NRL Report 6258

# Metallurgical Characteristics of High Strength Structural Materials

[SIXTH QUARTERLY REPORT]


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## ABSTRACT

A progress report covering research studies in high strength hull structural materials, conducted in the period September 1964 to December 1964, is presented. Included in the report are fracture toughness studies on: (1) titanium alloys as related to processing variables (including diffusion-bonding), thickness, and heat-treatment, (2) aluminum, (3) high strength steels relating primarily to the 5Ni-Cr-Mo-V steels, and (4) a nickel-beryllium alloy. Low cycle fatigue crack propagation studies on several titanium alloys are presented as well as a general review of the status of the titanium casting industry. The U.S. Naval Research Laboratory's new 240,000 ft-lb laboratory drop-weight bulge testing facility is described and a preliminary correlation of the results obtained with it and the explosion bulge test are presented. The results of a study using fatigue cracked and side notched Charpy specimens in impact and slow bend testing on a wide variety of steels is given. The results are compared to those obtained in the standard Charpy test using the same materials. A background statement describing the "philosophy" of testing and evaluation of high strength structural materials which is the basis of the program is included in this report.

## PROBLEM STATUS

This is a progress report; work is continuing

## AUTHORIZATION

Projects SR 007-01-02-0704, RR 007-01-46-5405, and WW 041-63RO5-24A NRL Problem MO1-05

Projects RR 007-01-46-5420 and SR 007-01-01-0856, and WW 041-63RO5-24B NRL Problem MO1-18

Projects SR 007-01-01-0850 and SR 007-01-01-0854 NRL Problem MO3-01

Manuscript submitted February 15, 1965.

# METALLURGICAL CHARACTERISTICS

# OF HIGH STRENGTH STRUCTURAL MATERIALS

(Sixth Quarterly Report)

## INTRODUCTION

This is the sixth status report covering the U.S. Naval Research Laboratory Metallurgy Division's long-range program of determining the performance characteristics of high strength materials. The program is primarily aimed at determining the fracture toughness characteristics of these materials using standard and newly developed laboratory test methods and at establishing the significance of the laboratory tests for predicting the service performance of the materials in large structures. The program is aimed at Navy requirements but the information that is developed is pertinent to all structural use of these high strength materials. Quenched and tempered (Q&T) steels, maraging steels, titanium alloys, and aluminum alloys are the principal high strength materials currently under investigation.

Since these status reports are now receiving a much broader distribution than they have in the past, a background statement which considers the evolution of engineering principles for optimization, design selection, and specification of new high strength structural materials is presented. The background statement also presents the "philosophy of testing and evaluation" which is followed in investigating these materials to provide useful and meaningful information at the earliest possible date.

The effect of processing variables on the fracture toughness of a Ti-7Al-2Cb-1Ta alloy is evaluated with the U.S. Naval Research Laboratory's drop-weight tear test (DWTT). The results show that sizable variations in toughness and strength are possible with this alloy.

through careful control of processing variables and that the full thickness DWTT is most useful in showing the variations in fracture toughness with processing. The effect of heat treatment and diffusion-bonding on the strength and toughness properties of some titanium alloys is presented along with some fracture toughness measurements on 2-in. thick titanium alloy plates.

The results of fracture toughness studies on aluminum alloys as developed to date are reported and a preliminary fracture toughness diagram is presented. The results presently show that most of the aluminum alloys studied are capable of developing high levels of plastic strain in the presence of flaws, even at fairly low DWTT energies.

A review was made of the 5Ni-Cr-Mo-V steel data developed under the Bureau of Ships 130-150 ksi yield strength hull steel contract. From this, it is seen that high quality electric furnace practice using two oxidizing and one reducing slags resulted in a material of high fracture toughness. Explosion bulge tests (performed at NRL) of undermatching, matching, and overmatching weldments of the material at the 140 ksi strength level (base plate) showed that the heat-affected-zone (HAZ) was not subject to as great a straining as was the unaffected plate material. These tests were aimed at defining the relative strain conditions in mismatched weldments of the 5Ni-Cr-Mo-V steel.

A brief study of the fracture toughness characteristics of a Ni-1.95Be-0.5Ti alloy (Berylco-Nickel 440) was made at two extreme levels of yield strength (47.8 ksi and 177.5 ksi). The study indicates that the alloy can develop a high level of toughness, as measured by the DWTT, at both strength levels and since simple heat treatments can be used to develop intermediate levels of strength and toughness, a more thorough investigation of this and related alloys could be of interest.

A new laboratory drop-weight bulge testing facility at NRL is described in which the capacity of the drop-weight machine (240,000 ft-lb) exceeds that available in the NRL explosion bulge test facility which is limited to the use of 7-lb. Pentolite charges. The facility was developed to evaluate weldments in steels, titanium, aluminum, and other high strength metals in the thickness range of 1 - 2 inches. Preliminary drop-weight bulge tests of 1-inch thick 150 ksi yield strength steels are presented and are correlated to the results obtained in the standard explosion bulge test.

Low cycle fatigue crack propagation studies of Ti-6Al-4V, Ti-7Al-2Cb-1Ta, and unalloyed titanium are presented. From these studies, it is concluded that the growth rates of low cycle fatigue cracks in the materials follow the form

$$\Delta L/\Delta N = K(\epsilon_T)^n,$$

the growth rates in the two alloys are very sensitive to small changes in applied strain and become quite rapid at cyclic strain levels below the proportional limit. In a salt water environment there is considerable effect on the crack growth rate of the Ti-7Al-2Cb-1Ta alloy.

A review of the titanium castings industry is included in this status report which points out the capabilities and limitations of the industry as is known today. This is a preliminary to possible future work in surveying the fracture toughness characteristics of titanium alloy castings.

The use of the fatigue cracked Charpy and side notched Charpy in impact and slow bend testing has been investigated for a variety of high strength steels ranging in yield strength from 80 ksi to 280 ksi. Correlations between the standard Charpy V and these modified tests have been determined. The results of the study show that no immediate advantage is gained by using the modified tests in lieu of the standard Charpy V test, and that discrimination between the steels is drastically reduced, especially at the lower levels of toughness, when fatigue notching procedures are used.

## BACKGROUND STATEMENT:

## CONSIDERATIONS IN THE EVOLUTION OF ENGINEERING PRINCIPLES FOR THE OPTIMIZATION, DESIGN SELECTION, AND SPECIFICATION OF HIGH STRENGTH METALS FOR NAVAL STRUCTURES

(W.S. Pellini)

As a consequence of a greatly increased interest in Navy, Department of Defense, and industrial circles in military structural application of high strength metals, the publication of the sixth quarterly report of this series is being provided to a considerably expanded distribution list of interested parties. Accordingly, these reports, which began as a method of quick dissemination of information to a limited circle of Navy specialists fully conversant with the general aims of the studies, now require redefinition of general aims to a much larger audience. Basically, the subject U.S. Naval Research Laboratory's program resulted from an assignment by the U.S. Bureau of Ships for studies to define engineering criteria by which the practicability of fracture-safe design of naval structures based on the "new" classes of high strength metals could be defined. It is important to recognize that the term "naval structures" is not restricted to a particular type of pressure hull for submarines (these may range from simple spheres to complex, large, "roughly" cylindrical structures) but includes ship hulls, hydrofoils, large internally loaded pressure tanks, large deep-ocean civil engineering structures, etc. In other words, the desired information should be such as to have general parametric applicability in design trade-off considerations for any type of structure. The principal difference, with respect to the use of these materials in naval applications, is that the general size of the structures and the thickness of the materials tend to be on the "massive" side and severe environmental conditions may be encountered. For the most part, high strength metals have been used previously in rather thin sections and under rather carefully controlled environmental conditions. The history of these previous applications also included full scale-up fabrication of sections of prototypes which were tested to destruction and the "weak" points redesigned until optimum performance was obtained; viz., rocket cases, aircraft forgings, etc. In contrast, the naval architect is often involved with structural problems which may exclude such approaches and simultaneously require meeting of severe service conditions including the development of localized or general plastic overloads. In many complex naval structures, localized plastic overloads are to be expected if reasonable general structural efficiency is to be attained and if shock loads are applied. The possibility of explosion attack resulting in plastic overloads is another matter of possible parametric consideration. If it is possible to define the level of stress (elastic or plastic) that critical structural element positions can withstand in the presence of flaws or cracks, we can then proceed to select materials within whatever particular limitations the designer is required to accept. In other words, the essence of the studies is the evolution of the parametric matrix of metal capabilities that apply not only to an idealized laboratory test element but also to a practically fabricated article, including weld joints and complex geometries.

We should remember that large structures built of low strength steel have failed catastrophically and approximately fifteen years of applied research was required to evolve solutions (1). In this case, the problem was simply one of "transition temperature" -- in other words, a single metallurgical failure mechanism (cleavage) was operative and such failures always occurred by crack propagation through the base metal. With the advent of quenched and tempered steels, failures have been noted which may be considered "anomalous" in the sense that a variety of other fracture modes -- fracture propagation through welds, heat-affected-zones (HAZ), or fusion line (2) -- may be encountered. Hydrogen embrittlement, stress corrosion cracking, and other environmental factors plagued the applications to aircraft and rocket cases. Age-hardened aluminum alloys have developed catastrophic failures in naval service due to metallurgical conditions that could be traced entirely to the "softened" HAZ of weldments. This listing is not meant to be all-inclusive but to highlight the fact that anomalies or "surprises" may be expected in the applications of these new metals and it is best to discover these in the laboratory rather than to be surprised in service. In projecting to the use of these new materials, we are faced with problems of greater complexity (because of the variety of failure mode possibilities) than for the brittle fracture problems of low strength steels.

The U.S. Naval Research Laboratory Metallurgy Division's program in high strength structural metals is but a part of the Navy-wide research effort in this field, but it is a crucial part because it is concerned with the nature of failure mechanisms and with the development of practical test methods for the study and evaluation of the inherent resistance of the various metals to their particular set of failure modes. The subject progress reports represent a gradual evolution of the parametric structure by which these analyses can be made; coupled to these studies there are other investigations covered by a quarterly report (3,4) series (recently initiated) dealing with corrosion aspects.

There is no escape from the complexity of utilizing high strength metals, simply by a switch to nonmetals or to glass composite structures. All of these have their crucial relationships between the "laboratory defined" strength of materials and the effective strength of a structure. These aspects have been generally reviewed and considered by the Undersea Technology Panel of Project Seabed (5). For many of these materials we are in the unfortunate stage at which "design confidence in their utilization is inversely proportional to the technological knowledge that has been accumulated." In other words, we do not know enough to realize what we should be concerned about. This may be described as Stage 1. Stage 2, which generally follows, involves disenchantment with the material because of difficulties that become apparent, and Stage 3 then brings us to building sound technical bases for use of the materials by the proper combination of design and fabrication techniques, within realistic limits. It is only at Stage 3 that "design confidence becomes proportional to knowledge." For example, for quenched and tempered steels of the HY-130/150 ksi yield strength range we are in Stage 3; for maraging steels in thick sections we have entered Stage 2; for thick titanium alloys we are approaching Stage 2; for glass and fiber-reinforced-plastics (FRP) we are in Stage 1. The particular stage may be 1, 2, or 3 depending on size and complexity of the structure contemplated. These estimates (the author's) are roughly qualitative and intended only for emphasis of a point of philosophy, without which we can readily confuse R&D "imagineering" with the practical realities of true engineering capabilities.

It is most important that we "stand off" and take a good perceptive look at where we are technically and the directions that are indicated for R&D in this field. There is much confusion developing from the "race" to design and build structures utilizing these new materials. Claims, counterclaims, partial information, and misinformation prevail -- the temptation is considerable to treat these materials as a general class subject to definitions covered by a few simple generalities. In fact, we are dealing with highly "individualistic" materials comprising several distinctly different families of quenched and tempered steels, maraging steels, titanium alloys, and aluminum alloys. The problem of utilizing these materials in "fracture-safe design" cannot be resolved solely from simple tests of selected "base material" meaning the plate or forging, as produced experimentally. Processing, fabrication, and design configuration variables have important interrelationships -- it is essential that these be understood in a general sense before much more can be said. As a start, let us consider the three major classes of "discrimination" factors which must serve as the starting point for any and all analyses of suitability:

(1) Structural discrimination factors: Is the contemplated structure subject to rigorous stress analysis or is it complex with a requirement for plastic hinge readjustment at points which are relatively "stress indeterminate"? Certain aerospace and aircraft components have been built with "near perfection" design and fabrication techniques. The ground rules which can apply to materials selection for such structures cannot be translated to more conventionally designed and fabricated structures. Moreover, as size increases the retention of "perfection" design and fabrication procedures become impossible or temporarily beyond the "state of the art".

(2) <u>Materials discrimination factors</u>: Under certain conditions (for example -- no welds) the base plate is the "discrimination factor" with respect to fracture-safe design. To the limits that structural design and fabrication permit "rigorous", "approximate", or "indeterminate" definition of the flaw size and stress, the analyses of required fracture toughness can be restricted to base metal characteristics. However, if welds are present, the crucial discrimination factors may become the weld, the heat-affected-zone (HAZ), or the fusion line zone. In other words, conditions may exist for which the fracture-safe, allowable local and general stress levels for the base material are greatly in excess of those permissible for the weld zone area.

(3) <u>Environmental discrimination factors</u>: For single cycle loading, fatigue is not a factor; however, for repeated loading, low cycle fatigue may play an important role in flaw generation and growth. Thus, the initial flaw size condition of the structure may not be the discrimination flaw size for fracture-safe design. In the presence of moisture or seawater, the rate of fatigue crack growth may be greatly increased; moreover, the critical stress for fracture initiation may be grossly decreased. These two factors can work in concert or separately; also, the response of the base metal may be specifically different from that of weld zones in these respects.

The permutation of these factors may suggest that we are involved with an "impossible" problem of utilizing high strength metals. This is not the case for the following simple reasons:

(A) For some modestly high level of strength (for any given metal), high fracture toughness will be "inherent" (built into) in the metal and the weld zone and environmental aspects will be of no significant consequence for a reasonable design. This simply means that relatively complex structures (within reason) can be constructed with relatively conventional fabrication and fracture safety will be assured. The "guarantee" comes from metallurgical accomplishment.

(B) For some intermediate (somewhat higher) levels of strength, the same can be said (as above) for the base material but not for the weld zone. Discrimination analysis may suggest location of welds at regions of low stress, etc., or elimination of welds. In this respect, it should not be forgotten that improved welding techniques are being evolved to cope with this problem. In fact, this is an area of utmost fruitfulness and high potential payoff. The "guarantee" must come from a combination of metallurgical, design, and fabrication factors.

(C) For a much higher level of strength (the highest range) the base metal becomes highly fracture sensitive at flaw size-stress combinations that are difficult to avoid and environmental factors become highly controlling in addition. These materials can and are being used within their limitations -- however, it is most important to recognize that perfection of design, flaw size control, fabrication finesse, and precise control of environmental aspects are of paramount importance. The "guarantee" must come from design and fabrication factors with possible severe penalties in factors of safety that must be utilized.

Now, a great deal of the effort conducted under the present program relates to characterizing the various new materials in a "first cut" approach as to strength level limits related to categories (A), (B), or (C) as outlined above. In these respects, it should be recognized that fracture toughness characteristics of a metal at a given level of strength are highly sensitive to process history, thickness, welding techniques, etc. Thus, the "first cut" approach must be aimed at evaluating the best that can be done for specific ranges of strength within realistic limits of production consistency attainability. The importance of understanding the virtues or limitations of the best materials within specified strength ranges lies in the prevention of the application of undue restrictions on design and fabrication (more perfection than is necessary thus leading to an expensive structure) for basically fracture tough materials. Obviously, there is the other possibility of using less than the required restrictions for materials of low fracture toughness. Another reason lies in the elimination of much misused "factors of safety" solutions which are self-negating and lead to mismatch between attained structural efficiency and the cost for the attainment. An example of this situation may be found in the construction of large diameter solid propellant booster cases of 250 ksi yield strength (nominal) maraging steel. At this strength level the steel (approximately 3/4-in. thickness) has a very high sensitivity for fracture initiation at high stresses (relative to yield) from very small flaws. If used at low hoop stresses (say approximately 0.6 of the biaxial yield strength), a relatively large flaw is tolerable, provided no adverse environmental effect is present. Now, the point is that fabrication of 250 ksi steel is much more difficult and the cost is higher than say for a 150-160 ksi steel. At the lower strength level, the steel and weld can be metallurgically constituted to have high fracture toughness which in turn allows hoop stressing to very high stresses exceeding the biaxial yield strength and environmental effects are relatively minor. The choice of either a high factor of safety applied to a high strength brittle material or a very low factor of safety applied to a lower strength, fracture tough material does not necessarily lead to a structure of equivalent fracture safety. In fact, environmental aspects considered, higher reliability could be attained with the lower strength steel. These factors become much more important for structures which are expected to have long life and which deviate from the utmost simplicity of design inherent to simple spheres or cylinders.

The foregoing discussions were intended as "prologue" for definition of the aims of the studies reported in this and in the previous five progress reports. The primary aims may be summarized as follows:

(1) To evolve frame-of-reference ("yardstick") categorization of the relative fracture toughness of competing high strength material -- from very low to very high levels of strength.

(2) To use this information for purposes of determining trends and directions in optimizing the strength-fracture toughness relationships of the various materials. That is to say -- as R&D direction guidelines.

(3) To evolve information of early usefulness to designers faced with near-term decisions in the applications of these materials. In this respect, "time is of essence" and data interpretation finesse is of secondary importance. Finesse in these respects will evolve with time -- for the present, the attainment of reasonably interpretable engineering information is the first order of business.

It is inconceivable that near-term design decisions for thick walled structures will be made with data of high discrimination finesse -- there simply is not enough time to develop such detailed experience. It is more conceivable that decisions must be made on relatively little data and the temptation will exist to proceed on this "too little" data basis, particularly for thick sections. The motivation of this program is to establish a reasonably sound basis for evaluation of relatively large size, thick walled structural elements of materials for which there is no prior experience in practical applications. Two opposing philosophies have to be contended with -- one involves projection (guessing) what the properties may be, candidly referred to as "extrapolation" -- the other involves "ivory tower" promises of the ultimate in R&D (given enough time) when everything will be known exactly. A moderate middle course program is of essence -- in this respect, the scope and magnitude of the subject NRL effort is at the maximum practical rate. To date, a rational exploration has been made for materials of 1-in. thickness; as rapidly as possible, we are extending these studies to 2-in. thickness and then to thicker sections. The problems of procuring such materials are severe, both from the point of view of commercial availability (many are considered experimental materials) and from the point of cost required to obtain samples representative of large scale production. For example, a test of a 3-in. thick plate of say 12-in. x 12-in. laboratory produced dimensions is meaningless insofar as representing a production plate.

As noted previously, a companion quarterly progress report involving corrosion (stress corrosion cracking, cathodic protection, etc.) aspects of high strength material is issued separately. In many cases the same materials reported in the present series of reports are "transferred" to the NRL group involved with corrosion studies.

The potentials of plane strain $K_{Ic}$ (linear elastic analysis -- fracture mechanics) categorization of material is being investigated. These procedures have been established for relatively brittle, ultrahigh strength metals. Unfortunately, indiscriminate use of these test methods for intermediate and high strength materials of higher fracture toughness has been made in some circles -- the results have been confusion with report quotations of invalid $K_{Ic}$ numbers -- qualified by the meaningless generality that these are "lower bound values". The present studies in fracture mechanics under this program are aimed at "working down" from the brittle range to the semi-tough range so as to establish the strength limit below which such linear elastic analysis determinations are no longer valid. In effect, this means establishing an OMTL (optimum materials trend line) based on plane strain fracture toughness. Effects of environment (say salt water) on $K_{Ic}$ determinations presently seriously confuse the issue of tolerable flaw sizes for materials which initiate and propagate fractures at stresses below yield. As of the present report, we are yet unable to present valid $K_{Ic}$ data for reasonably fracture tough materials under test, because of test interpretation difficulties. It is hoped that such data will be available for the next quarterly report, at least for materials of relatively low fracture toughness.

It should not be overlooked that high strength materials which have relatively small flaw sizes for fracture initiation and have fracture propagation capabilities in the elastic load range have been used to date only in highly refined (design and fabrication) structures. General use of high strength materials in ordinary structures (conventional design and fabrication) requires material that resists fracture initiation at localized positions of plastic overstressing and resists fracture propagation through the remainder of the structure which is subjected to normally elastic load levels. Thus, a critical first question becomes -- how high in strength can we go before fracture safety cannot be assured by materials characteristics? This is the same as saying -- above a certain strength level the fracture toughness of a material will be sufficiently low so as to require precise knowledge of stresses at all locations in the structure and of the flaw sizes existing at such locations. Conversely, below some critical strength level it will not be necessary to have such precise knowledge to assure fracture-safe design because the best materials can be demonstrated to require stresses above yield for fracture propagation. In effect, the cut-off points on the OMTL charts developed in this program are aimed at answering such elementary questions.

It does not follow that all structures will require material with "built-in" fracture-safe design assurance; however, the designer should know when he enters this territory -- above this point he must play the "factor of safety" game. It is important that the metallurgist should know when he has attained the maximum possible strength level with retention of "built-in" fracture toughness. The importance of this is simply that the designer does not have to concern himself with "factors of safety" questions to this level of strength. The metallurgist and the designer should both know specific fatigue crack propagation rates and specific environmental effects. There must be assurance that weld regions are not degraded significantly below the limits established for the base material. Metallurgical problems of welding heat response and weld metal characteristics are orders of magnitude more involved than those of the base plate. We are only beginning to examine the characteristics of such zones in an exploratory fashion.

There is no doubt that many more quarterly progress reports will be issued before a reasonably complete picture of these many interdependent variables is made available. At the present time, the picture is sufficiently clarified to establish the approximate strength limits for base plate fracture toughness which guarantees "built-in" fracture safety for reasonably complex structures subjected to slight plastic overloads, at least for sections of 1-in. thickness. Our immediate next aims are to define these limits for 2-in., 3-in., 4-in. sections and for weldments. At the same time, fracture mechanics studies will be made of the relatively brittle materials to define tolerable flaw sizes and stress levels. The techniques for such measurements have been evolved for materials of relatively high brittleness -- it is not clear how the intermediate range of semi-tough materials can be so defined. There is some evidence that this region covers a relatively narrow range of strength levels (on an OMTL basis) and, therefore, may not be a serious problem of definition. In other words, the application of lower factors of safety may provide the desired structural efficiency by using material of a lower level of strength. Conversely, a higher level of strength may be used with the application of a liberal factor of safety. The two approaches -- fracture mechanics, "working down" in strength to the less brittle materials and relatively simple engineering test approaches "working up" in strength from the very tough materials -- may leave a very narrow gap of strength range for which there is no well-defined design approach. The crucial problem will remain the definition of weld zone properties over a wide range of strength levels. In this respect none of the sophisticated laboratory test tools are adequate for proper definition and reliance may have to be placed on conservatism or on weld joint elimination procedures.

The relatively simple test tools utilized in this program to date (Charpy V test, drop-weight tear test, explosion tear test, explosion bulge test, and variations) have provided a frame-of-reference which does not exist in any other terms. These tests are not meant to compete with linear elastic analysis for materials which are sufficiently deficient in fracture toughness to permit such measurement. However, they generally indicate materials that may or may not be treated in linear elastic analysis terms. The challenge to the metallurgist is the optimization of the metals to as high a strength level as possible with properties that make the application of linear elastic analysis difficult if not impossible at this time. There is no problem in developing brittle metals at any strength level. The problem is one of achieving the best possible level of fracture toughness and directing the use of metals by the designer to these "best" categories. It follows that the design and fabrication job will be made much easier and more reliable by such attainments. As metallurgists, we are not satisfied with simply evolving and using fracture test methods -- the payoff is in optimizing the metal and particularly the weld joint region, hence our emphasis on the OMTL concepts discussed in the reports.

## TITANIUM ALLOYS

(R.J. Goode, R.W. Huber, and D.G. Howe)

Fracture toughness studies of a spectrum of 1-in. thick titanium alloys have provided a "yardstick" frame-of-reference of drop-weight tear test (DWTT) fracture energy in relation to the tensile yield strength (YS). This frame-of-reference is redefined as a "fracture toughness diagram" by correlation with results of the explosion tear test (ETT). Figure 1 illustrates test data for weak direction (WR) and strong direction (RW) -- i.e., along and across the direction of primary rolling. The "fracture diagram" aspects are illustrated by the correlation of "flat break" -- fracture below yield when ETT tested and by fracture propagation requiring 1-2%, 3-5%, and 5-7% strain. In other words, with increasing DWTT energy absorption ranging from 1500 ft-lbs and less to approximately 3000 ft-lbs, the material may be predicted to react to the ETT (2-in. crack flaw) by breaking flat or by deforming a certain specified amount prior to fracture. The optimum materials trend line (OMTL) marks the maximum (optimum) limit of the highest DWTT values noted as a function of yield strength. From a reliable producibility point of view the OMTL probably has to be "shifted back" approximately 10 ksi at any specified level of DWTT energy. Also, with increasing thickness the OMTL may be expected to be "shifted back" -- additional test data are required to develop a diagram relating to thick materials, materials subjected to special processing, and for welds of various thickness. It should be noted that Fig. 1 also presents approximate Charpy V values, related to the DWTT energy. These correlations have not been sufficiently precise to provide for the use of Charpy V definition of the performance in the ETT. For this reason, we favor the use of the DWTT as being a more exact and discriminating test method -- the Charpy V test can be used only as a "rough" index. Tests are underway to establish the plane strain fracture toughness ($K_{IC}$) for a number of alloys represented in Fig. 1. It is expected that data will be presented in the next progress report -- the principal difficulty has been the determination of $K_{IC}$ values that can be considered valid. Considerable contention exists as to the proper method of determining a crack instability point because titanium alloys apparently do not give a clear indication of a "pop-in" or crack instability. The fracture mechanics, $K_{IC}$ determinations are being conducted with the advisory assistance of experts in the field.

At the present time, the only reliable compendium of fracture diagram data are available only on terms of the DWTT. An example of the use of these data is provided by the 2000 ft-lbs index which represents a value which assures ETT toughness of a minimum of 1-2% strain (1-2% guaranteed and 3-5% expected) prior to fracture. On an OMTL basis it is possible to attain this performance with 135 ksi material -- on a practical producibility basis this performance may be expected in production at 125 ksi (on the average). If effects of increased thickness are factored into this analysis (from known decreases in yield strength with increased thickness) it should be reliably expected to attain such levels of fracture toughness at 2-in. and probably 3-in. thickness at a guaranteed minimum of 105 ksi yield strength. This appears to be a conservative estimate based on present knowledge -- additional tests may show that the fracture toughness -yield strength minima for purchase specifications could be relaxed. On the basis of the subject frame-of-reference, one may make other estimates based on higher or lower levels of fracture toughness and other levels of anticipated production controls, heat to heat variables, etc.

