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Fracture-Safe Design of Aluminum and Titanium Alloy Structures

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Strength of Metals Branch
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ABSTRACT

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PROBLEM STATUS

This is a special summary and interpretive report covering the results of investigations aimed at the general problem of fracture-safe design. These studies are continuing.

AUTHORIZATION

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FRACTURE-SAFE DESIGN OF ALUMINUM AND TITANIUM ALLOY STRUCTURES

SYMBOLS

DT	Dynamic Tear
K	Stress-intensity factor
K_{Ic}	Critical value of K for plane strain crack extension at slow loading rates
RAD	Ratio Analysis Diagram
Ratio	Refers to lines of constant K_{Ic}/σ_{ys} on the RAD
σ_{ys}	Yield strength of material
R	Resistance to fracture extension
C_v	Standard Charpy-V notch test
B	Thickness dimension of specimen or plate
EB	Electron beam
ETT	Explosion Tear Test
TL	Technological Limit
GFM	Graphical Fracture Mechanics
α	Alpha, low-temperature, close-packed-hexagonal phase of titanium
β	Beta, high-temperature, body-centered-cubic phase of titanium

INTRODUCTION

High-strength aluminum and titanium alloys are being considered for structural applications because of their high strength-to-density ratio. Other relevant factors include the good oxidation, corrosion, and nonmegnetic characteristics for specific alloys. However, the fracture resistance characteristics of these materials is an important factor in determining their fracture-safe reliability. Meaningful fracture resistance information for the full-strength spectrum of titanium and aluminum alloys has been developed only in the last few years. This fracture resistance characterization information is summarized for these materials commercially produced as large plates, and the procedures are presented that have evolved for translating this information into meaningful predictions of flaw size-stress level conditions for fracture. These procedures are similar to those evolved for control of fracture in structures of high- and ultrahigh-strength steels (1).

The general metallurgical characteristics of high-strength aluminum and titanium alloys are provided in Appendixes A and B, respectively, and Appendixes C and D list the various alloys and range of properties studied.

FRACTURE MODE CHARACTERISTICS

Titanium and aluminum alloys are considered "strength transition" materials. High-strength titanium alloys exhibit only a gradual rise in fracture resistance with increasing temperature; for high-strength aluminum alloys, increasing temperature has essentially no effect on fracture resistance. This behavior is in contrast to the abrupt increase in fracture resistance for the low- and intermediate-strength, temperature transition steels, as shown in Fig. 1. The microfracturing process for all fast-fracture conditions in titanium and aluminum alloys involves microvoid coalescence of the metal, even where brittle; i.e., intermetallic or precipitate particles fracture in advance of a moving crack forming microcracks in the material. The microcracks enlarge creating voids (dimples) of a size proportional to the matrix ductility (fracture toughness) of the material. This fracture mode is illustrated in Fig. 2.

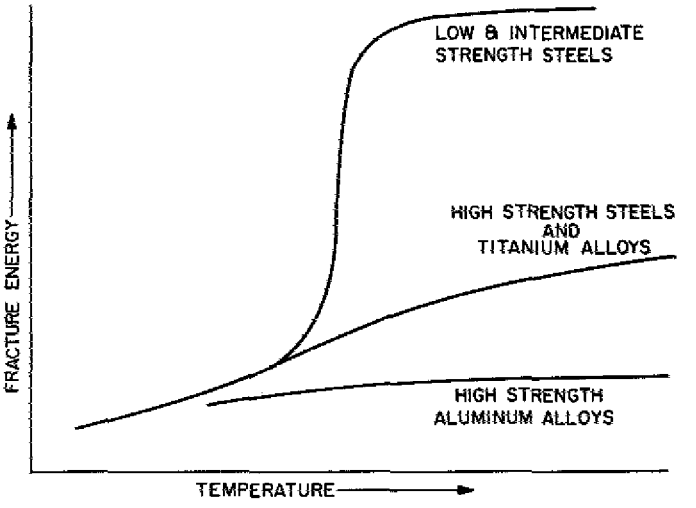

Fig. 1—Schematic illustration of the general effect of temperature on the resistance characteristics of structural metals

The increase in dimple size with increasing fracture toughness relates to increased resistance of the metal to extension of a crack. The fracture extension resistance process can be visualized in terms of a resistance (R) curve, which corresponds to the increase in plastic work absorbed per unit increment of initial fracture extension. No increase in plastic work energy absorption is required for fracture extension in the brittle (very small dimple) case. Initial instability of the crack results in fracture propagation through the total section of material without any requirement for further increase in nominal elastic stress level. Thus, the R curve is flat for brittle titanium and aluminum alloys.

For the ductile fracture case, an increase in plastic work energy is required for initial crack extension. The amount of increase, R-curve slope, is proportional to the intrinsic ductility of the metal.

At this point the reader is referred to Ref. (2) for detailed discussions of the relationships of R-curve slope to crack tip plastic zone size, crack-tip-plasticity conditions, metal ductility range, and section-size effects involved in the characterization of fracture resistance. Propelling force aspects that relate to structural considerations are also covered in the above reference.

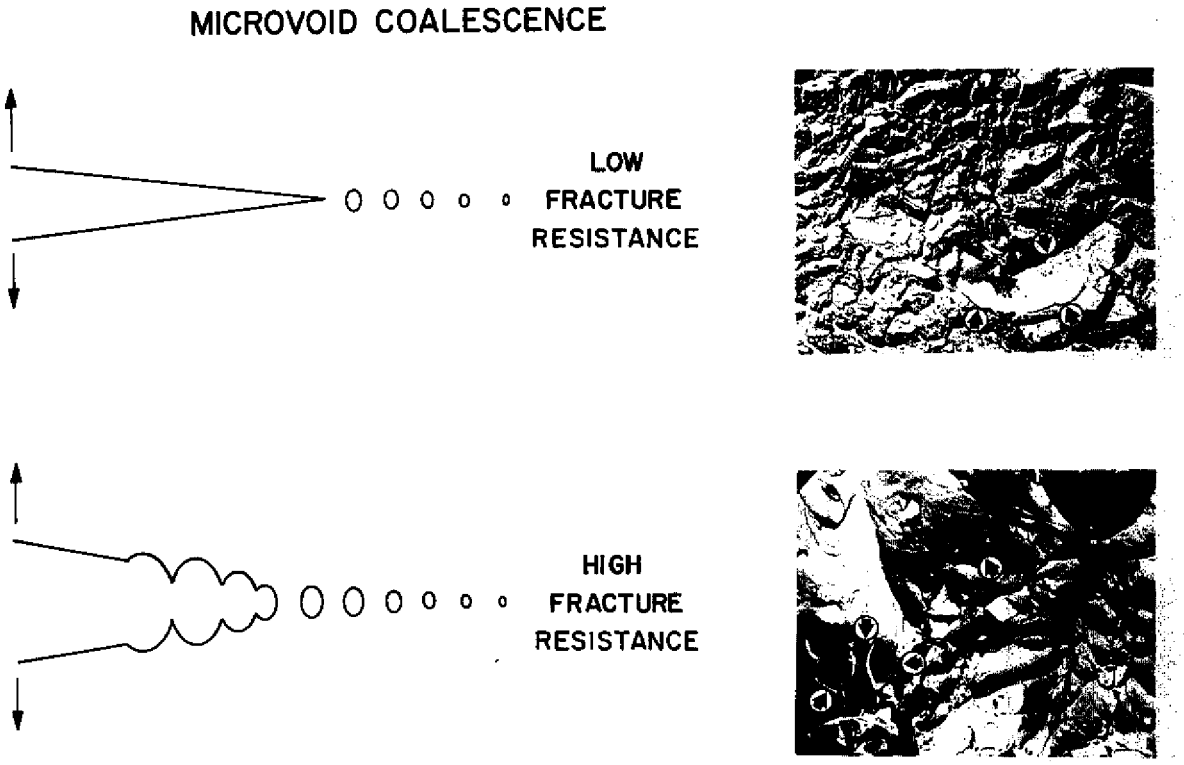

Fig. 2—The effect of increasing fracture resistance on microvoid coalescence processes. The arrows in the microfractographs indicate "brittle" particles which fracture ahead of the advancing crack front causing unit action of the void formation process. The size of the voids (dimples) is an index of the microfracture ductility of the matrix material.

The R-curve slope characteristics for brittle and high plastic fracture resistance for aluminum and titanium alloys are shown in Figs. 3 and 4. These curves were developed by modeling the fracture-extension process using DT test specimens of different crack lengths. The results of the test series for each material show the same characteristics as has been shown for steels (2). A flat R curve is characteristic of brittle alloys and a steeply rising R curve characterizes the highly ductile alloys. Intermediate and low R-curve slopes are obtained for materials of intermediate and low fracture extension resistance.

FRACTURE RESISTANCE TESTS FOR ALUMINUM AND TITANIUM

The principal laboratory tests that have been used to measure the fracture resistance characteristics of aluminum and titanium alloys include the DT, K_{Ic}, C_v (titanium), and notched tensile ratio (aluminum) methods. However, work at NRL during the last 5 years has established that only results of DT and K_{Ic} tests provide reliable definition of levels of fracture toughness in these materials.

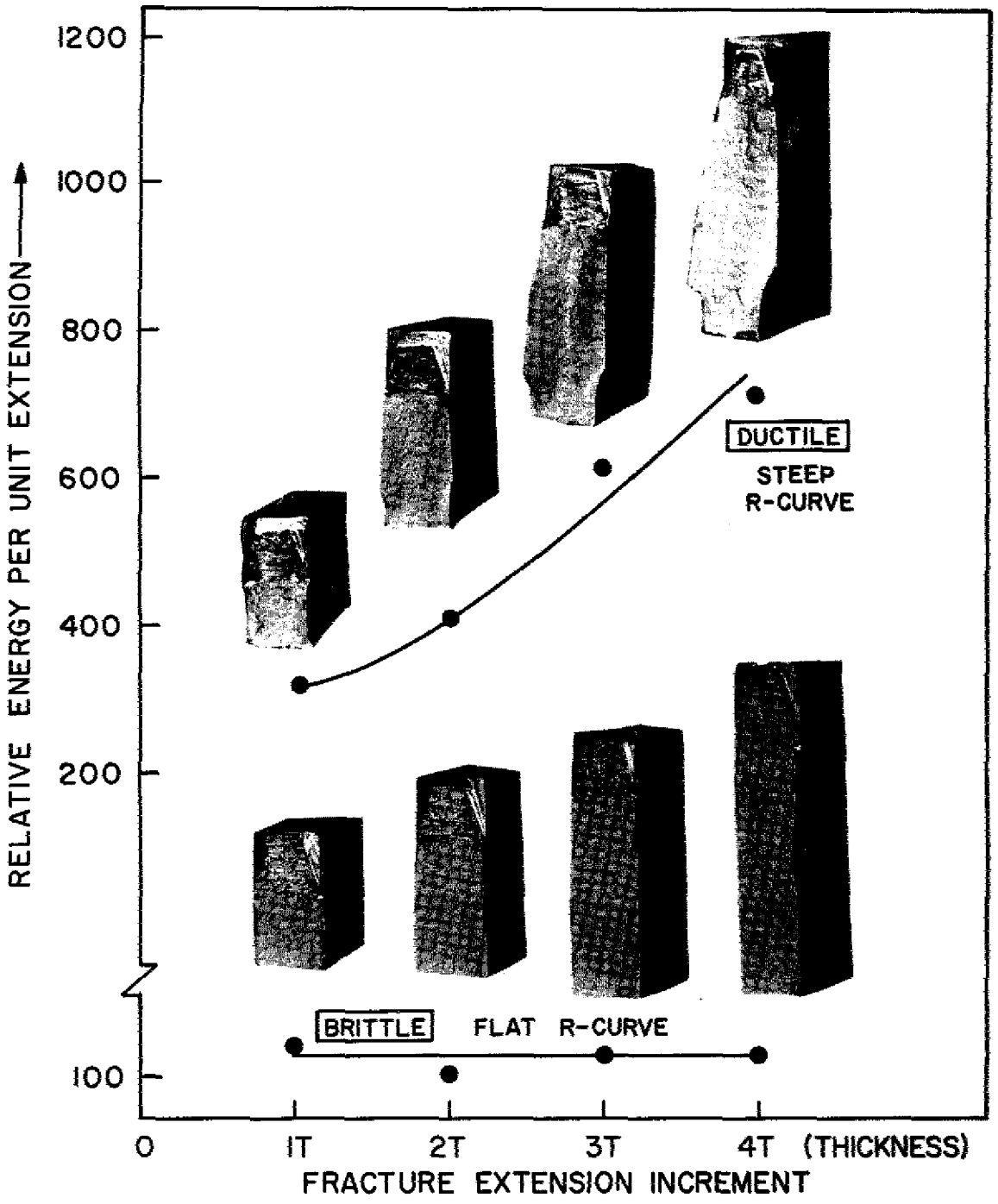

Fig. 3—Representative R-curve slopes for brittle and ductile aluminum alloys of 1-in. thickness. The relative increase in plastic work energy for fracture extension is modeled by Dynamic Tear (DT) test specimens of increasing fracture path lengths. The increased lengths provide for measurement of the effects of decreased triaxial constraint in the process of fracture extension.

