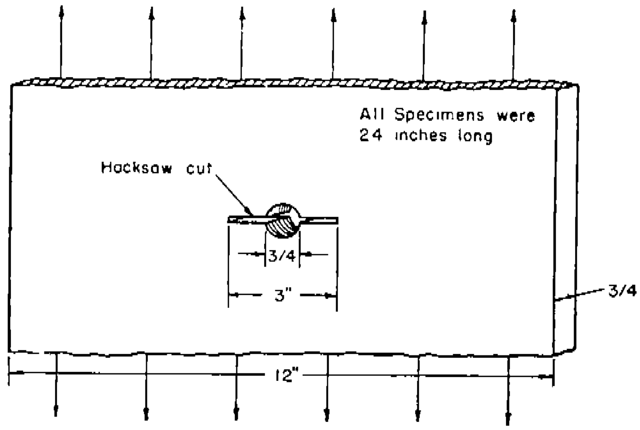


Fig. 1 *(a)*– Form of Tensile Test Specimens

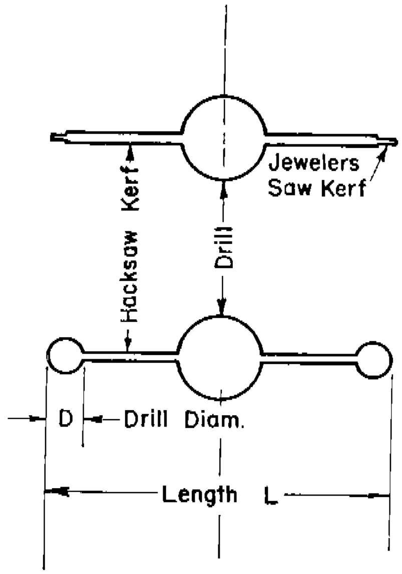


Fig. 1 *(b)*—Details of Slots

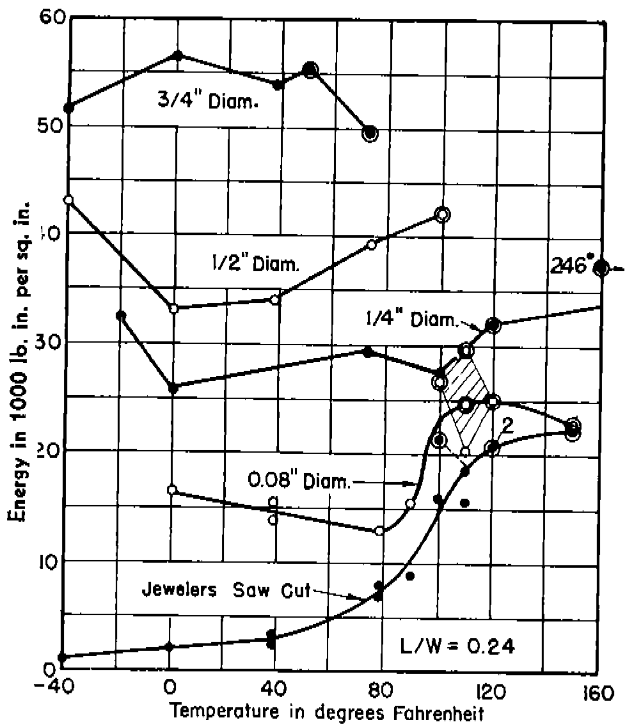


Fig. 2--Effect of Temperature on Energy Absorbed for Specimens of Steel E

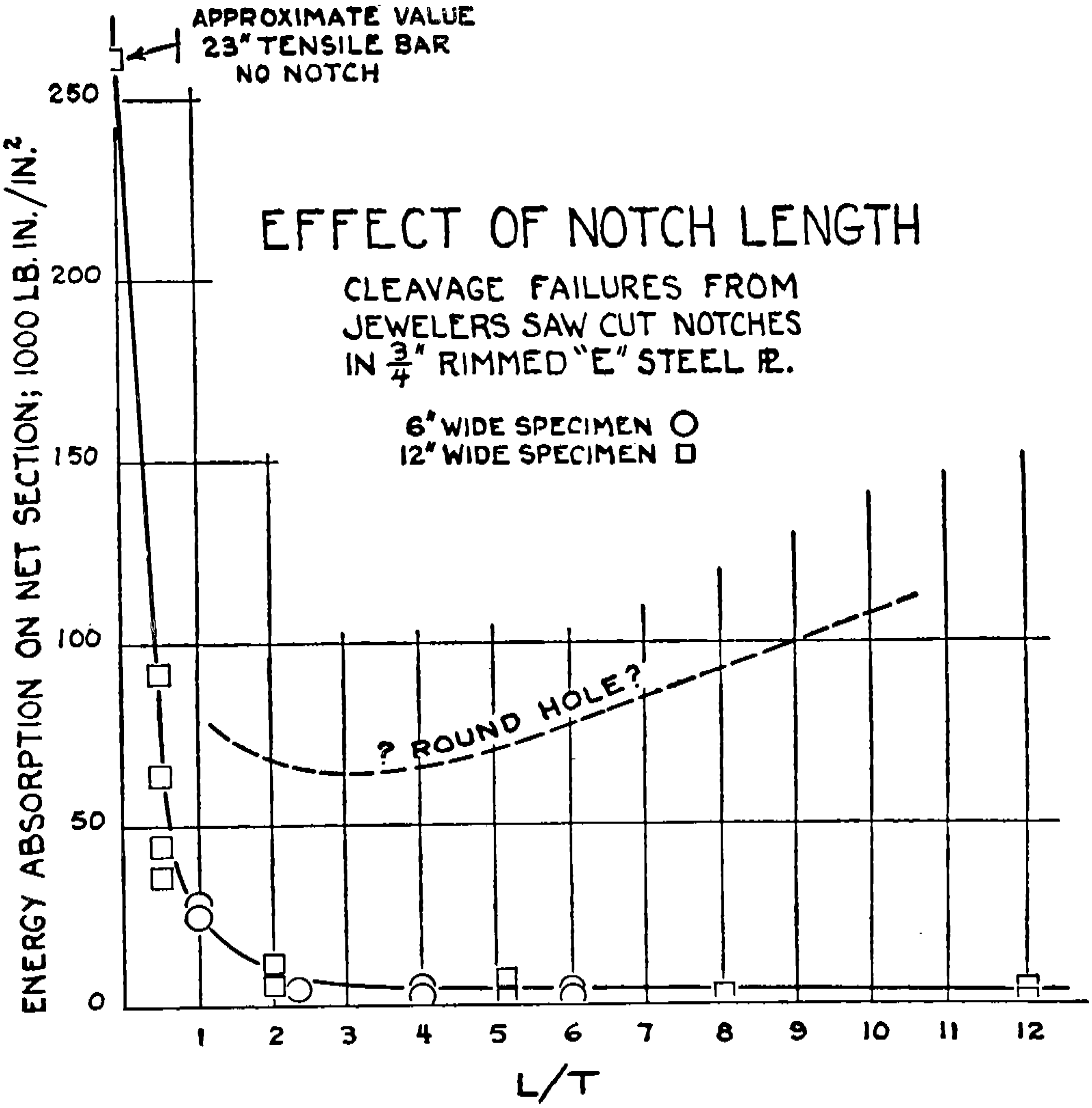


Fig. 3 —Data from Table 1 of the Report Plotted Against L/T Instead of L/W

was based upon extensions over a 23-inch base length. Energy absorption to fracture for specimens exhibiting granular fractures and to maximum load for fibrous fractures was reported. Thus the maximum energies for ductile failures are somewhat more than those obtained over the standard 9-inch base length used in later projects in the program, but the transition temperatures are approximately comparable.

The notch-root radius study covered a range of notches starting with a 3/4-inch diameter hole and running down to a jeweler's saw kerf of an indeterminate but small effective radius. The .08-inch diameter hole and the jeweler's saw kerf gave approximately the same results under test, with the energy-absorption transition temperatures nearly matching. All notches of larger radius resulted in a substantially lower transition temperature and an increase in the energy absorption both above and below the transition temperature. See Figure 2. The finding that a small radius, say 1/24-inch, for a notch in a thick plate, in this case 3/4-inch, results in substantially the same transition temperature as obtained with sharper notches is in conformance with the results from several other studies, including those by Kahn and Bagsar.

The tests with different lengths of slot included also two plate widths. Initial analysis of these data indicated no trend when the energy absorption was plotted as a function of length of the slot divided by the width of the plate. A reanalysis of these data employing the length of slot as independent variable revealed a distinct trend. See reference 1(a) and Figure 3. It showed that for cleavage fractures a substantial amount of extra energy is absorbed when the length of the slot is on the order of the thickness of the plate. It is possible that this may offer a clue as to the size of defect which may be tolerated in a steel structure.

Tests of plate specimens with slots at an angle of 45° with the plate surface resulted in a substantially lower energy-absorption transition temperature when compared to those with perpendicular slots. This is reasonable when it is realized that a 90° rotation would remove the notch effect entirely.

The second listed project was the last series of tests of 12-inch wide plates at the Model Basin, and it added eleven additional steels to the background of information on the subject. See references 2 and 2a. Fracture appearance was recorded and energy to fracture and to maximum load were reported. The energy was based on elongation over a base length of 9 inches, i.e., 3/4 of the specimen width, in conformance with the standard for other projects in the Ship Structure Committee program.

Employing the results of these eleven steels, along with the Ship Structure Committee "Pedigree" Steels D and E, it was possible to make a study of the so-called ΔT hypothesis. This hypothesis presumes that all notch tests arrange steels in the same order in terms of transition temperature and that the temperature difference between the transition temperatures of the different steels is the same regardless of the test employed. Correlation between the energy-absorption and appearance-of-fracture transition temperatures, both obtained with the 12-inch wide specimen, gave a mean deviation for thirteen steels of 2 1/2°F. See Table 1. This could mean that our determination of transition temperature is more precise than we realize.

A good correlation was found between the 12-inch wide specimen, the Kahn tear test, and the 1-inch edge-notched bar which is mentioned later. The basis of this correlation on the ΔT hypothesis is a quantity which may be described generally as energy absorption with the yardstick of energy measurement varying somewhat among the three types of tests.

It appears that the ΔT hypothesis is suitable as an interim means for evaluating roughly the toughness of steel. It can be used during the time that we are faced with such a variety of test specimens.

TABLE 1

Correlation of Transition Temperatures with 12-inch Wide Specimen Energy Absorption Transition Temperature

Specimen, Test	Criterion	No. of Steels	ΔT °F	Deviation, °F Maximum	Mean
12-in. wide specimen	Appearance	13	1	12	2 1/2
1-in. edge-notch, 1/8-in. hole	Thickness reduction	8	38	41 1/2	20 1/2
1-in. edge-notch HSC	Thickness reduction	8	65	34	11
1-in. edge-notch JSC	Thickness reduction	3	75	20 1/2	14
Keyhole Charpy	15 ft-lb	12	-52	40	14 1/2
V-Notch Charpy	15 ft-lb	3	- 1	35	23 1/2
Navy tear test	Energy	12	40	31 1/2	10 1/2

The aluminum alloy tests at New Kensington showed that the 61S-T6 alloy is not sensitive to temperatures within the ship operating range(3). All fractures were silky-shear in appearance, and no cleavage fractures occurred. However, the energy absorption by all specimens was substantially less than that absorbed by steel specimens when they fail with fibrous fracture. In fact, it was closer to the low energy absorption values observed when steel fails with brittle granular fractures. They energy absorption was based upon elongation over a 39.5-inch gauge length, and energy to maximum load and to fracture were reported. The tests made over a range of -60 to +127°F resulted in an energy absorption of approximately 3.5-inch kips per square inch to propagate a fracture. Corresponding tests on medium carbon steel gave about 1.0-inch kip per square inch at temperatures below the transition range but gave about 35.0 for temperatures above the transition range.

The third study at the Model Basin was exploratory in nature: to check a suggestion that the prestraining of notched-steel plates or structures above the brittle-to-ductile transition temperature is a means of increasing toughness in performance at lower temperatures(4). The tests were made with edge-notched bars cut from 3/4-inch plates of four different steels. See Figure 4. As a result of the prestrain procedure employed on the notched specimens, three of the four steels exhibited somewhat greater toughness at a temperature just below the transition temperature.

Not all of the physical effect (Distance A) in Figure 5 is useful for practical purposes. If the prestrain method is to be beneficial, the service strain must actually exceed the strain that would occur in a structure which had not been prestrained. Thus, the absolute amount of strain during the test must exceed the strain observed in tests of specimens with no prestrain. This performance is indicated by distance B in Figure 5.

The results on Ship Structure Committee Dn steel are shown on Figure 6. This steel exhibited the greatest improvement and was also the steel upon which most of the tests were performed. Two of the three other steels exhibited somewhat greater toughness at a temperature just below the transition temperature. The fourth steel did not show an improvement. It was decided that the prestrain method warranted further study.

Considering that the purpose of this symposium is to gather foundation data to support further elucidation of brittle fracture mechanics and that the ultimate hope is to formulate new concepts of structural design which embody the brittle fracture tendencies of materials, all potentially useful information should be assembled here in summary. For instance, it is probable that we shall

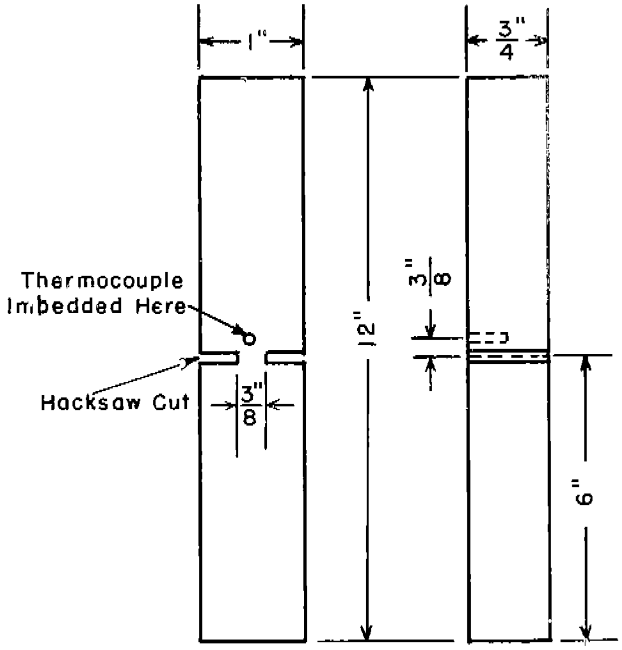


Fig. 4

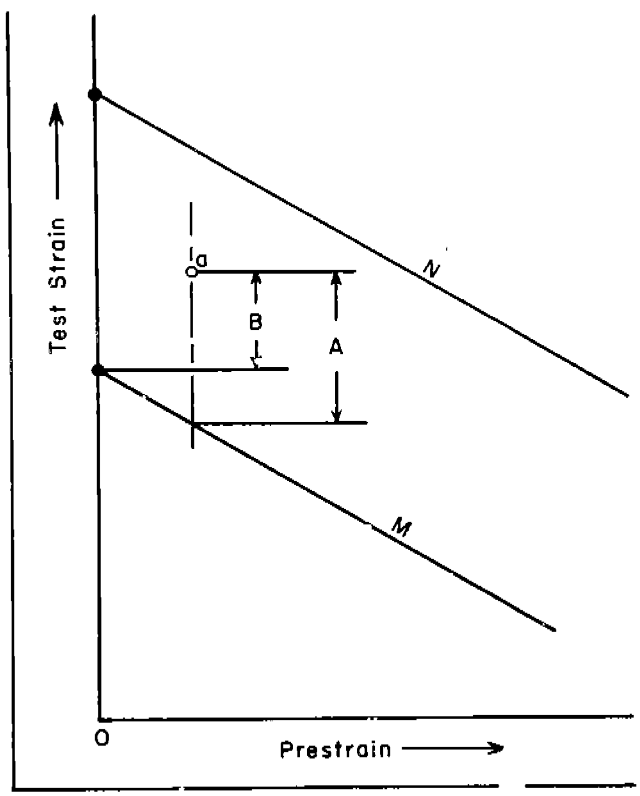


Fig. 5

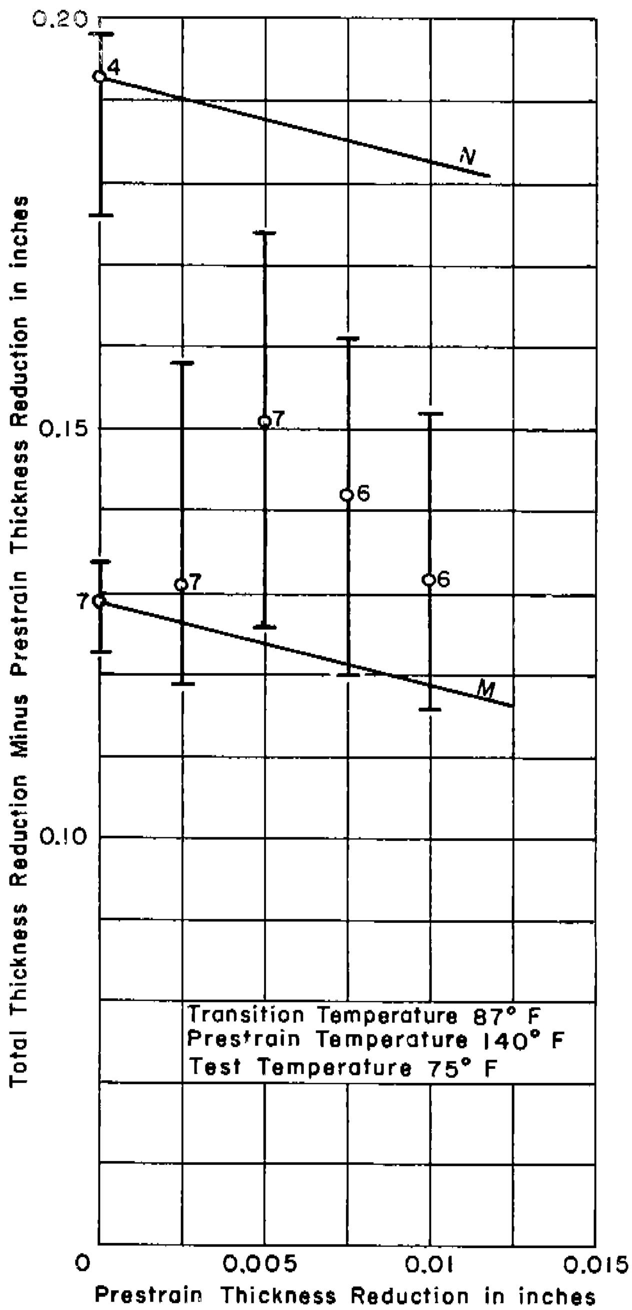


Fig. 6

be able to obtain a better understanding of this problem by further inquiry into the statistics of the Liberty ship fractures which occurred during World War II.

Initial steps in a study of the Liberty ship fractures are reported in reference 4. Figure 7 shows a very good correlation which was obtained between the actual reported service exposure of the Liberty ships to air temperature and a fitted mathematical curve. Figure 8 shows a mathematical deduction of the number of fractures which might have resulted had the fracture transition temperature distribution of the steel populations been shifted either way from whatever it was (zero on the abcissa) in the ships which were studied.

Figure 9 is an assemblage of all of the data which I have been able to bring to bear in an effort to estimate the tendency to fracture in terms of the two parameters, temperature and sea conditions. The trends presented on Figure 9 suffer from two shortcomings; first, the statistical data are not entirely homogeneous, and second, the presentation embodies implicitly the assumption that the independent variables (a) air temperature and (b) sea condition are not coupled with one another. This represents a rough first step only. It would be my earnest hope that we could at some time develop a more precise mathematical model of the fracture tendency as has been done for the still water condition in Figures 7 and 8.

I am sure that with the aid of results from active studies of ships and sea and with a more sophisticated treatment of the fracture statistics we shall obtain a powerful tool to make ship designs embodying the concept of brittle fracture. As an example of what I have in mind, I have prepared a fracture concept which could be useful to a designer. I should like to point out that no conclusions should be drawn from what follows as I intend it to be an example only. The basis of this concept is the hypothesis that an energy balance is a necessary (if not a sufficient) condition for fracture propagation and that the tendency for a failure to propagate in the form of a fast fracture varies directly as the potential elastic energy stored locally in the structure and inversely as the energy absorbed by the material during the fracture propagation. The reciprocal of the tendency to fracture was used for visual presentation because the resulting curves are similar in implication and appearance to the familiar notch-test temperature-transition curves.

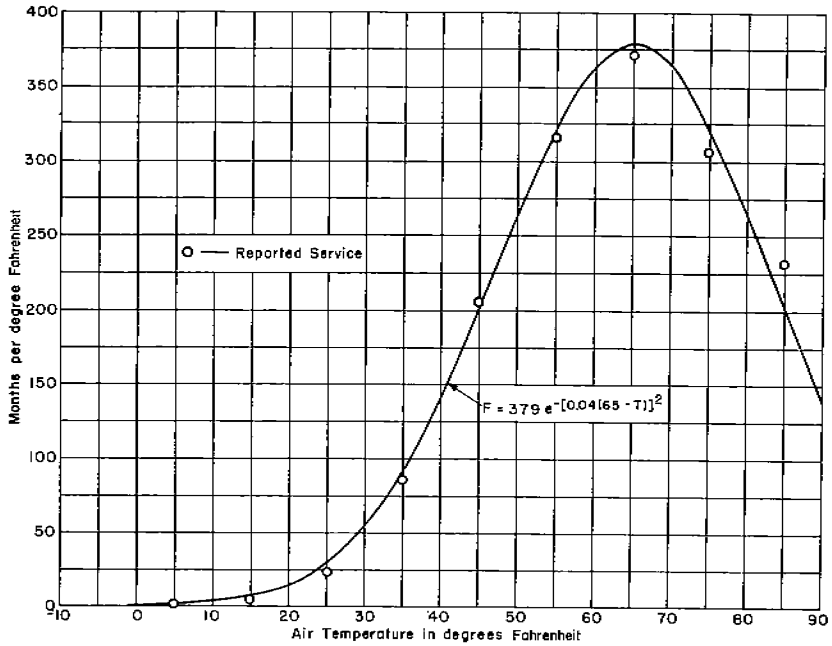


Fig. 7

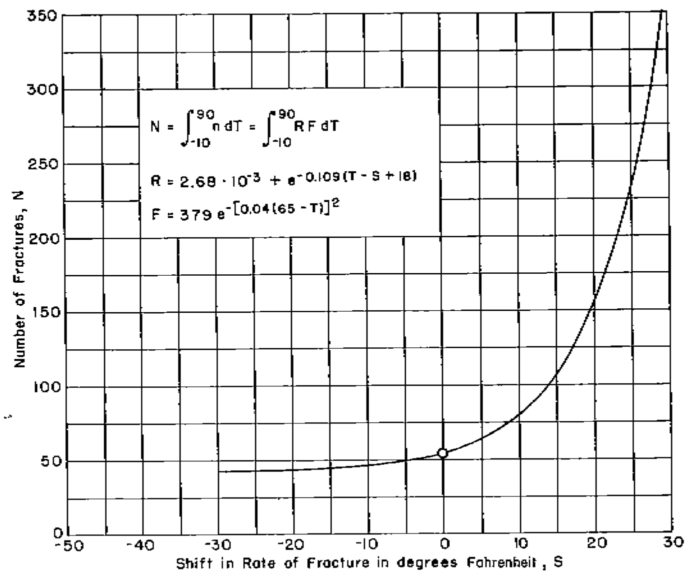
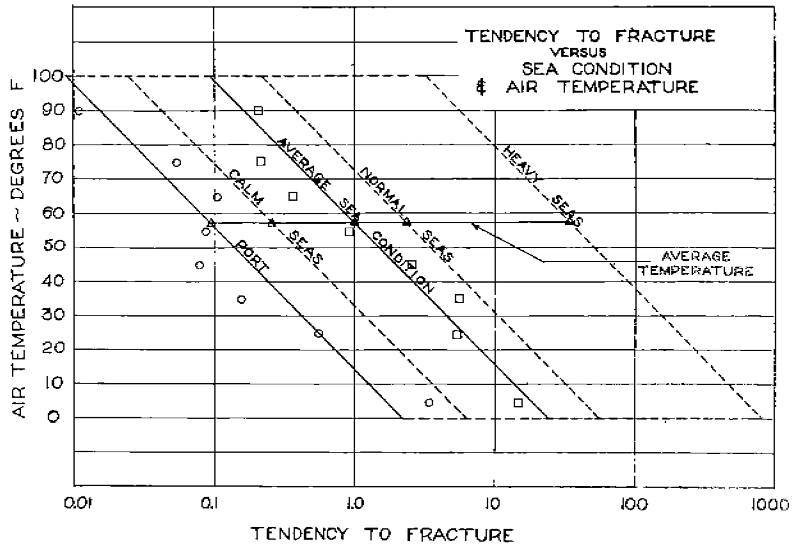


Fig. 8



LEGEND

(a) Tendency versus Temperature is based on 620 fractures of Class 1, 2 and 3 occuring before 1 Aug 45 on 667 EC2-S-Cls launched before 1 Feb 43.

(b)(1) Service time versus sea condition is based on 24,461 ship months accumulated before 1 Aug 45 on EC2-S-Cls built before 1 Feb 43.

(2) Tendency to Fracture versus sea condition is based on 664 potentially serious fractures (Class 1 and 2) on all EC2-S-Cls before 1 Apr 46. Average value = 664/76396 = .012 fractures per ship month.

(c) The EC2-S-C1 is the standard, dry cargo liberty ship. Class 1, 2 and 3 fractures are as defined in the final report of the "Board to Investigate the Design and Methods of Construction of Welded Steel Merchant Vessels."

Fig. 9

This can be expressed by $FF = \frac{U_a E}{\sigma^2 d}$ where

U_a = energy absorbed in propagating a fracture per unit of area of cross section of the structure fractured

E = modulus elasticity of the metal

σ = Stress

d = distance shock wave travels in the metal in a unit of time (time of 1 millisecond was used)

In reference 5 the authors point out that it is possible, with available test data, to obtain an idea of the energy absorbed in propagating a fracture in steel, U_a. On the basis of reference 6, a rough relation was established between U_a and the thickness reduction at the edge of a fracture in plate structures. Very roughly T. R. = U_a where T. R. is the thickness reduction at the edge of the steel plate in per cent and U_a is expressed as inch kips per square inch of fracture surface. From measurements made on actual service fractures, a thickness reduction frequency curve was obtained using data in reference 7. This showed that actual ship fractures are not brittle. Instead the thickness reductions ranged up to 4 per cent, and the greatest frequency was about 1 1/2 per cent.

Combining these results, it is seen that there is danger of fast fractures when U_a is less than 4 inch kips per square inch. For steel $E = 30 \times 10^6$ lb. per sq. in. and, $d = 197$ in. For the ships which fractured, the design stress averaged about 18,750 lb. per sq. in. Thus, there is danger of fracture when

$$FF = \frac{U_a E}{\sigma^2 d} < 2, \text{ and } \sigma \text{ is the design stress.}$$

Employing design stresses commensurate with conventional practice for various sizes of vessels and various types of steel, it was possible to devise Figure 10. This figure embodies the entire concept and points up the critical temperature for medium carbon steels.

Tests at the Naval Research Laboratory have demonstrated that aluminum alloys are capable of fast fractures but that the fracture surface is silky in appearance and never granular as in the fast fractures observed in steel. From a practical viewpoint the appearance of wreckage is unimportant. If a fast fracture can occur and if the energy balance is a controlling condition, then

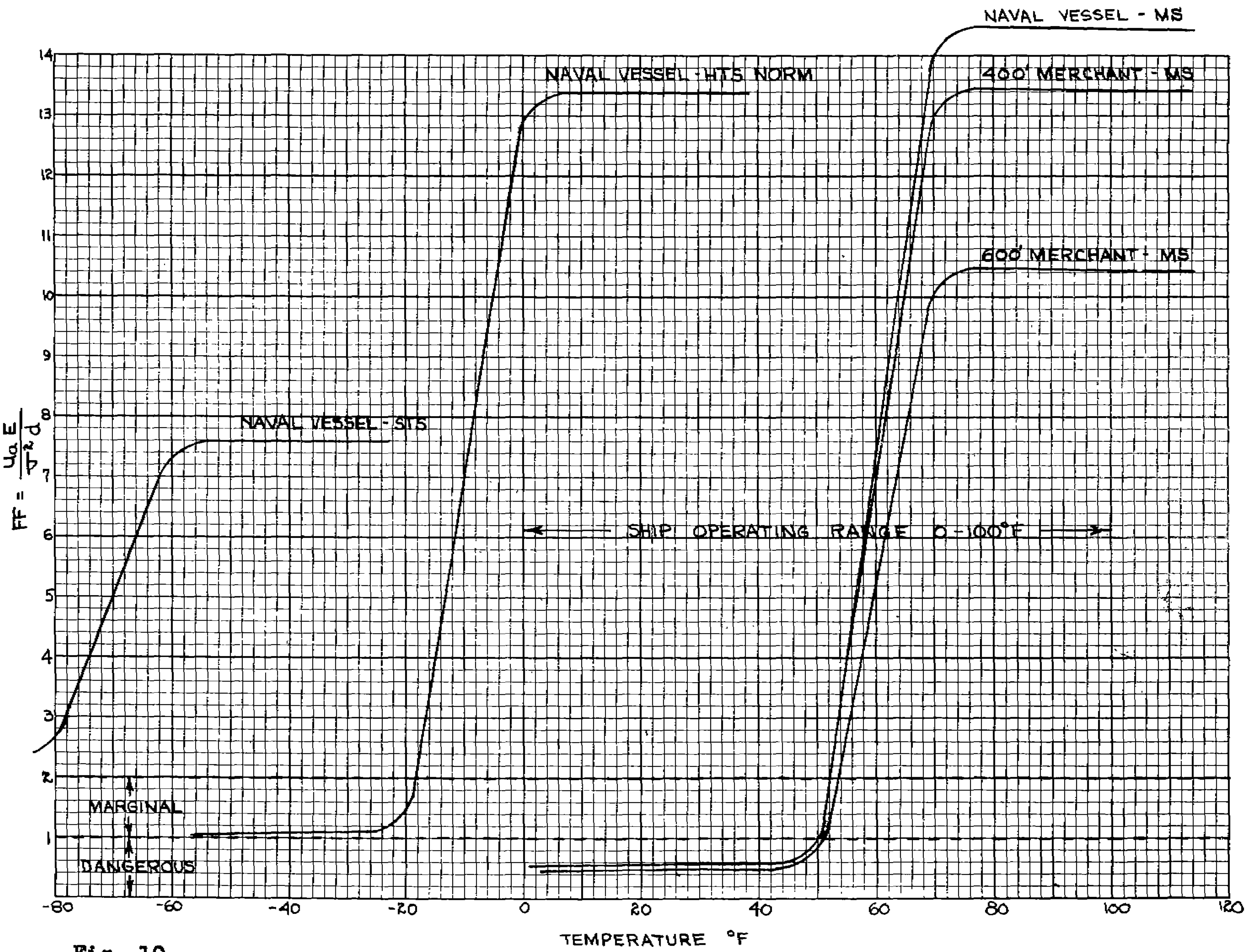

NAVAL VESSEL - MS
NAVAL VESSEL - HTS NORM
400' MERCHANT - MS
600' MERCHANT - MS
NAVAL VESSEL - STS
SHIP OPERATING RANGE 0-100°F
MARGINAL
DANGEROUS
FF = $\frac{u_a E}{\sigma^2 d}$
TEMPERATURE °F
14
13
12
11
10
9
8
7
6
5
4
3
2
1
-80
-60
-40
-20
0
20
40
60
80
100
120

Fig. 10

a look at aluminum is important. For typical design studies of aluminum ship structures, the following was found:

$$.6 < FF = \frac{U_a E}{\sigma^2 d} < 2.8.$$

This is not very comforting to a designer.

It is imperative that we determine whether there is any sense to such a fracture criterion. A sound and suitable tool for ship and other structural design must be found so that the concept of brittle fracture can be embodied into the design and in the characterization of metals to be employed in the structure. That is the purpose of the proposed fracture mechanics program.

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THE RELATION OF NOTCHED TENSILE TEST DATA TO PERFORMANCE IN SERVICE

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Nomenclature

Ductility: The local state of strain at the location and at the moment when a crack begins. Quantitatively, the local value of the largest principal strain. (Sometimes the reduction of the minimum cross section after separation of the specimen, and hence a measure of the average strain. This usage employs the term "reduction of area").

Tensile strength: In an unnotched tensile specimen, the maximum load divided by the initial cross section. Hence the average nominal stress at maximum load.

Uniaxial tension: Stress in only one direction. The other two principal stresses are zero.

Biaxiality: Two principal stresses are tensions and the third is zero. Quantitatively, the ratio of the smaller to the larger tension.

Triaxiality: All three principal stresses are tensile stresses. The quantitative measure which seems most pertinent to fracture is the ratio of the smallest to the largest tension.

b/h (breadth to height ratio): In unnotched bend specimens, the ratio of the transverse dimension to the radial dimension.

Notch section: The minimum cross section of a notch tensile test specimen.

Notch depth: Per cent cross section removed by the notch in machining the specimen.

Notch sharpness: Half the diameter of the notch section, divided by the root radius of the notch contour.

Notch strenghth: In a notched tensile specimen, the maximum load divided by the initial notch section. Hence the average nominal stress at maximum load.

Notch strength ratio: Notch strength divided by tensile strength.

Notch ductile behavior: A behavior observed in the more ductile metals, where the notch strength is greater than the tensile strength by a percentage that is approximately equal to the notch depth. Thus, notch strength ratio approximately equals unity plus the fraction of the cross section removed by the notch.

Notch brittle: A behavior characterized by a notch strength ratio visibly lower than that characterizing notch ductile behavior.

Disk strength ratio: Average tangential stress in a disk at bursting divided by tensile strength.

Abstract

First, there is a brief review of the information which currently can be obtained from notched tensile tests. Then there is a discussion of the fact that the applicability of this information to manufacturing problems is so limited as to be almost useless for engineering purposes. Finally, there is a consideration of what information is necessary for rational design and how such information might be obtained. For the purpose of being more specific about what information is needed for a rational design procedure, there is included a detailed discussion of several experimental results that raised fundamental questions. Some of these results were obtained from a wide variety of hitherto unpublished tests on a special steel. These tests include notched and unnotched tensile tests, notched and unnotched bend tests, notched and unnotched disk bursting tests, and Charpy impact tests.

Introduction

In notch tensile tests, as in any test to fracture, the results of the test are governed by two phenomena: the initiation of a crack, and separation of the specimen due to the propagation of that crack. The first of these phenomena is much better understood than the second; and most of the features of notch tensile testing which make the results intelligible are connected with the "ductility," or the local strain at crack initiation. The nature of the crack propagation, on the other hand, is almost completely obscure; and consequently, although crack propagation is exceedingly important in many applications, such as the failure of welded ships, this aspect of fracturing will not be treated in the following.

There has been very little change in the state of knowledge of notch tensile behavior since last reviewed(1), regarding either new experimental results or general concepts. It will not be the purpose of the present paper to summarize again the scientific status of the subject; and consequently, only those few new experimental facts subsequent to the earlier survey will be discussed. However, an effort will be made to review the state of affairs of notch tensile testing from an engineering viewpoint, inquiring in particular as to what degree of applicability the current scientific knowledge has to engineering problems. The unanswered question of how to design a part in such a way as to avoid fracturing in service is one of the most important engineering problems of today.

Recent Developments in Notch Tensile Testing

The principal scientific advance in the field of notch tensile testing since 1947 is due to the work of Sachs and Fried(2) on the local strains at the root of the notch and the strain distribution over the notch section (the minimum section perpendicular to the axis). This work of itself does not have direct engineering usefulness, but together with other information as yet not obtained, it should result in a considerable advance in the field of notch tensile behavior. How these data fit into the general scheme of things will be discussed later.

Most of the recent notch investigations have dealt with behavior at elevated temperature(3), where the results are very similar to those at room temperature in many respects.

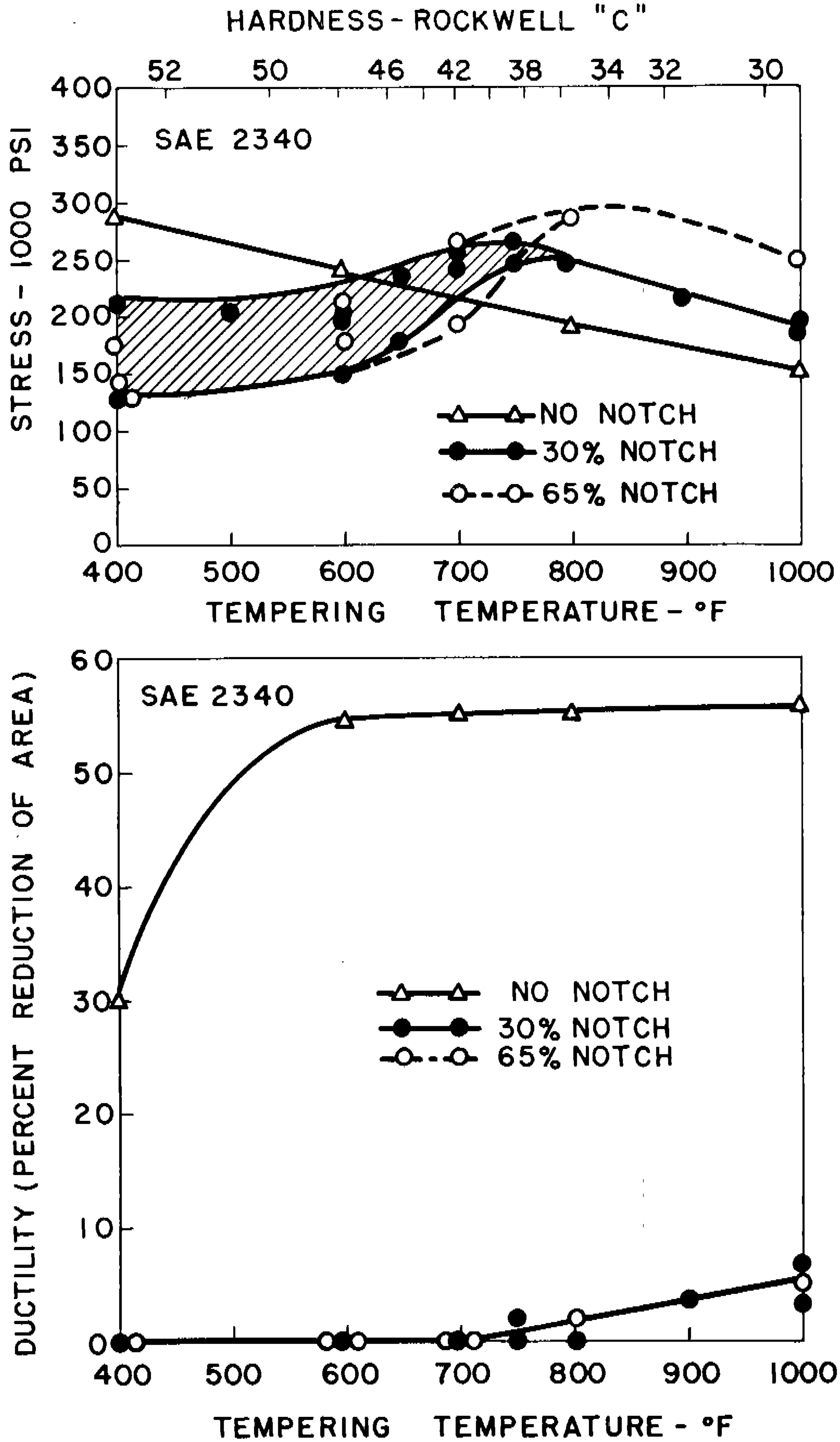

Fig. 1. Effect of tempering temperature on the notched and unnotched tensile properties of SAE2340 Steel. From Sachs and Lubahn(7).

