CRITICAL REVIEW

MECHANICAL PROPERTIES OF TITANIUM ALLOYS

J. B. Borradaile and R. H. Jeal
Rolls-Royce Ltd., Aero Division, P.O.Box 31, Derby, England.

Introduction

Titanium alloys are noted for their high specific strengths and although relatively expensive have been used as structural materials in the aerospace industry for over thirty years. In that time many alloys have been developed with a range of mechanical properties which depend primarily upon composition, processing history and heat treatment. The important properties are listed below:

(a) The zero time mechanical properties - tensile strengths, ductilities and fracture toughness.

(b) The real time static properties - creep strength and stability.

(c) The real time dynamic properties - high and low cycle fatigue and crack propagation.

The final choice of material for a particular application is usually dependent not just on one mechanical property but on the optimum combination of several competing characteristics. For example, above a certain temperature, creep strength is often the limiting parameter in design, but it still has to be accompanied by adequate elevated temperature tensile properties, LCF resistance and both metallurgical and surface stability. Below this temperature, which varies with the alloy, LCF rather than creep strength tends to be limiting but satisfactory fracture toughness and crack growth rates are still needed to guard against catastrophic failure.

To illustrate how different combinations of properties are achieved in various components, it is important to understand how the composition, processing and heat treatment of titanium alloys can modify the structure and control the mechanical properties.

The Range Of Compositions For Titanium Alloys

Titanium undergoes an allotropic phase change at 882°C, transforming from the lower temperature hexagonally close packed alpha phase to the high temperature body centred cubic beta phase. The alpha phase is stronger, particularly so in creep, but less ductile than the beta phase. However, in comparison with other HCP materials (e.g. Mg and Zn), the alpha phase is relatively ductile because of its low c/a ratio.

Titanium alloying additions can be divided into two groups, alpha and beta stabilisers. The alpha stabilisers can then be further sub-divided into:

i) "Pure" alpha stabilising elements which raise the beta transus (such as aluminium and interstitial oxygen) and

ii) "Alpha strengthening" elements which have a high solubility in the alpha phase (such as tin and zirconium) but which do not raise the beta transus.

The amount of alpha stabiliser which can be added to titanium is restricted by the formation of α_2, a coherent ordered phase based on Ti_3Al which causes embrittlement. Excessive α_2 can be avoided (1) if a limit of 9 wt. % is placed on the 'aluminium equivalent' of the alloy, this value being derived from the empirical summation:

Aluminium Equivalent = Al + ⅓Sn + Zr + 10 (O + C + 2N)

The beta stabilising elements can also be subdivided. They include:

i) The beta isomorphous additions such as BCC elements vanadium and molybdenum which continuously lower the beta transus and

ii) The beta stabilising eutectoid formers such as iron and chromium which also lower the beta transus until they are interrupted by compound formation.

(a) Alpha Alloys

Single phase alpha alloys are solid solution strengthened by additions of alpha stabiliser. They, therefore, exhibit low to medium strengths but have good stability and weldability if little scope for heat treatment. They are used primarily as sheet materials for fabrications and include the important commercially pure titanium grades and the alloy Ti-5Al-2½Sn. The cold formability of this group of alloys becomes more difficult as strengths increase. However, the elevated temperature properties of the substitutionally strengthened alloys are good.

(b) Near Alpha Alloys

The low strengths of the simple alpha alloys and poor hot workability of high aluminium compositions are alleviated by the addition of the beta stabilising elements. With only 1-2 wt. % of beta stabilising addition, a compromise is achieved between the higher strengths which are available with alpha + beta alloys and the high temperature properties of the simple alpha alloys. This group is called the near alpha alloys and includes IMI 679, Ti 6242, IMI 685 and IMI 829, the compositions of which are given in Table I.

(c) Alpha + Beta Alloys

Larger additions of beta stabilisers (4-6%) allow the development of alloys which can be heat treated to high strengths by martensitic transformation and fine precipitation of beta. This group, the alpha + beta alloys includes general purpose Ti 6/4, higher strength Ti 6-6-2 and IMI 550 and the very strong Ti 6246, the compositions of which are again given in Table I.