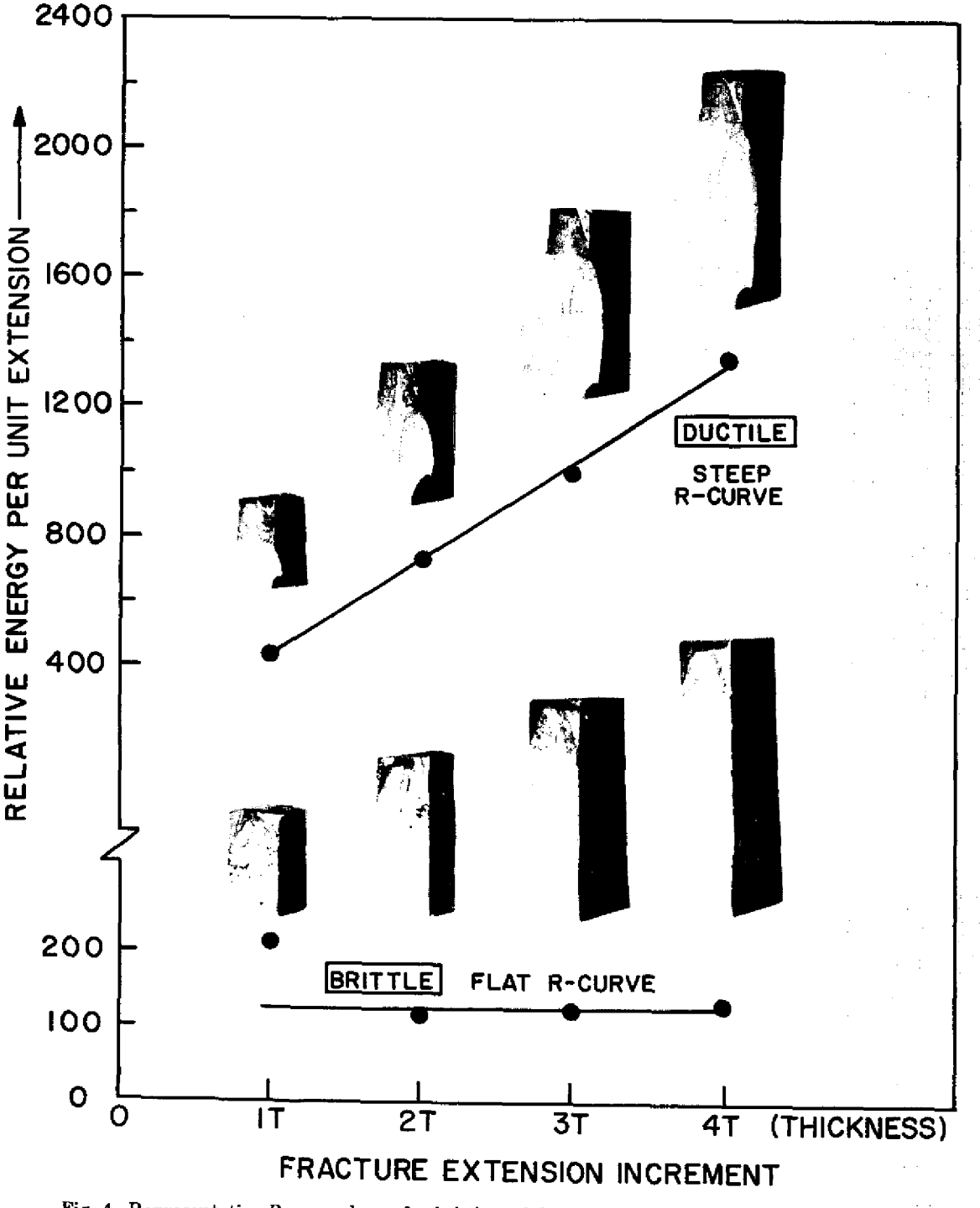

Fig. 4—Representative R-curve slopes for brittle and ductile titanium alloys of 1-in. thickness

Fracture Mechanics Tests

Fracture mechanics tests characterize fracture resistance in terms of the value of the elastic stress field intensity K in the region ahead of the crack-tip plastic zone. The critical stress-intensity level for crack extension (instability) is designated by the parameter K_{Ic}. The subscript I defines the condition of the opening mode for an applied stress normal to the plane of the crack.

The K_{Ic} value must be determined under the condition of plane strain (maximum possible mechanical constraint that can be applied to the metal at the crack tip). This value represents the lowest value of fracture resistance and is considered to be geometry independent. Plane strain is attained in K_{Ic} testing by imposing predetermined requirements on the width, length, thickness, and crack depth of the fracture mechanics specimens to ensure that minimum, i.e., valid, K_{Ic} values are obtained. Of these dimensions, thickness is most significant to valid K_{Ic} determinations because it is the basic dimension from which the other dimensions required for maximum mechanical constraint are determined. For example, an increase in thickness is required for an increase in fracture resistance to simultaneously maintain flaw width and depth dimensions sufficient to maximize constraint conditions at the crack tip. Once a valid plane strain K_{Ic} value has been established, it is then possible to calculate critical flaw size-stress level relationships for various kinds of flaws in the material, based on equations derived by linear-elastic fracture mechanics. The calculations, therefore, presuppose that equivalent conditions of maximum constraint are involved. To ensure that the requirements for plane strain can be met, the thickness of the K_{Ic} specimen must satisfy the conditions

$$B \geqslant 2.5(K_{Ic}/\sigma_{ys})^2,$$

where

B = thickness (in.),

K_{Ic} = critical plane strain stress-intensity factor (ksi$\sqrt{\text{in.}}$), and

σ_{ys} = yield stress (ksi).

It is important to note that the ratio $(K_{Ic}/\sigma_{ys})^2$ is proportional to the crack-tip plastic zone size. According to this ratio, a specific K_{Ic} value related to a metal of increased yield strength results in a decrease in the plastic zone size, i.e., a decrease in fracture resistance. An increase in the K_{Ic} value for a given level of yield strength corresponds to a rapid increase in the plastic zone size and the $(K_{Ic}/\sigma_{ys})^2$ ratio, thus, increased thickness is required for retention of the plane strain condition. Consequently, for materials of high fracture resistance the required specimen size becomes very large. For example, a material with a ratio of 2.0 requires a B value of at least 10 in.

If the thickness is less than the minimum B requirement, the mechanical constraint is then less than adequate for plane strain, even for a very deep crack. Thus, the measurement of fracture resistance must involve procedures that relate to defining fracture extension resistance characteristics.

Values of K are calculated from experimental determinations of crack tip opening displacements obtained by means of a clip gage spanning the crack opening. Load-clip gage

displacement plots indicate a crack-tip opening instability by a drop in the load with increasing clip gage displacement. The nominal stress is calculated from appropriate equations. The critical stress-intensity factor for instability, K_{Ic}, is derived from fracture mechanics equations that relate K to crack depth and nominal stress. Crack instability occurs at a K_{Ic} value which is unique to the material. Different combinations of crack depth and nominal stress can be used to obtain this value, provided plane strain conditions are satisfied.

At the present stage of development, fracture mechanics tests are not practical for routine use. The cost of K_{Ic} testing is high, due to the requirement of close tolerances on specimen dimensions, stringent conditions for producing acceptable fatigue cracks, and other aspects of research laboratory equipment and procedures required in the conduct of the tests.

Dynamic Tear Test

The Dynamic Tear (DT) test was developed to provide a practical laboratory procedure for measuring the fracture resistance associated with the characteristic fracture propagation mode of the metal. A principal requirement for this purpose, which is a feature of the DT test, is the simulation of a deep, sharp crack for initiating the fracture. This feature is obtained for the DT test by introducing an electron beam (EB) weld embrittled with iron (for titanium specimens) or phospher bronze (for aluminum), thereby eliminating the complications of fatigue precracking. Recently, it was established that equivalent results could be obtained with a deep, sharp crack produced by a machined or slot notch sharpened at the tip by a pressed knife edge (3).

The energy required to fracture the specimen serves as an index of the crack-tip plastic zone size associated with the natural process of fracture propagation. The width-to-thickness geometry of the DT test is sufficient to establish conditions for the development of the natural fracture mode related to intrinsic fracture resistance and thickness aspects. For relatively brittle materials, which fracture in the plane strain mode (flat fracture) with small plastic zones, low energy absorption is measured. With increasing resistance to fracture the plastic zone size increases, culminating in the development of a large plastic enclave preceding the fracture process. At high fracture energies, slant fracture (shear) is generally obtained. Mechanical constraint effects due to thickness, which influence the plastic zone size, may be evaluated directly by scaling the size of the DT specimen.

In effect, the DT test defines the fracture mode (plane strain, mixed-mode plane stress, or full plane stress) when conducted for the specific section size of interest. The effects of deviations from the specific size (DT tests of thickness less than plate thickness) may be interpreted from the intrinsic fracture resistance value and fracture mode that is observed. For example, a flat-break fracture which implies plane strain conditions would not be changed by increased thickness. Similarly, a full-slant plane stress fracture, associated with high energy absorption measured for a 1-in. DT test specimen, implies that an increase in thickness of 2 to 3 in. would not be expected to decrease the fracture resistance level significantly for the same metallurgical condition. Thus, the characterization of the thick-section fracture resistance as related to gross strain mechanical conditions would not change. For intermediate fracture resistance levels represented by mixed-mode plane stress fractures,

the effects of changes in section thickness require more detailed assessment for engineering interpretation to specific applications.

Figure 5 is a schematic drawing of the standard 1-in. DT specimen. Impact loading with a calibrated falling or swinging pendulum weight (similar to a large C_v machine) fractures the specimen, and the energy required to fracture a specified area of plate material is then measured.

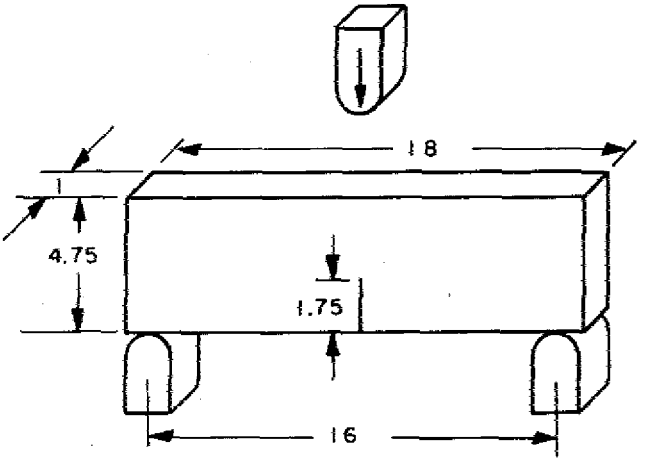

Fig. 5—Standard 1-in. DT specimen featuring a 1.75-in.-deep electron beam brittle weld crack-starter

An example of fracture surface characteristics of DT specimens for extremes in fracture toughness is shown in Fig. 6 for aluminum and titanium alloys. The flat fracture, indicated on the left for each alloy, corresponds to brittle (plane strain) behavior and low DT energy values. Very high levels of fracture resistance correspond to conditions of plane stress for which high DT energy values and slant fracture, indicated on the right for each alloy, are obtained.

C_v and Notched Tensile Tests

The C_v test has been used extensively by industry and research laboratories for measuring the fracture energy of titanium alloys. NRL work during the late 1960's has shown that the C_v test lacks adequate discrimination capabilities between different levels of fracture resistance. This inadequacy is due to the blunt notch and lack of sufficient specimen depth to establish the characteristic fracture mode. These deficiencies are manifested in C_v data that did not reliably indicate the true level of fracture resistance of the material tested. Further, the C_v test often erroneously indicates increasing fracture resistance (as a result of heat treatment, different processing, etc.) compared to the DT, K_{Ic}, and structural prototype tests. The DT test and K_{Ic} test (where appropriate) are now being used to provide the major guidance for such studies.