The tests of Davis and Manjoine(4) show that the notch strength ratio (for a given rupture time in this case) varies with notch radius in the same way as observed previously for room temperature behavior(6), namely, that for sufficiently ductile metals the notch strength ratio increases from unity at infinite radius (unnotched specimen), and levels off for sufficiently sharp notches at a value that depends on the notch depth:

$$\text{Notch Strength Ratio} = 1 + \text{Notch Depth}/100 \qquad [1]$$

while notch brittle metals exhibit low notch strength ratio values for sharp notches.

In the tests of Davis and Manjoine, it appeared that the notch strength ratio was the above function of the notch depth for any time to rupture. However, these data did not cover a very wide range of rupture times. Tests by Brown et al.(5), covering a much wider range of rupture times, for a particular notch depth, show that the notch strength ratio may pass through a minimum with increasing rupture times. The lower the testing temperature, the more pronounced is the minimum and the longer is the rupture time at which it occurs. These tests thus serve to show the importance of the new variable--time.

Current Applicability of Notch Test Results to Engineering Problems

The principal current value of notch tensile testing for engineering purposes arises from the ability of these tests to distinguish between "notch ductile" and "notch brittle" behavior in notched members carrying tensile load. The earlier work at room temperature(7) showed that the notch strength of heat-treated low alloy steels is very high and a constant percentage (conforming to Equation [1]) higher than the tensile strength, providing that the hardness is less than a certain boundary value, Figure 1. This behavior is called "notch ductile" behavior. It occurs when the ductility under conditions of notching is large enough to permit a smoothing out of the initial elastic stress concentration, and thus cause a rather uniform stress distribution. As the hardness increases above this boundary value, the notch strength falls off rather abruptly to the tensile strength and thence to still lower values, which scatter widely. This behavior is called "notch brittle" behavior. In the example of Figure 1 the boundary line between notch ductile and notch brittle behavior is perhaps 36 Rockwell "C" for the 30 per cent notch and perhaps 34 Rockwell "C" for the

65 per cent notch. Presumably for a still deeper notch the boundary hardness would be somewhat lower still. Thus, the metal is notch ductile below about 35 Rockwell "C" and notch brittle at higher hardness. One could design the material to fit the application by selecting the greatest hardness within the notch ductile region as being the most suitable hardness for an application involving notches. Unfortunately, the converse problem is much more difficult--that of designing a suitable geometry for a given material which is somewhat notch brittle. This problem will be discussed in considerably more detail later.

The above design procedure based on distinguishing between notch brittle and notch ductile materials is beginning to come into use in high temperature applications. However, for some reason that is not readily apparent, it has become customary in many cases to take the condition where the notch strength ratio is unity as being the boundary line between notch brittle and notch ductile behavior(3). According to the concept implied in Figure 1, the use of this condition is much less safe than that criterion discussed in connection with Figure 1. The curve of notch strength ratio vs. hardness is quite steep where the notch strength ratio is unity; and consequently, any unforeseen circumstances that would tend to shift the notch strength curve to the right (increase the tendency toward notch brittleness) would cause a considerable loss of the notch strength ratio. On the other hand, the notch strength ratio vs. hardness curve is not steep where the notch strength ratio is very high. Consequently, a small change in the tendency towards notch brittleness would not cause any appreciable change in the notch strength ratio near this point.

The above design procedure represents a definite, though small, improvement over older procedures. Prior to about fifteen years ago, notched parts were usually designed exclusively on the basis of classical elasticity. This procedure is suitable for completely brittle materials for which the fracture stress is known. The same method has been widely used also for metals with some ductility, using the yield strength instead of the fracture strength as the basis for limiting the stress. Such a procedure restricts the plastic flow to the small value of offset strain used in determining yield strength (usually 0.2 per cent). Where stress concentrations are present, even this small plastic strain is present only in the local regions of greatest shear stress, while the rest of the metal volume often is stressed far below its yield strength.

When the ductility is negligible, designing by elastic theory is very nearly correct. When there is a small amount of notch ductility so that some redistribution of the stress is permitted by this small plastic flow, designing by elastic theory is somewhat over-conservative, depending upon the increase in average stress permitted by the plastic deformation. When the metal is completely notch ductile, designing by elastic theory is very over-conservative; the working stress obtained by this procedure would be smaller than a suitable value of working stress by a factor which is equal to the stress concentration factor.

Thus we see that designing so that the average stress equals the yield strength for notch ductile metals remedies the most flagrant over-conservatism caused by designing against any plastic flow, even locally. However, a design procedure is still lacking for metals that have a little ductility but not enough to cause notch ductile behavior. The problem of developing a design procedure for these materials will be discussed in the next section.

The importance of designing for reasonable amounts of local plastic flow, rather than for elastic behavior only, arises primarily from the use of metals over long periods of time at elevated temperatures, where creep occurs. Under these conditions the stress to avoid any plastic flow is so small that the load-carrying capacity would be impractically low. Furthermore, it is permissible, for many high temperature applications, to allow some plastic flow, particularly when it is so localized that the overall deflections are small.

Requirements of a Suitable Design Procedure--Example: Elastic Design

A suitable design procedure to avoid fracturing must somehow relate the conditions of the engineering applications to the results of simple tests on the metal being considered for the application. It is imperative that the pertinent engineering conditions be determinable by calculation or measurement, and that the laboratory test be so devised that the test results can be correlated with performance under service conditions.

Consider an elastic structure, as an example, for which a suitable design procedure has long been in wide use. Theory of elasticity calculations(8) or photoelasticity(9) measurements permit the determination of the stress and strain distribution in the structure. For these operations, the necessary material constants (modulus of elasticity and Poisson's ratio) can be determined from a simple tensile test. In this case, the stress-limiting aspect of metal behavior which must be determined by test is the stress state for initiation of plastic flow. This metal characteristic is completely described by the yield strength, as determined by a simple tensile test, and the shear stress law or the energy-of-distortion condition of plastic flow(10).

Now let us consider how far present knowledge will permit the development of an analogous design procedure against fracturing of notch brittle metals and what features of such a procedure are still lacking.

The Importance of Stress Calculations to Engineering Design

A basic feature of the current design procedure against plastic flow is the use of elastic theory or photoelasticity to determine stress distribution. In principle, the stress and strain distributions in the plastic range can also be calculated using a combination of the theory of elasticity and classical plasticity(10, 11). Actually, however, this procedure is less than satisfactory in three respects. First, in nearly all the plasticity solutions in the literature, the elastic strains are neglected compared to the plastic strains; whereas in most engineering applications, the important problem is that in which the plastic strains are comparable with the elastic strains. When the plastic strains are quite small, the problem is much more difficult mathematically. Furthermore, experiments are not available to demonstrate the validity of the principles that one would presume in such a problem--namely, that Hooke's law applies to the elastic part of the strain, classical plasticity to the plastic part, and the compatibility condition(8) to the simple sum of the elastic and plastic parts. A second uncertainty in applying classical plasticity to manufactured parts arises when the temperature is high enough to allow creep. There is very little experimental evidence to show that the so-called "laws of plasticity" apply to creep. In particular, when most of the creep is anelastic(12), the influence of stress state might be entirely different than for plastic strain. Third, classical plasticity is still on a very shaky experimental foundation if the principal stress ratios vary during the deformation and especially if partial unloading occurs.

The Importance of Laboratory Test Results in Engineering Design

In designing against plastic flow, the test results which are necessary are the yield strength, the modulus of elasticity, and Poisson's ratio. The metal characteristic or characteristics which must be measured in laboratory tests in order to design against fracturing are not known. Many types of tests are currently made for this purpose, ranging from the simple tensile test to various simulated service tests. The simpler, "standard" laboratory tests, such as the tensile test and Charpy impact test, often do not contain the factor which governs failure under service conditions; and even when they do, the quantity which is pertinent to the fracturing is sometimes not measured. For example, the Charpy impact test imposes conditions of notching which may be similar to those in many engineering applications; but it is not customary to measure maximum bending load, bend angle at crack initiation, or other quantities which may be the governing features of the engineering application, rather than total energy absorbed. In simulated service tests, the pertinent factors are more likely to be present; but the complexity of such tests often makes them too expensive to be feasible.

The Problem of Developing a Design Procedure against Fracturing

The inability, described above, to relate laboratory test results to engineering performance is the most important stumbling block in developing a rational design procedure against fracturing in notch brittle materials. Currently, we do not even understand the relationships between various simple tests commonly performed in the laboratory; certainly, until we do, there is little hope of seeing the relationship between one or more simple laboratory tests and the performance of the metal under the complex conditions that prevail in most engineering applications. Obviously, the first step in developing a suitable design procedure must consist of a detailed investigation of the metal characteristics under a wide variety of laboratory tests. Gradually the relationship between different kinds of tests will be understood well enough to permit the prediction of one test result from the results of another kind of test. When this type of relationship has been established among tests that represent a sufficiently large number of factors that govern fracturing, it will be possible to predict engineering performance, which may be viewed merely as a more complex test.

Most of the rest of the discussion will be devoted to this subject of how to relate the results of different kinds of tests to each other. In doing so, however, one must be careful not to lose sight of the ultimate objective. It is altogether too easy to become so absorbed in discovering the relationships between two particular kinds of tests that one disregards other factors which may be more pertinent to engineering performance but which are not represented in the tests currently being studied. In the examples which are given below, there is a discussion of some of the factors which are currently considered to be important. Other factors, such as size and repeated loading, may prove to be just as important if not more so.

<u>The Degree to which Various Tests Can Be Correlated Currently</u>

Let us first consider one of the very few cases where there is some evidence that the relation between two tests is understood, namely, the unnotched bend test and the unnotched tensile test. The data(13) show that the biaxiality (transverse tension over tangential tension) at mid-width of a bend specimen varies with breadth-to-depth ratio (b/h) from 0.5 for a very wide specimen to zero for a very narrow one. Fracture begins at the mid-width for b/h ratios greater than unity, and the ductility decreases as the biaxiality increases. For b/h equal to or less than unity, fracture begins at the corners, where the stress is uniaxial (transverse stresses zero), and the data indicate that the ductility for corner fracture is the same as that in a tensile test. Thus, it appears that ductility is independent of stress gradient for uniaxial tension and that narrow bend tests can be related to tensile tests.

When notches are present, the situation is much more confused. There is no type of notch test whose result can be predicted consistently from the result of any other type of test, with or without a notch, although correlation does appear possible in certain cases where the ductility is exceedingly high. In the investigations of welded ship failures, there have been numerous comparisons between such tests as the Navy tear test, the internally notched wide plate, the keyhole Charpy test, and the V-notch Charpy test. So far, these comparisons have failed to reveal even what factors cause the differences among the different tests, much less the detailed way in which these factors control the test result.

Notch tensile investigations have been pursued to the point where several interesting speculations can be made. For example Schwartzbart and Brown(15) and Lubahn(1) have discussed the concept that fracture begins at the root of the notch in sharply notched specimens, where the very high local strain has more effect than the ductility-killing triaxiality* below the surface, whereas in mildly notched specimens, the crack starts below the root of the notch because the loss of ductility due to the triaxiality there is a more important factor than the small strain concentration that prevails at the root of a mild notch. So far, there have been few efforts to verify this concept. Sachs and Fried(2) showed that mild notches cause internal fracture, although their experiments failed to prove conclusively that surface fracture occurs for sharp notches in the same materials. However, the gradual development of surface fracture can be observed with the naked eye in sharply notched mild steel specimens. The above observations bear out the qualitative features of the concept, but efforts(16) to establish quantitatively a boundary value of notch sharpness between surface fracture and internal fracture seemed inconclusive(1), probably because the two surface conditions were not different enough. A more conclusive test would be to compare the local surface ductilities of various sharply notched specimens with the ductility of unnotched bend specimens designed to give the same biaxiality as the notched tensile specimens, using the technique of Sachs and Fried(2) for measuring local strain and biaxiality at the root of a notch. It would be particularly fruitful to make this comparison on SAE 2340 for which it was found(7) that the unnotched tensile ductility was practically independent of hardness below 48 Rockwell "C", but that the notch behavior varied within wide limits in the same hardness range, Figure 1, the harder steels being much more notch brittle than the softer ones. According to the above concept, these differences between notched and unnotched behavior are caused by susceptibility of the steel to either biaxiality or triaxiality, depending on where the fracture begins. If fracture

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*In general, the term "triaxiality" indicates that all three principal stresses are tensile stresses. Since brittleness is lack of plastic flow and since plastic flow depends(14) on the difference between the largest and smallest (algebraically) principal stresses, it will be suitable for the present purposes to define triaxiality quantitatively as the ratio of the smallest to largest principal stress.

begins at the surface, where biaxiality controls fracture, one should find the ductility in an unnotched bend test of suitable b/h ratio to be the same as in a notched tensile test and that ductility in unnotched bend tests should fall off with biaxiality much faster for the harder steels than for the softer ones. Such tests have not yet been made. The only reliable* investigations of the effect of biaxiality on ductility were made on the aluminum alloys(13). If triaxiality governs the notch tensile test result, on the other hand, causing fracture to begin below the root of the notch, one should find a considerably lower surface strain at the root of the notch in the tensile specimen than in the unnotched bend specimen of the same biaxiality. In sharply notched tensile specimens there is a very high strain gradient at the root of the notch: the strain at a point a little below the root of the notch is much smaller than the surface strain. If fracture begins at this small strain because of triaxiality, instead of at the much larger surface strain, this result would constitute the long-awaited experimental verification of the often-quoted supposition that triaxiality is a very potent embrittling factor. This supposition has been founded primarily on the plastic flow laws, which state that any metal, however ductile in a tensile test, would be completely brittle at a triaxiality of unity ($S_1 = S_2 = S_3$), where the absence of shear stress prevents any plastic flow at all. Although this may be true, the supposedly great degree of embrittlement for readily attainable triaxialities has been largely a matter of conjecture**.

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*Biaxiality results using bulged sheet or thin-walled tubes are usually confused by the fact that local necking rather than the applied load governs the biaxiality at fracture. Furthermore, the average strain which is usually reported as being the ductility is less than the local strain at the neck, which <u>is</u> the ductility.

**In 24ST aluminum alloy(16) the ductility was not less than half the uniaxial value even at a triaxiality greater than 0.4. In quite hard steels(20)--240,000 psi tensile strength--the reduction of area at fracture was only a few per cent for a rather sharp notch, but strain distribution measurements(2) show that the local surface strain may be more than ten times the reduction of area value for such a sharp notch. Thus, much of the appearance of "brittleness" for sharp notches might be due to high strain gradient of sharp notches, rather than a true effect of triaxiality on ductility.

Recent Notch Tests on a Special Steel

The difficulty of relating different kinds of tests, on the basis of current knowledge, can be illustrated by a few more examples taken from recent tests by the authors. These tests were made on a special alloy steel in which a wide variety of properties (as determined by standard tests) could be obtained by various heat treatments. The details of the tests are given in the Appendix, and the results are summarized in Table I. The conditions of heat treatment designated by A and B in Table I are particularly interesting because they have the same very high value of unnotched tensile ductility but tremendously different values of keyhole Charpy impact energy. In addition to the standard types of tests, five kinds of special tests were made for the A and B heat treatments.

The logical first supposition as to the different impact test results for the two heat treatments would be that the local strain at fracture was the same for both heat treatments because the tensile ductilities were equal but that the energy to propagate the crack was very large for heat treatment A as compared with practically none for heat treatment B. This supposition stems from the Navy tear test results of Kahn and Imbembo(17). Their tests showed that the energy to start the crack is practically independent of the test temperature. The energy to propagate the crack, however, varied from essentially zero below the transition temperature to a large value above the transition temperature. Also, one V-notch Charpy bend test(18) indicated that the energy to propagate the crack may be at least five times that required to initiate fracture, even though the local strain before crack initiation was very high (40 per cent), Figure 2.

Thus, previous tests on mild steel would indicate that the difference in impact values for conditions A and B in Table I would be due to differences in the mode of crack propagation. However, the results of slow notch bend tests for conditions A and B (Table I) prove otherwise. The difference in energy absorption is due almost entirely to an enormous difference in local ductility at the root of the notch-- a difference which is not indicated at all by the unnotched tensile ductility. In both specimens the crack propagated suddenly and with comparatively little energy, Figure 3. From these data it becomes evident that impact energy is not a criterion of any single metal characteristics; for one metal the level of impact energy may measure the ductility; in another metal the impact energy may be governed almost entirely

TABLE I

Material and Results

Composition: 0.12%C, 0.48%Mn, 5.87%Cr, 0.95%Mo, 0.37%Ti, 0.03%B

	A	B	C
Austenitizing Temperature, °F	1900	1800	2100
Quenching Medium	oil	oil	oil
Tempering Temperature, °F	1300	1200	1200
Tempering Time, hours	6	4	2
0.02% Yield Strength, 1000 psi	43--56	67--78	93--101
Tensile Strength, 1000 psi	75--83	90--102	145--149
Reduction in Area, %	77--79	73--76	1--3*
Keyhole Charpy Value, ft-lb	36--52†	2--4	1--2
Disk Str. Ratio = $\frac{\text{Av. Tang. Stress}}{\text{Tensile Strength}}$			
With no stress raiser	1.01	0.99	0.85
With crack in rim		0.76	
With holes and saw cut near rim	1.00		
With saw cut in rim	0.86	0.57	
Notch Tens. Tests: 50%, 60° V-notches; r/D = 0.02			
Notch Str. Ratio = $\frac{\text{Notch Str.}}{\text{Tens. Str.}}$	1.55--1.69	1.49--1.57	0.84--0.90
Notch Ductility, % red. in area	30--53	0.6--1.2	1.0--1.3
Unnotched Wide-Plate Bending Strain, %	>120	>120	
Notched (V-Notch Charpy) Slow Bend Ductility, %	165--180	3--12	
Maximum Bending Load--lbs.	2220--2440	1740--1980	
Energy Absorbed in Slow, Notch Bend Test, ft-lb	38.3**	1.2**	

* Two out of six tests gave 59 and 63 per cent.

** Of the 38.3 ft-lb, 9.4 ft-lb was expended in propagating the crack, while the crack propagation was completely brittle for heat treatment B.

† One out of five tests gave only 10 ft-lb.

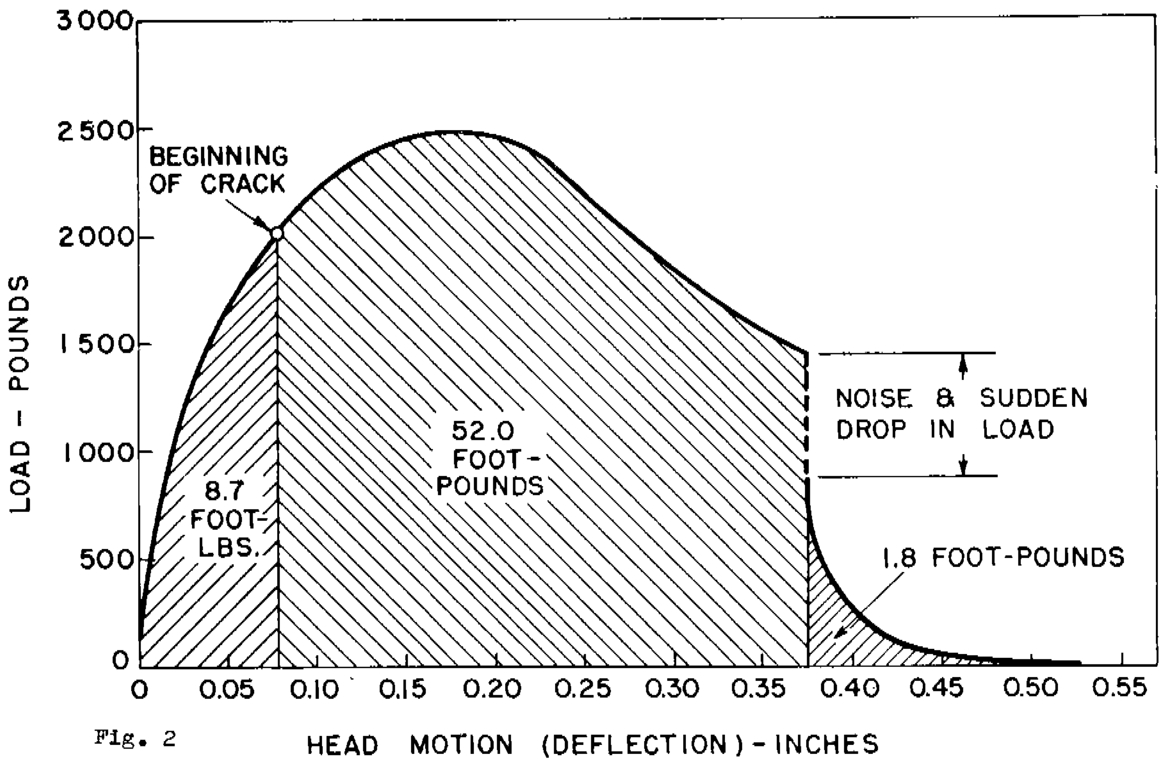
3000
2500
2000
1500
1000
500
0
LOAD - POUNDS
BEGINNING
OF CRACK
8.7
FOOT-
LBS.
52.0
FOOT-
POUNDS
NOISE & SUDDEN
DROP IN LOAD
1.8 FOOT-POUNDS
0 0.05 0.10 0.15 0.20 0.25 0.30 0.35 0.40 0.45 0.50 0.55
HEAD MOTION (DEFLECTION) - INCHES

Fig. 2

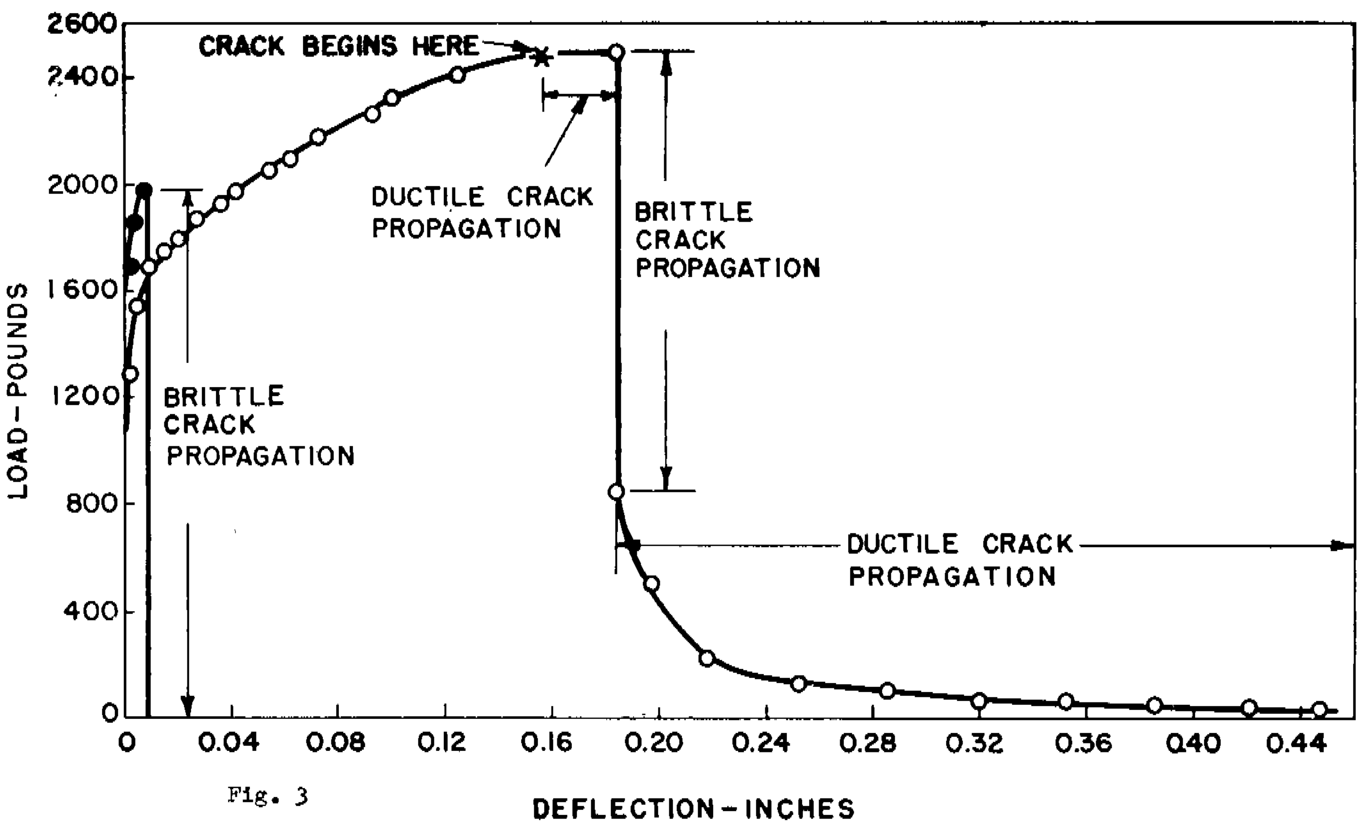
2600
2400
2000
1600
1200
800
400
0
LOAD-POUNDS
CRACK BEGINS HERE
DUCTILE CRACK
PROPAGATION
BRITTLE
CRACK
PROPAGATION
BRITTLE
CRACK
PROPAGATION
DUCTILE CRACK
PROPAGATION
0 0.04 0.08 0.12 0.16 0.20 0.24 0.28 0.32 0.36 0.40 0.44
DEFLECTION - INCHES

Fig. 3

by the mode of crack propagation, while the energy to initiate the crack, even for high ductility, may have been small.

These same slow notched bend tests show another rather disconcerting fact. If the high strain gradient below the notch root of a rather sharply notched tensile specimen causes surface fracture in spite of the lower ductility inside due to triaxiality, then one would certainly expect surface fracture in a notch bend specimen where the equally high notch-induced strain gradient is superimposed on a strain gradient due to the bending and having the same sign. If fracture begins at the surface in notch bending, the ductility should depend only on the biaxiality in that surface. This ductility should be the same as in a very wide unnotched bend test, where the biaxiality is the same as in the notch bend specimen. For example, some preliminary tests on mild steel indicate that the notch bend ductility equals the unnotched bend ductility, provided that the specimen is not oriented in such a way that the fracture anisotropy causes an abnormal location of crack initiation, Figure 4. This same equivalence may prevail for heat treatment A, where the notch bend ductility is even greater than the strain in an unnotched specimen that has been folded flat shut (about 120 per cent). In contrast, the bend ductility values for heat treatment B in Table I show that the local surface strain at fracture of the notch bend bar is many times less than the unnotched bend ductility. Thus, fracture apparently did begin below the root of the notch because of high susceptibility of the metal to triaxiality.

The ductility values in <u>notch tensile tests</u> also suggest that fracture begins below the surface for heat treatment B, while surface fracture may have occurred for heat treatment A, (see Table I). For this type of test, however, the state of affairs is not as obvious as for notch bending; there is not yet a feasible method of applying the technique of Sachs and Fried(2) to the measurement of the local surface strains in these small specimens. The reduction of area value can be related either to the local tangential strain at the notch root:

$$1 - \frac{\%\ \text{red. of area}}{100} = 1 + \frac{\%\ \text{tangential strain}}{100}$$

or to the average axial strain:

$$1 - \frac{\%\ \text{red. of area}}{100} = \frac{1}{1 + \frac{\%\ \text{axial strain}}{100}}$$

but not to the local axial strain. One can easily imagine that the local axial strain was large, even for a small reduction of area because of a high strain gradient. It is equally

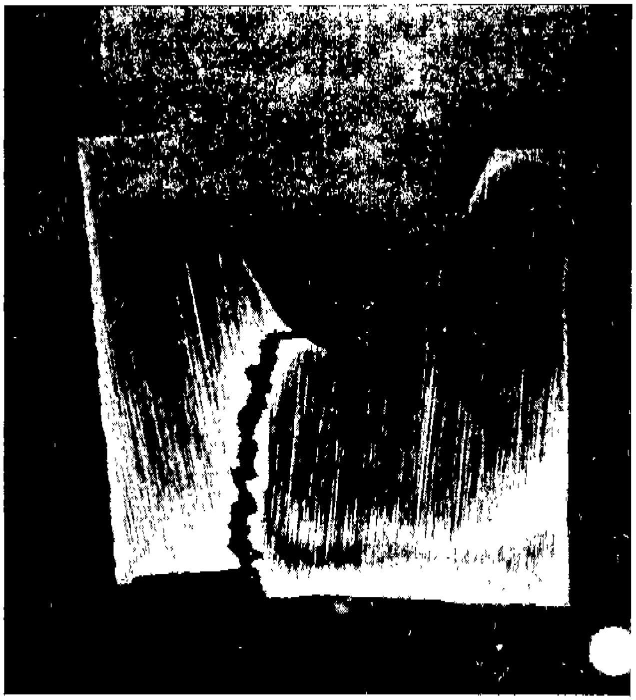

Fig. 4. Section through a portion of a mild-steel notch bend specimen cut from four-inch plate. Specimen was geometrically similar to a V-notch Charpy specimen, being 3.4 in. square and having a 0.091 in. notch radius. From Lequear and Lubahn(18).

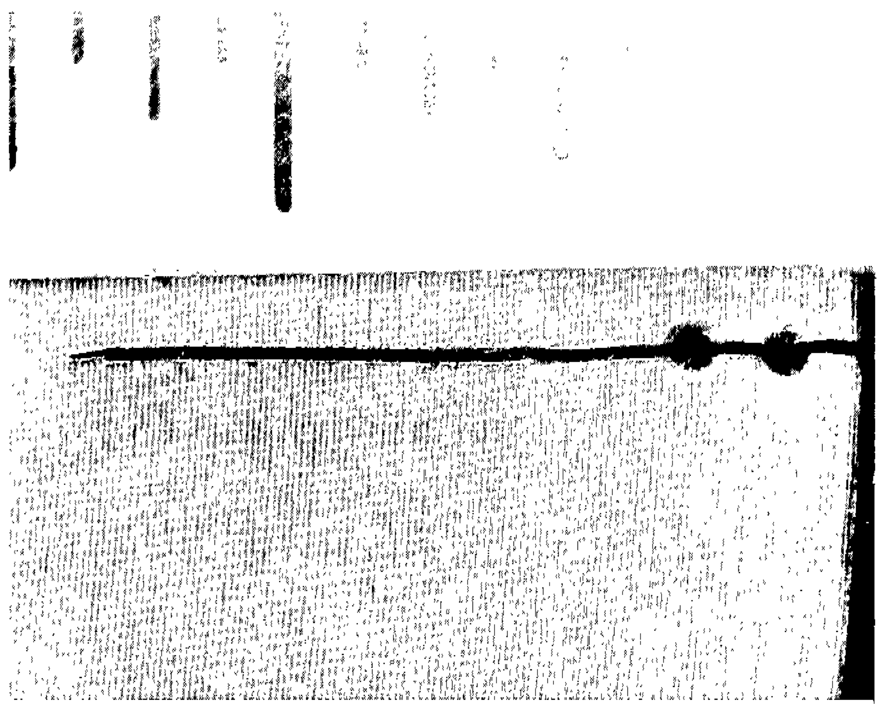

Fig. 5. Saw cut notch in rim of disk before spinning. (Heat Treatment A).

easy to imagine a high value of local axial strain coexisting with a small value of local tangential strain because of a biaxiality only slightly less than 0.5. On the other hand, a large reduction of area value does not necessarily mean a high value of local axial strain. Because final diameter is usually measured after complete separation, the reduction of area value might be fictitiously high if ductile crack propagation has reduced the notch diameter while the crack is propagating. This phenomenon has been called the "rim effect"(19).

Regardless of the uncertainties of interpreting the reduction of area values in terms of local axial strain, it is still clear from the high notch strength ratio that even for heat treatment B the local ductility is high enough to cause a very uniform stress distribution before fracture. For very ductile metals the stress apparently becomes quite uniformly distributed over the notch section by the time that maximum load occurs. Since maximum load occurs at about the same average strain in notched and unnotched specimens(15), the ratio of the nominal stresses at the maximum load for the notched and unnotched specimens, respectively, is a rough measure of the triaxiality present(16). For very ductile metals, it has been found that this notch strength ratio depends only on the notch depth, according to Equation [1], provided the notch sharpness is less than about 6. The amount by which a notch strength ratio value falls below this ideal value is a measure of the degree to which the stresses have failed to be uniformly redistributed before failure occurred(6). For the fifty per cent notches discussed in Table I, the notch strength ratio values are at least as high as the "ideal" value of 1.5 for very notch ductile metals. Thus, one must conclude either (a) that the low notch ductility indicated by the notch bend tests for heat treatment B is still enough to smooth out the initial elastic stress concentration in a notch tensile test, or (b) that a less severe state of stress results in a higher ductility in a notch tensile test than in a notch bend test.

The <u>bursting tests on unnotched disks</u> (Table I) show about what one would expect. For heat treatments A and B, where the unnotched tensile ductility is very high, the disk strength ratio is untiy. This value means that the initially non-uniform (elastic) stresses have been sufficiently uniformly redistributed by large amounts of plastic flow that the average tangential stress at bursting of the disk equals the average stress in a tensile specimen at maximum load. This condition would be expected if the necking strain for a disk were about the same as that in a tensile specimen. Indeed, this appears to be the case from measurements on both types of specimens after fracture at locations away from the necked regions (see Appendix). Thus, the unnotched disk tests conform to current concepts, at least qualitatively.

For <u>notched disks</u>, or disks containing a stress raiser of any kind, the test results are not as easy to understand. For this type of test, the criterion of the load-carrying capacity is the "disk strength ratio." For metals with considerable ductility, the disk strength ratio is observed to be in the neighborhood of unity for the design of disks being used here. If the maximum load strain is about the same in a notched disk as in an unnotched tensile test, as is the case for notched tensile specimens, a disk strength ratio of unity is to be expected for these thin disks, since the triaxiality and therefore the increased flow stress would be expected to be small.

The disk strength ratios for notched disks of heat treatment A are considerably higher than those for heat treatment B, and in some cases near unity. These data tend to indicate two things: (1) that the B heat treatment material is the more susceptible to notch testing, and (2) that the notched bursting test is more severe than the notch tensile test. The fact that the disk strength ratio values are considerably smaller for heat treatment B than A is not surprising, considering that the ductility in both notch bend and notch tensile tests is smaller for heat treatment B. The thing that is surprising, however, is that the disk strength ratio is less than the ideal value of unity for both heat treatments, while the notch strength ratio assumes the ideal value of 1.5 or more in the notch tensile tests for both heat treatments. Thus the disks suffer more loss of load-carrying capacity because of notching than do notch tensile specimens. The first reason that comes to mind for the low values of disk strength ratio is that the disk contains a sharper notch than the tensile specimen and that a given notch ductility can smooth out a mild stress concentration but not a severe one.

One might at first doubt the above speculation that a sharper notch reduces the load-carrying capacity more than a mild notch, because of the comparison of load-carrying capacity of notch tensile and notch bending tests, where the notch sharpnesses are equal. In the notch tensile tests the stress redistribution causes a high load-carrying capacity for both heat treatments. For the notch bend tests the lower maximum bending load for a higher tensile strength in the case of heat treatment B would seem to indicate that a less favorable stress redistribution has taken place in this specimen than for heat treatment A. However, the bending load is not a valid criterion of the completeness of stress redistribution. It would be a valid measure of this characteristic only if the strain hardening at maximum load were equal to that in an unnotched tensile specimen, as is the case for notch tensile tests. The maximum load in bending is not controlled by instability: in very ductile specimens, with increasing strain, the load still continues to rise long after the maximum load strain in a tensile test has been exceeded. Thus, the apparently greater severity of notch disk tests than notch tensile tests might well be due entirely to the greater notch sharpness, though this fact could be ascertained only by testing notch tensile specimens with equally sharp notches.

Before concluding, one other aspect of the test results should be mentioned. Under conditions of ductile crack propagation it has been observed in both notch bend tests(18), Figure 2, and notch tensile tests that the load may increase while a crack is tearing open. This phenomenon also occurred in the notched disk test with heat treatment A. This disk was stopped and examined at 95.5 per cent of the bursting speed, and a crack was observed. Then, there was a total increase of 4.5 per cent in speed or 9 per cent in average stress before failure finally occurred. Thus, the higher notched disk strength ratio for heat treatment A than for B apparently is partly due to the more ductile nature of the crack propagation in this material. (The slow notched bend tests, Figure 3, also indicated that heat treatment A was characterized by a larger fraction of ductile crack propagation than heat treatment B.)

Discussion and Summary

For the present, the principal value of notch tensile testing to engineering problems is to distinguish between "notch ductile" and "notch brittle" behavior in a notched member carrying tensile load. When the metal is very notch ductile, a satisfactory design procedure for notched parts subjected to tension is to put the average stress on the minimum section equal to tensile or rupture strength, divided by a suitable safety factor. This procedure is superior to the classical method using elasticity theory and the condition of yielding to avoid any plastic flow, even locally. This classical method is applicable to very brittle metals, but is over-conservative for metals that have even a small ductility. When the metal has some ductility, but not enough to permit the use of the average stress in design, there is no known design procedure that is applicable. The development of a design procedure for these notch brittle metals awaits a more fundamental understanding of the various factors that govern fracturing--size, stress state, strain rate, residual stress, strain gradient, etc. For engineering purposes, this understanding need not be detailed enough to be able to state the mechanism of fracture on an atomic or crystallographic level; but it must be complete enough to permit the prediction of one test result from the results of other tests. After all, engineering design is basically the relating of service performance to laboratory test behavior; and how can one relate complex service performance to laboratory test behavior if one cannot relate the results of two different simple laboratory tests?

The most obvious basis for correlating different test results is that fracture occurs at the same strain for the same local conditions of stress state, etc., regardless of the conditions present in neighboring regions. Test result correlation on this basis appears rather difficult for the sub-surface crack initiation, because the conditions below the surface are hard to determine, and also because it is hard to determine the exact location of crack initiation.

A more practical, though less scientific basis for test correlation has been suggested by Sachs(21). This suggestion is based on the hypothesis that a particular elastic stress distribution will be transformed by plastic flow into a particular strain distribution and a particular triaxiality distribution, regardless of the conditions of geometry and loading that caused that elastic stress distribution. This hypothesis is reasonable, and if true, would permit comparison between test results and engineering performance without knowing the local conditions at the place where fracture begins but only knowing the more easily measured surface strains as follows. A notched test specimen can be devised so as to have an elastic stress distribution essentially the same as a manufactured part as determined by photoelasticity. The test specimen fails when the surface strain reaches a certain value, corresponding to some smaller ductility value below the root of the notch, where triaxiality has caused failure. In the manufactured part, the same triaxiality should cause failure at the same ductility at the same distance below the surface; and because the neighboring strain distribution is the same as in the test specimen, the manufactured part also has the same surface strain at fracture as the measured value from the test specimen. This surface strain can be converted to load-carrying capacity by means of a model test.

