CREEP

Critical Review

DEVELOPMENTS IN HIGH TEMPERATURE ALLOYS

P.A. Blenkinsop, IMI Titanium Ltd., Birmingham, England

Introduction

The development of high temperature titanium alloys can be traced back over thirty years. While the driving force for these developments has been the gas turbine engine, major engineering acceptance did not materialise until the early 1960's with the widespread inclusion of the alloys in engine design programmes. Many titanium alloys have been developed to match specific targets linked with major engine programmes, and the story of titanium alloy development in many ways parallels the advances made in engine performance over the same period.

Initially this paper will attempt to trace the development of alloys both within Europe, in particular the UK, and contrast them with those in the US. Then report on the present state of the alloy technology before closing with a review of the future, in particular highlighting the obstacles that need to be overcome for titanium alloys to challenge the superalloys in the hottest stages of the engine.

Early Alloys

The targets set for the alloy developers in the early 1950's were modest when measured against today's aeroengine requirements and were limited mostly to tensile and creep strength properties. Earliest work was carried out in the USA and was based on a very broad front. The US programmes covered the range of titanium alloy types :

1) α phase alloys
2) α+β phase alloys
   (i) Ti alloys containing only β stabilising elements
   (ii) Ti alloys containing both α and β stabilising elements
3) meta-stable β phase alloys

Titanium + α Stabilising Elements

It was soon recognised that solid solution strengthening of the alpha phase with additions of Al, Sn and Zr gave improvements in room and elevated temperature tensile and creep properties. One such single phase alloy to emerge from this work was Ti-5Al-2½Sn which is still available commercially today and is the only one of its type to survive. The alloy has useful creep stability and tensile properties up to 300°C, it is weldable but difficult to cold form.

Attempts in the mid to late 1950's to develop alloys with much higher amounts of α stabilising elements (known as 'super alpha' alloys) led to problems. In the UK alloys in this group included Ti-13Sn-2¾Al while in the US Ti-7Al-12Zr and Ti-5Al-5Sn-5Zr were alloys that did not succeed to beyond the development stage. These alloys were found to be difficult to fabricate and to embrittle either on processing or during exposure at elevated temperatures.

While the creep properties of these 'super alpha' alloys were good, with the maximum working temperature being pushed to 400°C, the reasons for their severe embrittlement took time to establish. It was found that ordering took place within the α phase and the formation of $Ti_3Al(\alpha_2)$[1]. It was demonstrated that although Al was the most potent alloying element to cause ordering the other α stabilising elements were also effective. An empirical relationship was established for the Al equivalent[2] below which ordering does not occur.

$$Al + \frac{Sn}{3} + \frac{Zr}{6} + 10 \times O_2 < 9 \text{ (Levels measured in weight \%)}$$

This equation is now regarded as one of the corner-stones in defining alloy compositions for high temperature applications.

α+β Phase Alloys

(i) Ti+β stabilising element

The simple experiment of adding β stabilising elements to titanium was one that was explored extensively in the USA in the early 1950's. The elements considered were Mn, Fe, Cu and Cr. It was believed that by strengthening the β phase significant strength improvements could be achieved. The alloys consisted of primary α together with β phase which could be strengthened by ageing to precipitate secondary acicular α. Additional strengthening of the α phase occurred by the limited solubility of the β stabilising element.

In general such alloys were not particularly successful, although Ti-8Mn[3] found favour for a time. If this alloy is heated in the β field and quenched the microstructure will appear 100% β phase, and could therefore be regarded as a metastable β Ti alloy. However, in commercial use it was processed in the α+β phase region. It could not be used above 300°C because it suffered from metallurgical embrittlement owing to the formation of omega phase and the intermetallic TiMn phase on heat treatment.