Based upon these studies an Interim Guide issued by the U.S. Marine Engineering Laboratory (6) has modified the previous 21 ft-lb Charpy V at -80°F specification requirements for titanium intended for submarine hull construction to a 105 ksi minimum YS and a minimum fracture toughness of 2000 ft-lb DWTT energy. As seen from the fracture toughness diagram (Fig. 1), this level of toughness corresponds to an expected development of 3-5% plastic strain prior to fracture propagation in the ETT in the presence of a 2-in. crack-like flaw.

The 1-in. thick plate alloys that have been investigated which fall within these revised specifications are shown in Fig. 2 -- this figure does not include the results obtained for the material from the special processing program underway at Reactive Metals, Inc. (RMI), which are reported in a separate section of this report. The data points were obtained from material in the hot-rolled condition and in various heat treated conditions, depending upon the alloy, and, as can be seen, a number of alloys are represented. The remaining principal deciding factor on the usefulness of any of these alloys in the region of interest as a hull material is dependent upon the weldability of the material.

It is hopeful that the OMTL can be moved to higher levels of fracture toughness through introduction of new alloys resulting from alloy development studies and through heat treatment. An example of this is the Ti-6Al-3V-1Mo alloy which was made in the vacuum arc skull melting facilities at NRL in the form of a 65-lb vacuum arc remelt cast into a 4 x 7 x 12 in. copper chill mold following which a one-half section of the billet was forged and rolled at NRL. The oxygen level is approximately 0.04%, and through heat treatment it has been possible to develop over 4300 ft-lb DWTT energy at a 109 ksi YS level. As seen in Fig. 2, these properties exceed the previously established OMTL. The heat treatment used was annealing at 1675°F for two hours and water quenching, followed by an aging treatment at 1300°F for one hour then air cooling.

## EFFECT OF PROCESSING VARIABLES ON STRENGTH AND TOUGHNESS OF A Ti-7Al-2Cb-1Ta ALLOY

A study of processing variables on the mechanical properties of a Ti-7Al-2Cb-1Ta alloy (Ht 291488) is being conducted by RMI under sponsorship of the U.S. Navy Bureau of Ships. NRL is evaluating the plate material evolved in this study in the DWTT and ETT to provide guideline information on the full plate thickness strength-toughness combinations developed in relation to the OMTL for titanium. Some results were presented previously in the Fifth Quarterly Report (7) on plate produced by forging and hot rolling. Drop-weight tear test information has been recently obtained on extruded plates. The earlier reported results are included in this report with more detailed processing and heat treatment information.

The average chemical composition of the alloy as reported by RMI is:

| Weight - % | | | | | | |
|---|---|---|---|---|---|---|
| Al | Cb | Ta | Fe | C | $N_2$ | $O_2$ |
| 6.9 | 2.5 | 1.1 | 0.13 | 0.01 | 0.006 | 0.063 |

The forging and hot-rolling procedure used by RMI for producing this 1-in. thick forged-and-rolled plate material was as follows: Ingots were forged from 24-in. diameter to 4 x 17 x 19-in. slabs using three different unspecified forging techniques at a starting temperature of 2200°F (furnace temperature) and 1820°F finishing temperature; die temperature was 1980°F. The slabs were then rolled to 1.1-in. thick plate using an 1850°F initial and 1700°F final rolling temperature. The effects of annealing and aging temperatures on the tensile YS and fracture toughness as measured by the Charpy V notch and DWTT are given in Table 1.

The extruded material was produced in the form of bars 5 x 1-in. by length, using a 5500-ton press. The extrusion temperatures were 1900°F ($\beta$ extrusion) and 1700°F ($\alpha$+$\beta$ extrusion). The billet diameters were 10.6-in. and 9-in. for the 1900°F and 1700°F extrusions, respectively. The effects of heat treatment on the strength and toughness properties are given in Table 2.

Forged-and-Rolled Plate

Within the scope of this investigation several interesting general observations can be made from the results obtained with the forged-and-hot-rolled plate material. These are:

(1) The effect of processing on the "as-forged" and hot-rolled fracture toughness properties as determined by the DWTT was greatest in the RW fracture direction (8). Here the spread of values is approximately 500 ft-lb, whereas in the WR fracture direction the spread is about 150 ft-lb. The effect of processing on the tensile YS seems to be slightly greater in the transverse direction than in the longitudinal direction -- 8.5 ksi and 5.5 ksi spread, respectively.

(2) Fracture toughness, as measured by the DWTT, is essentially the same in the RW and WR directions, independent of processing procedures, when the plates are annealed above the $\beta$ transus regardless of whether the annealing treatment is followed by air cooling or water quenching (except for forging process C). Water quenching does result in generally lower DWTT energy values and higher tensile YS compared to air cooling.

(3) Annealing slightly below the $\beta$ transus ($\alpha+\beta$ field) produces the highest levels of fracture toughness (DWTT data) compared to either the $\beta$ annealed or as-forged and hot-rolled plate. The yield strengths obtained are comparable to those obtained in the as-forged and hot-rolled condition. However, except for forging process A, the spread in DWTT energy for the two directions remains the same or is slightly greater than that of the forged-and-hot-rolled material.

(4) Annealing the material subjected to the forging process A produced essentially the same DWTT energies in both the RW and WR orientations regardless of whether it was an $\alpha+\beta$ anneal or $\beta$ anneal followed by air cooling or water quenching. This was probably the result of the particular forging procedure used in the initial breakdown of the ingot.

(5) The Charpy V notch test does not provide a satisfactory discrimination of the process variables, as compared to the DWTT.

Specifically, the better combination of strength and toughness is afforded by the $\beta$ anneal followed by air cooling, giving approximately 2700 ft-lb DWTT energy in the 106-110 YS range. Water quenching following this same annealing treatment raised the strength to 109-122 ksi range at some expense to the fracture toughness level.

### Extrusions

The results obtained with the extruded and heat treated material suggest the following general conclusions:

(1) Considerable anisotropy in fracture toughness is present in these materials; from the few measurements made, the DWTT energies in the RW direction is approximately double those in the WR direction.

(2) The anisotropy in YS values is not much more than is normally found in commercially-produced forged-and-rolled plate but, in most instances, the specimens taken parallel to the extrusion axis had the higher values than those taken perpendicular to the extrusion axis. Generally, in rolled plate, even after heat treating, the reverse situation exists in relation to the direction of principal rolling.

(3) Higher fracture toughness levels and possibly slightly higher strength levels are developed with the heat-treated $\alpha+\beta$ extrusions (process D) than are developed with the correspondingly heat-treated $\beta$ (process E) extrusions. The anisotropy in YS may also be a little greater in the $\alpha+\beta$ extruded material.

(4) As with the rolled plate material, the Charpy V ($C_V$) data are not as discriminatory as the DWTT results.

The materials provided from this study have shown without exception a better combination of strength and toughness (Fig. 3) over that seen in any of the previously produced Ti-7Al-2Cb-1Ta or Ti-8Al-2Cb-1Ta alloys investigated at NRL (Fig. 4). Several of the processing procedures coupled with heat treatment for forged-and-rolled plate have produced material which approaches the estimated OMTL for titanium. The heat-treated 1700°F extrusions exceed the OMTL when tested in the strong (RW) direction. However, the OMTL represents "rolled plate" weak direction properties, and on this basis the weak (WR) direction fracture toughness properties of the extrusions lie considerably below it.

## DIFFUSION-BONDED TITANIUM ALLOY PLATES

The notch fracture toughness properties of laminated plate prepared from diffusion-bonded titanium alloy sheet stock are being studied because (1) enhanced mechanical properties can be developed through work-hardening or strengthening mechanisms, (2) anisotropic properties associated with texturing might be used to advantage, and (3) medium strength, fracture tough materials may be combined with brittle, high strength alloys to give an optimum combination of properties.

Previously reported(7) $C_V$ data obtained from a small sample piece of diffusion-bonded plate indicated comparatively good toughness was retained at a tensile YS level that was about 15% above the nominal for the Ti-6Al-4V alloy.

Titanium alloy sheet stock remaining from the DOD Titanium Sheet Rolling Program was procured for preliminary diffusion-bonded plate experiments. The six alloys used in these tests represent the all alpha or super-alpha, alpha plus beta, and all beta titanium alloys rolled to 0.062-in. and 0.090-in. sheet.

Hot sheath rolling of the titanium alloy laminates was employed to accomplish the diffusion-bonding. Sheared 6 x 6-in. squares (thirty of the 0.062-in. and twenty-two of the 0.090-in.), cleaned by $HNO_3$-HF pickling, were stacked in a mild steel box with the rolling direction of each piece rotated 90° with respect to its neighbors. A 1/4-in. thick cover plate was heli-arc welded to the box after evacuating and backfilling with an inert gas. The laboratory rolling mill limited the starting sheath size to 2-in. thickness and the reduction to 0.025-in. per pass.

The two all alpha alloys were too stiff to deform at 1850°F and it was necessary to raise the furnace temperature to 2000°F. All of the beta containing alloys were hot-rolled at 1750°F furnace temperature.

Test results are shown in Table 3. Drop-weight tear tests were made on the as-rolled plate after removal of the iron sheath. The fracture surfaces show evidence of good bonding with only minor amounts of delamination. It is doubtful that the "desirable" sheet properties can be retained in the bonded plate when temperatures much above 1700°F are employed in roll-bonding as evidenced by the two alpha alloys.

Additional diffusion-bonding experiments are being planned.

## FRACTURE TOUGHNESS OF 2-IN. THICK TITANIUM ALLOY PLATES

Drop-weight (vertical drop, large machine) tear tests on 2-in. Ti-7Al-2Mo plate using the bracketing technique (Charpy-type pendulum machine limited to 5000 ft-lb capacity) give a good limit value of 9200 ft-lb. Previously reported DWTT data on 2-in. Ti-7Al-2Cb-1Ta plate using the same technique was 7500 ft-lb (7). Further testing has shown that the actual value is closer to 7300 ft-lb DWTT energy.

Cross-section size scale up from the 1-in. to the 2-in. DWTT specimen size involves a ratio of about 3.8 increase in area; the corresponding ratio of increase in fracture tear energy is in the order of 4.7 to 5. These comparisons are to be considered preliminary and are subject to more exact comparisons in future tests.

## HEAT-TREATMENT STUDIES ON SOME TITANIUM ALLOYS

Heat-treatment studies (7,9,10,11) on a number of titanium alloys have been continued in order to develop information on the stability of the alloys and to determine the heat treatments which will produce an optimum combination of strength and toughness.

Beta transus determinations on the alloys Ti-6.5Al-5Zr-1V (T-36), Ti-6Al-2Sn-1Mo-1V (T-37), Ti-6Al-4Zr-2Mo (T-55), Ti-6Al-4V-2Sn (T-67), Ti-6Al-4Zr-2Sn-.5Mo-.5V (T-68), and Ti-7Al-2.5Mo (T-71) were made as a preliminary step to full scale heat-treatment studies and the results of these determinations are shown in Figs. 5-10. The beta transus temperatures are shown in Table 4.

The results of solution-annealing and aging treatments for the alloy Ti-8Al-1Mo-1V (T-19) are shown in Table 5. The treatment includes solution-annealing at temperatures below the $\beta$ transus for one hour in an argon atmosphere followed by air cooling or water quenching. The specimens were in most cases aged at 1100°F or 1200°F for two hours in an argon atmosphere and water quenched or air cooled.

The data shows that strength of this alloy is relatively insensitive to differences in aging treatment if the annealing treatment is followed by air cooling regardless of annealing temperature. If after the annealing treatment a faster quench rate is used, water quench, then the strength of the material increases with increasing annealing temperatures between 1750°-1850°F. This is seen in Fig. 11 which shows the effect of solution annnealing temperature, followed by water quenching or air cooling, on the YS of the material which had been aged for two hours at 1100°F and 1200°F, and then air cooled or water quenched.

Figure 12 is a summary of a preliminary survey based on Cv energy and YS relationships for the alloy Ti-8Al-1Mo-1V (T-19). The line indicates the optimum "weak" direction properties. Earlier DWTT studies showed that this alloy, heat treated at 1825°F for two hours, then air cooled, absorbed 2500 ft-lb energy before fracturing, indicating an ETT capability in the order of 5% plastic strain.

Of all the alloys investigated to date in the heat-treating studies, the alloys Ti-8Al-1Mo-1V (T-19), Ti-6Al-4Sn-1V (T-20), and Ti-6Al-2Mo (T-22) have shown the best combinations of strength and toughness. However, the properties of welds with some of these alloys will probably be difficult to maximize both as to strength and fracture toughness.

## ALUMINUM ALLOYS

(R.W. Judy, Jr.)

The testing of aluminum alloys during this period was confined to explosion tear testing (ETT). This is a large scale structural prototype element test which utilizes explosive loading in the presence of a sharp flaw. The object of the test is the determination of the amount of plastic strain which can be developed by a particular alloy in the presence of a sharp 2-in. crack-like flaw. It has been shown that this strain value can be related to the drop-weight tear test (DWTT) energy for steels and titanium alloys (9).

The preliminary fracture toughness diagram (Fig. 13) illustrates the results of the testing done to date. The crosshatched lines indicate correlations between DWTT energy and the amount of plastic strain that can be developed in the ETT before failure occurs. Preliminary ETT work had indicated that the 750 ft-lb level of DWTT energy was indicative of approximately 4% allowable plastic strain and that below 300 ft-lb, fracture below yield strength could be expected (7). However, a

6061-T651 (DWTT - 750 ft-lb) specimen was tested to 7.3% plastic strain (Fig. 14) without completely running the fracture, which indicated a capability for plastic strain in excess of 7% at this DWTT energy level.

In other tests, a 2024-T4 (DWTT - 367 ft-lb) fractured completely at a 3.2% plastic strain level, which value placed it near the lower extreme of the 1-7% band. A 2219-T87 specimen (DWTT - 281 ft-lb) fractured completely at 2.1% plastic strain and a 7075-T6 (DWTT - 139 ft-lb) specimen shattered and delaminated when loaded to less than 1% plastic strain (Fig. 15). Both of these were considered to have fractured near or below yield strength levels. It is obvious that the response of aluminum alloys in the ETT follows a distinctly different relationship to the DWTT than observed for steels and titanium alloys. Certainly, the very low DWTT and Charpy values of many of these alloys which indicate low fracture resistance for steels and titanium do not translate to the same low levels of fracture resistance in the ETT. The importance of developing specific correlations for various materials is emphasized by these findings. Additional studies are planned for aluminum alloys to evolve more exact relationships. These will include $K_{Ic}$ plane strain fracture toughness determinations for alloys that provide a valid basis for measurement at the specified thickness.

## HIGH STRENGTH STEELS

(P.P. Puzak, K.B. Lloyd, and E.A. DeFelice)

High strength steel studies have been continuing to evolve fracture toughness vs. yield strength (YS) frame-of-reference charts in terms of drop-weight tear tests (DWTT) and Charpy V ($C_V$) tests conducted at 30°F. Correlations with explosion tear test (ETT) data have served to define the prefracture strain levels similar to those described previously in this report for the case of titanium alloys. The highlights of the studies for steels have been the pronounced effects of melting and rolling process variables on fracture toughness properties for some "new" steels in the 150+ ksi YS range over that of the "old" types. The "old" types are generally characterized as HY-80 compositions heat treated to a range of high strength levels and the various 4330 - 4340, H-11, D-6, etc., as compared to the "new" types characterized by the maraging, the 9Ni-4Co, 5Ni-Cr-Mo-V, and HP varieties.

Additional material is being procured to develop more detailed data on the effects of process variables for the various types of steels.

From these studies, it has become clear that the fracture toughness of steels at a particular strength level is a critical function, not only of the composition of the steel, but also the process history -- data for a particular analysis and thickness are relatively meaningless without simultaneous definition of the process practices. It is also necessary to specify the test direction with respect to the rolling direction. In general, the characterization of relative fracture toughness for specified directions and process conditions has evolved as a major contribution of this program which clarifies an otherwise completely confusing picture of relative fracture toughness quality for a particular composition. In the absence of this information it would not be possible to "index" the quality of a particular steel with respect to others.

Additional studies are required to ascertain the effects of increased thickness compared with the 1-in. thickness for which most of the data to date have been obtained. In effect, thickness becomes a third dimension to the DWTT and $C_V$ frame-of-reference data in relation to yield strength. Future studies will be aimed to evolve significant information relative to thickness effects, to similar explorations of frame-of-reference charts for weld metals and to determinations of $K_{Ic}$ plane strain fracture toughness at yield strength levels that provide for obtaining meaningful data of this type. In addition, explosion tests of weldments will be required to evaluate heat-affected-zone (HAZ) properties. The weldment studies will be conducted on selected or "more promising" materials.

A summary of the DWTT results as a function of the yield strength of the test material for 1-in. thick steel plates is presented in Fig. 16. The curves shown in Fig. 16 separate these data into characteristic groups relating to the processing variables (melting practice and/or crossrolling) of the steels. It should be noted that these data represent the lowest level of fracture toughness for the indicated material, i.e., fracture propagation in the "weak" direction of a rolled plate, providing such a direction exists. Straightaway rolling of a steel plate results in pronounced "fiber" direction in the plate with relatively low fracture toughness in the fiber (longitudinal or "weak") direction and significantly higher (3000 to 4000 ft-lb) fracture toughness in the "strong" direction. Conventionally processed steel plates feature approximately a 3 - 1 crossrolling ratio and wide differences in fracture toughness can still be obtained as a function of specimen orientation. Special 1 - 1 crossrolling results in plates having isotropic (same fracture toughness) properties in the two directions, and the indicated values represent this fact.

For each characteristic group of steels shown in Fig. 16, it should be noted that a wide range of fracture toughness may be developed by different alloy steels of the same yield strength level; however, the maximum depicted by the curve is seen to indicate that fracture toughness decreases with increasing strength level for all groups of steels. The limiting ceiling curves have been designated as the "optimum" materials trend line (OMTL). The OMTL may be recognized as the "yardstick" for evaluation of new steels with respect to the practicable upper limits of fracture toughness for any given strength level as a function of conventional or special processing variables. Of particular interest is the fact that the limiting, ceiling curve in Fig. 16 depicts the apparent practical upper limits of toughness for any strength level attainable with special melt practice and 1 - 1 crossrolling of the new alloy steels developed within the past five to seven years. The OMTL indicated for the "old" steel alloys that have long been in use for high strength applications (SAE 4340, etc.) represent vacuum consumable-electrode-remelt (CER) practice and 1 - 1 crossrolling. This OMTL indicates better strength and toughness levels than that developed by conventionally processed steels, but considerably lower levels than the OMTL in Fig. 16 for the recently developed new alloys.

The standard method for evaluating fracture toughness available generally to industry has been the Charpy V ($C_v$) test. Correlations between the DWTT and the $C_v$ test have provided a means of calibrating the significance of the standard industry "small specimen" test measurement. For the steels studied to date, Fig. 17 depicts the relationships between DWTT and $C_v$ tests. It should be emphasized that the surprisingly good correlation of DWTT energy values with $C_v$ values applies to steels studied to date in this program, at least for $C_v$ values exceeding 35 ft-lb. A summary of 30°F $C_v$ tests for steel is presented in Fig. 18 as a function of the yield strength of the test material. The curves in this figure separate these data into characteristic groups relating to the processing variables of the steels similar to that described for the data given in Fig. 16.

## REVIEW OF 5Ni-Cr-Mo-V STEEL DATA

A detailed review has been made of the extensive data provided by the United States Steel Corporation's development program under the Bureau of Ships 130-150 ksi yield strength hull steel contract. The results of this review may be summarized in reference to the $C_V$-YS frame-of-reference master diagram discussed previously. The solid block areas in the diagram of Fig. 19 encompass the 0° to 30°F test data obtained by the United States Steel Corporation for 1/2-to 4-in. thick plates for the first two large scale production heats (Nos. X53185 and X53588) of 5Ni-Cr-Mo-V alloy steel. These materials were produced by a high quality electric furnace practice employing two oxidizing and one reducing slags; and all final deoxidation and alloy additions for adjustment of final composition were made in the furnace -- not in the ladle or mold. The effects of other melting practices and processing variables on the mechanical and toughness properties of this alloy steel composition are being investigated and will be reviewed in a later quarterly progress report. The data shown in Fig. 19 indicate that the first two production heats resulted in material of high fracture toughness, close to the upper OMTL for 1/2 and 1-in. plates. The data for 2-in. and 4-in. plates fell in an intermediate position -- it should be emphasized that the frame-of-reference used is that which was developed for 1-in. plate data. These comparisons again emphasize two aspects of the problem of comparing properties of a given composition with those of other types -- (1) the strong effect of melt practice, and (2) the effects of plate thickness. The chemical compositions of the first two production heats are given in Table 6.

A summary of MIG and stick weld data obtained under this same contract are represented by the additional dashed block areas in the diagram of Fig. 19. It should be noted that the properties of 1-in. and 2-in. plates of heats Nos. 1 and 2 reported by the United States Steel Corporation were checked by NRL and found to be in close agreement.

## EXPLOSION BULGE TESTS OF 5Ni-Cr-Mo-V WELDMENTS

Explosion bulge tests have been conducted for United

States Steel Corporation's 5Ni-Cr-Mo-V 1-in. steel plates welded with 140 ksi yield strength MIG weld metal developed under the U.S.S.-BuShips contract. General experience with the explosion bulge tests to date has demonstrated that undermatching welds results in a considerable decrease in weldment performance, and as such is not desirable for hull applications. A cooperative investigation of the question of relative performance of overmatching -- matching -- undermatching weldments of the new 5Ni-Cr-Mo-V steel was undertaken. Several specimens were prepared by U.S.S.-Airco for explosion bulge tests with the base-plate steel heat treated to nominal yield strength levels of 130, 140, and 150 ksi and joined with a MIG weld metal having a nominal yield strength of 140 ksi. The weld crown on each specimen was fully ground smooth and photogrids were applied to the specimens.

Figure 20 illustrates the appearance of three bulge test specimens (undermatching, top; matching, center; overmatching, bottom) that withstood four explosive shots (7 lb, 15-in. standoff) with no visible signs of failure. Reliable surface strain measurements were precluded by deterioration of the photogrids on the first explosive loading of the plates. However, visual examination and a carefully measured profile of each bulge specimen, Fig. 21, provided evidence confirming previous experience concerning relative performance of undermatching -- matching -- overmatching weldments. In each specimen, the heat-affected-zone (HAZ) was found to be in relief indicating greater straining in unaffected plate material than in the HAZ. Even though the weld was not centrally located in the undermatch weldment sample, this weld metal was visibly strained more than the others, and a depression or ridge approximately 10-mils deep had formed in the center of the weld. Past explosion bulge test experience has indicated that if the weld crown is not ground smooth, the strain concentration at the toe of undermatch welds tends to promote complete separation of the bulge test specimen on the first or second explosive shot.

These tests were primarily aimed at defining the relative strain conditions in mismatched weldments. Additional tests with weld reinforcement and with complex weldment assemblies would be required to judge the relative suitability of matched and undermatched welds. Equally important, it is essential that tests be conducted with flaws (such as crack-starter weld cracks) located in weld and HAZ regions. It is important to develop a "failure" of the weld zones to determine if this is of high or low fracture toughness. For example, a complete fracture following a weld or HAZ path, as compared to a short fracture resistant tear, has the same meaning as for plate tests. Explosion bulge tests of flaw-free weldments do not provide information of fracture toughness of weld and HAZ regions -- no more so than tests of flaw-free plates.

The samples shown in Fig. 20 will be subjected to additional explosion tests to develop failure -- from which the extent of fracture resistance can be evaluated. Additional tests will be conducted with new weldments featuring flaws. In addition, fatigue data should be obtained for undermatching welds.

## FRACTURE TOUGHNESS CHARACTERISTICS OF A NICKEL-BERYLLIUM ALLOY

(R. J. Goode, R.W. Judy, Jr., and R.W. Huber)

Recently, two 1-in. thick plate specimens of Berylco Nickel 440 -- a nickel-2% beryllium alloy -- were obtained from the Beryllium Corporation for preliminary evaluation as a distinctly "different" material and therefore of interest in relation to the development of fracture test correlations. The nominal composition of this material is 1.95 Be, 0.5 Ti, and the remainder Ni.

A series of small laboratory tests, including the drop-weight tear test (DWTT), the tensile test, and the Charpy V ($C_V$) notch test, were conducted on the two specimens; one of these being in the annealed condition, and the other in the full-hardened condition. The DWTT specimens were 5-in. wide, 1-in. thick, and 17-in. long (the standard geometry for titanium and aluminum specimens) with the direction of principal rolling in the 5-in. dimension. Titanium was used to embrittle the electron beam crack-starter weld (9). This is the same technique described previously for the introduction of a brittle zone in titanium alloys for which iron wire is used as a contaminant. All DWT tests were conducted in the WR orientation (8).

In the annealed condition, the energy values obtained from the fracture toughness tests (DWTT and $C_V$) exceeded the rating of the respective testing machines. In the DWTT, 5000 ft-lb of energy was absorbed by the specimen at 30°F with the resulting crack produced by the brittle weld extending approximately one inch into the test material. The specimen was broken at -40°F for study of the fracture surface in the electron microscope. Even at this temperature the specimen required two blows for complete fracturing, one consuming 5000 ft-lb and the second consuming 3950 ft-lb for a total of 13,950 ft-lb for complete fracture of the annealed DWTT specimen. This shows an extremely high level of fracture toughness. Standard $C_V$ specimens cut from the annealed DWTT specimen were tested in a 264 ft-lb Tinius Olsen machine with the result that none of the specimens would break, even at liquid nitrogen temperature. Figure 22 shows the deformed specimens tested at -320°F. The tensile strength of the annealed specimen was 47.8 ksi; the ultimate tensile strength was 100 ksi.

The initial DWTT energy value obtained for a specimen solution-treated to the full hard condition was about 1840 ft-lb. Further substantiating DWT tests were unsuccessful due to failure in the welds used to join steel tabs onto the specimen to provide the proper specimen span for testing. The full hard condition of the specimen was developed by heat treating for 1-1/2 hours at 970°F during which precipitation hardening occurred. The crack-starter welds and tab welds were incorporated in the specimen after heat treatment. The $C_V$ tests were conducted on specimens with the result that the $C_V$ energy values were essentially constant at 15 ft-lb (RW) and 12 ft-lb (WR) over the -320°F to 212°F temperature range (Fig. 23). Yield strength in the full hard condition was 177.5 ksi; the ultimate tensile strength was 242.5 ksi.

The fracture surfaces of the DWTT specimens were examined in the electron microscope. Figures 24 and 25 show comparable areas of the fracture surfaces of the annealed specimen and the full hard specimen, respectively. In both cases, fracturing occurred primarily by the same ductile mode; however, considerable differences in the two fractographs can be seen. In Fig. 24, the failure mode was by dimpled rupture (12) with gross plastic deformation in evidence. The dimples were very large and deformed by stretching and serpentine glide (12).

The full hard specimen also failed by dimpled rupture (Fig. 25), but the dimples were considerably smaller than those seen in the annealed specimen, and very little evidence of plastic deformation of the dimples was in evidence. The differences in appearance described indicate a difference in plastic strain imparted to the specimen during the fracture; this same factor probably accounts for a large portion of the gross differences in the fracture toughness energy values found in the DWT and $C_V$ tests.