Regardless of what basis is found ultimately to be suitable for relating different tests to each other, one requirement for the efforts of the immediate future seems clear: it is imperative to make more kinds of tests on a few materials so as to understand the tests better rather than accepting the value of some given test on faith and making tests only of this kind on a great variety of materials. Any serious efforts to compare different tests in terms of some basic concept, such as the comparisons for the special alloy steel described above, will be the beginning of the development of a rational procedure for designing against fracturing.

So far, this approach to the problem has brought to light a mass of confusing and conflicting data, only a small part of which can be rationalized satisfactorily in terms of current concepts. Out of this mass of data are emerging a few tentative generalizations, and a host of ideas as to how to pursue these concepts further, using various experimental methods to track down the factors that cause differences in the results of different types of tests.

The following tentative generalizations can be made from all the notch tensile testing work that has been done:

1. Fracture begins at a given local strain for a given set of local conditions, irrespective of conditions in the surrounding material or the overall size.

2. Fracture may begin either at the surface or within the volume of metal, depending upon the relative importance of two factors: strain gradient in the vicinity of the surface, and the susceptibility of the metal to triaxiality.

3. Ductility values determined from measurements of dimensions after complete separation of the specimen may be misleading if the crack propagates in a ductile manner. For example, the reduction of area measured in this way in a notch tensile specimen will be too large if the central core of metal continues to stretch while the crack is slowly progressing inward from the root of the notch ("rim effect"). In many cases, this phenomenon will even affect the quantity which is being used to characterize the load-carrying capacity, such as the notch strength of a tensile specimen or the bursting speed in a disk test.

4. Values of energy to cause complete rupture, such as in the Charpy impact test, may be misleading; a high value of energy may be due to high ductility in one kind of metal, or to the mode of crack propagation in another metal.

Some of the experimental methods that can be used to track down further the factors that cause differences in the results of different kinds of tests are as follows:

1. In different kinds of tests the same quantity (such as local stress or local strain) should be measured in all cases, rather than measuring one quantity in one test and a different quantity in another test. For example, it is difficult to compare the Charpy impact test, where the total energy to rupture the specimen is measured, with the unnotched tensile test, where ultimate strength, reduction of area, etc., are measured.

2. Any suspicions as to what variable might be governing the result of a test should be checked by comparing two tests that differ only in that variable. In this way the various pertinent factors can be isolated and studied individually. For example, biaxiality is presumed to be the controlling factor in a notch tensile test when fracture begins at the surface; if the strain gradient in the normal direction is not a factor, the local strain at fracturing should be the same as in an unnotched bend test having the same biaxiality. If the comparison of these two kinds of tests proves this to be the case, the fact that biaxiality, not strain gradient, is the governing variable will be proved to be the case.

3. A valid generalization must be based on tests of a wide variety of metal behaviors such as hardness, strain hardening rate, tendency toward brittle crack propagation, ductility, etc.

4. The method of calculating stress and strain distributions from classical plasticity must be improved. Particularly, methods must be developed for treating situations where the elastic and plastic strains are comparable in magnitude. Some

of the poorly verified assumptions used in such calculations should be checked by experiment.

5. The wider the variety of tests compared, the broader will be the engineering applicability of the generalizations derived from these tests.

6. A direct experimental method is needed for finding the location of crack initiation when the fracture begins below the surface.

7. Several other situations have been discussed throughout the text, and the specific clarifying experiments which suggest themselves have already been discussed there.

ACKNOWLEDGMENT

The authors would like to acknowledge the interest and encouragement of B. M. Wundt and also express appreciation to R. Sakamoto, W. E. Jones, and H. A. Lequear for assistance in making the tests.

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APPENDIX

Details of Procedure and Test Results

Several types of tests were made on a special alloy steel. The composition and heat treatments are shown in Table I. This material was selected because it could be heat treated to produce either high or low impact strength while still maintaining high ductility in a tensile test. These heat treatments are designated by A and B, respectively, in Table I. Standard tensile and keyhole Charpy impact tests and five types of special tests were made for each of these two heat treatments. Some of the special types of tests were also made for a third condition of heat treatment (C) which resulted in both low tensile ductility (except for sporadic high values) and low impact strength.

The steel was supplied in the form of four-inch-square billets by the Simonds Saw and Steel Company. Disk forgings were obtained by upsetting at a maximum temperature of 2100°F. The forgings were annealed at 1700°F and furnace cooled before final heat treatment. Some of the forgings were cut up for impact tests and notched and unnotched tensile tests, and the rest were machined to disk bursting specimens. The smaller types of specimens were also cut out of failed disks, provided that locations away from the fractures could be found in which the tangential strain due to the bursting tests was small. (Less than 3 per cent).

Standard Tests

Unnotched tensile specimens and keyhole Charpy impact specimens were machined and tested according to the ASTM standards. In addition to the standard measurements, the reduction in diameter away from the neck after fracture was also measured (necking strain). The tensile specimens had a 0.250-inch diameter, one-inch long smooth section. The results are shown in Table II. Smaller specimens cut from broken disks give results that agreed within the limit of scattering with results of tests on specimens cut from disks not tested by spinning.

Disk Bursting Tests

After heat treatment the upset forgings were machined to disks of the following dimensions: 10-inch outside diameter, 1 1/2-inch central hole diameter, and 0.375-inch thick. Two disks were severely notched by saw cutting with a jeweler's saw along a diameter to a depth of 0.7, the final 0.025 inch of which was 0.005-inch thick, Figure 5. One disk has a forging crack of about 3/4-inch depth in the rim, as determined by Magnaflux, Figure 6. One disk has a special stress concentration near the rim, as shown in Figure 7. The disks with cracked or slotted rims also had two holes at the opposite edge, as shown in Figure 6.

The disks were tested by spinning to destruction in a high-speed vacuum pit. The energy from the air turbine drive was transmitted through a slender spindle to the disk by means of a tulip-type expansion adapter (Figure 8). The speed of the assembly was controlled by a hand throttle on the air supply and was measured with an electromagnetic pickup. The speed indicator could be read easily to five cycles per second.

The average tangential stress in the disk was calculated from the bursting speed as follows. The radial force exerted by one-half the disk upon the other is:

$$\text{Force in pounds} = \frac{8\pi^2 R^2 \alpha t(b^3 - a^3)}{3g}$$

where $\underline{R}$ = revolutions per second, α = density in pounds per cubic inch, $\underline{t}$ = thickness in inches, $\underline{b}$ = radius to outside in inches, $\underline{a}$ = radius of the hole in inches, and $\underline{g}$ = acceleration of gravity in inches per second per second. This force acts over an area of:

$$\text{Area} = t(2b - 2a)$$

in the case of the unnotched disks, and over a smaller area, that can be readily calculated, in the case of the disks with various stress concentrations.

The results of the disk bursting tests are shown in Table II. The disk strength ratio for unnotched disks is unity for ductile metal and less than unity for brittle material. In other words, for ductile metal the average tangential stress at bursting is as high as the tensile strength while this average stress is below the tensile strength for metal with less than a certain ductility.

These results confirm the generalization made by Holms and Repko(22) that the disk strength ratio is unity for ductility values greater than 5 per cent. On the other hand,

Fig. 6. Rim cracks outlined by Magnaflux.

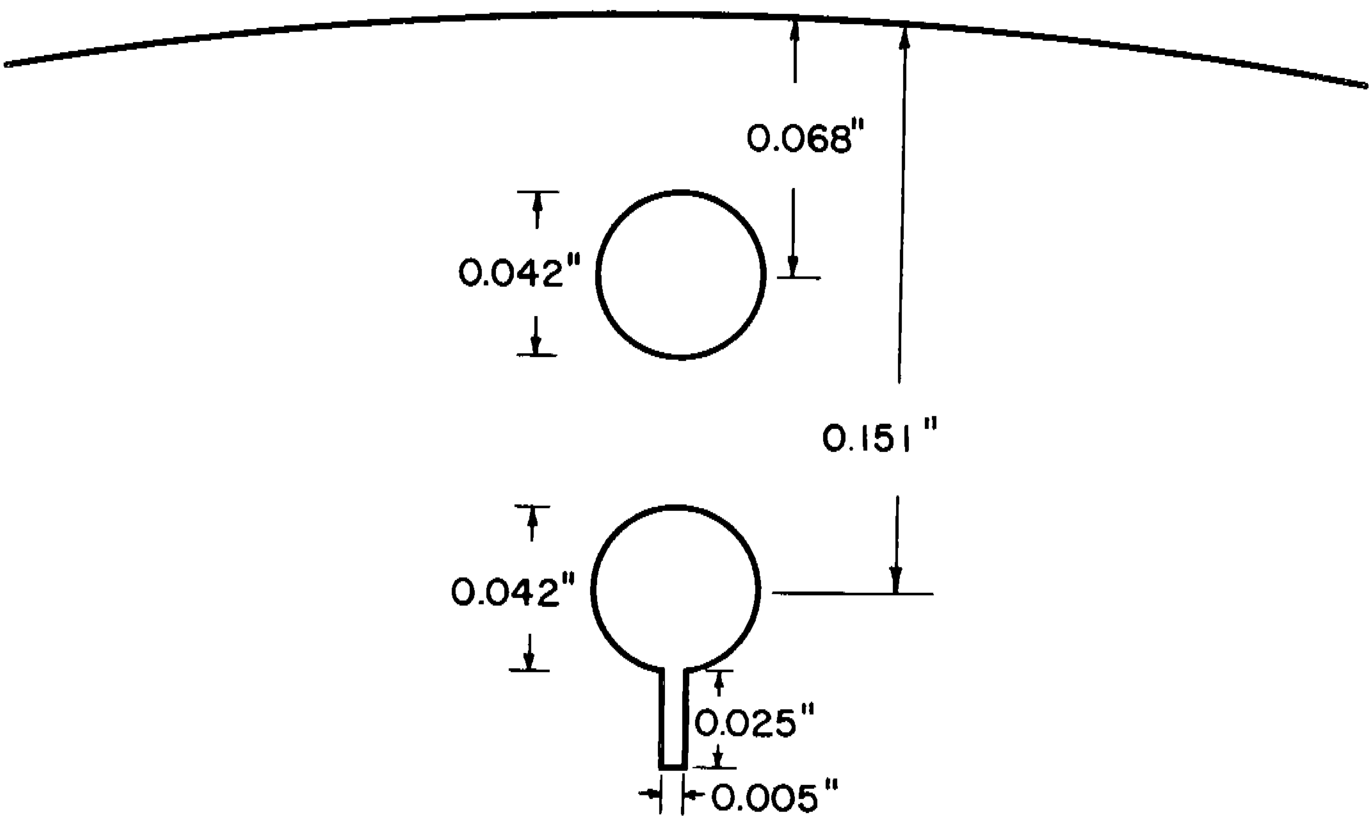

Fig. 7. Special stress concentration at rim of disk. Two identical holes at 180 degrees from these, but without saw cut.

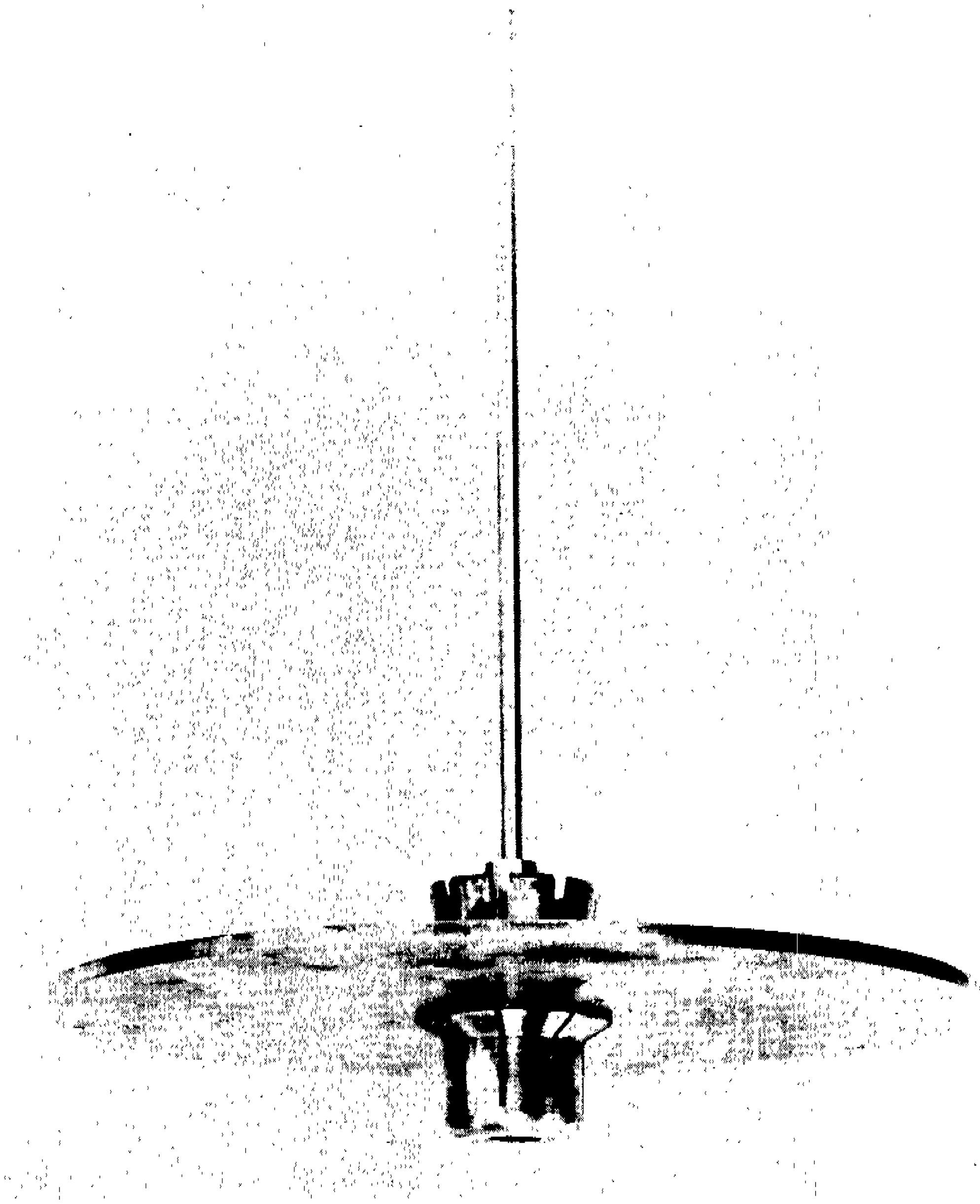

Fig. 8. Tulip-type expansion adapter used to attach spinning disk to rotating shaft.

Fig. 9. Ductile fracture in an unnotched disk, showing 45 degree shear fracture along radial neck.

this is not a valid generalization; Robinson(23) reports tests on metals with more than 15 per cent elongation in tension and disk strength ratios as low as 0.7.

The ductile disks broke into three pieces with 45-degree shear fractures along radial necks, Figure 9, except for the last inch or so at the outside surface. These disks apparently suffered 6 to 8 per cent uniform reduction of thickness before local necking, according to measurements on failed disks away from the necked areas. The brittle disk broke into more than fifty pieces. The fracture surfaces were normal to the wheel surface. There was no evidence of any ductility preceding or accompanying the propagation of the cracks.

For heat treatment A, the disk strength ratio is as high when a hole and short saw cut is present as when not; only a deep saw cut in the rim causes appreciable loss of strength. For heat treatment B, the presence of a notch causes much more loss of bursting strength than for heat treatment A. Sharp cracks are apparently no more severe in this respect than a saw cut (Table II).

The slotted disk of heat treatment A showed significant widening of the slot after spinning at 26,400 r.p.m. (compare Figures 5 and 10), and there was evidence of considerable deformation at the root of the slot, Figure 10. After spinning at 27,000 r.p.m., the disk had thinned to 0.357 inch at the root of the saw cut, and a crack had started. The crack was 1/8-inch deep on the surfaces and estimated to be at least 3/8-inch deep at mid-thickness of the disk.

The cracked disks and slotted disks of heat treatment B exhibited brittle primary crack propagation, starting at the base of the crack or slot, as indicated by the chevron markings, Figure 11. The secondary cracks were ductile, as in the unnotched disks. On the other hand, the disks of A heat treatment with stress raisers present exhibited ductile crack propagation. In the slotted disk the crack changed from normal to 45-degree shortly after leaving the notch bottom. Its propagation was accompanied by deep necking. In the disk with the special hole-and-slot stress raiser, Figure 7, the crack did not go through the saw-cut stress raiser, Figure 12. The slot closed up during spinning.

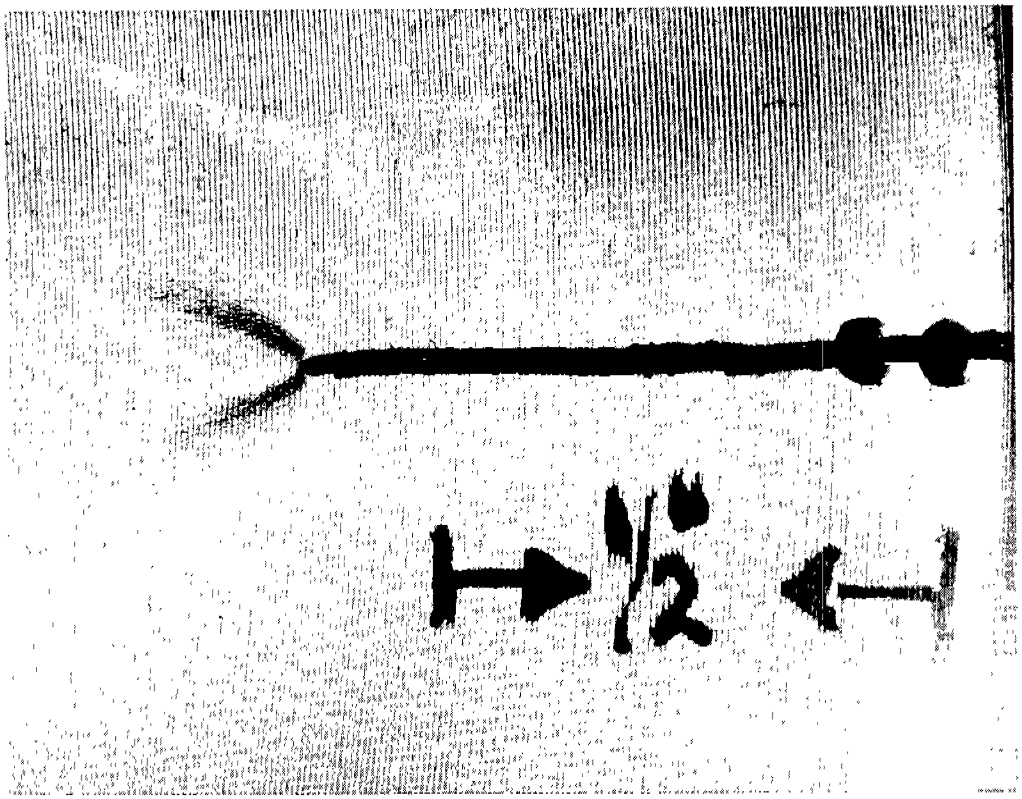

Fig. 10. Saw cut in rim of disk (shown in Fig. 5) after spinning at 26,400 R.P.M. for one minute. Saw cut has widened, and considerable local deformation has occurred.

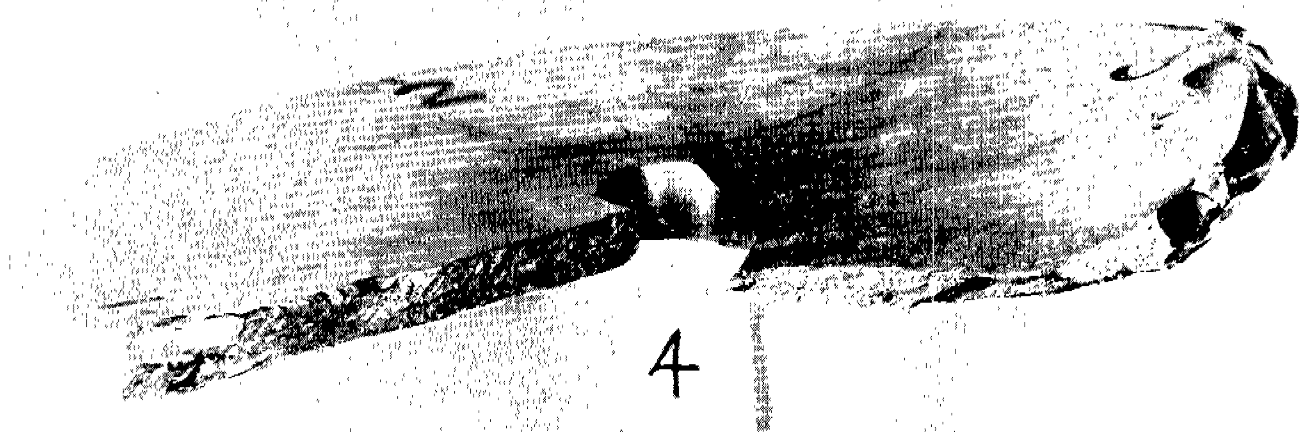

Fig. 11. Brittle crack propagation starting at base of initial crack, as indicated by chevron markings (Heat treatment B).

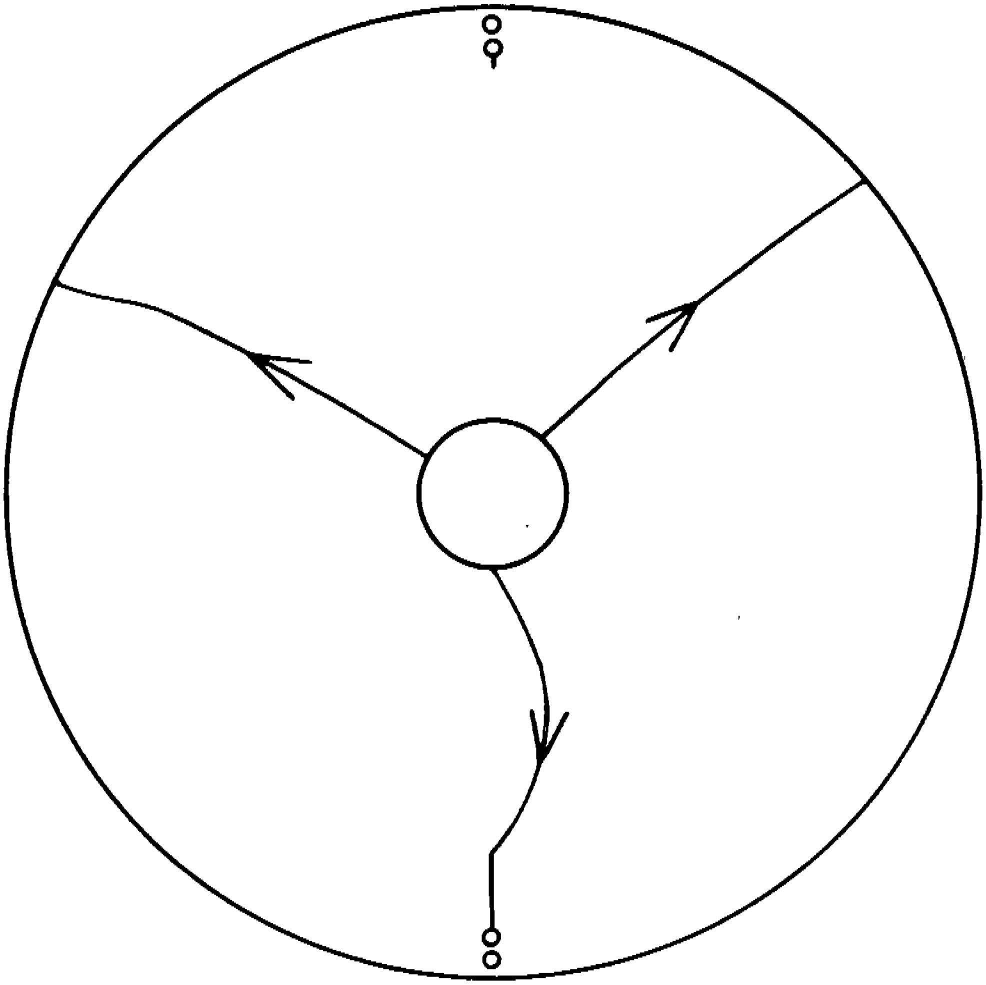

Fig. 12. Fracture pattern for disk with special stress concentration near rim. (Heat treatment A).

TABLE II

Details of Test Results

	Heat Treatment A	Heat Treatment B	Heat Treatment C
Tensile Test Results			
0.02% Yield Strength--1000 psi	56.4	75.0	94.0
	53.9	75.0	95.9
	51.4	78.4*	101.9
	53.4	73.0	98.0
	42.8*	75.5	96.1
		57.2*	93.1
		67.4*	
Tensile Strength--1000 psi	77.6	94.7	148.0
	76.8	102.0*	147.8
	76.0	91.8	145.0
	83.3*	92.3	145.3
	79.6	89.6*	149.4
		92.6*	147.0
		95.1	
Per Cent Reduction of Area after Complete Separation	77	74	3
	78	76	63
	79	73*	3
	78	73	2
	78*	73	1
		74*	59
		76*	
Per Cent Reduction of Diameter at Beginning of Necking	4	3	
Keyhole Charpy Impact Energy--ft-lb	42	2	1
	52	4	1
	43	2*	2
	36	3	2
	10*	2	2
		3*	
		4*	
Notch Tensile Test Results	134.0*		
Notch Strength--1000 psi	124.1	146.8	132.1*
	122.5	139.5	122.8*
Per Cent Reduction of Area at Crack Initiation	19*		
	21*		
Per Cent Reduction of Area after Complete Separation	38	1.2	1.3*
	53	0.6	1.0*
	30*		
	40*		

*Cut from failed disks

TABLE II (continued)

	A	B	C
Disk Bursting Tests			
Bursting Speed--1000 r.p.m.			
No Stress Raiser	31.8	35.7	39.9
3/4-inch Deep Crack in Rim		28.2	
Saw Cut in Rim	28.2	24.6	
Special Stress Raiser (Fig. 7)	32.1		
Average Tangential Stress--1000 psi			
No Stress Raiser	80.1	100.9	125.1
3/4-inch Deep Crack in Rim		69.2	
Saw Cut in Rim	68.2	53.0	
Special Stress Raiser (Fig. 7)	83.5		
Tensile Strengths of Actual Disks--1000 psi			
No Stress Raiser	79.0**	102.0	147.0**
3/4-inch Deep Crack in Rim		89.6	
Saw Cut in Rim	79.0**	92.6	
Special Stress Raiser (Fig. 7)	83.3		
Disk Strength Ratio			
No Stress Raiser	1.01	0.99	0.85
3/4-inch Deep Crack in Rim		.76	
Saw Cut in Rim	.86	.57	
Special Stress Raiser (Fig. 7)	1.00		
Per Cent Reduction of Thickness away from Local Necks in Unnotched Disks	8	6	
Wide, Unnotched Bend Test Ductility, %	>120*	>120*	
Notch Bend Tests*			
Local Strain at Root of Notch when Crack First Appears	165 180	3 12 4 5	
Energy Absorbed--ft-lb	38.3	1.2	
Bending Load--pounds	2220 2440	1740 1980	

* Cut from failed disks.
**Average of all tests (no test from the disk in question).

Unnotch Bend Tests

The specimens were 3/16 by 2 inches in cross section and 3 inches long. According to Sachs et al.(13), the breadth-to-depth ratio (b/h) of 2/(3/16) = approximately 11 corresponds to a biaxiality of 0.48 to 0.50, depending on the curvature. One specimen each for heat treatments A and B was cut from a failed, notched disk in such a direction that the principal bending stress was in the tangential direction of the disk, and such that the transverse direction of bending was in the radial direction of the disk.

The specimens were first bent to 120 degrees in four-point loading fixtures that were especially designed for use in a furnace or liquid bath, Figures 13 and 14. Then they were squeezed between compression heads until the machine limit of 60,000 pounds was reached. No cracks were present in either specimen after the specimens had been bent flat shut, Figure 15. This condition corresponds to about 120 per cent conventional tangential strain on the convex side.

Notch Bend Tests

Standard V-notch Charpy specimens were cut in the tangential direction of failed, notched disks. The notch was usually cut across the specimen in the axial direction, giving orientation relationships like those in the disk tests, but a few specimens had the notch in the radial direction, giving orientation relationships like those in the unnotched bend tests. No significant behavior difference was observed for the two notch orientations.

Bending was performed in special fixtures designed for use in a furnace or liquid bath, Figures 16 and 17. The geometry of the anvils and punch was the same as in a standard Charpy impact machine.

The load was increased in small increments beginning with a load estimated to be low enough not to give any plastic strain. As plastic flow progressed, the load increments were decreased as necessary to keep the strain increments sufficiently small. After each load increment the specimen was unloaded and removed from the fixtures in order to measure bend angle and to look for cracks.

Small angles were measured with a curvature gauge, Figure 18. For this application, the difference between the

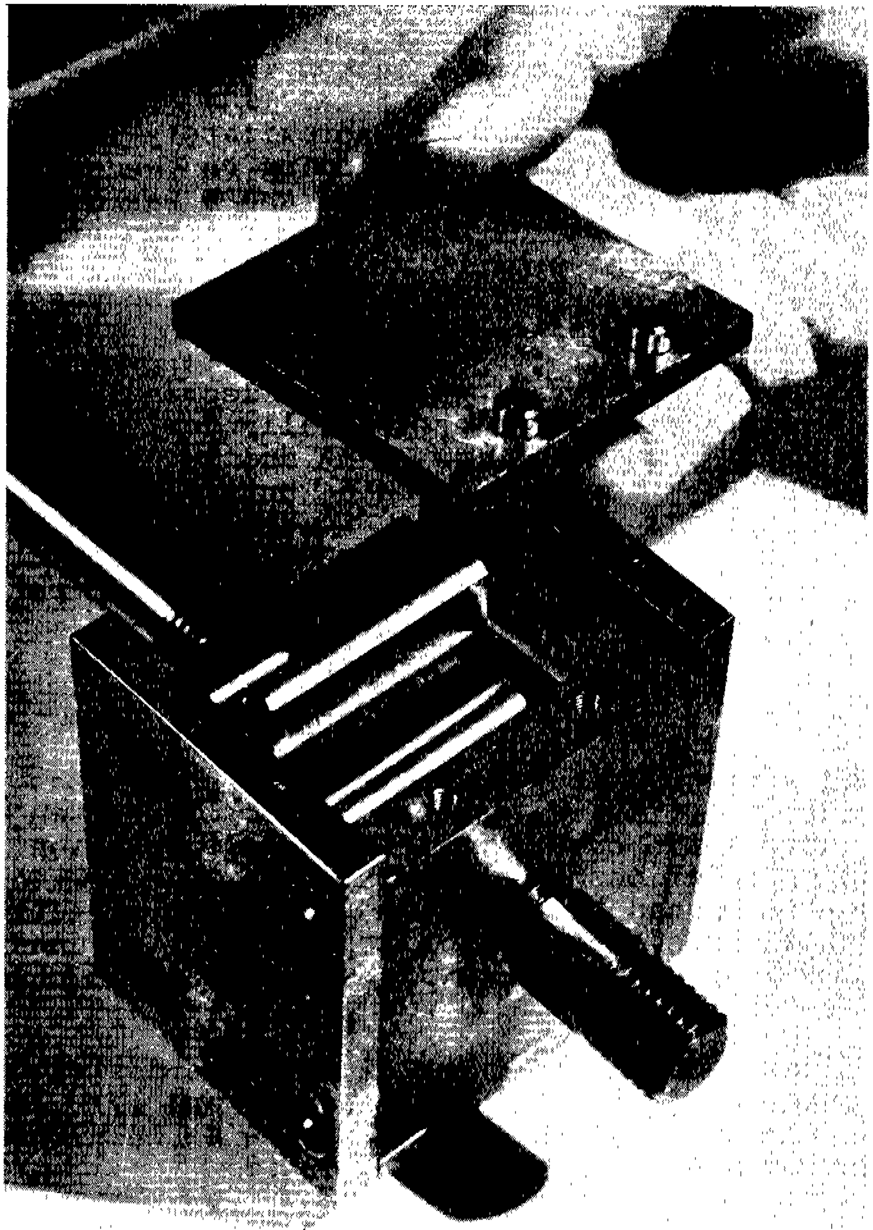

Fig. 13. Unnotched Bending Fixture.

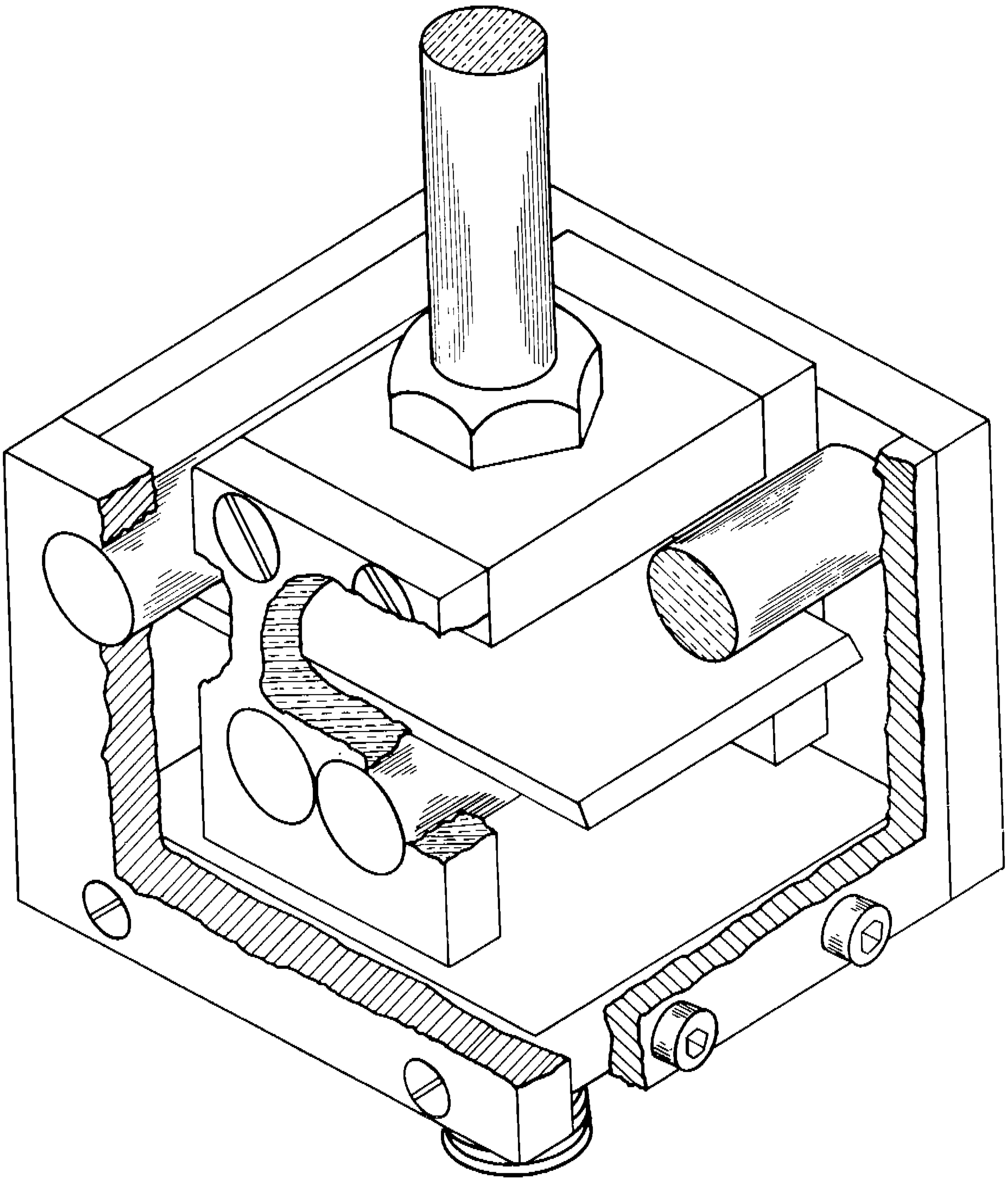

Fig. 14. Unnotched bending fixture.

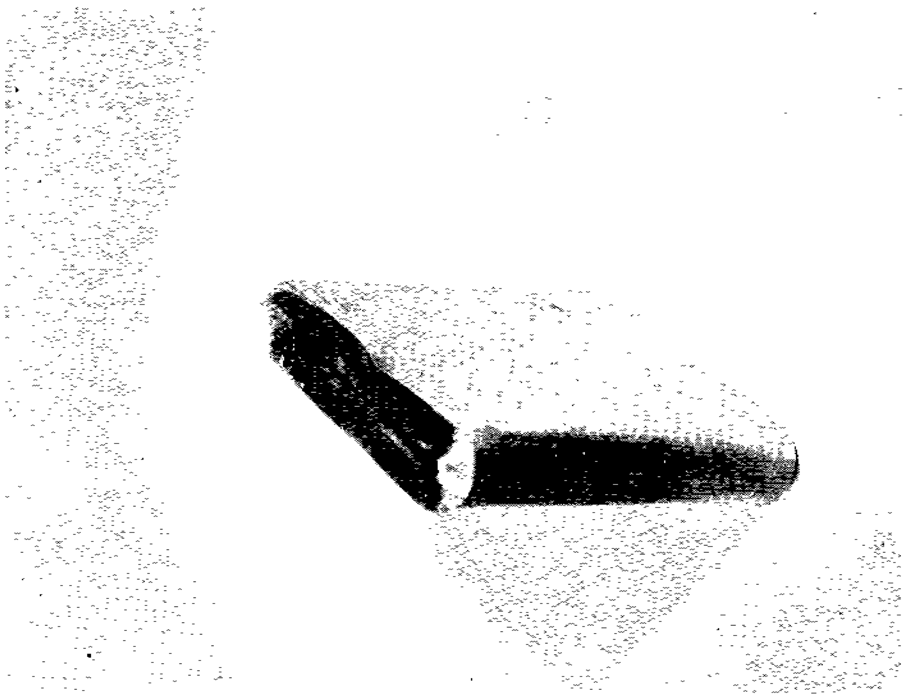

Fig. 15. Unnotched bend specimen folded flat.