(d) Metastable Beta Alloys

With greater additions of beta stabilising elements, the beta phase can be retained in a metastable form at room temperature and then aged to form very fine Widmanstatten alpha in a matrix of enriched beta. This group is called the metastable beta alloys and includes B120 VCA, Beta III and Ti 10-2-3.

(e) Beta Alloys

Very large beta stabiliser additions can give a structure of stable beta. Such alloys, however, more closely resemble the refractory metals with their high densities, poor ductilities and poor oxidation resistance than common engineering titanium alloys and commercial applications have yet to be found.

(f) Titanium Aluminides

Titanium aluminides can be classed as another group of alloys which have received a lot of recent attention. They are based on the intermetallic compounds Ti_3Al and TiAl and are generally lighter, stiffer and more temperature resistant than conventional titanium alloys. However they are exceedingly brittle, hard to form and prone to segregation and these deficiencies preclude their use in all but very specialised circumstances.

The Effect Of Processing And Heat Treatment On Microstructure

The mechanical properties and microstructures of titanium alloys are strongly dependent on processing history and heat treatment. For example, when an alloy is hot worked and heat treated below the beta transus, in the (alpha + beta) phase field, it generates a duplex microstructure of primary alpha and transformed beta. The primary alpha forms by nucleation and growth during the alpha + beta working operation and its morphology can vary from elongated plates in lightly worked material to equiaxed grains in heavily worked material. The equiaxed morphology is a result of recrystallisation. The transformed beta, on the other hand, refers to regions which were beta phase at the solution treatment temperature, annealing temperature, or finishing temperature of the hot working operation and can also have a varied morphology which depends primarily on cooling rate and composition. On cooling the beta can transform martensitically to alpha or on slower cooling can transform by nucleation and growth to Widmanstatten alpha + beta. Sometimes metastable beta can be retained. When the finishing or solution treatment temperatures are above the beta transus, the microstructures consist entirely of transformed beta.

Alpha martensite decomposes on subsequent ageing to precipitates fine beta which gives useful increments in strength. Martensite is, however, difficult to form to thick sections because of the low thermal conductivity of titanium. It is also difficult to form martensite in lean beta stabilised alloys because the kinetics of the competing Widmanstatten nucleation and growth mechanism are more rapid.

The Widmanstatten alpha can have several morphological variations. Slow cooling favours the formation of colonies of similarly aligned alpha platelets, together with prior beta grain boundary alpha, whereas faster cooling and higher beta stabiliser contents favour a more basketweave morphology.

Metastable beta decomposes on subsequent ageing to precipitate fine alpha which also gives useful increments in strength. The more weakly stabilised the beta, the larger the volume fraction of alpha which can be precipitated and the higher the strength which can be achieved. Two other decomposition reactions may occur at low temperatures; these are ω formation which is generally limited to relatively lean beta alloys and a phase separation reaction $\beta \rightarrow \beta_1 + \beta_2$ which occurs in the richer alloys. These two reactions are generally considered undesirable and need to be avoided.

Alpha + beta solution treatment of alpha + beta processed material allows control over the final duplex microstructure, the higher the solution treatment the more transformed beta there is in the microstructure. Slow cooling from this alpha + beta region coarsens the primary alpha and progressively enriches the remaining beta, which can be retained at grain boundary triple points as in a recrystallisation anneal.

The Effect Of Composition And Microstructure On Mechanical Properties

The mechanical properties vary considerably with different microstructures and the variations are discussed below:

(a) Near Alpha Alloys

The most practical creep resistant titanium alloys are the near alpha alloys. These fall into two groups.

1) The $\alpha + \beta$ processed alloys such as IMI 679 and Ti 6242 which were developed to have a duplex microstructure and

2) The β heat treated alloys such as IMI 685 and IMI 829 which were developed to utilise the increased temperature capability of acicular microstructures.