The two heat treatments used in this study represent to some degree two extreme combinations of strength and fracture toughness. Since combinations of strength and fracture toughness lying between these values are attainable by proper heat treatments and could compare very favorably with structural materials now being studied, a more thorough investigation of the properties of this alloy and related alloys could be of interest.

## NEW LABORATORY DROP-WEIGHT BULGE TESTING FACILITY

(R.J. Goode, E.A. Lange, and P.P. Puzak)

Reliable knowledge concerning weldment performance is essential to fracture-safe design of structures fabricated with high strength steel, titanium, and aluminum alloys. The problem with these high strength materials is the potential susceptibility to develop low energy-absorption tear fractures, particularly in the heat-affected-zone (HAZ) regions of weldments. Such susceptibilities may not be inherent to the mill produced materials, but may be developed in response to specific metallurgical treatments on weld fabrication in the HAZ regions. One example of a catastrophic low energy tear failure in which the initiation and propagation of the terminal fracture were uniquely associated with the HAZ regions of a "lean-analysis", quenched and tempered (Q&T) steel pressure vessel has been fully documented (13,14)

The NRL explosion bulge test developed in 1949-50 has been the only reliable test method for evaluating the potentials of heat-affected-zone (HAZ) fractures in thick plate weldments. The method has been used by the Navy for the past fifteen years to preclude the use in submarine hulls of Q&T steel weldments with low energy tear (or brittle) fracture properties in the HAZ. The NAVSHIPS Standard (250-637-6) procedure derived from these studies relies not only on performance exhibited by as-welded samples, but also on weldments to which a crack-starter bead is added. The crack-starter bead introduces a brittle weld-crack flaw providing a realistic structural flaw condition. The fracture toughness of the HAZ is then judged on the extent of HAZ tearing which results from the application of low or high levels of bulge deformation. In addition to the Navy test facilities, several commercial bulge test facilities have been established within the past five years.

Within the past decade, the most versatile and valuable laboratory tool for evaluating the fracture toughness of prime plate and weld metal has proved to be the simple drop-weight test machine. Since there were no practical laboratory means for evaluating fracture toughness of HAZ, a feasibility study was conducted with an existing one-ton drop-weight machine and 1/2-in. thick steel weldments. Upon the successful completion of this study, a large drop-weight test facility (Fig. 26) was designed, constructed, and placed in operation. The capacity of the machine is 240,000 ft-lb which is obtained by dropping a 6-ton weight a distance of 20 feet. This new test facility will provide a laboratory means of conducting HAZ fracture toughness evaluations of 1, 2, and possibly 3-in. thick aluminum and titanium alloy weldments of up to 160 ksi yield strength levels, and 1 and possibly 2-in. thick steel weldments of 150 to 180 ksi yield strength levels.

The drop-weight bulge test procedure entails the use of an expendable aluminum "punch" casting. Its general shape is that of a right truncated cone with a base diameter of 5-in. A hardened T-shaped steel pin designed to cover the top diameter of the casting and fit snugly in a 2-in. deep hole is used to transmit the drop-weight loading to the soft aluminum casting. The resulting deformation developed by drop-weight loading of the assembly forms a hemispherical bulge of the weldment as it is forced into the open cavity of a die.

One-inch thick prime plate specimens of 150 ksi yield strength steels were used for the initial correlation of results with the new drop-weight machine and the standard explosion bulge test. The surface strains (bulges) developed by one full-capacity drop-weight test with the new machine were slightly greater than those developed in other specimens of the same steels by one seven-pound Pentolite charge exploded at a 15-in. standoff (Fig. 27). Using the same aluminum casting and pin assembly and three blows at full-capacity, the average surface strain was approximately twice that in another specimen of the same steel subjected to three explosive loadings of seven-pound charges at 15-in. standoff. The shape and geometry of the casting are easy to cast in a foundry, and used castings can be remelted to provide material for new castings.

The use of this equipment will be expanded following additional correlations with explosion tests -- duplicate welded specimens are to be tested in explosion and drop-weight. It is also planned to modify the aluminum block to provide for simulated explosion tear tests. For this purpose, a cylindrical surface is developed by the use of a "shoe-like" aluminum block which deforms into a cylindrical surface on deformation loading.

## LOW CYCLE FATIGUE CRACK PROPAGATION STUDIES OF Ti-6Al-4V, Ti-7Al-2Cb-1Ta, AND UNALLOYED TITANIUM

(T.W. Crooker, R.E. Morey, and E.A. Lange)

The safe and dependable application of modern high strength materials to large cyclically-loaded structures, such as pressure vessels and submersible vehicles, requires an improved knowledge of slow crack propagation resulting from low cycle fatigue. Small flaws and cracks invariably are formed during fabrication and manufacture of a large welded structure, despite the use of the best available processing and inspection techniques. Since fabrication flaws are unavoidable, the only practical recourse is to provide design criteria for preventing the growth of such cracks to a critical size from repeated service loads.

The aim of this investigation is to define and evaluate the factors which control the growth of cracks under low cycle fatigue conditions. The results of the current phase of this investigation are based on studies of crack propagation in center-notched plate bend specimens loaded in cantilever fashion. Preliminary evaluations of the low cycle fatigue characteristics of a variety of quenched and tempered steels, Monel Ni-Cu alloys, and a 2024 aluminum alloy, have been made (7,9,10,15,16). Briefly, it has been observed that for a specific environment and strain ratio, the growth rate of a low cycle fatigue crack is dependent upon applied total strain range, as expressed by the relationship

$$\Delta L/\Delta N = K(\epsilon_T)^n \quad \text{where:}$$

$L$ = total length of fatigue crack

$N$ = cycle of loading

$K$ = constant

$\epsilon_T$ = total strain range

$n$ = exponent

This relationship remains valid in the presence of mean strains other than zero and in the presence of aqueous corrosive environments. However, it has been observed that both of these factors affect crack growth rate. Mean strain can either accelerate or retard crack growth rate, depending upon whether it is tensile or compressive. Corrosive environments tend to accelerate crack growth rate, depending upon such factors as corrosion resistance, stress corrosion, and hydrogen embrittlement.

## MATERIALS AND PROCEDURE

The materials considered in this report are three titanium alloys; two samples of Ti-6Al-4V (T-5 and T-27), one sample of Ti-7Al-2Cb-1Ta (TA-2), and one sample of unalloyed titanium (T-16). Chemical compositions and mechanical properties of these materials are shown in Tables 7 through 9. This assortment of titanium structural alloys offers a wide spectrum of variations in chemistry, strength level, and fracture toughness for comparisons of fatigue performance. The strain deflection characteristics of these materials in the plate bend specimen are shown in Fig. 28.

The experimental procedure employed for this series of tests is the same as that described in Refs. 7,9,10,15, and 16. Experimental data are based on the observed macroscopic growth of fatigue cracks across the surface of center-notched plate specimens. These specimens are cantilever loaded under fully-reversed cycling in both air and 3.5% salt water environments. Constant total strain range loading conditions are maintained by adjusting deflection, and the corresponding crack growth rate is measured. Nominal surface strains are measured with electrical resistance strain gages. Each specimen is successively tested at a specific total strain range value for an interval of several hundred to several thousand cycles until the crack growth rate can be established, and then loaded to a higher strain level. In this manner, a series of crack growth rate versus total strain range data points are obtained from each specimen. For tests conducted under a salt water environment, a corrosion cell is added to the specimen, allowing the salt water solution to flow over the crack from a reservoir during testing.

## STRAIN RANGE EFFECTS

Crack growth rate versus total strain range data for each of the materials in air are shown in Fig. 29. It can be seen that the data fall into two distinct groups which can be expressed as separate mathematical relationships of the form $\Delta L/\Delta N = K(\epsilon_T)^n$. The unalloyed titanium data form one relationship with a slope of 4:1 and data from the three high strength titanium alloys form another relationship with a much steeper slope of approximately 8:1.

The slope of these lines is an indication of the sensitivity of the macroscopic crack growth rate to changes in applied strain. Such changes in strain would occur in the presence of a growing crack under constant load conditions or would result from accidental overload. Thus, materials showing a steep slope possess lesser ability to accommodate excursions beyond nominal design stresses without grossly increasing the cyclic crack growth rate.

Similar variations in the slope of such curves, i.e., increase in sensitivity to strain with higher strength materials, have been observed in other alloy groups, specifically Monel alloys (16). Such variations appear to be dependent upon microstructure and this possibility is discussed in Ref. 17. The apparent rapid increase in the macroscopic crack growth rate of high strength alloys at higher strain values may be due to precracking of secondary phase constituents ahead of the crack front in the primary matrix. However, confirmation of this theory will have to await further investigation of a wider variety of materials and a study of the fracture surfaces.

## COMPARATIVE FATIGUE PERFORMANCE

A more practical comparison of the relative fatigue crack propagation resistance of materials can be obtained from a log-log plot of the crack growth rate versus the ratio of total strain range to proportional limit strain range. Such a plot is shown in Fig. 30, where the proportional limit is defined as 500 micro-inches/inch plastic strain range as determined from plate bend measurements. In addition to the titanium alloys under discussion, Fig. 30 indicates also the low cycle fatigue characteristics of 2024 aluminum alloy and HY-80 steel for comparison.

An important point to note is the wide difference in crack growth rates that exist in various materials at applied strain values near their respective proportional limits. These are the strain conditions that would exist at structural discontinuities and around growing fatigue cracks at critical locations in large structures. It can be seen that low modulus, high strength materials are at a distinct disadvantage under such conditions. Studies to date indicate that low cycle fatigue crack growth rate is highly sensitive to applied cyclic strain and such materials develop relatively large strains when loaded to their respective proportional limits. Data such as this indicates that the possibility of failure from low cycle fatigue crack propagation will be much more acute in future structures employing higher yield strength and/or lower elastic modulus materials.

## CORROSION FATIGUE EFFECTS

The log-log plot of crack growth rate versus total strain range data for the titanium alloys obtained under salt water environment is shown in Fig. 31. Superimposed on Fig. 31 is the relationship from similar data obtained in an air environment for purposes of comparison.

The low cycle fatigue performance of both samples of Ti-6Al-4V appears to be unimpaired by the presence of a salt water environment. In contrast, Ti-7Al-2Cb-1Ta is unfavorably affected by the salt water environment and exhibited inferior fatigue performance. Cracking in Ti-7Al-2Cb-1Ta alloy became very rapid and erratic under salt water conditions when cyclic strain range values exceeding approximately 60% of the proportional limit were applied. Figure 31 shows that significant acceleration in the growth rate exists at all strain levels in Ti-7Al-2Cb-1Ta; however, the value of 60% of the proportional limit (10,000 micro-inches/inch) appeared to be a threshold value beyond which a form of"corrosion cracking" appeared to be the dominant crack growth mechanism. These fatigue data suggest that a corrosion problem may very well exist in the structural application of Ti-7Al-2Cb-1Ta alloy, at least for structures containing flaws.

## CONCLUSIONS

(1) Macroscopic growth rates of low cycle fatigue cracks in unalloyed titanium, Ti-6Al-4V, and Ti-7Al-2Cb-1Ta follow exponential, strain dependent relationships of the form $\Delta L/\Delta N = K(\epsilon_T)^n$.

(2) Growth rates in Ti-6Al-4V and Ti-7Al-2Cb-1Ta are highly sensitive to small changes in applied strain and develop very rapid crack growth rates (>1000 micro-inches/cycle) at cyclic strain levels below the proportional limit.

(3) The presence of a salt water environment had no measurable effect on the fatigue crack growth performance of Ti-6Al-4V. However, Ti-7Al-2Cb-1Ta exhibited markedly inferior fatigue crack growth performance under salt water conditions at cyclic strain levels in excess of 60% of the proportional limit.

## TITANIUM CASTINGS

(E.J. Chapin)

It is planned to conduct a survey of fracture properties of titanium castings at a future date. In preparation for this future study (to include 100-lb castings produced at NRL as a start and hopefully larger castings to be procured under the Bureau of Ships contracts to follow at a later date),a brief review was made of titanium casting status. Because of the general interest this information is presented in the present progress report. It should be noted that similar questions of foundry producibility problems may develop for other materials of competing strength-density ratio.

The reactive nature of molten titanium toward atmospheric gases and toward all known refractory materials constitutes a major problem in melting and casting of the metal. Since the metal can be seriously contaminated in this process, the metallurgy of melting and casting must be restricted to environment, processes, materials and equipment that do not affect initial metal quality. This requirement makes the casting of titanium a relatively costly operation and limits the use of titanium castings to applications where forging production may involve other problems, particularly with respect to complexity.

The production of ingot titanium material for subsequent reduction to fabricated forms differs from the production of titanium castings. Ingots are produced by arc melting a consumable electrode in a cold mold in which a shallow molten pool is maintained that is progressively solidified until the desired length of ingot is obtained. In this process only a relatively small amount of metal is molten at any given time. The casting of titanium into desired shapes however, requires that a sufficiently large volume of molten metal be accumulated and poured at one time to fill the required mold cavities. This requirement makes the problems of scaling-up melting capacity for increased sizes of castings far more formidable than those involving the scaling-up of melting capacity for production of larger sizes of ingots.

An essential requirement in the successful use of nonmetallic refractory mold materials for casting of titanium into desired shapes is that the molten metal must not wet the mold surfaces and the metal must freeze and cool relatively rapidly to avoid harmful surface contamination. The solidification time for a given section of the casting must not exceed a certain critical value if contamination or mold breakdown is to be avoided. This value depends solely upon the stability of the refractory mold material and the length of time the mold surface is in contact with molten metal. In the case of casting very heavy sections, involving increased periods of solidification, special precautions must be taken to keep the mold surfaces relatively cool to avoid contamination of the casting surfaces.

The melting procedure in present use for production of titanium castings consists of arc melting in vacuum one or more consumable electrodes in a water cooled copper crucible. Rapid melting occurs from the use of high current densities and a large volume of molten metal is produced with only a relatively thin skull of solidified titanium being formed at the copper crucible-melt interface. The solidified skull remaining in the crucible after pouring represents about 35% of the total metal charged.

It is rather difficult to retain adequate amounts of superheat in the melt to assure fluidity because of severe heat losses. These include: (1) continuous heat loss from the hottest region under the arc through the liquid and the solid skull into the water cooled crucible, and (2) radiation losses from the arc and from the surface of the melt. The amount of superheat that can be retained in a molten titanium bath for a given power input is therefore low for two reasons: (1) there is a limit to the amount of heat that can be transferred through the molten bath -- the steep temperature gradients rapidly dissipate when the arc is extinguished prior to pouring, and (2) the molten bath tends to approach an equilibrium condition with the skull which means that pouring must be done very quickly to avoid loss of fluidity.

The amount of superheat in a melt cannot be determined accurately because of the lack of a suitable method for measuring temperatures of a titanium melt under the conditions of the melting process. As a result it is not possible to establish optimum pouring temperatures for casting and the temperature conditions can only be roughly estimated from either the power consumed after the titanium is completely melted or from optical pyrometer measurements. This circumstance makes it necessary to rely upon trial and error for obtaining the necessary fluidity to meet a specific casting requirement. As a result, the physical and mechanical properties may vary considerably between castings from different heats of metal. It is obvious that a method is needed for determining the temperature of titanium melts with reasonable accuracy in order to produce castings of uniform quality and properties.

Graphite has been found to be the most reliable mold material for casting titanium. Although graphite is not entirely inert to molten titanium, the carbon contamination is mostly confined to the surfaces of castings while only a very minor amount is added to the composition of the metal. The limited reaction occurring between molten titanium and the cold graphite mold is attributable to the fact that under the conditions of casting,molten titanium does not wet graphite and the high thermal conductivity of the graphite causes solidification of a thin skin of metal almost immediately upon contact at the interface. This circumstance inhibits any large take up of carbon into solution in the bulk of the titanium.

Two forms of graphite molds have been developed which include machined graphite and expendable rammed graphite. Machined graphite produces surface carbon contamination in castings which is generally no greater than .01-in. deep. This type of mold, however, has some disadvantages such as: impermeability to gases; it is costly to make; it has short life; and it is not suitable for castings of complex geometry. A second type of mold that was developed to overcome some of the disadvantages of machined graphite consists of a graphitic material composed of a combination of graphite powder and carbonaceous additives used to prepare expendable rammed molds on a pattern using conventional foundry techniques. The expendable graphite material can be used for internal cores in castings and such cores can be readily knocked out of the castings. Castings produced in this type of mold are comparable in surface finish and soundness of castings produced in machined graphite, however, the depth of carbon contamination is greater, amounting to about .04-in. maximum for a 1-in. thick section. The advantages of this type of mold include: improved permeability; ability to handle complex casting geometry; and lower mold cost. A characteristic common to both machined graphite and expendable rammed graphite type molds is the high thermal conductivity of the mold material which introduces shrinkage problems that involve production of gross porosity and centerline shrinkage defects. The extent to which surface contamination of castings by carbon impairs ductility is a question that needs to be resolved by service tests. In cases where such contamination is considered deleterious, the surface contamination is removed to some predetermined depth by either sandblasting or chemical milling.

Metallic molds composed of copper or of iron are sometimes used for casting titanium where casting geometry permits unrestricted contraction. The high thermal conductivity of the metallic molds causes rapid freezing of the casting and avoids reaction of the mold with the casting. This is particularly advantageous in the case of heavy sections although rapid chilling aggravates the shrinkage problems.

Titanium alloys may be cast centrifugally or statically. In general, both processes utilize the same melting and mold technology; the only difference is that in the centrifugal mode of casting the mold is spun about a vertical axis. This procedure is best suited for production of cylindrical shapes. Parts cast by this method have a considerably finer grain size than statically cast parts and are easier to machine.

Unalloyed titanium casts readily into relatively complex shapes with good surface finish, soundness and good reproduction of details. Alpha and alpha-beta alloys appear to have a good castability as unalloyed titanium. Meta-stable beta alloys appear to be difficult to cast because of their long freezing ranges. Consequently, control of shrinkage is difficult due to the poor feeding characteristics.

The development of titanium alloys with good mechanical properties has been almost entirely concerned with wrought material. These alloys are not necessarily good casting alloys and much of the experimental data that has been obtained in the development of wrought titanium alloys may not be directly applicable to casting alloys. Although a great number of experimental alloys have been cast, the alloys on which the most property data have been developed include only one alpha type (Ti-5Al-2.5Sn), and one alpha-beta type (Ti-6Al-4V).

Although alpha-beta alloys respond to strengthening by quenching from temperatures above the beta transus, the preferred treatment is to employ a solution treatment followed by aging. Alpha-beta alloys are stronger than alpha alloys on a strength-to-weight ratio basis, however, they suffer from limited weldability. It is doubtful that weldable alpha-beta alloys can be satisfactorily strengthened by heat treatment in thick sections since weldable alloys with low beta stabilizer content require very rapid quenching from solution treatment temperatures to achieve heat treatment response. The required rapid cooling would be very difficult to obtain in thick cast sections. Where requirements involve high strength and/or greater ductility for heavy section castings, special casting alloys may have to be developed to meet these requirements.

Present industrial casting furnace capacity permits casting of titanium alloy shapes with maximum dimensions of 54-in. diameter by 31-in. height and a maximum weight of 400 lbs. Metal needed in the casting furnace in the form of consumable electrodes is produced in 20-in. diameter ingots. Titanium alloys are cast by centrifugal, centrifuge or static methods, depending upon size, shape and requirements. Present centrifugally cast bar capability is 8-in. O.D. by 6 ft. length or 24-in. diameter by 30-in. length. Casting yield is about 40% of total metal charged.

The carbon content of commercial titanium alloy castings made in machined graphite is about 800 ppm maximum and about 2000 ppm maximum for castings made in expendable rammed graphite molds. The surface finish of castings produced in these molds is comparable to smooth sand cast finishes. Castings can meet all classes of X-ray quality depending upon design, and pressure tightness is classified as excellent. Where defects occur due to gross porosity and centerline shrinkage, the defects are repaired by a weld repair method. This consists of drilling the casting to the defect and filling the cavity with weld metal.

Present melting and casting capacity for producing shaped castings could be scaled up, using present technology to melt 3500 lbs. of titanium alloy material from which a casting yield of 2500 lbs. could be obtained using either a static or centrifugal mode of casting. The scaled up furnace equipment would accommodate shapes up to 12 ft. diameter by 6 ft. height.

Since the production of large thick flat plate by casting into molds is considered to be not feasible because shrinkage defects cannot be satisfactorily controlled by risering, the increased melting and casting capacity mentioned above would be only for the production of shaped castings in graphitic type molds. Shaped castings could be scaled up in size and weight to the limits indicated above without involving any extensive developmental work in the matter of melting and casting furnace equipment or in the preparation of molds, providing graphitic type materials would be acceptable for molds.

As was discussed previously, it is not known whether satisfactory wrought titanium alloys would be directly applicable for use as casting alloys. Thus, the problem of obtaining satisfactory casting alloys which are weldable and which can be satisfactorily strengthened in heavy section needs to be investigated and the mechanical properties of cast and wrought forms of the same alloy compared.

## MODIFIED CHARPY TESTING PROCEDURES: THE FATIGUE CRACKED CHARPY AND SIDE NOTCHED CHARPY IN IMPACT AND SLOW BEND TESTING

(L.J. McGeady)

Because of long experience with the use of the conventional Charpy V test for carbon steels, most of the interrelations of results from various specimen types and test methods are rather fully understood or documented as a result of the sharp transition from tough to brittle failure. Consequently, specification for these steels can usually be drawn with a remarkable degree of confidence and precision. The newer steels do not, in general, exhibit this sharp transition. Therefore, as they develop, they need to be examined closely to determine whether the testing results from various tests and specimens are interrelatable.

More pertinently, the relevance and applicability of laboratory tests to service behavior need to be evaluated with circumspection and need to be based on the best experience available. The NRL drop-weight tear test and modifications offer specimen type and behavior on a geometric scale large enough to be considered prototype. The conventional Charpy V-notched specimen offers different characteristics in size and possibly test response. It is important to know whether the Charpy test can be correlated with the larger test. Recently it has been demonstrated at NRL that there is a high degree of correlation possible between the drop-weight tear test measurement of steel notch toughness at 30°F and standard Charpy test shelf energy measurements. It has been suggested that improved correlation of results from the Charpy test may be obtained if a fatigue crack is induced at the root of the standard notch prior to test. The philosophy behind so doing is that because a crack has already been formed by fatigue in such a test, the breaking energy in the Charpy test is limited to that energy required to force the fatigue crack to run and does not include a large component of the energy necessary to cause a crack to initiate. Therefore, it was deemed important to explore whether the modified Charpy tests can provide information not available in the standard test by making the Charpy notch cracklike and more severe. It has been suggested too that if the testing conditions in the Charpy test are made more severe by intensifying the notch, as by notching two additional sides of the test bar as well as by inducing a fatigue crack, testing will produce more meaningful results because of minimization of the shear-lip type fracture encountered in the conventional test. Still another philosophy is that a slow-bend method of testing the Charpy specimen may be more pertinent to service behavior and more discriminatory of steels than the standard impact test method.

To investigate these questions a test correlation program was evolved which included various tests of 34 high strength steels ranging in yield strength from 80,000 psi to 280,000 psi. These represent steels of a variety of types and compositions; some are conventionally quenched and tempered, some "lean alloy" commercial materials, some maraging steels, some from commercial production heats, and some from specially processed heats, etc., representing therefore various current methods of manufacture and states of development of these steels. Further, steels for the present program were selected on the basis of NRL drop-weight tear test performance. That is, both high and low toughness steels and several levels of yield strength that have been tested to date. All have been tested using the standard Charpy test specimen and two modifications, the fatigue-cracked Charpy and the side-notched Charpy test specimens. Two types of loading have been used, standard impact and slow-bending. The latter testing was performed by personnel of the Research Center of Republic Steel Corporation, Messrs. S. J. Matas and S.J. Pascover, who have experience in conducting such tests. The cooperation of the Republic Steel Corporation is highly appreciated.

Figures 32 and 33 illustrate the correlations determined between results from conventional Charpy tests and results from modified specimens and procedures. The W/A results have been used as the basis for plotting the figures to take into account the cross-sectional area of specimen broken since this varies between the specimen types. Figure 32 portrays the correlation of results from conventionally notched and tested specimens with results of impact tests of fatigue cracked specimens and also with results of tests of specimens side notched and fatigue cracked. Figure 33 illustrates the relation of conventional impact test results to slow bend test results of modified Charpy tests. Although there is considerably more scatter shown in Fig. 33, some of the scatter may be due to the fact that specimens for some of the steels were tested in slow bending at 90°F while other steel specimens were tested at 30°F.

The energy absorbed in fracturing specimens can be reported in terms of W/A, work or energy expended per unit area of fracture surface produced. This system has been used in analysis of the data from this test program. However, since the W/A values are merely the measured ft-lbs energy absorbed in the Charpy test divided by a constant, use of this figure in itself does not reveal anything not already available from the test data from which it is derived.

The principal effect of intensifying the standard Charpy V notch, by addition of a fatigue crack alone or by addition of a fatigue crack and side notches combined, is to cause less energy absorption in specimen fracture. The effect has been shown to be predictable and to be independent of the type of steel used in these tests. When the steels tested were ranked in decreasing order of toughness (Fig. 34) using as a basis either the conventional Charpy test results or the results from modified specimens, the orderings with minor exception were the same irrespective of the test specimen used. The most serious exception to this occurred in the results of slow bend tests wherein the instance of five steels W/A and/or E values of side notched specimens were higher than those of corresponding steels fatigue cracked only.

This may have occurred because some of the specimens tested in slow bending were tested at 90°F while others were tested at 30°F.

The data obtained by slow bending of the modified Charpy test specimens are characterized by considerable scatter. However, average results of energy absorption in the slow bend tests correlate reasonably well with values obtained in conventional impact testing of the same steels.

In addition to not being of particular advantage, the use of the notching procedures more severe than in the conventional Charpy tests causes a reduction of low values of energy absorption for all steels such that ease of discrimination among the steels is seriously reduced for the lower levels of energy absorption.

It is likely that if all specimens had been tested at 30°F, the correlations obtained between slow bend tests and conventional impact tests would have been even closer. However, despite this feature of the testing program, the correlation lines of Figures 32 and 33 are identical. This is further substantiation that the results of the modified tests can be predicted with reasonable accuracy from the conventional Charpy test results. Hence the value and necessity of modified Charpy testing procedures appear questionable.