(ii) Ti+α and β stabilising elements

This type of alloy system (α+transformed β) provided the greatest advance during the 1950's. Many of the alloys developed then still have widespread applications today. The earliest commercially available titanium alloys of this group were Ti-2Al-2Mn and Ti-4Al-4Mn[3]. With these alloys serious attempts were made to strengthen both the α and β phases simultaneously. They had good room and elevated temperature strength and were capable of operating up to 400°C for limited periods. These and similar alloys (Ti-4Al-3Mo-1V) found widespread applications in engines throughout the world. However, in the 1950's two alloys were developed, one in the USA and the other in the UK, which laid the ground rules for future developments and marked a significant shift in emphasis between alloy developers in the two countries.

The first alloy, Ti-6Al-4V[4] was developed in the USA at the start of the decade. It was rated up to about 350°C and since its introduction has become the universal workhorse of the aerospace industry. However, it was soon recognised that the alloy had some limitations in terms of tensile and creep strength and that there was a need to develop a stronger alloy with higher temperature capabilities. The US approach was to reply on a traditional alloying approach and resulted in the development of Ti-7Al-4Mo.

However, in the UK while the alloy development philosophy was based on $\alpha+\beta$ structures there was a significant difference in alloy composition compared to the US alloys. The major discovery was the beneficial effect of Si on the creep strength. A series of alloys were developed in the UK of which Ti-4Al-2Sn-4Mo-0.5Si[5] (IMI 550) gave not only 10% increase strength over Ti-6Al-4V, but also extended the operating capabilities up to about 400°C. IMI 550 is still widely used in Europe in both engine and airframe components. The use of Si as a major alloying addition has remained as one of the key factors in the high temperature alloy development within the UK.

Meta-stable β Phase Alloys

The work described earlier on the $\alpha+\beta$ titanium alloys where only β stabilising elements were added was extended in the US to systems in which the amount of β stabilising elements was such that the β phase could be retained at room temperature. The β phase was however meta-stable and could by heat treatment be strengthened by the precipitation of the α phase. Studies were made of the additions of V, Mo, Fe and Cr. Very high strength (up to 1500MPa) was achieved but in this condition alloys suffered embrittlement due to the precipitation of intermetallic phases e.g. $TiCr_2$ and the formation of the omega phase. Furthermore the alloys became metallurgically unstable during exposure at temperature and had little practical use beyond about 250°C. The alloy Ti-13V-11Cr-3Al was one of the early commercially available alloys and although a number of new alloys (6,7) have been developed more recently none have found high temperature applications in aero-engines.

Subsequent Developments

The major advances in the development of creep resistant titanium alloys came in the late 1950's. It was recognised that to obtain maximum creep strength it was necessary to have an α titanium base highly strengthened by solute elements (with the upper limit set by the avoidance of embrittlement by ordering) and that the relatively highly stabilised β phases of alloys such as IMI 550 and Ti-6Al-4V did not have the required creep stability. Therefore a new family of near-α alloys were investigated in which the maximum amount of α stabiliser was used, with sufficient β stabilising element to give medium strength levels, sufficiently low to avoid metallurgical instability and high temperature creep problems. Research was directed towards the study of stable α titanium bases with the addition of the optimum β stabilising element levels to give fabricability, strength and creep resistance. In the UK an α titanium base of Ti-11Sn-2¼Al was studied to which Zr, Mo and Si additions were made which led to the commercial alloy IMI 679[8] (Ti-11Sn-2¼Al-5Zr-1Mo-0.2Si). The addition of Zr strengthened the α titanium base, particularly at high temperature while the silicon level, which was fixed at an amount just in excess of the level of supersaturation gave improved creep resistance. The effects of silicon on the creep resistance of the alloy are detailed in Table 1[8] and the general benefits of Si are well documented in the subsequent literature [9,10,11]. The alloy which was $\alpha+\beta$ processed and heat treated represented a major breakthrough on its introduction in 1961, pushing the high temperature capabilities of titanium alloys to 450°C. At about the same time the US launched a near-α titanium alloy Ti-8Al-1V-1Mo. This alloy was capable of operating up to about 400°C and although it is still used, and has many attractive properties, including high modulus, it does suffer from an ordering embrittlement problem and therefore has to have its chemistry and thermo-mechanical processing particularly carefully controlled.