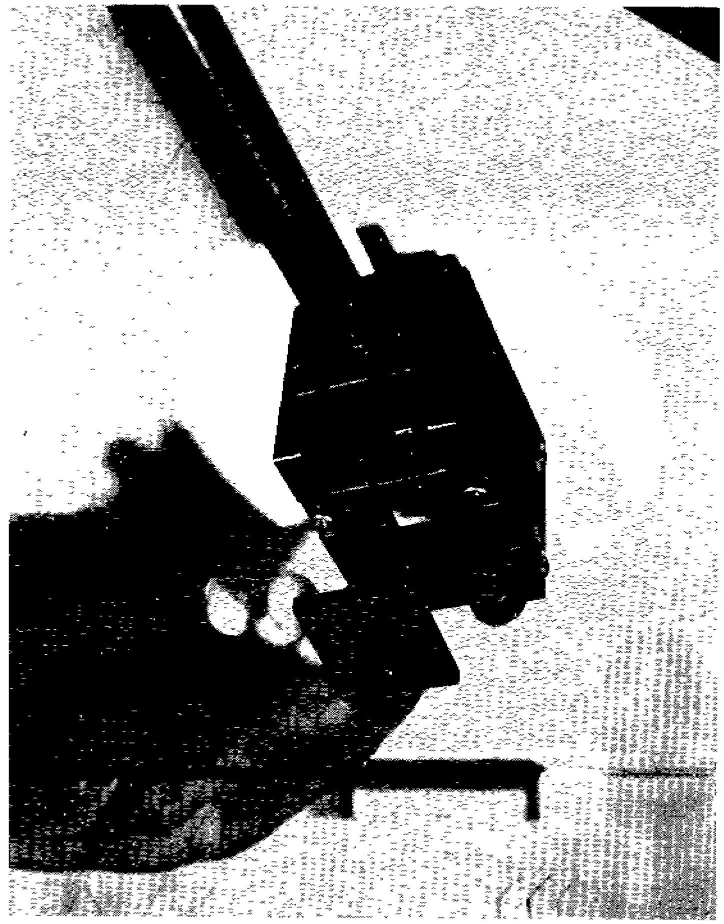

Fig. 16. Notched bending fixture.

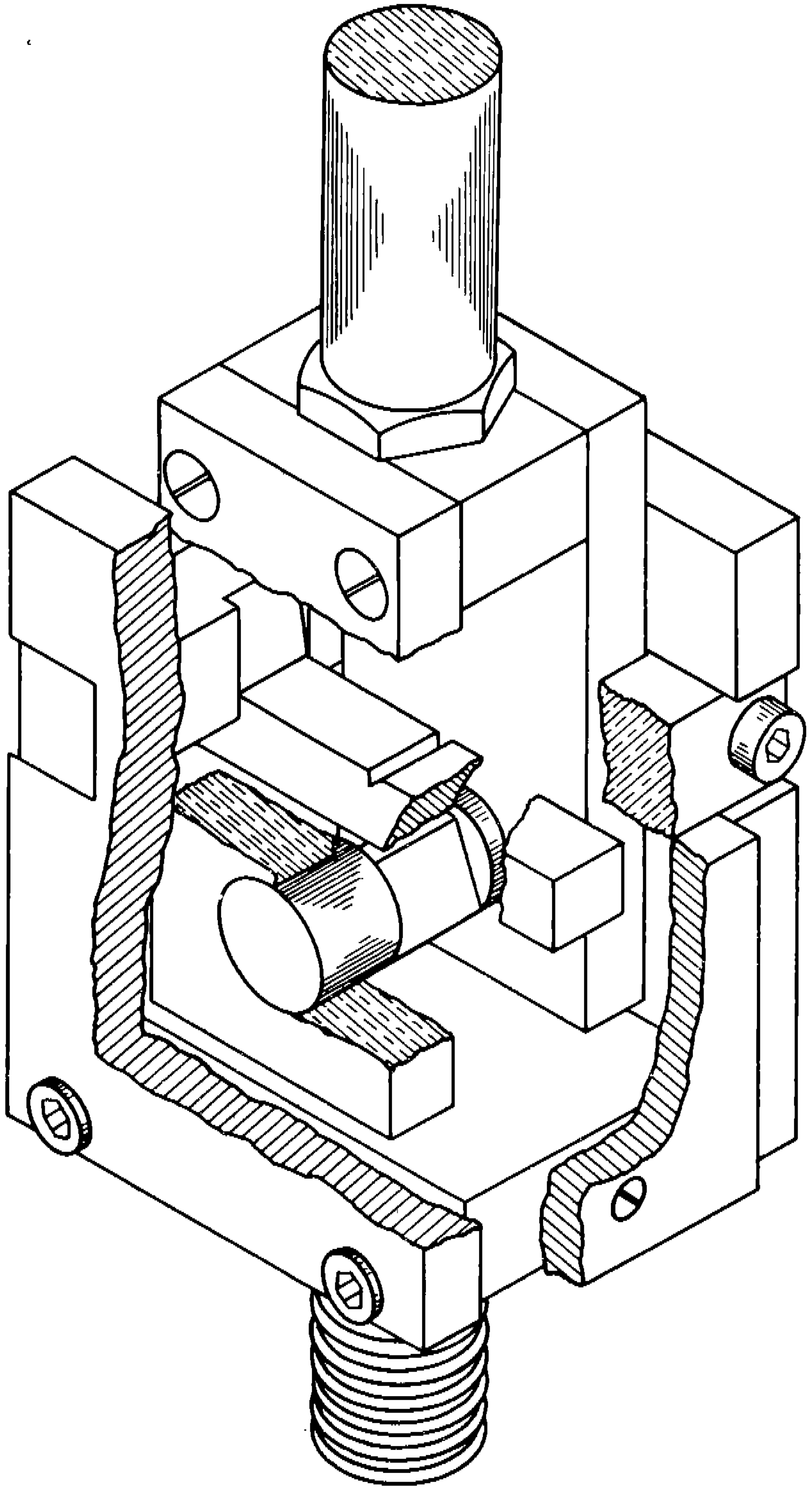

Fig. 17. Notched bending fixture.

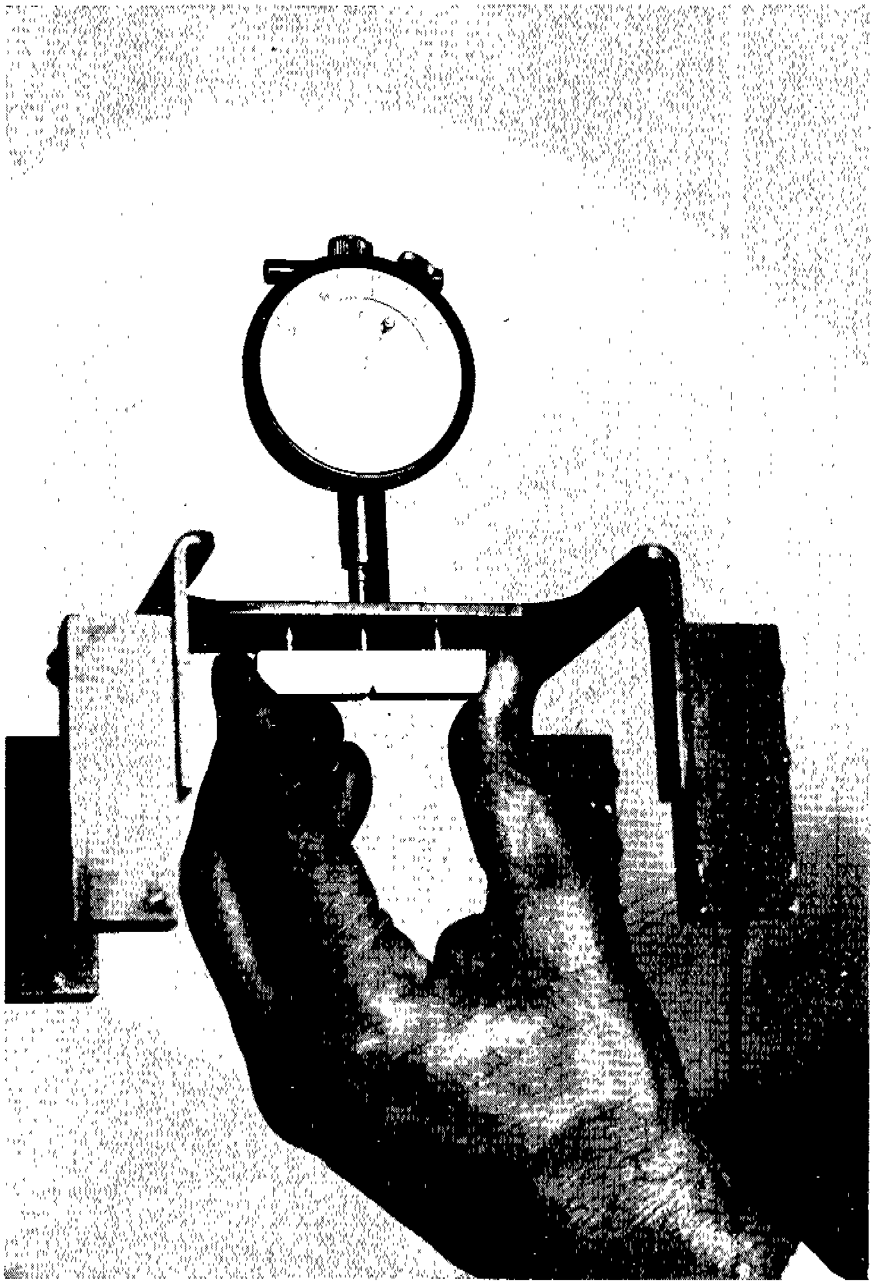

Fig. 18. Curvature gage being used to determine the bend angle of a notched specimen.

dial gauge reading for a specimen and for a flat plate, divided by half the distance between the fixed anvils, equals the tangent of half the bend angle. The bend angle is needed to determine the local strain at the root of the notch, Figure 19. Large bend angles were measured with a protractor which could be read to ±0.2 degrees. For sudden type fractures, the load vs. bend angle curve could be extrapolated a short distance beyond the last point to the observed fracture load, thus giving the bend angle at fracture. In some cases, however, failure occurred suddenly at a load smaller than that reached prior to the previous unloading, and then the last measured bend angle was taken as the bend angle to fracture. For the most ductile specimens, the calibration curve did not go far enough to read off local strain for the very large bend angles observed at fracture. In these cases, the approximate slope of ten per cent local conventional strain per one degree of bend angle (Figure 19) was used to convert bend angle into local strain. The values of local ductility at the bottom of the notch are given in Table II.

For the purpose of determining energy to fracture, the bend angles up to crack initiation were converted to deflections using the bending span: deflection = (1/2)(bending span)(tan. 1/2 bend angle). The additional deflection occurring during the propagation of the crack was determined from the motion of the heads of the testing machine. Figure 3 shows load-deflection curves for heat treatments A and B. The area under such a curve is the energy to fracture, and these energy values are given in Table II.

For the brittle specimens, the exact moment when the crack started was very definite. There was a noise and sudden drop in load, after which a crack of considerable depth, Figure 20, could be observed. For the ductile specimens, the final crack began to open up unobtrusively; and so the fracture bend angle was taken as the angle when a definite crack was first observed. The crack was considered to be "definite" when it was significantly deeper than the initial machining marks which had spread open into characteristic "furrows," Figure 21.

For heat treatment B, the crack propagated in a brittle manner across the entire specimen, Figure 22. For heat treatment A, the crack propagated most of the way in a brittle manner, Figure 22, but there was a gradual tearing for a short distance at the beginning and end of the process. Also, there were shear lips on the sides, which were not present for heat treatment B. Figure 22 shows the marked distortion of the

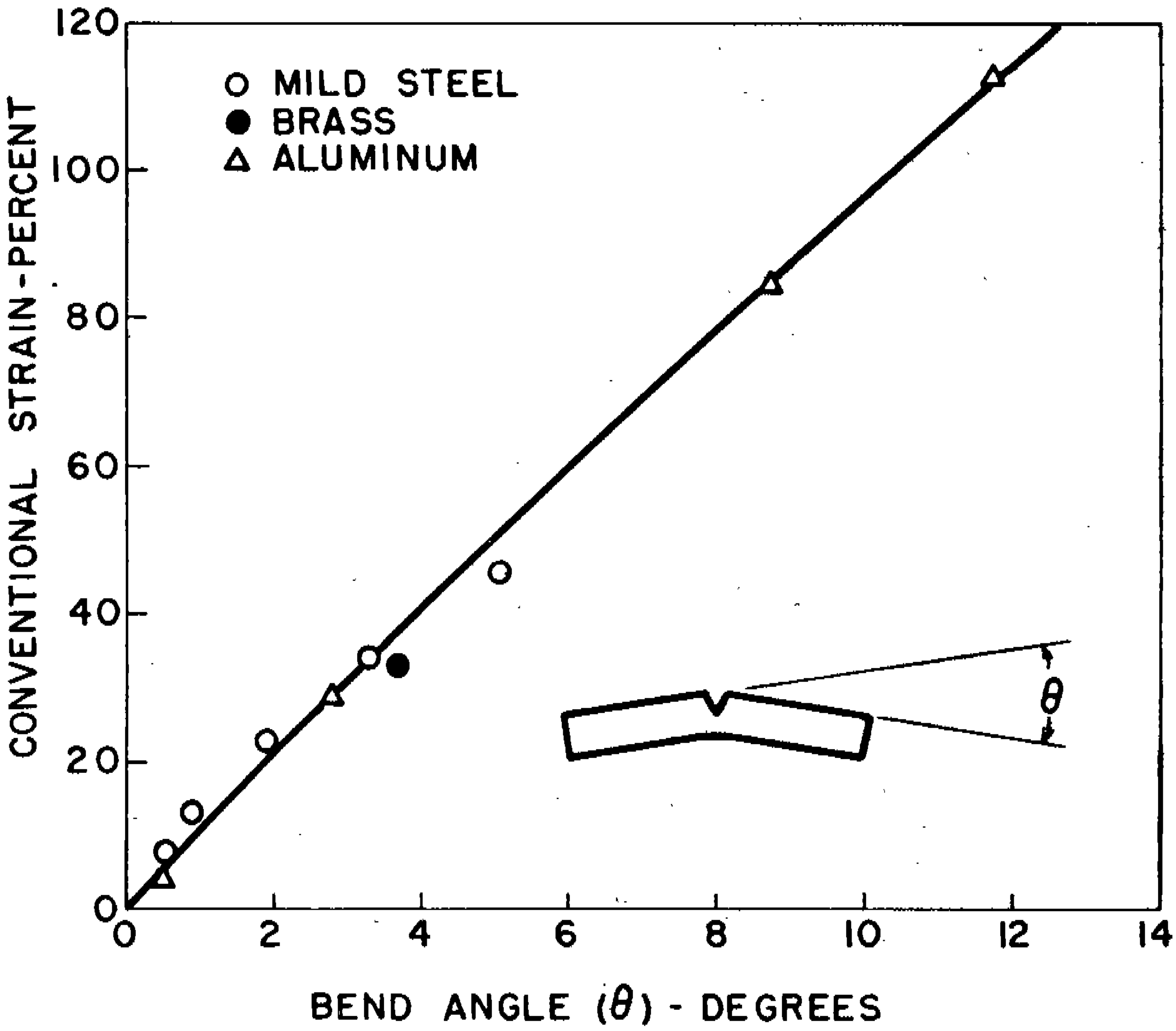

Fig. 19. Calibration curve of bend angle vs. local strain at the notch bottom. (18)

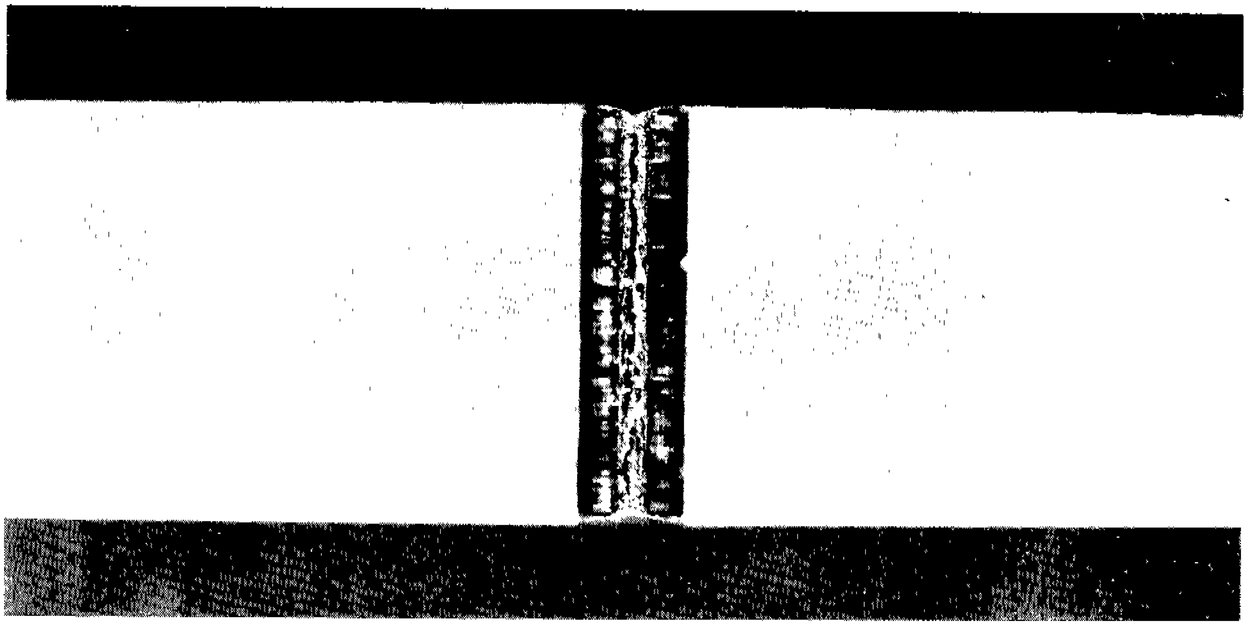

Fig. 20. Brittle notch bend specimen immediately after initiation of a crack (Heat treatment A).

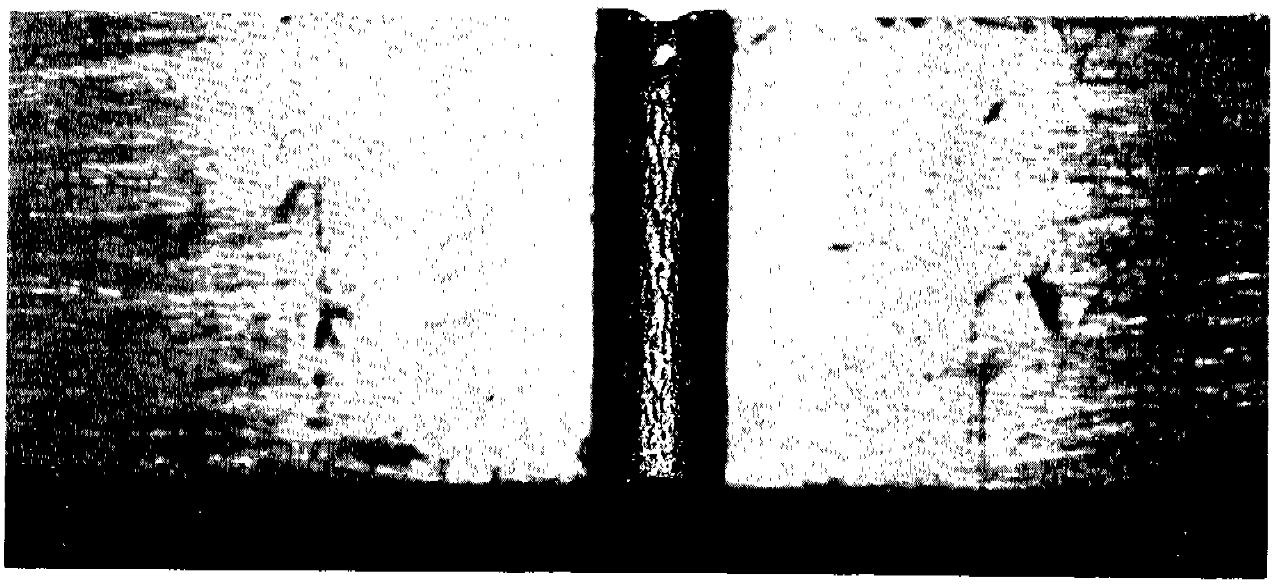

Fig. 21. Ductile notch bend specimen after considerable strain but before initiation of a crack. "Furrows" in notch bottom are machining marks which have opened up into blunt grooves.

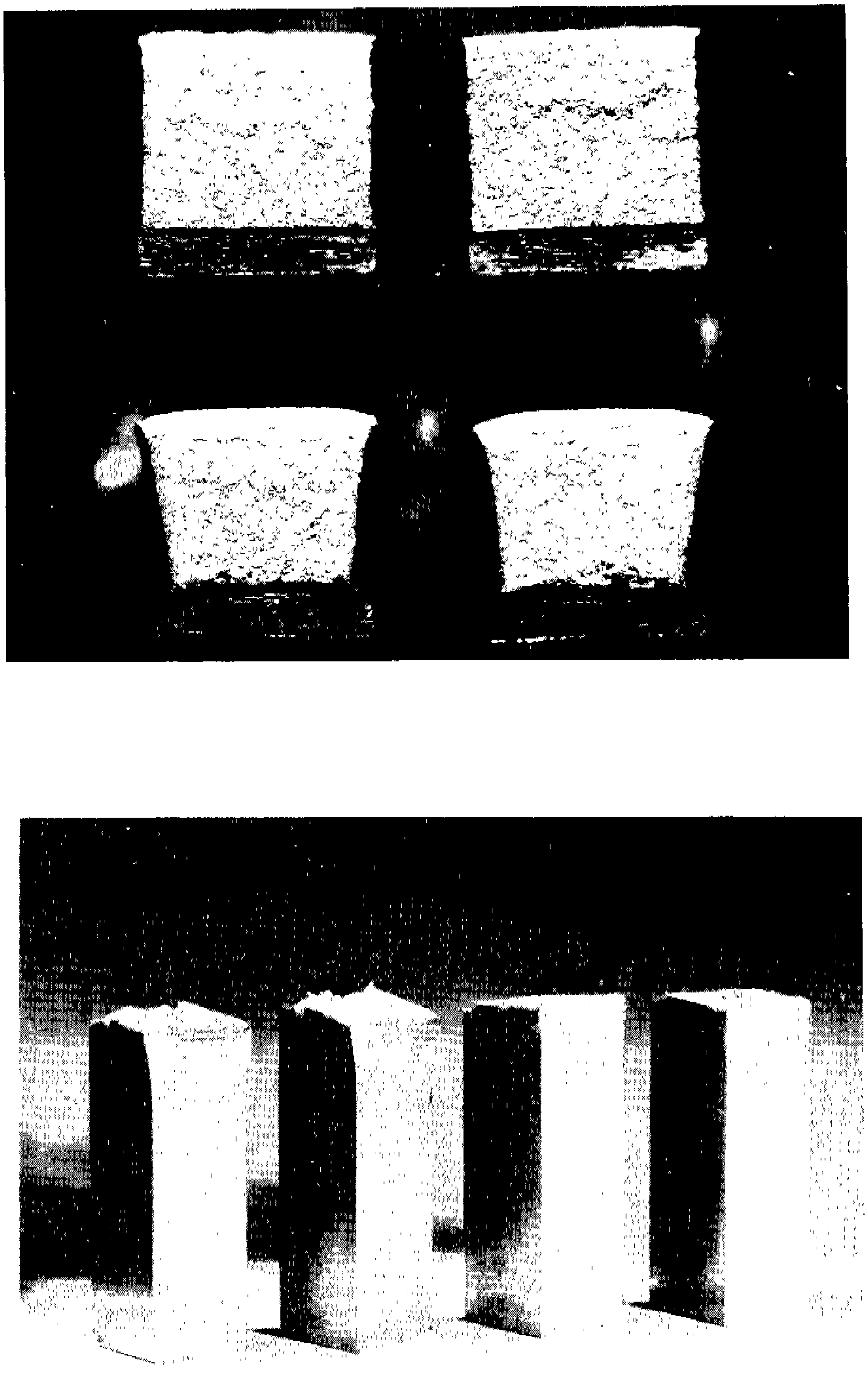

Fig. 22. Comparison of fracture surfaces of notched bend specimens for Heat Treatments A and B.

cross section due to the large ductility associated with heat treatment A. The lack of such distortion for heat treatment B is also shown by the other specimen in this figure. The difference in mode of crack propagation between specimens of A and B heat treatment is also indicated by the load-deflection curves, Figure 3. For heat treatment A, there were 5.4 ft-lb associated with the initial tearing and 3.2 ft-lb accompanying the final tearing. For heat treatment B, there was no indication of any slow tearing prior to complete, sudden rupturing. Thus, the energy of crack propagation was less than that stored elastically in the system.

Notch Tensile Tests

The notch was a 60-degree, annular V-groove cut in a 0.250-inch diameter, one-inch long cylindrical section. The notch had a 0.005-inch root radius. The specimen diameter at the root of the notch was 0.175-inch, thus giving a notch depth of 51 per cent.

In most of the tests, only the maximum load and reduction of area after separation were measured. In a few tests on ductile specimens, however, the load-diameter curves were determined, and the point of crack initiation was observed. These latter tests (Table II) showed that part of the "ductility," measured after separation, is fictitious. It is due to the "rim effect"(19).

The results of all the tests are given in Table II. They show that the ductility under notch conditions is much lower for heat treatment B than for heat treatment A. However, the ductility in either case apparently is high enough to smooth out the initial stress concentration because the notch strength ratio is very high for both A and B heat treatments. This is not so for heat treatment C. In this case, the notch strength ratio is far below the ideal value of 1.5. This must be due to a still lower notch ductility for heat treatment C than for B, although the measured values do not show it. (However, they are too small to be very reliable.)

SOME REMARKS ON THE RELATION OF THE GEOMETRY OF WELDED DETAILS TO THEIR SUSCEPTIBILITY TO BRITTLE FRACTURE

by

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University of Washington

I. INTRODUCTION

Certain aspects concerning the design of welded structural steel details are discussed in this paper, particularly such factors as the following which have a bearing on their susceptibility to brittle fracture:

1. The inherent notch severity of a welded detail.
2. The rigidity of the detail.
3. The maximum ductility which we can probably attain in welded structures.
4. The effect of temperature upon the mode of plastic flow and failure of structural members and details.

The information obtained from the tests of plates with openings(1) will be frequently used for purposes of illustration because of their familiarity to the speaker. Other data could also be cited, and it is hoped that a paper of wider scope on this same subject can be prepared for publication in the future.

II. RELATION OF THE GEOMETRY OF THE PLATES WITH OPENINGS TO THE RESULTING STRENGTH AND ENERGY ABSORPTION

Let us examine some of the data of the tests of plates with openings(1), especially those which relate to their ultimate strength and energy-absorbing capacity. The openings in these specimens were circular, square with a 1/32-inch corner radius, or square with a 1-1/8-inch corner radius. The opening was centrally located in a parallel-sided body plate which was either 36 inches by 1/4 inch, 36 inches by 1/2 inch, or 48 inches by 1/2 inch in gross cross section. The sides of the square openings were parallel to the plate edges. The tension load was concentrically applied. The specimens were fabricated from semi-killed steel meeting ASTM Designation A7-49.

A summary of the effect of the notch severity upon the various mechanical properties of the plates with openings is given in Figure 1. Increasing the notch severity resulted in a little

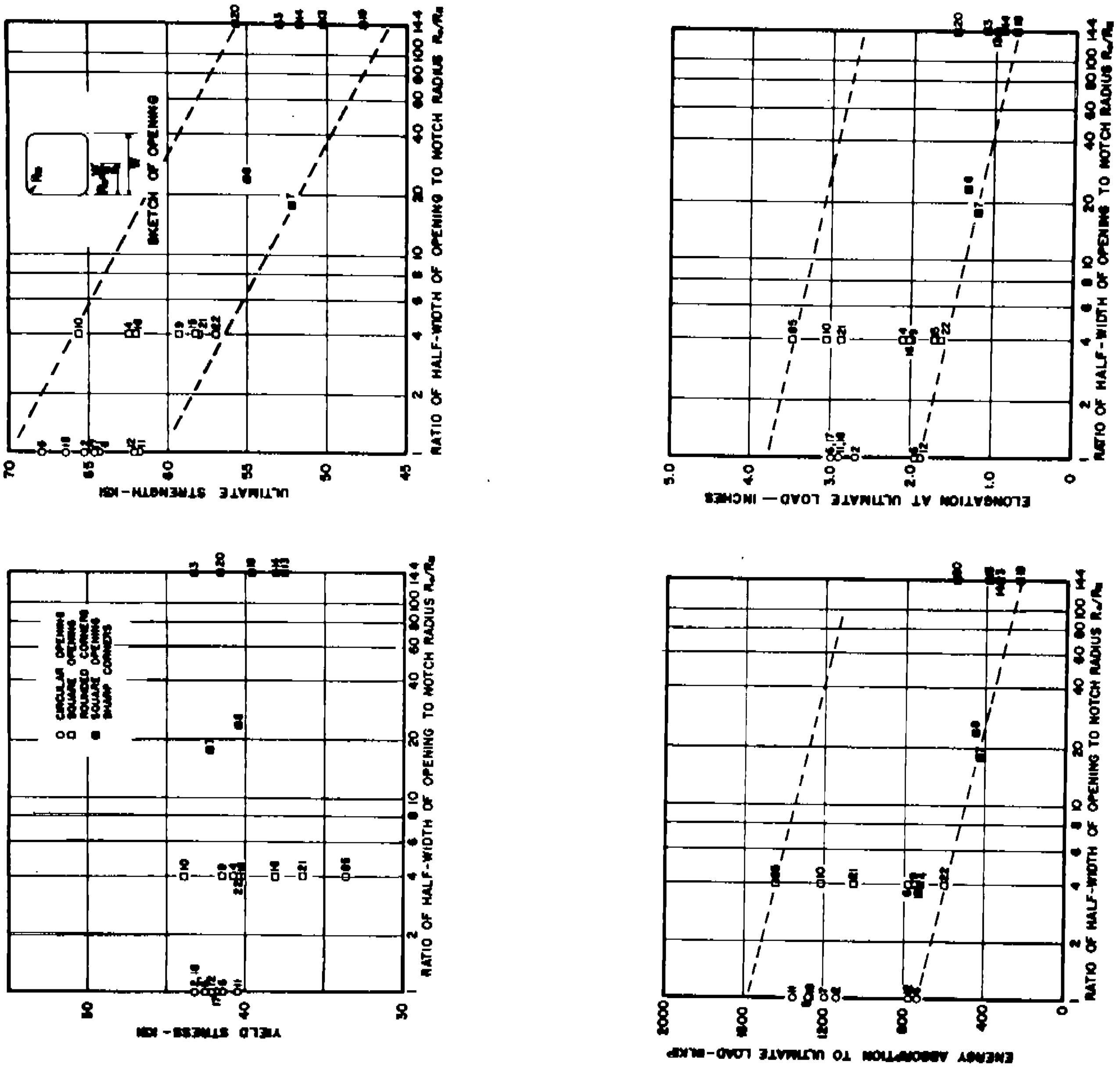


FIG I. EFFECT OF NOTCH ACUITY UPON PROPERTIES OF PLATES WITH OPENINGS.

effect upon the average stress on the net cross section at general yielding of the plates, but was much more effective in reducing the ultimate strength and the energy absorption to failure. It is important to add that through good design it was possible to restore by means of reinforcement the ultimate strength which would be found in a plate which had no opening, but the energy-absorbing capacity of the plates with openings never exceeded 25 to 30 per cent of that of the plates without an opening. It appears possible in the case of shear fractures to design a welded detail which will approach the ultimate strength of the steel. However, the openings, sharp radii, and discontinuities present in these details will prevent them from absorbing more than a small fraction of the potential energy-absorbing capacity of the steel. When brittle fracture occurs in a welded detail, a further drastic reduction in the energy-absorbing capacity results.

It was also found that for shear fractures increasing the severity of the notch was much more effective in reducing the ultimate strength and the energy absorption to ultimate load or failure than changes in either the design or the cross section area of the reinforcement. In fact, among those plates with the square opening with a 1/32-inch corner radius the unreinforced plates gave a better performance than the reinforced plates. When the notch radius was increased to 1-1/8 inches or the opening became circular in shape, the reinforced plates developed a higher ultimate load and energy absorption than the unreinforced plates. However, the reinforced plate which developed the highest energy absorption developed only 32 per cent of the energy absorption of a plate without an opening.

Another observation with respect to the plates with the circular opening or the square opening with rounded corners which underwent shear fractures was this: that increasing the percentage of reinforcement, that is, the cross section area of the reinforcement, brought about improved performance only when the width of the reinforcement in the third direction, the direction of the thickness of the body plate of the specimen, was kept small. For the higher percentages of reinforcement, a single doubler plate was more effective, for example, than a face bar. It has already been stated that for the plates with a square opening with a 1/32-inch corner radius, any reinforcement reduced the performance of the plates. These observations suggest two conclusions:

1. That as the notch severity of the welded detail increases, the danger increases of losing some of the capacity to carry load and to absorb energy

because of too much rigidity in the detail in the third direction.

2. That because most of the structural members and welded connections in a ship have a more complex geometry, a higher stress concentration and a greater width in the third direction than the rather simple structural element used in the tests of plates with reinforced openings, it is not difficult to understand why many details in the welded ship have demonstrated low strength and energy absorption and an inherent tendency towards brittleness.

Translated into terms of design, these statements mean simply that the amount of rigidity, particularly in the third dimension, that can be built into a detail must be decreased as the notch severity of the detail increases.

One of Mr. DeGarmo's conclusions(2) with respect to the hatch-corner tests may well be applied to all welded structures. It is:

> "There are two basically different approaches to improve welded hatch corner design. One results in a very rigid structure wherein improved performance is obtained by the addition of structural members and the reduction of points of high multiaxial stress concentration so far as possible (a problem which is difficult with increased rigidity). The second approach is to design for a minimum of rigidity so that plastic flow may occur naturally and easily, with the result that high stress concentrations do not occur. This second type of design appears to be the superior."

The results of the tests of plates with openings confirm this conclusion.

One of the interesting results of the hatch corner tests(2) was the excellent strength and energy-absorbing capacity of the Kennedy type detail, although it ultimately underwent a cleavage fracture. Similar behavior was observed in the tests of plates with reinforced openings(1). With respect to the latter the following was found: if the original notch in the specimen was not sufficiently sharp to initiate cleavage fracture at the testing temperature and if the testing temperature was below the transition temperature of the steel as determined by the Kahn tear-test or Charpy keyhole test, the first crack to form was a shear fracture; however, this crack then became the predominate stress-raiser and was immediately sufficiently severe to cause a cleavage fracture to pass completely through the plate. The plates with reinforced openings which developed this type of failure absorbed as much energy up to their ultimate load, where the crack was initiated, as those which underwent a 100-per cent shear fracture. It should be noted however that these plates absorbed very little energy during fracture, which occurred in the same explosive manner as in the extremely brittle plates. This type of cleavage fracture which is preceded by a considerable amount of plastic flow has been termed "ductile cleavage fracture." The apparent explanation is that specimens undergoing a ductile cleavage fracture fell in the fracture transition range, and not in the ductility transition range for the particular combination of geometry and steel. It is important to realize that we should not be concerned about the cleavage type of fracture except when it is accompanied by low ductility. Cleavage fracture of itself is not necessarily an evil to be avoided.

III. WHAT MAXIMUM DUCTILITY CAN WE ATTAIN IN SHIP STRUCTURES WITH OUR PRESENT DESIGNS?

It would be well at this point since energy absorption is dependent primarily on ductility to consider what maximum ductility we can expect to attain in our present ship structures. We can use test data to arrive at some indication of what this maximum may be. The measure of ductility used here is the overall average unit strain developed by the welded detail being tested. In tests of a rather simple structural element, the plate with an opening with welded reinforcement(1), it was found that the average unit strain in the region of the opening at the moment fracture was initiated ranged from 2 to 11 per cent for shear fractures. However, most of the values fell between 2 and 6 per cent. This relatively low ductility is quite different from the high ductility of the tensile coupon or of simple plate specimens with longitudinal or transverse butt welds and demonstrates the effect of introducting notches or abrupt changes of cross-section in our structural members. In rather simple types of welded framing and welded structural elements, it is reasonable to say that we would be certain of a maximum ductility at first crack not greater than around 5 per cent.

Now let us examine the ductility of more complex structures, structures where the third dimension in the direction of the thickness of the main plate is of appreciable magnitude. DeGarmo(2) reported that cleavage fractures occurred in all the hatch corner tests except in the case of a portion of one specimen. The ductility of even the best hatch corner was low.

Tests of other welded members with sizable dimensions in the third direction have demonstrated a similar lack of ductility.

It is reasonable to estimate the maximum ductility of the best details in a ship's structure as not greater than around 5 per cent for shear fractures and of the majority of the details considerably less. Many ships with these welded details are giving good service. If the inherent low ductility of steel structures is understood, one may realize that the problem of improving the ductility of our welded hulls is not one of making the present plastic strain-absorbing capacity of our structures tremendously greater, but only a little greater--of increasing the strain-absorbing capacity of some of our bad details and those at critical points from one per cent to about five per cent.

A steel capable of developing an elongation to failure of the magnitude of 25 or 30 per cent in the tensile coupon is required to permit under favorable circumstances this overall ductility of around five per cent in the structural member.

Is an overall ductility of five per cent adequate for ship structures? It would appear that it is. Many types of riveted joints whose plates have been found to undergo a permanent deformation to failure around five per cent in tests have performed satisfactorily in bridge structures. In aircraft structures aluminum alloys with much less potential ductility than structural steel perform satisfactorily. If structural steel were not prone to brittle fracture, undoubtedly less than five per cent ductility would be sufficient. Low notch sensitivity rather than high ductility is the prime requirement for structural steel.

IV. PLASTIC FLOW AND FRACTURE IN PLATES WITH OPENINGS

Let us return to the data of the tests of the plates with openings to examine those data pertaining to plastic flow and fracture. One method of analysis used was Nadai's octahedral theory for the determination of the unit energy distribution throughout the region around the notch. A typical plot of this distribution at the ultimate, or maximum load is shown in Figure 2. It was at this load that fracture was always initiated. Two interesting observations may be made with respect to these data:

1. The higher values of the unit energy absorption became more localized around the notch as the severity of the notch increased.