IMI 679 (Ti-11Sn-2¼Al-5Zr-1Mo-0.25Si) has been used for many years at temperatures up to 450°C. It has a large amount of alpha strengtheners but its aluminium equivalent is less than Rosenthal's stability limit and it therefore has good ductility. It has sufficient beta stabiliser to make the alloy heat treatable and forgeable without significantly reducing its creep resistance. Silicon is also added in excess of its solubility limit to ensure the maximum is in solution in order to boost creep resistance. At the time of its development it was taken for granted that the heat treatment should be similar to alpha + beta alloys, namely a solution in the alpha + beta phase field followed by quench and a stabilisation anneal and as the silicon is wanted in solution the stabilisation anneal temperature is kept low at 500°C. (2)

Although IMI 679 has good creep resistance it forfeited some of its low density and later alloys of this general type have attached more importance to this property. In Ti 6242 (Ti-6Al-2Sn-4Zr-2Mo-0.1Si) for instance, high aluminium is preferred to high tin, while the increase in molybdenum improves forgeability. It is usual to follow the solution anneal of this alloy with an air cool, thus reducing strength but giving better fracture toughness (see later) and lower residual stresses.

The two alloys, IMI 679 and Ti 6242 represent close to the ultimate in alpha + beta processed and heat treated creep resistant alloys and further improvements must come by producing acicular microstructures. To this end, in the United States Ti 6242 has been used in a beta processed condition when creep resistance is of prime importance. In the United Kingdom, where greater emphasis tends to be placed on temperature capability, special alloys for beta heat treatment have been developed, the first of these being IMI 685 (Ti-6Al-5Zr-0.5Mo-0.3Si). In this, low alloy density was considered important and the alpha strengtheners were restricted to aluminium and zirconium with no addition of the heavier element, tin. Creep has been improved by the acicular microstructure, by reducing the beta stabiliser to 0.5% molybdenum and raising the silicon level to make use of the higher solubility of silicon on beta heat treatment. (2)

IMI 829 (Ti-5½Al-3½Sn-3Zr-1Nb-0.3Mo-0.3Si) was developed from IMI 685. Its higher alloying additions of both alpha and beta stabilisers slow down diffusion and make the alloy both more creep resistant and more hardenable.

The general trends in the mechanical properties of titanium alloys can be observed by comparing the data for these alloys in their normal conditions (Figures 1-5). The duplex microstructures of IMI 679 and Ti 6242 exhibit better strengths, better LCF & better tensile ductilities compared to the acicular microstructures of IMI 685 and IMI 829. In comparison the acicular microstructures possess better creep, better fatigue crack propagation resistance and better fracture toughness. (3)

The improved properties of the duplex microstructures are due to the general refinement of the microstructure by the alpha + beta working. Typically, an $\alpha + \beta$ processed structure will have a primary α grain size of 0.01-0.05mm whereas an acicular microstructure will have an alpha colony size, controlled by the prior β grain size and cooling rate, of 0.05-0.4mm with a mean of usually 0.1mm. The smaller mean free slip paths of the duplex structures improve the tensile properties as can be predicted by the Hall-Petch relationship. In the same way the smaller mean free slip paths reduce the magnitude of local plastic strains and make crack initiation more difficult which gives improved LCF.

Ductilities are also improved because of the refined microstructure. Void formation takes place at the primary alpha - transformed beta interfaces and these same interfaces are easy crack paths. Numerous but very small interfaces are therefore required for good ductility. Grain boundary alpha, alpha stringers and long alpha plates allow early void link up and reduce ductilities. In the same way acicular structures have lower ductilities because the long α/β interfaces act as easy paths for void growth. (4, 5)

The duplex structures are however lower in fracture toughness than the acicular structures, predominantly because the refined structure leads to a more direct crack path. This crack path can be made more tortuous by going to acicular structures (6). For good fracture toughness the platelets need to be thick enough to turn a crack and short enough and sufficiently closely spaced to cause frequent changes in crack direction. Small prior beta grain sizes are therefore preferred. Interplatelet β can also help by blunting the crack. In commercial practice duplex microstructures usually have a fracture toughness (K_{IC}) in the range 40 - 60 $MNm^{-3/2}$ while the acicular microstructures are above 60 $MNm^{-3/2}$.