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TABLE 1

MECHANICAL PROPERTIES OF
SPECIALLY PROCESSED Ti-7Al-2Cb-1Ta ALLOY PLATE

| Heat Treatment and Mechanical Properties | Process A | | Process B | | Process C | |
|---|---|---|---|---|---|---|
| | Long. | Trans. | Long. | Trans. | Long. | Trans. |
| As Forged & Hot Rolled | | | | | | |
| DWTT (ft-lb) | 3032 | 2410 | 2676 | 2498 | 2560 | 2353 |
| $C_V$ at 30°F (ft-lb)* | 39.8 | 37.7 | 38.7 | 43.7 | 48.0 | 41.5 |
| 0.2% YS (ksi)* | 102 | 105 | 107.5 | 112 | 106 | 113.5 |
| Annealed 1650°F/1hr/AC | | | | | | |
| DWTT (ft-lb) | 2950 | 2965 | 3148 | 2526 | 3150 | 2705 |
| $C_V$ at 30°F (ft-lb)* | 43.5 | 38.5 | 39.7 | 37.5 | 48.5 | 46.5 |
| 0.2% YS (ksi)* | 102.8 | 114.5 | 104.7 | 114.2 | 108.7 | 111.2 |
| Annealed 1940°F/1hr/AC | | | | | | |
| DWTT (ft-lb) | 2728 | 2790 | 2733 | 2733 | 2733 | 2566 |
| $C_V$ at 30°F (ft-lb)* | 48.5 | 45 | 47.8 | 49.8 | 48.0 | 48.5 |
| 0.2% YS (ksi)* | 110 | 111.2 | 105.7 | 106.7 | 106.5 | 108.7 |
| Annealed 1940°F/1hr/WQ | | | | | | |
| DWTT (ft-lb) | 2705 | 2742 | 2325 | 2266 | 2266 | 2026 |
| $C_V$ at 30°F (ft-lb)* | 41 | 44.6 | 45 | 39 | 31.5 | 28.0 |
| 0.2% YS (ksi)* | 112.5 | 120.5 | 108.7 | 115.7 | 120.5 | 122 |

* Data furnished by Reactive Metals Inc.

TABLE 2

MECHANICAL PROPERTIES OF
EXTRUDED Ti-7Al-2Cb-1Ta ALLOY PLATES

| Heat Treatment and Mechanical Properties | Extrusion Temperature | | | |
|---|---|---|---|---|
| | 1700°F (Process D) | | 1900°F (Process E) | |
| | Long. | Trans. | Long. | Trans. |
| Annealed 1750°F/1hr/FC | | | | |
| DWTT (ft-lb) | 3282 | 1356 | 2245 | ND |
| $C_V$ at 30°F (ft-lb)* | 24 | 26.8 | 48.7 | 35.2 |
| 0.2% YS (ksi)* | 118.5 | 121.5 | 117.8 | 105.8 |
| Annealed 1750°F/1hr/WQ | | | | |
| DWTT (ft-lb) | 4350 | 2443 | 3574 | ND |
| $C_V$ at 30°F (ft-lb)* | 39.5 | 46.5 | 57 | 43.5 |
| 0.2% YS (ksi)* | 109.8 | 121 | 106.8 | 98.6 |
| Annealed 1925°F/1hr/FC | | | | |
| DWTT (ft-lb) | 3023 | ND | 1935 | ND |
| $C_V$ at 30°F (ft-lb)* | 62 | 37 | 43 | 39 |
| 0.2% YS (ksi)* | 119 | 89.2 | 100.3 | 97.3 |
| Annealed 1925°F/1hr/WQ | | | | |
| DWTT (ft-lb) | 3093 | ND | 2324 | ND |
| $C_V$ at 30°F (ft-lb)* | 46 | 43.5 | 34 | 34.8 |
| 0.2% YS (ksi)* | 122.5 | 113.5 | 116.5 | 114.0 |

ND - Not determined due to insufficient material

* Data furnished by Reactive Metals, Inc.

TABLE 3

MECHANICAL PROPERTIES OF
DIFFUSION-BONDED TITANIUM ALLOY PLATES

| Nominal Composition | Heat Treatment | Heat No. | DWTT (32°F) (ft-lb) | UTS (ksi) | YS (ksi) |
|---|---|---|---|---|---|
| *Ti-7Al-12Zr | Hot Rolled 2000 °F | RMI-T32558 | 870 | 120 | 110 |
| Ti-5Al-5Sn-5Zr | Hot Rolled 2000°F | TMCA-V1913 | 1200 | 134 | 112 |
| Ti-6Al-4V | Hot Rolled 1750°F | TMCA-M5257 | 2054 | 129 | 115 |
| Ti-4Al-3Mo-1V | Hot Rolled 1750°F | TMCA-T9046 | 1568 | 125 | 108 |
| Ti-16V-2.5Al | Hot Rolled 1750°F | CRU-R4813 | 455 | - | - |
| *B120VCA | Hot Rolled 1750°F | CRU-R6759 | 250-300 | - | - |

* Twenty-two pieces of 0.090-in. sheet rolled to 1-in. thickness; all other thirty pieces of 0.062-in. sheet rolled to 1-in. thickness.

TABLE 4

BETA TRANSUS TEMPERATURES
OF SEVERAL TITANIUM ALLOYS

| Nominal Alloy Composition | Beta Transus Temperature |
|---|---|
| Ti-6.5Al-5Zr-1V (T-36) | 1835°F ± 15°F |
| Ti-6Al-2Sn-1Mo-1V (T-37) | 1835°F ± 15°F |
| Ti-6Al-4Zr-2Mo (T-55) | 1840°F ± 15°F |
| Ti-6Al-4V-2Sn (T-67) | 1815°F ± 15°F |
| Ti-6Al-4Zr-2Sn-.5Mo-.5V (T-68) | 1865°F ± 15°F |
| Ti-7Al-2.5Mo (T-71) | 1865°F ± 15°F |

TABLE 5

TEST DATA FOR SOLUTION ANNEALING AND AGING TREATMENTS ON THE ALLOY Ti-8Al-1Mo-1V (T-19)*

| Solution Heat Treatment | Aging Heat Treatment | Longitudnl (RW) Charpy V Energy (ft-lb) -80°F | Longitudnl (RW) Charpy V Energy (ft-lb) +32°F | Transverse (WR) Charpy V Energy (ft-lb) -80°F | Transverse (WR) Charpy V Energy (ft-lb) +32°F | YS (0.2%) (ksi) | UTS (ksi) | Elong (%) | RA (%) |
|---|---|---|---|---|---|---|---|---|---|
| As Received | | 19.0 | 23.0 | 19.0 | 24.5 | 122.2(L) | 132.6(L) | 12.1 | 20.0 |
| | | | | | | 122.7(T) | 130.9(T) | 10.7 | 21.2 |
| 1850°F/1hr/AC | | 30.5 | 40.5 | 32.0 | 45.0 | 108.9(L) | 125.2(L) | 12.1 | 22.2 |
| | | | | | | 110.9(T) | 126.5(T) | 15.7 | 17.7 |
| 1850°F/1hr/AC | 1100°F/2hr/WQ | 31.0 | 37.0 | 30.0 | 40.0 | 118.9(L) | 128.8(L) | 10.0 | 14.9 |
| | | | | | | 117.7(T) | 128.4(T) | 10.7 | 14.8 |
| 1850°F/1hr/AC | 1200°F/2hr/WQ | 31.5 | 34.0 | 28.0 | 37.5 | 117.0(L) | 126.5(L) | 7.1 | 14.8 |
| | | | | | | 118.7(T) | 126.8(T) | 7.1 | 17.7 |
| 1800°F/1hr/AC | | 29.0 | 40.0 | 32.5 | 46.0 | 110.9(L) | 126.3(L) | 10.7 | 25.1 |
| | | | | | | 111.6(T) | 127.6(T) | 14.3 | 25.7 |
| 1800°F/1hr/AC | 1100°F/2hr/WQ | 27.5 | 37.0 | 31.5 | 36.5 | 116.8(L) | 128.3(L) | 12.9 | 21.2 |
| | | | | | | 117.5(T) | 129.9(T) | 12.1 | 18.3 |
| 1800°F/1hr/AC | 1200°F/2hr/WQ | 28.0 | 36.0 | 27.5 | 34.0 | 120.2(L) | 129.1(L) | 11.4 | 21.9 |
| | | | | | | 120.9(T) | 130.4(T) | 10.7 | 18.3 |
| 1750°F/1hr/AC | | 34.0 | 47.0 | 32.0 | 46.5 | 110.6(L) | 127.3(L) | 14.3 | 25.1 |
| | | | | | | 108.7(T) | 125.2(T) | 10.7 | 29.0 |
| 1750°F/1hr/AC | 1100°F/2hr/WQ | 30.5 | 38.0 | 30.0 | 36.0 | 117.9(L) | 129.4(L) | 10.7 | 21.2 |
| | | | | | | 118.2(T) | 129.4(T) | 13.6 | 21.2 |
| 1750°F/1hr/AC | 1200°F/2hr/WQ | 28.0 | 37.0 | 28.0 | 41.5 | 120.4(L) | 129.3(L) | 8.6 | 16.6 |
| | | | | | | 119.6(T) | 128.8(T) | 9.3 | 22.4 |
| 1750°F/1hr/WQ | 1100°F/2hr/AC | 18 | 37 | 20 | 26 | 127.6(L) | 135.8(L) | 8.6 | 11.1 |
| | | | | | | 128.9(T) | 136.1(T) | 5.0 | 11.8 |
| 1750°F/1hr/WQ | 1200°F/2hr/AC | 23 | 30 | 24 | 28 | 128.1(L) | 136.2(L) | 5.7 | 10.0 |
| | | | | | | 129.8(T) | 137.0(T) | 5.0 | 10.0 |

* Beta transus 1885°F ±15°F

(Table continues)

TABLE 5 (Continued)

TEST DATA FOR SOLUTION ANNEALING AND AGING TREATMENTS
ON THE ALLOY Ti-8Al-1Mo-1V (T-19)*

| | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|
| 1700°F/1hr/AC | | 33 | 50 | 42 | 51 | 111.3(L) | 127.0(L) | 11.4 | 22.9 |
| | | | | | | 111.6(T) | 127.3(T) | 11.4 | 22.9 |
| 1700°F/1hr/AC | 1100°F/2hr/WQ | 33 | 44 | 32 | 41 | 115.2(L) | 127.6(L) | 14.3 | 24.0 |
| | | | | | | 118.1(T) | 127.3(T) | 10.7 | 22.9 |
| 1700°F/1hr/AC | 1200°F/2hr/WQ | 32 | 39 | 28 | 40 | 120.1(L) | 129.6(L) | 12.1 | 21.7 |
| | | | | | | 121.4(T) | 130.9(T) | 12.1 | 21.2 |
| 1850°F/1hr/WQ | | 20 | 25 | 22 | 30 | 124.2(L) | 153.4(L) | 6.4 | 17.7 |
| | | | | | | 128.6(T) | 158.7(T) | 5.7 | 11.1 |
| 1850°F/1hr/WQ | 1100°F/2hr/AC | 12 | 15 | 14 | 18 | 147.3(L) | 159.7(L) | 4.3 | 8.1 |
| | | | | | | 152.2(T) | 162.3(T) | 3.6 | 8.8 |
| 1850°F/1hr/WQ | 1200°F/2hr/AC | 14 | 20 | 16 | 23 | 147.3(L) | 156.1(L) | 4.3 | 9.4 |
| | | | | | | 142.7(T) | 152.1(T) | 3.6 | 8.1 |
| 1800°F/1hr/WQ | | 33 | 36 | 30 | 35 | 116.8(L) | 146.3(L) | 10.0 | 25.1 |
| | | | | | | 119.6(T) | 149.5(T) | 10.0 | 19.4 |
| 1800°F/1hr/WQ | 1100°F/2hr/AC | 19 | 23 | 20 | 23 | 138.7(L) | 150.9(L) | 5.7 | 9.4 |
| | | | | | | 137.5(T) | 150.8(T) | 4.3 | 11.2 |
| 1800°F/1hr/WQ | 1200°F/2hr/AC | 22 | 22 | 23 | 28 | 136.8(L) | 146.6(L) | 5.7 | 10.6 |
| | | | | | | 135.6(T) | 145.6(T) | 4.3 | 10.5 |
| 1750°F/1hr/WQ | | 33 | 42 | 33 | 42 | 103.4(L) | 139.5(L) | 12.1 | 22.2 |
| | | | | | | 107.0(T) | 140.4(T) | 10.0 | 27.3 |
| 1750°F/1hr/WQ | 1100°F/2hr/AC | 18 | 37 | 20 | 26 | 127.6(L) | 135.8(L) | 8.6 | 11.1 |
| | | | | | | 128.9(T) | 136.1(T) | 5.0 | 11.8 |
| 1750°F/1hr/WQ | 1200°F/2hr/AC | 23 | 30 | 24 | 28 | 128.1(L) | 136.2(L) | 5.7 | 10.0 |
| | | | | | | 129.8(T) | 137.0(T) | 5.0 | 10.0 |
| 1700°F/1hr/WQ | | 34 | 38 | 34 | 42 | 98.2(L) | 136.5(L) | 11.4 | 23.4 |
| | | | | | | 102.1(T) | 137.1(T) | 10.0 | 17.7 |
| 1700°F/1hr/WQ | 1100°F/2hr/AC | 18 | 22 | 20 | 22 | 129.9(L) | 138.4(L) | 5.0 | 10.6 |
| | | | | | | 130.0(T) | 138.3(T) | 3.6 | 11.1 |
| 1700°F/1hr/WQ | 1200°F/2hr/AC | 22 | 25 | 22 | 33 | 128.3(L) | 135.8(L) | 5.7 | 15.4 |
| | | | | | | 128.1(T) | 136.5(T) | 7.1 | 14.8 |

* Beta transus 1885°F ± 15°F

TABLE 6

CHEMICAL COMPOSITION OF 5Ni-Cr-Mo-V STEELS

| Steel No. | Chemical Composition - Wt. - % | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | C | Mn | S | P | Si | Ni | Cr | Mo | V | Sol Al |
| Melt range | 0.08/ 0.13 | 0.65/ 0.90 | 0.010 Max. | 0.010 Max. | 0.20/ 0.35 | 4.80/ 5.30 | 0.45/ 0.65 | 0.47/ 0.62 | 0.04/ 0.07 | 0.015/ 0.040 |
| Heat #1 | 0.11 | 0.73 | 0.005 | 0.006 | 0.27 | 5.00 | 0.57 | 0.53 | 0.06 | 0.018 |
| Heat #2 | 0.10 | 0.83 | 0.008 | 0.005 | 0.22 | 5.20 | 0.52 | 0.53 | 0.08 | 0.015 |

NOTE: The above are averages of the range determined for the different plates produced from these heats.

TABLE 7

NON-INTERSTITIAL CHEMICAL COMPOSITION
OF TITANIUM ALLOYS*

| Code | Material | % Composition by Weight | | | | | | |
|---|---|---|---|---|---|---|---|---|
| | | Al | V | Cb | Ta | Fe | Mn | Mg |
| T-16 | Unalloyed Titanium | - | - | - | - | .1 | - | - |
| T-5 | Ti-6Al-4V | 5.85 | 3.85 | - | - | .19 | .03 | .0005 |
| TA-2 | Ti-7Al-2Cb-1Ta | 6.9 | - | 2.5 | 1.1 | .13 | - | - |

*Producer's Data

TABLE 8

INTERSTITIAL CHEMICAL COMPOSITION
OF TITANIUM ALLOYS*

| Code | Material | % Composition by Weight | | | |
|---|---|---|---|---|---|
| | | $O_2$ | $N_2$ | C | $H_2$ |
| T-16 | Unalloyed Titanium | .07 | .009 | .029 | .006 |
| T-5 | Ti-6Al-4V | .06 | .06 | .042 | .74 |
| TA-2 | Ti-7Al-2Cb-1Ta | .063 | .006 | .01 | - |

*Producer's Data

TABLE 9

MECHANICAL PROPERTIES OF TITANIUM ALLOYS

| Code | Material | Tensile Test Data | | | | $C_v$ Energy (32°F) (ft-lb) | DWTT Energy (32°F) (ft-lb) |
|---|---|---|---|---|---|---|---|
| | | 0.2% YS (ksi) | UTS (ksi) | Elongation (%) | RA (%) | | |
| T-16 | Unalloyed Titanium | 34 (.313-in. diam) | 47 | 40 (1.4-in. gage length) | 84 | 163 | 9625 |
| T-5 | Ti-6Al-4V | 125 (.505-in. diam) | 131 | 12 (2-in. gage length) | 44 | 29 | 2075 |
| T-27 | Ti-6Al-4V | 116 (.505-in. diam) | 128 | 12 (2-in. gage length) | 26 | -- | 1228 |
| TA-2 | Ti-7Al-2Cb-1Ta | 107* | 120* | 13* | 28* | 39 | 2675 |

* Data received with material from Reactive Metals, Inc.

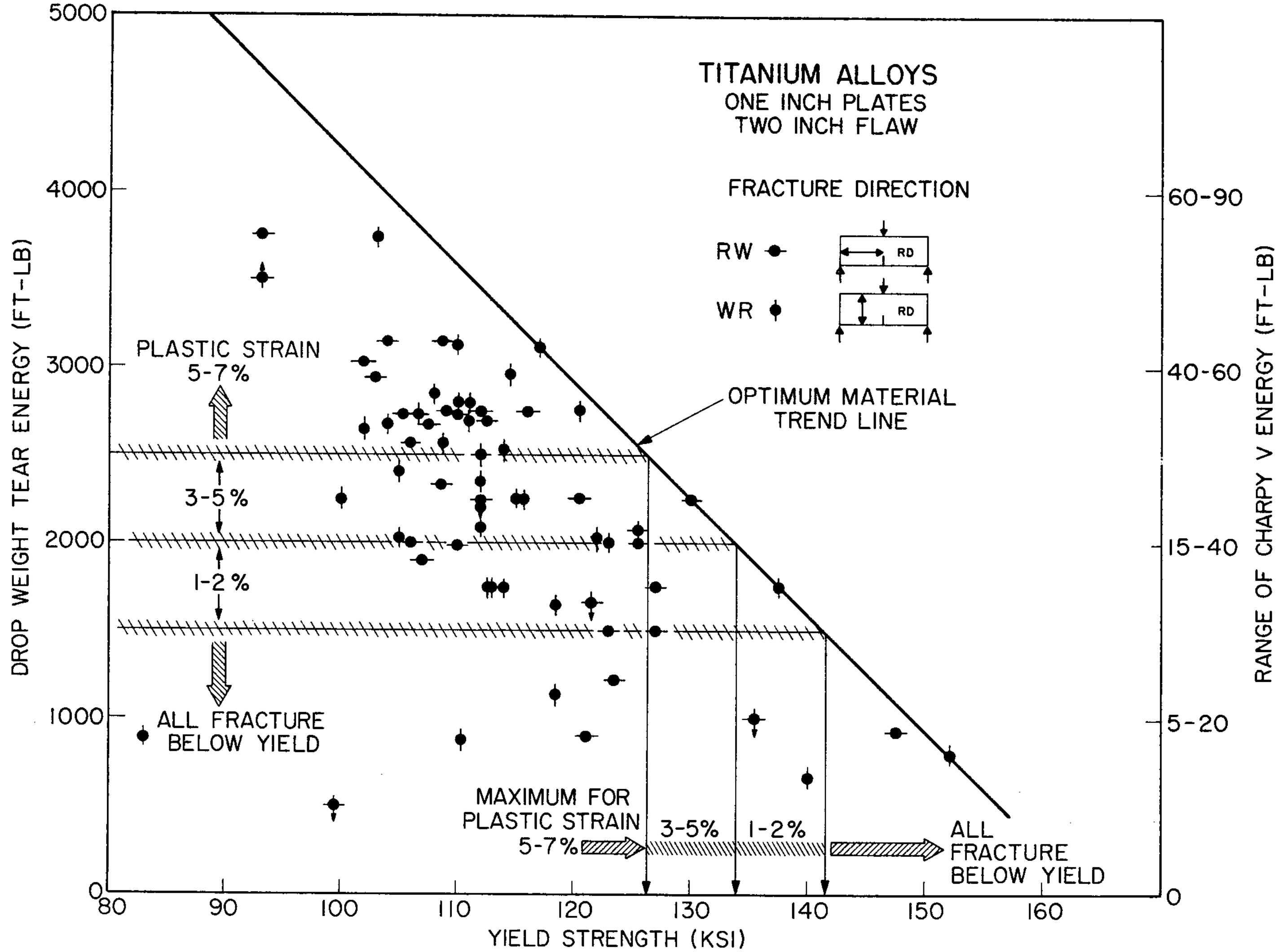



Fig. 1 - Fracture toughness diagram for titanium. Correlates drop-weight tear test, Charpy V, explosion tear test, and yield strength data for 1-inch thick titanium alloy plates. Optimum materials trend line indicates estimated highest level of strength for any given level of toughness.

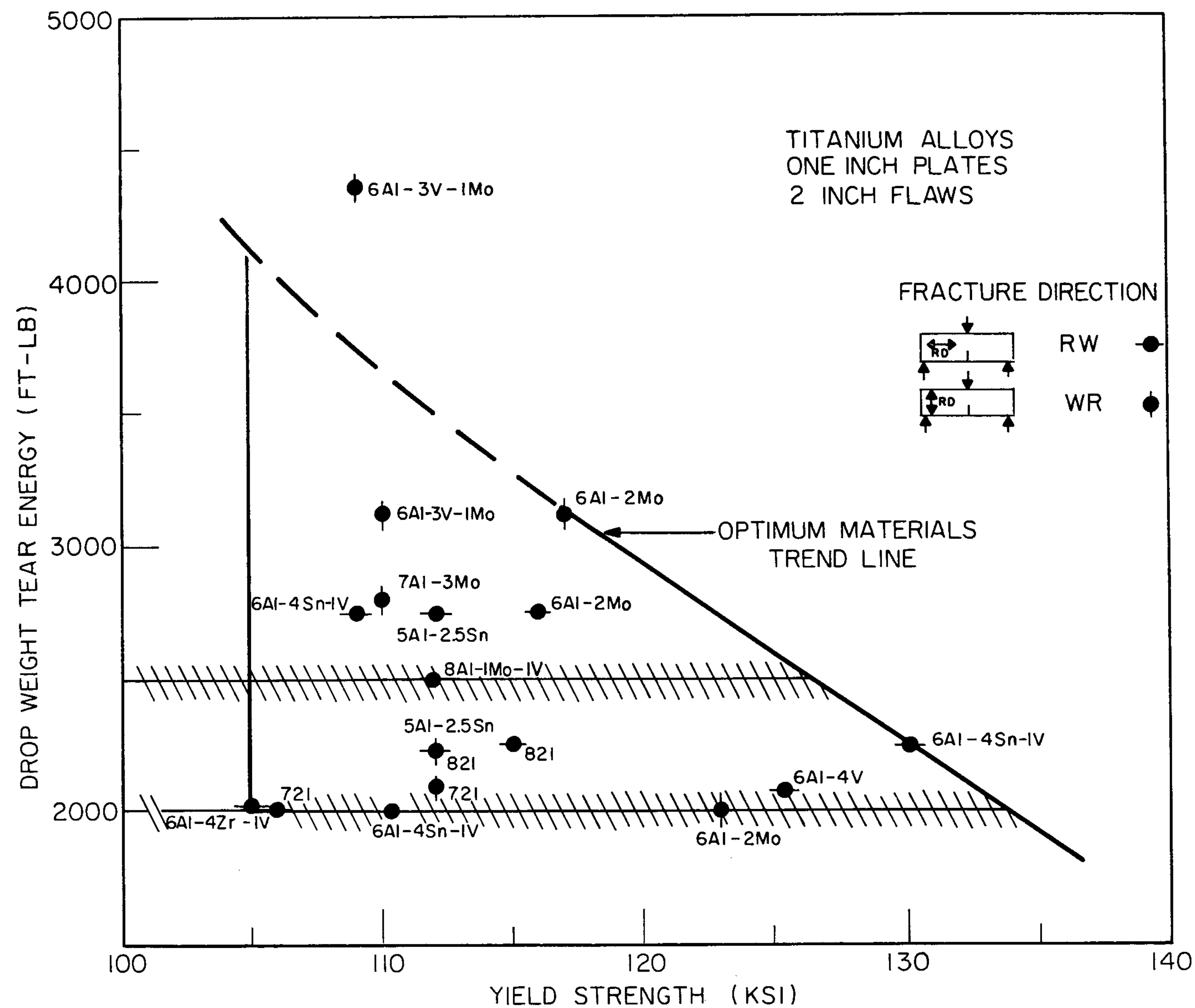



Fig. 2 - Portion of fracture toughness diagram for titanium showing variety of alloys that are within desirable limits for hull materials

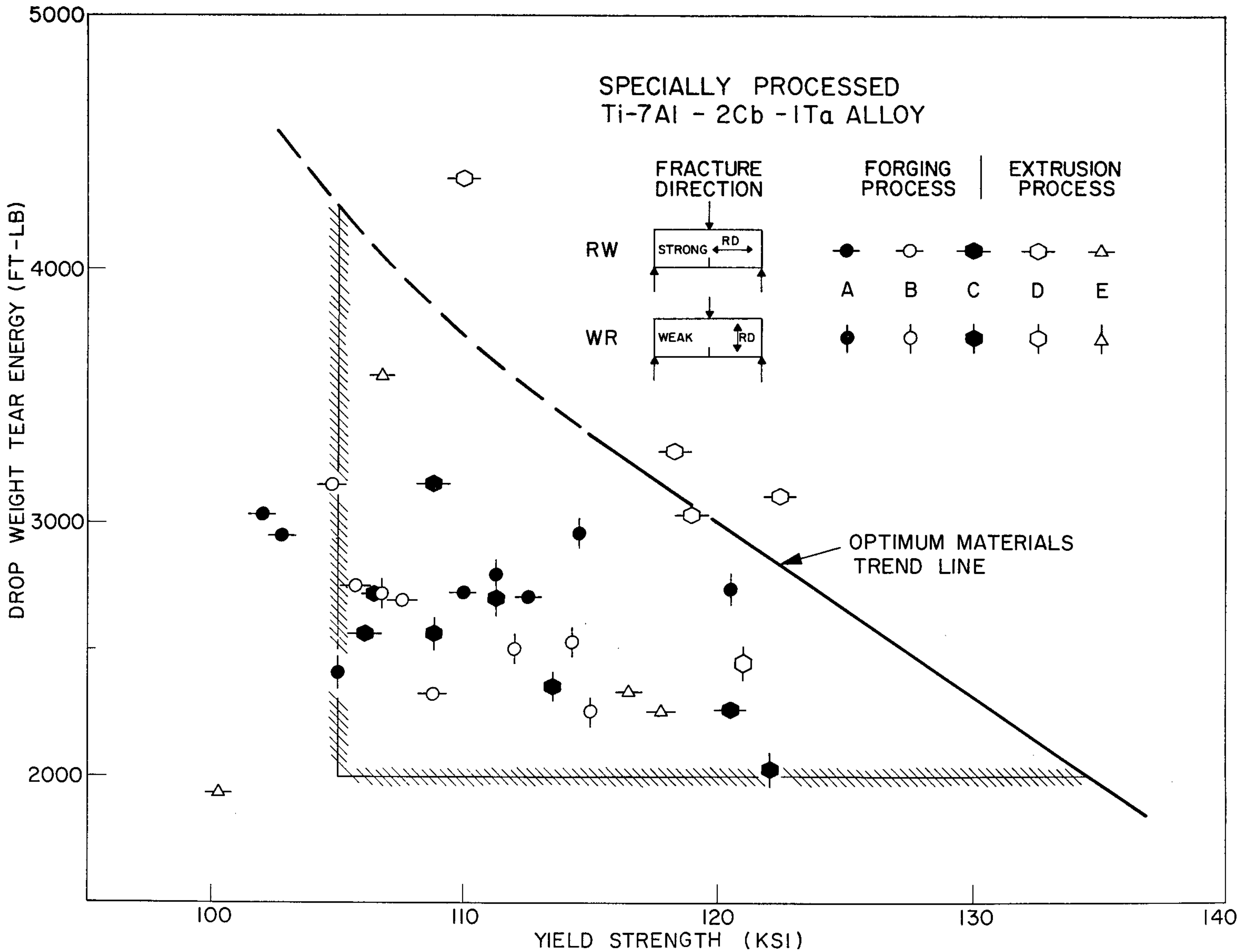



Fig. 3 - Results obtained with specially processed Ti-7Al-2Cb-1Ta alloy. Optimum materials trend line established for forged-and-rolled plate material in earlier studies.