Si level	Creep strain % after 300 hours at 400°C - 544MPa
0	0.225
0.1	0.111
0.25	0.098
0.5	0.093

Table 1 : Effect of Si on the Creep Performance of IMI 679 (8)

Three years later Ti-6Al-2Sn-4Zr-2Mo (2) became available with a service temperature of about 470°C. This alloy has remained the most widely used high temperature titanium alloy in US manufactured engines. Reflecting UK practice of adding Si to high temperature alloys a study was made of adding up to 0.25Si to the alloy. However, work in the US on the effect of Si level and heat treatment on the creep properties of the alloy were somewhat inconclusive. The data is shown in Table 2(12) and indicates that only beneficial effects of Si are in the beta heat treated condition. This observation, together with general concern over the addition of Si to titanium alloys, precluded the element's addition. As will become apparent later it was not until 1975 that work in the US on 6-2-4-2 finally convinced researchers of the beneficial effects of Si in the alloy.

Heat treatment	Creep Deformation % (540°C - 200MPa - 150 hours)		
	Base alloy	0.125%Si	0.25%Si
Low $\alpha + \beta$	0.546	0.949	1.860
High $\alpha + \beta$	0.331	0.380	0.507
β	0.134	0.071	0.037

Table 2 : Effect of Heat Treatment Cycle and Si Level on the Creep Performance of Ti 6-2-4-2 (12)

The 1960's saw major developments in aero-engines and the widespread acceptance of titanium alloys for discs, blades and other components in the compressor stages. This breakthrough also heralded major alloy development programmes aimed at high temperature applications. Over the next 10 to 15 years considerable effort was mounted on the development of high temperature alloys. Many passed from the research to production scale but failed, either for technical considerations or because the engine programme for which they were designed did not progress into full production.

Within the UK the need to develop a weldable alloy capable of operating up to about 520°C for the Rolls Royce RB211 resulted in a major change in alloy philosophy. Whereas up to the mid-1960's all alloys had been alpha + beta heat treated, the realisation of the very significant improvements in creep resistance that could be achieved in near-α alloys by β heat treatment i.e. changing from the accepted $\alpha+\beta$ microstructure to one of fully transformed β (acicular α) was the basis of the alloy development strategy. While major improvements in the creep resistance of the then existing near-α alloys were obtained by β heat treatment, the level of tensile ductility was unacceptably low. Compositions had therefore to be explored which, while maintaining the creep resistance imparted by β heat treatment, would do so at acceptable levels of tensile strength and ductility maintained after exposure at temperature.

Out of this work came first IMI 684 (Ti-6Al-5Zr-1W-0.3Si) and then IMI 685 (Ti-6Al-5Zr-0.5Mo-0.25Si). Commercially IMI 685 is β heat treated to produce an acicular microstructure. For comparison Fig.1 shows the difference in creep behaviour of the alloy in β and α+β heat treated conditions. The benefits of the former heat treatment are evident.

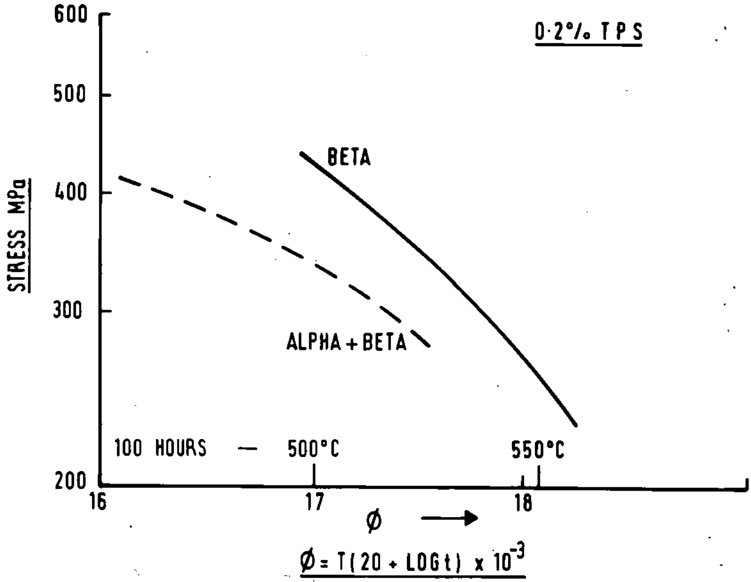

Fig.1 : Comparison of Creep Data of α+β and β Processed IMI 685 (Larson-Miller plot).