Fatigue crack growth rates are lower in the acicular than duplex microstructures as the acicular morphology is more conducive to crack arrest and branching. Variations in the acicular microstructures, however, do themselves lead to variations in crack growth rates, especially at low temperatures. Aligned alpha platelet colonies can be prone to slip banding across the colonies which cause higher local stress concentrations which can accelerate fracture, especially if environmental factors assist. For example, the deleterious low temperature 'dwell on load' fatigue effect is currently thought to operate by a stress corrosion cracking type mechanism along such slip bands. Basketweave or very small colonies are therefore preferred as they exhibit more predictable behaviour.

(b) Alpha + Beta Alloys

The alpha + beta alloys are generally used for low temperature applications. Ti 6Al 4V is the most commonly used alloy having modest quantities of both alpha stabiliser (aluminium) and beta stabiliser (vanadium) and thereby combining reasonable strength with a good forgeability. For most applications, Ti 6Al 4V is used in the alpha + beta processed condition followed by a simple anneal. If necessary, however, additional strengths can be obtained by an alpha + beta heat treatment. This simple and versatile alloy has also been used for castings, powder products and beta processed products in an attempt to optimise the metal usage, although such processing techniques have as yet been only partially successful.

The higher strength alpha + beta alloys use greater quantities of alpha and beta stabiliser, alpha + beta processing and full heat treatments to obtain their properties, IMI 550 (Ti-4Al-4Mo-2Sn-Si) and Ti 662 (Ti-6Al-6V-2Sn) being typical examples. Still stronger alloys such as Ti 6246 (Ti-6Al-2Sn-4Zr-6Mo) have been developed but are not widely used because of their low fracture toughness.

Generally the alpha + beta alloys have good LCF properties which are best achieved in a quenched fine grained alpha + beta structure with a relatively high proportion of primary alpha. Fracture toughness, on the other hand is improved by raising the solution treatment temperature to give a microstructure with more coarse Widmanstatten transformed beta. A compromise between the two is achieved by controlled heat treatment. (2)

There are, however, considerable difficulties in obtaining good LCF in large scale forgings with alpha + beta alloys. Firstly, it is difficult to achieve the necessary refinement with an economical forging route and secondly the alloys have inherently low hardenability. Some improvements in the fatigue strength of large scale Ti 6Al 4V forgings have been achieved by beta processing and beta heat treatment, presumably by slowing the crack propagation rate. (7). Also an increase in fracture toughness, accompanies this processing and low fracture toughness can be another limitation of heat treated alpha + beta alloys. Solutions to these limitations may be offered by the metastable beta alloys which are discussed below.

(c) Metastable Beta Alloys

Metastable beta alloys offer good ductility in the solution annealed condition and high strength, high toughness and deep hardenability in the aged conditions.

The first alloy of this type and the most frequently used is B120 VCA (Ti-13V-11Cr-3Al). It is more formable than earlier beta alloys and can be heat treated to high strengths. The vanadium and chromium stabilise the beta whereas the aluminium segregates to and improves the strength of, the alpha precipitates. The high chromium content, however, leads to segregation in the ingot which in turn can cause a variable aging response and an embrittling chromide phase. (8)

The newer alloys, Beta III (Ti-11½Mo-6Zr-4½Sn), Beta C (Ti-3Al-8V-6Cr-4Mo-4Zr), Ti-8Mo-8V-2Fe-3Al and Ti-10V-2Fe-3Al contain either no, or reduced amounts of, compound forming additions.

All these beta alloys exhibit good strength/toughness combinations in large scale forgings. They also have potential as high strength/high toughness sheet materials. The alloys, however, do have deficiencies; they exhibit low aged tensile ductility, particularly in the transverse direction, in coarse grained components. They also have higher densities, higher costs, less oxidation resistance, lower moduli and are non-weldable. In addition they have a low temperature capability and this coupled with concern about their low temperature stability has restricted their use. The alloys have found applications in the fastener market but as yet have no significant applications as either forgings or sheet.

Summary

The choice of material for a particular application involves the critical appraisal of such criteria as:

- economic factors,
- previous experience,
- temperatures,
- stresses,
- type of construction,
- environmental conditions,
- lifing philosophy.