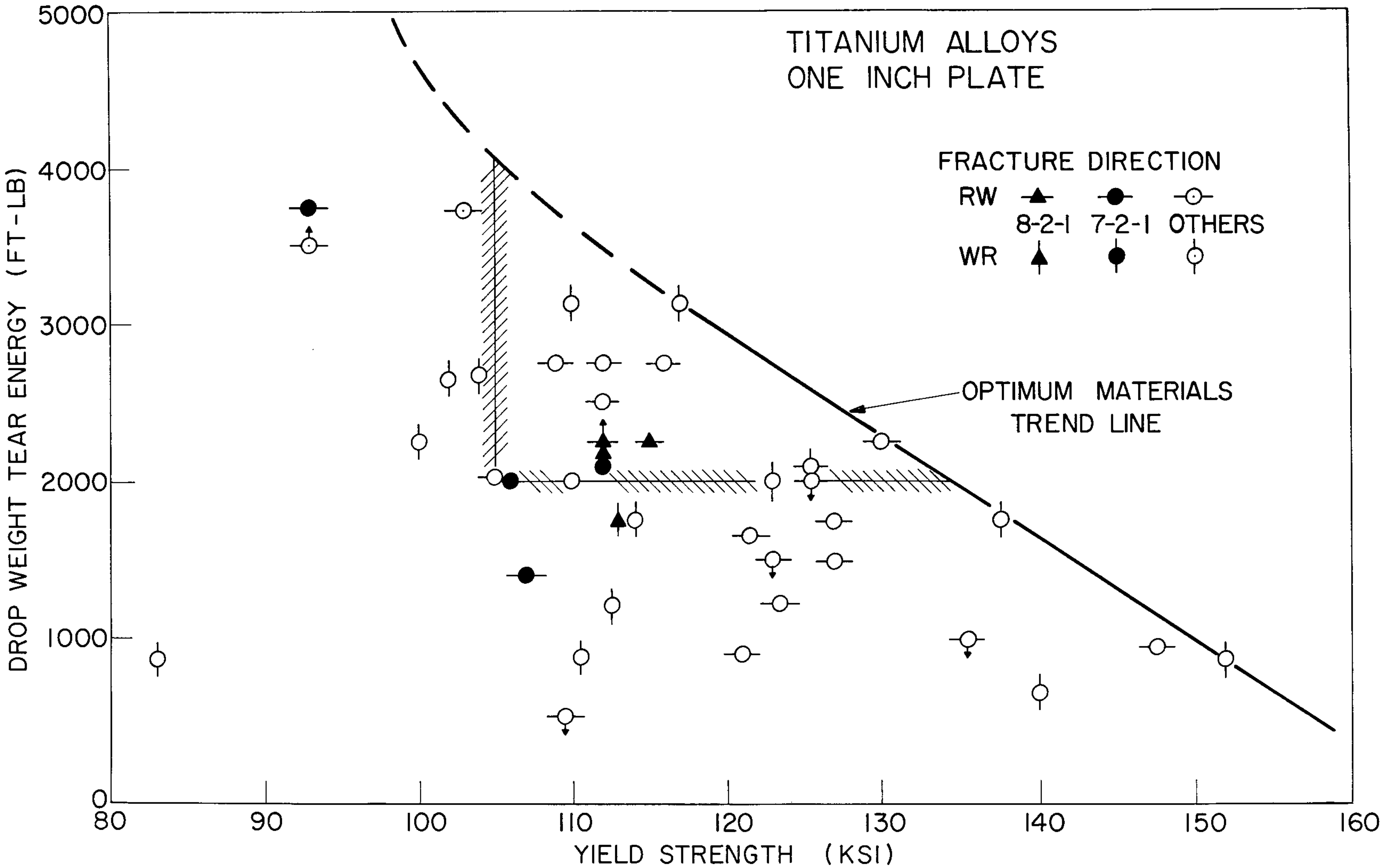
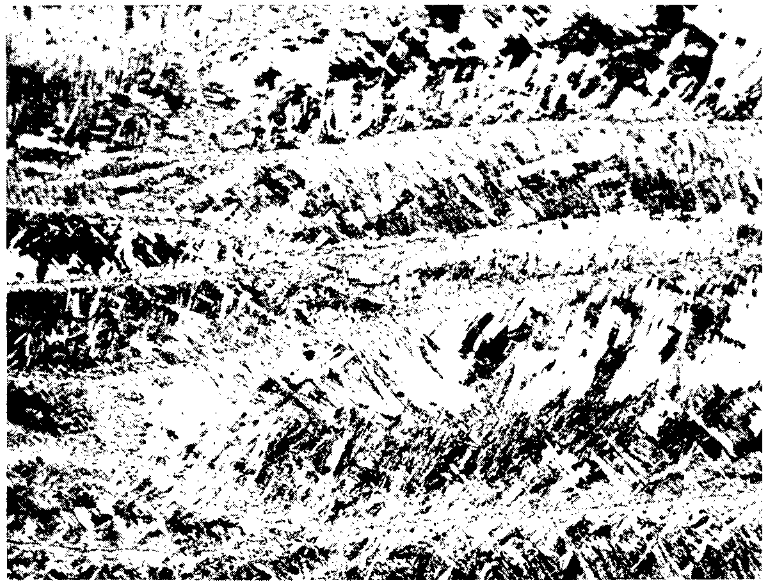



Fig. 4 - Fracture toughness properties of Ti-7Al-2Cb-1Ta and Ti-8Al-2Cb-1Ta alloys investigated prior to special processing program. Shows fracture toughness properties relative to all other alloys investigated.

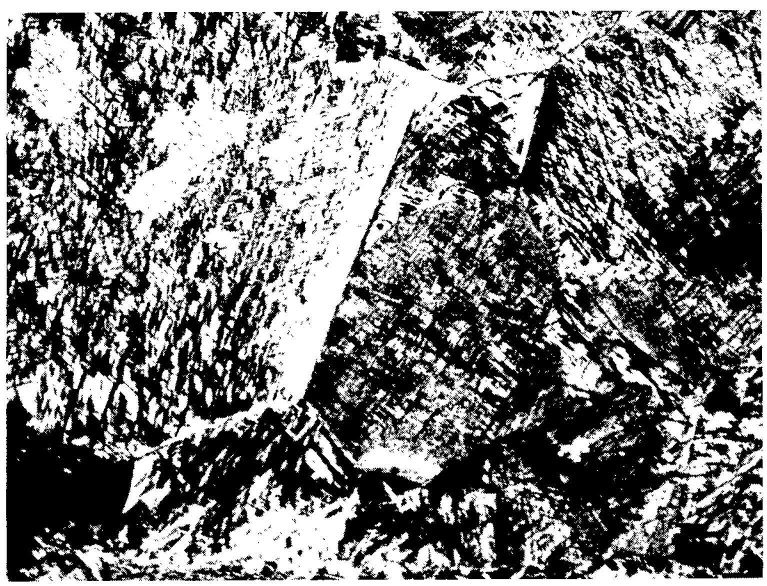
(a)

(b)

Fig. 5 - Microstructure of Ti-6.5Al-5Zr-1V (T-36) heat treated above and below the $\beta$ transus. (a) 1825°F for one hour in an argon atmosphere. 50X. (b) 1850°F for one hour in an argon atmosphere. 50X.

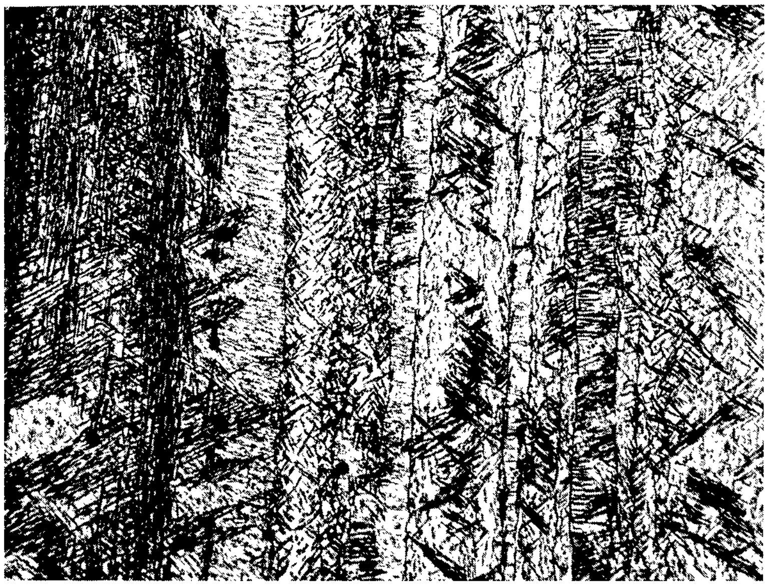


(a)

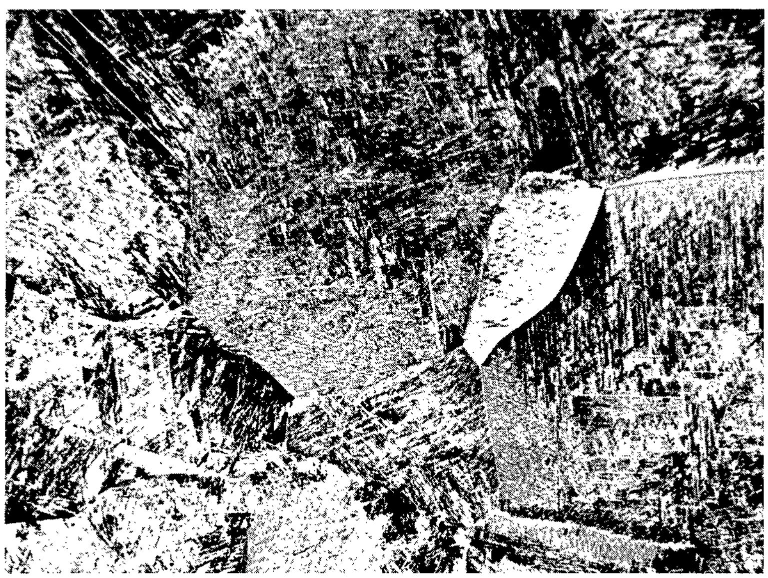


(b)

Fig. 6 - Microstructure of Ti-6Al-2Sn-1Mo-1V (T-37) heat treated above and below the $\beta$ transus. (a) 1825°F for one hour in an argon atmosphere. 50X. (b) 1850°F for one hour in an argon atmosphere. 50X.

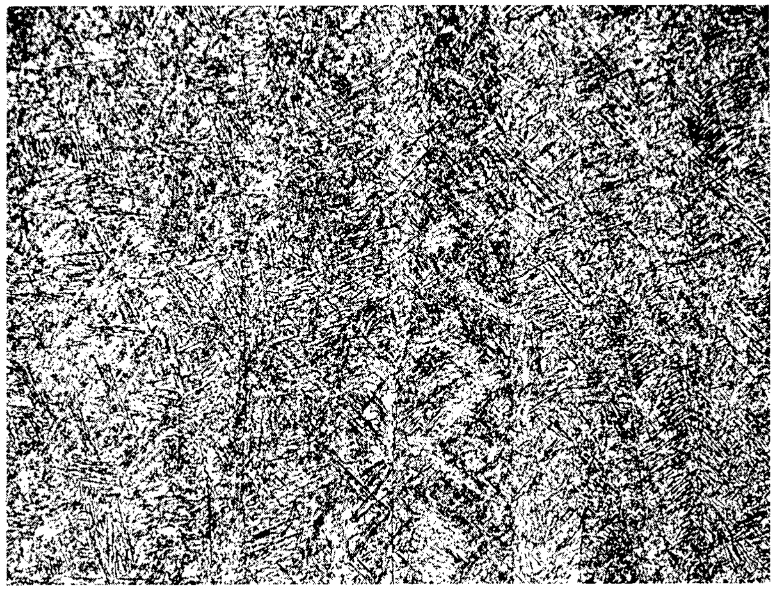


(a)

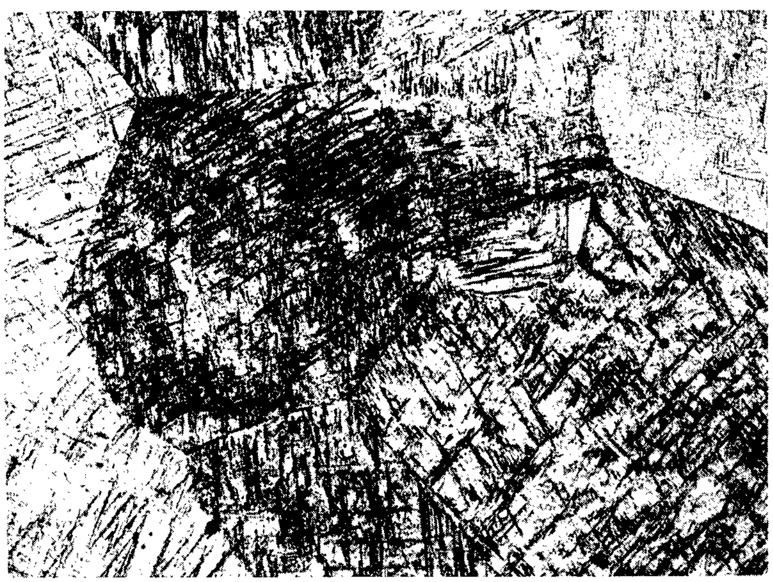


(b)

Fig. 7 - Microstructure of Ti-6Al-4Zr-2Mo (T-55) heat treated above and below the $\beta$ transus. (a) 1825°F for one hour in an argon atmosphere. 50X. (b) 1850°F for one hour in an argon atmosphere. 50X.

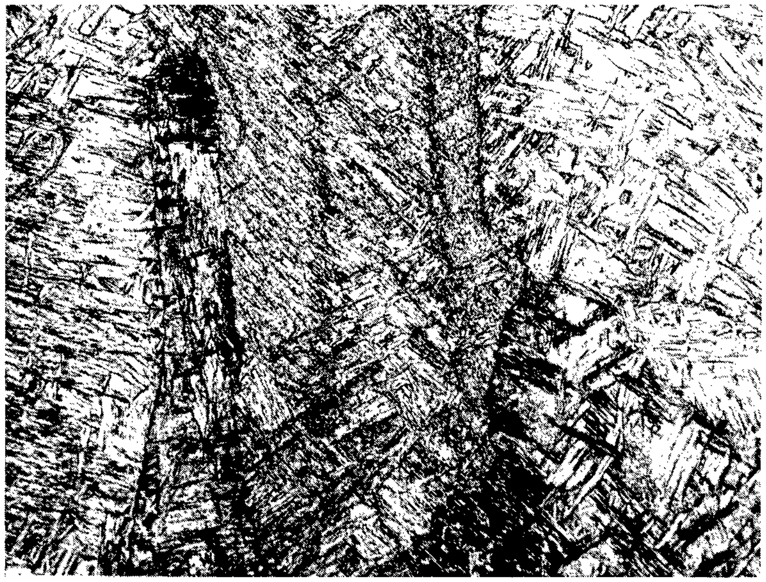


(a)

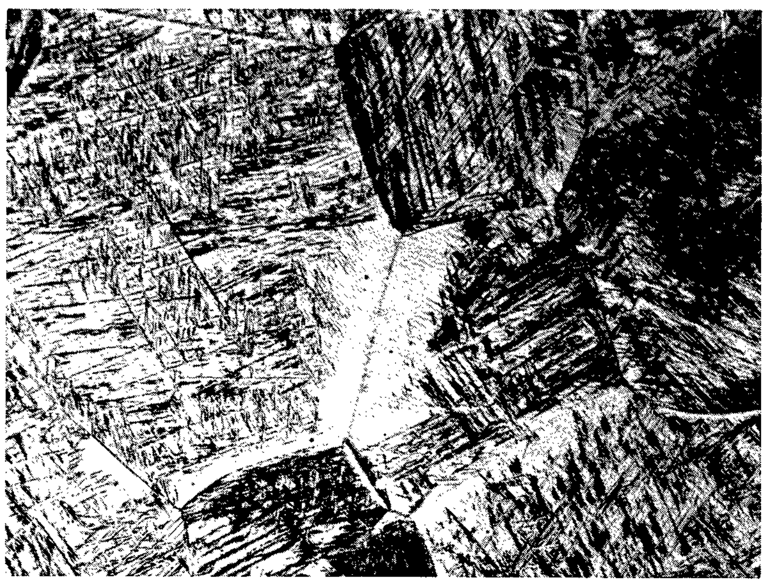


(b)

Fig. 8 - Microstructure of Ti-6Al-4V-2Sn (T-67) heat treated above and below the $\beta$ transus. (a) 1800° F for one hour in an argon atmosphere. 50X. (b) 1825°F for one hour in an argon atmosphere. 50X.



(a)



(b)

Fig. 9 - Microstructure of Ti-6Al-4Zr-2Sn-.5Mo-.5V (T-68) heat treated above and below the $\beta$ transus. (a) 1850°F heat for one hour in an argon atmosphere. 50X. (b) 1875°F for one hour in an argon atmosphere. 50X.

(a)

(b)

Fig. 10 - Microstructure of Ti-7Al-2.5Mo (T-71) heat treated above and below the $\beta$ transus. (a) 1850 °F for one hour in an argon atmosphere. 50X. (b) 1875° F for one hour in an argon atmosphere. 50X.

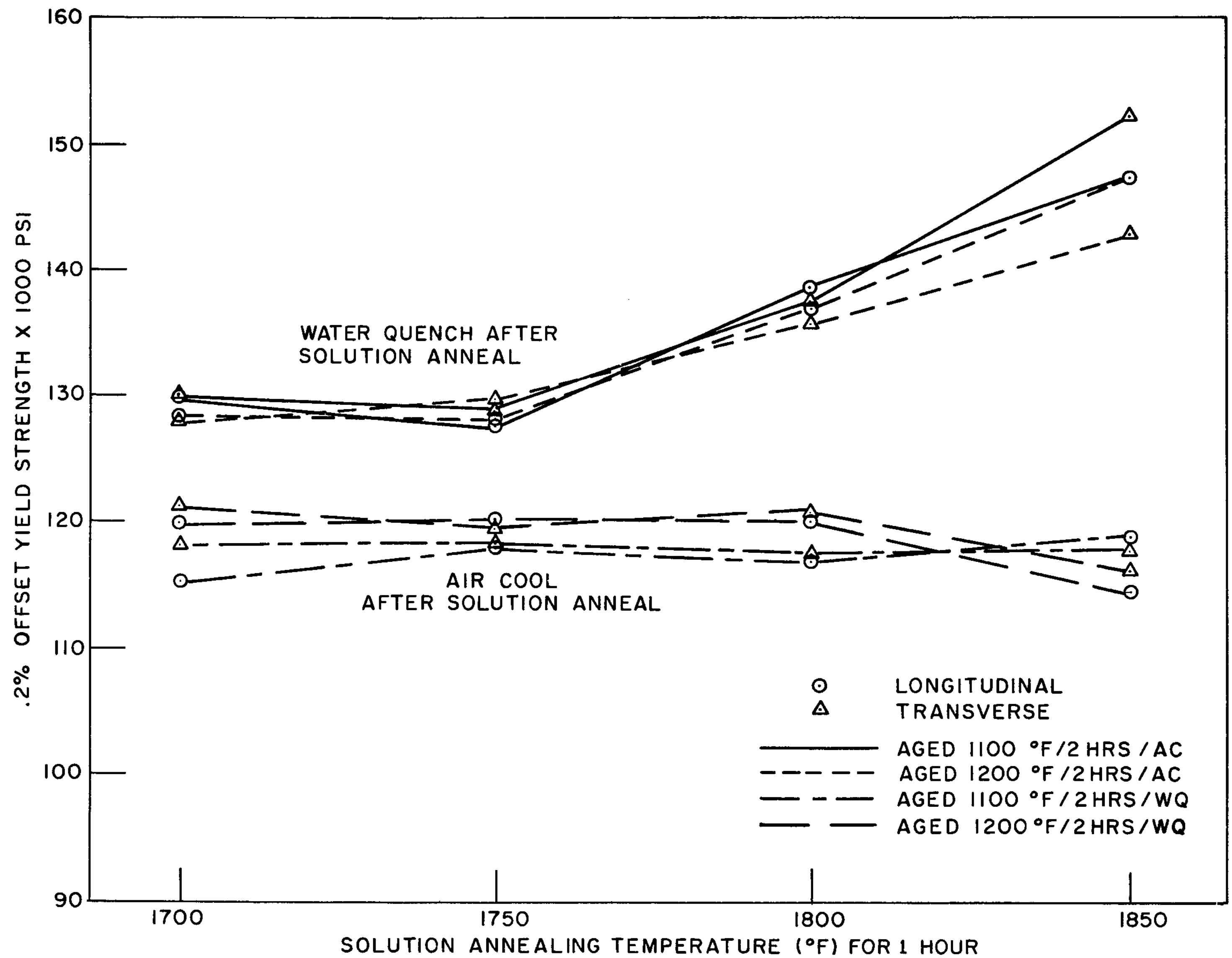



Fig. 11 - Effect of variations in solution annealing temperatures on the yield strength of the alloy Ti-8Al-1Mo-1V (T-19) with aging treatment

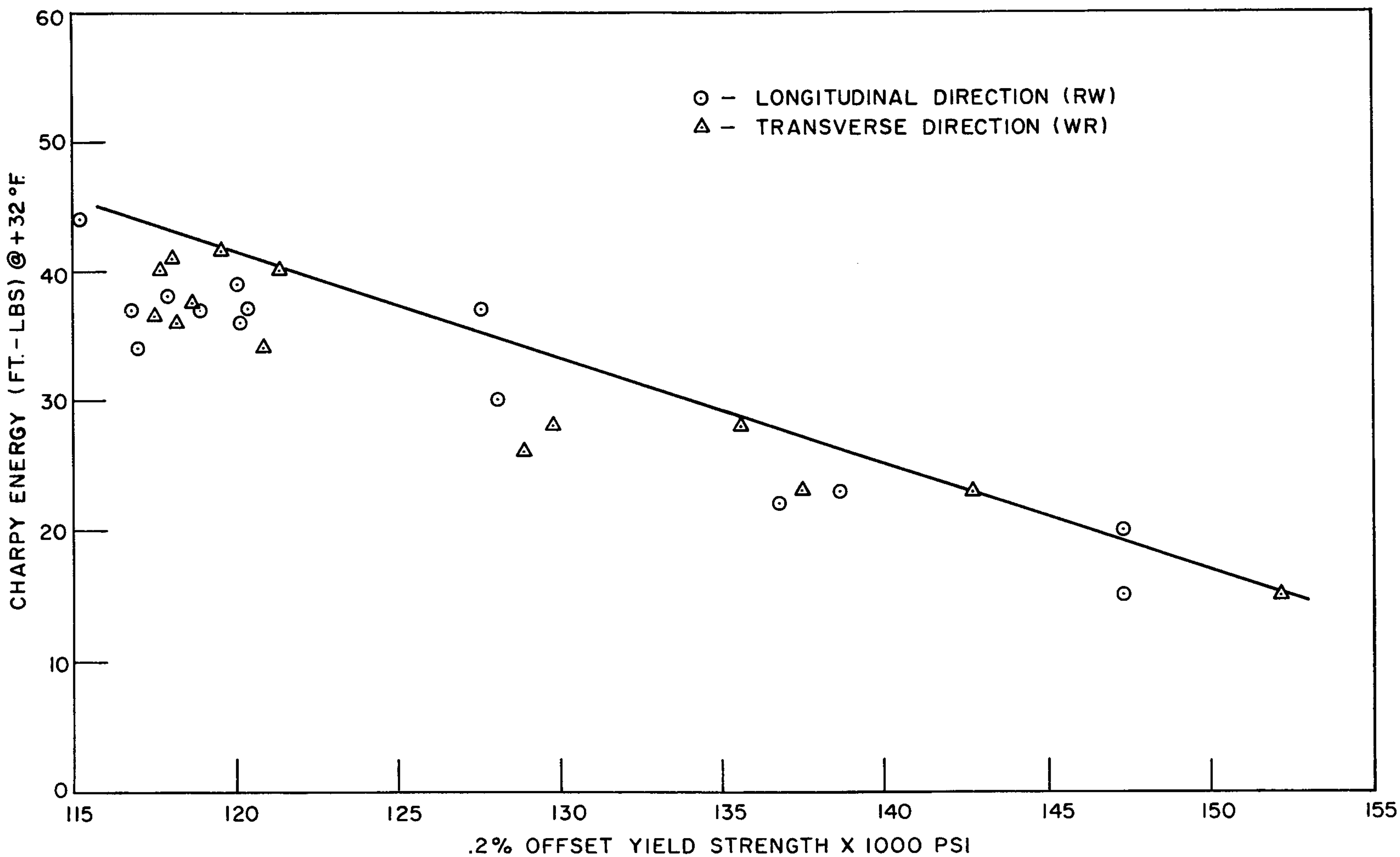



Fig. 12 - Summary of Charpy V energy and yield strength relationships for the alloy Ti-8Al-1Mo-1V (T-19)

