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REVIEW PAPER

High-temperature Titanium Alloys

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ABSTRACT

The development of high-temperature titanium alloys has contributed significantly to the spectacular progress in thrust-to-weight ratio of the aero gas turbines. This paper presents an overview on the development of high-temperature titanium alloys used in aero engines and potential futuristic materials based on titanium aluminides and composites. The role of alloy chemistry, processing, and microstructure, in determining the mechanical properties of titanium alloys is discussed. While phase equilibria and microstructural stability consideration have restricted the use of conventional titanium alloys up to about 600 °C, alloys based on Ti_3Al (α_2), Ti_2AlNb (O), $TiAl$ (γ), and titanium/titanium aluminides-based composites offer a possibility of quantum jump in the temperature capability of titanium alloys.

Keywords: High temperature materials, aero engines, high-temperature titanium alloys, titanium aluminides, composites, titanium/titanium aluminides-based composites, Intermetallic alloys, metal-matrix composites

1. INTRODUCTION

Titanium has exceptional properties of strength, specific strength, low-thermal expansion coupled with low modulus in relation to its high-low cycle fatigue strength, corrosion resistance, workability, and weldability. No other material approaches this combination of engineering properties over a temperature range spanning from ambient temperature to about 550 °C. Accordingly, this single material has been dominant in aero engine application in the compressor stages. It is noted that the aluminium and steel used for the earliest aero engines have been superceded almost entirely in today's advanced military and civil aircraft engines by titanium-and-nickel-base alloys. The use of titanium alloys in the present-day aero engines has made it possible to bring out further improvements in aerodynamic component, cycle and propulsive efficiency coupled with improvements in thrust-to-weight ratio. The popularity of this material for aero engine design can be gauged from the fact that titanium content[1] has been increased from 3 per cent to 31 per cent of the aero engine weight since 1950 (Fig. 1) and contrary to some expectations, the trend is likely to continue. The predictions for material requirements in the aero engine into the new century[2] are shown in Fig. 2. One of the most interesting features of the prediction, compared with a similar[3] one made in 1983, is the significantly increased demand for titanium alloys, mainly at the expense of earlier optimistic predictions for ceramic and metal-matrix composites.

The driving force for the use of titanium in aerospace remains one of weight reduction, but of

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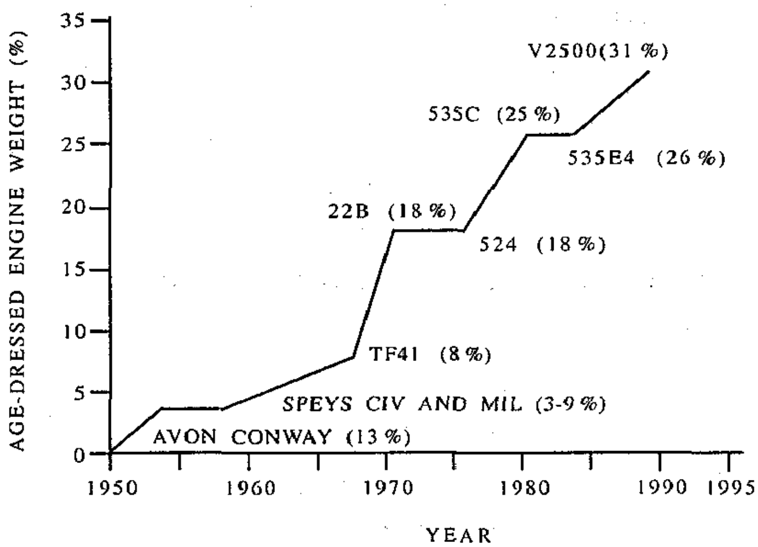


Figure 1. Titanium content–aero engines

equal importance to all sectors of the industry is the need to reduce costs. The introduction of castings, superplastic forming (SPF) and superplastic forming with diffusion bonding (SPF/DB), as near-to-net shape technologies, in the manufacture of titanium alloy components had a profound effect on the increased use of titanium, with resulting technical and cost-saving benefits, over alternative designs in competitive materials.

In aero engines, one of the most interesting development is to replace the nickel-base alloys at 550 °C–700 °C or conventional disc/blade structur with blings. In airframes, titanium becomes competiti and necessary, in relation to aluminium, at supersor and hypersonic speeds (eg, SR71) as skin temperatu exceeds 250 °C.

Aerospace being the major market, the titaniu alloys that have attracted the maximum attenti have been those targeted towards high-temperatu applications. The area can be split into two distir groups. The immediate and medium-term requiremen which are addressed by the conventional allo

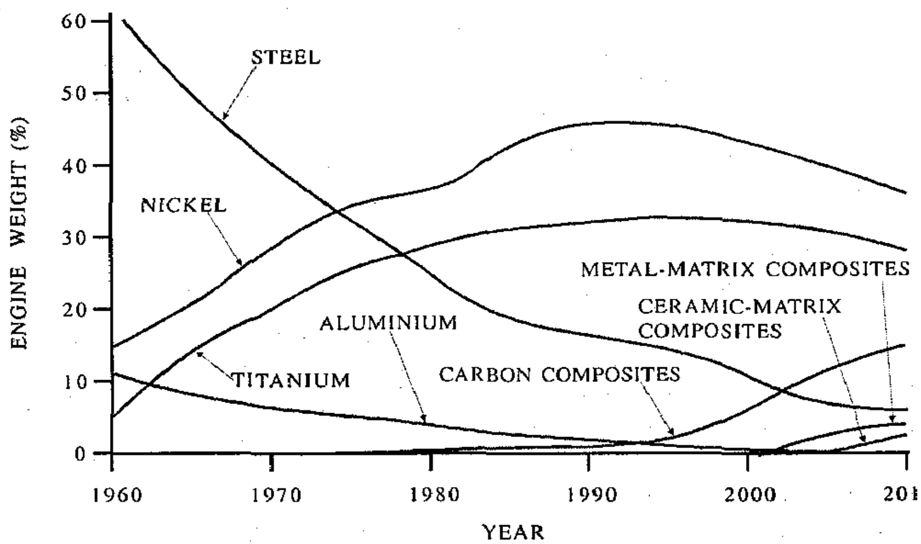


Figure 2. Evolution of materials used in aero gas turbines

Table 1. Temperature range and chemical composition of conventional titanium alloys

Alloy designation	Country of origin/year	Max useful temperature (°C)	Chemical composition (Wt %)								
			Aluminium	Tin	Zirconium	Molybdenum	Niobium	Vanadium	Silicon	Others	$[Al]_{Eq}$*
IMI 318** (*Ti*-64)	UK (1954)	300	6.0	--	--	--	--	4	--	--	6.0
IMI 550	UK (1956)	425	4.0	2.0	--	4.0	--	--	0.50	--	4.7
Ti-811	USA (1961)	400	8.0	--	--	1.0	--	1	--	--	8.0
IMI 679	UK (1961)	450	2.0	11.0	5.0	1.0	--	--	0.20	--	6.8
Ti-6246	USA (1966)	450	6.0	2.0	4.0	6.0	--	--	--	--	7.4
Ti-6242	USA (1967)	450	6.0	2.0	4.0	2.0	--	--	--	--	7.4
IMI 685**	UK (1969)	520	6.0	--	5.0	0.5	--	--	0.50	--	6.8
Ti-6242S	USA (1974)	520	6.0	2.0	4.0	2.0	--	--	0.10	--	7.4
IMI 829	UK (1976)	580	5.5	3.5	3.0	0.3	1.0	--	0.30	--	7.2
IMI 834	UK (1984)	590	5.5	4.0	4.0	0.3	1.0	--	0.50	0.06 *C*	8.1
Ti-1100	USA --	600	6.0	2.8	4.0	0.4	--	--	0.40	--	8.2
VT 3-1	CIS --	--	6.5	--	--	2.7	--	--	0.25	1.60 *Cr*	6.5
VT 9**	CIS --	520	6.5	--	2.0	3.5	--	--	0.25	--	6.8
VT 18	CIS --	--	6.0	--	11.0	1.0	1.0	--	0.10	--	7.8
VT 18Y	CIS --	550	6.9	2.7	4.1	0.9	1.0	--	0.15	--	8.5
Ti-633G	China --	--	5.5	3.0	3.0	0.3	1.0	--	0.30	0.20 *Gd*	7.0
Ti-55	China --	--	5.5	4.0	2.0	1.0	--	--	0.25	1 *Nd*	7.2
DAT 54	Japan --	550	5.8	4.0	3.5	2.8	0.7	--	0.40	0.06 *C*	8.3

* $[Al]_{Eq}$ shown here does not include *O*, *N*, *C* content.

** Designation of equivalent Indian alloys produced by the Midhani (Mishra Dhatu Nigam Ltd, Hyderabad) are: IMI 318-Titan 31; IMI 685-Titan 26A; VT9-GTM900.

like IMI 834 and *Ti*-1100, and the longer-term objectives, which are mainly focused on intermetallics systems based on titanium aluminides (Ti_3Al and $TiAl$) and titanium and intermetallic-based composites.

2. CONVENTIONAL HIGH-TEMPERATURE ALLOYS

In titanium, alloying elements tend to stabilise either the low temperature, close-packed hexagonal alpha (α) phase, or the high-temperature allotrope, body-centred cubic beta (β) phase. Titanium alloys for aerospace applications contain both the α and β stabilising elements in various proportions, depending on the application, and therefore, the required mechanical properties. To fully optimise the mechanical properties, titanium alloys are worked to control the microstructure –the shape, size, and distribution of both the α and β phases.

2.1 Compositional Effects

Titanium alloys developed for high-temperature applications have been presented in Table 1. All these alloys have variation on the same theme, starting with the $\alpha+\beta$ *Ti*-6*Al*-4V alloy. However, the more heat-resistant alloys contain much less β phase than *Ti*-6*Al*-4*V* and are generally referred to as near-α alloys. The α phase, by virtue of its close-packed structure, exhibits far lower diffusivities than the body-centred cubic β phase, and thus, forms a major constituent (> 95 %) in high-temperature titanium alloys.

The α phase is solid solution strengthened with elements which either stabilise it (such as aluminium) or are neutral (such as zirconium and tin) to α phase. The addition of aluminium (*Al*) increases tensile strength and creep strength, and moduli while reducing alloy density[4]. Tin (*Sn*) is used as a solid solution strengthener, often in conjunction with aluminium to achieve the higher strength without embrittlement. Zirconium (*Zr*) forms a continuous solid solution with titanium and increases strength at low and intermediate temperatures.

However, the use of zirconium above 5-6 per cent may reduce ductility and creep strength[5]. The oxygen content is usually kept fairly low (0.1–0.5 %) in this class of alloys. Although it strengthens titanium, the beneficial effects deteriorate at temperatures above 300 °C. Increased oxygen content also tends to lower ductility, toughness, and long-term high-temperature stability[6].