The aluminum industry has long recognized the inadequate discrimination capability of the C_v test for aluminum alloys. In general the C_v test is relatively insensitive to changes

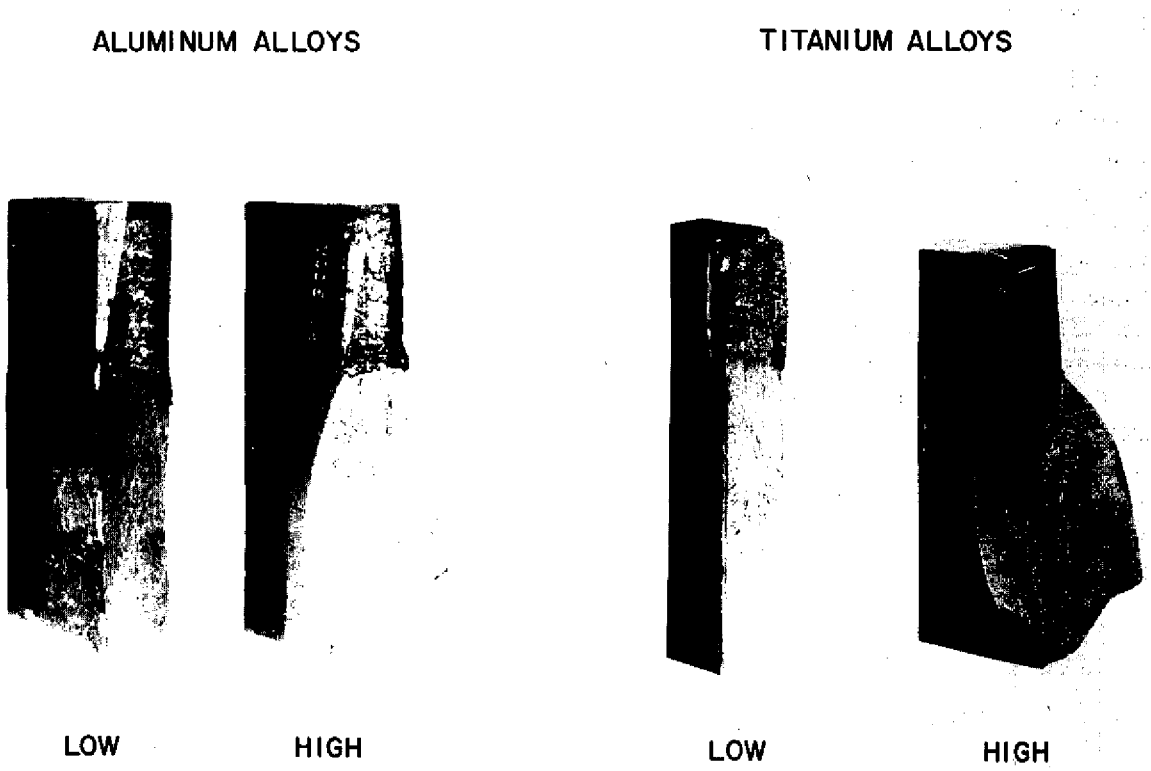

Fig. 6—Features of DT test fracture surface for aluminum and titanium alloys of low and high fracture toughness

in fracture resistance, and the total range of fracture energy for the spectrum of aluminum alloys corresponds to low values. Thus, for many years the principal method for fracture resistance evaluation of aluminum alloys has been based on tests of notched round tensile bars. The parameter of fracture resistance comparison is the ratio of the breaking strength of the notched bar to the tensile yield strength of the material, for a specific notch acuity. Increasing values of the ratios above 1.0 were considered to indicate increasing ability for the material to plastically deform at the notch tip. Although, compared to the C_v test, this test was felt to provide better discrimination capability for changes in fracture resistance, the improvement was not great. Other difficulties, which were also a result of the wide variety of notched tensile specimen geometrics and notch acuities used, related to translation of the various ratio values obtained for a material to a specific level of fracture resistance and to structural performance.

Application of DT test procedures to the fracture toughness characterization problems for the full yield strength range of aluminum alloys is expected to replace the notched tensile test for heavy-section material. K_{Ic} tests provide the accurate, vernier scale, plane strain definition of fracture resistance for only the highest-strength materials. For example, the Metals Property Council currently is conducting a program to develop a handbook of fracture resistance data for guidance to designers. In this program, DT energy values will be used for those materials for which valid K_{Ic} data cannot be obtained.

CORRELATIONS OF DT AND K_{Ic} FRACTURE RESISTANCE TESTS

Early interpretations of DT and K_{Ic} data depended on extensive indexing of the results of the Explosion Tear Test (ETT). The ETT is a prototype structural element test designed to simulate elastic and/or plastic loading of a flawed panel of prime plate material. Explosive loading methods are employed to deform the 22-by-25-by-1-in.-thick test panel into a cylindrical form with the primary stress being normal to a centrally located flaw. A 2-in.-long, through-thickness, embrittled EB weld, similar to the DT crack-starter weld, initiates a propagating crack into elastically or plastically loaded material, depending on fracture resistance. The indexing parameter of the ETT is the amount of surface plastic strain required to cause extensive fracture of the test piece. Thus, the ETT yields an indication of the material's tolerance for large flaws. The typical performance of low, intermediate, and high fracture resistance material is shown in Fig. 7 for titanium alloys. The fracture modes correspond, respectively, to plane strain, mixed-mode plane stress, and full plane strain. Similar performance is displayed by aluminum alloys for the same fracture modes.

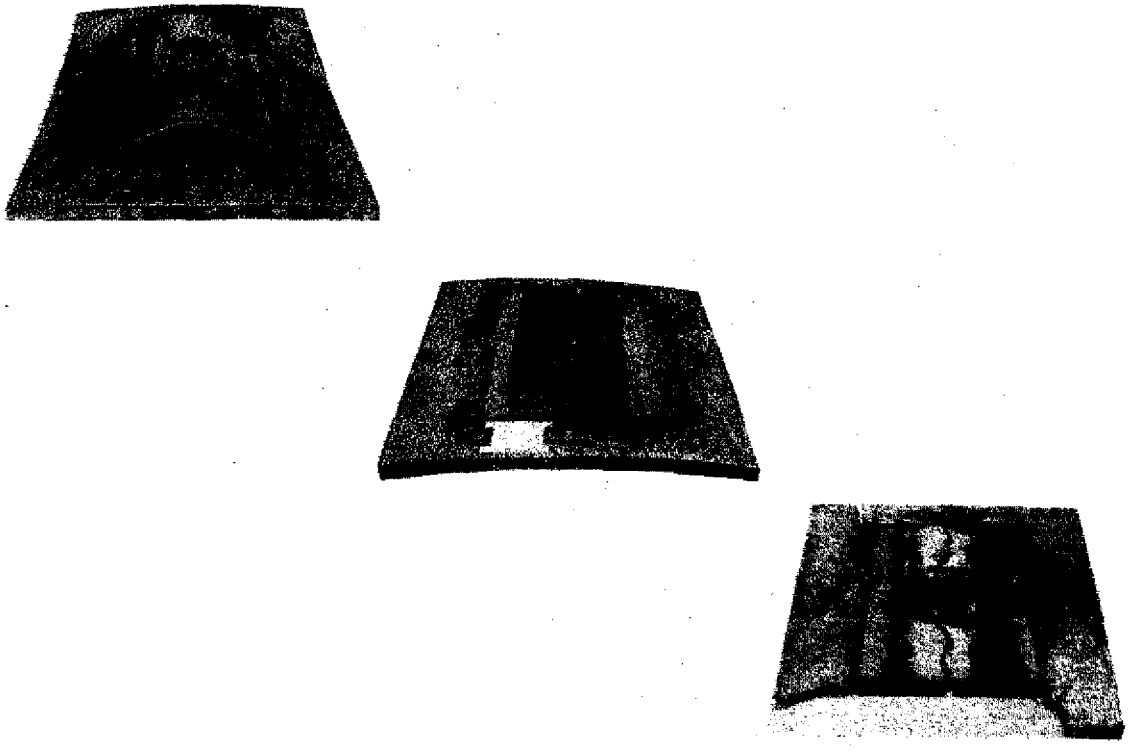

Fig. 7—Series of ETT plates illustrating the structural prototype performance of high, intermediate, and low fracture resistance (left to right) as measured by DT tests

DT-K_{Ic} Correlations

The recently developed DT-K_{Ic} correlations for aluminum and titanium alloys provided for a major advancement in interpreting fracture resistance data for fracture-safe design with these materials. The inexpensively obtained DT data could be easily translated into flaw size-stress level conditions for fracture for the plane strain case. The DT-K_{Ic} correlations

shown in Figs. 8 and 9 for aluminum and titanium alloys, illustrate that DT energy values increase as K_{Ic} values increase. However, for the 1-in.-thick (full plate thickness) specimen tested, the valid K_{Ic} values obtained were limited to about 30 ksi$\sqrt{\text{in.}}$ and 85 ksi$\sqrt{\text{in.}}$ for aluminum and titanium alloys of about 50 ksi σ_{ys} and 130 ksi σ_{ys} respectively, according to current ASTM E399-70T criteria (4).

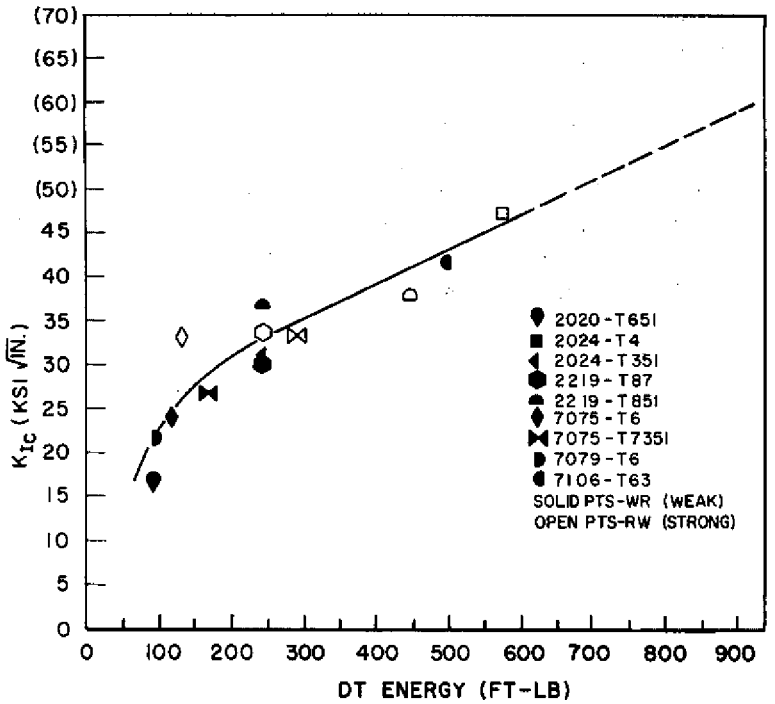

Fig. 8—Relationship of K_{Ic} to DT energy for aluminum alloys

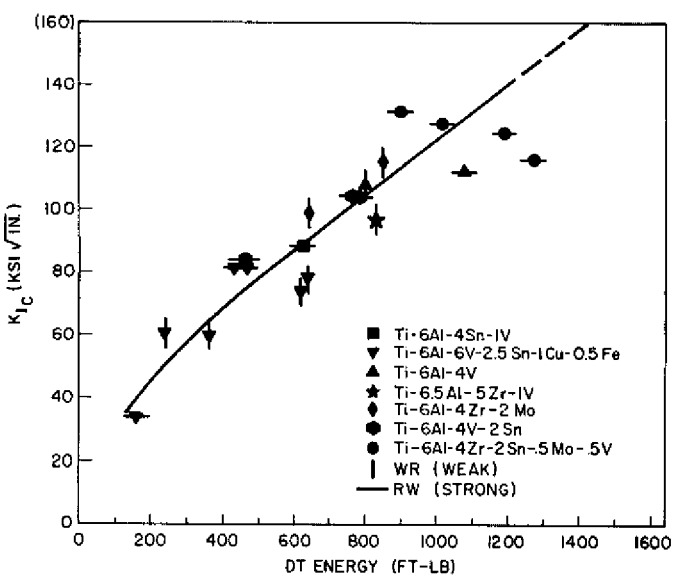

Fig. 9—Relationship of K_{Ic} to DT energy for titanium alloys

RATIO ANALYSIS

The basic data which led to development of the RAD diagrams for 1-in.-thick plate material are shown in Figs. 10 and 11 for aluminum alloys and Figs. 12 and 13 for titanium alloys. The upper limit of highest metallurgical quality material as defined by fracture resistance properties is represented by the Technological Limit (TL) curve. A similarly defined curve relating to the lowest metallurgical material is represented by the lower bound curve. The limit curves are indexed by DT energy values for the ductile fracture region, Figs. 10 and 12, and by K_{Ic} data for the plane strain region, Figs. 11 and 13. The data points represent the latest (1970) DT and K_{Ic} information evolved from metallurgical research studies at NRL and at other research laboratories (K_{Ic} data only). Note that each alloy system contains a relatively narrow yield strength range over which the TL trend line drops sharply from a high level of obtainable fracture resistance to a low level of resistance. This sharp drop, which occurs at 30- to 50-ksi yield strength for aluminum alloys and 110-140 ksi yield strength for titanium alloys, defines the strength transition range for the best quality material. If a series of similar curves were drawn for intermediate levels of metallurgical quality, they would fall between the TL and lower boundary lines. Decreasing metallurgical quality causes a downward shift of the strength transition to a lower range of yield strength.