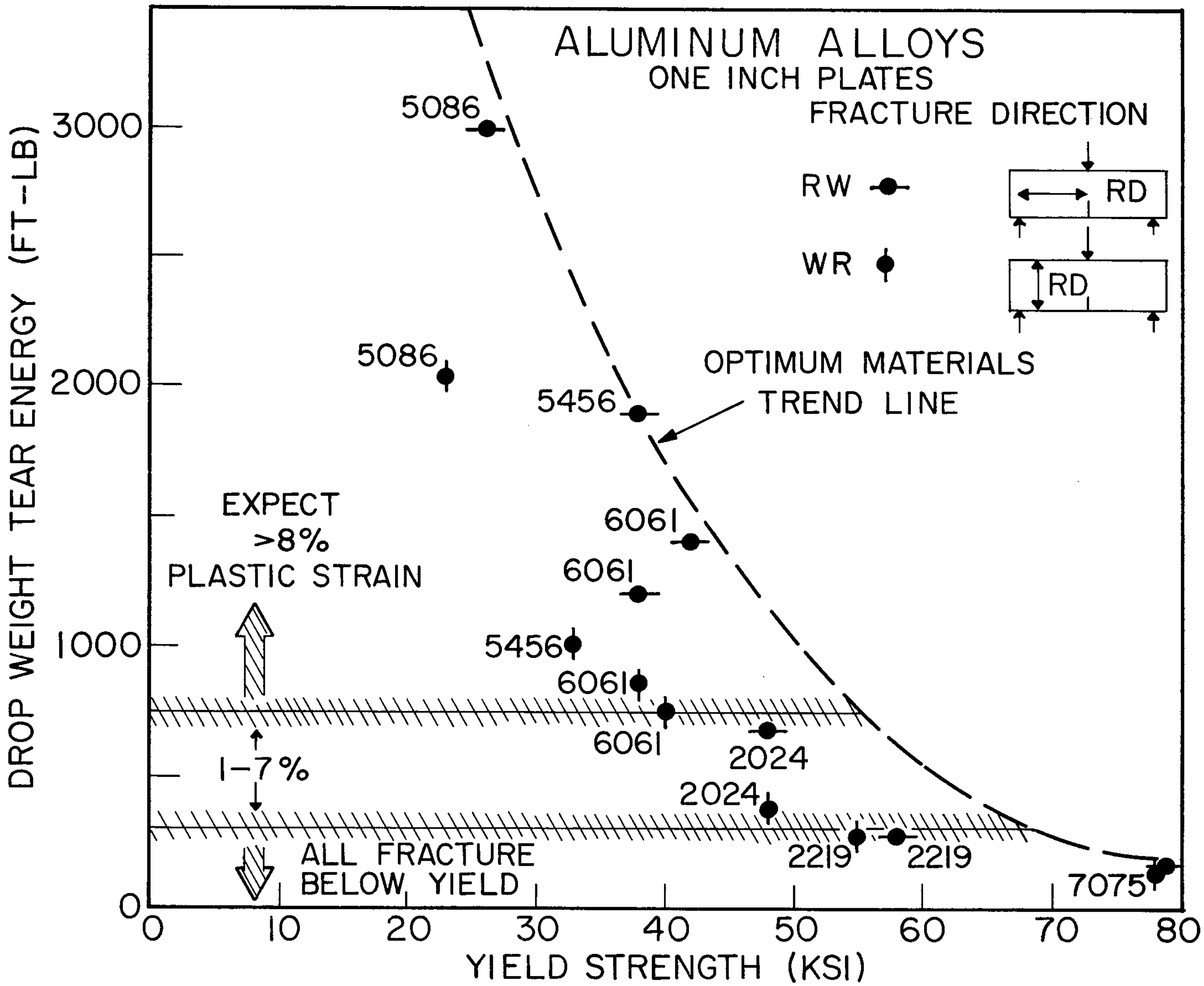



Fig. 13 - Preliminary fracture toughness diagram for aluminum; relates drop-weight tear test energy with performances in the explosion tear test and with yield strength

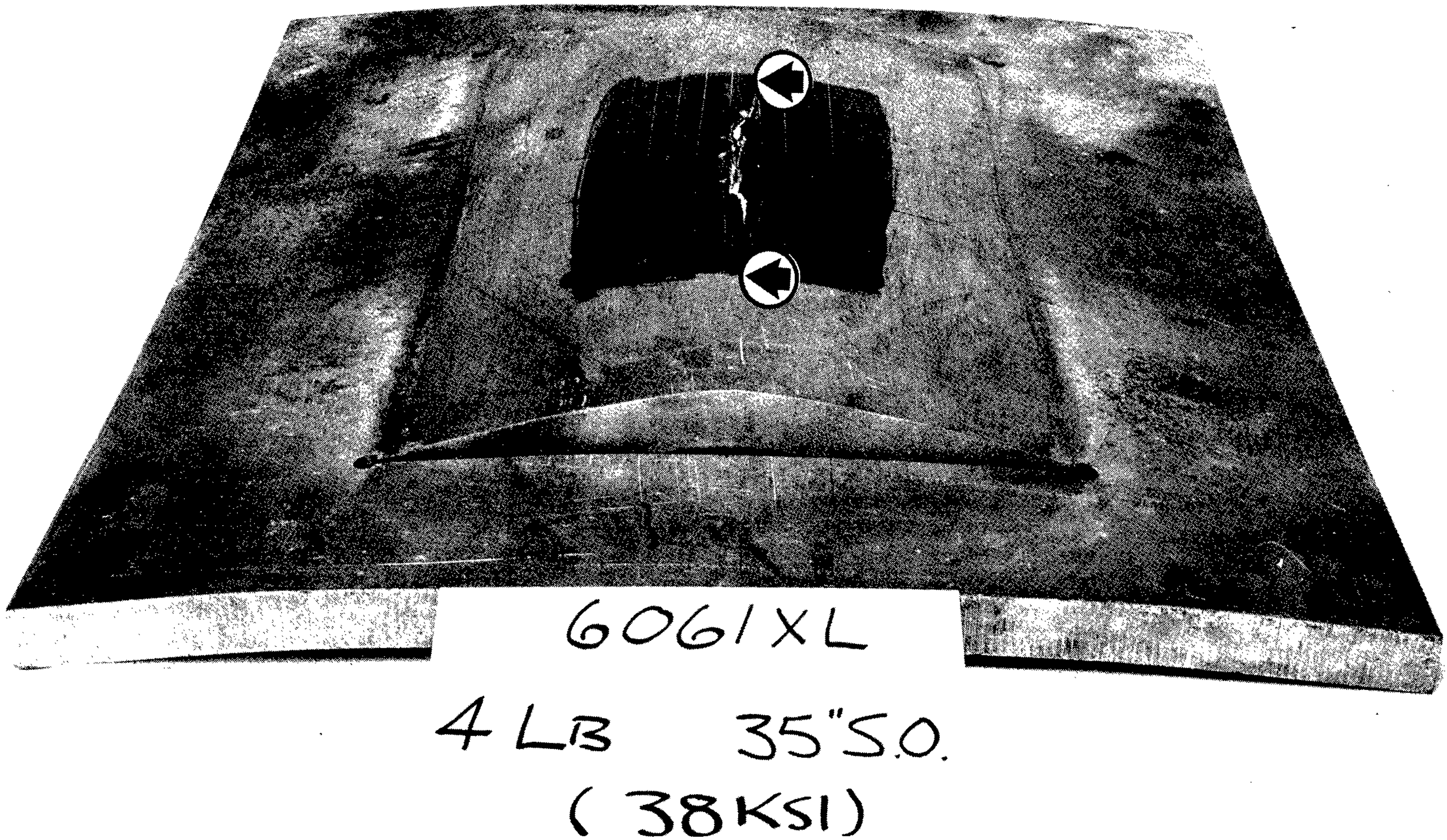



Fig. 14 - Explosion tear test plate of 6061-T651 tested to 7.3% plastic strain in the presence of a 2-in. flaw. Arrows indicate limit of crack extension.

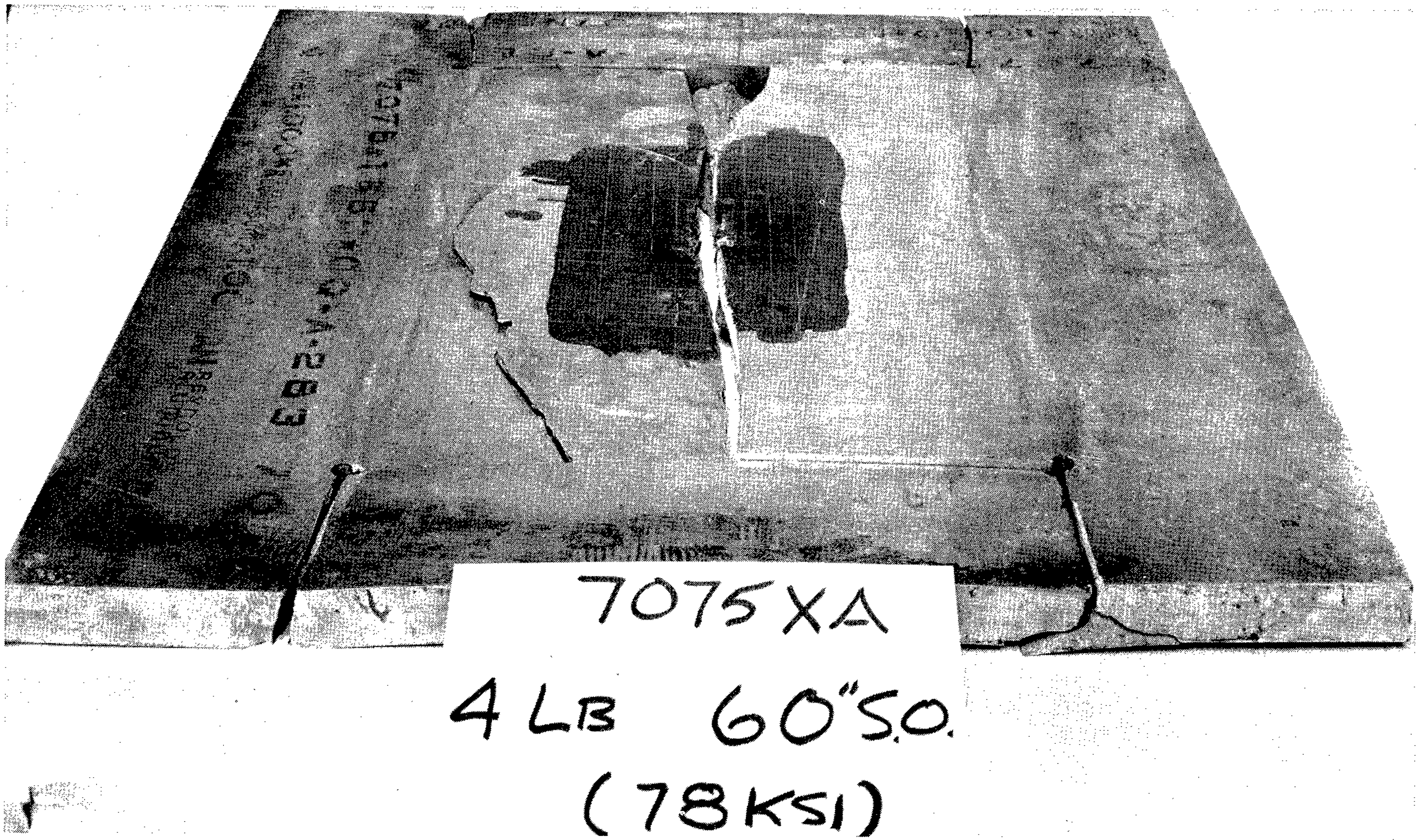



Fig. 15 - Explosion tear test plate of 7075-T6. Failure occurred under elastic conditions in the presence of a 2-in. flaw.

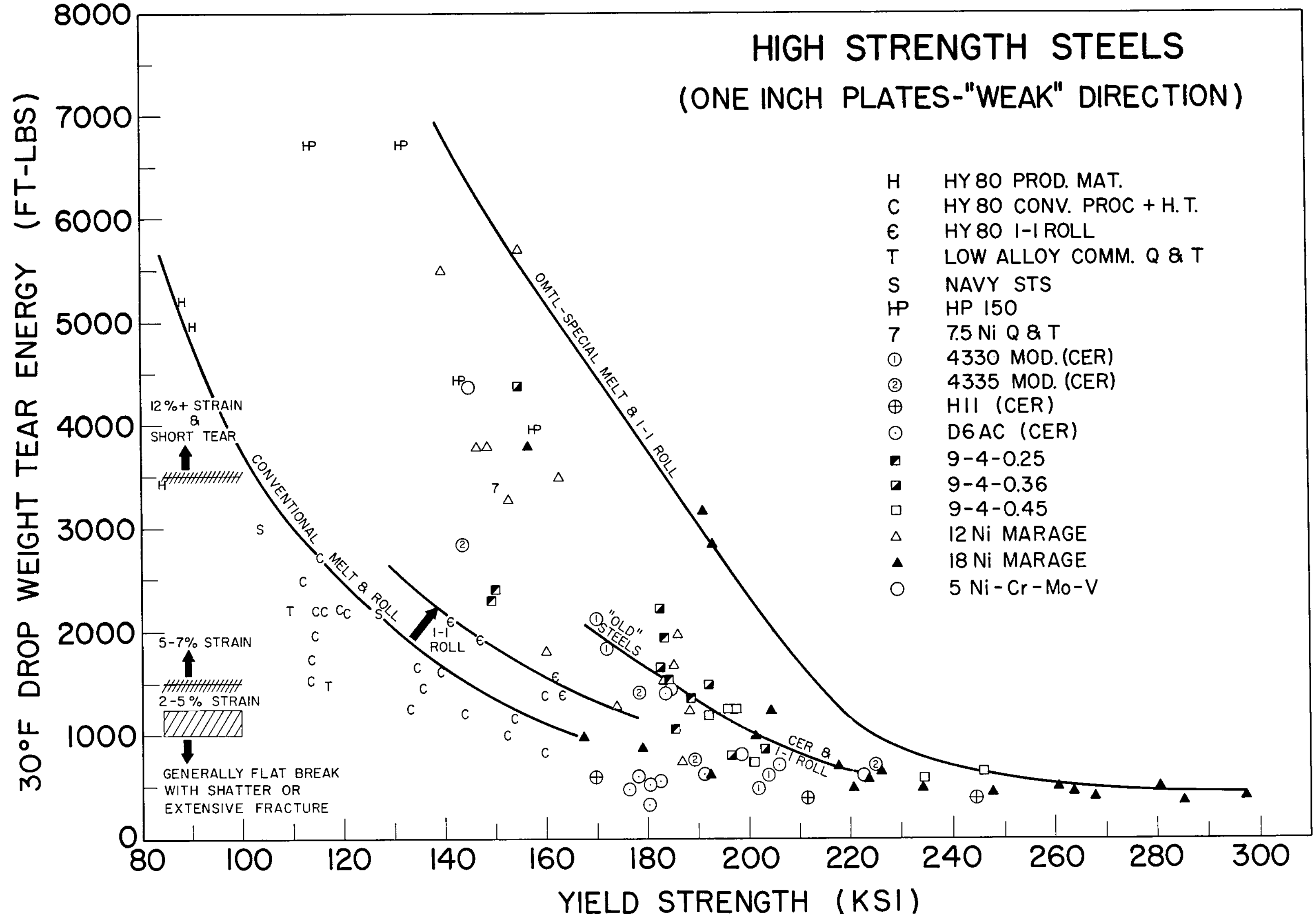



Fig. 16 - Summary of the drop-weight tear test results for high strength steels as a function of the yield strength of the test material

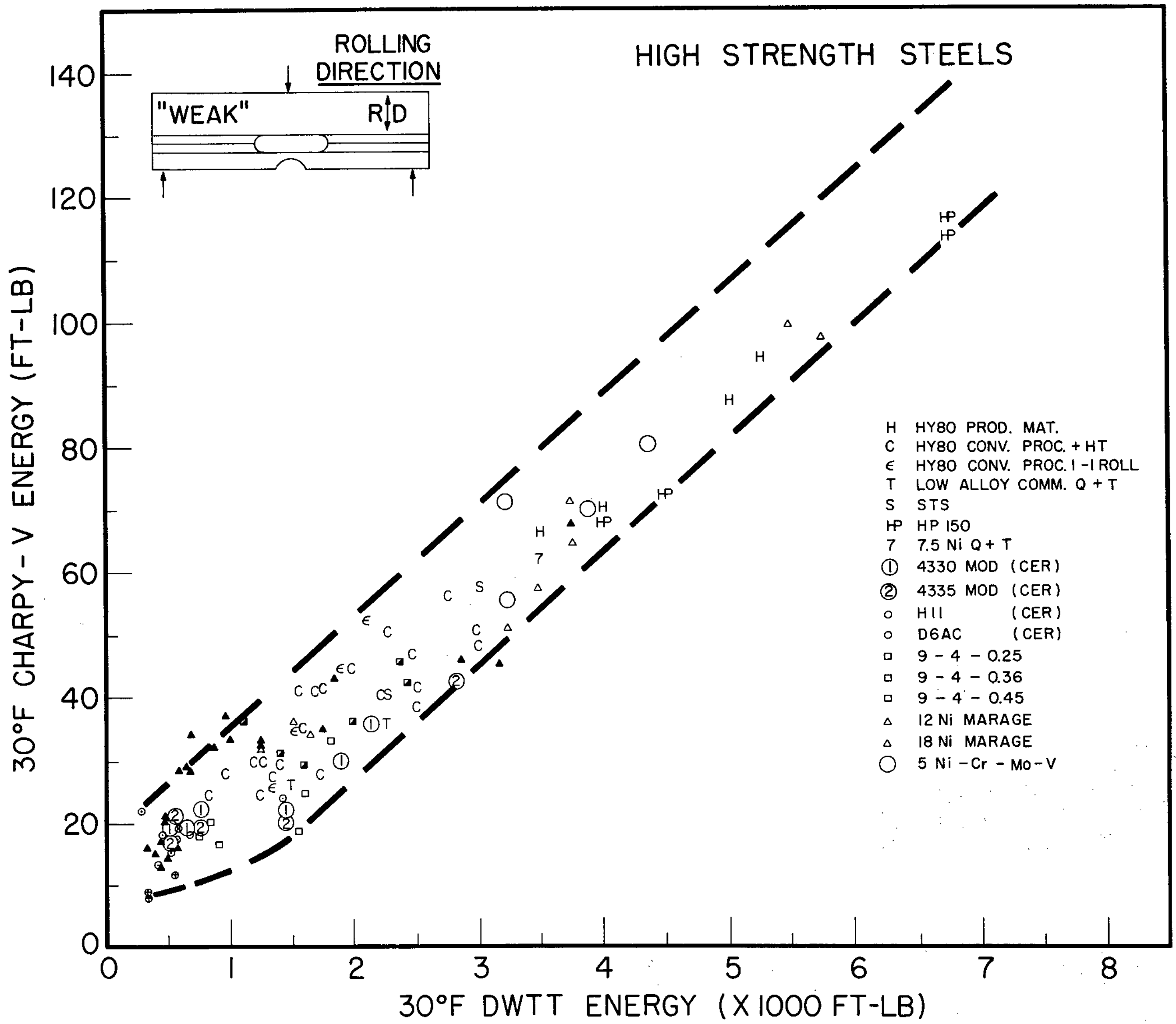



Fig. 17 - Correlation of Charpy V-notch energy with drop-weight tear energy at 30° F

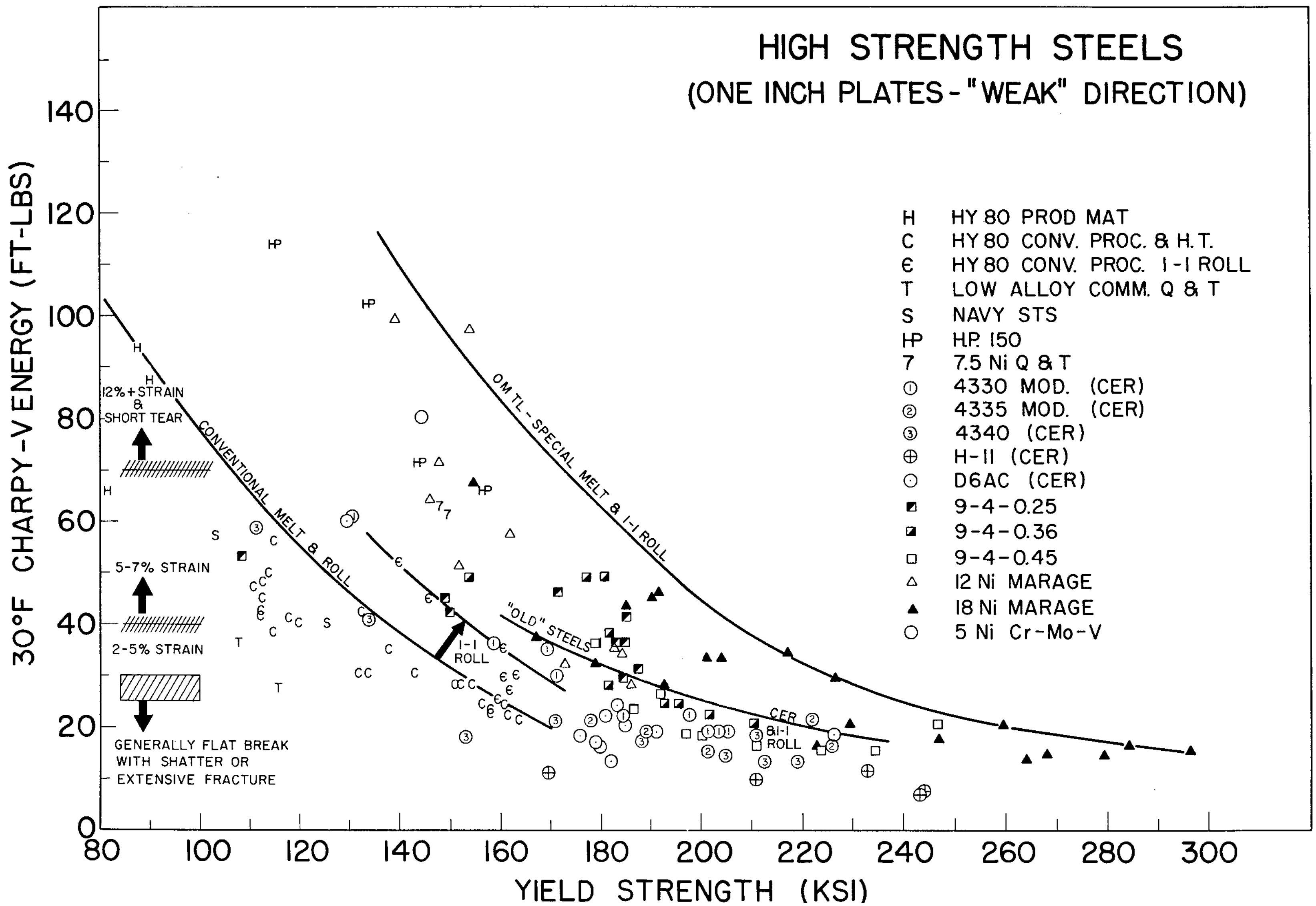



Fig. 18 - Summary of Charpy V test results for high strength steels as a function of the yield strength of the test material

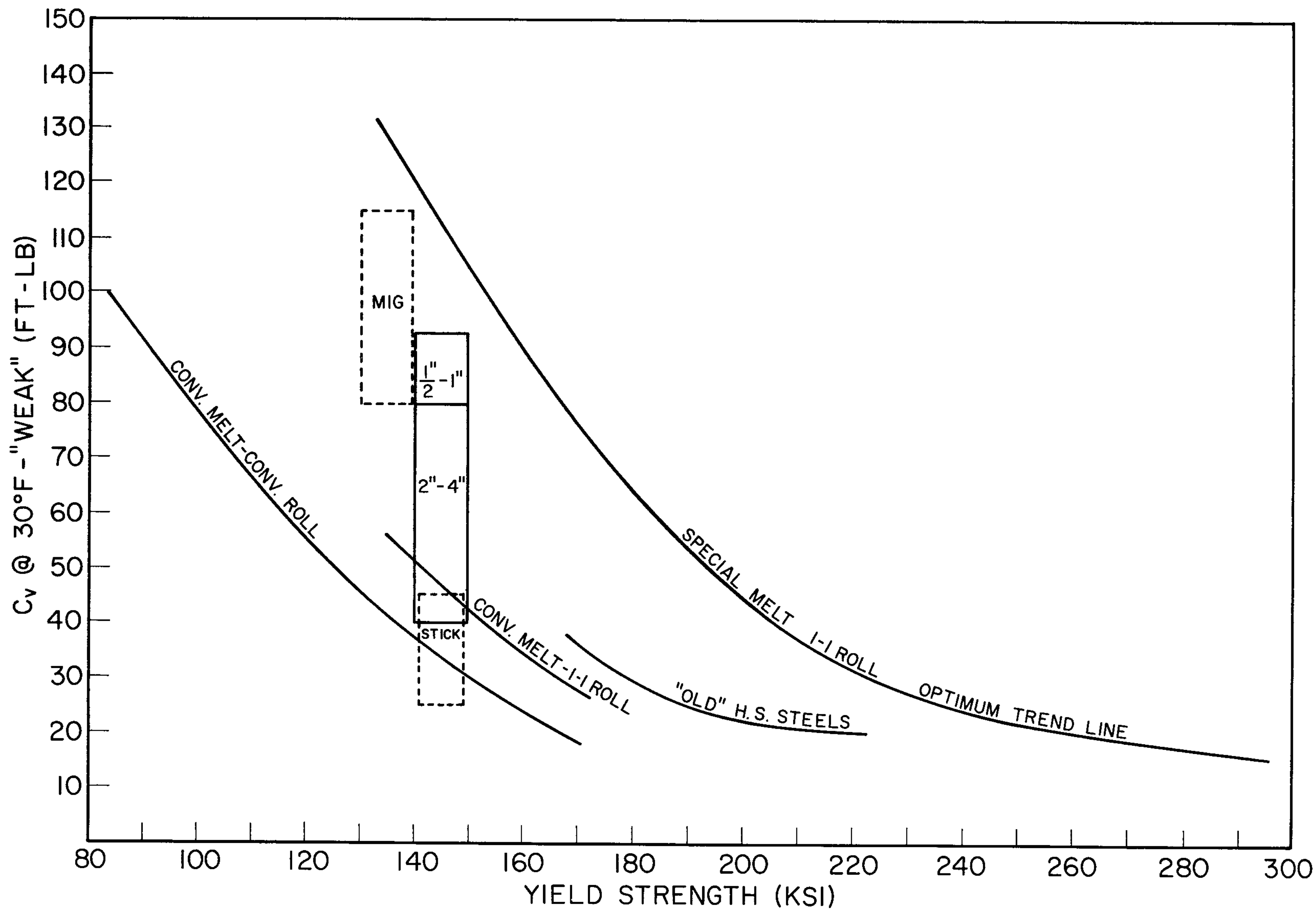



Fig. 19 - General summary of present Charpy V and yield strength relationships for production heats of 5Ni-Cr-Mo-V steel and for experimental welds