However, a limit exists to these strengthening additions beyond which the α phase decomposes to form fine precipitates of an embrittling phase Ti_3Al (α_2). Rosenberg[7] empirically arrived at an upper-limit value for the α equivalent (aluminium equivalent) content of the alloy below which Ti_3Al (α_2) phase is not likely to form.

$$Al_{eq} = Al + 1/3\ Sn + 1/6\ Zr + 10(O+C+2N) \leq 9$$

Based on the recent work, gallium is included in terms of gallium/2. All commercial alloys presently in service still meet this requirement (Table 1).

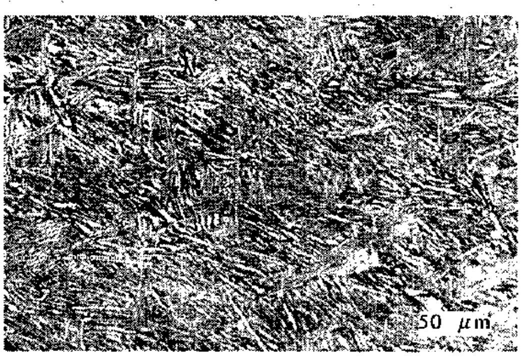
The level of β stabilisers is very low–sufficient to confer some microstructural strengthening without endangering metallurgical stability. It is notable that the level of heat resistance goes up as the β stabiliser content decreases in proceeding from *Ti*-6242 to IMI 834 and TIMETAL1100. Molybdenum (*Mo*) has been used as the prime β stabiliser in these alloys as it increases heat-treatment response of the alloys[8]. Molybdenum also increases short-term high-temperature strength and alloys aimed at short-term high-temperature strength (such as *Ti*-6246) have greater molybdenum content. Another β stabiliser, niobium (*Nb*) is added to improve the surface stability at high temperature. Also, substituting niobium for some of the molybdenum provides β phase strengthening with the least possible lowering of the β transus[9] in IMI 829 and IMI 834.

Silicon (*Si*) is an important element in high-temperature titanium alloys since it increases strength at all temperatures and has a marked beneficial effect on creep resistance[10,11]. The silicon content was usually restricted to 0.1–0.2 per cent in alloys of US origin, while a higher silicon content was used in alloys of UK and Russian origin. However, the recent alloys IMI 834 (UK) and *Ti*-1100 (US) have higher silicon contents, up to 0.5 per cent.

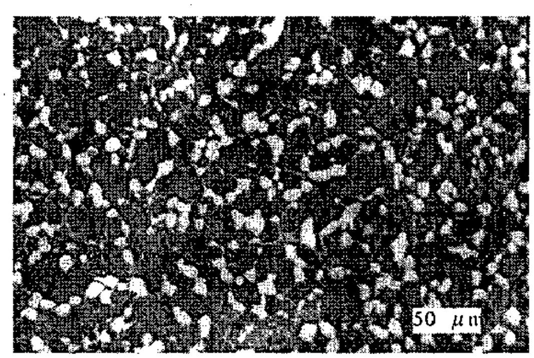
The role of trace elements such as iron, nickle, Chromium (*Cr*) and cobalt (*Co*) in the degradation of creep properties has now been well-recognised[12-15] and many aero engine manufacturers and alloy producers have restricted these elements to less than 0.01 per cent for creep-specific applications.

2.2 Microstructural Effects

In near-α phase and $\alpha+\beta$ phase titanium alloys, the creep strength is increased by the heat treatment or processing the material above the β transus temperature. On cooling, this results in a lenticular α structure [Fig. 3(a)]. In contrast, the equiaxed α structure [Fig. 3(b)], resulting from both hot working and subsequent heat treatment in the $\alpha+\beta$



(a)



(b)

Figure 3. Microstructural effects: (a) β heat-treated structure showing acicular α in alloy GTM 900 and (b) $\alpha+\beta$ -processed and $\alpha+\beta$ heat-treated structure showing equiaxed α in alloy GTM 900.

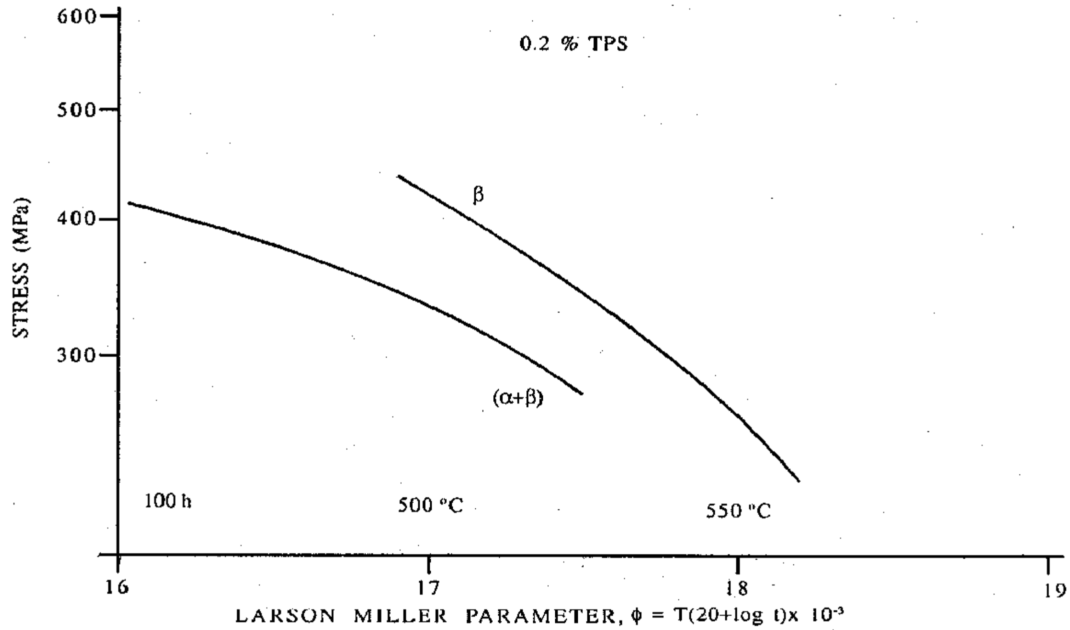


Figure 4. Comparison of creep data of $\alpha+\beta$ and β-processed IMI 685 (Larson Miller plot)

phase field, exhibits the lower creep resistance[16,17] An example of this is shown in Fig. 4 for alloy[17] IMI 685. For β-processed material, the creep strength depends on the morphology of the α phase, which can be controlled by cooling rate from the β heat-treatment temperature. The effect of cooling rate on creep strain[18] is shown in Fig. 5.

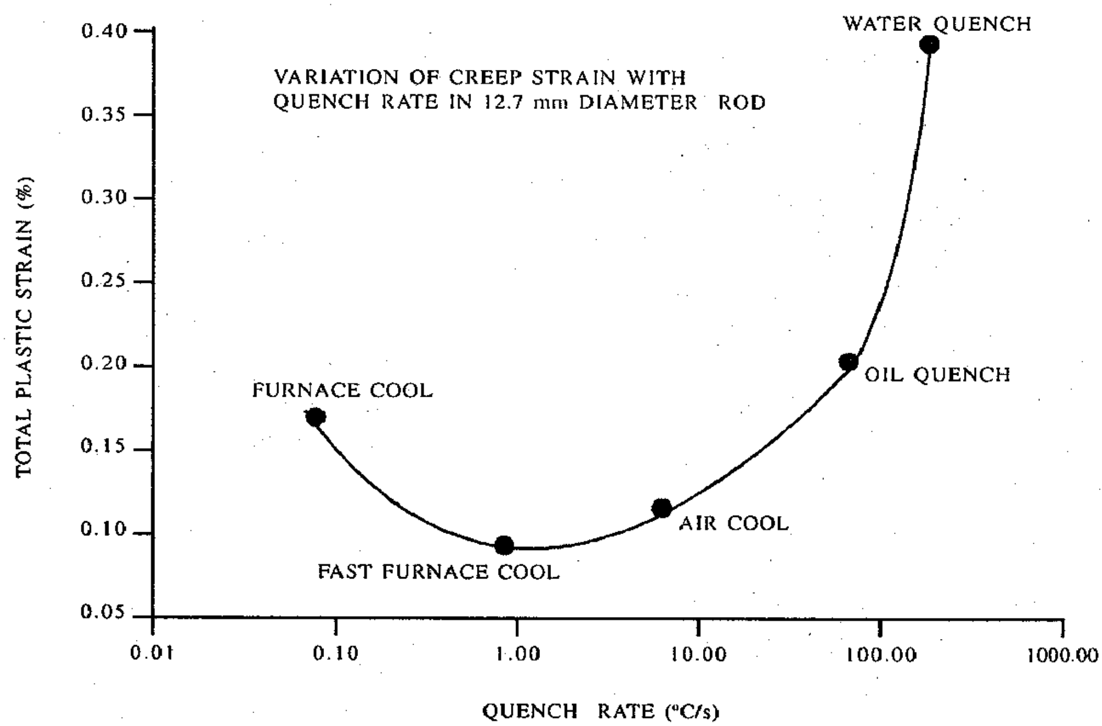


Figure 5. Effect of cooling rate on the creep strain in IMI 685

The β heat-treated structures also show superior fracture toughness and fatigue-crack growth resistance[19,20]. On the other hand, $\alpha+\beta$ treated equiaxed structures exhibit higher ductility and low-cycle fatigue properties. The alloys IMI 685 and IMI 829 used in many current European aero engines such as RB2111, 535E4 are both fully β heat-treated to maximise creep resistance. However, the need to optimise both creep and fatigue is well-recognised now and a recent alloy such as IMI 834 is used in $\alpha+\beta$ heat-treated condition with a 5-15 per cent of equiaxed α in the structure. Close control of equiaxed α fraction in this alloy has been made possible by widening the $\alpha+\beta$ phase field by the carbon addition[21]. The alloy is used in RR Trent800 in the last two stages of the LP compressor and the first four stages of HP compressor. The *Ti*-1100, a competitive alloy to IMI 834, is designed to be used in the β heat-treated condition. The composition of *Ti*-1100 has been optimised to take the maximum advantage of its creep potential without shortcomings seen in the earlier alloys of the same type. The alloy is under investigation for the Allison gas turbine engine T406/GMA3007/ GMA2100 family of engines, primarily for castings[22].

The current temperature limit of the near-α alloy class is 590 °C for IMI 834, while efforts are continuing for enhancing the temperature capability of these alloys by adding[23,24] tantalum (*Ta*), and bismuth (*Bi*), but no significant improvements have been achieved. The temperature capability of titanium alloys, relative to its melting point (Tm), is surprisingly low (0.4 Tm as compared to 0.8 Tm in nickle-base alloys). This is mainly due to the fact that the metallurgical stability is governed by $\beta \leftrightarrow \alpha+\beta$ transformation rather than by melting. Surface oxidation above 600 °C and consequent embrittlement as well as burn resistance are other limiting factors for the use of titanium alloys at higher temperatures.