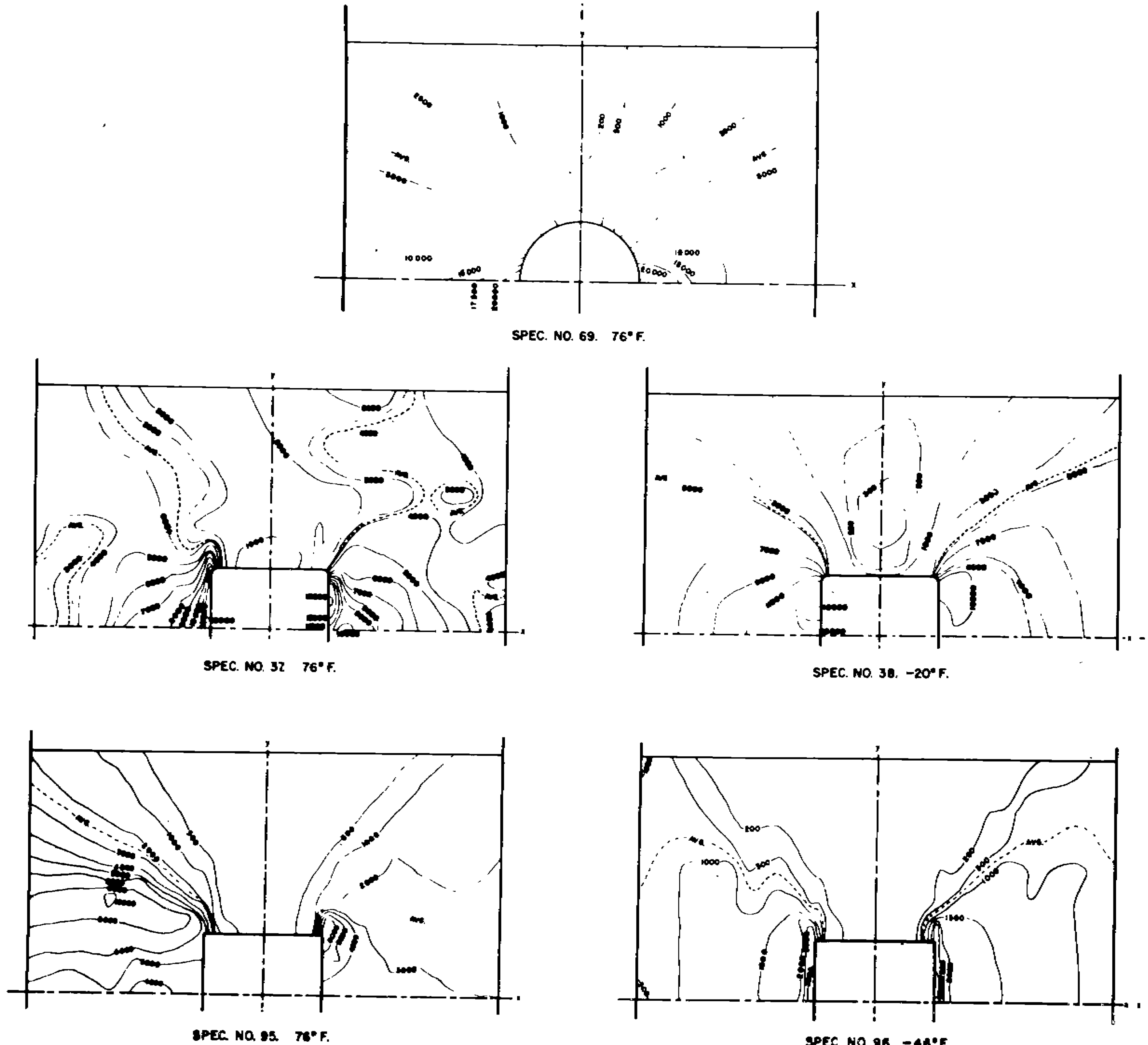


FIG 8. UNIT STRAIN ENERGY CONTOURS AT ULTIMATE LOAD FOR UNREINFORCED PLATES AS DETERMINED FROM MEASURED STRAINS.

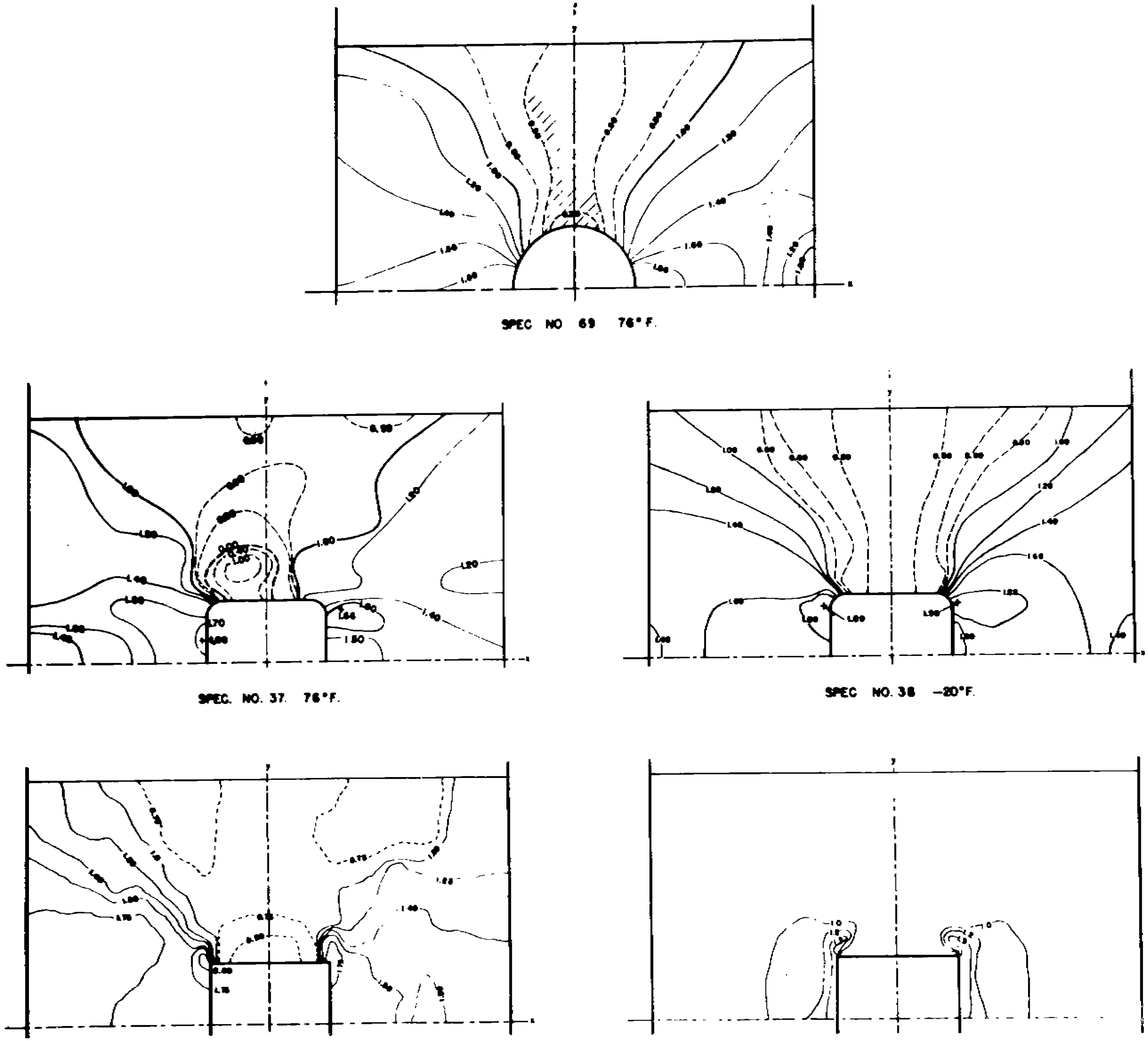
2. For identical plates tested at two temperatures, the first of which produced shear fracture and the second cleavage fracture, the higher values of the unit energy absorption were more localized around the notch for the cleavage fracture than for the shear fracture.

Similar statements could be made about the plastic stress distribution at the ultimate load as shown in Figure 3. The values on the stress concentration contours are the ratio of the true stress at the particular point to the average stress on the gross section of the plate in the region remote from the opening. Increasing the severity of the notch or changing the mode of fracture from shear to cleavage tended to localize the higher values of the stress to a greater degree around the notch.

This investigation indicated that two identical specimens, one of which finally sustained a shear fracture and the other a cleavage fracture, both specimens having the same elastic stress distribution, did not develop in the plastic range the same stress and unit energy distribution. This difference in behavior increased as the ultimate load was approached. Cleavage fracture was accompanied by a less efficient stress and energy distribution. Moreover, this difference in plastic deformation for the two modes of fracture precludes the use of tests resulting in shear fractures to predict the probable behavior of identical elements at temperatures which would produce cleavage fracture.

Some data are plotted in Figure 4 which relate to the initiation of fracture. The plots on the left side of this figure include the specimens with either a circular opening or a square opening with a 1-1/8-inch corner radius. All these specimens ultimately sustained a shear fracture or a ductile cleavage fracture. The data fell in two groups, one for the reinforced and one for the unreinforced specimens. However, when the notch became sharp as in the case of the unreinforced plates with a square opening with a 1/32-inch corner radius, shear fracture and brittle cleavage fracture were the modes of fracture at the two temperatures, and the plotted points were segregated according to the testing temperature as shown in the diagrams on the right. Thus again, the line of demarcation between brittle cleavage fracture on one hand and ductile cleavage fracture and shear fracture on the other hand is evidenced.

It may be seen in Figure 4 that the effect upon the value of the stress concentration factor of increasing the plastic stress level up to the ultimate strength of the plate was a tendency for the stresses to approach uniformity, and the plastic stress concentration factor to decrease toward a constant



SPEC NO 69 76°F.

SPEC. NO. 37. 76°F.

SPEC NO. 38 -20°F.

SPEC. NO. 95. 76°F

SPEC. NO. 96 -46°F.

FIG 3. PLASTIC STRESS-CONCENTRATION CONTOURS IN y DIRECTION FOR UNREINFORCED PLATES AS DETERMINED FROM MEASURED STRAINS.

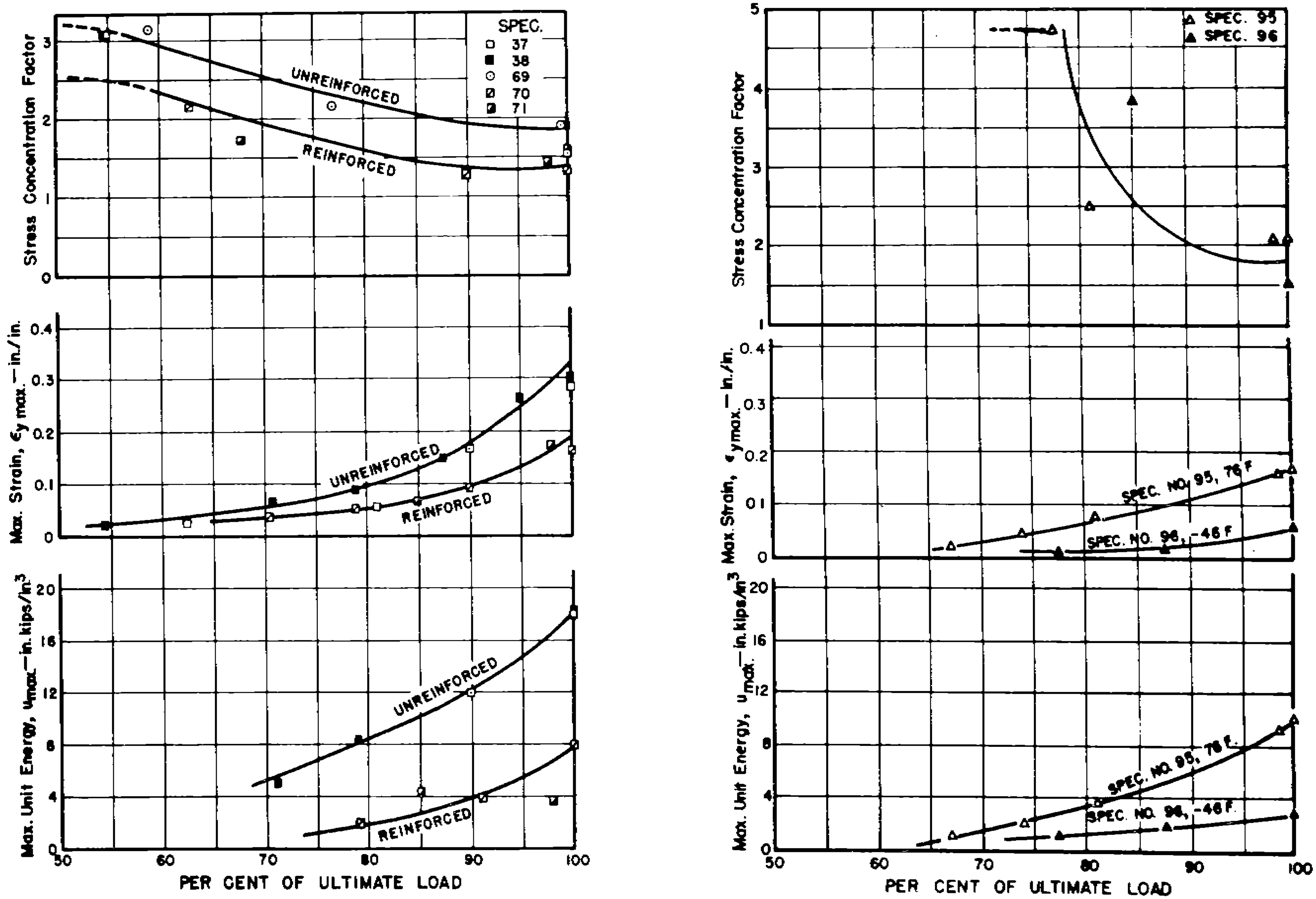


FIG. 4. PLASTIC STRESS CONCENTRATION, MAXIMUM UNIT STRAIN AND MAXIMUM UNIT STRAIN ENERGY AS ULTIMATE LOAD WAS APPROACHED.

and also minimum value. This finding suggests that the low-energy cleavage fracture of some welded ships, which is often accompanied by low ultimate strength, may result in part because the amount of plastic flow which has occurred is not large enough to bring about a sufficient reduction in the plastic stress concentration factor.

The wide variation in Figure 4 in the maximum observed values of the unit strain and the unit energy absorption, and an equally wide variation found in the maximum observed true stresses, which ranged from 69 to 105 ksi, suggests that no simple theory of failure would apply in the case of these tests. While inability to find and observe the absolute maximum values in these tests must account for some of the variation, it is felt that the previous statement is substantially correct.

One brief general comment should be made about the research in the field of plasticity in which Dr. D. Vasarhelyi and the speaker have been engaged. It is admittedly crude in character. The application of the methods of analysis is tedious, sometimes difficult, and not always precise. However, the work has substantiated the conjectures of some previous investigators and given quantitative results which lead to what we hope are significant conclusions.

V. SUMMARY

Let us summarize some of the principal points of this paper as they apply to the design of welded details:

1. Increasing the cross-section area of the reinforcement of an opening in a plate specimen brings about increased ultimate strength and energy absorption only when the width of the reinforcement in the third direction is kept small.

2. The amount of rigidity that can be built into a welded detail must be decreased as the notch severity of the detail increases.

3. The best design incorporates a minimum of rigidity so that "plastic flow may occur naturally and easily," to use Mr. DeGarmo's words.

4. The openings, sharp radii, and discontinuities present in welded details prevent them from absorbing more than a small fraction of the potential energy-absorbing capacities of the steel.

5. It seems likely that we can attain the ductile cleavage type of fracture in our ships' structures by adequately reducing the notch severity inherent in our designs. A structure designed in this manner would be very resistant to the formation and extension of a crack and perhaps could therefore never be brought to failure.

6. The problem of improving the ductility of our welded hulls is not one of making the present plastic-strain-absorbing capacity of our structures tremendously greater, but only a little greater--of increasing the strain-absorbing capacity of some of our bad details from one per cent to about five per cent.

7. If structural steel were not prone to brittle fracture, undoubtedly less than five per cent overall ductility would be sufficient.

8. Cleavage fracture in the plates with openings is accompanied by a less efficient stress and energy distribution than shear fracture.

9. This difference in plastic deformation for the two modes of fracture precludes the use of tests resulting in shear fractures to predict the probable behavior of identical elements at temperatures which would produce cleavage fracture.

10. The low-energy, low-strength cleavage fracture of some welded ships may result in part because the amount of plastic flow which has occurred is not large enough to bring about a sufficient reduction in the plastic stress concentration factor.

11. No simple theory of failure appeared applicable to the data of the tests of plates with openings.

The speaker would like to comment on one more point before closing. We must be very careful that the apparent complexities of the results of our research, old habits of thinking which cannot solve the present problem, the many proposed theories, and the many voices of explanation which are raised, including that of the present speaker, do not confuse the investigator into thinking that the causes of brittle fracture are complex. The problem is undoubtedly susceptible to a simple explanation.

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BRITTLE FRACTURE MECHANICS, AS REVEALED BY TESTS OF LARGE STRUCTURES

by

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The series of tests of 36 welded hatch-corner type specimens which was conducted at the University of California from 1944 to 1947 constituted one of the most extensive sets of laboratory tests ever made on large size weldments using full-scale plate thickness and welded details. These tests, therefore, are particularly significant in connection with brittle fracture mechanics, since they afforded an opportunity to study such fractures on real structures which contained those characteristics, such as extreme rigidity, structural discontinuities, and welding defects, which frequently are present in real welded structures, some of which have failed by brittle fractures.

Since these tests grew out of the failure of some welded ships, the basic specimen used was similar to a corner of a hatch from the early Liberty ships. The details of this specimen are shown in Figure 1 and are familiar to most of you. The original objective in using this type specimen was to have one which would be full-scale and would fail with a brittle fracture. Parenthetically, it will seem amazing to many people that in 1944 there was considerable doubt in many quarters whether a laboratory specimen could be developed which would produce the same type of brittle fracture that had been experienced in ships. As will be noted in Figure 1, the primary characteristics of the specimen are: (a) a sharp structural discontinuity in each of the main members, (b) extreme rigidity, (c) some bad welding details resulting in voids in critical areas, and (d) plate thicknesses up to 3/4 inch.

Subsequent events proved the choice of an existing hatch corner as the model for the laboratory specimen to be a fortuitous one since it was possible to extend the scope of the tests to include some hatch corner design studies, and some very significant correlations of the results of these tests and actual ship experience have since been made by other investigators (1). These correlations, shown in Table 1, verify the possibility of predicting large structure performance by much smaller laboratory tests. Honesty forces me to the admission that the project investigators were more than slightly influenced in their decision to use the hatch-corner type

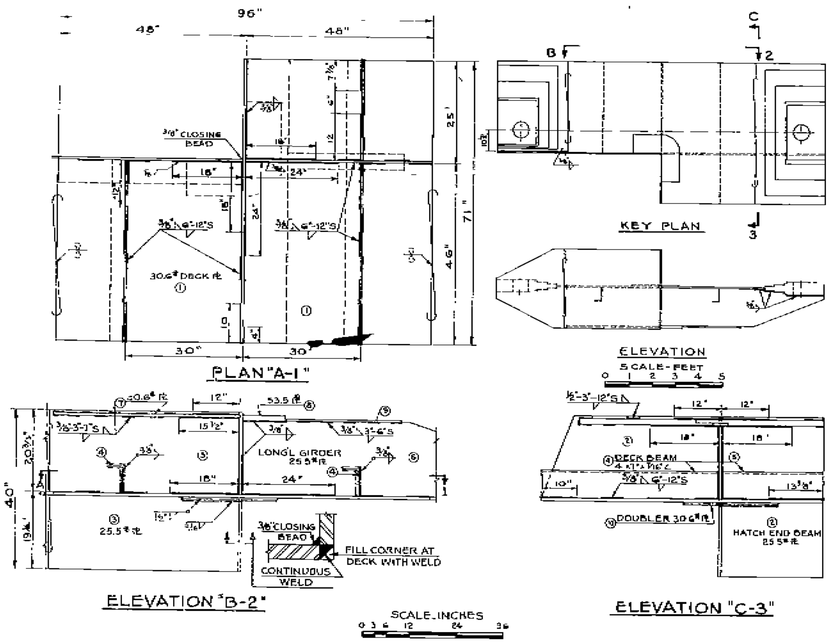

Figure 1. Details of Hatch-Corner Type Specimen

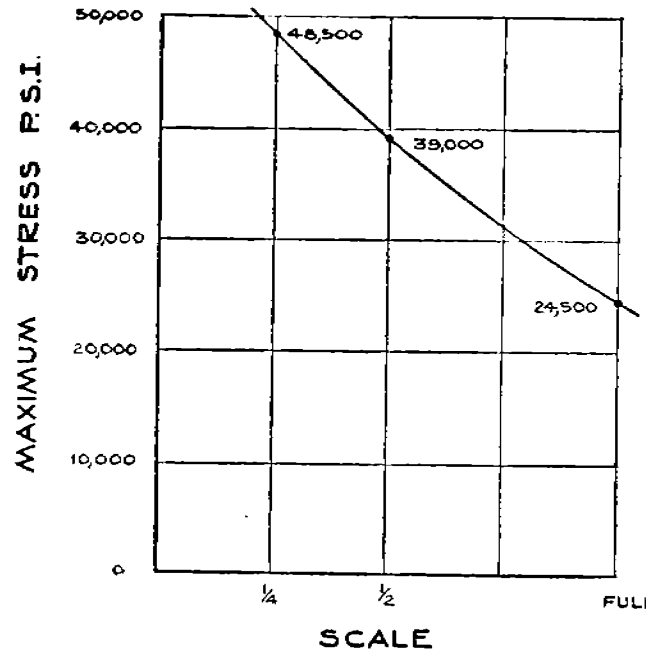

Figure 2. Relationship between Size and Breaking Stress--Hatch-Corner Type Specimens

TABLE 1

RESULTS OF LABORATORY TESTS AND ANALYSIS OF SHIP SERVICE RECORDS (APPROXIMATELY TO END OF 1950) OF VARIOUS HATCH CORNER DESIGNS

	Record of Structural Performance in Service			Data from Laboratory Tests (*)	
Type	Ship Years in Service	Fractures Reported at Hatch Corners	Hatch Corner Fractures per 100 Ship-Years	Energy Absorption at Failure, In-lb.	Nominal Stress at Failure Pounds/sq. in.
Liberty Ship with Square Hatch Corner	2,110	224	10.60	230,000	24,000
Liberty Ship as Altered with Rounded Bracket	4,400	31	0.70	921,000	31,450
Liberty Ship with Rounded Deck Plate and Doubler	3,750	1	0.03	3,627,000	35,500
Victory Type	2,100	0	0.00	5,800,000	33,200
Kennedy Type	-----	---	-----	6,786,000	54,100 (**)

(*) Laboratory specimens varied somewhat from the actual ship detail to permit practical fabrication and testing procedure.

(**) Failure started from a notch resulting from an arc struck inadvertently where no weld was intended.

specimen by the hope that some design studies could eventually be undertaken, but we did not expect that such excellent correlation with actual ship experience could, or would, ever be obtained.

I would now like to present some of the results of the tests which I believe are significant to the subject of brittle fracture mechanics and simultaneously point out certain problems which I feel are not yet completely answered. I shall present here only fragmentary results, since the complete details are voluminous and can be found in the reports listed as items 2--6, inclusive, in the appended Bibliography.

The first result, which I want to dispose of quickly since it will not appear to have a direct bearing on brittle fracture, relates to the very considerable effect of size upon breaking stress. This effect is shown in Figure 2. The full-scale specimens, made of "C" steel welded in the ordinary manner, had a nominal breaking stress of 24,200 psi. It will be noted that the strength was inversely proportional to size, giving a breaking stress of 48,500 psi for the quarter-scale specimen. The curve of breaking stress vs. size is such that one wag remarked that if we had built a specimen to twice full size we could have obtained a brittle failure without having to put the specimen in a testing machine. This remark is probably of more significance than was intended in the light of certain known failures which have occurred in structures having plate thicknesses above 1 1/2 inches. This apparent size effect, coupled with these recent failures, makes it appear that in very thick plate welded structures a slight defect may bring about brittle fracture at very low working stresses. The serious question that follows from this situation is, "What is the designer of a large, rigid, welded structure going to use as the design stress when all he has available are data from the usual 0.505 physical tests?"

A second significant result from the hatch corner specimen tests was that with a given notch-sensitive steel, welded into a rigid structure having severe structural discontinuities, the service temperature and type of electrode used had relatively little effect upon the breaking strength when the structure was tested in the as-welded condition. For example the use of 25-20 electrode increased the strength only 15%. This implies that the stress at which a given structure in the as-welded condition will fail with a brittle fracture will not be altered materially by service temperature or by the electrode used. Thus improved performance would have to be sought through better steels, thermal treatment or improved design. It must be pointed out that no low-hydrogen electrodes were used in these tests and information about their use in this respect would be desirable.

Both high-temperature, post-welding stress relieving at 1,000°F. and 400°F. preheating on all passes were found to provide substantial improvement in strength and ductility but did not change the fracture type. Preheating at 400°F. was somewhat more effective than the stress-relief heat treatment and was more effective than any other procedure tried except design change. While the increase in breaking stress was about 25% and 33%, for high temperature stress-relief and preheating, respectively, the improvement in energy absorption was much greater, being over 500% in the case of preheating. These results are particularly significant in that they indicate feasible procedures which can greatly improve the performance of welded structures even though they do not change the mode of fracture from brittle to ductile. Furthermore, where the size or location of a structure does not make post-welding stress-relief heat treatment practicable, 400°F. preheating is a good substitute. Numerous other and subsequent tests(7) have verified the acceptability of substituting preheating for post-heating in many conditions.

One of the most significant conclusions to be drawn from the tests of the preheated and post-heated specimens was that residual welding stresses are not a significant factor in brittle fractures. It has been established that 400°F. preheat does not reduce the residual stresses to any great extent, yet the preheated specimens out-performed the stress-relieved specimen. I have yet to find any case of brittle fracture in a welded structure where the facts have been conclusive that residual stresses were a major cause of the failure. I believe the statement contained in the Final Report(8) of the Board of Investigation to Inquire into the Design and Methods of Construction of Welded Steel Merchant Vessels, issued in 1946, which said, "...no evidence has been found to indicate that these stresses are important in causing the fractures in welded ships," is just as valid today. However, there are many responsible engineers who are not convinced of this and still believe that the problem of brittle fracture is closely connected with residual stresses. In the last few weeks I received a communication from such an engineer which contained the following statement: "This is of considerable importance in such structures that are exposed to low temperatures and where the presence of residual stresses may cause a sudden rupture."

In connection with preheating, it would be helpful to know more about its effectiveness on less notch-sensitive steels and on structures fabricated with low-hydrogen electrodes. Some such information has recently become available but not from tests of large structures.

Certainly one result of the hatch corner tests was a conviction that more notch-resistant steels are desirable in critical areas of structures. As will be mentioned later, it is of greatest importance that brittle fractures never get started in large welded structures. However, we should remember that the steel must continue to show notch resistance <u>when combined with a weld</u>. If the steel plus the weld and weld-affected zone are not notch resistant, we may achieve just as good or better results with a poorer steel and suitable auxiliary procedures such as preheating or post-heating. For example, one specimen was made of a steel having 3.34% nickel. This steel was undoubtedly more notch-resistant than the "C" steel. While the nickel steel specimen had greater strength than the preheated specimens, (44,400 psi vs. 32,700 psi), the preheated specimen absorbed over 23% more energy.

Again, while the use of a more notch-resistant steel would undoubtedly decrease the effects, the extreme importance of notches was brought out by the hatch corner tests. The use of full-penetration welds instead of fillet welds, thus eliminating a marked void at the intersection of the three main members, resulted in a strength increase of 24% and gave energy absorption amounting to 280% of that of the fillet welded specimen.

Another serious problem brought out by the hatch corner tests is that of predicting, through the use of small laboratory tests, what the transition temperature of a large structure constructed of a particular steel will be. For example, the difference between the transition temperature indicated by Charpy tests and the transition temperature of the full-scale hatch corner specimens was approximately 95°F for "C" steel. However, for a second steel, "B", it was about 60°F. Many tests have shown that there are numerous small specimens which will <u>rank</u> steels as to their relative transition temperatures. However, they do not give the value the designer needs, namely, the transition temperature to be expected in the full-scale, as-welded structure. The Kinzel test is probably the most satisfactory small-size test now available, but it too leaves much to be desired in this respect.

Two facts of brittle fracture mechanics as demonstrated in the hatch corner tests point out the importance of preventing such fractures from initiating. One is that such fractures propagate at speeds up to 5,000 feet per second. Thus there is no opportunity to take remedial measures as often is done in slowly-progressing cracks in riveted structures. The second fact is that these rapidly traveling brittle fractures appear to have a dynamic effect which causes them to spread through even very heavy members which are apparently unstressed by the live loading of the structure. This was demonstrated by the fracture of a restraining bar in two of the hatch corner tests, one of which is shown in Figure 3. The heavy 3-inch by 3-inch bar was fractured by reason of its being connected to the 3/4-inch plate by a small 1/8-inch fillet weld although analysis indicated that it carried little, if any, stress. Similar situations have been found on ships which have failed. Thus one must conclude that nearly every brittle fracture in a large welded structure is a serious matter because of the speed and extent of the failure which results.

While design is not directly a factor in brittle fracture mechanics, it cannot be divorced from the subject. Undoubtedly, the most important result of the hatch corner tests was the demonstration of the tremendous effect design could have upon the strength and ductility of welded structures, even though none of the design changes investigated caused a change in the mode of fracture from brittle to ductile. Thus it appears valid to conclude that below a certain temperature a structure welded from a particular steel will fail with a brittle fracture regardless of design modifications, <u>if a fracture starts</u>. However, design can probably do as much as, if not more than, any other single thing to prevent a fracture being initiated under a given set of load conditions. For example, in the case of the hatch corner specimens, the simple expedient of adding a triangular extension to the longitudinal hatch coaming increased the strength by 34% and the energy absorption by 1645%. More extensive design changes resulted in specimens which had over twice as much strength and absorbed 25 to 30 times as much energy as the original specimen did.

While it certainly is desirable to have more notch-resistant steels, these alone are not the answer to the brittle fracture problem. In fact, the existence of such materials may lead poor designers to produce bigger and better failures through a sense of false security. In conclusion, it is interesting to speculate just a bit as to how good a hatch-corner type specimen could have been produced by utilizing a combination of a slightly better steel, such as steel "B", 400°F preheat, and the Victory ship design. If one assumes that the better design would again produce a 38% improvement in strength and increase

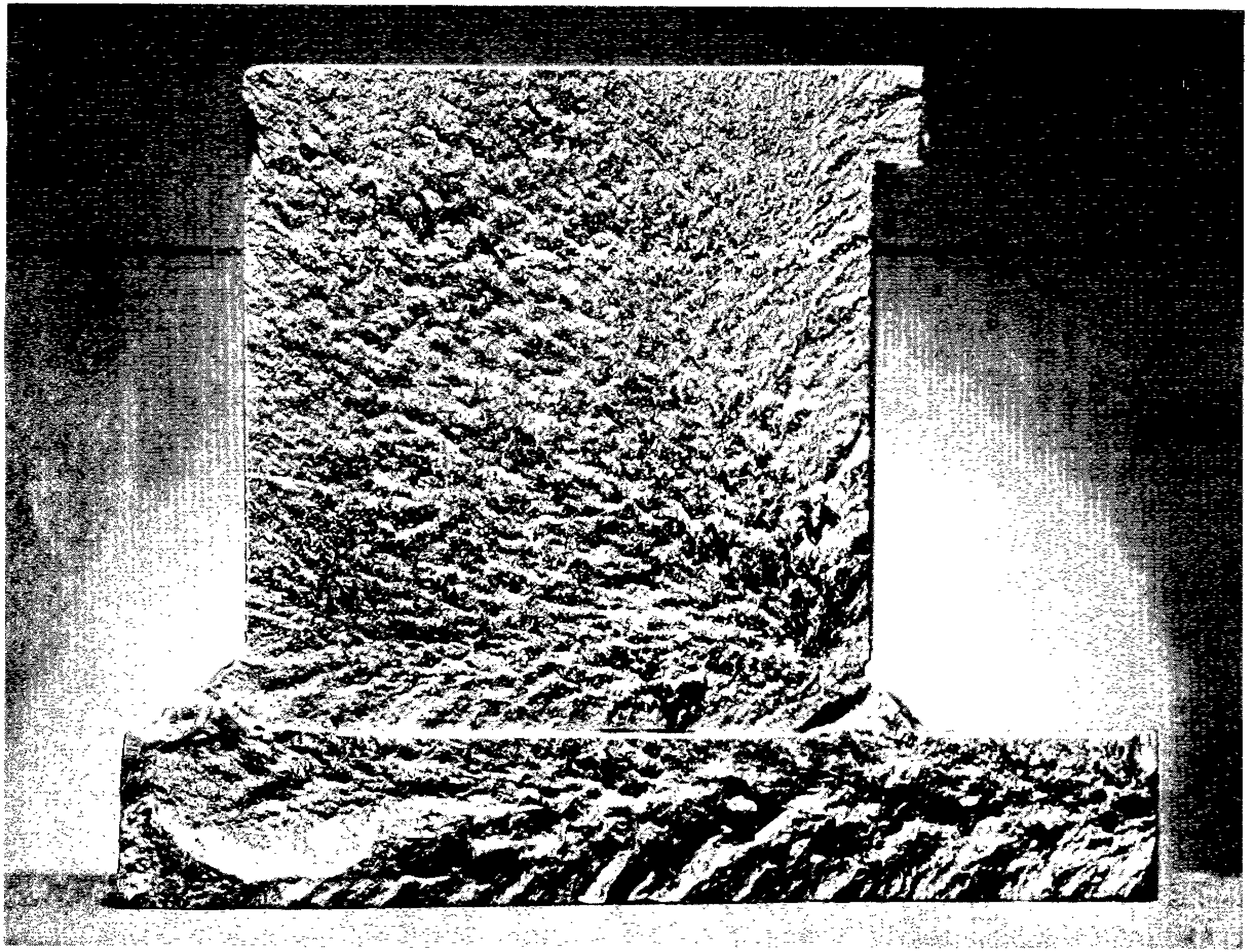

Figure 3
3" x 3" Restraining Bar Broken by a Dynamic Effect of Brittle Fracture

the energy absorption by a factor of 25, starting with a stress of 32,400 psi which was obtained with a preheated "B" steel specimen and an energy absorption of 340,000 in-lb which was obtained from a non-preheated "B" steel specimen, the result would be a breaking stress of 44,600 psi and energy absorption of 8,500,000 in-lb. When this is compared with the corresponding values of 24,000 psi and 230,000 in-lb for the basic hatch-corner type specimen, it is apparent that proper use of the information which we now have could do much to avoid future failures. Thus while there is need for more information regarding certain aspects of brittle fracture, it appears that with the information now available there is little reason why welding should not be used freely nor why this type of failure should be expected or condoned in the future if designers and fabricators will make use of the information which is now known.

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FRACTURE DYNAMICS AND FRACTURE STRENGTH OF LARGE WELDED STRUCTURES

by

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The first topic which will be discussed is the interrelatedness of size effect, of time of load application, and of unstable fracturing relative to strain energy release.

A simple example is the failure of a liquid column under negative pressure. Dr. John Fisher's theoretical description of this process is well known, and I will only sketch it briefly. We imagine the liquid contained in a rigid tube closed at the top, completely filled, and fitted with a piston in its lower regions for our convenience in applying a strain. We conduct the tensile strength experiment by applying a load to the piston, then fixing its position while we wait a certain length of time for the liquid to develop a hole of critical size. If nothing happens, we move to higher load. We repeat the process until failure occurs. When the experiment is done in this way, the chances are that our most careful looking in the time period within which the tension in the column suddenly drops does not permit observation either of the critical size hole or of the bubble which developed from it. To estimate the critical bubble size, we may write as the energy of bubble formation

$$\Delta E = A\sigma - vT \quad - - - - \qquad (1)$$

In the first term $\underline{A}$ is the surface area of the bubble, and σ is the surface tension of the liquid. In the second term $\underline{v}$ is the bubble volume, and $\underline{T}$ is the negative pressure or hydrostatic tensions. The second term is the loss of strain energy in the whole volume tested because of the release of negative pressure. When the area of the bubble increases by the amount $\underline{dA}$, the surface tension energy increases by σdA, and the increment of strain energy release is $\frac{rT\,dA}{2}$. $\underline{E}$ increases until $\underline{r}$ equals $2\sigma/T$. Thereafter strain energy release pulls the bubble open to about a thousand times its critical size. Even this final bubble would be scarcely visible since the critical size is only about 10^{-7} cm. Besides virtual invisibility another difficulty with this experiment lies in the fact that there is no mechanism other than local temperature fluxuation by means of which the energy to form the critical size bubble is furnished. This is a matter of probabilities. For similar probabilities of failure the required tension is less, the longer the time under tension and the larger the volume. By putting values for CCl_4 into a rate theory formula, one finds that a doubling of the time or of the tested volume should decrease the observed fracture strength by about 2 per cent. This is qualitatively similar to the relations between test volume, time, and strength observed in metals. The magnitude of the loss of strength with doubling of the volume actually turns out to be comparable in a quantitative sense for several sets of experimental results on steels.

The fracture of a pure chemical compound in its liquid state as discussed above is probably the easiest example of fracturing from the viewpoint of understanding the event in fundamental terms. The Griffith crack theory of fracturing as applied to glass is nearly as good. However, glass must be imagined to contain flaws, and the strength probability calculation must make assumptions about the unknown flaws. The fracture instability relation is quite similar to that of the liquid example. There are important effects of humidity on the growth rate of crack flaws in glass. When these are taken into account, it is clear that time and test piece size influence fracture strength in ways qualitatively similar to those indicated above.

Fracturing of metals is still more complicated. There are inherent pre-existing flaws as is the case with glass. In addition so much plastic deformation accompanies fracturing that the surface energy concept must be replaced with something more appropriate in order to represent the conditions for instability of fracturing relative to strain energy release. None of these modifications are such as would be expected to eliminate the size and time effects which accompany the probability of fracture instability in the simple models discussed above.

Figure 1 shows the results of some trials of an effect of time upon the progress of an already started fracture in mild steel. These trials used testing machine loading of notched bars from a 3/4-inch steel plate of ship steel type. The bars were nearly 3/4-inch square rods notched on all four sides. The notches removed 50% of the area and had sufficient depth and acuity so as to eliminate the lag of the fracture normally present at the sides of the

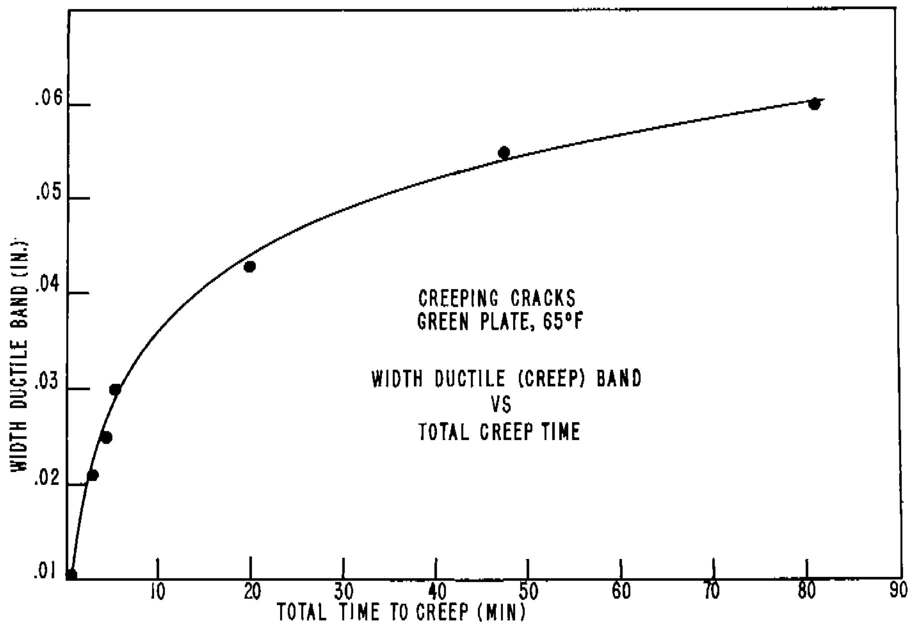

Fig. 1 WIDTH DUCTILE (CREEP) BAND VS TOTAL CREEP TIME-GREEN PLATE, 65°F Univ. of N. Carolina Report No. 39. NRL Contract N6ori-227, 1948.