For high temperature applications a fine grained transformed β near α alloy offers a weldable material with good creep resistance, fracture toughness and lower crack propagation rates than the $\alpha + \beta$ alloys. However, at low stress levels, a case can be made for $\alpha + \beta$ alloys since their reserves of ductility allow greater instability losses.

For low temperature applications and small section sizes ($<$4") $\alpha + \beta$ processed alloys (both near α and $\alpha + \beta$) are preferred because of their better LCF, tensile strengths and ductilities. However, if a lifing philosophy is developed based on fracture mechanics then a strong case can be made for beta heat-treated alloys since these materials have low crack propagation rates at low temperatures and would therefore be more defect tolerant.

References

(1) Rosenberg - The Science, Technology and Application of Titanium Jaffee & Promisel, eds. 1970 p851

(2) Morton - Phil. Trans. R. Soc. Lond. vol 282 p401

(3) Neal - Proceedings of the Joint Conference on the Forging and Properties of Aerospace Materials. University of Leeds, 1977.

(4) Greenfield and Margolin - Met. Trans. 1972 vol 3 - 2649

(5) Eylon, Hall, Pierce and Ruckle - Met. Trans. 1976 vol 7a - 1817

(6) Chesnutt, Rhodes and Williams - Proceedings of Fractography -microscopic cracking processes symposium. Montreal 1975 ASTM publication STP 600 p99.

(7) Lucas - Titanium Science and Technology - Proceedings of the second international conference vol 4 - 2081

(8) Froes, Malone, Williams, Greenfield and Hirth - Proceedings of the Joint Conference on the Forging and Properties of Aerospace Materials, University of Leeds, 1977.

Table I Alloy Compositions

IMI 679	Ti - 11Sn - 2¼Al - 5Zr - 1Mo - 0.25Si
Ti 6242	Ti - 6Al - 2Sn - 4Zr - 2Mo - 0.1Si
IMI 685	Ti - 6Al - 5Zr - 0.5Mo - 0.3Si
IMI 829	Ti - 5½Al - 3½Sn - 3Zr - 1Nb - 0.3Mo - 0.3Si
Ti 6/4	Ti - 6Al - 4V
Ti 662	Ti - 6Al - 6V - 2Sn
IMI 550	Ti - 4Al - 4Mo - 2Sn - 0.5Si
Ti 6246	Ti - 6Al - 2Sn - 4Zr - 6Mo
B120 VCA	Ti - 13V - 11Cr - 3Al
BETA III	Ti - 11½Mo - 6Zr - 4½Sn
Ti 10-2-3	Ti - 10V - 2Fe - 3Al