The possible approaches to increase the temperature capability of titanium alloys are:

- To strengthen the conventional titanium alloys in a different manner such as rare earth dispersoids
- To develop alloys based on a fine dispersion of ordered precipitates in α matrix or in an $\alpha+\beta$ matrix
- To develop alloys based on intermetallics Ti_3Al (α_2) and $TiAl$ (γ)
- To develop composites based on titanium and/ or intermetallic alloys
- To develop coatings and inherently oxidation-resistant alloys.

Alloys with dispersoids have been explored both by conventional melting[24,25] and by rapid solidification processing[26,27]. Several rare earth additions such as yttrium (*Y*), erbium (*Er*), neodymium (*Nd*) and gadolinium (*Gd*) and elements such as carbon (*C*) and boron (*B*) have been investigated. However, control of distribution of dispersoids and their stability are problematic and the alloys have showed only limited promise.

Alloys based on $\alpha+\alpha_2$ phase have been studied[28] but have not shown the expected results. Alloys containing dispersion of silicides (Ti_5Si_3) have been developed in Ukraine[29] for applications up to 800 °C in turbocharger turbines and internal combustion engine components. However, concern about poor room temperature ductility (0.4 % elongation at room temperature) and fracture toughness (18-20 MPa$\sqrt{m}$) preclude their widespread use.

The most promising research has been the development of intermetallic-based alloys.

3. INTERMETALLIC ALLOYS

Titanium aluminides (Ti_3Al and $TiAl$) are the attractive structural materials for the aerospace industry due to their low density, high specific strength and modulus retention, and excellent creep resistance. The potential of the aluminides can be seen in Table 2, which compares these with the conventional titanium alloys and superalloys. However, extreme brittleness of these intermetallics made their use impractical, if not impossible. Extensive alloy development efforts to ameliorate their shortcomings have brought intermetallic alloys very close to the use, and several components of Ti_3Al and $TiAl$-based alloys are under evaluation for the commercial use.

Table 2. Properties of titanium alloys, titanium aluminides, and superalloys

Property	*Ti*-base	Ti_3Al base	*TiAl* base	Superalloys
Structure	hcp/bcc	DO_{19}	$L1_0$	fcc / $L1_2$
Density (g/cc)	4.5	4.1-4.7	3.7-3.9	7.9-8.5
Modulus (GPa)	95-115	110-145	160-180	206
Yield strength (MPa)	380-1150	700-990	350-600	800-1200
Tensile strength (MPa)	480-1200	800-1140	440-700	1250-1450
RT ductility (%)	10-25	2-10	1-4	3-25
HT ductility (% / °C)	12-50	10-20 / 660	10-60 / 870	20-80 / 870
RT fracture toughness (MPa $\sqrt{m}$)	12-50	13-30	12-35	30-100
Creep limit (°C)	600	750	750-950	800-1090
Oxidation limit (°C)	600	650	800*- 950**	870* - 1090**

* Uncoated, ** Coated/actively cooled

3.1 Titanium-aluminides Alloys

Early efforts at alloy development[30-33] were primarily directed towards improving the ductility, strength, and toughness. Niobium was established as the primary alloying addition and was found to impart ductility by modifying slip behaviour[34-36] in Ti_3Al (α_2), and stabilising a relatively ductile ordered B2 phase into the structure[37,38]. At high enough levels (>12 atom %), it also led to distortion of the Ti_3Al (α_2) phase to an orthorhombic structure, the *O* phase[39]. While early alloys such as *Ti*-24*Al*-11*Nb* and *Ti*-25*Al*-10*Nb*-3*V*-1*Mo* were based on the α_2 phase, alloy development in later years focussed on the *O* phase. In fact, presence of several phase transformations in *Ti-Al-Nb* system, offers vast scope for designing the alloys with varying microstructures and properties.

Several *Ti-Al-Nb* alloys based on α_2+B2/*O*+B2 have been investigated over the years, mainly in the US and India. The effect of quarternary additions like molybdenum, vanadium, tantalum, zirconium, and silicon has also been explored. The microstructure and mechanical properties of these alloys have been presented in several review papers[40-42]. One of the most remarkable features of α_2/*O*-based alloys is the similarity in their thermomechanical processing-structure-property relationships to those obtained in conventional $\alpha+\beta$ alloys. Processing of these alloys above the β transus results in acicular morphology of α_2/*O* phase while a subtransus processing leads to the equiaxed morphology[43,44]. The arrangement of α_2/*O* laths, on cooling from the β heat-treatment temperatures changes from basketweave to colony like, as in the conventional titanium alloys. However, much more complex microstructures can be generated in these alloys due to (i) retention of B2 phase on quenching, and subsequent decomposition to α_2/*O* laths in a variety of transformations, and (ii) $\alpha_2 \rightarrow O$ transformations[45,46].

A survey of mechanical properties and microstructure in these alloys suggest the following trends:

(a) In equiaxed microstructures containing α_2+B2 phase or *O*+B2 phase, the strength, ductility, and fracture toughness of the alloys increase with increasing B2 volume fraction (Fig. 6). The ductility and toughness of the B2 phase increase with decreasing aluminium as well as grain size[47,48]. The B2, in these microstructures, is unstable and decomposes on the lower temperature ageing. Increasing the ageing temperature results in decreasing strength and increasing ductility for a given equiaxed α_2/*O* fraction (Fig. 7).

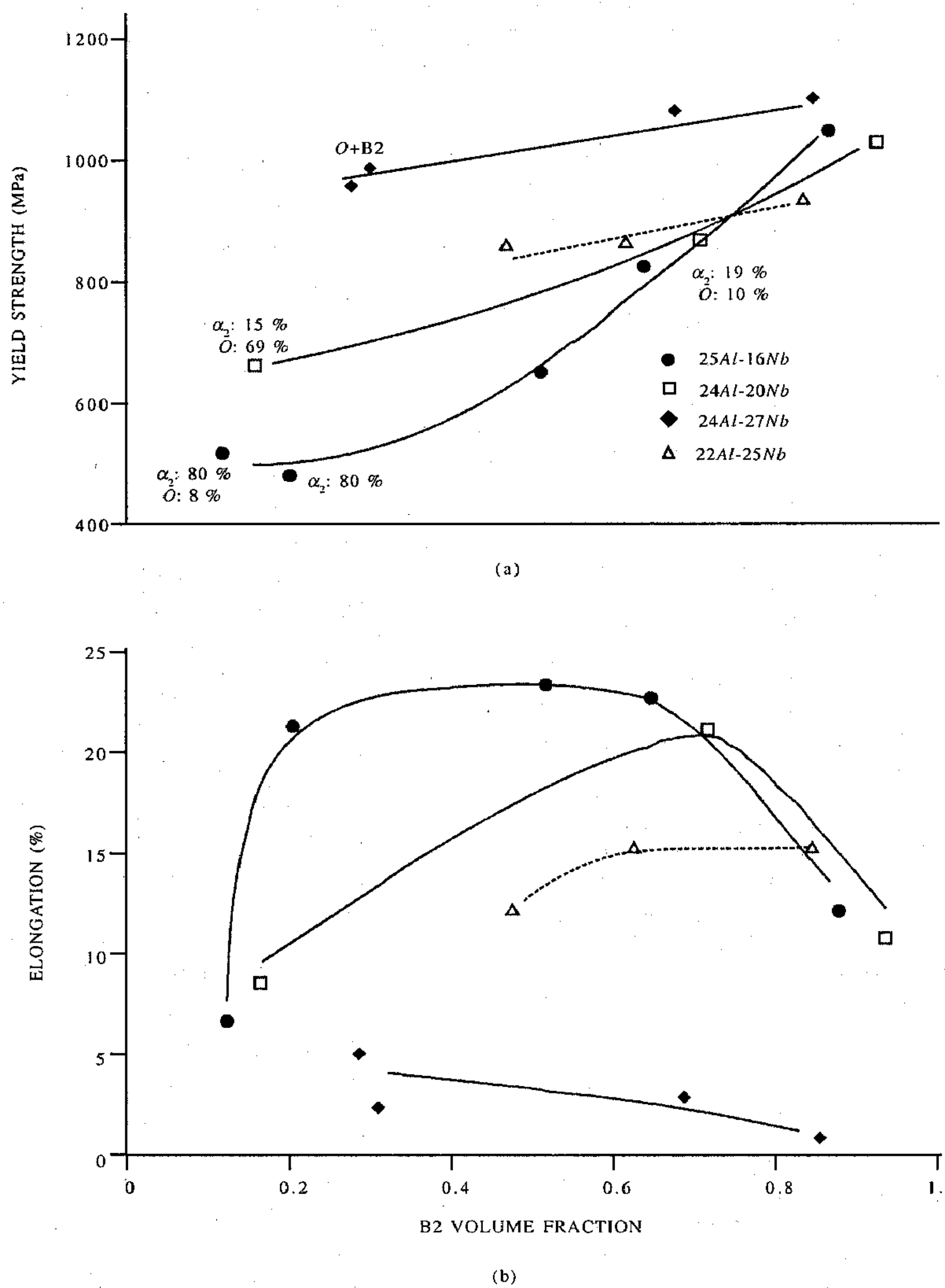


Figure 6. In equiaxed microstructures: (a) yield strength and (b) elongation of orthorhombic alloys as a function of B2 volum fraction.