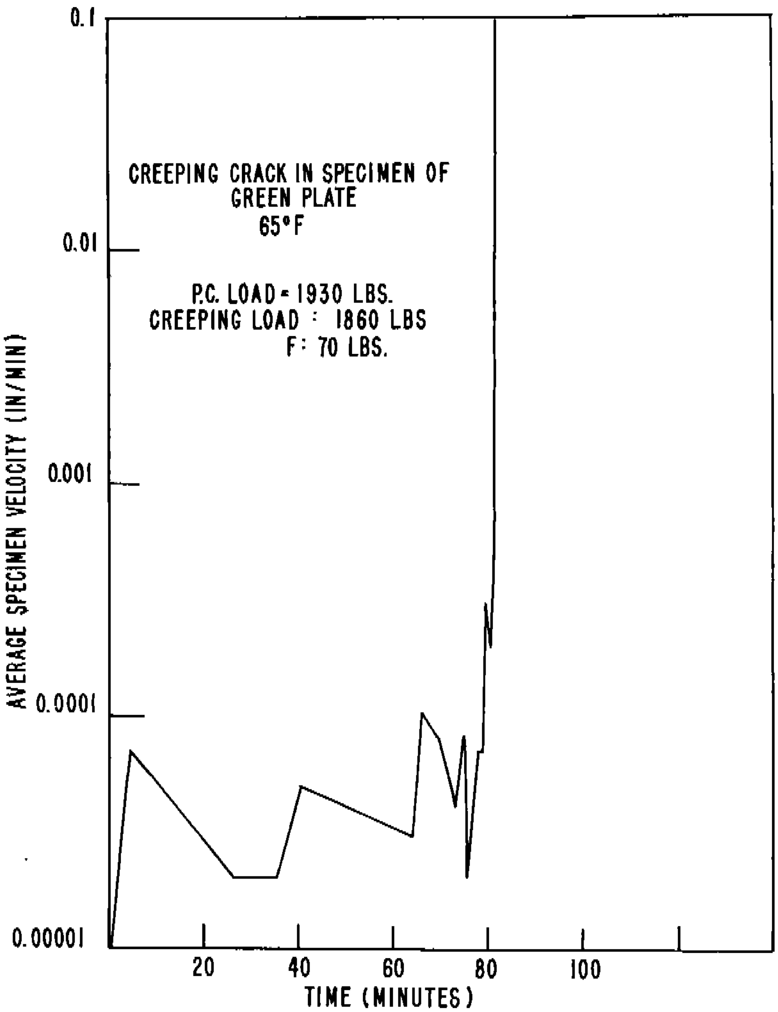

Fig. 2. CROSSHEAD SPEED FOR A CREEPING CRACK IN SPECIMEN OF GREEN PLATE VS TIME-65°F Univ. of N. Carolina Report No. 39. NRL Contract N6ori-227, 1948.

specimen. The specimens were bent so that a well started crack was causing a drop-off of the load. At a predetermined load the specimen was partly unloaded; a stain was applied to the crack to indicate its depth, and the load was reset to several per cent less than the value just before unloading. By continual adjustment of the speed of the machine, the load was maintained very nearly constant until either the speed of the machine was not sufficient to compensate for load relaxation by advance of the crack or unstable fast fracturing occurred. For the results shown in Figure 1, the end point of each test was a fast cleavage fracture. Figure 2 shows the very large change of velocity of fracturing as indicated by the speed of the crosshead of the machine. Trials of this nature on the same steel at a higher temperature showed essentially similar results. However, at this higher temperature maximum crosshead speed failed to maintain the load, and a constant load approach to unstable fracturing could not be demonstrated.

In connection with understanding brittle fractures of large structures, it appears desirable to know how a small crack may grow in size to a length sufficient for instability all in a field of tensile stress much below the yield stress of the material. The load differential of only a few per cent as used above could be increased several fold by allowing days rather than hours of time for the experiment. Much larger reductions of the load for advancing the crack might have been obtained by repeating the unloading and reloading of the specimen many times during the period of the experiment. A trial of this was not made. However, the effect of stress cycling is commonly assumed to be much greater than that of time under steady load. In fact the damaging effects of fatigue for a fixed load-strain cycle are usually considered to be independent of the frequency employed. One may note from Figure 1, as well as from Figure 2, that the speed of the creeping advance is greater, the closer the crack is to its point of instability. We have made studies of slow growth of cracks under constant load in various fracture model experiments and have found that the fractional increase of crack length per unit of time is an increasing function of the crack length throughout its whole time history except for temporary fluxuations. These growth rate fluxuations constitute a descriptive feature of fracturing which was always noted in the University of North Carolina experiments as well as those at NRL. Both the speed of growth and the fluxuations are related to the size effect in a very fundamental way because each depends upon the chance number and arrangement of local weaknesses participating at any given time in the motions which advance the crack length.

For a structural metal one cannot express the relations between the effects of test volume, time, and load so well as can be done for fracturing of a pure liquid. However, the relations can be outlined in general terms. The size effect in the liquid fracture was based upon probability of certain local energy fluxuations. In engineering metals this is replaced by probability of certain local weaknesses or flaws. The direction of the effect of increasing the test volume size is to make the material appear weaker. Under tensile stress the flaws develop into little openings or cracks. Among these, the larger ones tend to grow faster because they are overstressing larger volumes of material. From this same consideration even their fractional time rate of growth on the average must be larger. Because of the uncertainty of local weaknesses, the growth rates are variable in time. In the most strained regions ahead of an advancing crack, new cracks are always forming, so that progressive fracturing is just the growth and joining of new fracture origins. As the dominant crack extends in length, the zone which can contain new fracture origins also extends. The increasing probability of bad flaws in this zone, as indicated above, means that the dominant crack not only grows more rapidly but also with ever diminishing amounts of plastic deformation in a relative or scale model sense. Since the strain energy release tends to maintain its magnitude relative to the crack dimensions, there is the possibility of sufficient strain energy release per unit of fracturing to provide all of the associated deformation work. The danger of unstable fast fracturing from this cause is just a matter of time and of having a large enough test structure. A welded ship appears to satisfy these conditions on an experimental basis.

In order to estimate the magnitude of the danger and ways of diminishing it, it is helpful to consider whether we are to regard all of the growth of the crack to unstable size as observable or as unobservable. In the latter event the problem is in the general class of fatigue failure. Estimates of danger and control procedures can be determined as sufficient statistical information becomes available. However, a more reassuring degree of safety can be planned for if we can provide observability of the crack prior to the unstable size. This situation is assumed to be a reasonable possibility in the balance of this discussion. Actually, efforts to improve fracture strength of large structures would have much in common regardless of our decision on this observability feature.

The general idea of our suggestion on the control of observable growing cracks is very simple. It is merely that one should select materials and control the design so that the unstable length of each growing crack is longer than some predetermined amount which we decide may be regarded as observable. In order to do this, it is necessary to know what work rate values can be counted on to accompany extension of a crack under various conditions. It is also necessary to have a procedure for calculating the strain energy release rate which accompanies crack extension.

Figure 3 shows information of the above type for the University of North Carolina notched bend tests mentioned above. As already stated, these 3/4-inch square bars of ship plate type steel were notched on all four sides. The crack maintained nearly a straight front as it was moved through the piece by the bending load. Successive depths of fracture were marked by staining. From the load records and from the unloading curves at the time of staining, work values and strain energy values were computed and plotted against the depth of the crack. The slopes of these curves furnished work per unit area of fracture and strain energy release per unit area of fracture. Figure 3 is a typical sample of the information obtained. The strain energy release includes the strain energy in the testing machine. Dropping out the portion of strain energy release contributed by the machine would lower the strain energy release rate curve by 40 per cent at the left of the figure and by lesser amounts as the fracture area increases. The values of strain energy release rate shown here are the amounts of driving energy per unit area of fracture available for self-propagation of the crack within a structure of which the testing machine is a component part.

Also shown in Figure 3 are values of the measured work per unit area. The strain energy was subtracted in computing these. These work rates are referred to as dW/dA and the strain energy release rates as dE/dA. When dW/dA is greater than dE/dA, then the motion of the crack is being assisted by the forces which turn the screws of the testing machine. Only when dW/dA is equal to or less than dE/dA, can the structure be considered to be breaking itself by conversion of stored elastic energy. A necessary condition for self-fracturing is

$$\frac{dW}{dy} \leq \frac{dE}{dy}$$

where <u>y</u> is <u>A</u> or some other parameter of the experiment which increases regularly with A.

It is at first disconcerting to find that an average curve through the dW/dA values lies so far above the dE/dA curve. However, three factors must be considered together in this connection. First, the work rate values are average values for a substantial increment of crack depth. Second, the quantity averaged is known to have large fluxuations. These were present in all of our work rate measurements using centrally notched thin sheets. They are a natural consequence of the unequal local weaknesses of the material mentioned previously. The third point is that fractures of mild steel under various combinations of fracture velocity, constraint, and temperature tend to change from ductile to cleavage fracturing. Thus regardless of the magnitude of average ductile fracture work rate values, one cannot count on larger effective work rates than those indicated by the values of dE/dA at the point of instability. The instability events which occurred at from 20 per cent to 50 per cent of the total fracture area showed effective dW/dA values of 300 to 400 in-lb per sq. in. for a typical steel of ship plate type and for a similar plate which had been aluminum killed. The observed quality difference between these steels was less in the effective dW/dA values for similar crack depths than in the testing temperatures required to produce unstable fracturing at similar depths. The aluminum killed steel at 10°F appeared similar in this regard to the ship steel at 80°F.

The approach of fracture extension to the unstable or self-fracturing condition can be visualized in terms of the information presented in Figure 4. The test piece represented in the upper right of this figure is a flat plate with an extending central crack of length, X. For such a test the strain energy release rate rises steadily with crack length, as shown in the lower left of the figure. The self-fracturing structure must include the testing machine and specimen gripping devices. However, if these components are very stiff compared to the test specimen, they will contribute very little strain energy during unstable fracturing, and the unstable fracturing process may be thought of as one in which the separation, ℓ, of the grips does not change. Thus the load-extension curve, as shown in the upper left of Figure 4 must become a straight down drop of the load, F, during onset of fracture instability. An instability of this kind can occur with the load-extension curve bending continuously over into a straight down drop, as illustrated by curve 1. In the NRL tests, fractures extending in thin sheets of ductile metals have shown this type of approach to instability. When the test piece is a plate of mild steel of

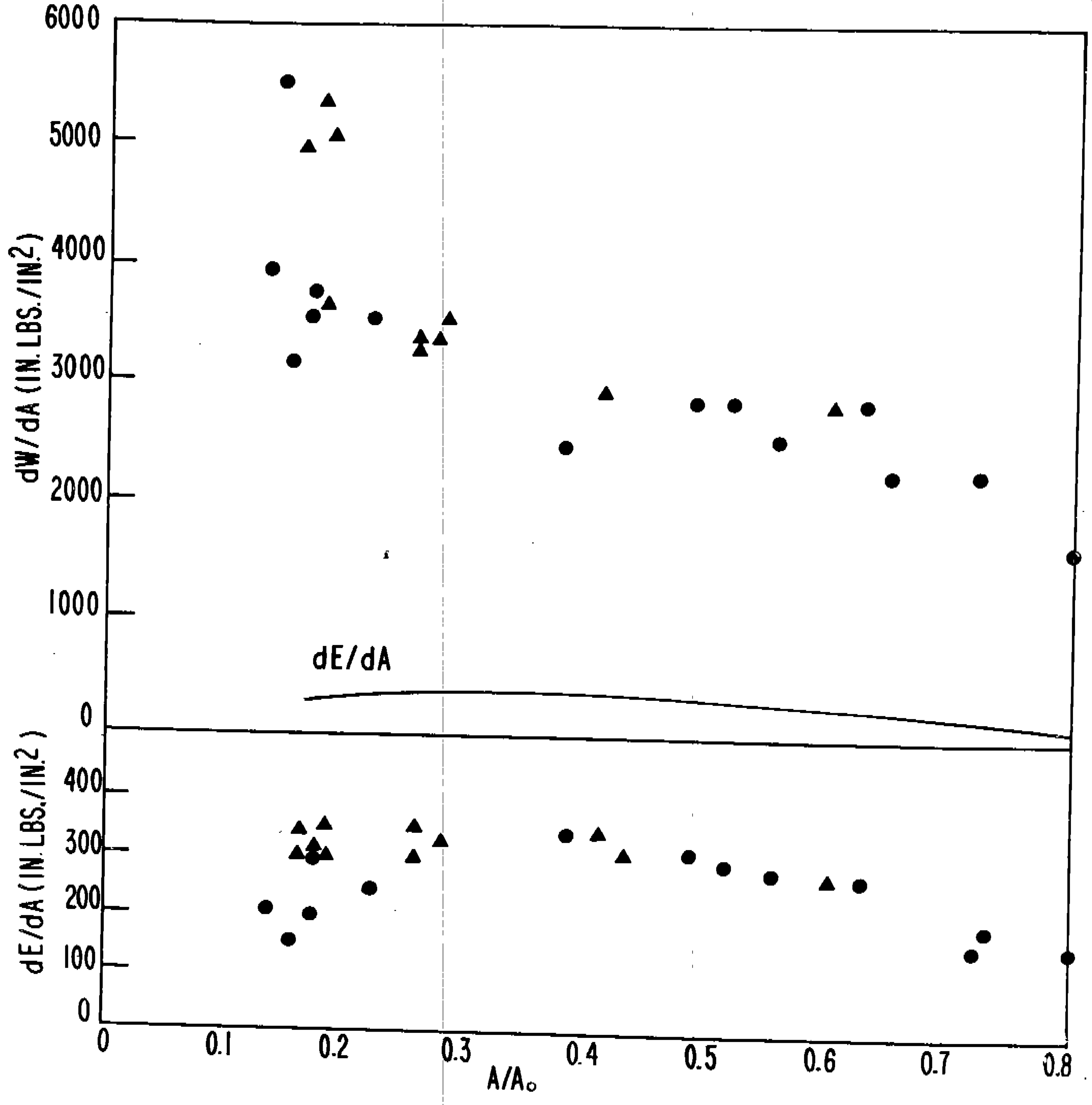

Fig. 3. Comparison of dW/dA and dE/dA vs A/A_o for 3/4" slow bend specimens. Color code-Green Univ. of N. Carolina Report No. 38. NRL Contract N6Ori-227, 1947.

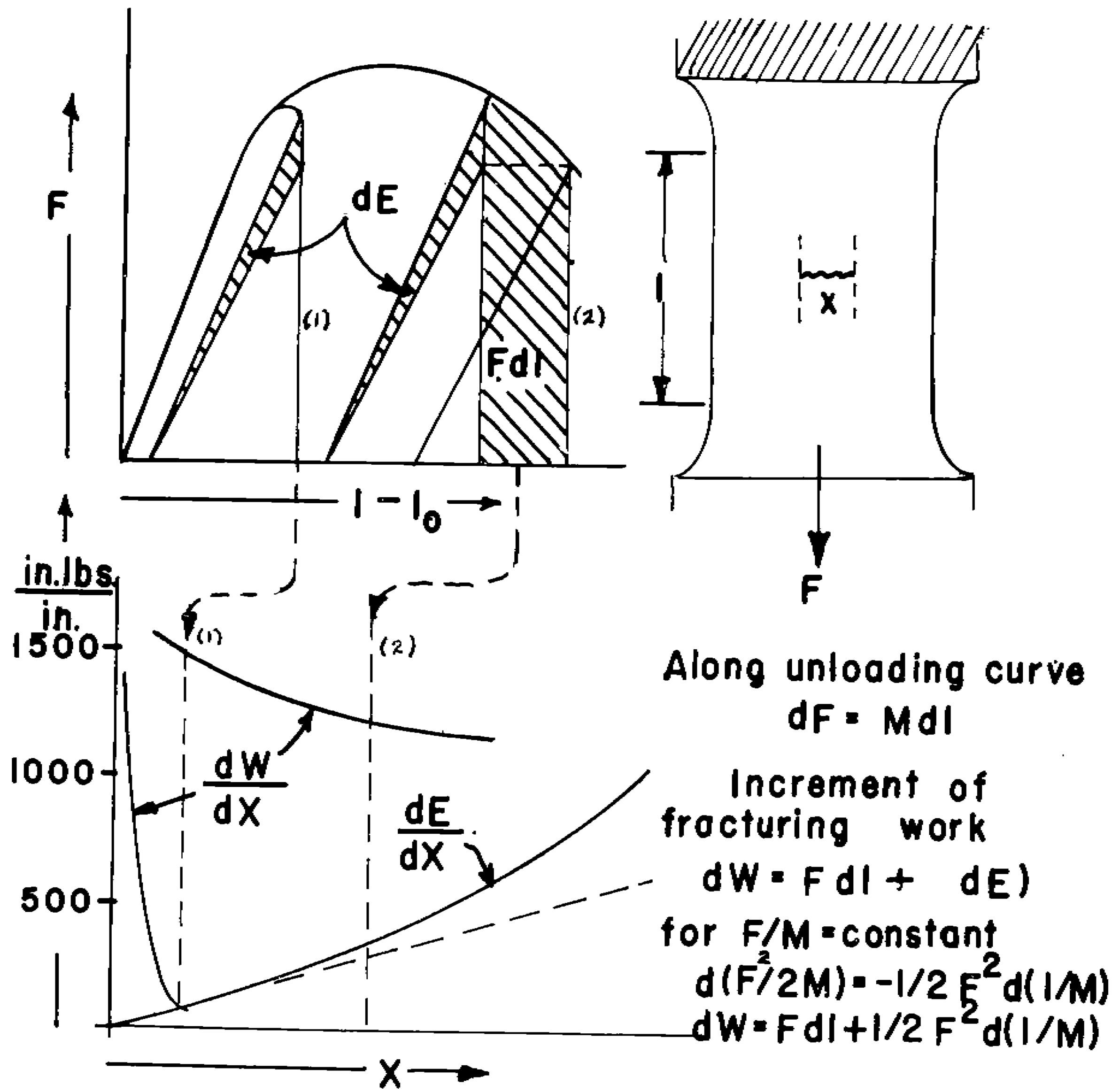

Figure 4. Schematic diagrams showing continuous and discontinuous types of onset of fracturing instability.

average ship steel quality and 1/2 inch or more in thickness, onset of self-fracturing may be quite abrupt with a discontinuity in slope of the load-extension curve, as shown by the second case illustrated in the schematic curves of Figure 4. Ample motivation for the abrupt change from predominantly ductile to semi-brittle cleavage fracturing may be found in the snapping of tough sections which always accompanies ductile fracturing of mild steel coupled with the sensitivity of that material to impact.

If extensive sections of cleavage fracturing of mild steel plate began and moved along with work rate values much less than the release rate of strain energy, then such instability events would be to a large extent capricious and unpredictable. It must be borne in mind that cleavage fractures ordinarily run with velocities less than half of the transverse sound velocity. The kinetic energy of directed particle motion associated with the changing pattern of strain in the test plate, in this velocity range, is about $(1/2)\ (C/C_2)^2$ times the strain energy release or less. (Here C is the crack velocity and C_2 is the transverse sound velocity). Both of these energies are proportional to the same power of the crack length. Thus the portion of the strain energy release rate which must be consumed by the fracturing work rate is much larger than the portion given to kinetic energy of directed particle motion. One concludes that the strain energy release rate and the fracturing work rate are not only equal at onset of instability but remain close in magnitude as fracturing continues. The pattern and appearance of the fracture which occurs in a service or a laboratory case of unstable cleavage fracturing tells the extent to which one may rely in that instance on this near equality of dE/dA and effective dW/dA values. If dW/dA becomes and continues substantially less than dE/dA, the crack velocity increases steadily; the fracture appearance changes; and repeated forking or shattering occurs. On the other hand if extensive lengths of cleavage fracturing are observed with little change of fracture appearance and with much less extensive branching than in shatter, then one concludes that a near equality of dE/dA and dW/dA existed all along except close to such points of forking or branching as did occur. In regions of cleavage fracturing such that dE/dA and dW/dA are most nearly equal, as at the start or halting of a segment of cleavage fracturing, calculation or experimental knowledge of dE/dA furnishes our best information of the effective fracturing work rate, dW/dA.

A useful equation for the strain energy release rate in terms of the load, F, and the specimen spring constant, M, is

$$\frac{dE}{dA} = \frac{1}{2} F^2 \frac{d}{dA}\left(\frac{1}{M}\right)$$

The calculations leading to this expression assume the screws of the loading machine are temporarily stopped and that the separation of the specimen grips which accompanies dA is limited to $d\ell'$ where $d\ell'$ is the incremental change of (F/M). On Figure 4 the expression for the inelastic work of crack extension, dW, which accompanies dA assumes, in addition, that $d\ell'$ is zero as appeared to be nearly true in the notched bar studies discussed above.

In general the rate of loss of energy from the machine-specimen system, dW/dA, is given by

$$\frac{dW}{dA} = F\left(\frac{d\ell}{dA} - \frac{d\ell'}{dA}\right) + \frac{1}{2} F^2 \frac{d}{dA}\left(\frac{1}{M}\right)$$

Where $d\ell$ is the actual grip separation increment. To examine the effect of scaling the dimensions of the test up or down this equation can be rewritten as

$$\frac{1}{B}\frac{dW}{dA} = \left(\frac{F}{A_o}\right)\frac{d\mathcal{E}}{d(A/A_o)} + \frac{1}{2}\left(\frac{F}{A_o}\right)^2 \frac{d}{d(A/A_o)}\left(\frac{A_o}{BM}\right)$$

where $Bd\mathcal{E}$ replaces $(d\ell - d\ell')$ and A_o is the area severed when the fracture is completed. We refer to (A/A_o) as the relative fracture area.

Figure 5 shows a typical set of results from carefully scaled notched bend tests. On this figure equal values of relative fracture area correspond approximately to equal fractional drops of the load. The 15/8-inch test shows onset of unstable fracture. For the 3/8-inch test size instability fails to occur;at 80 per cent of maximum nominal stress, the inelastic bending strain is 50 per cent greater than that for the same relative fracture area in the case of the largest test size. In the above relation for (1/B)(dW/dA), when $d\mathcal{E}/dA$ is zero, as is uaually the case at instability, the second term on the right represents all of (1/B)(dW/dA). The effect of test size on this term is confined to that in the square of the nominal stress. Thus there would be very little fracture size effect in (1/B)(dW/dA) except for the fact that decreasing the test size causes delay in the onset of instability until a larger relative fracture area has been attained. The mechanism for this delay is the pronounced size effect in the plastic work rate represented by the first term on the right of the equation. A speeding up of the generally distributed plastic flowing as by increasing the temperature may be expected to delay onset of instability to a larger value of

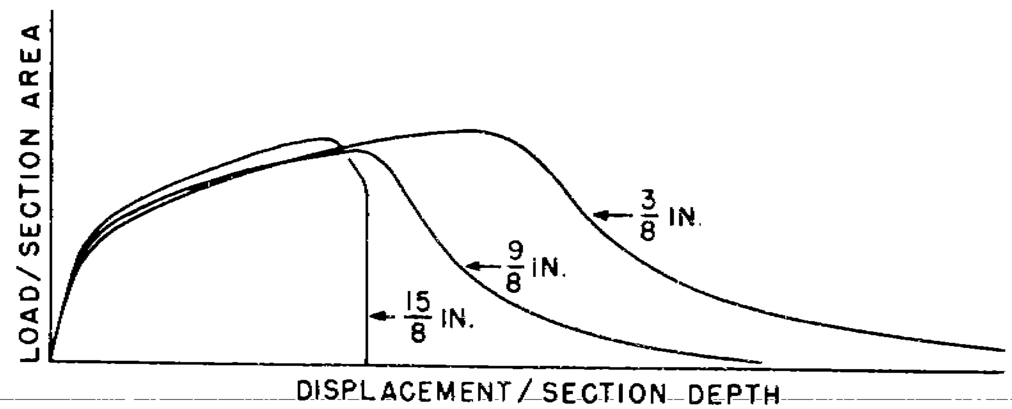

Figure 5 - Typical load-deformation curves for three sizes of notched bar bend tests of a steel. From Shearin, Ruark, and Trimble, reference (4)

its cracked extensions is just the strain energy which would be contained in an ellipse having the separation of the crack tips as its minor axis, twice this length as its major axis, and located in a position where the stress is the nominal stress. Thus if X is the crack length,

$$\frac{dE}{dA} = \frac{1}{t}\frac{dE}{dX} = \frac{1}{t}\frac{d}{dX}\left(\frac{\pi X^2 t}{2}\frac{\sigma^2}{2Y}\right)$$

where Y is Young's modulus and t is the plate thickness. For X = 3.5 inches

$$\frac{dE}{dA} \cong 200 \text{ in-lb per sq. in.}$$

Kies and Irwin published a more exact method for making this calculation in the research supplement to THE WELDING JOURNAL of February 1952.

The difficulty with having too much plate deformation prior to fracture might be avoided in centrally notched plate tests by lowering the stress and working the crack length out to the instability point by fatigue or repeated loading. An indication of what can be done along these lines is furnished by the Cornell University ship plate steel fatigue tests. These tests do not, actually, exhibit results in which the unstable fracture length was attained. If we assume that each test, in which well developed cracks appeared, was carried to the point where the speed of extension of the crack, say per 1000 cycles, indicated to the test operator serious danger of a quick break, then the dE/dA values for these crack lengths represent limiting values of practical interest. The tests of simplest geometry were with 3/4-inch plates of 17-inch width containing a 4.25-inch diameter central hole. Cracks extending one to two inches to each side of the hole were obtained in 20 to 100 thousand cycles using a nominal stress of 22,500 psi in nearly one million cycles for a nominal stress of 15,000 psi. More than half the test time was required to get any kind of a crack started. Estimates of dE/dA were made by several rough approximation methods. All of these computed values were in the range of 150 to 400 in-lb per sq. in. The ability of ship steel plate to resist extension of fairly long cracks with effective dW/dA values in the range of 150 to 400 in-lb per sq. in. suggests that cleavage fracturing of mild steel plates requires dE/dA values which are far from vanishingly small. Apparently, for 3/4-inch mild steel plates, more than 100 in-lb per sq. in. of fracture must ordinarily be supplied from the strain energy field.

Another notable fact which carries a similar message is that mild steel cleavage fractures have been observed to stop relative fracture area. One can view the effect of increasing the size of the test as an increasing of the growth rate of relative fracture area by increasing the probability density of flaws. Increase of crack growth rate relative to the rate of deformation by distributed plastic flowing causes onset of instability at a smaller relative fracture area.

In the steels used in the notched bend tests at the University of North Carolina, the difference in measured values of effective dW/dA was moderate and may not seem to represent fully the relative toughness of the steels when employed at a temperature where one is ductile and the other brittle. One may note that these tests were all conducted against dE/dA energy release capabilities limited to the range of about 200 to 400 in-lb per sq. in. Toughness values outside this range would not have been measured. Centrally notched plates, such as were employed in wide plate tests in fracture studies at Swarthmore and in model experiments at NRL, are preferable. These tests can be operated without side notching and have the possibility of fracture instabilities at depths of cracking more realistic with respect to observability than those permitted by the University of North Carolina bend tests. Values of toughness corresponding to critical crack lengths less than the width of the central notch, of course, are not measurable. A more serious disadvantage is the general deformation which accompanies the growth of fibrous fractures of substantial length in such tests. This deformation may be thought to change the material properties too much before onset of unstable fracturing permits the measurement which is of interest to us. By looking through the Swarthmore results for cracks which went unstable at lengths only a small amount longer than the original three-inch long central slot, Mr. Kies selected cases and calculated results which at least represent the minimum of plastic deformation consistent with the assumption that the critical length was not being limited by the length of the slot. These calculations resulted in values of about 200 in-lb per sq. in. Again the numerical result for steels of similar yield strength is affected very little by plate toughness. One can measure only the temperature at which this condition of instability occurs for steels of various quality.

A procedure in agreement with the Griffith crack theory of fracture can be used for obtaining an approximate value of dE/dA for the above experiment. The test was made with 12-inch wide plates of 3/4-inch steel having a central notch three inches wide. The nominal stress, σ, was 32,000 psi. We assume the strain energy relieved by the central slot plus in plates subjected to 15,000 psi or less of tensile stress. The expected mechanics of having a fracture change from an unstable to a stable condition relative to self-fracturing on released strain energy does not differ greatly from the reverse transition as represented in the experiments discussed above. However, information available to the writer does not permit calculation in this paper of effective dW/dA values from fracture arrest data.

Estimates along the lines discussed above using a dW/dA of 200 in-lb per sq. in. indicate that if the nominal stress in a structure of mild steel plates never exceeds 30,000 psi then a crack forming and moving out into a plate of this structure should not go unstable until it has developed to a length of four inches. The critical length for 15,000 psi would be four times as large. Cleavage fracture origins as large as this are not often found because the starting crack length can be produced in ships in so many ways other than by creep or by slow growth in fatigue. Most of these have to do with poor weldments and the residual stresses associated with welds. Efforts to find out more about the seriousness of various kinds of flaws are in progress at several places. In order that the information from these studies may be helpful in more than a qualitative sense, it is desirable that the experiments be designed so as to show the strain energy release rate considerations applicable either to the start or to the arrest of unstable fast fracturing. The same remark applies to studies of transition temperatures. In the latter case so much has been done with standard impact tests that an effort might be made to correlate results of these standard impact tests with some test capable of furnishing effective dW/dA values at crack lengths appropriate to specific applications. A centrally notched plate in fatigue is an example of a suitable kind of measurement which might be employed for this purpose.

Several ways suggest themselves of using dW/dA values in connection with estimating danger of brittle fracture in large structures of mild steel plate. For example where a welded connection between two plates runs parallel to the direction of greatest tension, cracks to either side perpendicular to the weld and extending through the heat-affected zone might be considered to indicate a probable starting crack size. Allowing something for unnoticed extension of such a crack during periods of stress and temperature variation and assuming this happens in poor quality ship plate, a calculation can be made based on say, a 3-inch long starting crack and an effective dW/dA of, say, 150 in-lb per sq. in.

For this situation it should not be far from correct to take

$$150 \text{ in-lb per sq. in.} = \frac{\pi}{2} \frac{(3 \text{ in.})(T^2)}{30 \times 10^6 \text{ psi}}$$

Thus we would find T equals about 30,000 psi as a condition for fracture instability. It has been suggested that brittle fractures of welded ships might be eliminated if no stresses large enough to cause yielding were permitted in the structure. The suggestion appears to be quantitatively appropriate for the starting crack situation just described.

If the weld parallel to the tension direction is joined by a crosswise weld coming in from one side, the situation is somewhat more alarming because we do not know how much of the crosswise weld participates as part of the starting crack. It may be a few inches or as much as a foot. If the latter is the case, then a nominal stress of less than half of the material yield strength is a dangerous condition.

Unless some limitation can be placed upon the starting crack sizes which have fair probability of occurrence, the stress limitations for safety against unstable fracturing become quite unreasonable. The writer has thought that inspection procedures might furnish upper bounds on the lengths of the starting cracks which need to be assumed as potentially present in the structure during its load time history. Granting that we have or can find out the critical strain energy release rate for various cracks and flaws, then one can estimate their relative seriousness. The inspectors can then be told what flaws are not tolerable, and the design drawings can be checked in a realistic way for estimating danger of unstable fracturing. The checking procedure would be to suppose cracks of the maximum non-observable length to exist in the worst location of each region of high stresses. One would then calculate dE/dA for each and compare with whatever appropriate effective dW/dA values are known.

The proposals listed above appear to be direct applications of certain information about fracturing which has emerged in recent years. It is realized that practical construction difficulties associated with large welded structures make a situation far from ideal for any theoretical considerations, that inspectors may have limited diligence or poor vision, and that the workmen never cease from producing unpredictable flaws. However, it is believed that the introduction into engineering design of rational quantitative procedures for control of unstable fracturing will nevertheless have many beneficial effects.

REVIEW OF BRITTLE FRACTURE RESEARCH UNIVERSITY OF ILLINOIS

by

Professor N. M. Newmark
University of Illinois

INTRODUCTION

1. Summary of Programs

This paper is a brief review of a number of research programs at the University of Illinois on the subject of mode of fracture of structural steel. The main features of several of the programs are described, and a brief summary is given of the general trend of the results. More complete information on the various programs is given in the references indicated in the bibliography.

The first program, on the so-called wide plate tests, was conducted by Professor W. M. Wilson under the auspices of the War Metallurgy Committee (later the Ship Structure Committee) and involved static tests of internally notched steel plates of various widths(1).

Additional studies of the factors influencing the fracture of structural steel were conducted under the sponsorship of the Office of Naval Research, under Contract N6ori-71, Task Order5, and the results are reported in a number of theses(2,3,4,5), and in a paper(6) in which the major results of the experimental programs were summarized in terms of a nondimensional parameter.

As a part of a program on the strength of riveted joints connecting steel plates, a number of tests were made, under the direction of Professor Wilson, which resulted in brittle fractures. The implications of these tests are of importance in connection with the practical problems of brittle fractures in structures, and a brief survey of the important test results are given herein. A detailed report of these tests is available(7).

2. Factors of Importance in Engineering Design

In connection with the interpretation of the fractures in ship structures which have occurred, and in any study of the importance of brittle fracture as a factor in engineering design, there are three principal phenomenological concepts which require consideration. These are as follows:

a. The strength of the material, in either brittle or ductile fracture, and the relation of the strength to both the design stresses of the structure and to the actual stresses existing under the most severe conditions of service.

b. In structural applications where dynamic loads are of importance, the energy absorbing capacity of the material becomes the factor of major importance, rather than the strength, whether the fracture be brittle or ductile.

c. The factor of greatest practical importance concerns the question as to how the strength and energy absorbing capacity can be measured in order to determine whether a given material has the required properties to resist the loads which are applied to the structure. This engineering approach, which involves the use of a given material whatever may be its properties, is somewhat different from the more fundamental approach of developing a material which has the properties necessary for a given condition of service.

In most applications both the strength and the ductility are important, and neither can be considered independent of the other. An indication of the variation of both of these quantities with temperature is shown by the nominal stress-elongation curves of Figure 1, taken from the results of tests of very carefully made specimens in Reference 4. Although it appears from Figure 1 that the strengths of all of the specimens are about the same, this arises from the fact that for the same amount of elongation the stresses are higher for the lower temperatures, but the amount of elongation at fracture is reduced as the temperature is decreased. Where brittle and ductile fractures occur at nearly the same temperatures, the ductile fracture generally occurs at a higher stress. However, where the brittle fracture occurs at a much lower temperature than a ductile fracture, the strengths are likely to be the same, or in some cases the brittle fracture may show a higher strength than the ductile fracture.

WIDE PLATE TESTS

3. Effect of Plate Width on Strength

One of the striking features of the wide plate tests is connected with the maximum stress which could be carried by the plates before fracture occurred. In order to indicate the general nature of the results, Table 1 shows the fracture strength of internally notched plates the results of which were reported in Bulletin 388(1). Values are given for three steels, for fractures which are either primarily ductile or primarily brittle in character, and for two widths of plates, namely 12 inches and

Table 1. Fracture Strength of Internally Notched Plates
(Average of Values from Table 1 and Fig. 24, Bulletin 388)

Type of Steel	Type of Fracture	Fracture Strength, ksi	
		12-in.wide plates	72-in.wide plates
Rimmed Steel E	Primarily Ductile	49.2	41.4
(as Rolled)	Primarily Brittle	38.7	32.8
Killed Steel D	Primarily Ductile	53.7	45.7
(as Rolled)	Primarily Brittle	49.6	42.3
Killed Steel D	Primarily Ductile	50.5	42.8
(Normalized)	Primarily Brittle	48.0	38.3

72 inches. The values for plates of intermediate width ranged between those reported for the widths selected in the Table. The strengths ranged generally from values slightly above the yield point strength of the material to values slightly below the ultimate strength of the material as determined by standard coupon tests, The higher strengths were obtained for the narrower specimens and the lower strengths for the wider specimens. The values reported in Table 1 indicate a differential between the 12-inch and the 72-inch plates ranging from about 6,000 to 10,000 psi, whether the fracture is ductile or brittle. Moreover, the difference between the strengths for ductile and for brittle fractures was relatively small for Steel D, ranging from 2,500 to 4,500 psi, and although somewhat larger for Steeel E, the difference was only of the order of 9,000 to 10,000 psi. Moreover, the strength of the 72-inch wide plates was not appreciably greater than the yield point strength of the material, even for ductile fractures.

The strengths reported in Table 1 are determined from the average stress on the net section at the notch. Obviously, there is a stress concentration at the root of the notch, but the significance of the maximum stress after general yielding occurs is not a simple matter to evaluate.

4. <u>Effect of Strain Concentrations</u>

Although the nominal stress on the net section is a convenient measure, it is valid only as a sort of index rather than as a measure of the true stress at the point of failure, with the sharp notches of the types used in these plates, or of the type occurring at accidental cracks in structures. The theoretical stress concentration factor applying to elastic behavior is seriously misleading. Even if the material were to remain elastic, the stress concentration factor varies approximately inversely as the square root of the radius of curvature at the end of the notch or crack. As the plate deforms, the deformations become large enough to change this radius of curvature appreciably. It may be the instantaneous radius at which the crack forms which determines the stress concentration or strain concentration factor which is of greatest significance, although this is not necessarily the case. An indication of the change in the shape and character of the notch is shown by the photographs in Figures 2 and 3 of the local conditions existing at the root of the notch for both brittle and ductile fractures in wide plates.

An indication of the non-uniform distribution of strain at the net section is shown in Figure 4. These values are unreliable to the extent that the strains were measured by gages

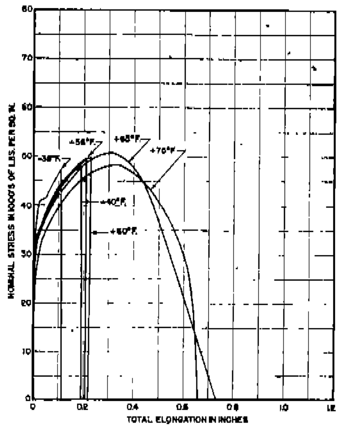

Fig. 1. Stress-Elongation Curves for Internally Notched Plates 1 1/4 in. Thick by 7 1/2 in. Wide

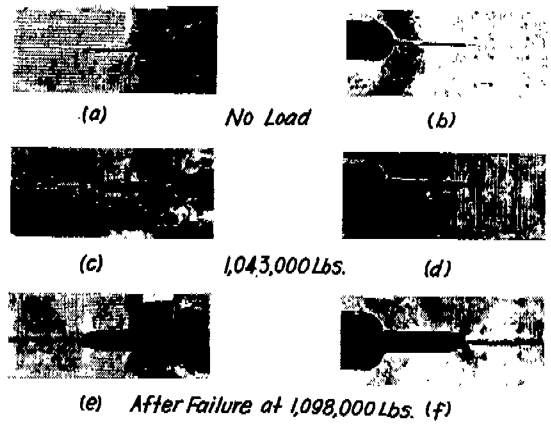

Fig. 2. Local Deformations at Root of Notch in Brittle Fracture of 48 in. Plate of Steel Dn.

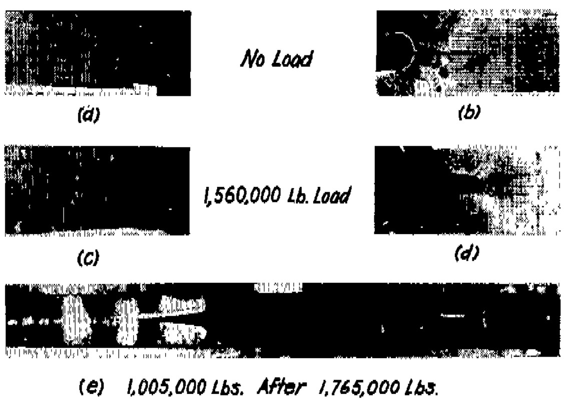

Fig. 3. Local Deformations at Root of Notch in Ductile Fracture of 72 in. Plate of Steel Dn.