IMI 685 has good creep resistance and stability up to about 520°C and is used extensively by aero-engines manufacturers throughout Europe[13].

In the late 1960's work was also carried out in the USA on beta heat treated near alpha alloys. The alloy Ti-5Al-6Sn-2Zr-0.8Mo-0.25Si[14] was specifically designed for use in the engine for the SST aircraft. When the programme was terminated the alloy failed to secure an alternative permanent place in the US engine design.

In both the US and Europe other alloys were developed in the early 1970's that moved from the research scale to production but failed to secure a place in aero-engines. These alloys included Ti-3Al-6Sn-5Zr-0.5Mo-0.5Si (Hylite 65)[15], Ti-6Al-2Sn-5Zr-1Mo-0.25Si (UT 651A)[16], Ti-6Al-2Sn-1.5Zr-1Mo-0.1Si-0.3Bi (Ti-11)[17], Ti-4Al-6Sn-5Zr-1Mo-0.5Si and Ti-6Al-2Sn-2Zr-1Mo-0.1Si-1.5W

Basic alloy research was also carried out, particularly in the US, on ways of increasing the Al level (or Al equivalent level) in high temperature alloys but at the same time modifying the deleterious effect of ordering. Gallium containing alloys[18,19] were investigated extensively and more recently alloys with high niobium levels have attracted attention[20,21]. Some improvements in creep resistance were obtained but they were not matched by corresponding enhancement of metallurgical stability and ductility.

While most research has been directed towards increasing the maximum temperature capability of titanium alloys using near-alpha alloys some parallel research was carried out on $\alpha+\beta$ titanium alloys to achieve high strength for moderate temperature applications (i.e. up to about 400°C). Many of these alloys grew out of the fundamental work that was carried out on the near-α alloys. For example in the UK Ti-2¼Al-11Sn-4Mo-0.2Si (IMI 680) arose from the understanding of the behaviour of the Ti/Al/Sn base which resulted in IMI 679; Ti-6Al-5Zr-4Mo-1Cu-0.2Si (IMI 700) was a derivative of IMI 685. In the US Ti-6Al-2Sn-4Zr-6Mo[22] is a high strength development of 6-2-4-2 and has found use in impeller and disc applications while Ti-17 (Ti-5Al-2Sn-2Zr-4Mo-4Cr) was developed by GE for intermediate temperature disc applications up to about 350°C.

Recent Developments

The developments in high temperature alloys in the recent past have reflected the increased interest for the alloys by the aero-engine designers. While the initial mechanical property targets were basically concerned with tensile and creep strength, the greater engineering demands placed on titanium alloys in the modern engines has highlighted the need to maximise the performance of the components and therefore fully extend the high temperature alloys. The fatigue strength of the alloys is now of the utmost importance and therefore the development of existing and new alloys has reflected this need to maximise the all-round capabilities of the alloys not just their tensile and creep strength.

The last 10 years has seen the developments, in both the US and Europe, directed towards establishing the optimum compositions and macro/micro-structures.

With increasing concern on structures and their relationship with properties, particularly fatigue, the concept of alloy development now embraces not only composition and heat treatment, but the whole package from ingot melting and processing to thermo-mechanical processing of the final component.

The structural factors and the way they interact with properties, particularly fatigue have been well documented. However, it is for the purposes of this review to highlight some of the observations.