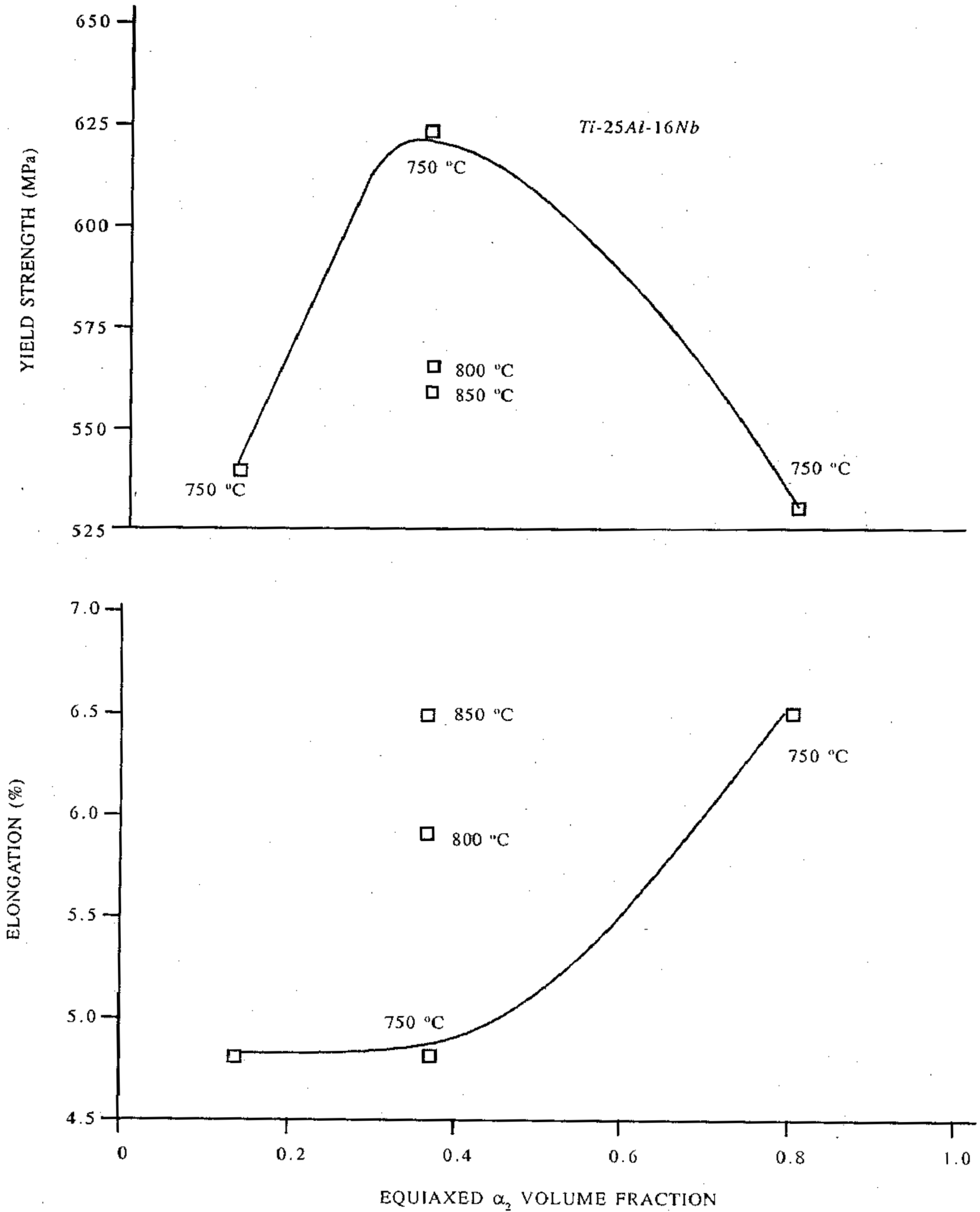


Figure 7. Effect of equiaxed α_2/O phase and ageing temperature on tensile properties of orthorhombic alloys

(b) In β heat-treated structures, the strength and ductility are strongly influenced by the cooling rate. Strength is higher at the higher cooling rate due to structural refinement, while ductility maximises at some intermediate cooling rate (Fig. 8). Increase in niobium results in increased strengthening.

(c) As in conventional titanium alloys, lath structures with α_2/O are superior in creep to structures

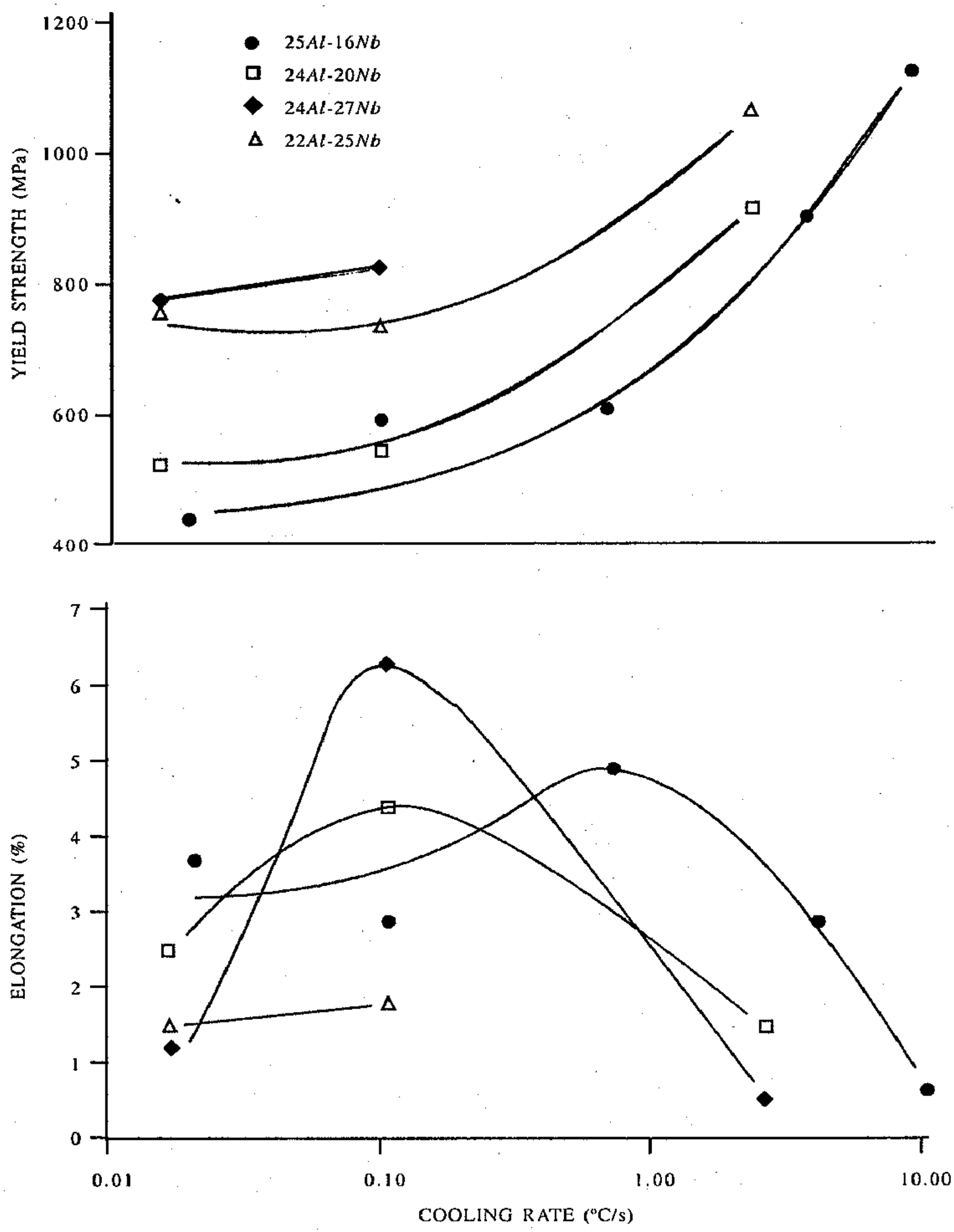
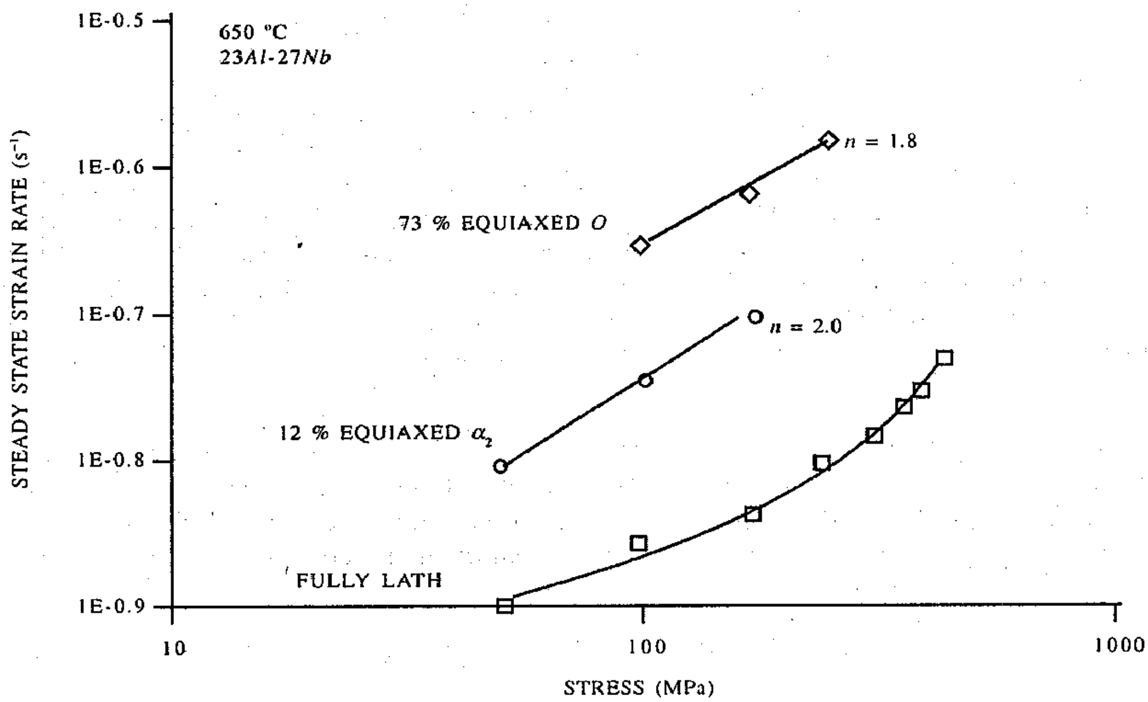


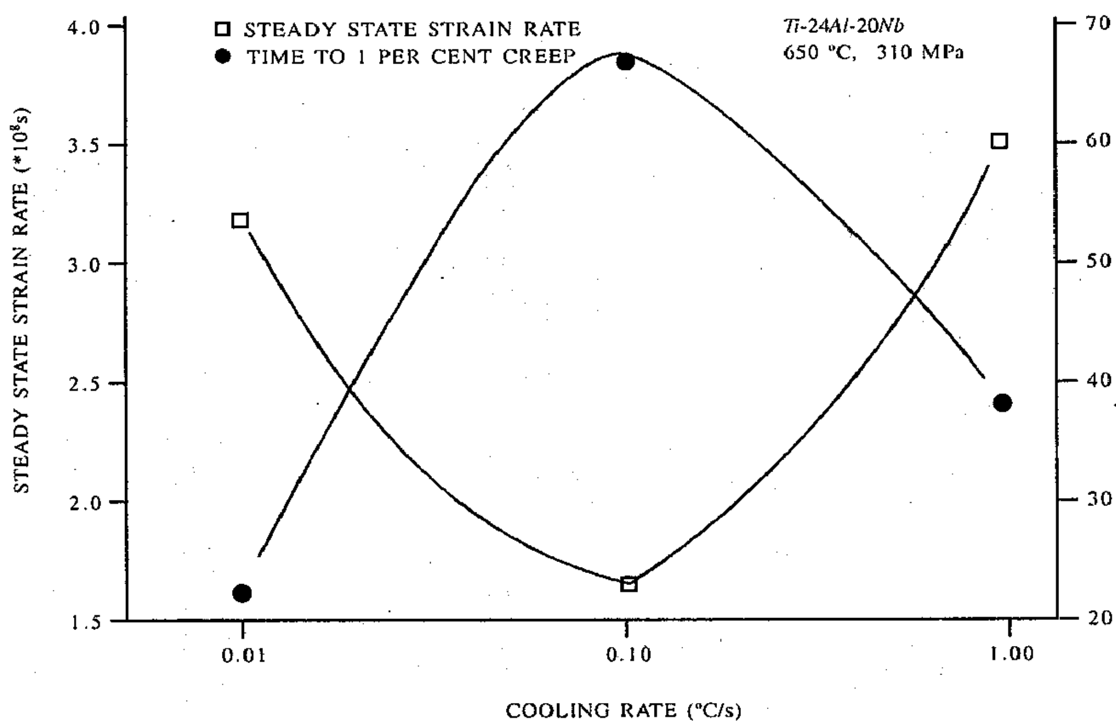
Figure 8. Effect of cooling rate on the tensile properties of β-treated O phase alloys

containing equiaxed α_2/O alloys. Also, in a manner similar to near-α alloys, creep performance is maximised at intermediate cooling rate from β heat treatment (Fig. 9).