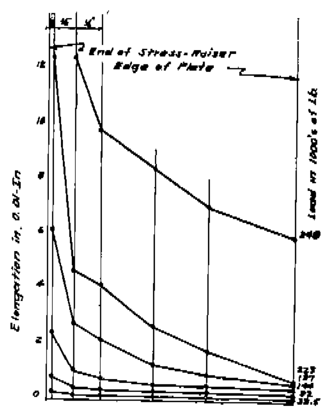

Fig. 4. Strain Distribution on Net Sect, 12 in. Plate of Steel E.

having a finite length and width; and consequently, the values shown are average values over some area rather than the values at the point plotted. However, in spite of this fact, the indications are that the strain at the root of the notch is of the order of four times the average strain on the net section throughout the range of loading in this particular specimen.

It is apparent from the results of tests that have been reported that the stress concentration or strain concentration at the root of the notch has an effect on the strength and on the ductility. However, no simple relationship appears in the data. The strain distributions given in Reference 1 indicate nothing which would cause brittle fractures to occur at lower stresses than the ductile fractures, except for the possibility that the distorted shape of the notch or crack at high loads may have resulted in somewhat smaller crack widths, and consequent higher stress concentrations, or strain concentrations, at the lower temperatures where brittle fracture occurred.

REVIEW OF DATA FOR WIDE PLATES AND OTHER NOTCHED SPECIMENS

5. Introduction of Dimensionless Parameter

As part of a study for his thesis, Mr. W. C. Hoeltje reviewed the data from the wide plate tests at Illinois, and from the various tests reported(2,3,4), as well as other similar data both at the University of Illinois and elsewhere in an attempt to correlate the various test results. The various types of specimens which were studied are shown in Figure 5. Type A represents the specimens used in the wide plate tests as well as in other supplementary tests at the University of Illinois. Among these latter tests were several series of tests by Randall, reported in Reference 4, on specimens having complete geometrical similarity both in size, thickness, and notch radius. Following a suggestion in a paper by Bagsar (The Welding Journal, Vol. 31, No. 3, pp. 97-s--123-s), Mr. Hoeltje plotted a number of test results against a parameter having the form of the radius of the notch divided by the net area at the section of failure. This quantity, designated by the symbol ρ, is not dimensionless but varies inversely as the specimen size for geometrically similar specimens. However, if this parameter is multiplied by a quantity such as the mean diameter of the grains in the structure of the material, it does become a dimensionless parameter. However, because of the fact that the grain size is not known for the various steels and particular conditions in all of the tests reported, the plots made by Hoeltje are given in terms of the parameter ρ rather

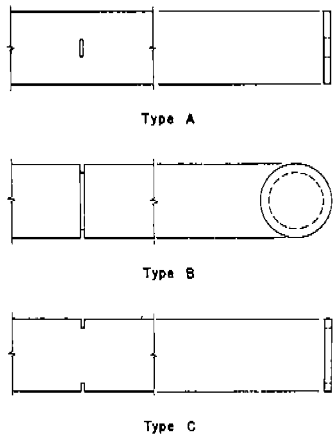

Fig. 5 Types of Specimens Investigated

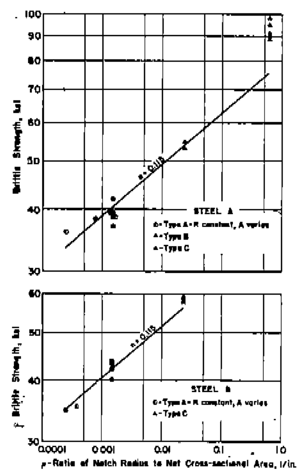

Fig. 6 Relation Between ρ and Brittle Strength
Steels A and B.

than the dimensionless parameter which can be obtained from ρ by multiplying it by the grain size. The results of these studies are reported(5,6). Typical results are given herein in the next two sections.

6. Brittle Strength

When the parameter ρ is plotted against the nominal strength on the net section for brittle fracture on a logarithmic scale, linear relations are obtained for various types of steel and are shown in Figures 6,7, and 8 for some of the steels considered. All of the lines have the same general slope which corresponds to the relationship,

$$\text{Strength} = \rho^n,$$

where n has the value 0.115. The curves for the different steels differ in their location; but all have the same slope, regardless of whether the radius varies and the cross-section remains constant, the cross-section varies and the radius remains constant, or both vary.

Most of the results are for specimens of Type A of Figure 5, but some data are shown for specimens of Types B and C. These were plotted on the same diagrams, with a modification in accordance with the theoretical stress concentration factor. This rather arbitrary modification cannot be entirely justified, but it does appear to make the points representing the other types of specimen fall on the same curves as those for Type A.

There is considerable scatter in the results, but the general trend is impressive. It is, of course, probable that the relationship does not remain linear for very large or for very small values of ρ. The strength certainly cannot become appreciably larger than the coupon strength of the material, and there are other indications that the strength cannot become appreciably lower than the yield point of the material. The relationships indicated are more striking when one recalls that the range in parameters covers the following values: thickness from 0.18 to 1.50 inch; width from 1.5 to 72 inch; and notch radius from 0.007 to 0.125 inch. The relationship certainly is only an empirical and approximate one, but it may be indicative of the fact that the parameters which enter into it are the most important in determining the brittle strength. The other variables probably have only minor importance in this regard.

7. Transition Temperature

Similar plots of the parameter against transition temperature are shown in Figures 9 and 10. Additional plots are given

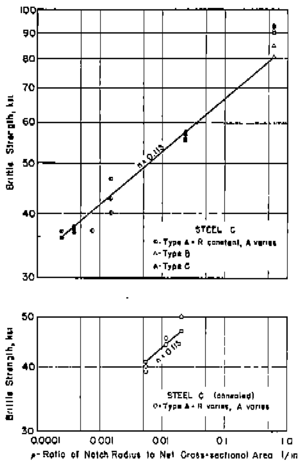

Fig. 7. Relation Between ρ and Brittle Strength Steels C and C (annealed)

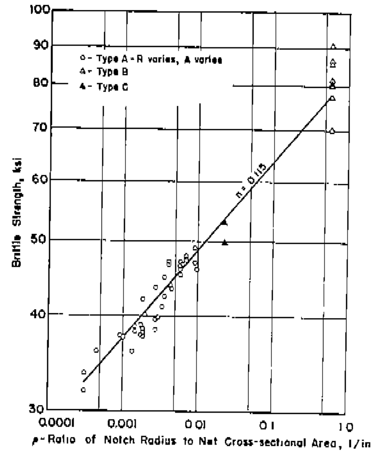

Fig. 8. Relation Between ρ and Brittle Strength Steel E

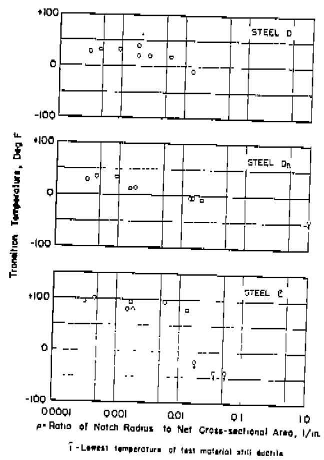

Fig. 9. Effect of ρ on Transition Temperature

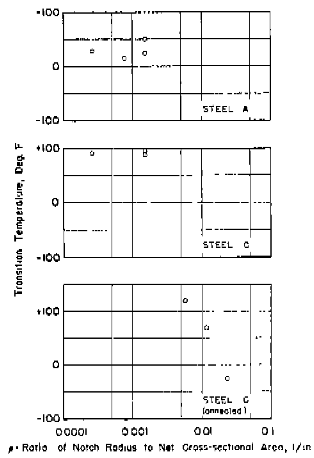

Fig. 10. Effect of ρ on Transition Temperature

in Reference 6. These results are not as striking as those in the preceding section, but there is an indication that below a certain value of the parameter ρ the transition temperature is unaffected by changes in the geometrical design of the specimen. Above the value the transition temperature shows a decrease in some sort of functional relationship with the parameter ρ. If this relation is borne out by other studies, it would appear that the ceiling value of transition temperature has some important significance as a material parameter. This would appear to be the highest value of transition temperature which can be obtained in a static test, regardless of the kind of notch or size of specimen. It therefore may serve as an indication of the relative merit of different materials. Comparisons of materials at other than the maximum value of transition temperature may in some cases be misleading.

TESTS OF RIVETED JOINTS

8. Brittle and Ductile Fractures of Riveted Joints

The data reported in Reference 7 cover a wide range of types of riveted joints, both of the double strap butt-type and the lap-type, at temperatures ranging from -26°F to +120°F. Although most normal fractures of riveted joints in ordinary structural steel are of a ductile type, under certain conditions brittle fractures were obtained in these tests even at ordinary temperatures, and in most cases brittle fractures were obtained at the lower temperatures. In general, however, the strengths of the net section even when brittle fracture occurred were not appreciably lower than the corresponding strengths for ductile fractures.

A ductile fracture of a typical riveted joint is shown in Figure 11, and a brittle fracture in Figure 12. In the latter figure, the typical shattering or multiple fracturing is shown in the top plate. The strengths of these two joints were of the same order of magnitude.

Brittle fractures of lap joints are shown in Figure 13. These are interesting in that they show the typical tendency for multiple fractures to occur either because of the stress waves generated in the specimen at failure or because they are initiated prior to failure.

A fracture with a considerably smaller elongation of the material is shown in Figures 14 and 15. The plates in this riveted joint were made of rimmed steel and had sheared edges. At the section of failure the reduction in area was only 1.3

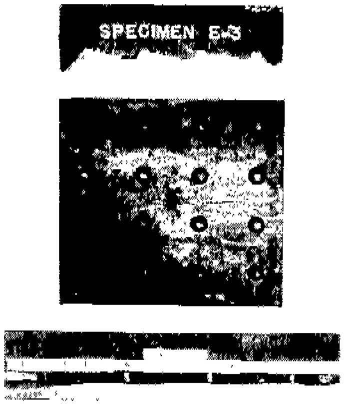

Fig. 11. Typical Ductile Fracture of Riveted Joint at 120°F.

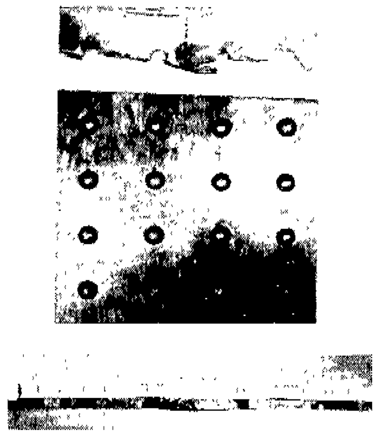

Fig. 12. Typical Brittle Fracture of Riveted Joint at -8°F.

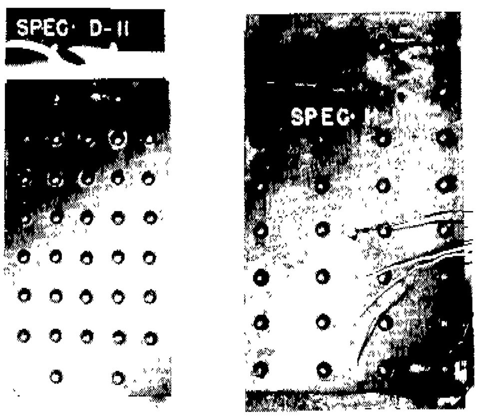

Fig. 13. Brittle Fractures in Lap Joints at -20° F. Showing Multiple Fractures.

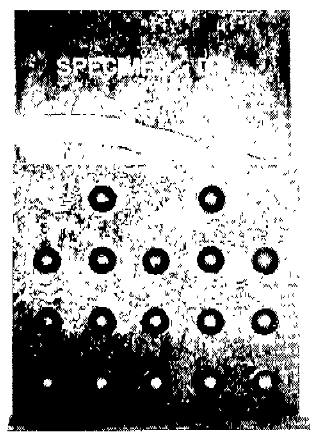

Fig. 14. Brittle Fracture Initiating at Sheared Edge of Plate and Passing Around Rivet Holes.

per cent. It is particularly interesting that the complete fracture occurred on a section not through rivet holes, at a stress about 73 per cent of the stress corresponding to the ultimate coupon strength of the material on the gross section.

9. Effective Edge Condition of Plates

The remaining figures illustrate unexpected types of failures in the test program. Figure 16 shows the first fracture of a specimen, in which the failure occurred at a welded connection attaching one of the plates to a pulling head. This failure occurred at a stress of 45,500 psi on the gross section, near the weld, in a plate having sheared edges. The temperature of test was -22°F. The pulling head was rewelded, and the edges of the plates were machined in the neighborhood of the weld in order to avoid difficulties from the sheared edge in the subsequent test. The next failure is shown in Figure 17. Here one of the two plates that failed tore through the rivet holes and the other through the gross section above the rivet holes. The stress in the joint at the section of failure was computed as 43,900 psi. In this specimen the theoretically weakest section was in the center plate shown at the bottom in Figure 17. The stress in this plate at the time that failure occurred in the previous tests reached values of 64,800, and 72,100 psi, respectively, for the two previous tests. The lower strength of the outer plates was probably due to the fact that they were made of rimmed steel and had sheared edges.

Another inadvertent failure indicating the effect of sheared edges is shown in Figure 18. Here the specimen under test was an internally notched plate 24-inch wide made of rimmed steel 3/4 inch in thickness, with a 1/4-inch drilled hole as a stress raiser. The test was conducted at a temperature of 88°F. The upper and lower sections of the test plate were welded to special pulling heads also 24 inches wide, but 7/8 inch thick. Since this plate was not one of the test plates and was not expected to fail, its properties were not completely known. Examination after the failure shown in Figure 18 indicated that the plate was a semikilled steel. Its yield point was 34,300 psi, and the reduction of area of coupon specimens was pactically 50 per cent. The pulling plate had been used for several other tests. The fractures shown in Figure 18 apparently occurred simultaneously and suddenly. Both fractures were brittle in general except for the portion of the fracture near the internal notch. The average stress on the net section through the notch was 45,100 psi, and that on the section of failure in the pulling head was only 28,300 psi. The pulling heads had sheared edges.

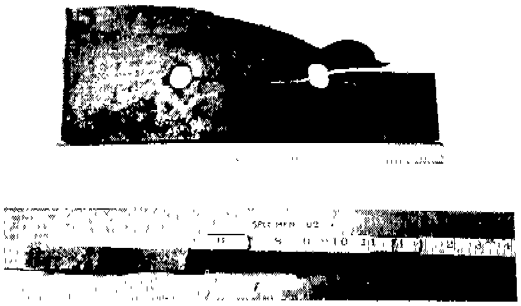

Fig. 15. Branching of Path of Fracture Shown in Fig. 14.

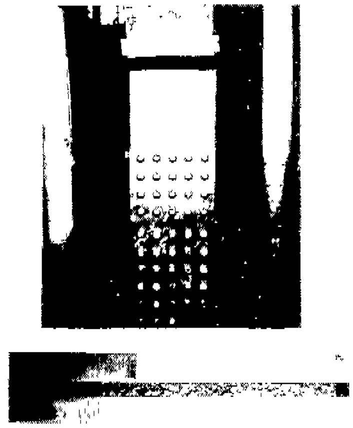

Fig. 16. Fracture Near Weld at Stress of 45,500 psi.

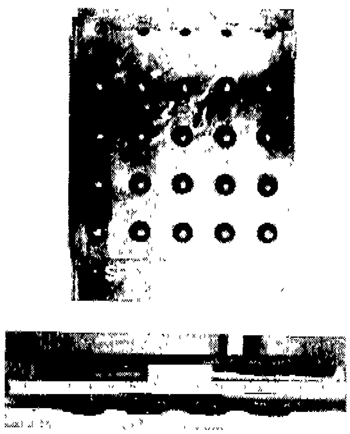

Fig. 17. Fracture Through Plate Away from Rivet Holes at 43,900 psi.

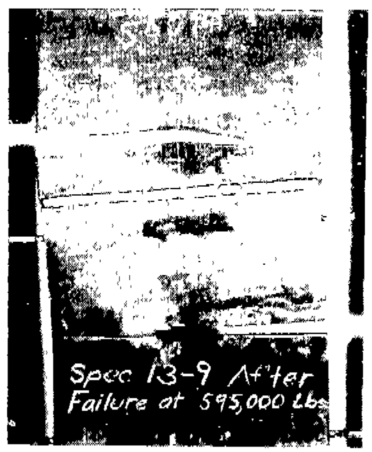

Fig. 18. Brittle Fracture of Plate at Nominal Stress of 28,300 psi.

The generally unsatisfactory behavior of plates with sheared edges at low temperatures indicates the importance of the method of fabrication. Brittle fractures of riveted joints apparently occur rarely if at all when the plates have machined edges or universal mill edges and the holes are drilled instead of punched.

GENERAL DISCUSSION AND SUMMARY

10. Minimum Stress at Fracture and Dependence on Yielding

In general, subject to certain limitations which will be discussed below, no fracture has been obtained in tests at the University of Illinois where the nominal stress on the net section at failure has been less than the yield point of the original material. The test shown in Figure 18 appears to contradict this statement. However, in this test, the failure in the test plate apparently developed on the right hand side somewhat before that on the left hand side, and the subsequent dynamic release of stress may have produced an eccentricity in the loading on the pulling head. This eccentricity would produce a much larger nominal stress at the left hand edge where the fracture in the pulling head started, as indicated by the herringbone markings. Brittle fractures below the yield point of the material were also obtained by Randall, and reported in Reference 4, for rimmed steel "E" which had been prestrained approximately 3 per cent and aged. The brittle fractures so obtained occurred at stresses below the new yield point of the material, but the strengths were still considerably above the yield point of the virgin material.

Where the stress on the net section is eccentrically applied in such a way that the nominal stress cannot be uniformly distributed even without stress concentrations, then it is reasonable to expect that the average stress on the section at failure can be below the yield point. However, under these conditions, one would expect the nominal stress at the root of the notch to govern the fracture. The internally notched plates were considerably less sensitive to nominal stress increases at the root of the notch because of eccentricities of loading than other types of notched plates. Externally notched plates, for example, would show a very much greater effect of eccentricities of loading. Whether it is possible for an extremely wide plate, much greater in width than 72 inches, of the type tested in the wide plate series, to fail with a strength less than the yield point of the material is a question that has not yet been settled. If it can fail at a lower stress than the yield point, when eccentricities and other than static loading are avoided, it would have to be for relatively high strain concentrations at the root of the notch. However, if the length of the plate is short and the loading consequently cannot be uniformly distributed over the net section even after yielding takes place, one would expect that the stress at failure might be decreased.

So far as the writer is aware, yielding on the net section precedes failure except where loading is eccentric, dynamic forces affect the stress distribution, or prior straining and aging have occurred.

11. <u>Initiation and Propagation of Cracks</u>

The factors influencing the initiation of a crack are very much different, apparently, from those influencing its propagation. It appears likely that the initiation of cracks both in test specimens and in structures are similar in character. However, the propagation of cracks in test specimens and in actual structures must be very much different in character because the dynamical behavior of a specimen differs from that of a structure. The differences arise because of differences in mass distribution and in storage of energy in the various parts. Unless a specimen is complicated enough to represent the action of the structure in respect to mass distribution and energy storage, its behavior will very likely be different insofar as propagation of cracks through the specimen is concerned.

The initiation of a fracture requires the application of a local stress probably in every case above the yield point of the material. However, it is possible to apply the stress locally by means of a blow or an impact, without applying a general stress distribution higher than the yield point over the entire net section.

When a fracture is initiated by means of an impact or other dynamic influence, the action may be different from that under a static influence because of the difference in yield point of the material. The effect of the delay in yielding of structural steel when subjected to a rapidly applied load may be extremely important for a dynamic load and of negligible importance under static conditions insofar as initiation of the crack is concerned. However, when it becomes a matter of propagation of the crack, the phenomena are all dynamic phenomena in any case. The suggestion was made by Asrow(2), that material with a marked upper yield point might show more notch sensitivity than material without such a marked upper yield point. If this is indeed the case, the explanation is probably valid only insofar as propagation of the brittle crack is concerned. It certainly does not apply to the initiation of a crack. However, if after a crack has started, the yield point is increased because of the speed of loading or because of the time delay to produce yielding, it is possible for cleavage to occur before yielding, with a consequent brittle fracture.

ENERGY CRITERIA OF FRACTURE

by

Professor E. Orowan
Massachusetts Institute of Technology

1. THE GRIFFITH ENERGY PRINCIPLE

In the course of the last few years, it has become clear that the Griffith equation for the tensile strength of a brittle solid cannot be applied in its original form to brittle fracture in normally ductile steels. X-ray back reflection photographs show (1) that a thin layer at the surface of apparently quite brittle fractures of low carbon steels contain significant plastic distortion; the plastic work p in this layer amounts to roughly 2×10^6 ergs per sq. cm. if the fracture has occurred not too far below room temperature. Compared with this value, the surface energy α (a few times 10^3 ergs per sq. cm.) is negligible; consequently, if an expression of the Griffith type can be used at all in this case, the surface energy (representing the work for creating unit area of the surface of fracture) has to be replaced by the plastic surface work p. Thus the crack propagation condition would be(2)

$$\sigma \approx \sqrt{\frac{Ep}{c}}. \qquad [1]$$

The presence of considerable plastic distortion at the surface of fracture raises the question under what conditions the Griffith principle of virtual work can be applied to fractures accompanied by plastic deformation. This principle can be stated in the following manner: Let dW be the free energy required for increasing the length* of a crack from c to

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*As in the original work of Griffith, only two-dimensional cases (cracks in plate-specimens) will be considered here for simplicity. The general results can be easily extended to three-dimensional cases.

Professor Orowan has kindly contributed two papers to this volume. These papers, while generally summarizing his remarks at the conference, also include the results of additional study and analysis following the conference. Ed.

c + dc, and -dU the elastic energy released simultaneously in the specimen if this is held between rigidly fixed grips so that the external forces cannot do work. The critical length of the crack above which it can propagate spontaneously is then determined by the condition

$$dW = -dU. \qquad [2]$$

It is easily seen that the assumption of rigidly fixed grips is not essential; the same result is obtained if the crack propagation is assumed to occur under constant load. Let M(c) be the elastic compliance, i.e., the reciprocal spring constant, of a specimen containing a crack of length <u>c</u>; thus,

$$x = MF \qquad [3]$$

where <u>F</u> the tensile force acting upon the specimen and <u>x</u> its elastic elongation. The elastic energy of a specimen containing a crack of length <u>c</u> is

$$U = \int_{x=0}^{x=MF} F \cdot dx = \frac{M(c)F^2}{2} \qquad [4]$$

and

$$dU = \frac{F^2}{2}dM + MF \cdot dF; \qquad [5]$$

$dM = \frac{dM}{dc}dc$ is the increment of the elastic compliance due to the increase by <u>dc</u> of the crack length.

If the crack length increases while the specimen is held between rigidly fixed grips, x = MF = const. and

$$dx = MdF + FcM = 0; \qquad [6]$$

substitution of MdF = -FdM in [5] gives

$$(dU)_x = \frac{-F^2 dM}{2}. \qquad [7]$$

On the other hand, if the crack propagates while the load is kept constant (dF = 0), eq. [5] gives

$$(dU)_F = \frac{F^2 dM}{2}; \qquad [8]$$

at the same time, the force $\underline{F}$ does the work

$$dL = F \cdot dx = F^2 dM, \qquad [9]$$

since, at constant F, $dx = F dM$.

Equations [8] and [9] show that, if the crack propagates at constant load, half of the external work is stored as additional elastic energy of the specimen, and the other half is available for increasing the free energy of the crack. If the length of the crack exceeds the critical value at which eq. [2] is just satisfied, the work of the applied force is more than sufficient to provide the increment of its free energy; the balance creates kinetic energy and accelerates the rate of crack propagation.

If on the other hand, the crack propagates between fixed grips, the elastic energy of the specimen decreases according to eq. [7], and its decrement is available for increasing the free energy of the crack and the kinetic energy. Comparison of [7], [8], and [9] shows that the energy available for crack propagation at fixed load is the same as at fixed grips; in the former case, $-dU$ in eq. [2] has to be replaced by $dL - (dU)_F$ which is numerically equal to $-(dU)_x$ for the same increment $\underline{dc}$ of the crack length.

In the present paper, two questions will be treated that have been widely discussed in connection with the brittle fracture of structural and ship steel, and on which a wide divergence of opinions has arisen. They are:

(A) Does the Griffith equation

$$\sigma \approx \sqrt{\frac{E\alpha}{c}} \qquad (\alpha = \text{surface energy}) \qquad [10]$$

represent a necessary and sufficient condition of completely brittle fracture? And is the present writer's equation [1] a necessary condition of brittle fracture in low carbon steels?

(B) Under what conditions can the Griffith principle [2] be applied to fractures involving plastic deformation?

2. THE GRIFFITH EQUATION AS A NECESSARY AND SUFFICIENT CONDITION OF COMPLETELY BRITTLE FRACTURE

It is obvious that eq. [10] is a necessary condition of crack propagation in a completely brittle specimen under tension. If it is not satisfied, propagation of the crack with the accompanying increase of its (free) surface energy would violate the first or the second law of thermodynamics. In particular, thermal fluctuations (disruption of atomic bonds at the tip of the crack by thermal activation) cannot propagate the crack if the Griffith equation is not satisfied, because any such process would result in the creation of free energy from thermal energy without heat flowing from one reservoir to another of a lower temperature. Of course, thermal fluctuations of free energy do occur; however, they cannot lead to any significant crack propagation because the greatest energy fluctuation that may arise with any probability amounts to a few electron volts which is equivalent to the disruption of a few individual atomic bonds at the tip of the crack.

From the fact that the Griffith equation is a necessary condition of completely brittle fracture, it does not follow that it is also a sufficient condition. However, it can be proved that once the condition is satisfied, crack propagation is not merely possible but is bound to follow. This can be shown by proving that, if the applied stress has the value given by the Griffith equation, the stress concentration at the tip of the crack reaches the value of the molecular cohesion (theoretical strength) at which fracture is bound to take place.

The molecular cohesion of a brittle material can be estimated in the following well known way. When a rod of unit cross sectional area breaks with a smooth surface of fracture perpendicular to the axis of the rod, two new surfaces of unit area are created; the work required for this is 2α (α = surface energy). This work is done against the intermolecular attractive forces as the two fragments are pulled apart. Figure 1 shows the variation of the molecular forces between the two fragments, per unit of cross sectional area, as a function of the distance $\underline{d}$ between the layers of molecules in the two fragments that are adjacent to the surface of separation. The force is zero when $d = b$ = the molecular spacing in the absence of stress; it rises to a maximum σ_m which is the molecular cohesion and then falls to zero with increasing separation of the fragments. The area below the curve is the work of fracture per unit of the cross sectional area, i.e., it is equal to 2α. At the maximum of the curve in Figure 1, the amount of energy

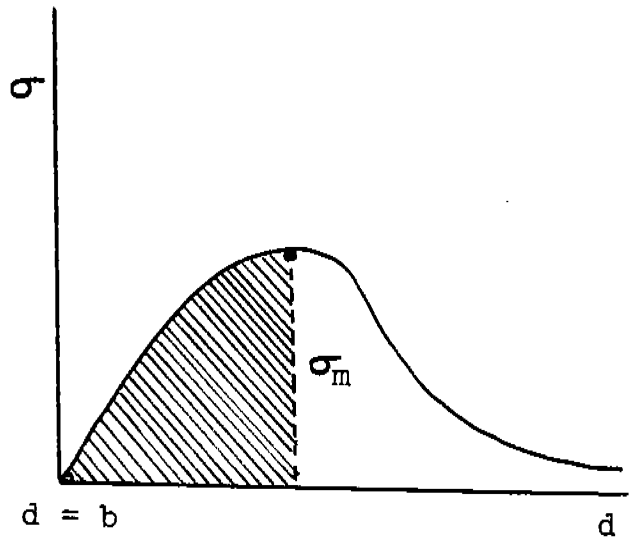

Figure 1

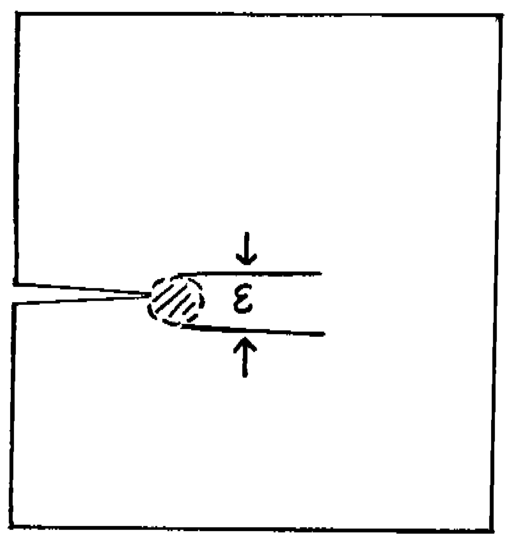

Figure 2

represented by the shaded area below the curve must be present between all neighboring pairs of molecular (or atomic) planes perpendicular to the tension; it is identical with the elastic energy stored in the material between two adjacent atomic planes. If, for an order-of-magnitude estimate, Hooke's law is assumed to be valid up to the theoretical maximum σ_m of the stress, the density of elastic energy at this point is $\sigma_m^2/2E$, and the elastic energy between two atomic planes of unit area, spaced at b, is $b \cdot \sigma_m^2/2E$. If it is assumed that the shaded area is about one-half of the total area below the curve and therefore approximately equal to α, the relationship

$$b \cdot \frac{\sigma_m^2}{2E} = \alpha \qquad [11]$$

gives the order of magnitude of the molecular strength as

$$\sigma_m \cong \sqrt{\frac{2E\alpha}{b}}. \qquad [12]$$

The next question is: what is the value of the applied tensile stress at which the critical value σ_m is reached at the tip of the crack? The stress concentration factor of a surface crack of depth c and root radius ρ is(3)

$$q = 2\sqrt{\frac{c}{\rho}}; \qquad [13]$$

this relationship shows that the maximum stress would be infinitely high for any finite value of σ and c in an elastic continuum containing a perfectly sharp crack, and therefore the tensile strength would be zero. The reason why brittle solids have a finite strength lies in the atomic structure of matter. Figure 1 shows that Hooke's law breaks down when the increment of the atomic spacing becomes comparable in magnitude with the atomic spacing itself: near the tip to the crack the stress vs. strain curve levels out, and the situation can be regarded roughly as if a certain region at the tip, comparable in linear dimensions with the interatomic spacing, would be under the constant stress σ_m instead of obeying Hooke's law.

This case of the laws of elasticity ceasing to be valid in a region at the tip of the crack has been treated by L. Föppl(4) and, in particular, by Neuber(5). Neuber proved the following theorem: Let there be a region of linear dimensions ε at the tip of the crack (Figure 2), so that the specimen is

Hookean elastic outside this region, whereas the stress in the region is approximately constant at the value existing at its boundary; the ratio of the stress in the region to the tensile stress applied to the specimen is then equal to the stress concentration factor of a crack of the same length and of the root radius $\varepsilon/2$ in a purely Hookean elastic material. (The quantity ε is assumed to be small compared with the length c of the crack which itself must be small compared with the dimensions of the specimen).

In the present case, the diameter of the region in which Hooke's law breaks down and the stress levels out is obviously of the order of magnitude of the interatomic spacing b; if it is assumed to be approximately 2b, Neuber's theorem indicates that the effective stress concentration factor is that of a crack of tip radius b in a purely Hookean specimen. According to eq. [13], this is

$$q = 2\sqrt{\frac{c}{b}}. \qquad [13a]$$

Thus, the value of the applied tensile stress at which the molecular strength is reached at the tip of the crack is given by

$$\sigma_m = \sigma \cdot 2\sqrt{\frac{c}{b}}; \qquad [14]$$

if σ_m is replaced from [12], the tensile strength σ is obtained as

$$\sigma \approx \sqrt{\frac{E\alpha}{2c}} \qquad [15]$$

which, within the accuracy of the estimate, is identical with the Griffith equation [10].

This derivation of the Griffith equation directly from the stress concentration factor of the crack shows that, when the applied tensile stress has the value given by the equation, the stress at the tip of the crack reaches the highest value that can be withstood by the interatomic forces in the material. Any further straining is bound to produce crack propagation and fracture. In other words, the Griffith equation represents not only a necessary but also a sufficient condition of fracture in a completely brittle specimen.

3. CAN THE GRIFFITH PRINCIPLE BE APPLIED TO DUCTILE FRACTURE?

In recent years the view has been expressed that the Griffith energy principle eq. [2] may be applied to all types of fracture, not only to essentially brittle ones. In what follows, it should be pointed out that this is not so: the principle can only be applied if plastic deformation is either absent or confined to a thin layer at the crack walls so that the bulk of the specimen is still elastic.

Figure 3 indicates the manner of crack propagation in a purely elastic material: owing to elastic strain release around the crack, its walls are pulled apart, and its length increases. Figure 4, on the other hand, shows one of the simplest types of ductile fracture(6), such as is observed in aluminum single crystals or (polycrystalline) plates of ductile metals in tension. The crack (which in this case has a square cross section) is propagated by slip in the planes AB + CD and EB + CF; in the course of this process the cross section of the crack increases until fracture is complete.

The fundamental difference between the propagation of the brittle crack shown in Figure 3 and the ductile mechanism of Figure 4 is that the former is based essentially on the elasticity of the material, while the latter could work in the same way even if the elastic moduli were infinitely high. The Griffith equation [10] shows directly that the tensile strength of a brittle material would rise to infinity with an infinite increase of the value of Young's modulus: in such a material, the crack could not open up because there would be no elastic strains to release. On the other hand, the slip mechanism shown in Figure 4 is quite independent of the elastic moduli.

It is immediately obvious that the force required for propagating the crack in Figure 4 cannot be derived from the Griffith principle eq. [2]. Its value is simply

$$F = Y \cdot A \qquad [15]$$

where $\underline{Y}$ is the yield stress of the material in tension and $\underline{A}$ the projection of the areas $\underline{AB}$ plus $\underline{CD}$ on the plane perpendicular to the direction of the tension; if $\underline{F}$ satisfies [15], the plastic deformation that opens up the crack can progress, and the crack propagates. The elastic moduli do not appear in [15]; they could be infinitely high without any consequence to the propagation of the crack. On the other hand, infinitely high elastic moduli would make the right hand side of eq. [2]

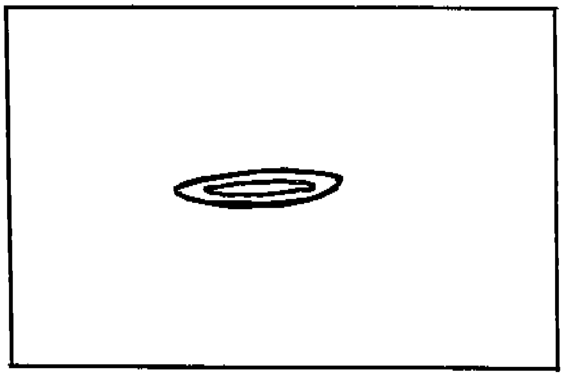

Figure 3

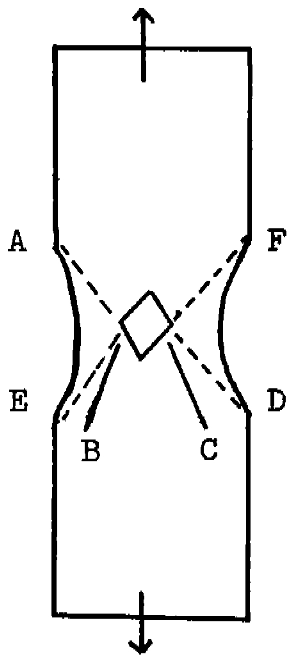

Figure 4

vanish: this shows that the tensile strength obtained by any application of the Griffith energy principle would rise to infinity with the elastic moduli.

The conclusion is, then, that the Griffith energy principle can only be applied to fully or substantially brittle fractures; ductile fractures are quite outside its scope.

In arguing the applicability of the elastic energy release principle to ductile fractures, occasionally the point has been made that if the specimen is long enough, the elastic energy stored in it should be sufficient to produce rapid crack propagation even if the energy absorption of the crack is as high as it is in typically ductile fractures. The answer to this is that a <u>fast</u> fracture is not necessarily a <u>brittle</u> fracture (i.e., a fracture involving very low energy absorption). Any ductile fracture can be made to run fast, at least from a certain stage onwards, if the specimen is connected in series with a large enough spring (or, what is the same, if the specimen is long enough). It can be shown that the condition for a ductile fracture to become a fast fracture is not eq. [2] but equality of the <u>second</u> derivatives of <u>W</u> and <u>U</u>*.

4. THE WRITER'S CRACK PROPAGATION CONDITION FOR BRITTLE FRACTURE IN NORMALLY DUCTILE STEELS

As mentioned in the first Section, the present writer has suggested that brittle fracture in ductile steels may obey the crack propagation condition

$$\sigma \approx \sqrt{\frac{Ep}{c}} \qquad [1]$$

which results if, in the Griffith equation [10], the surface energy is replaced by the surface plastic work <u>p</u>. It can be obtained by starting from the Griffith principle of elastic energy release eq. [2] and equating the free energy required for producing unit area of the crack wall to <u>p</u> instead of α.

The first question is: Can the Griffith energy principle be applied to a fracture process that involves plastic deformation? It was seen in Section 2 that the Griffith equation can be derived from the elastic stress concentration factor of the

*To be published in a separate paper.

crack; however, can this be done if plastic deformation takes place and re-distributes stresses at the tip of the crack? The Neuber principle, mentioned in Section 2, shows that the stress concentration factor can be calculated on the basis of the classical theory of elasticity if the plastically deformed region is small compared with the length of the crack. In that case it can be treated in the manner explained in connection with Figure 2: the stress concentration factor is the same as that of a crack in a purely elastic body with a tip radius equal to half of the diameter of the plastically deformed region. In fact, this case is only quantitatively different from that of the completely brittle material in which, in order to take into account the atomic structure of matter, the same consideration had to be applied to the region at the tip of the crack in which the stress distribution flattens out owing to the maximum of the force-displacement curve, Figure 1. The only difference is that in the Griffith case the diameter of the non-Hookean region is of the order of the interatomic spacings, while in the brittle fracture of steel it is about twice the thickness $\underline{t}$ of the plastically deformed layer at the surface of the crack. According to the Inglis equation [13], the stress concentration factor is then

$$q = 2\sqrt{\frac{c}{t}}. \qquad [16]$$

X-ray measurements indicate(1) that $\underline{t}$ is of the order of 0.2 to 0.4 mm. in low carbon steels broken not too far above or below room temperature.