Alloy design is of necessity a compromise, to maximise a particular property, for example, creep resistance can only be achieved at the expense of some other property. The significant improvements that can be achieved in creep resistance in changing from an $\alpha+\beta$ structure to a fully transformed β one has already been highlighted. Such changes introduced to many alloys would result in significant losses of tensile ductility therefore compositions have to be tailored to take account of this loss.

However, beta heat treatment alone is not sufficient to maximise creep resistance. The effect of cooling rate on the transformation product and the creep performance needs to be taken into account. While oil or even water quenching may be optimum for one component another in the same alloy may require an air cool. An example of the effect of cooling rate on the creep performance of beta heat treated IMI 685[13] is shown in Fig.2. These effects are not only associated with morphological changes but the composition of the alpha and of the interplate phases.

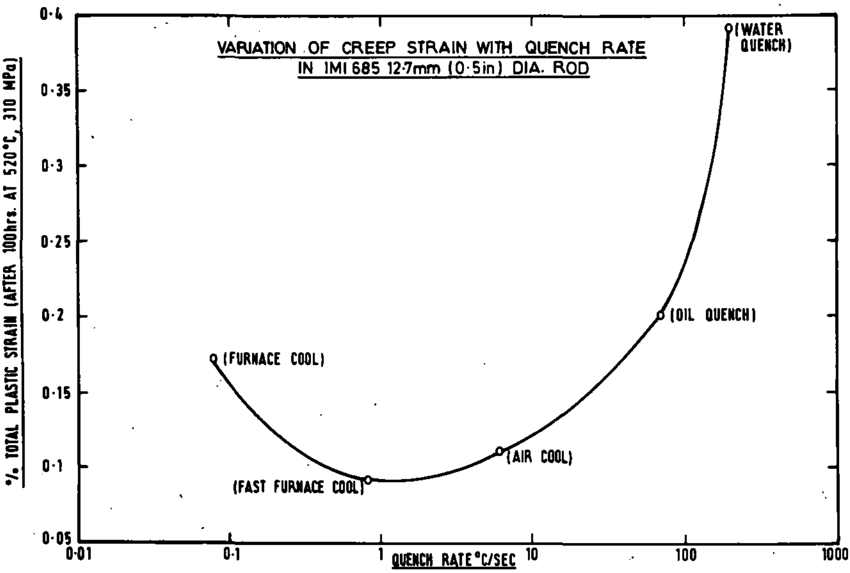

Fig.2 : Effect of Cooling Rate on the Creep Strain in IMI 685[13]

Other well established differences between beta and alpha + beta heat treated alloys are the high fracture toughness and slower fatigue crack growth rate associated with the former structure. On the debit side the beta heat treated materials tend to have lower room temperature tensile ductility and fatigue strength than the corresponding alpha + beta structures.

Much has been published on the effect of structure on fatigue and it is not within the scope of this paper to review those studies. However, the essential details are that the fatigue strength is basically a measure of the fatigue initiation resistance of the material. It has been shown that initiation occurs in the alpha phase[23]. With beta heat treated alloys the effective initiation unit is either the prior beta grain or the alpha colony, which can vary in size from 0.5mm up to about 2mm compared to the alpha grain size in an $\alpha+\beta$ structure which is typically 10-15 μm. How these differences are reflected in fatigue terms is shown in Fig.3 which compares the low cycle fatigue strength of $\alpha+\beta$ and β processed Ti-6Al-4V; similar differences are seen in other $\alpha+\beta$ and near-α alloys.