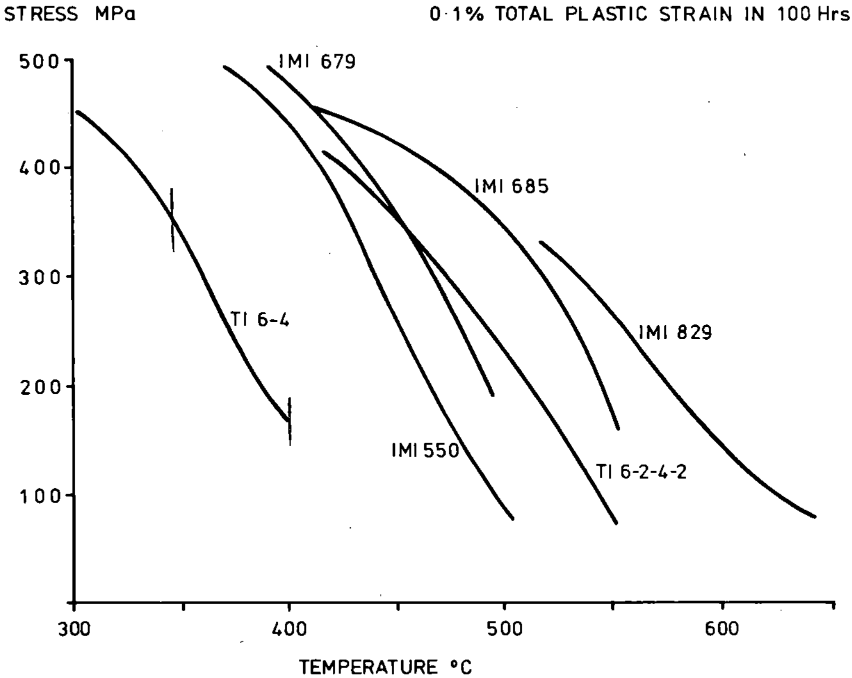

FIG. 1 MINIMUM CREEP STRAIN PROPERTIES FOR VARIOUS TITANIUM ALLOYS

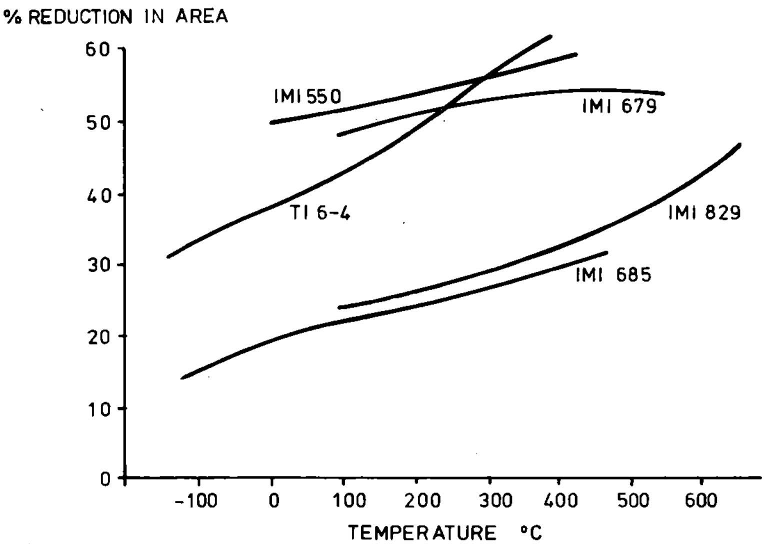

FIG. 2 EFFECT OF TEMPERATURE ON THE % REDUCTION IN AREA (DUCTILITY) FOR VARIOUS TITANIUM ALLOYS

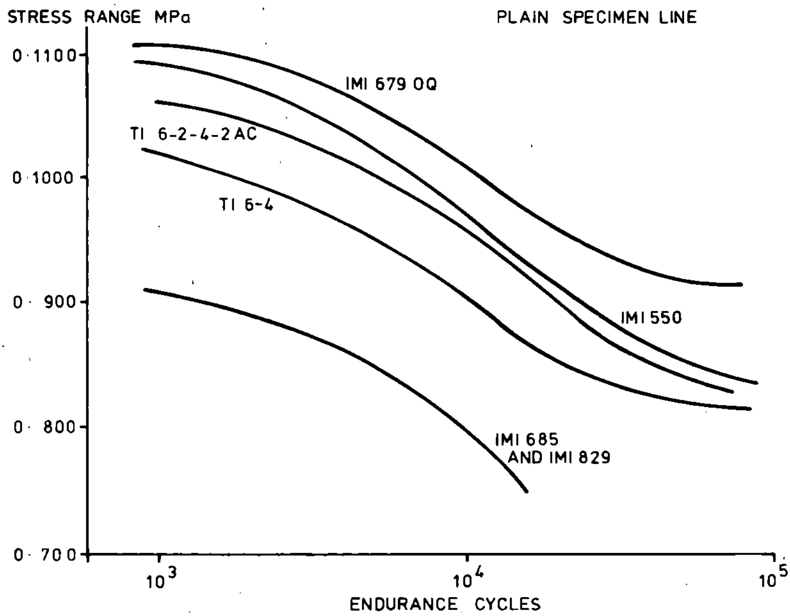