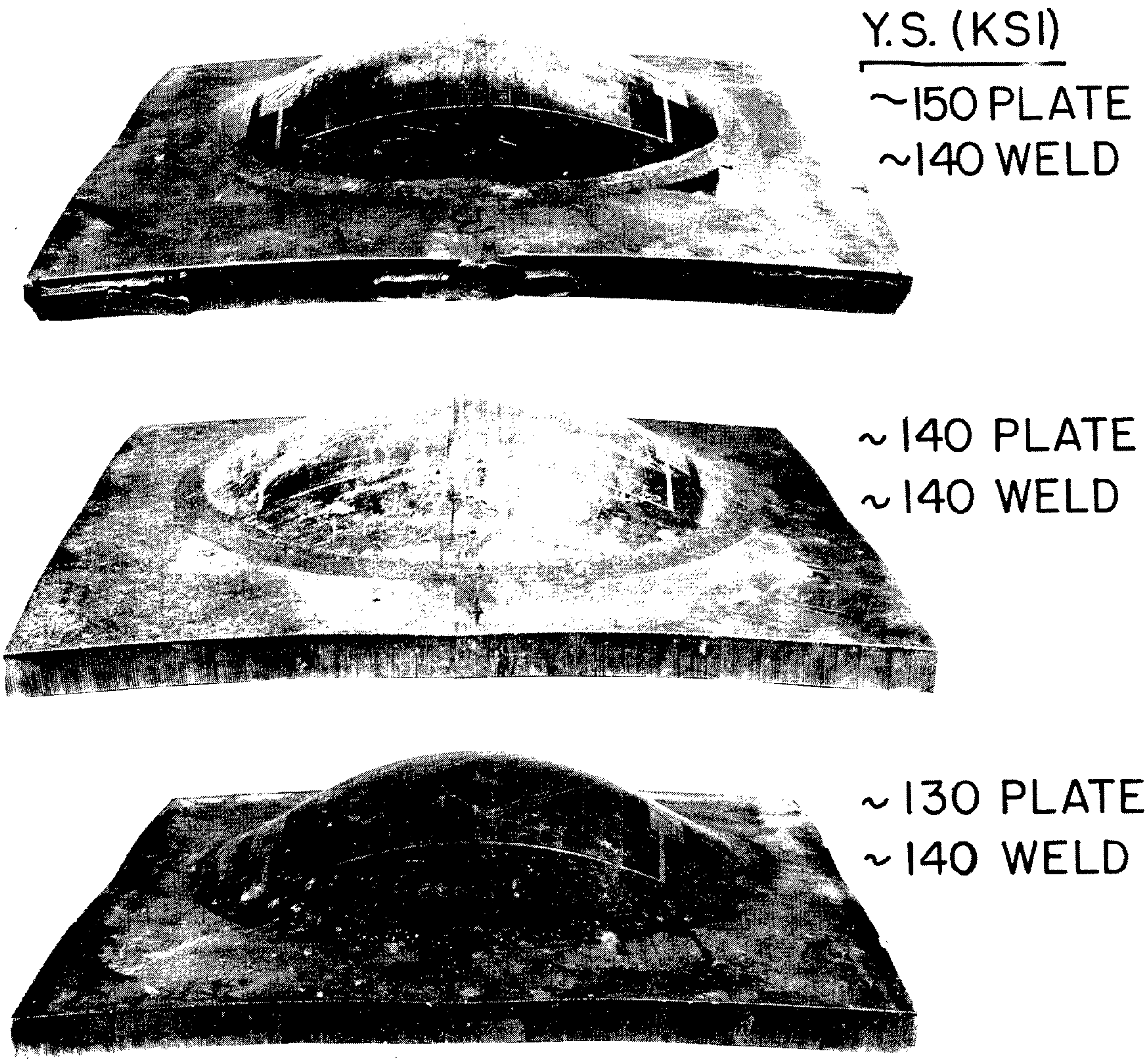



Fig. 20 - Explosion bulge test plates, after four shots, of 140 ksi steel with undermatching, matching, and overmatching welds

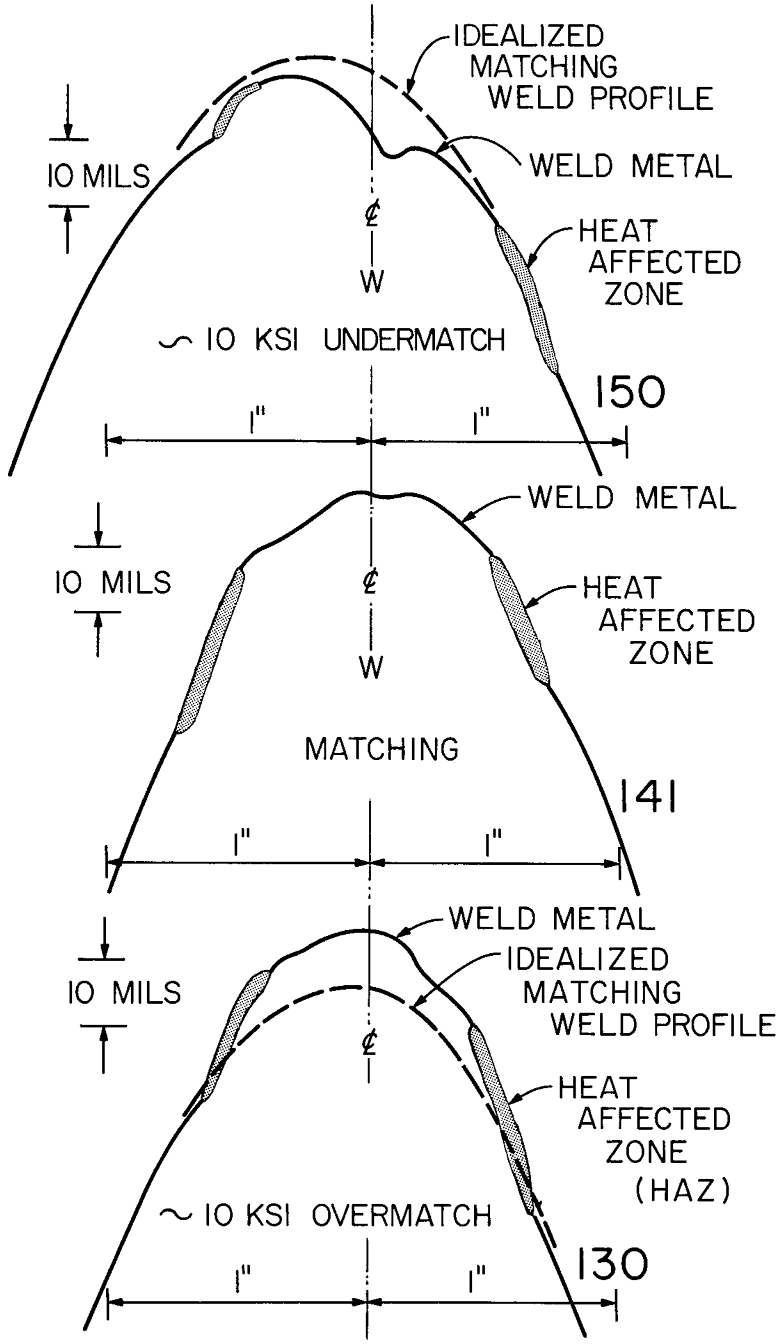



Fig. 21 - Weld profile diagrams of bulge plates shown in Fig. 20; differences produced by overmatching or undermatching the welds are indicated by the dashed lines which indicate expected contour in the absence of the weld.

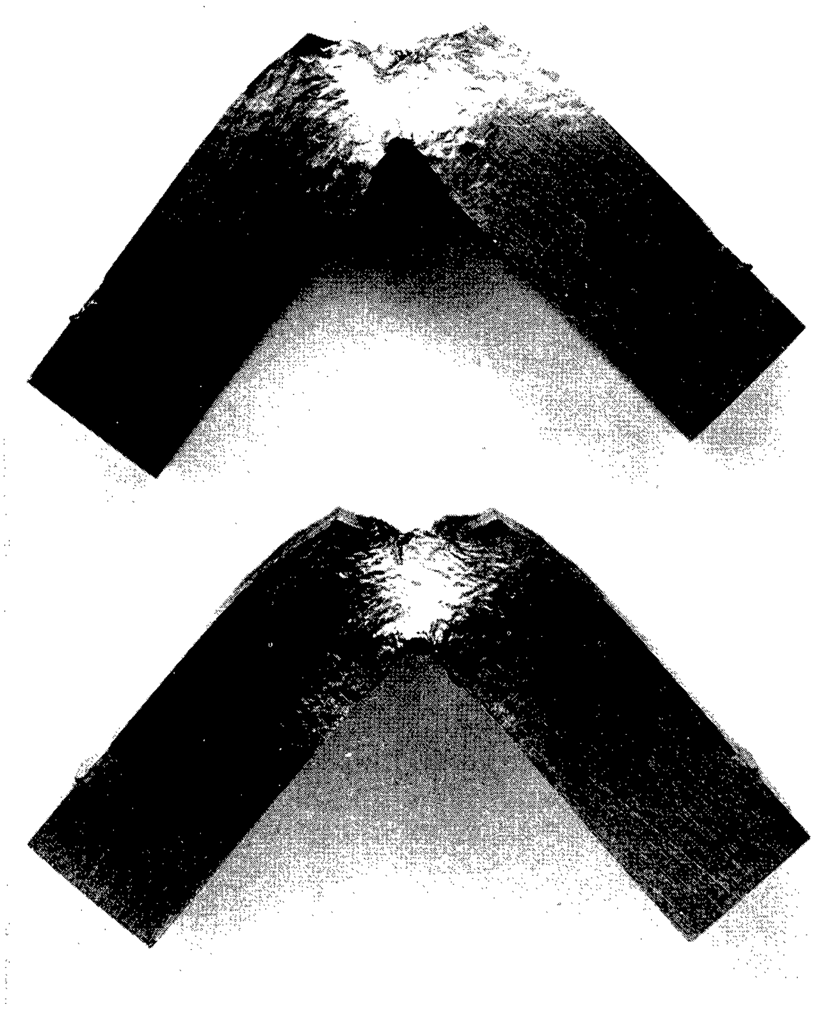


Fig. 22 - Charpy V-notch specimens of Ni-2Be alloy in annealed condition tested at -320°F (258+ ft-lb)

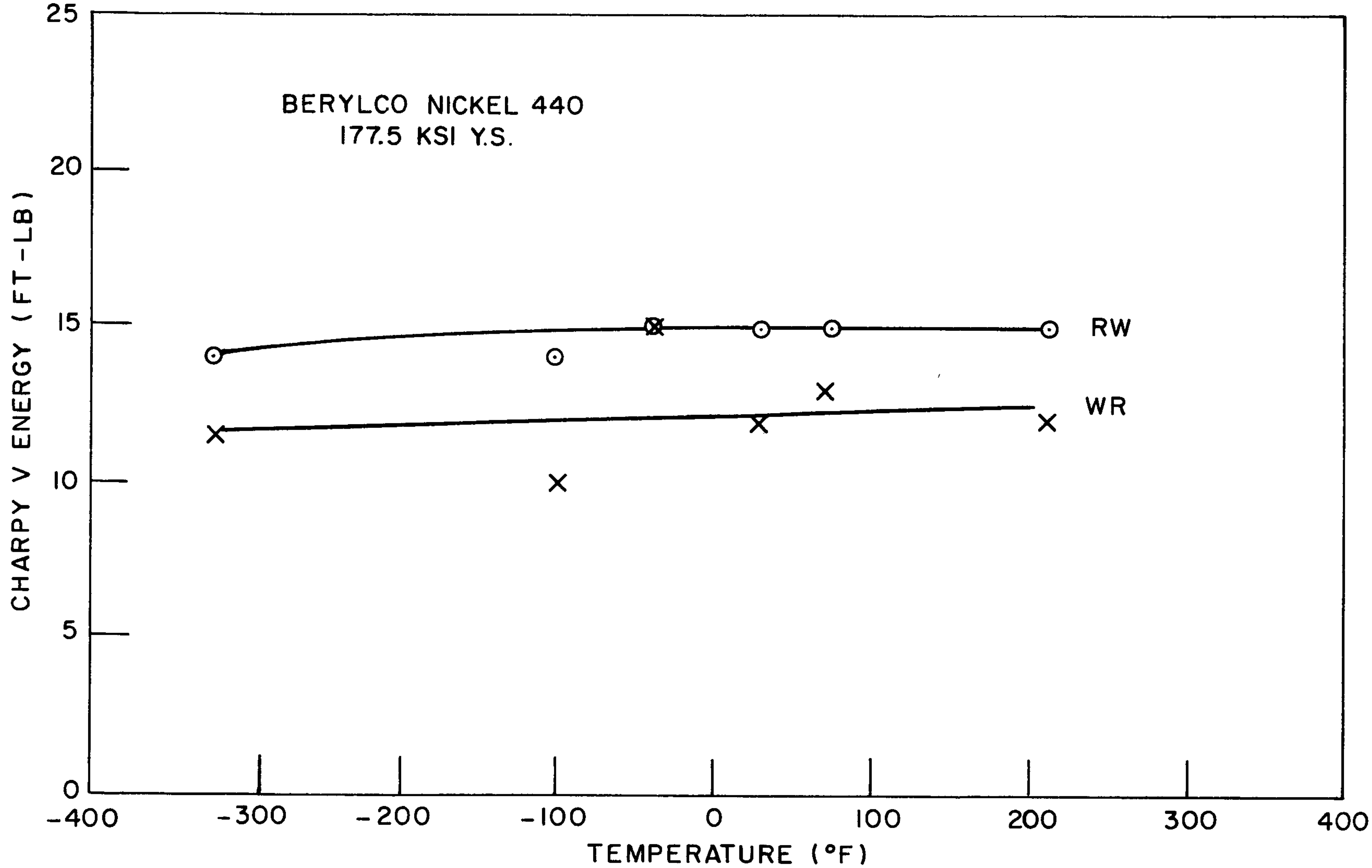



Fig. 23 - Charpy V-notch properties of Ni-Be alloy in full hard condition


Fig. 24 - Electron fractograph of Ni-2Be in annealed condition. Note large dimple size and stretching marks. 3000X. (Reduced approximately 21% in printing)

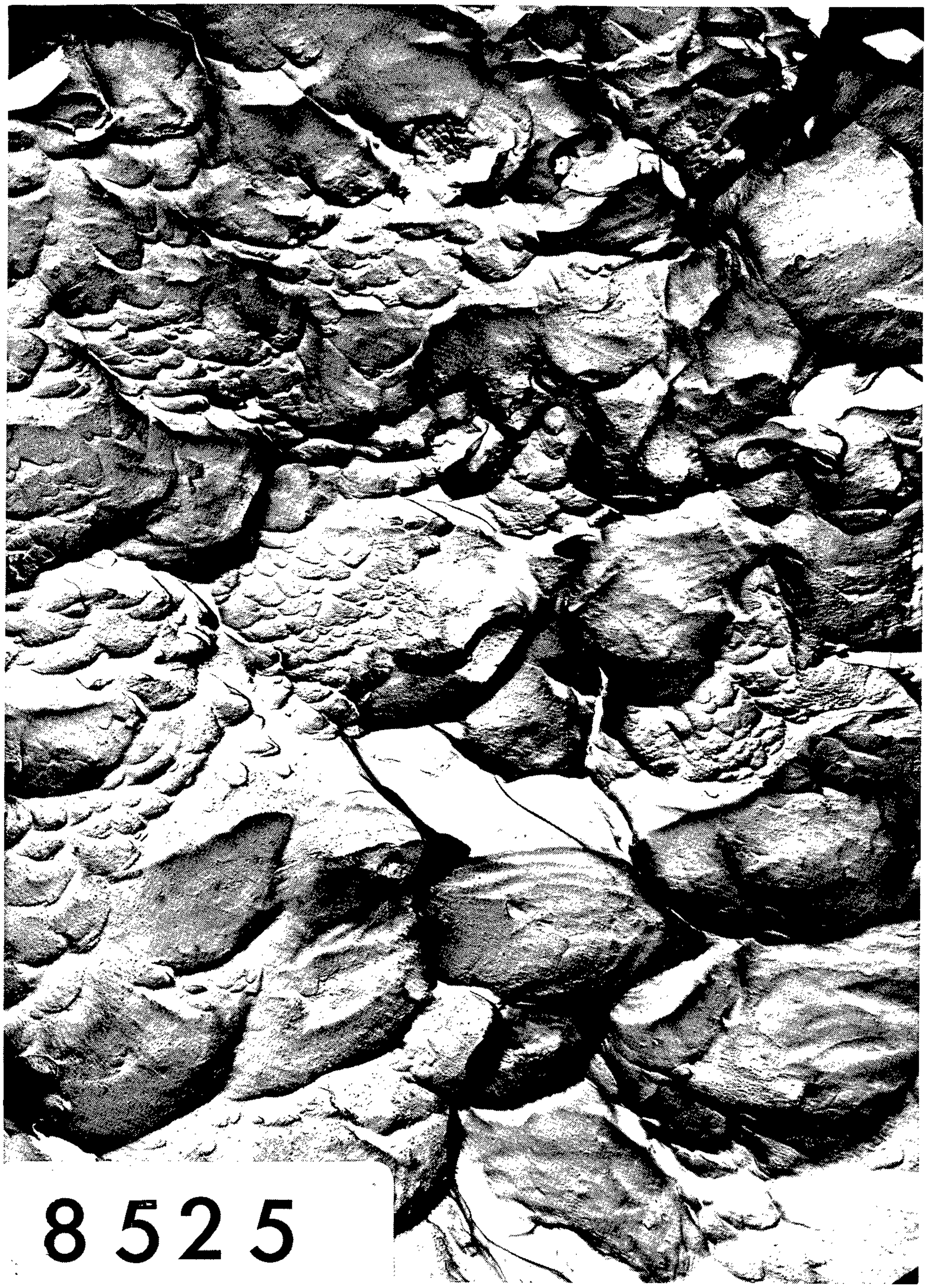



Fig. 25 - Electron fractograph of Ni-2Be in full hard condition. Note small dimple size and lack of evidence of gross plastic deformation. 3000X. (Reduced approximately 20% in printing)

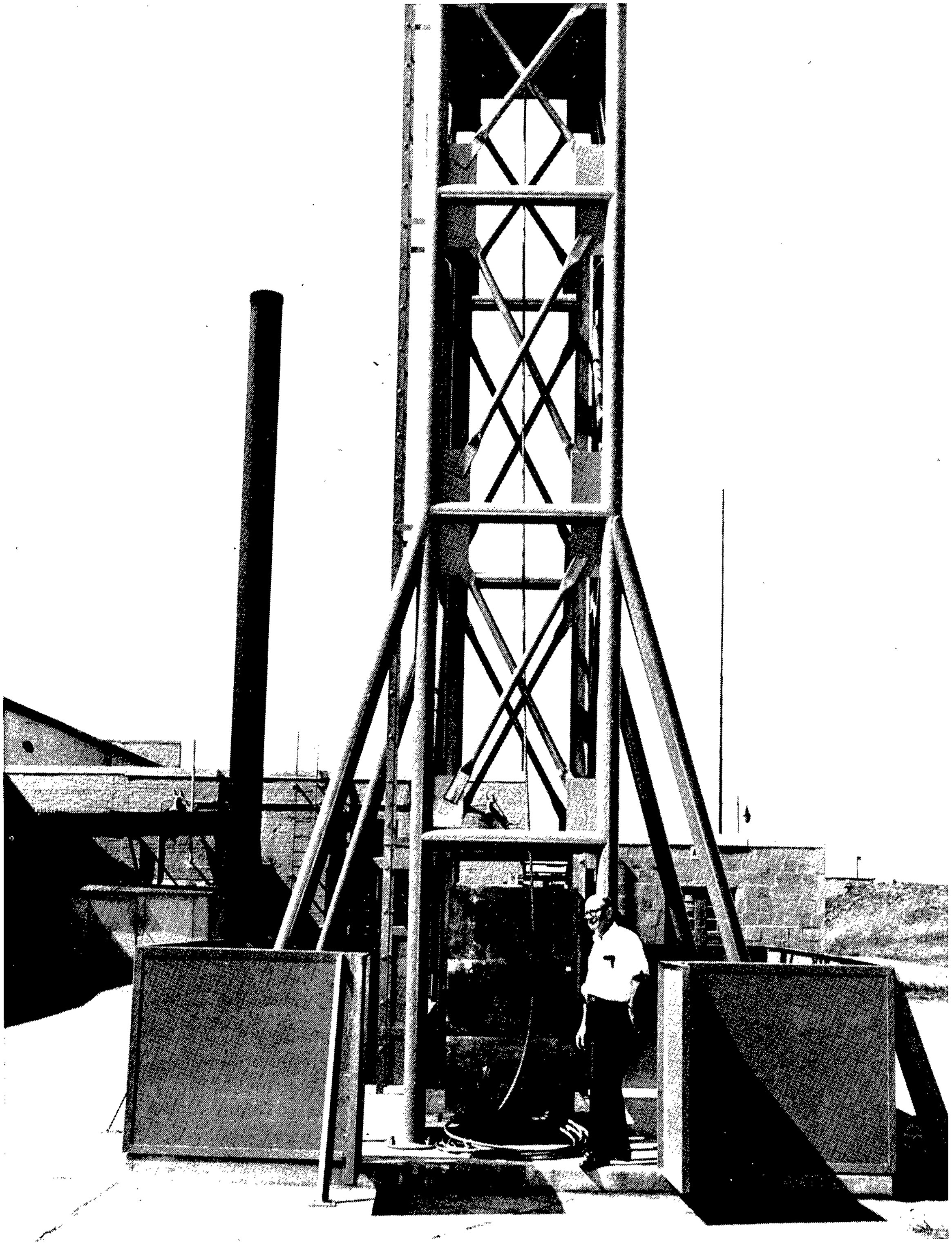


Fig. 26 - NRL 240,000 ft-lb capacity drop-weight test facility

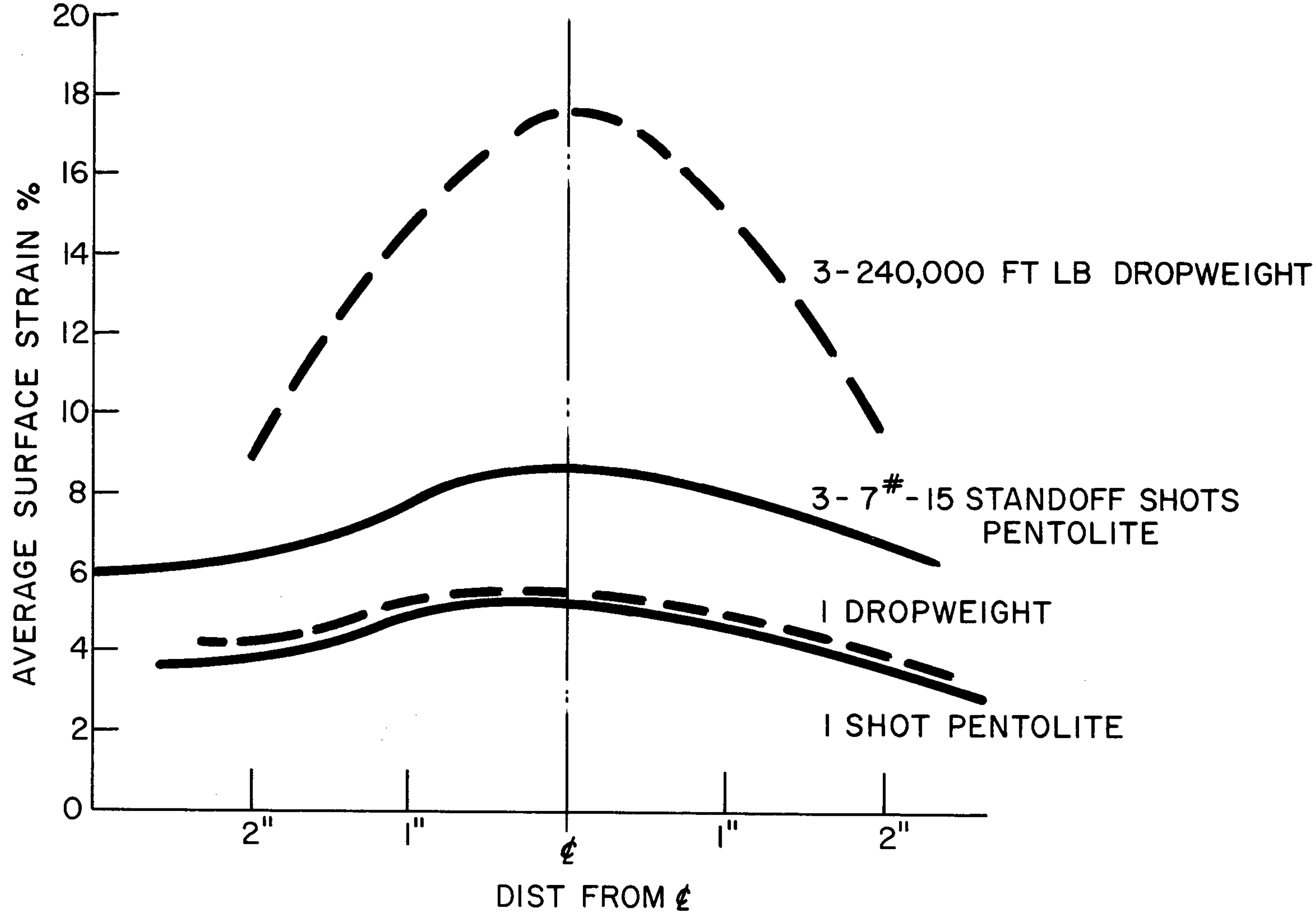



Fig. 27 - Comparison of surface strains produced by 7-lb explosive shots and tests in the 240,000 ft-lb machine

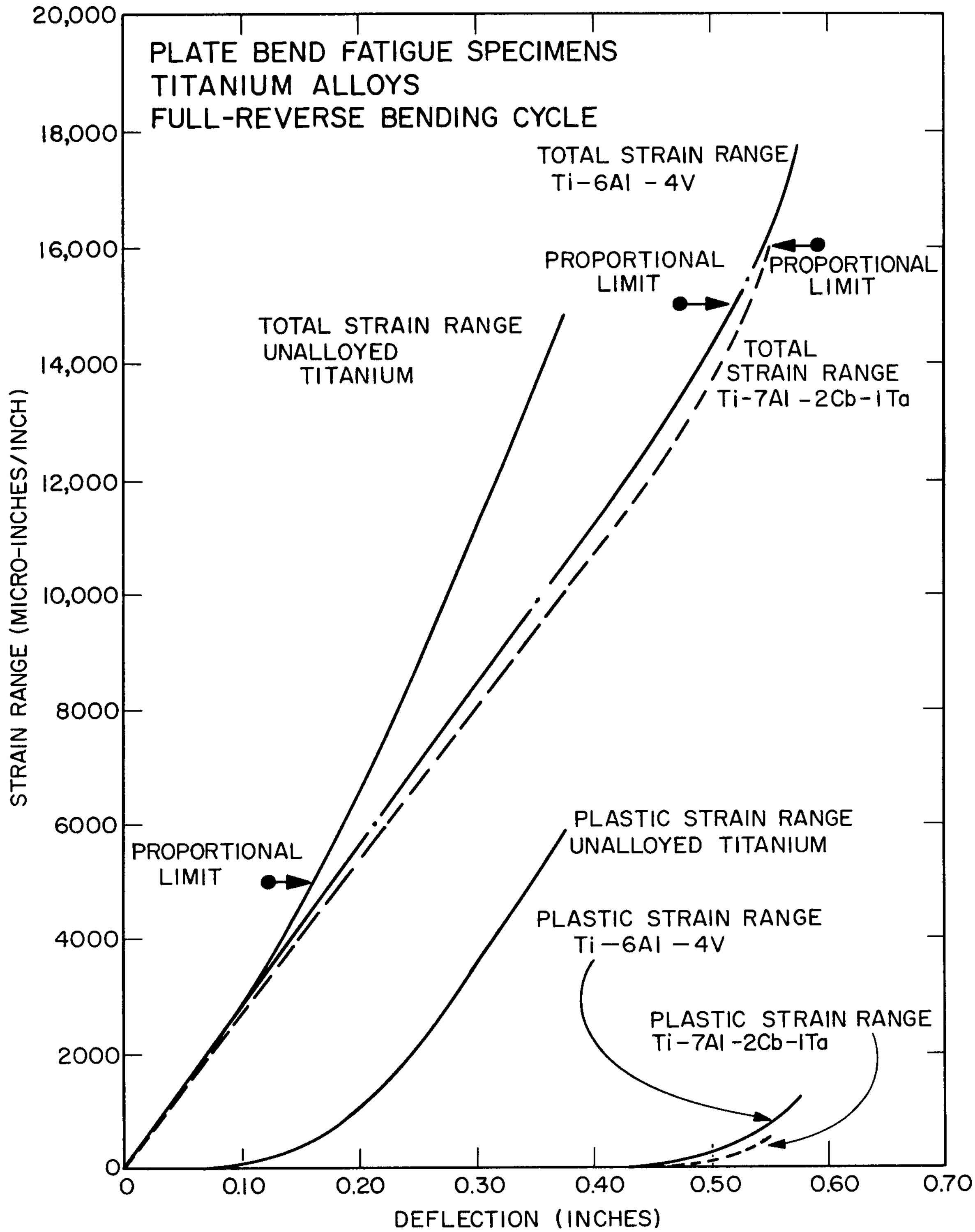



Fig. 28 - Strain range-deflection characteristics of unalloyed titanium, Ti-6Al-4V and Ti-7Al-2Cb-1Ta alloy plate bend fatigue specimens in full-reverse loading cycle