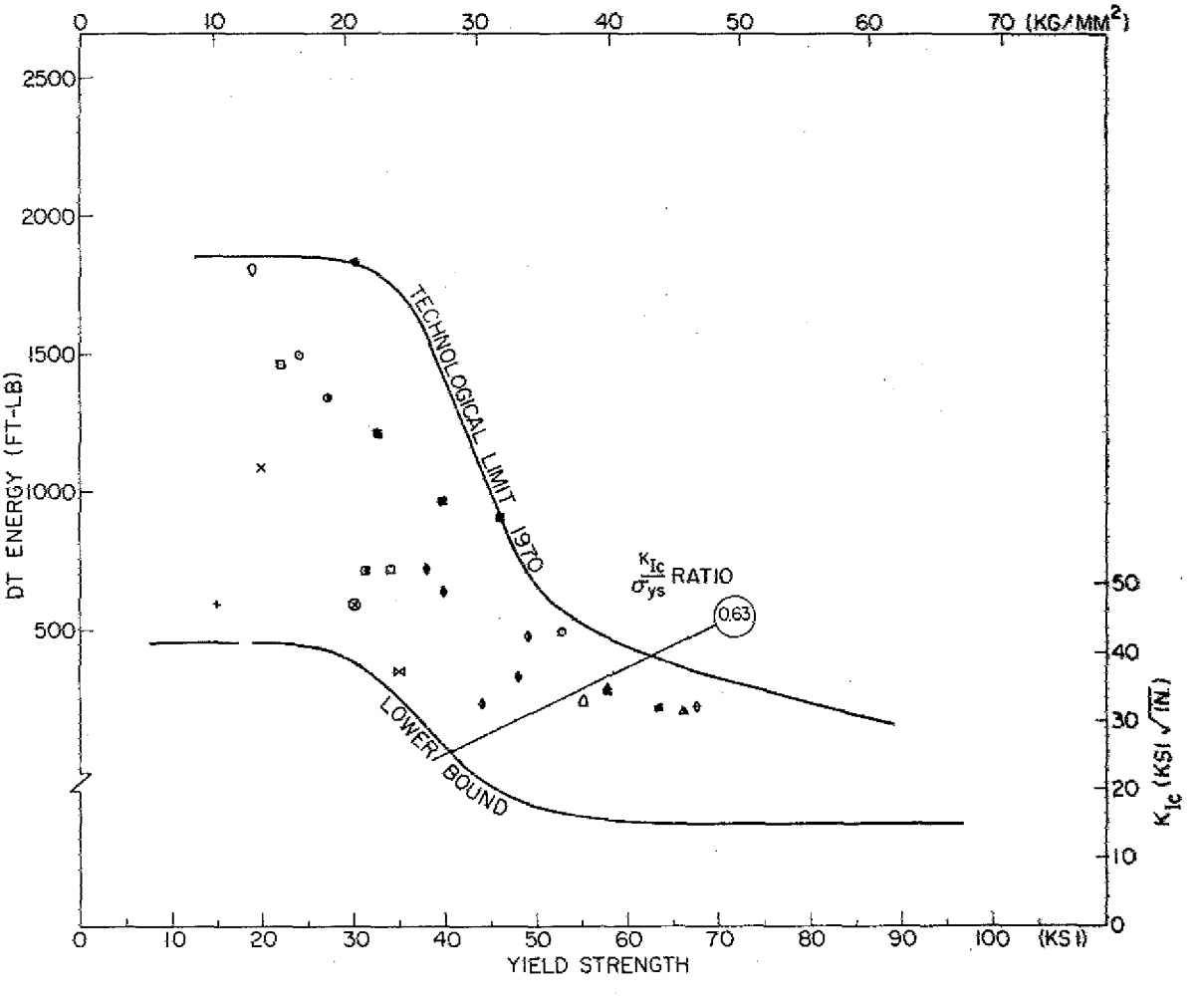

Fig. 10—Summary of 1-in. DT test data for a variety of aluminum alloys

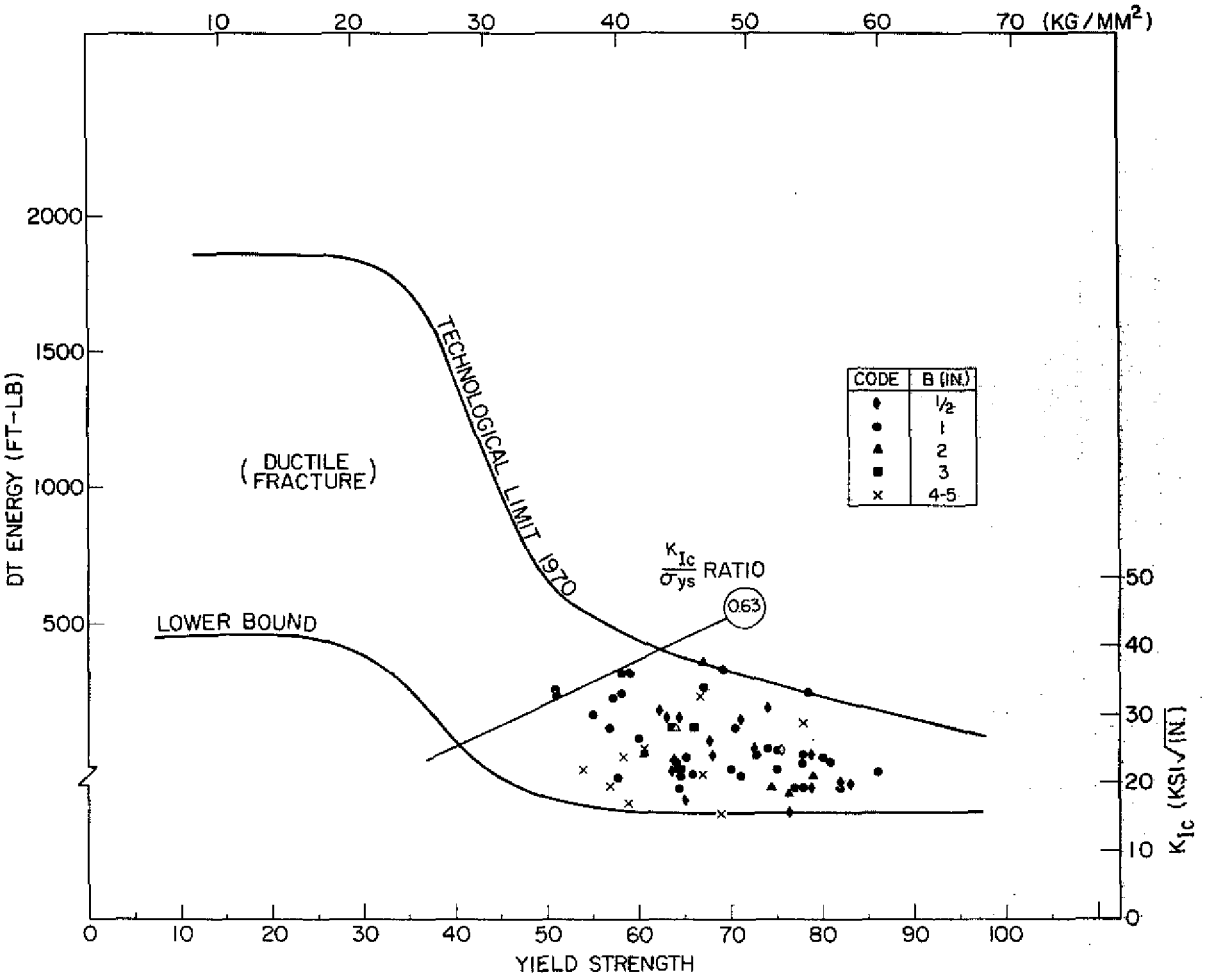

Fig. 11—Summary of K_{Ic} data for high and ultrahigh strength aluminum alloys

The relative position of the DT and K_{Ic} scales on these plots is provided by the DT-K_{Ic} correlation plots described earlier. The relationship of K_{Ic} to yield strength locates the K_{Ic}/σ_{ys} ratio lines on the plot for translation of DT and K_{Ic} data to critical flaw size-stress level predictions, a described below. The 0.63 ratio line, noted in Figs. 10 to 13, relates to the upper limit of plane strain for a 1-in. section thickness.

Ratio Analysis Diagrams for Aluminum and Titanium Alloys

The K_{Ic}/σ_{ys} Ratio Analysis Diagrams (RAD) for high-strength aluminum and titanium alloys are shown in Figs. 14 and 15. These diagrams combine the metallurgical information (TL and lower bound lines) with a structural mechanics interpretative system of critical flaw size-stress level for fracture based on a system of K_{Ic}/σ_{ys} ratio lines. The critical flaw size noted for full yield strength and 0.5 yield strength at each ratio level was obtained using the Graphic Fracture Mechanics (GFM) plot, Fig. 16, for a long, thin surface flaw in a tensile stress field. The GFM procedures are discussed in detail in Ref. 2.

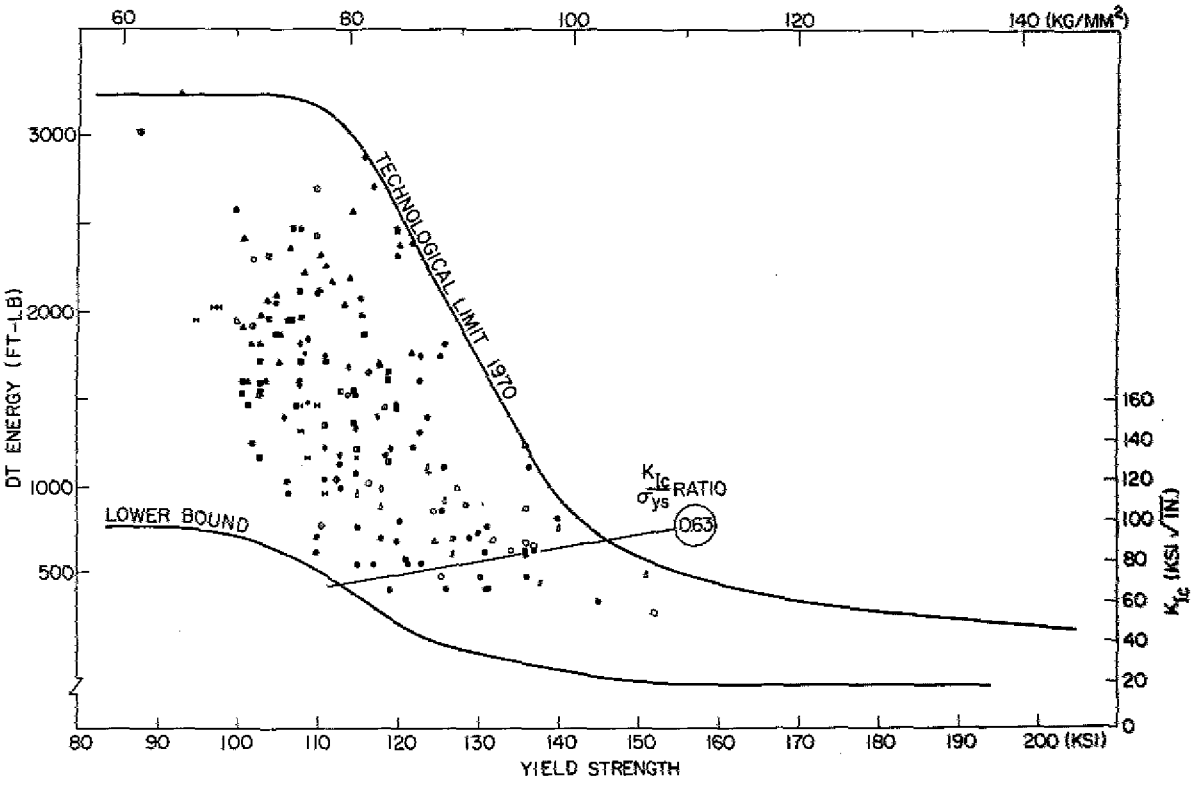

Fig. 12—Summary of 1-in. DT test data for titanium alloys produced as plates of 1- to 3-in. thickness. The alloys feature variations in chemistry, processing, and heat treatment.