(d) Aluminium increases creep resistance but reduc fracture toughness and aluminium content great than 25 atom per cent are unacceptably detriment to the fracture toughness.



(a)



(b)

Figure 9. (a) Effect of volume fraction of equiaxed α_2/O on creep in a *Ti-23Al-27Nb* alloy and (b) effect of cooling rate on creep of *Ti-24Al-20Nb* alloy in β treated structures.

(e) Molybdenum addition of about 1 per cent is desirable for high-temperature strength. Zirconium and silicon are effective in increasing creep resistance[49,50].

Based on the above, most of the alloys under investigation are complex alloys based on *Ti-Al-Nb* and containing one or more of the elements such as molybdenum, zirconium, silicon, tantalum. In addition, the effect of alternative processing techniques such as extrusion, isothermal forgings are under investigation*.

A comparison of the mechanical properties of α_2/O alloys developed at the Defence Metallurgical Research Laboratory (DMRL), Hyderabad, and the most advanced conventional titanium alloy IMI 834 and a nickel-base alloy INCO 718 is presented in Fig. 10.

No real problem with processing by the conventional technologies exist. In fact, the alloys have been shown to exhibit superplasticity like conventional alloys[51,52] and can also be superplastically formed. Figure 11 shows compressor blades of a *Ti*-24*Al*-15*Nb* alloy for an experimental gas turbine GTX engine produced by the conventional close-die forging route at the Hindustan Aeronautics Ltd (HAL), Bangalore.

With all their attractive properties, the α_2 and *O*-based alloys still present significant challenges. The characteristics of fatigue-crack growth resistance and fracture toughness of this class of alloys bring these up short of the conventional alloys relative

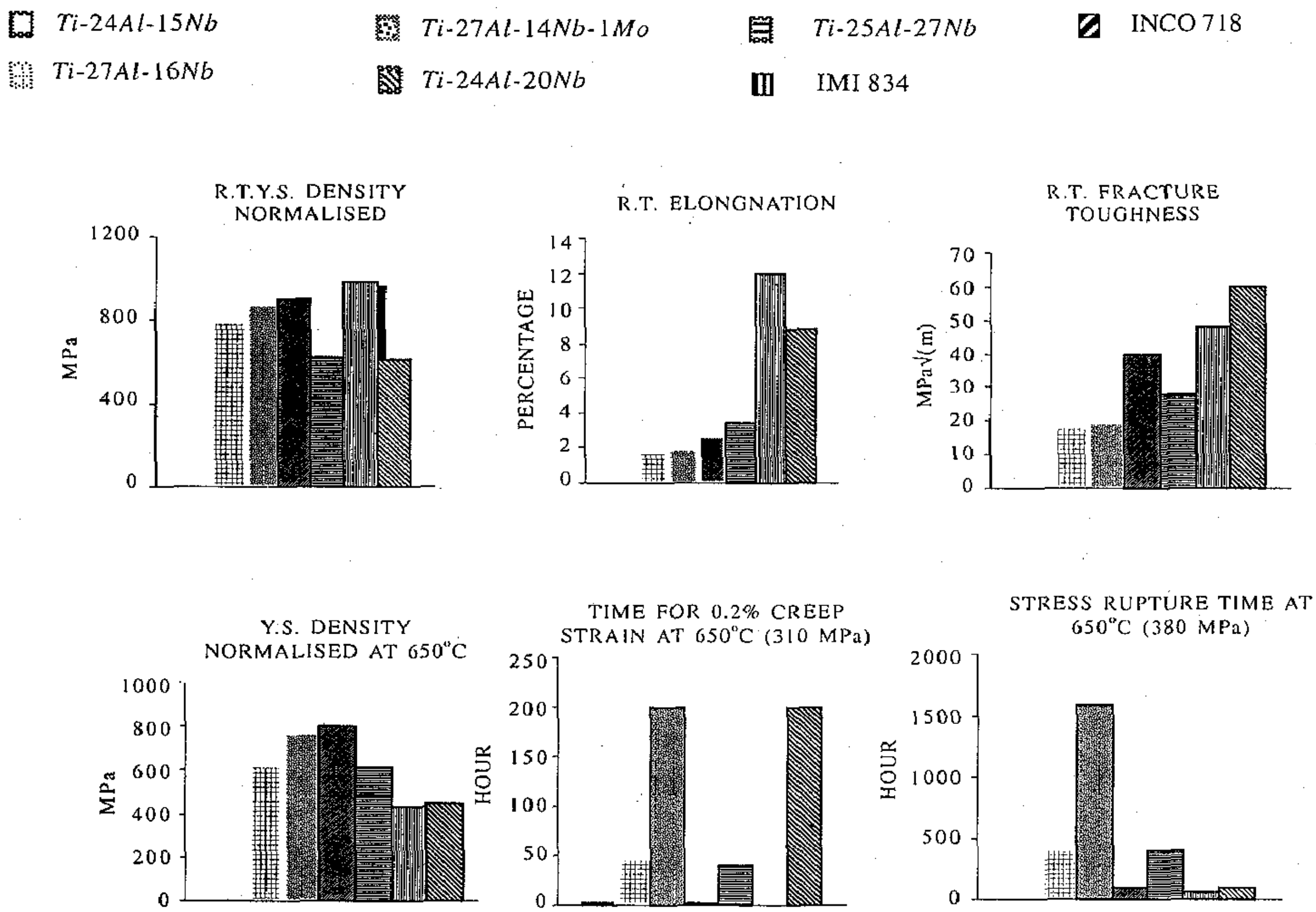


Figure 10. Comparison of Ti_3Al/Ti_2AlNb alloys with IMI 834 and INCO 718

* Personal communication with Centre des Materiaux, France.

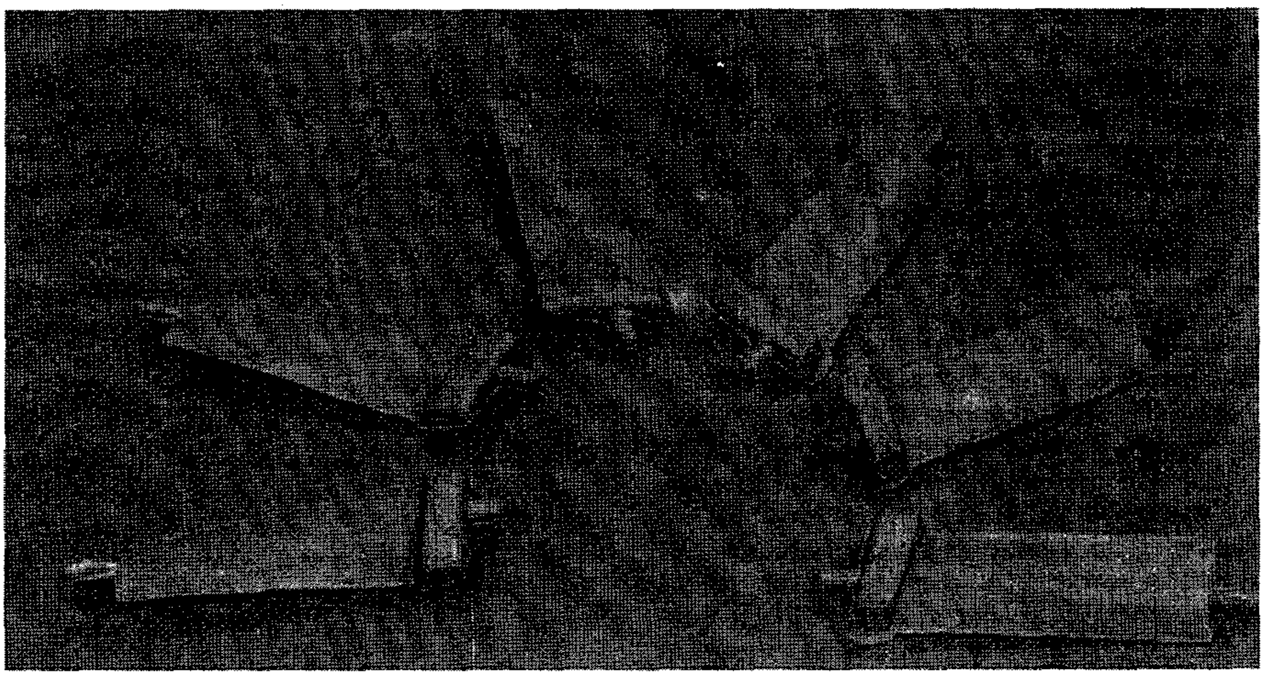


Figure 11. Compressor blade of a *Ti-24Al-15Nb* alloy forged at the Hindustan Aeronautics Ltd, Bangalore

to damage tolerance. While interest still remains in *O*-based alloys, there is a growing view that in their present state of development, their advantage over titanium alloys such as IMI 834, *Ti*-1100 are less attractive to justify widespread application in aerospace projects. Their use as matrix material in composites, however, is still considered attractive.

3.2 Titanium-aluminide (γ) Alloys

A great deal of the current interest in γ alloys arises from potential application areas in gas turbine engines and for valves in internal combustion engines for automobiles. Austin and Kelly[53] have presented an excellent review of the development and implementation of cast γ alloys. All of the main turbine engine manufacturers, including General Electric Aircraft Engines (GEAE), Pratt and Whitney (PWA), and Rolls Royce (RR) have been involved with the development of γ alloys for rotating and static engine components in the compressor, combustor, turbine, and nozzle[54,55].

Research and development on γ-*TiAl* alloys have progressed significantly within the last decade and a great deal of work has been done on alloying and microstructural effects and innovative processing techniques[56-59]. Like its counterparts in conventional titanium alloys and α_2/O alloys, these alloys exhibit a large variety of microstructures depending upon composition and processing. The alloys usually consist of Ti_3Al (α_2) and $TiAl$ (γ) in various proportions and varied morphologies.