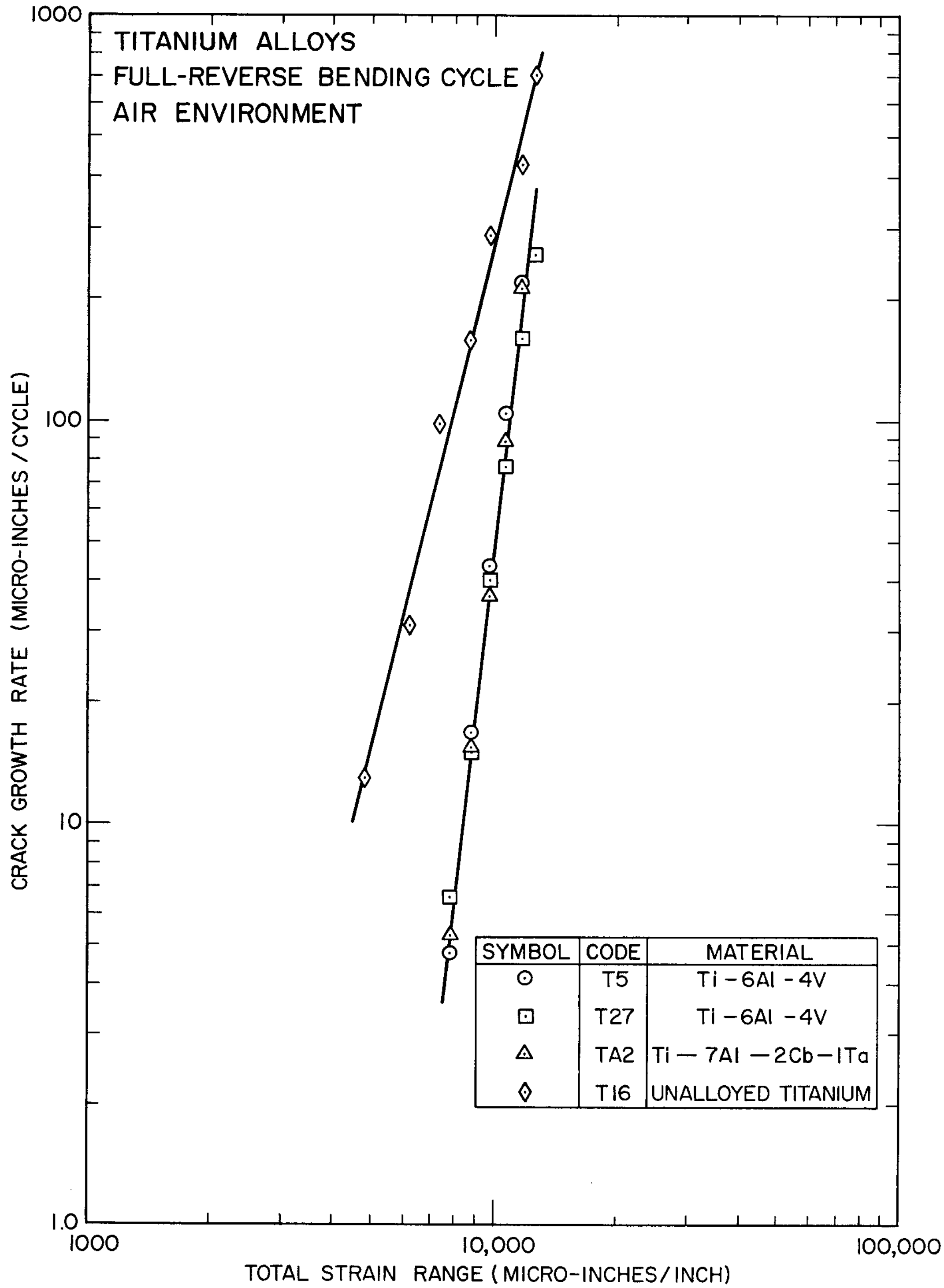



Fig. 29 - Log-log plot of fatigue crack growth rate vs. applied total strain range data in full-reverse bending for unalloyed titanium, Ti-6Al-4V and Ti-7Al-2Cb-1Ta alloys in an air environment.

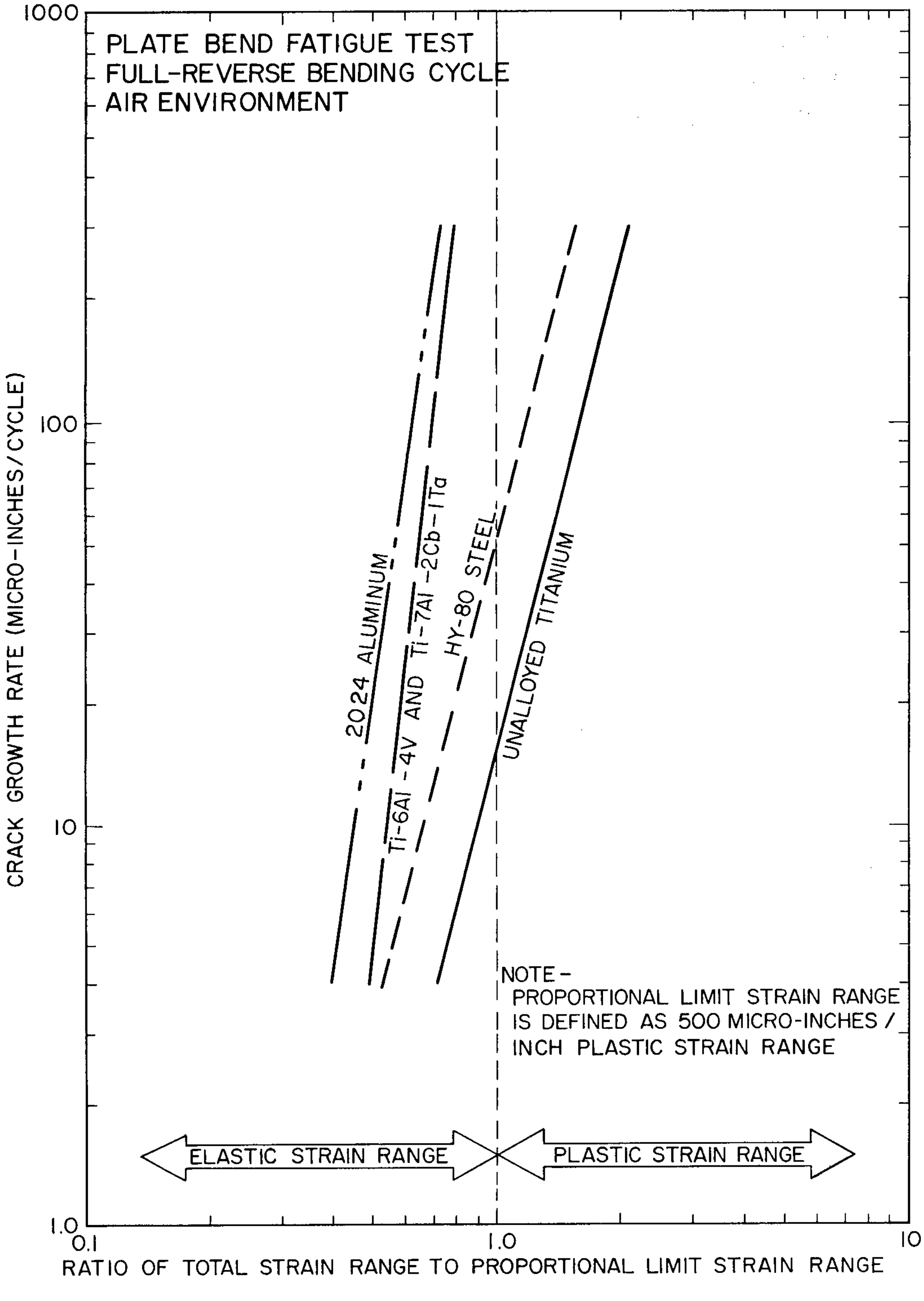



Fig. 30 - Relationship between fatigue crack growth rate vs. ratio of total strain range to proportional limit strain range for three titanium alloys. The 2024 aluminum and HY-80 steel relationships are shown for comparison.

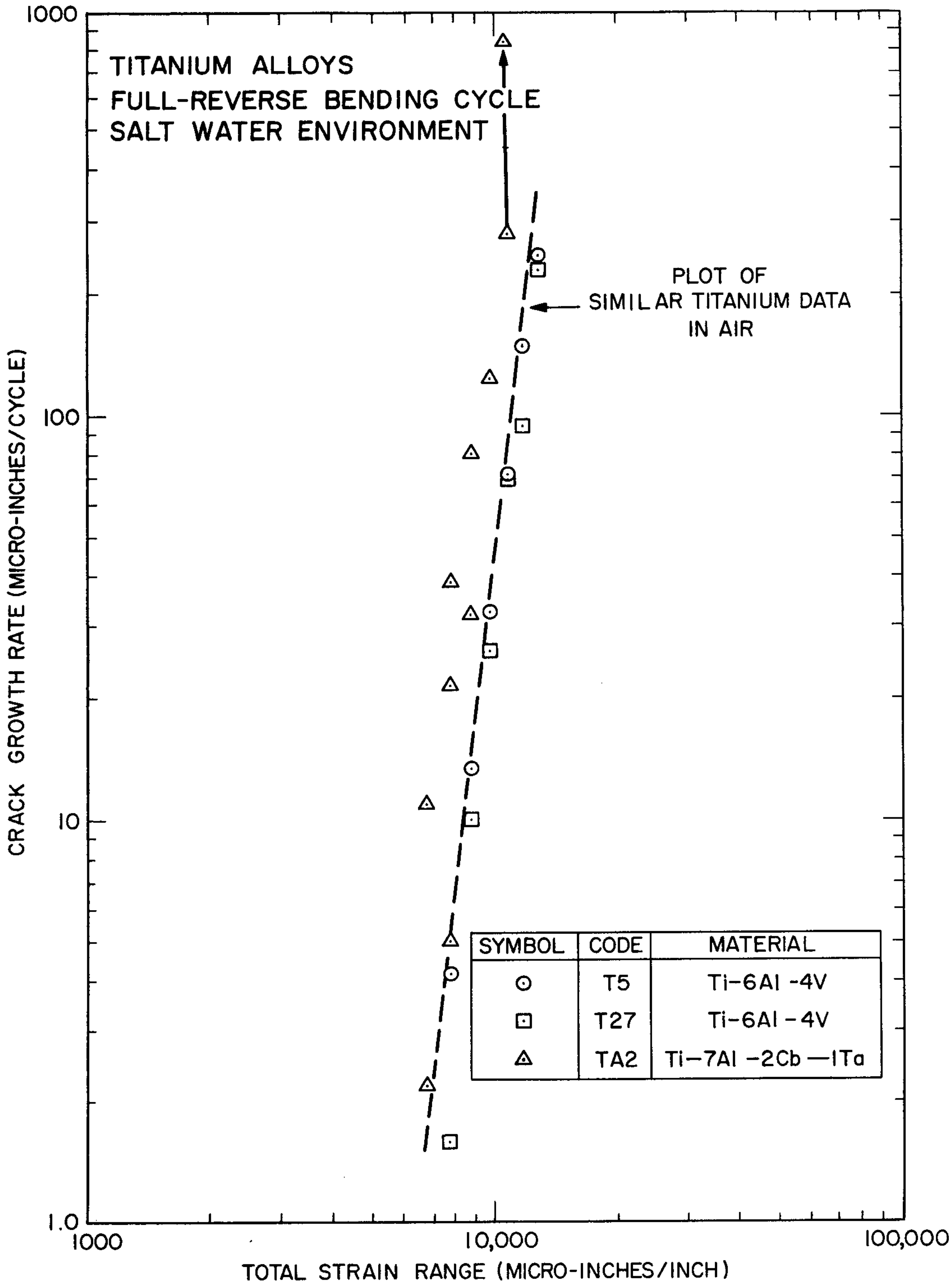



Fig. 31 - Log-log plot of fatigue crack growth rate vs. applied total strain range data in full-reverse bending for Ti-6Al-4V and Ti-7Al-2Cb-1Ta alloys in a 3.5% salt water environment

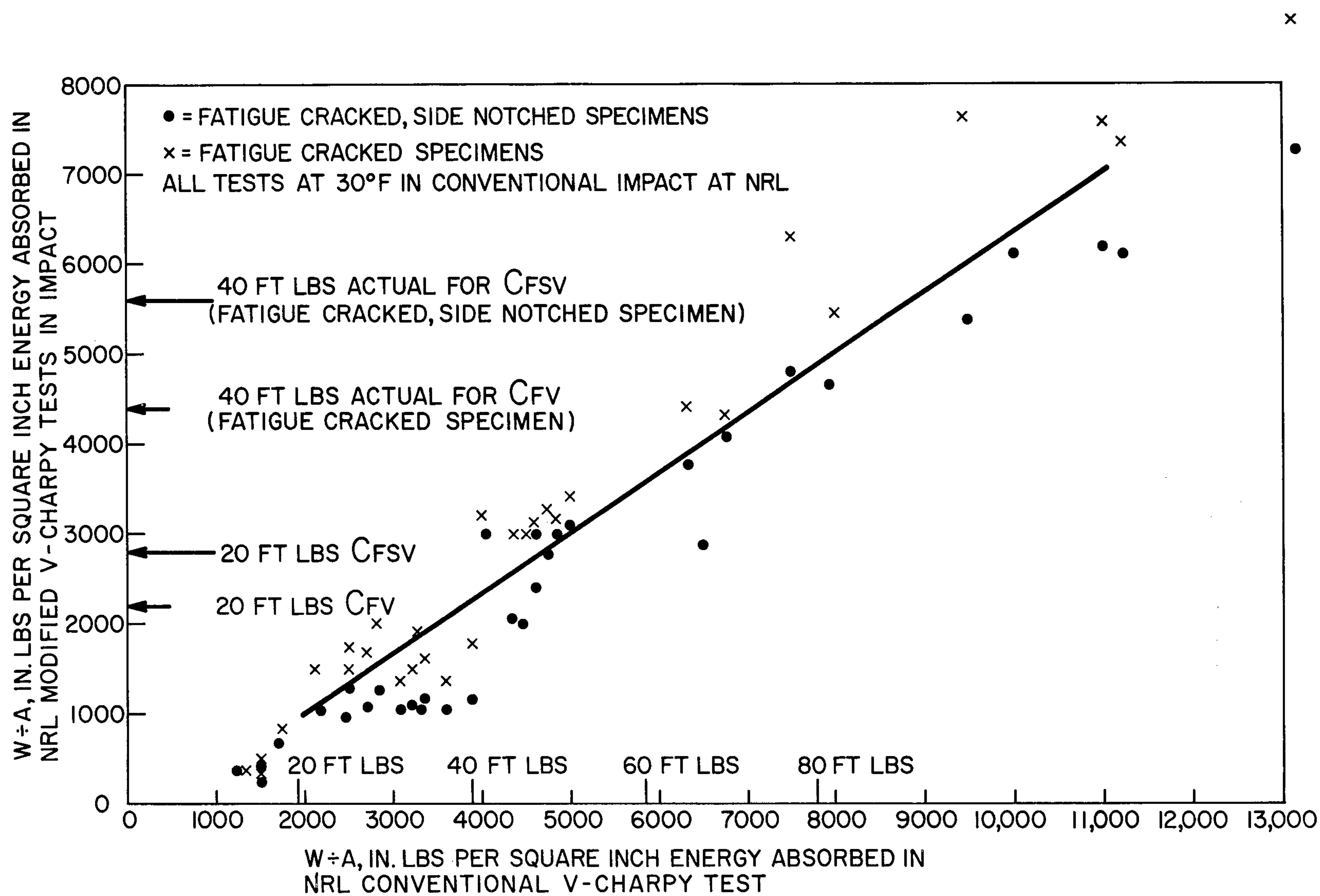



Fig. 32 - Correlation of conventional Charpy V data with data from fatigue crack modified Charpy test specimens, impact tested

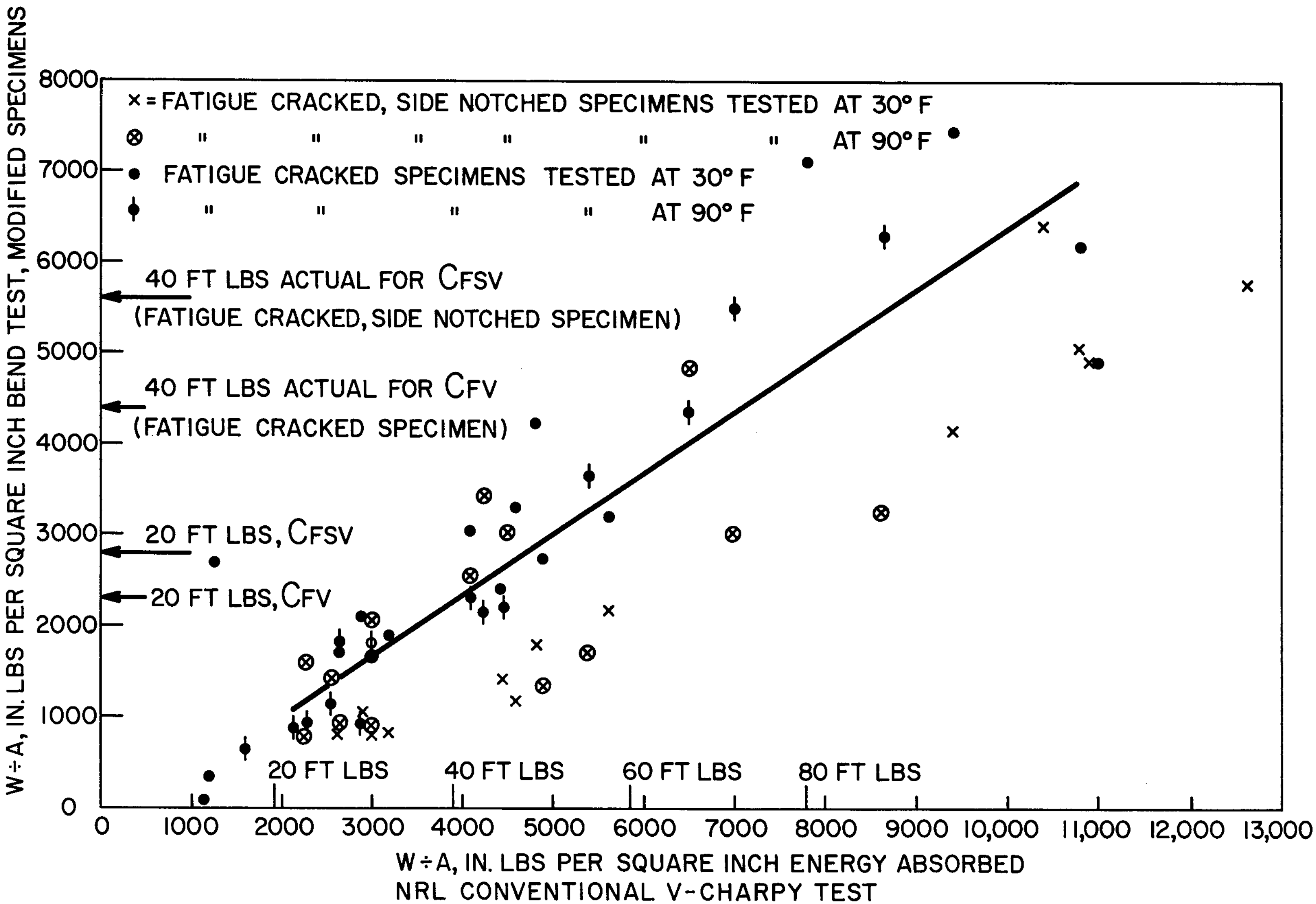



Fig. 33 - Correlation of conventional Charpy V and fatigue crack modified Charpy V test data from slow bend tests

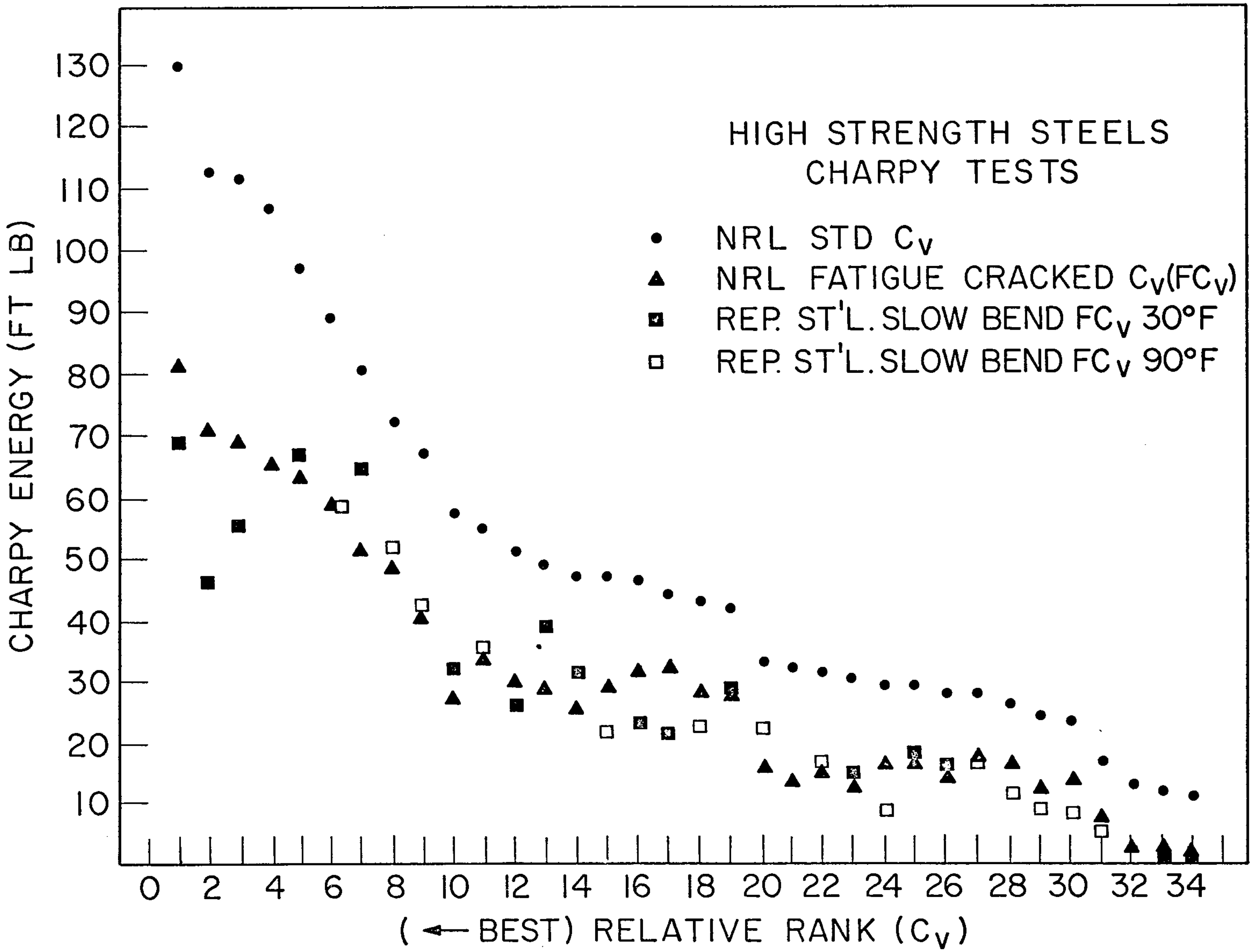



Fig. 34 - Results of conventiona. fatigue crack and modified Charpy tests of high strength steels ranked in decreasing order of toughness

Security Classification

## DOCUMENT CONTROL DATA - R&D

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13. ABSTRACT A progress report covering research studies in high strength hull structural materials, conducted in the period September 1964 to December 1964, is presented. Included in the report are fracture toughness studies on: (1) titanium alloys as related to processing variables (including diffusion-bonding), thickness, and heat-treatment, (2) aluminum, (3) high strength steels relating primarily to the 5Ni-Cr-Mo-V steels, and (4) a nickel-beryllium alloy. Low cycle fatigue crack propagation studies on several titanium alloys are presented as well as a general review of the status of the titanium casting industry. The U.S. Naval Research Laboratory's new 240,000 ft-lb laboratory drop-weight bulge testing facility is described and a preliminary correlation of the results obtained with it and the explosion bulge test are presented. The results of a study using fatigue cracked and side notched Charpy specimens in impact and slow bend testing on a wide variety of steels is given. The results are compared to those obtained in the standard Charpy test using the same materials. A background statement describing the "philosophy" of testing and evaluation of high strength structural materials which is the basis of the program is included in this report.

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Security Classification

| 14. KEY WORDS | LINK A ROLE | LINK A WT | LINK B ROLE | LINK B WT | LINK C ROLE | LINK C WT |
|---|---|---|---|---|---|---|
| Engineering Principles | | | | | | |
| Titanium Alloys | | | | | | |
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| High Strength Steels | | | | | | |
| Fracture Toughness | | | | | | |
| Nickel-Beryllium Alloy | | | | | | |
| Drop-Weight Bulge Test | | | | | | |
| Low Cycle Fatigue | | | | | | |
| Corrosion Fatigue | | | | | | |
| Titanium Castings | | | | | | |
| Fatigue Cracked Charpy | | | | | | |
| Side Notched Charpy | | | | | | |
| Mechanical Properties | | | | | | |
| Diffusion Bonding | | | | | | |
| Explosion Tear Test | | | | | | |
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