The 0.3 ratio line (dashed) of the RAD in Figs. 14 and 15 represents a conservative definition of lower flaw size limit for reliable flaw detection with current techniques. For material below the 0.3 ratio line, proof test is required to insure that small cracks are not present in the structure. The 0.63 ratio limit and 1.0 ratio lines indicate the upper limits of attainable plane strain state for 1- and 2.5-in.-thick plate, respectively. When significantly above the ratio line for the specified thickness, fracture extension requires stresses above the yield strength of the metal; thus, fracture mechanics does not apply. For example, above the 0.63 ratio line, increasing DT energy values indicate increasing resistance to plastic fracture for 1-in.-thick material as defined by the R-curve slope (2). The shaded region between the 0.63 and 0.3 ratio lines denotes the range of the K_{Ic}/σ_{ys} ratio that can be used for practical critical flaw size-stress calculations for 1-in. section thickness.

A compendium RAD, which summarizes all the fracture resistance data for the generic families of aluminum and titanium, is presented in Figs. 17 and 18, respectively. The zonal locations for each family are based on all available DT and K_{Ic} data and cover the yield strength and fracture resistance range generally encountered with current alloys and processing techniques.

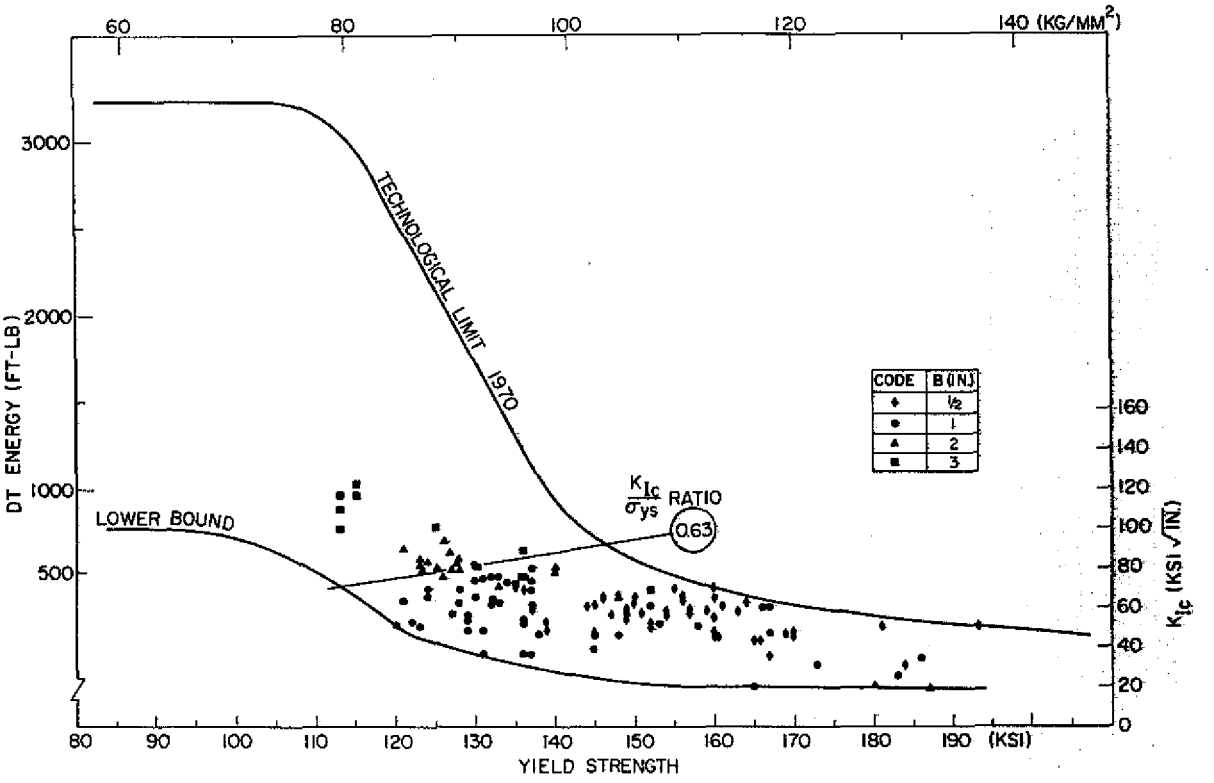

Fig. 13—Summary of K_{Ic} data for high-strength titanium alloys produced as plates of 1/2- to 3-in. thickness

Compendium Ratio Analysis Diagram for Aluminum Alloys

The general fracture toughness characteristics of each aluminum alloy family (described in Appendix A) is indexed according to the significance of RAD K_{Ic}/σ_{ys} ratio lines in Fig. 17. For example, the DT zonal location of the 5000 series alloys with respect to the ratio lines indicates very high fracture resistance capabilities (high R-curve slope) for these alloys below 40-ksi yield strength. Thus, large to huge size flaws and general plastic yielding are required for fracture. The 5000 series alloys define the TL line up to about 35-ksi yield strength. Due to the metallurgical nature of these alloys, yield strengths much above 45 ksi would be difficult to obtain in thick sections.

Alloys in the 6000 series can be expected to range in yield strength from about 20 to 65 ksi depending on the chemistry and tempering treatment. The 6061 alloy in the T651 temper is the only one of this family for which DT data are available. This alloy has a yield strength around 40 ksi for this temper condition and a level of fracture resistance below that of the TL line but still in the region where intermediate R-curve slope would be expected.

The 2000 series alloys are designed for the general yield strength ranges of 40 to 65 ksi and from RAD definitions would be expected to be brittle at the higher end of this range. Low R-curve slope characteristics would be expected for 2000 series alloys at the low end of the strength range for 1-in.-thick sections.

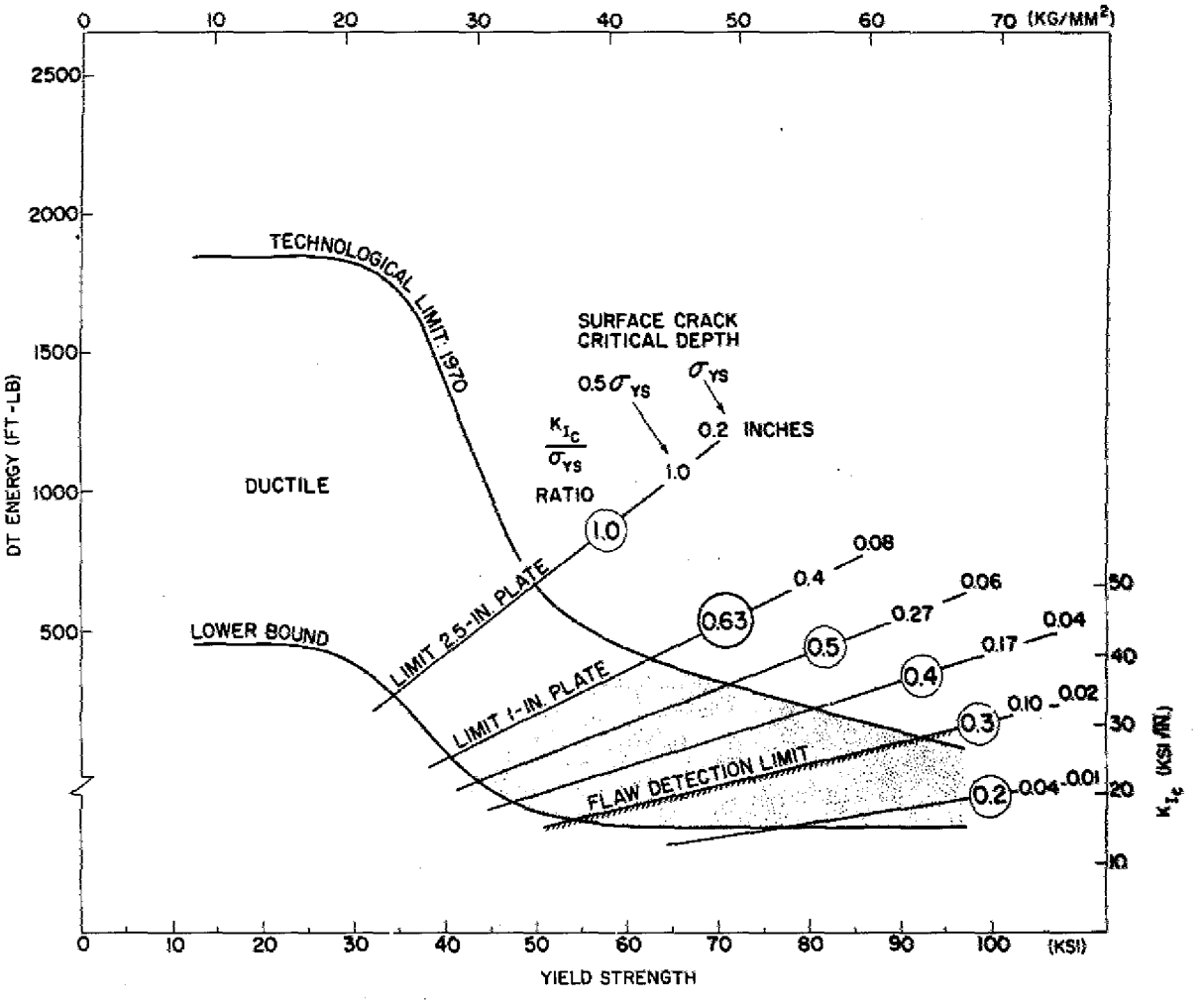

Fig. 14—RAD for aluminum alloys produced in 1- to 3-in. plate thickness. The RAD combines the metal data-bank summarization (TL and lower bound lines) with a structural mechanics interpretative system defining critical fracture conditions based on K_{Ic}/σ_{ys} ratio lines.

In general, only alloys in the 7000 series are used above 65-ksi yield strength. These alloys are the high-strength aircraft alloys, such as 7075, and are relatively brittle even for thicknesses significantly less than 1 in. The special-purpose 7000 series alloys designed for lower strength levels define the higher toughness portion of the DT zonal region. In the region of rapid TL line transition, intermediate levels of fracture resistance for these special-purpose alloys may be possible provided close control of chemistry and processing practice is exercised.

Compendium Ratio Analysis Diagram for Titanium Alloys

The compendium RAD for titanium, Fig. 18, summarizes all fracture resistance information for the generic families (see Appendix B) of titanium alloys.