FIG. 3 TYPICAL LCF PROPERTIES FOR VARIOUS TITANIUM ALLOYS AT 20°C

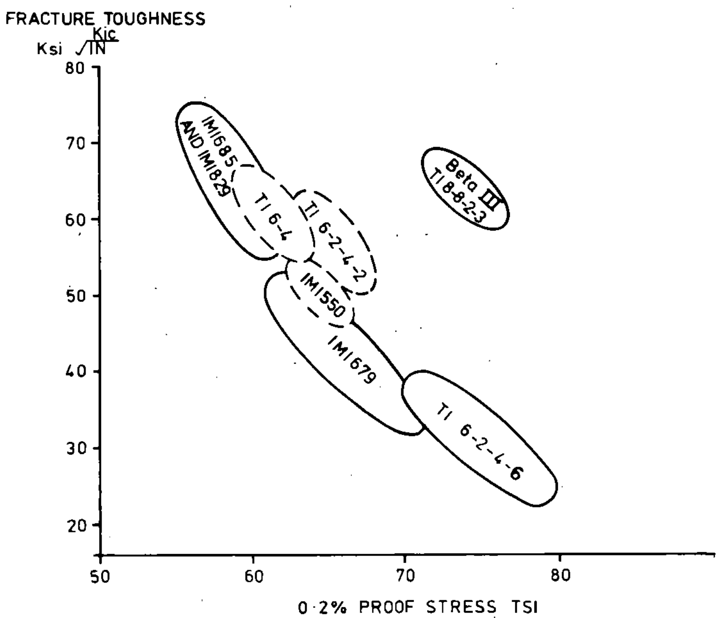

FIG. 4 COMPARISON OF FRACTURE TOUGHNESS VERSUS STRENGTH FOR VARIOUS ALLOY TYPES

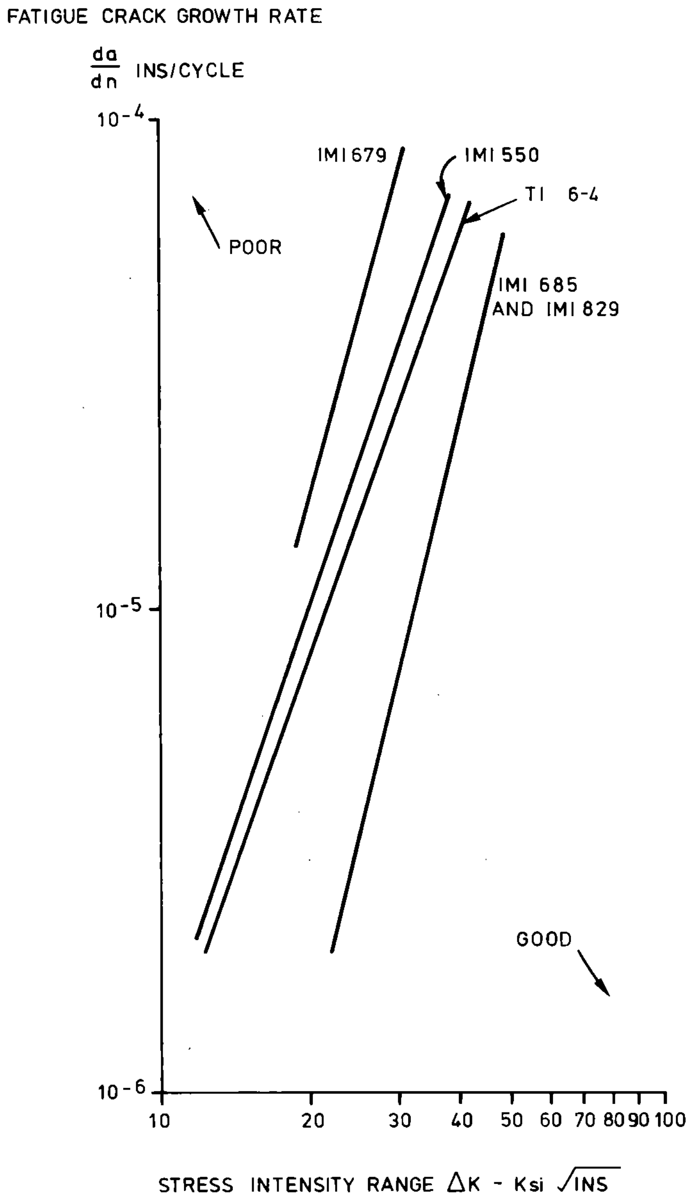

FIG. 5 COMPARISON OF THE CRACK PROPAGATION RATES FOR VARIOUS TITANIUM ALLOYS AT 20°C