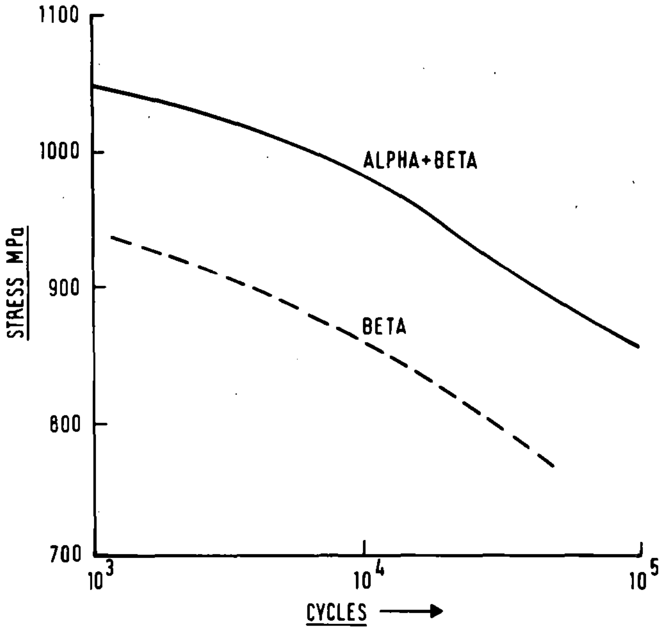

Fig.3 : Comparison of Low Cycle Fatigue Data of Ti-6Al-4V in Both $\alpha+\beta$ and β Heat Treated Conditions.

The need to maintain the beneficial effects of acicular structures while at the same time embracing the improvements associated with the fatigue strength of the alpha + beta structures has been the driving force behind much of the recent developments.

In the UK IMI 829[24,25] (Ti-5.5Al-3.5Sn-3Zr-1Nb-0.3Mo-0.3Si) is an alloy that was designed with the need to reduce the size of structural features. Not only was the composition chosen to maximise the creep performance of the alloy, but also to refine the macro and microstructure. Forging routes have also been developed which are aimed at minimising the beta grain size in the final forging. Postans and Jeal[26] detailed the thermo-mechanical processing approach. The technique applied involves the use of multi-recrystallisation stages.

The beta grain size in billet is typically in the range 1-1.5mm. (Fig.4 shows the macrostructure of a beta heat treated billet slice). By further controlled processing β grain sizes of 0.5-0.75mm are now being consistently produced in IMI 829 forgings.

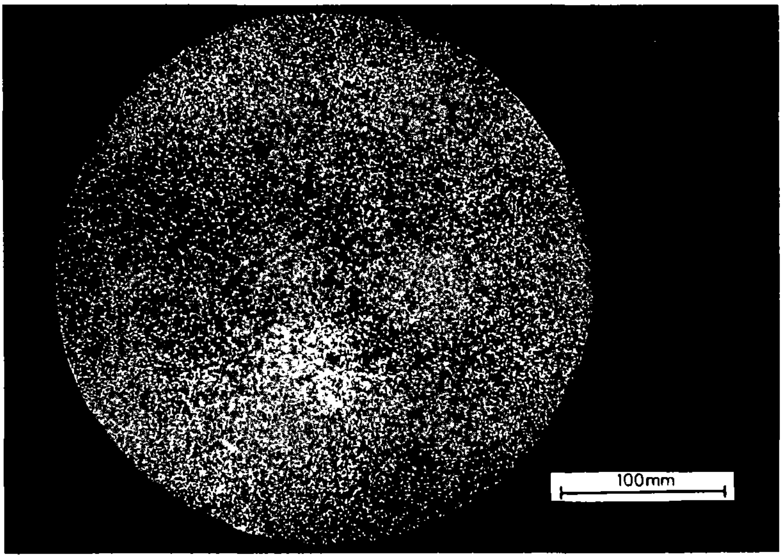

Fig.4 : Beta Heat Treated IMI 829 Billet Slice

In the US 6-2-4-2 has been the alloy which has received greatest attention. Chen(27) and Hall(28,29) have both reported studies aimed at maximising creep resistance in the alloy while at the same time maintaining the fatigue strength associated with equiaxed α+β structures. The work has involved both β and α+β forging combined with β and α+β heat treatments. Some of the results are summarised in Table 3(27).