Technological Limit line quality material is obtained with α and near-α alloys up to about 100-ksi yield strength. Above 100-ksi yield strength, the fracture resistance of both

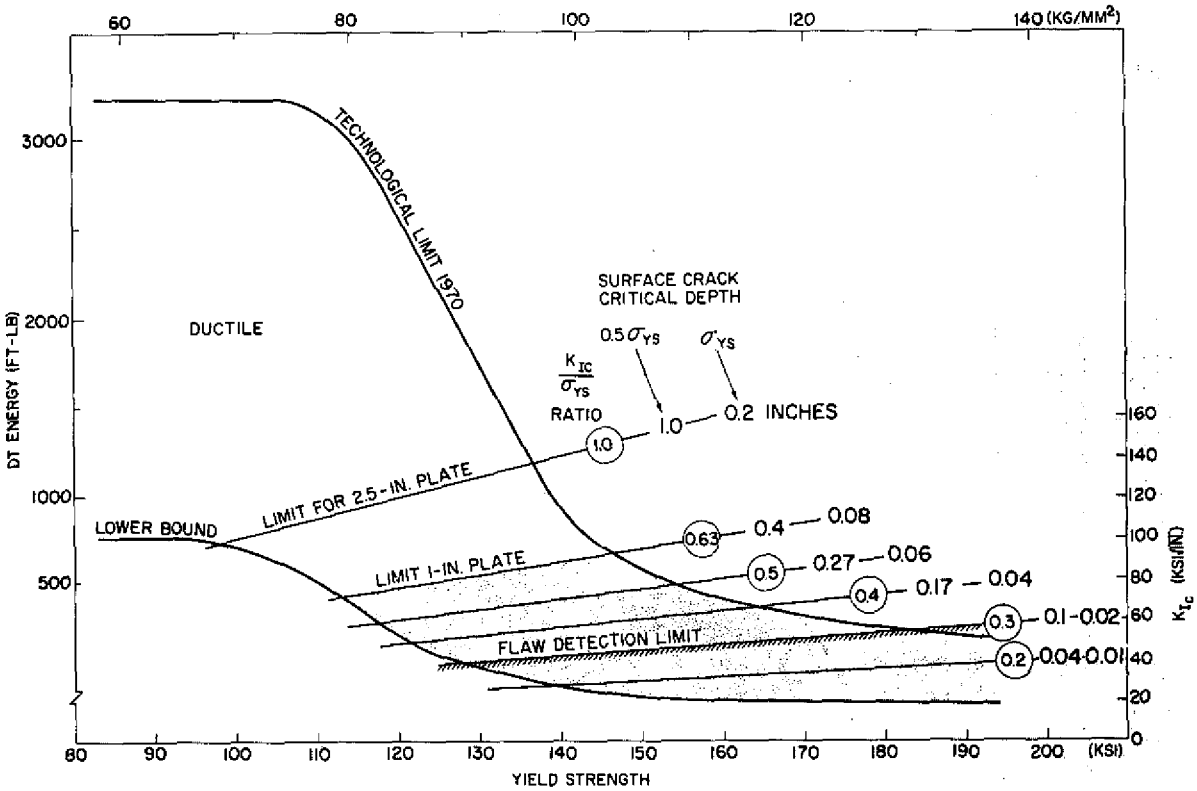

Fig. 15—RAD for high-strength titanium alloys produced in 1- to 3-in. plate thickness

types of α alloys begins to fall rapidly away from the TL line and optimum properties are no longer obtainable. This rapid drop in fracture resistance for these alloys is illustrated by the upper zone region dropping from the TL line at about 100-ksi yield strength to below the 1.0 ratio line at about 130-ksi yield strength.

Alloys of the $\alpha+\beta$ type define the TL line from 110- to about 190-ksi yield strength.

The β alloys are also expected to follow the TL line in the ultrahigh yield strength range. Limited data for several β alloys indicate that in the lower yield strength range of 110 to 130 ksi, the level of fracture resistance will be below that of the best $\alpha+\beta$ alloys but that K_{Ic}/σ_{ys} ratios of above 1.0 should be obtainable.

Many titanium alloys are capable of a wide range of yield strength and fracture resistance (ductile to plane strain) properties as a result of heat treatment, processing, and chemistry factor variations (see Appendix D). Figures 19 and 20 present available yield strength and DT fracture resistance information for several near-α and $\alpha+\beta$ alloys representing commercial practice for 1-in.-thick plate. For each alloy shown, a wide variety of metallurgical factor variations have been studied to establish the zonal regions defining the yield strength and fracture resistance limits. Details concerning the specific processing, heat treatment, and chemistry variations studied for each alloy and resultant data obtained are beyond the scope of this report.

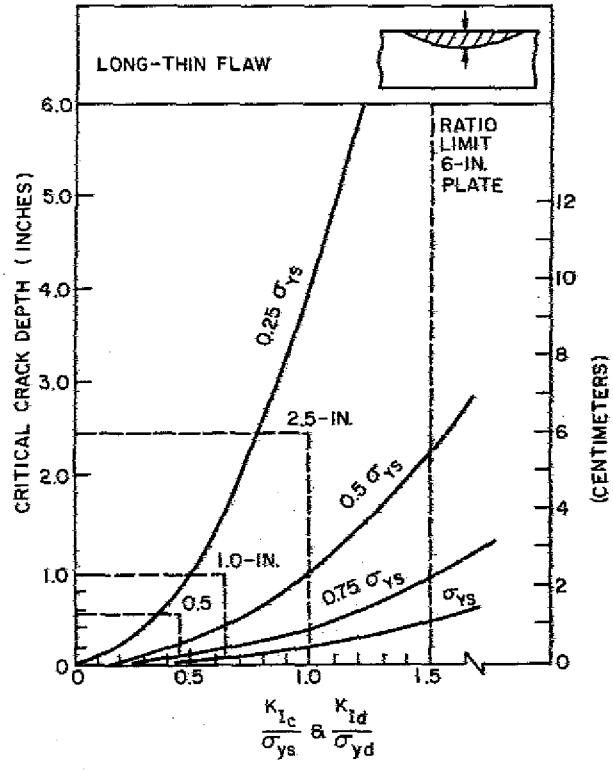

Fig. 16—GFM plot of critical crack depth versus K_{Ic}/σ_{ys} as a function of nominal tensile stress relative to yield strength for the case of long, thin surface flaw geometry. The vertical dashed lines indicate the upper ratio limit for plane strain for the specified section thickness.

SUMMARY

Fracture mechanics analytical procedures provide for characterizing the fracture toughness of high-strength titanium and aluminum alloys in terms of critical flaw size-stress level for fracture. However, valid characterizations are restricted to relatively brittle alloys at the present state of development of fracture mechanics. These restrictions derive from the transition from plane strain to plane stress as fracture resistance is increased with retention of section size. For both aluminum and titanium alloys, the DT test provides for a full spectrum definition of fracture resistance.

Interpretive procedures are evolved from the correlation of K_{Ic} with the DT test for 1-in.-thick plate. In the region of low fracture resistance corresponding to a condition of high mechanical constraint (plane strain), the correlations are reasonably accurate. Beyond the region of strictly valid K_{Ic} measurements, the correlations related to limited extrapolation to provide for interpretation of extending the K_{Ic} scale on the RAD up to 3-in.-thick section size.

Fracture resistance data have been presented which summarizes all information available for 1-in.-thick plate in terms of DT energy and K_{Ic} values. The DT energy is indexed to K_{Ic}