Titanium aluminide alloys, in general, fall in a composition range (in atom %) which can be expressed as follows:

$$Ti - Al_{45\text{-}52} - \mathrm{M1}_{1\text{-}3} - \mathrm{M2}_{1\text{-}10} - \mathrm{M3}_{<1}$$

$$\mathrm{M1} = Cr, Mn, V$$

$$\mathrm{M2} = Nb, Ta, W, Mo$$

$$\mathrm{M3} = Si, B, C$$

Additions sof chromium, mangnese, and vanadium increase the ductility[57-59] at room temperature due to higher propensity for mechanical twinning, whereas niobium, tungsten, and to a smaller extent, tantalum, improve the oxidation resistance[60, 61]. In contrast to the ductilising elements such as niobium, tungsten and tantalum enhance the creep properties due to significant substitutional solid solution strengthening. Additions of boron, carbon, and silicon are known to yield either dispersion or precipitation hardening very effectively[60]. Interstitial elements such as oxygen, nitrogen, carbon, and boron reduce the ductility when their concentration[62] increases beyond 1000 Wt ppm. Especially in wrought alloys, small additions of boron help to control colony size in fully lamellar microstructures[56].

Microstructure-property studies seem to indicate trends similar to those observed with conventional alloys regarding phase morphology and size[63,64]. By appropriate thermomechanical processing, the phases can be adjusted to lamellar or equiaxed morphologies, or a mixture (duplex structure) of the two. Hot working at temperatures below the α transus generally results in recrystallised fine-grained microstructure. Post-working heat treatment in a single phase α field results in fully lamellar structures, while heat treatment in two-phase $\alpha+\gamma$ field results in a mixture of equiaxed γ and lamellar γ structure[65-67]. Therefore, the α transus temperature (T_α) is of particular importance in designing the processing/heat-treatment temperatures for controlling the microstructure. The α transus temperature in these alloys has the same significance as the β transus temperature in conventional titanium alloys.

In two-phase $\alpha_2+\gamma$ alloys, the volume ratio of equiaxed-to-lamellar γ influences properties strongly and a lamellar volume fraction of about 30 per cent gives rise to optimum combination of properties with desirable high-temperature creep resistance and acceptable levels of tensile strength and ductility[65]. As in conventional titanium alloys, fully lamellar microstructures exhibit superior creep and fracture toughness, while equiaxed microstructures result in improved ductility. Refining the prior α-grain size and control of lamellar spacing significantly affects strength and ductilities. Table 3 presents tensile properties and fracture toughness as functions of microstructure and temperature for some important alloys.

The γ alloys have been processed by conventional methods, including casting, ingot metallurgy, powder metallurgy, and also by novel means[63]. Near-net-shape components such as turbine blades and turbochargers, have been successfully cast by the investment casting. In the ingot metallurgy, cast

Table 3. Properties of engineering γ alloys

Alloy designation composition (atom %)	Processing & microstructure	Test Temp (°C)	Tensile properties Y.S. (MPa)	UTS (MPa)	El (%)	Fracture toughness K_Q (MPa $\sqrt{m}$)	Creep T(°C) / σ(MPa)/e(%)/t(h)/$\dot{e}_m$
48-1-0.2C *Ti-48Al-1V-0.2C-0.140*	Forging + HT Duplex - NL	RT 815	480 360	530 450	1.5 --	-- --	760/276/1/4.5/- 815/103/1/264/-
48-2-2 *Ti-48Al-2Cr-2Nb*	Casting + HIP Duplex	RT 760	331 310	413 430	2.3 --	20-30 --	700/138/0.5/63/- 760/138/0.5/13/- 815/138/0.5/2.4/-
ABB Alloy *Ti-47Al-2W-0.5Si*	Casting + HT Duplex	RT 760	425 350	520 460	1.0 2.5	22 --	760/138/0.5/650/-
47 XD *Ti-47Al-2Mn-2Nb-0.8TiB₂*	Casting + HIP + HT NL + TiB_2	RT 760	402 344	482 458	1.5 --	15 – 16 --	650/276/0.5/460/- 760/138/0.5/63.3/- 815/138/0.5/10.5/-
GE Alloy 204b *Ti-46.2Al-XCr-Y(Ta, Nb)*	Casting + HIP + HT NL	RT 760 840	442 382 381	575 560 549	1.5 12.4 12.2	34.5	700/207/0.12/150/- 760/138/0.16/150/- 840/69/0.31/150/-
Alloy 7 *Ti-46Al-4Nb-1W*	Extrusion HT NL	RT 760	648 517	717 692	1.6 --	--	760/242/0.1/3/-
Alloy K5 *Ti-46.5Al-2Cr-3Nb-0.2W*	Forging + HT Duplex / RFL	RT 800 870	462/473 345/375 --/362	579/557 468/502 --/485	2.81/1.2 40/3.2 --/12	11/20-22	800/138/--/--/3X10^{-3} (duplex) 760/138/0.5/421/1X10^{-5} 800/138/1.0/158/4X10^{-3}

NL=Near-lamellar, T=Temperature, $\ddot{e}_m$=Min creep rate (h^{-1}), FL=Fully lamellar, σ =Stress, e=Strain (%), RFL=Refined FL, t=Time(h) ingots are HIPed and then hot worked by the isothermal forging, extrusion or hot-die forging[68-71]. Rolling of γ-*TiAl* alloys to sheets and foils has also been demonstrated using forging and pack-rolling process[56,60]. While processing of γ-*TiAl* alloys via ingot and powder metallurgical routes on industrial scale has been successfully demonstrated, it has necessitated the use of highly specialised processing techniques such as isothermal forging and extrusion, isothermal rolling, or pack rolling.

One of the most interesting development in the field of γ aluminide is the XD™ process developed by Martin Marietta[72,73], where the addition of up to seven volume per cent of TiB_2 has resulted in the significant improvement in strength, modulus, and structural refinement (Table 3). Development of ultrafine grain size by equal-channel extrusion in γ alloys is being explored**. The ultrafine grain size is expected to give improved strength, ductility, and toughness. The γ alloys have also been shown to exhibit superplasticity[74,75] at 900 °C-1150 °C and complex shapes have been formed[60,75]

The γ titanium aluminide alloys are undergoing extensive testing for the last stage of the LPT of 4084 by PWA and fifth stage of LPT of CFG-80C engine by General Electric Aircraft Engine (GEAE). A transition duct beam for the GE90 has been approved to be introduced into production pursuant to the completion of engine testing. This is not a critical part, but does provide the opportunity to gain some production and service experience with minimum risk. The alloys are being considered for high speed civil transport vehicles and single-stage-to-orbit concepts like reusable launch vehicle[76-78]. However, further research in alloy development and processing is required before widespread use of γ alloys becomes a reality.

4. TITANIUM-BASED METAL-MATRIX COMPOSITES

In the field of advanced materials, one of the most significant development is in titanium-based metal-matrix composites (MMCs). The MMCs can be divided into two classes. The discontinuous or particulate-reinforced MMCs and the continuous fibre-reinforced MMCs. Particulate MMCs can be fabricated using conventional processes and a variety of low-cost net shape processes, of which main applications are likely for automotive wear-resistant and heat-resistant components. These MMCs are similar to dispersoids strengthened materials, perhaps with the higher volume fractions of dispersoids such as *TiC*, *TiN*, *BN*, *SiC*, TiB_2, etc.

Of particular interest, however, are metal-matrix composites reinforced with continuous fibres. The application of these composites will allow radical changes in aero engine compressor design, from the conventional disc and dovetail blade arrangement to a MMC-bladed ring, termed a BLING[79]. This will lead to about 70 per cent weight reduction as shown in Fig.12.

The development of metal-matrix composites straddles the boundaries between alloy, process, and product development. Considerable effort has been directed towards the process of production of the titanium matrix with continuous fibre reinforcement and the study of the interactions between the fibre and the different alloys. While a number of processes are being evaluated for the production of titanium-based MMCs (TMCs), most of the development thrust is still centred on the use of titanium alloy foil as the starting material. Development has taken place to produce foil in a range of alloys from *Ti*-6*Al*-4*V* to aluminides[80]. Processes have been developed to manufacture, on a production basis, panels of *Ti*-6*Al*-4*V*, 100 μm thick up to 1.0 m×0.5 mm in area[83]. Other alloys such as IMI 834, *Ti*-1100, *Ti*-6242, *Ti*-24*Al*-11*Nb*[81,82], *Ti*-22*Al*-23*Nb*[83], *Ti*-22*Al*-25*Nb*[84,85] have been successfully produced on a development scale. Section of a 100 μ m thick foil of *Ti*-22*Al*-25*Nb* developed at the DMRL is shown in Fig. 13, and the process used involves the combination of hot rolling, pack rolling, and cold rolling, and is very cost competitive. The composites of various foil materials with silicon-carbon (*SiC*) fibres are generally produced by foil-fibre method, which involves laying up foil and fibre in specific arrangements, followed by vacuum hot pressing[86]. The fibres are generally held in place during this process either by a ribbon cross-weave or by an