In the Griffith theory, the tensile strength of the specimen was obtained by dividing the molecular cohesion by the stress concentration factor. What is the quantity corresponding to the molecular cohesion in the brittle fracture of steels? The clue is given by the important observation(7) that in steels the crack does not propagate continuously: before it has broken through a grain boundary, unconnected small cracks arise in grains ahead of the tip of the main crack. This shows at once that the brittle strength of steel cannot have the order of magnitude of the theoretical strength (molecular cohesion); in fact, it must be quite low if independent fracture processes can start ahead of the main crack at points where the stress cannot be much above the yield stress. This may be due to the presence of numerous invisible cracks scattered in the material; or to the well known fact that plastic deformation can produce high microscopic internal stresses and subsequently crack formation. It seems that the cleavage strength of the material at the tip of the crack is not, or not much, higher than the ordinary brittle strength of steel obtained experimentally as the stress at which brittle fracture occurs. Since the cleavage strength of steel depends on the plastic strain which is difficult to estimate in the small region around the tip of a crack, only a rough idea of its magnitude can be obtained; it is probably somewhere between 100,000 and 200,000 lb. per sq. in. for a low carbon steel. For a tensile stress of, say, 20,000 lb. per sq. in., therefore, a stress concentration factor between 5 and 10 would be needed. If the thickness t of the cold worked layer in eq. [16] is taken as 1/100 inch, the necessary crack length

$$c = \frac{1}{4}t\left(\frac{\sigma_m}{\sigma}\right)^2 = -t \cdot q^2 \qquad [17]$$

is between 1/16 inch and 1/4 inch; for a tensile stress of 10,000 lb. per sq. in., the stress concentration factor is four times higher, and the necessary crack length is between 1/4 inch and 1 inch. These orders of magnitude appear quite reasonable in the light of experimental observations.

The last question to be discussed is whether eq. [1] represents a sufficient as well as a necessary condition of crack propagation. At this point a significant difference appears between the fracture, say, of glass and of low carbon steel. The stress concentration in glass is not limited by plastic deformation; in steel, however, the stress at the tip of the crack cannot exceed the yield stress multiplied by a plastic constraint factor which probably has a value between 2 and 3(1). If, therefore, the cleavage strength is higher than 2 or 3 times the yield stress Y in tension, the tensile stress at the tip of the crack cannot reach the fracture level no matter how high the stress concentration factor (i.e., no matter how low the applied stress is that can produce the highest possible stress 2Y or 3Y at the tip of the crack). An additional point of great importance is that the yield stress of steel increases with the rate of deformation more rapidly than the yield stresses of most metals; between the usual rates of "static" tests and the fastest rates at which measurements could be carried out it seems to increase by a factor approaching 3. It seems that, in typical cases of brittle fracture in low carbon steels, the velocity increase of the yield stress is the salient feature of the phenomenon. Although cleavage fracture can arise at slow deformation rates, it then requires so much plastic deformation for producing the necessary plastic constraint that the resulting cleavage fracture is anything but brittle; its energy absorption may be almost equal to that of a ductile fracture. Typical brittle fracture in a low carbon steel, therefore, can occur usually only after the crack propagation has reached a sufficiently high velocity; in laboratory experiments, the fracture is almost always initiated by some ductile (fibrous) cracking, accompanied by considerable local plastic deformation.

It can be said, therefore, that a characteristic feature of brittle fracture in ductile steels is the enormous decrease of the crack propagation work with increasing velocity of the crack. The crack propagation condition [1] may well be fulfilled for a rapidly running crack with its low value of p but not for a stationary crack, the propagation of which may require per unit of crack length, an energy of a higher order of magnitude. In such cases, cleavage fracture is initiated in laboratory experiments by large deformations producing strong plastic constraint and usually some fibrous cracking; the plastic deformation may have to extend across the entire specimen, so that the yield load has to be reached before cleavage cracking can start. After a cleavage crack has arisen, it may accelerate rapidly provided that the condition eq. [1] is satisfied, so that there is sufficient elastic energy released during the crack propagation to increase the kinetic energy around the running crack. In this sense, it may be assumed that eq. [1] represents the condition for the fast, and therefore, brittle, propagation of a cleavage crack. The initiation of the cleavage crack, however, may have to be done by ductile crack propagation not governed by [1] or any other brittle crack propagation condition derived from the Griffith principle [2].

It should be remarked that many service fractures seem to start without significant plastic deformation in spite of static loading. An interesting possibility for understanding this has recently arisen and should be discussed in a subsequent paper.

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THE CONDITION OF HIGH-VELOCITY DUCTILE FRACTURE

by

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1. PLASTIC-ELASTIC INSTABILITY

T. W. George(1) observed some time ago that a large sheet of thin aluminum foil, provided at its center with a knife-cut crack, burst under tensile load with a suddenness usually associated with brittle fracture. Although genuine cleavage fracture has never been observed in aluminum, and the tensile fracture of the aluminum foil was of the common ductile type preceded by necking, the phenomenon observed by Irwin and George has much in common with brittle fractures. When necking starts, plastic deformation ceases elsewhere in the foil; the deformation in the neck is confined to a narrow strip, the width of which is of the order of the foil thickness. When the crack starts to propagate along the neck from the ends of the initial knife-cut gash outwards, practically all of the work required for extending it is plastic work concentrated in a narrow belt adjacent to the outlines of the crack; Figure 1 shows by shading the plastically distorted zones of necking which later become the outlines of the propagating crack. Since the width of the distorted zone is small compared with the length of the crack, the plastic work per unit length of the crack outlines can be treated on the same basis as the surface energy of the crack walls in the Griffith theory; an exactly corresponding treatment for the brittle fracture of ductile steels has been given by the present writer(2). Consequently, the fracture of a thin ductile foil can be treated by means of the Griffith energy criterion, although the fracture mechanism is essentially ductile.

While the case of the aluminum foil represents an essentially brittle fracture with a narrow zone of plastic deformation playing the role of the surface energy in the Griffith theory, instances of essentially ductile fractures progressing with high velocity are also quite common. Tensile tests on ductile metals usually end with a sharp bang due to acceleration

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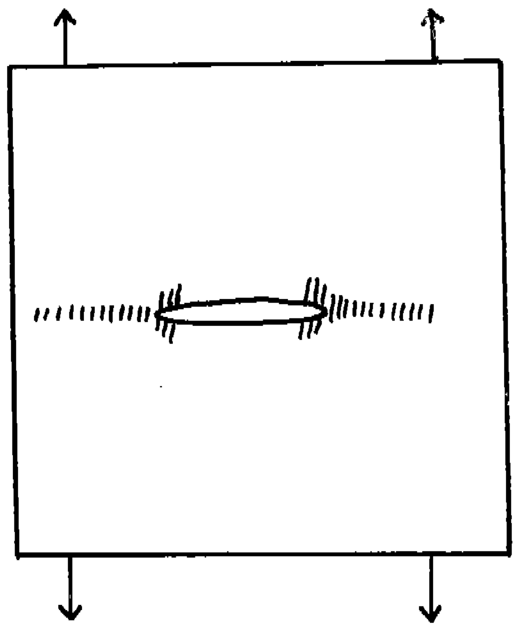

Figure 1

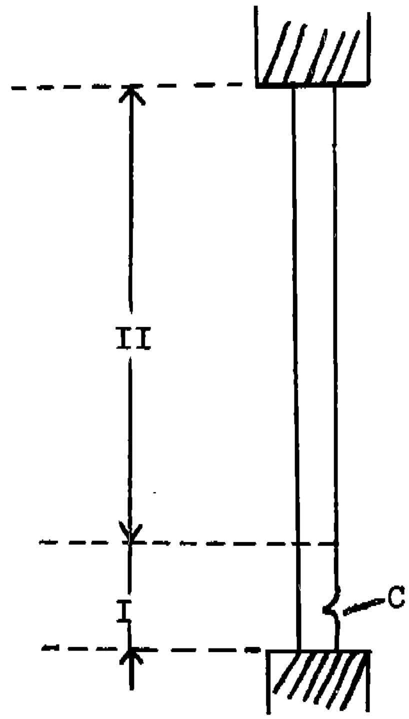

Figure 2

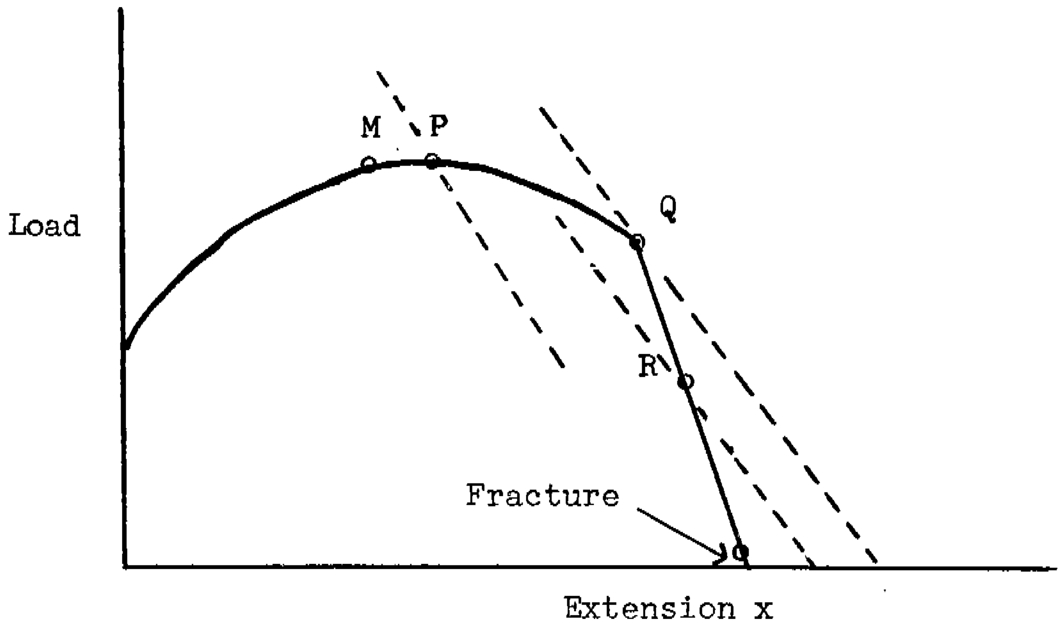

Figure 3

to high speed by the elasticity of the testing machine. If the specimen is very long, its own elasticity can produce the same effect.

The fact that the elasticity of the specimen can cause high-velocity crack propagation even if the fracture mechanism is basically ductile has led to the suggestion(3) that such processes would be governed by the Griffith energy criterion of fracture(4), according to which fracture occurs when the work dW required for extending the length c of the crack by a small amount dc is just covered by the accompanying release -dU of the elastic energy U stored in the specimen:

$$dW = -dU. \qquad [1]$$

This form of the Griffith principle applies to the case where the process of extension of the crack by dc takes place while the specimen is held between rigidly fixed grips. For the treatment of the case where the crack propagation is assumed to occur at constant load, and for a general discussion of the Griffith criterion, see reference 5.

On the other hand, it can be demonstrated(5) that the Griffith criterion is applicable only to essentially brittle fractures, i.e., to fractures where plastic deformation is either absent, or confined to a thin layer at the walls (in a foil, the outlines) of the crack while the bulk of the specimen is purely elastic. This can be recognized already from the circumstance that a typical ductile fracture, such as the cup-and-cone or the shear type fracture of a ductile metal, progresses by plastic deformation practically uninfluenced by the values of the elastic moduli. It would take place in the same way if the moduli were infinitely high; in this case, however, the right hand side of [1] would vanish, and the equation could not be satisfied.

Since the energy criterion [1] cannot be applied to essentially ductile fractures, the question arises, what is the condition for the self-acceleration of a ductile fracture by the release of elastic energy in the specimen or in structures connected in series with it, such as a testing machine?

Let Figure 2 represent a long tensile specimen in which, at the point C, a ductile crack, or a neck leading to cup-and-cone fracture, develops. Figure 3 gives schematically the load plastic extension curve of the specimen: its abscissa is the increase of the "natural length", i.e., of the length measured after removing the load and with it the elastic extension. Initially, the load increases; however, when necking starts, or when the crack has progressed to a certain depth, a load maximum (the "ultimate stress") is reached, and then the load drops with further extension. The maximum of the curve is a point of "plastic instability", beyond which the load required for plastic deformation decreases.

In addition to the plastic extension, the specimen also suffers elastic extension. The former is localized around the neck or the crack; to the latter, all parts of the specimen contribute. In the cases when rapid fracture driven by the release of elastic energy is likely to be noticeable the specimen is very long, so that the elastic extension of the plastically deforming region I is negligible compared with that of the purely elastic region II; this means that the spring constant of the specimen is identical with that of the elastic portion II and therefore remains practically constant during the propagation of the crack or the contraction of the neck. The elastic extension is then proportional to the load. Suppose now that the specimen is extended to the point P and then the grips are held rigidly fixed, so that any further plastic extension of I has to take place at the cost of an equal decrease of the elastic extension in II. In the course of this process, the stress must drop according to the elastic relationship,

$$dF = -C \cdot dx \qquad [2]$$

where dF is the change of the load, C the spring constant, and dx the increase of the plastic extension in region I, so that $-dx$ is the change of the elastic extension of region II. In Figure 3, the dashed line through P represents the elastic release of load that accompanies a virtual plastic extension of the specimen between fixed grips. With the assumed value of C, the load would drop more rapidly than the force required for further plastic yielding, so that the condition is stable and no plastic extension can take place unless the grips are moved apart. However, with further extension the point P, moving along the plastic curve, arrives at the position Q where the elastic load release line is a tangent to the curve. At Q, the condition of the specimen becomes unstable; any further extension leads to a point R at which the yield load drops more rapidly with further plastic extension than the load available after elastic release. Beyond Q, therefore, the specimen is unstable and fractures with high velocity under its own elastic tension. The point $\underline{Q}$ marks the beginning of the "plastic-elastic instability"; plastic instability, defined by the drop of the yield load, starts already at $\underline{M}$. The longer the specimen, the lower the value of the spring constant $\underline{C}$, and thus the slope of the elastic load release line. With decreasing $\underline{C}$, therefore, the point of plastic-elastic instability $\underline{Q}$ moves towards that of plastic instability $\underline{M}$; in the limiting case of an infinitely long specimen (or of infinitely high "elastic compliance" $1/C$ of the spring connected in series with the specimen), the two points coincide.

The point of plastic-elastic instability is of importance in testing: if a testing machine is not "hard" enough, elastic instability occurs soon after the load maximum, and the load-extension curve cannot be followed much beyond this point. Weight-loading and hydraulic machines, of course, have a tendency to "run away" already at the load maximum; however, if the machine is otherwise rigid enough, the descending branch of the load-extension curve can be followed at least approximately by the use of stops for interrupting the extension.

2. THE ANALYTIC CONDITION OF PLASTIC-ELASTIC INSTABILITY

The geometrical condition of instability explained in Figure 2 can be translated into an analytic form. Let $\underline{dW}$ be the plastic work of crack propagation or neck contraction during an increment $\underline{dx}$ of the plastic extension; if the yield force is $\underline{F}$,

$$dW = F \cdot dx$$

or

$$F = \frac{dW}{dx}. \qquad [3]$$

Consider now the purely elastic part II of the specimen (Figure 2); for simplicity, let it be assumed that all but an insignificant fraction of the elastic energy is contained in it. Since the specimen is between fixed grips, the extension $\underline{dx}$ of the plastic part I causes contraction of the elastic part by $\underline{-dx}$. If $\underline{G}$ is the tensile force in the elastic part and $\underline{U}$ the elastic energy,

$$dU = -G \cdot dx$$

or

$$G = \frac{-dU}{dx} \qquad [4]$$

If the specimen is in equilibrium, F = G and sc

$$\frac{dW}{dx} = \frac{-dU}{dx} \qquad [5]$$

or

$$dW = -dU \qquad [5a]$$

This is formally identical with the Griffith energy principle of brittle crack propagation [1]; but the meaning of [5a] is entirely different. It is merely an expression of Newton's third principle, stating equality of the forces acting upon the elastic and plastic parts of the specimen; it is satisfied identically from the beginning of the plastic crack propagation (or necking) to the point of plastic-elastic instability.

The tangent criterion of instability requires that the derivatives of the yield force $\underline{F}$ and of the elastic tension $\underline{G}$ with respect to the extension $\underline{x}$ must be equal, the differentiation being carried out at constant specimen length. With the values given by equations [3] and [4], the expression of the tangent criterion is

$$\frac{d^2W}{dx^2} = \frac{-d^2U}{dx^2}. \qquad [6]$$

It is seen that the criteria of rapid brittle fracture and of rapid ductile fracture, equations [1] and [6], respectively, are fundamentally different; the Griffith principle equation [1] does not govern high-velocity ductile fracture.

There is a significant practical difference between the two energy criteria [1] and [6]. Applied to brittle fractures, the Griffith criterion [1] leads to an expression for the tensile strength of a body containing a crack of given length. On the other hand, the tensile force required for ductile fracture cannot be obtained from the criterion [6] of plastic-elastic instability. The ductile breaking force is always the maximum of the load-extension curve, whether or not plastic-elastic instability with rapid fracture occurs. It has to be fed into the criterion, instead of being obtained from it; it can only be obtained as the force needed for producing the particular type of plastic deformation which ultimately results in crack propagation and fracture.

SUMMARY

The Griffith energy criterion

$$dW = -dU$$

(dW = crack propagation work, $-dU$ = released elastic energy) cannot be applied to essentially ductile fractures. In particular, it does not represent the condition of rapid ductile fracture propelled by the elastic energy of the specimen. The condition of such fractures is

$$\frac{d^2W}{dx^2} = \frac{-d^2U}{dx^2}$$

where x is the plastic extension accompanying the propagation of the crack.

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CONTRIBUTION TO THE DISCUSSION

by

Mr. G. M. Boyd
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This conference is both opportune and auspicious. Opportune because it comes at a time when the importance of "fracture mechanics" is becoming apparent, and auspicious because it will, we hope, place the stamp of respectability on that relatively new branch of science.

It might be said that fracture mechanics is not new, since it was originated by A. A. Griffith in about 1924, but this is only true in the sense that "nothing is new under the sun". The true birth of this approach was probably in the adaptation by George Irwin(1) of Griffith's ideas to the modern problem of brittle fracture.

Since then the subject has attracted many devotees, and has flourished. We must be careful, however, to prevent this tender plant from being choked by diffuseness, complication and digression--weeds which have beset and handicapped the whole problem of brittle fracture since the epoch-making episode of the "Schenectady". Let us keep our ideas clear, and be sure of our fundamentals as we proceed, and above all, let us keep our nomenclature precise and definite.

The first of the fundamentals which must be critically examined is Griffith's theory itself. This theory was devised for application to elastic materials which are incapable of plastic behavior. It was therefore permissible to evaluate both sides of Griffith's famous inequality simply by placing on one side the "surface energy", i.e., the energy theoretically required to create the two new surfaces, and on the other side the reduction in elastic potential energy due to a corresponding enlargement of the initial notch. This latter quantity could be calculated directly from elasticity theory, using certain plausible assumptions with regard to the shape of the notch.

This process has been adapted, by recent workers in the field, to the case of a material capable of plasticity simply by adding to the "surface energy" a quantity equal to the work done in plastic deformation. The other side of the inequality has been left untouched, apart from the use of different numerical factors to allow for differences in the assumed shape of the notch. This process logically leads, as Griffith's method did, to the idea of a "critical crack length", i.e., a certain length of crack above which the system containing it becomes mechanically unstable. This idea of a critical crack length is not, however, supported by the experimental evidence. It also led to the idea that, once such instability sets in, the rate of growth of the crack accelerates indefinitely. Appreciating that an infinite crack velocity was inconceivable, Mott(2) introduced another term to take account of the kinetic energy of the parts set in motion by the advancing crack, and by this means set an upper limit to the velocity, which he found to be that of sound waves in the material. Mott's additional term was, however, difficult to calculate; and moreover, it was soon found experimentally that actual crack velocities were far below that of sound.

In these circumstances it seems futile to go on tinkering with Griffith's theory, which is admirable in its proper place, and it seems preferable to go back to the fundamentals from which Griffith started and build upon them a structure consistent with recent experimental evidence and with the properties of the material now considered.

These fundamentals are somewhat as follows: If as a fracture progresses the potential energy of the system of which it is a part diminishes, then the system is unstable, and the fracture will progress until stability is restored. Otherwise, the system is stable, and fracture will not progress.

In order to determine whether a given system containing a crack, is stable or unstable we must evaluate the following inequality:

$$\frac{dF}{dA} - \frac{de}{dA} \neq \frac{dw}{dA} \qquad 1$$

in which, using the notation of Irwin, de/dA is the reduction in elastic potential energy due to an increment dA in the extent of the crack, dw/dA is the work which must be done against the resistance offered by the material to the extension dA, and dF/dA represents the external energy supplied to the system during the formation of the increment dA. If the left hand member of this inequality is the greater of the two, the fracture will progress. Otherwise it will not.

It is generally agreed and in accordance with experience that we are here concerned most with the case in which dF/dA is zero or negligible, i.e., the case in which a fracture can progress under the influence of the elastic energy alone, without the need for a supply of external energy. With this limitation we may write

$$\frac{de}{dA} > \frac{dw}{dA} \qquad 2$$

which is in fact Irwin's statement of the case with which we are concerned.

Coming to the evaluation of the two sides of this expression, it is apparent that dw/dA can be evaluated experimentally by methods such as that of Wells(3), and applied generally, provided that it is independent of the shape and extent of the fracture; of the shape of the body; and of the type of loading. It is at this point that present knowledge is sadly deficient and confused. Intuitively one suspects that dw/dA should depend strongly on these factors, but on the other hand indications are emerging that it is in fact independent of them. These indications emerge from the work of Robertson(4), Pellini(5), and the present writer(6), but they are as yet some way from being established. It would be extremely convenient if they were established, since the quantity dw/dA could then be regarded as a property of the material which could be determined by any convenient experimental technique. It must always be remembered, however, that this property is strongly dependent on temperature.

The left hand member of the inequality could presumably be evaluated by elasticity theory, provided that the entire stress distribution at each stage of progress of the crack were known. This would entail not only a knowledge of the geometry and loading of the body but also a knowledge of the actual shape of the "notch", i.e., of the crack, and in particular of the fracture front. Even with such knowledge the task of calculation would be formidable, except in some very simple cases.

Fortunately, there are indications among the available evidence which may considerably simplify the task. The first of these is that estimates of the total elastic energy in actual cases, and of the reduction in this due to small increments of fracture, show that in nearly all practical structures at normal working loads, there is more than sufficient elastic energy to propagate a crack if once initiated. It seems to follow, therefore, that the conditions for stability are not determined by the total elastic energy in the system but by the energy contained in a smaller region, the extent of which is at present unknown in the vicinity of the fracture front. This means that we should be cautious about what we mean by terms like "available energy." It seems that not all of the elastic energy in the system is "available", but at present little is known about what proportion is in fact "available." Several workers have considered the limitations on availability, for example by relating the velocity of elastic waves to the observed velocities of fracture or by estimating the volume required to balance the measured work done in fracturing. These efforts are as yet embryonic, but it is a promising line of thought.

The other experimental indication tending to simplify the evaluation of de/dA is that actual fractures tend to attain and maintain a constant "terminal velocity" which is considerably less than that of sound. Moreover, actual fractures tend to settle down soon after their commencement to a "steady state", which in the simple case of a fracture in a wide plate under tension, has been accurately described and accounted for theoretically(6). This suggests that of the total elastic energy "released" by the fracture only a small proportion actually intervenes in the process, the remainder being dissipated as shock waves, noise, heat, and the kinetic energy of the moving parts. It may be possible, therefore, to limit attention to that part of the released energy which in fact relates only to the propagation. This energy, however, must be equal to dw/dA, so it may be found sufficient to consider this latter quantity alone. If this happy conclusion were reached and established, it would only be necessary to find the critical value of dw/dA below which "unstable" or "brittle" fracture could occur and above which it could not. Recent work by Robertson, Pellini, Schaub(7) and others, much of which is still unpublished, strongly suggests that this is more than a bare possibility.

There is one crucial question which the Conference will no doubt consider, i.e.: "Is the character of the fracture influenced by the total elastic potential energy in the system before fracture?" The considerations given earlier in this contribution suggest that the answer is in the negative, but against this there is the evidence of DeLeiris(8) and considerable intuitive bias towards a positive answer. At first sight it appears that this central question has not yet been answered definitely by experiment, but in fact there are considerable difficulties, mainly due to the distinction between the separate phenomena of initiation and propagation of fractures. It seems clear that even with the sharpest conceivable initial notch a higher load is required to initiate a fracture than to extend one which is in progress. So much is this the case that one is led to think that initiation and propagation are controlled by separate properties of the material. This circumstance has been expressed in the idea that the resistance to initiation forms an "energy barrier" which must be overcome and which in many cases may protect the structure from catastrophe. On the basis of this idea one may conceive that the higher the barrier, the more catastrophic would be the fracture once it was overcome. In other words, this "initiation barrier" would itself determine the load and therefore the total elastic energy at the commencement of fracture. This is an interesting thought, and one which would probably repay investigation, but it would seem, on purely rational grounds, that such an effect would be limited to the early stages of the fracture, which should rapidly "settle down" to its natural form and velocity.

In conclusion, I wish to thank the organizers of the Conference for the opportunity to express these ideas, ill-formed though they are, and to wish the Conference the success which it augurs.

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DISCUSSION

by

Dr. J. M. Frankland
United Aircraft Corporation

In the examples of brittle failure of which we have heard, there is always present some degree of ductile behavior, even though it may be highly localized. This suggests that we may be faced with two simultaneous mechanisms, one of which is a ductile process and one which is a brittle process. It is possible that brittle and ductile failures might not then be essentially different but should rather be viewed as the extremes of a single type of behavior.

Support for the concept of a dual mechanism in failure can be found in ductile fractures of metals. For example, the average stress at fracture in circumferentially notched round tensile specimens increases at first with the severity of notch, reaches a peak value, and then falls as the notch becomes sharp. An example of this is shown in Figure 1. The data are for 75S-T6 aluminum alloy and have been taken from NACA Technical Note 1831, by Dana, Aul, and Sachs. The abscissa is "notch sharpness", defined as the ratio of radius of the minimum section divided by the root radius of the notch.

This behavior can be explained on the assumption of a maximum shear stress criterion for failure, which Dorn and Thomsen (Journ. Aero Sciences, 1944) have shown to apply for magnesium. The stresses at failure in mildly notched specimens can be approximated by Bridgman's formulas for the stresses in a necked tensile specimen. The dashed curve shows what Bridgman's theory predicts for the average normal stress at failure when the maximum shear stress is constant. There is initial agreement with the experimental results at low values of the notch sharpness. Near the maximum fracture stress the character of the failure changes from initial fracture on the axis to fracture starting at the surface. Apparently, at the sharper notches the plastic flow has not been sufficient to wipe out entirely the original elastic stress concentration at the surface.

Now Bridgman's theory says that the maximum shear stress is constant over the cross section. Why, then, should fracture occur always in the interior in the case of very mild notches? I suggest that microfractures are nucleated by shear stress, and that the hydrostatic tension component of the stresses (greatest at the axis) causes growth of these microfractures to produce gross failure. Microfractures of this character have been observed by Dr. Tipper in mild steel.

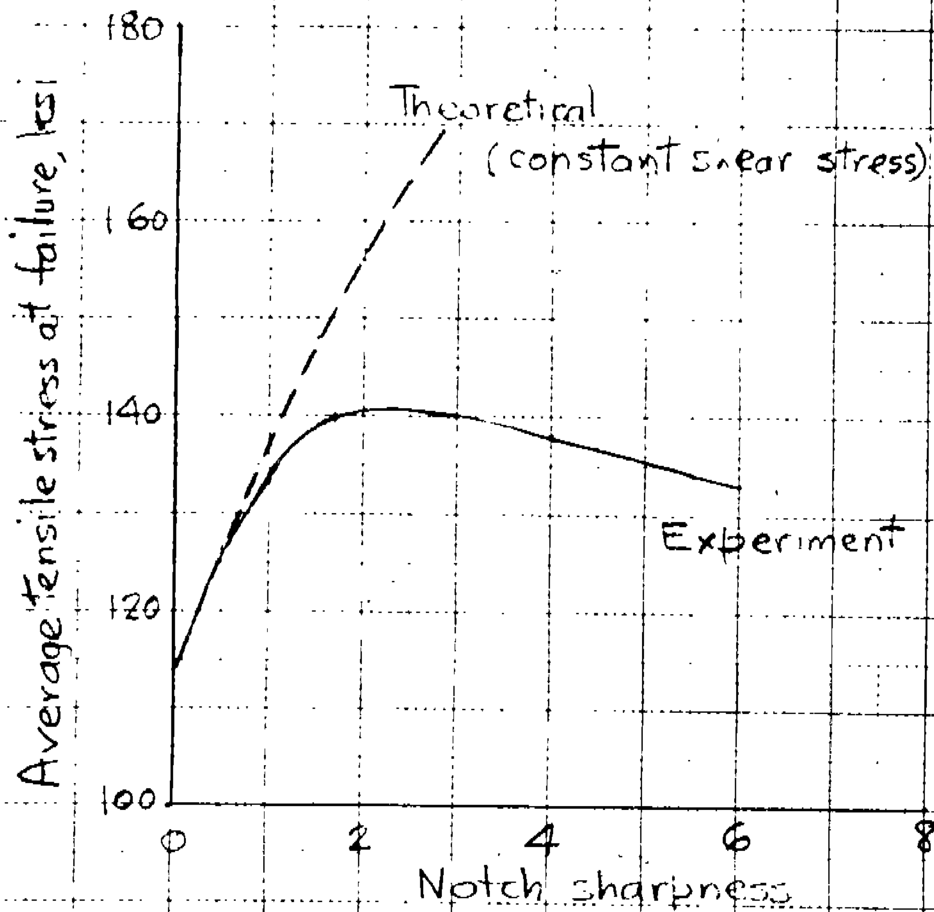

Fig. 1
Notched Tensile Tests of 75S-T6

Crack initiation in fatigue is often found to start along planes of maximum shear, and then after a short distance to turn into planes of maximum tensile stress. Static fractures normal to a principal tensile stress are sometimes on closer examination found to contain many small forty-five degree facets. These observations would be explained by a ductile shearing mechanism for initial local fracture followed by crack growth induced by hydrostatic tension. This crack growth process would be of a brittle character. The failure would be called ductile if crack growth started only after appreciable plastic flow had already taken place.

The development of unusually high ductility and strength in tensile tests under hydrostatic pressure might be explicable also on this hypothesis.

DISCUSSION

by

Professor W. R. Osgood
Illinois Institute of Technology

First, two general remarks occur to me. In all tests in which an initial crack is produced by cooling the specimen some 400°F. below room temperature and then bringing the specimen up to the temperature of the test, I raise the question whether the mechanical properties may not have been changed by the drastic cooling itself. It seems possible that differential contraction and expansion of the constituents of the metal might give rise to plastic flow and even microscopic cracks which might have a bearing on the results of subsequent tests. It would be interesting and perhaps informative to make fatigue tests on identical specimens some of which had been drastically cooled and others not.

Residual stresses should not be forgotten or left out of account. The presence of residual stresses in a structure, for example, may explain or help to explain discrepancies between computed stresses in the structure and test stresses with which they are being compared. In any event, ignoring residual stresses because one cannot determine them does not get rid of them. They must be reckoned with just as any other stresses are.

The "Report on Brittle Fracture Stresses", by F. J. Feely, Jr., D. Hrtko, S. R. Kleppe, M. S. Northup, presents a new technique that has much to commend it. Deductions from tests of specimens with real cracks raise fewer questions than those drawn from specimens with saw cuts or other simulated cracks. There follow a few specific comments and questions on the paper.

Stress Gradient Test

The formation of thumbnails observed here was probably, as the authors suggest, accompanied by shifts in the pattern of stress; but the thumbnails suggest also, in accordance with Irwin's theory, delays awaiting the arrival of sufficient energy to continue propagation of the crack.

S. O. D. Test

The last paragraph of this section may be very significant. Presumably the "critical value" of the temperature below which identical results are obtained is a temperature in the transition range--perhaps it is the best definition yet of the transition temperature. From a fundamental point of view experiments with a number of impacts, both below and above the critical temperature, and with metallographic examinations of their effects may shed light on the mechanics of fracture.

Notch Sharpness and Length

Was the radius at the root of the saw cut .020 inch or was the root merely formed by an ordinary saw?

It is not clear that the ratio of length of crack to width of specimen has no effect. This may be true up to some length of crack below which the energy available is ample to propagate the crack. As the initial, induced, crack gets longer, the excess energy available becomes less. It is suggested that in the tests cited, with the longer cracks, the breaking stress may have been found to be higher because less energy was available in the specimen for propagating the crack--according to Irwin's theory.

Impact

In the fourth paragraph the question still remains "What initiates propagation of a crack in a tank failure?" Could it be working of the material by changes of temperature and by changes of load, aided and abetted by residual stresses?

Geometry and Size

Width Effect

How long was the initial crack? Figure 11 appears to be consistent with Irwin's theory.

Length Effect

This effect seems to tie in directly with Irwin's theory. It is not inconsistent that below the critical temperature the breaking stress becomes independent of length, at least for lengths longer than the shortest tested, 24 inches or so. (For shorter specimens dependence on length might be found.) Below the critical temperature the energy required for fracture is so low that the necessary energy available can be stored in a short length of specimen.

Size Effect and Shape Effect

Figure 13 again would appear to be consistent with Irwin's theory, although scatter might be found at large ratios of length of induced crack to width of specimen.

In their presentation of "The Application of Notched Tensile Test Data to Engineering Design", E. M. Lape and J. D. Lubahn point the way to a more fundamental approach to an understanding of brittle fracture than the purely phenomenological approach largely used in the past. With the possible exception of the Standard Oil Development transition temperature, all transition temperatures have been arbitrarily defined, almost as arbitrarily as the yield stress of a material. For a better understanding of brittle fracture, it would seem to be highly desirable to relate the transition temperature and associated phenomena to mechanical (and possibly other physical) properties of the material. This, Lape and Lubahn have begun to try to do, and it is a very much worth-while effort. Incidentally, B. G. Johnston a few years ago started the study of a possible relation between brittle-fracture strength and reduction of area in the tensile test. The study was discontinued, not for lack of promising results, but for lack of money.

G. R. Irwin in his paper, "Fracture Dynamics and Fracture Strength of Large Welded Structures", has, in my opinion, made the greatest contribution to date toward an understanding of the mechanics of brittle fracture. It seems to me that in other work he and his associates have done (THE WELDING JOURNAL, Research Supplement, February 1952), perhaps a better case is made for the intersection of curves like those of Figure 3 than in the present paper. Do not the dW/dA-curves of Figure 3 include the kinetic energy released? The statement is made "The strain energy was subtracted in computing these", but as I understand it, this energy was taken as the difference between the energy available before a fracture dA and the remaining energy available after the fracture dA. This difference would include the energy lost as kinetic energy. If so, the dW/dA-curves as computed are too high. In the brittle fracturing of ships the strain energy converted to kinetic energy would seem to be many times the energy absorbed in the process of fracturing. I strongly recommend that Irwin's work be supported munificently.

DISCUSSION

by

Dr. Edward Wenk, Jr.
David Taylor Model Basin
Navy Department

The papers themselves certainly provided an interesting and stimulating commentary on the brittle behavior of metal structures, and I think the authors and the two discussors are to be congratulated on their very fine presentations. As well, the Chairman and those who planned the conference deserve our thanks for a well-organized and executed meeting.

The subject itself is admittedly baffling; if that were not true there would hardly be cause for such a meeting some eight or nine years after investigations into the field were initiated. In providing any comment, the discussor is in the position of himself not having personally contributed to this endeavor. Serving as an intensely interested bystander, however, and as one directly engaged in the field of structural mechanics of ships, he feels as keenly the need for a general understanding and solution of the brittle fracture problem as those who are intimately participating in the research. He has either the advantage or the disadvantage of only knowing what he has heard reported. He has not had the opportunity of personally checking either the completeness or the validity of the data or for carrying out any further investigations which he feels may throw light on the subject. In that situation he is overwhelmed by the staggering mass of data already collected and somewhat confused by the absence of any comprehensive summary or sorting of the facts. As a consequence he feels a very strong thirst for the separation of the material into various pertinent categories of information and secondly an evaluation and comparison of the available facts with proper recognition of the many apparent discrepancies which now devastate any general hypothesis of fracture.

At the outset it would seem that the most important data would be those from the full-scale casualties themselves. Any hypothesis which would be advanced for an explanation of brittle fracture should be consistent with ship or pressure vessel performance. Such information, although generally available, was not made the topic of any presentation on this basis.

Even before all of the data are so organized to permit comparison of laboratory with full-scale data, a summary of certain observations described in the symposium reveals sharp discrepancies. If the discussor recalls correctly, the Standard Oil tests indicate notch sharpness influences fracture stress whereas at one time opinions were offered that this was not an important variable. These tests also appeared to indicate that wide specimens were relatively stronger than narrow ones, contrary to other results presented by Capt. Roop. These S.O.D. results also appear to indicate length of crack to be insignificant which disagrees with the Griffith theory. In one case the opinion was expressed that the reservoir of energy in the ship was important whereas Professor Orowan offered some remarks indicating that in a brittle material enough energy was available in a ship one-inch long to propagate a crack. Professors Hechtman and DeGarmo felt strongly that evidence supports the development of moderate plasticity (4--5%) near brittle fractures so that at least this should be developed in structures to be satisfactory whereas Professor Orowan indicated such was not observed in fiberglass structures which demonstrated no brittleness yet fail without much local flow. There was also expressed by Mr. MacCutcheon the opinion that a theory of failure should make appeal to the energy level whereas Professor Drucker indicated that such an averaging process should be discarded in favor of a criterion based on stress alone. A good deal of discussion revolved around results from plates with central notches whereas another opinion was expressed that such a geometry was sufficiently unlike that of a ship that such results would be invalid.

A number of similar apparent differences in interpretation developed in the course of discussions, but these alone are not necessarily disturbing in view of the fact that the data which may be in conflict were not collected under identical test conditions. That is to say, all of the data presented may have some bearing on the brittle fracture problem.

<u>The discussor does believe, however, that when new data are developed the investigator has some obligation for commenting on discrepancy with earlier work</u>.

Admitting from these results that nature has behaved in an apparently mischievous way, it has been observed there would appear to be one other element which thus far seems unresolved. In most cases, engineers think of failure as occurring when the stress has reached a certain ultimate level. Except in the important case of fatigue it is presumed that failure occurs the first time this level is reached. With ships, at least, there is some indication that many if not most of the failures occurred when the nominal load level was low or moderate; from Mr. MacCutcheon's graphs only very few ships failed when in heavy seas. Yet it has been admitted that those failing under moderate load conditions almost certainly had experienced more severe loading at an earlier time. The discussor may have missed this point, but he doesn't recall much discussion of this particular feature of failure other than by Mr. Feely. That is to say, there has been no mention of the possible cumulative effects of damage which might bear on the explanation for observed capricious behavior.

In evaluating the problem, it would appear as though this question of the load which was associated with actual fracture would be of extreme importance. It is understood that all possible information has been extracted from ships' logs, etc., related to the loading history and condition at failure in ships. Such an analysis has up to this time produced only the general observation that in no known case was the nominal stress <u>above</u> the yield point. This, of course, is an important result if only that it is to be compared with laboratory data where, up to the S.O.D. tests, in no known case was the stress <u>below</u> the yield point. It is suggested that perhaps those data available from pressure vessels, particularly pipe lines, may facilitate a correlation between the fracture and load which is believed to be ultimately necessary for any practical application of scientific information.