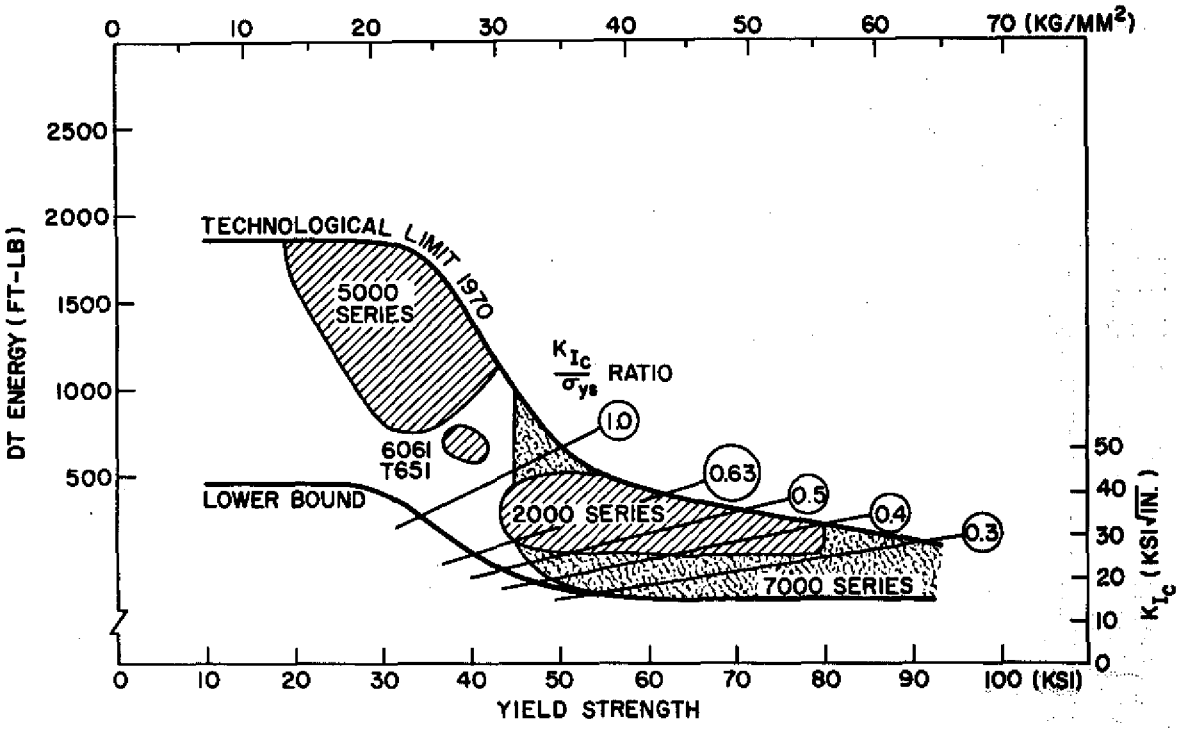

Fig. 17—RAD for generic families of structural aluminum alloys

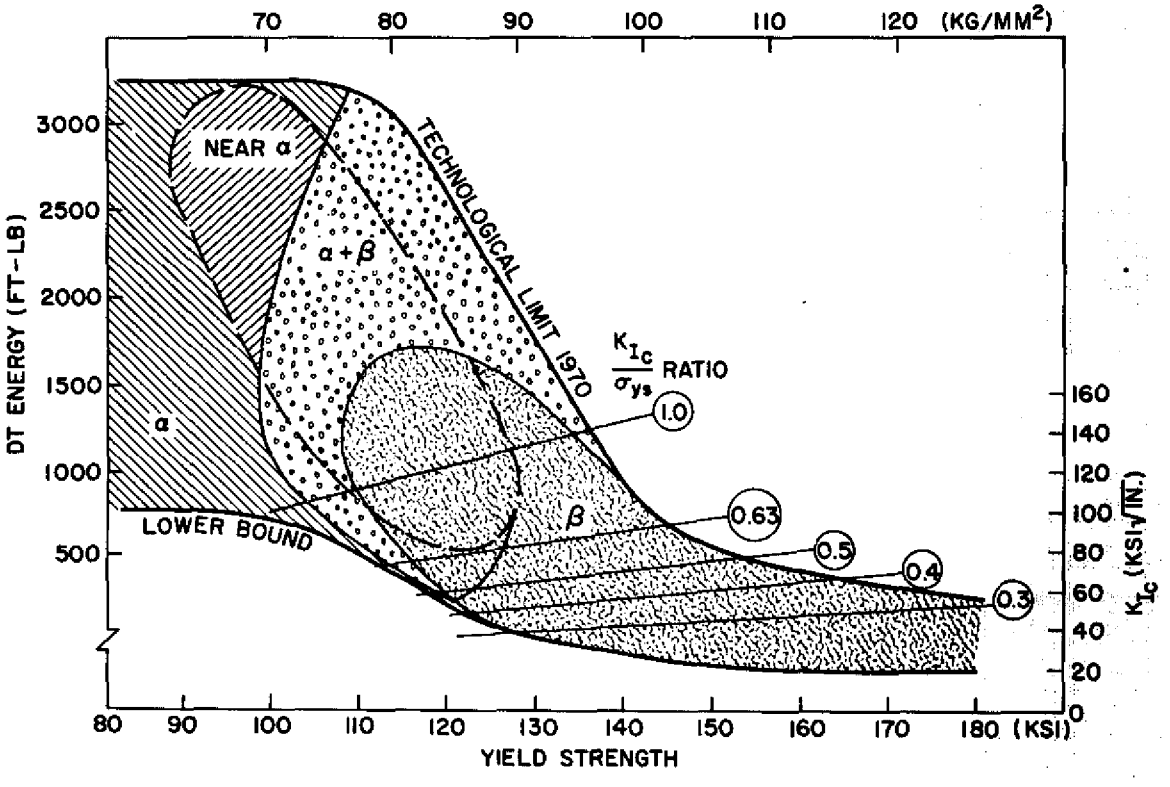

Fig. 18—RAD for generic families of structural titanium alloys

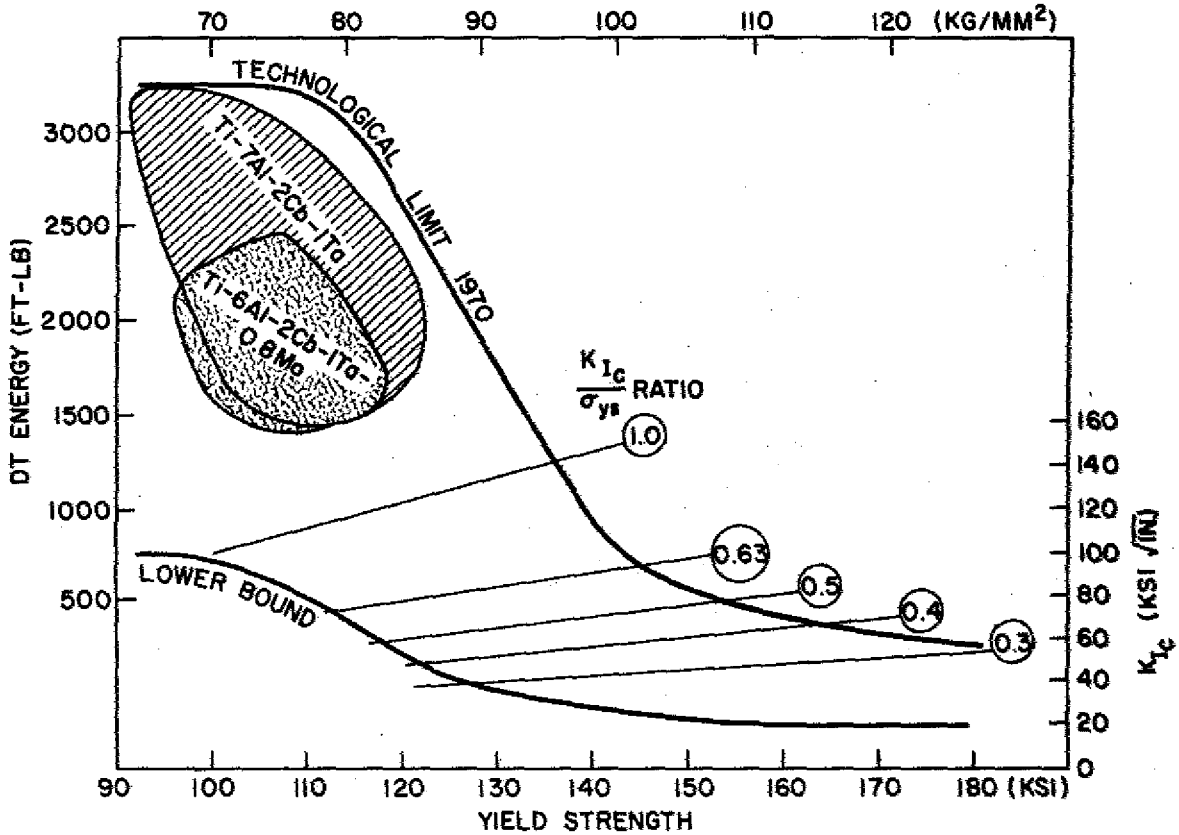

Fig. 19—RAD zonal regions which define the strength and fracture toughness limits established for several near-α titanium alloys

according to the correlation plot. A TL curve indicating the highest level of fracture resistance attained for any level of yield strength is defined by the DT and K_{Ic} data. Similarly, a lower bound curve defines the lowest level of fracture resistance observed for commercial material. A dynamic drop of the TL line from high to low fracture resistance in a relatively narrow range of yield strength characterizes the strength transition behavior of these materials.

The K_{Ic} scale on the diagram provides for translation of the DT and K_{Ic} data to flaw size-stress definitions for fracture initiation. This is expressed in terms of K_{Ic}/σ_{ys} ratio lines, which relate to flaw size calculations for plane strain fracture. For the ductile region of the RAD, DT energy corresponds to increasing fracture extension resistance as defined by the R-curve slope. Critical flaw sizes determinations cannot be made for this region since fracture mechanics K_{Ic} procedures do not apply. However, an increasing R-curve slope signifies rapidly increasing resistance to ductile fracture for large flaws.

The spectrum of commercial aluminum and titanium alloys covers a wide range of fracture resistance. Compendium RAD diagrams are presented to summarize the current strength and fracture resistance limitations for generic families of aluminum and titanium.

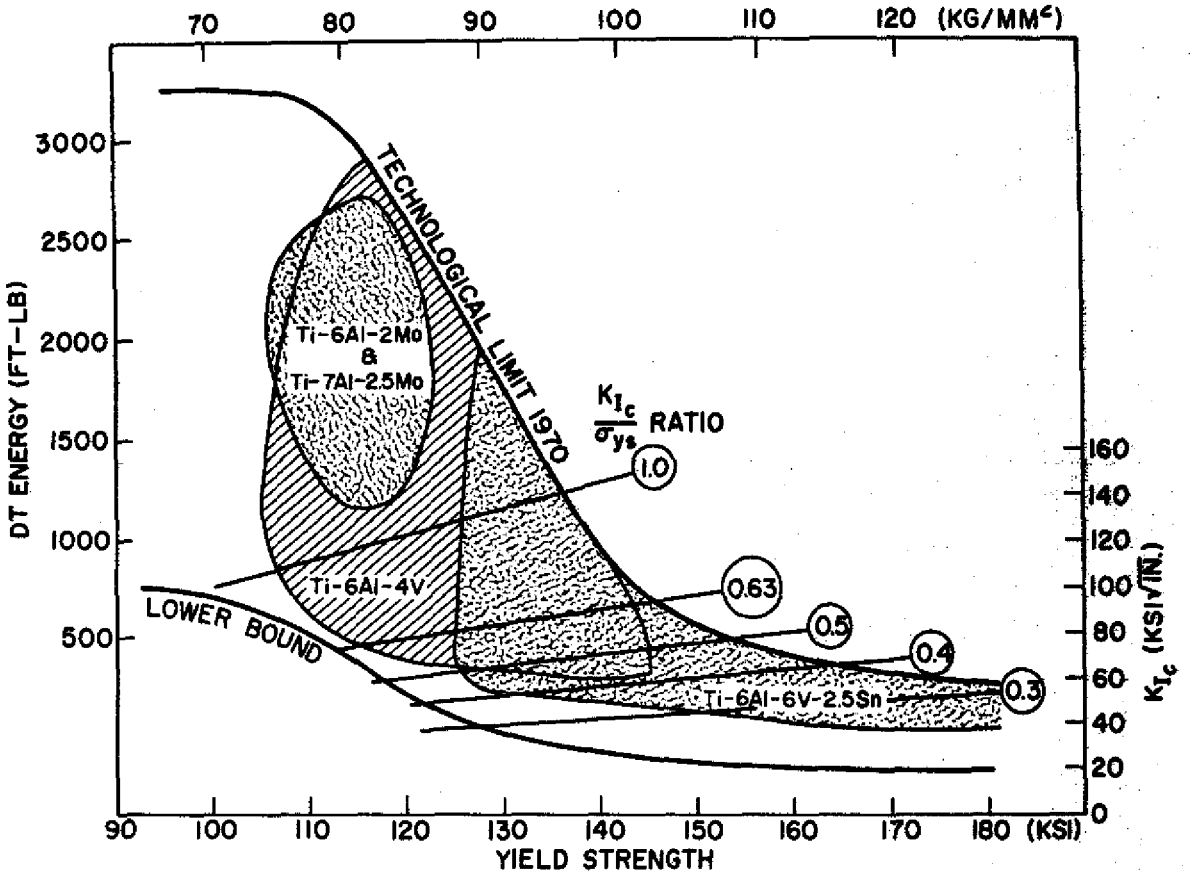

Fig. 20—RAD zonal regions established for a variety of high-strength $\alpha+\beta$ titanium alloys

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1. W.S. Pellini, "Evolution of Engineering Principles for Fracture-Safe Design of Steel Structures," NRL Report 6957, Sept. 23, 1969.
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3. E.A. Lange, P.P. Puzak, and L.A. Cooley, "Standard Method for the 5/8-Inch Dynamic Tear Test," NRL Report 7159, Aug. 27, 1970.
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APPENDIX A

PHYSICAL METALLURGY CHARACTERISTICS OF ALUMINUM ALLOYS

The principal strengthening mechanism for most high-strength aluminum alloys is precipitation hardening, or the dispersion of a system of fine intermetallic particles throughout the softer aluminum matrix. Another strengthening mechanism of importance for certain alloys (to be described) is strengthening of the matrix by solid solution of alloying elements.

Aluminum alloys are classified according to alloy systems and temper, i.e., heat treatment and/or mechanical working. The various alloy systems are number-coded according to their primary compositional elements under the guidance established by the Aluminum Association:

1XXX — Al
2XXX — Al-Cu
3XXX — Al-Mn
4XXX — Al-Si
5XXX — Al-Mg
6XXX — Al-Mg-Si
7XXX — Al-Zn-Mg

Specific alloy designations, such as 2024 and 7075, describe the major alloy system and indicate additions of other specific alloying elements.

Temper designations of aluminum alloys refer to the heat-treated or worked condition of the material. The *H* designation (as in 5086-H112) refers to strengthening by strain hardening with or without subsequent thermal treatment to produce softening. The digits following the *H* indicate the specific processes involved. The *T* temper designates a thermal strengthening process followed by an aging treatment or mechanical work or both. Again, the digits indicate the specific processes involved. A comprehensive treatment of alloy systems and temper designations is presented in Refs. A1 and A2.

Principal alloys in the 2000, 5000, 6000, and 7000 aluminum systems were covered in the studies at NRL. The following sections provide a brief review of the general metallurgical nature of each of these families and their commercial importance.

2000 SERIES ALLOY SYSTEM

Alloys of the 2000 system contain copper as the major alloying element; thus, their microstructures are characterized by copper-containing phases, such as $Cu\text{-}Al_2$ or $Cu\text{-}Mg\text{-}Al_2$, if the alloy also contains magnesium. These alloys can be heat treated to high levels of

strength. At the higher strength levels of their heat-treatable range, these alloys feature low resistance to fracture. At intermediate levels of strength the formability and machinability of these alloys make them desirable for machined parts, forgings, and structural members in aircraft.

5000 SERIES OF ALLOYS

Magnesium is the major alloying element for this family of alloys. Strengthening of these alloys is primarily by solid solution of magnesium in the aluminum matrix. Often, particles of the Mg_2Al_3 phase will be observed in the microstructure either as undissolved constituent particles or as a fine precipitate developed due to heat treatment. Particles of magnesium combined with other elements, such as silicon and chromium, will also be evident generally in proportion to the total amount of these elements added to the basic Al-Mg alloy.

The 5000 series alloys can be processed to yield strengths of approximately 30 to 40 ksi. These alloys have a high inherent resistance to fracture. The good corrosion resistance and welding characteristics of most of these alloys have led to their extensive use in various structural applications.

6000 SERIES ALLOYS

The 6000 series of alloys is based on alloying aluminum with magnesium and silicon. The microstructure contains particles of the Mg_2Si phase which is largely dissolved during solution heat treatment, then precipitated during aging. Thus, the character and amount of this phase present in the microstructure highly depend on processing variables. Usually, silicon particles are also observed in the microstructure due to additions of slightly excessive amounts.

In general, these alloys have properties similar to the 5000 series and are used in many of the same applications.

7000 SERIES ALLOYS

The 7000 series of alloys features zinc and magnesium as the basic alloying elements. For some alloys in this series, such as 7075, copper is also a major alloying element. Since these alloys represent the highest strength commercial alloys, they are usually used in the solution heat-treated and aged condition. Most of the zinc, magnesium, and copper is retained in solid solution in the matrix or precipitated into the microstructure as extremely fine particles. Usually, other elements, such as chromium, of which 7075 contains a small amount, and silicon, will cause discrete particles containing these elements to appear in the microstructure. However, these particles have little influence on the strengthening of the alloy.

The 7000 series alloys, particularly the alloy 7075, are favored by the aircraft industry. These alloys are heat treatable to yield strength levels of 70 to 80 ksi and in this condition are generally of low fracture resistance, especially for 1 in. or greater section thickness.

REFERENCES

A1. American Standards Association, Alloy and Temper Designation System for Wrought Aluminum (ASA H35.1—1962).

A2. K.R. Van Horn, editor, "Aluminum," Vol. 1, Metals Park, Ohio:ASM 1967.

APPENDIX B

METALLURGICAL CHARACTERISTICS OF HIGH-STRENGTH TITANIUM ALLOYS

The three basic types of commercial titanium alloys are alpha, alpha+beta, and beta. The alpha (α) phase has a hexagonal-close-packed crystal structure and is the low-temperature allotropic transformation phase. The beta (β) phase is the high-temperature phase and has a body-centered-cubic crystal structure.