** Private communication with IMSP, Ufa, Russian Federation

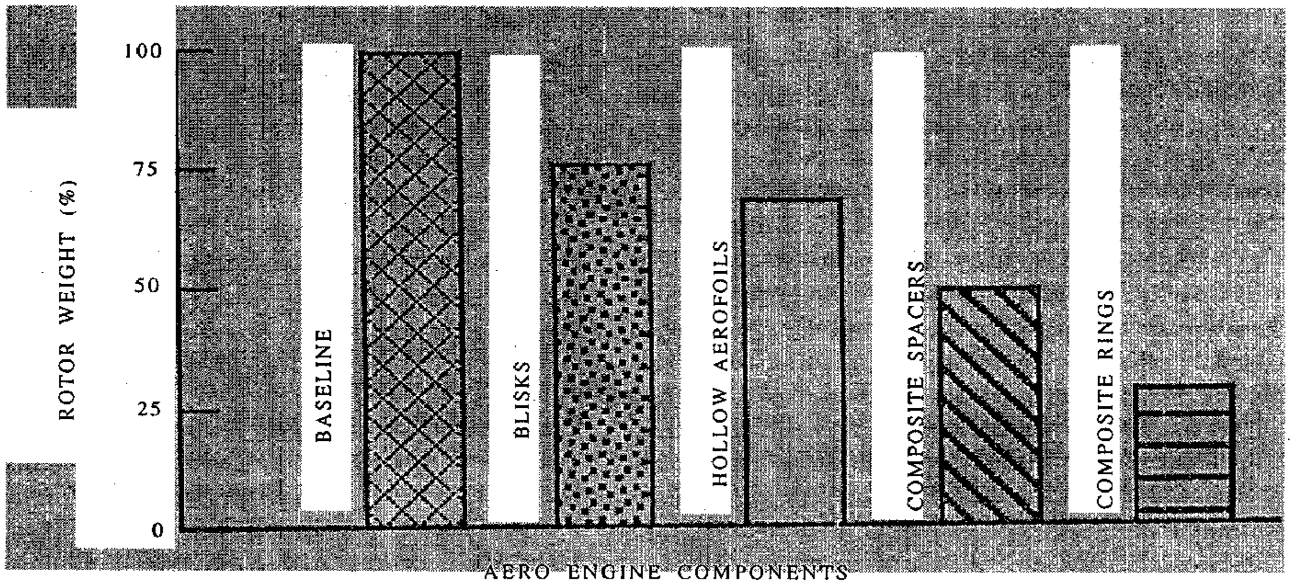


Figure 12. Potential weight savings in using *Ti*-MMCs in aero engine components

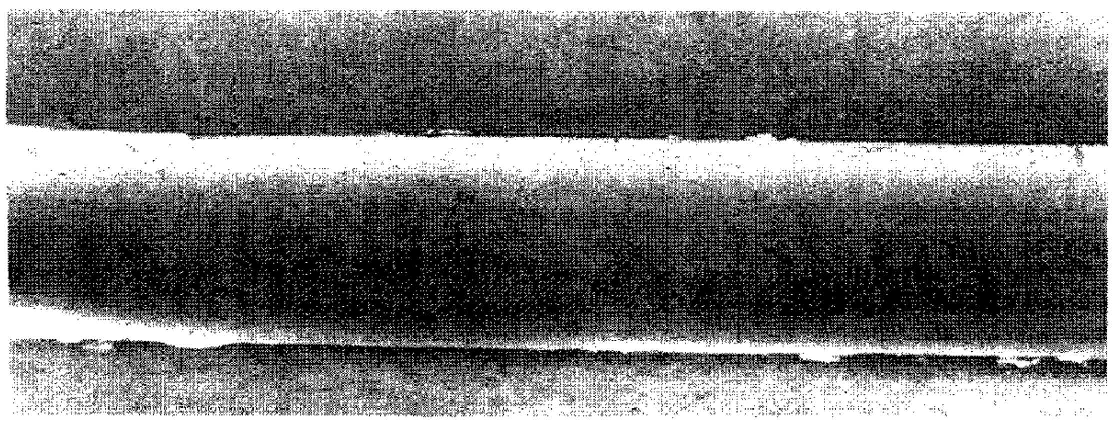


Figure 13. Foil of *Ti*-22*Al*-25*Nb* (100 μm thick) developed at the DMRL

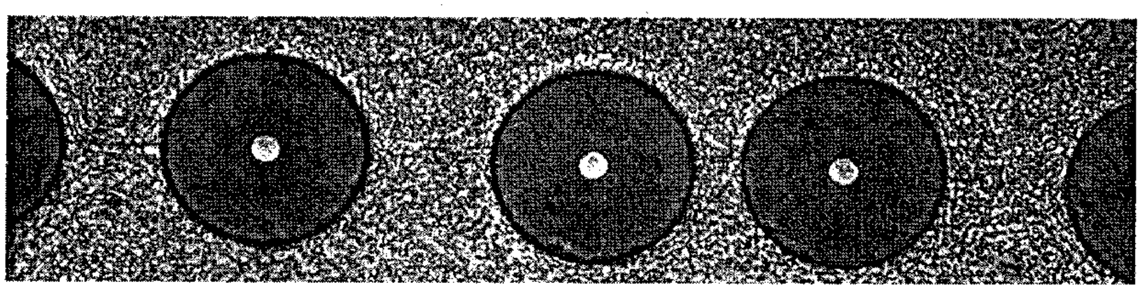


Figure 14. *Ti*-22*Al*-25*Nb*/*SiC* composite produced by fibre foil method utilising foil developed at the DMRL

organic binder that is later removed by vacuum degassing. Some of the other methods explored are powder cloth method[87], slurry tape casting method, and plasma spray method[88]. A composi (*Ti*-22*Al*-25*Nb*/*SiC*) produced by fibre foil methc is shown in Fig. 14.

Table 4. Properties of titanium-matrix composites

Composite type	Test temperature (°C)	Tensile properties				Creep
		YS	UTS	El (%)	E (GPa)	T(°C)/σ(MPa)/t(h)/e(%)
SCS-6/*Ti6Al4V*	RT	--	1932	1.09	202	538/965/26/0.2
	538	--	1370	0.87	183	650/758/258/0.2
	650	--	1221	0.86	167	538/1103/11/0.2
SCS-6/*Ti*1100	RT	--	1219	0.73	198	538/689/482/0.2
	538	--	1004	0.65	178	650/758/6/0.2
	650	--	971	0.68	166	538/827/8/0.1
SCS-6/*Ti*-25*Al*-10*Nb*-3*V*-1*Mo*	RT	--	1517	0.79	217	650/758/284/0.20
	538	--	1472	0.85	200	650/827/54/0.19
	650	--	1360	0.82	188	
SM1140/*Ti6Al4V*	600	--	1100	0.91	--	600/900/1/0.80
SM1240/*Ti6Al4V*	600	--	1090	0.85	--	600/750/60/0.78
SM1140/IMI 834	600	--	1050	0.75	--	600/900/25/0.80
SM1240/IMI 834	600	--	1060	0.81	--	600/750/90/0.65

A new exciting development for making TMC is matrix-coated fibre process developed at the DRA, Farnborough[89-91], UK. The process uses high rate physical vapour deposition (PVD) to precoat continuous fibres with a thick layer of matrix alloy. The coated fibres are then laid up and hot pressed into the finished TMC. The process has been scaled up to pilot plant level, and up to 1000 m lengths of *Ti*-6*Al*-4*V* coated *SiC* fibres for a 35 vol. per cent TMC are now being produced[91]. Technically, the process has many attractions. There is no restriction on alloy matrix composition. The matrix thickness around the fibre can be varied without difficulty and consolidation problems of spacing and fibre movement are eliminated.

One of the major factors limiting the use of TMC is interfacial reaction between the fibre and the matrix, which results in property degradation[92-95]. Coatings of TiB_2, CaO, Al_2O_3 and C have been found effective in reducing interfacial reaction. These coatings are applied on *SiC* fibre. The Ti_3Al base alloys show much less reaction with fibres such as SCS-6 than the conventional alloys[92] and is one of the reasons for growing interest in aluminides as matrix materials. A new coating technique, known as FUWT coating, which is based on a compliant layer/reaction barrier concept consisting of Gd/GdB_x, has also been reported[96] which prevents reaction of *SiC* fibre with titanium up to 1000 °C.

Table 4 lists some of the properties of titanium-based composites to present an idea of typical properties achieved in composites.

Another interesting development is laminated composites of two or more alloys to take advantage of superior properties of the two. An example is to combine superior high temperature properties of titanium aluminides with excellent toughness of β alloys. An example of this concept is shown in Fig. 15. The composite has been produced by combining the foils of *Ti*-22*Al*-25*Nb* alloy and a β alloy (*Ti*-10*V*-5.5*Fe*-1.5*Al*) by hot pressing and rolling, and is presently being evaluated at the DMRL. The possibilities of designing laminates with required properties in this manner are limitless.

The key factors governing the introduction of TMC into application are low cost, high performance, and feasibility of mass production[97,98].

5. CONCLUSION

The revolutionary development of high-temperature titanium alloys is linked to the progress in aero engine design. The use of titanium alloys has increased from 3 per cent to about 31 per cent in the last 5 decades, and is still increasing. During this era, capability of conventional high-temperature titanium alloys has increased from 300 °C in IMI 318 to 600 °C in IMI 834 and *Ti*-1100. At present, only conventional titanium alloys are in use in aero engines.

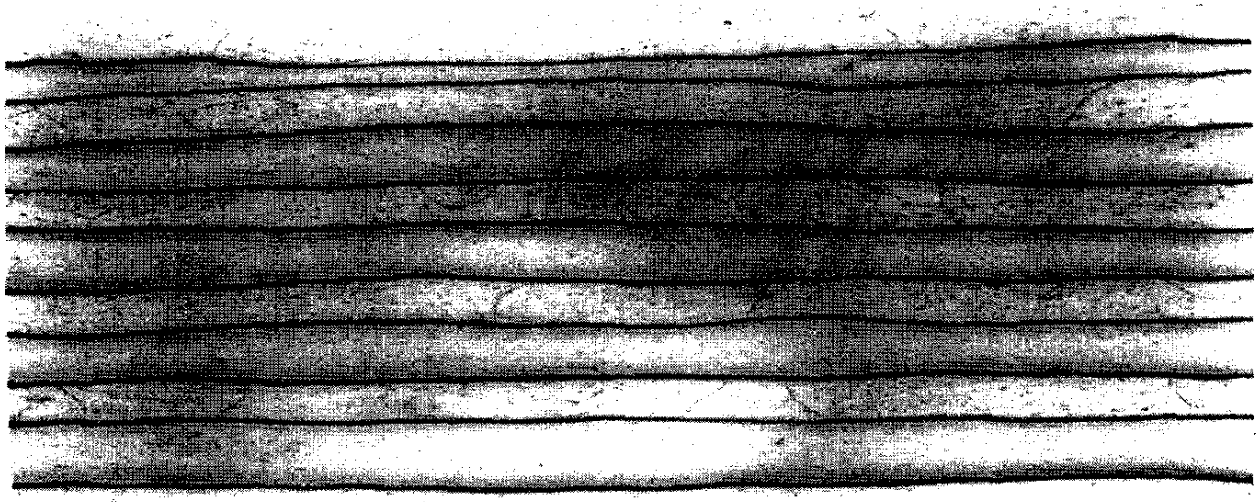


Figure 15. Laminated composite (*Ti*-22*Al*-25*Nb*/*Ti*-10*V*-4.5*Fe*-1.5*Al*) produced by hot pressing and rolling the foils of the two allc

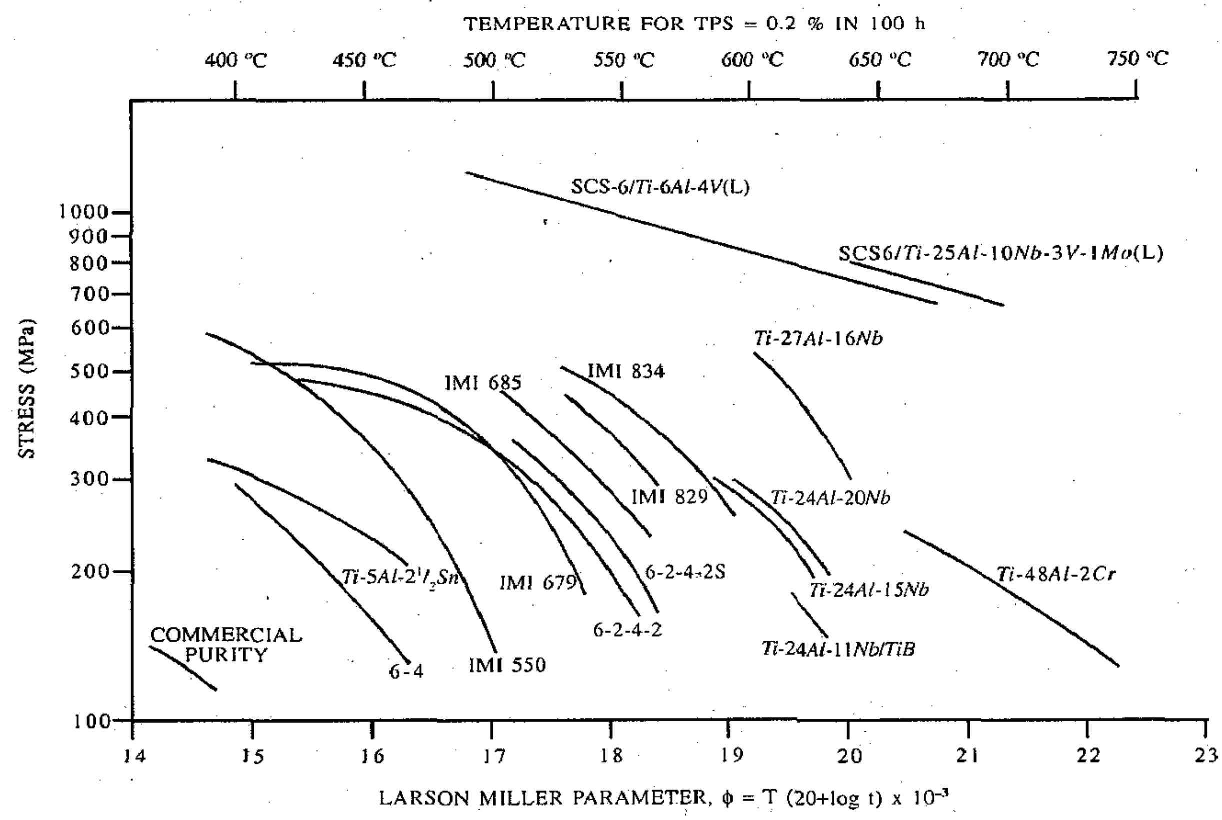


Figure 16. Evolution of high-temperature capability of titanium alloys–from conventional alloys to intermetallics and composi

A revolutionary increase in the temperature capability of titanium alloys can only be achieved by realising the potential of the aluminides and the composites. The evolution of high-temperature capability from the conventional alloys to intermetallics, to composite, is shown in Fig. 16. A comparis of some the important properties[99] is also present in Figs 17 and 18. Apart from some technologic issues related to environmental degradation a optimisation of properties, the major issues of conce

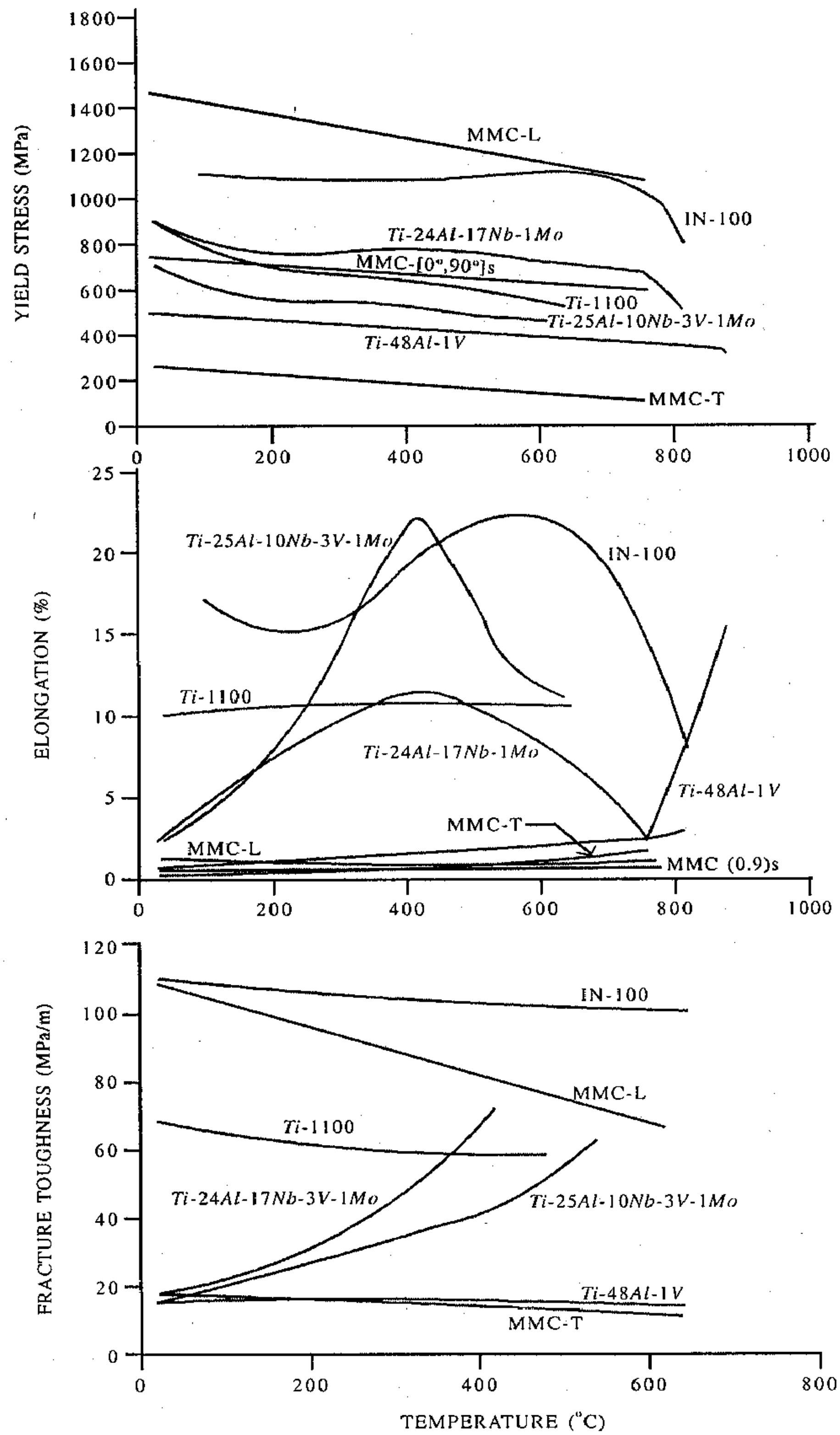
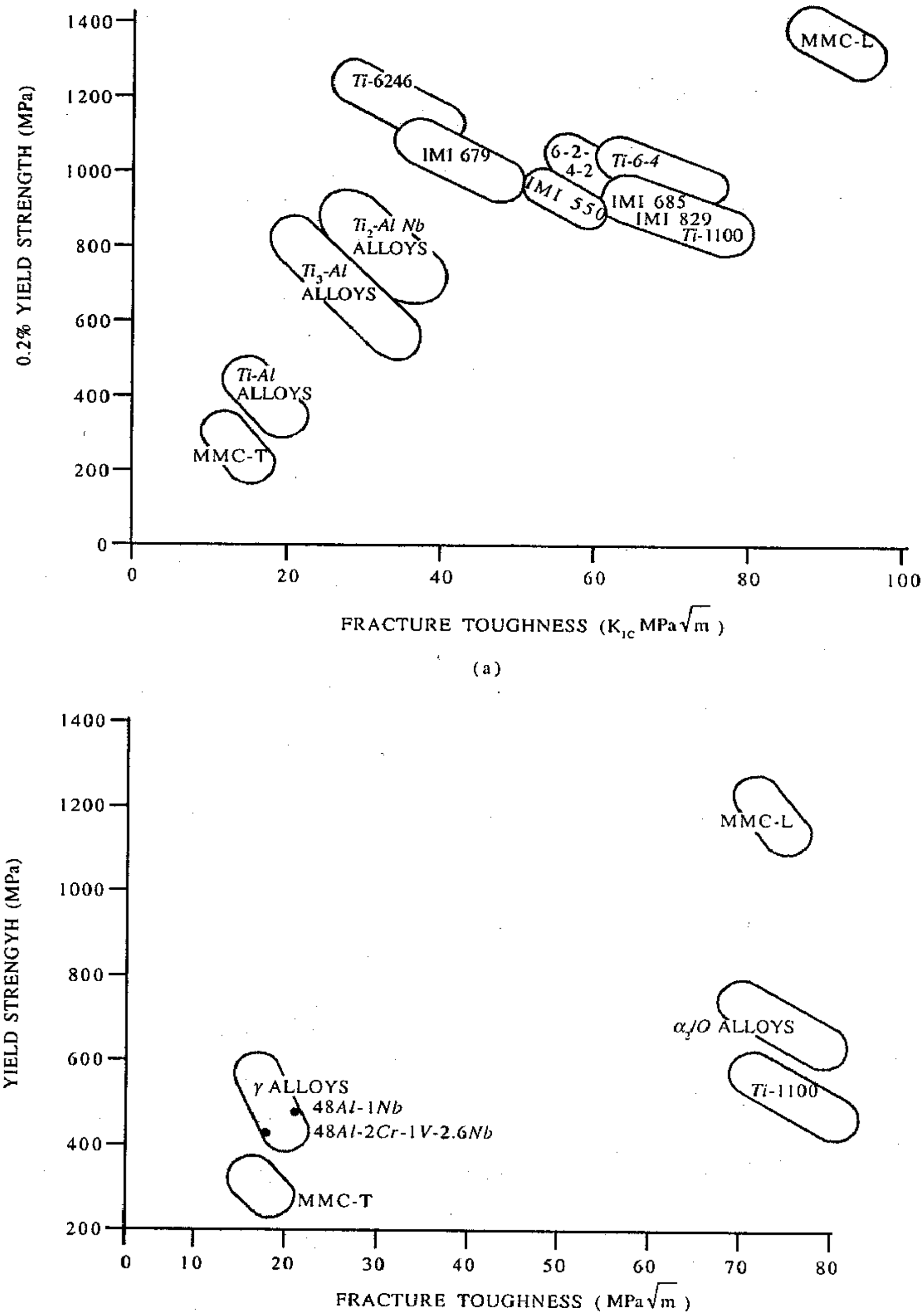


Figure 17. Comparison of mechanical properties of high-temperature titanium alloys



(b)

Figure 18. Range of yield strength and fracture toughness in conventional titanium alloys, intermetallics, and composites: (a) at room temperature and (b) at 600 °C.

relate to reliability, cost, and willingness of designers to adopt their designs using these new materials.

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