The α peritectoid system is typical for titanium alloyed with the α stabilizers aluminum, oxygen, nitrogen, and carbon. These elements have a high solubility in α titanium, causing an increase in a solute content which raises the β transus temperature. The elements are all solid solution strengtheners in the α phase. Aluminum enters the titanium atomic lattice as a substitutional element, replacing titanium atoms. Oxygen, nitrogen, and carbon from interstitial solid solutions with titanium and occupy the vacant spaces existing between the titanium atoms in the crystal structure. Aluminum is a basic alloying element for practically all titanium alloys since it decreases the density of the alloy and at the same time substantially strengthens the α phase without degrading weldability and fracture toughness.

Beta stabilizing elements have a high solubility in β titanium and limited solubility in α titanium. They lower the β transus temperature with increasing alloy content. The elements iron, manganese, silicon, chromium, copper, and hydrogen are all strengtheners of the β eutectoid type. The elements molybdenum, vanadium, columbium, and tantalum are strengtheners which form the β isomorphous-type system and have unlimited solubility in the β phase.

The neutral alloying elements tin and zirconium have high solubility in both α and β titanium. They are relatively moderate solid solution strengtheners which do not appreciably change the β transus temperature.

PHYSICAL METALLURGY ASPECTS

Alpha Alloys

Alpha alloys are essentially all α phase at room temperature, even when rapidly quenched from temperatures well above the β transus. Usually small amounts of the β phase can be found due to impurities, such as iron, managanese, and silicon. If the combined amount of these elements is above 0.3%, the β phase may affect the metallurgical character of the α alloys. Heat treatment of these alloys can have considerable effect on fracture toughness, but the response with respect to strength is relatively small. The useful yield strength range of thick-section α alloys (0.08 maximum oxygen) is below 120 ksi. In general, α alloys are considered

to have good welding characteristics. Typical α alloys are titanium containing fixed amounts of oxygen for increased strength, usually referred to as "commercially pure" titanium, Ti-3Al, and Ti-5Al-2.5Sn.

Alpha-Beta Alloys

Alloys of the $\alpha+\beta$ type can be subdivided into two groups. Alloys containing β stabilizing elements in amounts sufficient to impart only a limited amount of heat-treatment response with respect to strength are termed near-α (or super-α) alloys since their metallurgical character closely resembles that of α alloys. The factors discussed for α alloys apply equally well to near-α alloys. Ti-6Al-2Cb-1Ta-0.8Mo is a typical near-α alloy.

Alloys containing β stabilizing elements sufficient in amount to impart a high degree of heat-treatment response are termed $\alpha+\beta$ alloys. Those alloys containing less than about 5% β stabilizing elements range from about 115-ksi yield strength in the annealed condition to about 150-ksi yield strength in the heat-treated condition. Alpha+beta alloys containing greater amounts of β stabilizing elements are heat treatable to close to 200-ksi yield strength from an annealed yield strength of approximately 125 to 130 ksi. Intermediate strengths are obtained by varying the heat-treatment conditions. In general, an increase in the amount of β stabilizing elements in $\alpha+\beta$ alloys (i.e., increase in β phase present) results in a decrease in fracture resistance.

Alpha+beta alloys are strengthened by solution annealing high in the $\alpha+\beta$ region or in some cases above the β transus temperature, followed by a rapid cool to room temperature. The solution-annealing temperature determines the amount and composition of the β phase which is retained in an unstable condition at room temperature. Aging treatments in the range of 900° to 1200°F precipitates α titanium from the β phase, which enriches the β phase in β stabilizing elements and results in increasing its stability.

The weldability of $\alpha+\beta$ alloys decreases with increasing amount of β stabilizing elements. For example, welds in the alloy Ti-6Al-4V can be made to match the plate properties, if proper protection against oxygen contamination is applied, whereas weldments with inferior ductility and fracture resistance properties compared to plate are generally obtained for an alloy such as Ti-6Al-6V-2.5Sn-1Cu-0.5Fe (normally designated Ti-6Al-6V-2.5Sn).

Beta Alloys

Beta alloys contain sufficient alloying elements to allow only the β titanium phase to exist at ambient temperature. These alloys are metastable and are classified as eutectoid or compound-free alloys. When moderately heated for extended times, the β phase of the eutectoid-type alloys will sluggishly transform and precipitate a compound due to eutectoid decomposition. The α phase is precipitated from the β phase of the compound-free-type alloys. In general, β alloys can be heat treated into the same strength range as $\alpha+\beta$ alloys.

Few β alloys have been produced on a commercial basis. In general, they have been designed for ultrahigh strength and display relatively poor fracture resistance properties at moderate strengths. The low resistance is due primarily to the high content of β stabilizing elements in these alloys.

One commercial β alloy, which is considered weldable, is the eutectoid-type Ti-13V-11Cr-3A1. The as-deposited welds are fairly ductile with a fracture resistance equivalent to annealed plate. However, it has the problem of low ductility and low resistance to fracture in the weld compared to the plate when heat treated to high-strength levels.

Recent studies on the newly developed compound-free β alloys, such as Ti-11Mo-5.5Sn and Ti-11Mo-5SN-5Zr, indicate that significant improvements in fracture resistance and weldability may be expected in β alloys in the near future.

Oxygen in Titanium

All commercial titanium alloys contain oxygen in substantial amounts. Some of the oxygen is contained in the titanium sponge used for melting into ingots. It is not eliminated or reduced in the vacuum melting process. Furthermore, oxygen is often added to increase the strength of the material. Thus, for all practical purposes, oxygen should be considered an alloying element in titanium.

In line with its role as a potent strengthener, oxygen seriously affects the fracture resistance of titanium alloys—especially in thick sections, Fig. B1. As a general approximation, it is desirable to keep the level of oxygen below 0.8% for optimization of fracture resistance in thick sections. Fracture resistance deteriorates rapidly with increasing oxygen above 0.08%; the critical level of oxygen appears to vary only slightly for different alloys. To date, all available data indicate that the commercially designated extra-low-interstitial (ELI) grade alloys, which contain from 0.10% to 0.14% oxygen and especially the alloys of standard commercial purity (0.15% to over 0.2% average oxygen content), fall well below the maximum level of fracture resistance observed for the very low oxygen-containing alloys.

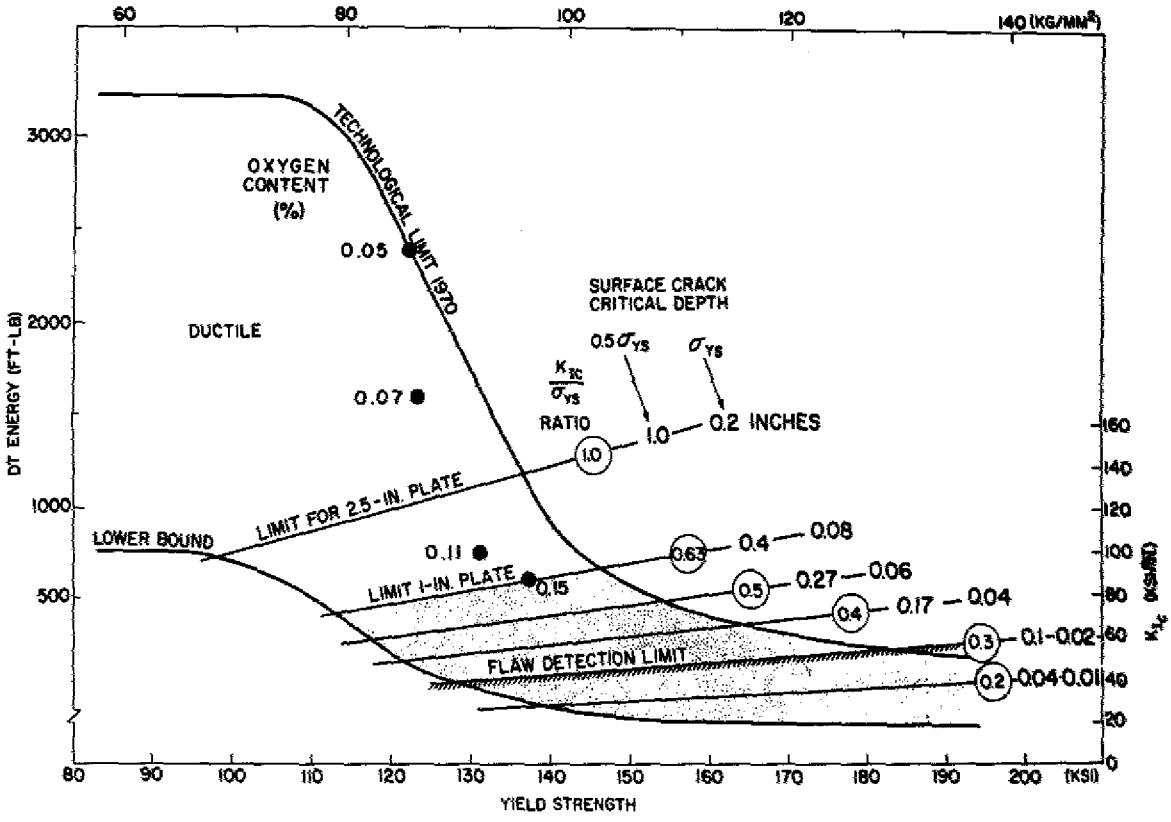

Fig. B1—Data illustrating that increasing oxygen content decreases fracture resistance and increases yield strength. The data are for Ti-6Al-4V alloys involving a fixed processing practice.

APPENDIX C

CHARACTERISTICS OF ALUMINUM ALLOYS COMMERCIALLY PRODUCED AS 1-IN.-THICK PLATE

Alloy	Temper	σ_{ys} (ksi)	1-in. DT Energy (ft-lb)
2020	T651	76	90
2021	T8151	63.4	234
2024	T351	49	472
2024	T4	48	330
2024	T851	67.6	234
2219	T851	55	240
2219	T87	58	290
5083	0	19	1790
5086	H112	24	1480
5086	H116	27	1330
5086	H117	30	1832
5456	0	22.2	1454
5456	H116	31.2	714
5456	H117	32.5	1205
5456	H117	39.6	958
5456	H321	34	726
6061	T651	40	640
6061	T651	38	720
7005	T63	46	900
7075	T7351	66	210
7075	T6	78	120
7075	T6	75	71
7079	T6	75	93
7106	T63	53	490

APPENDIX D

CHARACTERISTICS OF TITANIUM ALLOYS COMMERCIALLY PRODUCED AS 1-IN.-THICK PLATE

Alloy	Nominal Yield Strength Range (ksi)	Nominal DT Energy Range (ft-lb)
α Alloys		
Ti-5Al-2.5Sn	110-125	1500-1900
Ti-5Al-5Sn-5Zr	110-115	1000
Unalloyed	80	800
Near-α Alloys		
Ti-7Al-2Cb-1Ta	90-120	1600-3200
Ti-8Al-2Cb-1Ta	95-115	1500-3000
Ti-6Al-2Cb-1Ta-0.8Mo	95-120	1500-2500
Ti-6Al-2Cb-1Ta-1.2Mo	105-115	1200-1700
Ti-6.5Al-5Zr-1V	125	800
Ti-6Al-4Zr-2Sn-.5Mo-.5V	110-115	1500
Ti-6Al-4Sn-1V	130	1900
Ti-6Al-4Zr-1V	110	800
Ti-7Al-3Cb-2Sn	110	1500-1800
α+β Alloys		
Ti-6Al-4V	105-140	300-2990
Ti-6Al-2Mo	105-125	1200-2700
Ti-7Al-2.5Mo	100-120	1400-2600
Ti-7A1-1Mo-1V	105-115	800-1500
Ti-8A1-1Mo-1V	115-125	1000
Ti-6Al-4V-2Sn	120	1000
Ti-6Al-3V-1Mo	110-130	1000-2700
Ti-6Al-4V-2Mo	120-150	500-1100
Ti-6Al-2Sn-1Mo-1V	100	1900
Ti-6Al-4Zr-2Mo	120-135	700-1400
Ti-4Al-3Mo-1V	95-100	2000
Ti-6Al-6V-2Sn-1Cu-.5Fe	125-135	500-700
Ti-6Al-6V-2Sn-2Mo	125-175	300-600
Ti-5Al-2Sn-2Mo-2V	105-120	1100-2300
β Alloys		
Ti-11Mo-5Sn	90-100	3200-3600
Ti-11Mo-5.5Sn-5Zr	100-110	1800
Ti-10Mo-5.4Sn	115-135	1200-1600
Ti-2Al-4Mo-4Zr	90-100	2700-3100
Ti-3Al-3Mo-3Zr	90-100	1700
Ti-3Al-13V-11Cr	136	820