Forging/heat treatment condition	*Creep hours to 0.2% strain	Tensile RT Y.S. MPa	Tensile RT U.T.S. MPa	Elong %	RinA %
(α+β)/β	199	903	1034	13	21
(α+β)/β	169	924	1076	10	15
(α+β)(α+β)	56	952	1062	14	27
(α+β)/β	19	1069	1241	3	5
β/(α+β)	185	931	1062	11	23
β/(α+β)	160	938	1076	10	22
β/(α+β)	116	896	1034	11	25
β/β	134	841	972	7	17

*Creep condition : 566°C - 172MPa

Table 3 : Effect of Forging/Heat Treatment Variables on Properties of 6-2-4-2S(27)

In order to maximise creep resistance in 6-2-4-2 Seagle(30) re-examined the effect of silicon. This work indicated that in the α+β processed and heat treated condition significant improvements in creep resistance could be achieved by additions up to about 0.1%. (Fig.5). The existence of a minimum in the creep curve is unexpected. Observations in other alloys indicate that the effect of silicon is to increase creep resistance with increase in Si level up to the supersaturation limit for the alloy.

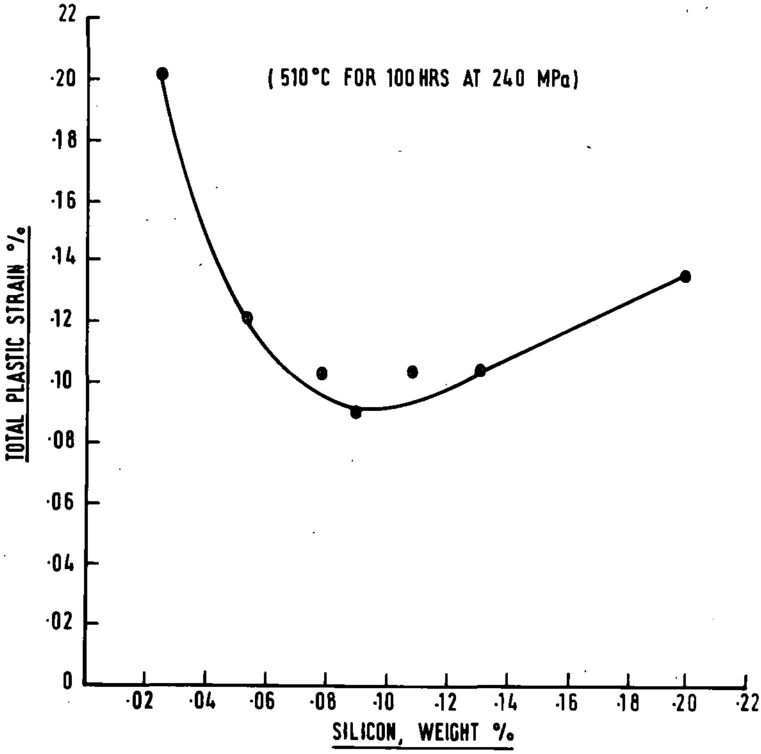

Fig.5 : Effect of Silicon on the Creep Strain in 6-2-4-2[30]

The need to maximise tensile, creep and fatigue strength is behind most of the new development alloys. In the UK a target temperature of 600°C has been set for an alloy having a strength of 1030MPa. Structural refinements in the alloy designated IMI Ex 834 have been achieved by compositional, heat treatment and processing control[31].

These developments are in line with similar trends in the USA. Much of the work was carried out on modifications of Ti-5Al-6Sn-2Zr-0.8Mo-0.25Si (5621S). Two alloys Ti-5Al-5Sn-2Zr-2Mo-0.25Si (5522S) and Ti-5Al-5Sn-2Zr-4Mo-0.25Si (5524S) were developed against a tensile strength (yield strength 378MPa min. at 483°C) and prolonged exposure targets at 483°C with potential operation up to 540°C. Data on these alloys can be found elsewhere in these proceedings.[32]

Present Status

The developments in alloys over the last 30 years are best illustrated in the form of a Larson-Miller graph. Fig.6 shows the increased creep capabilities achieved in the principal alloys. A more simplified illustration is shown in Fig.7 which details the year of introduction and the typical working temperature of alloys that are used in present generation aero-engines. By alloying and process development, conventional titanium alloys have now been produced that are capable of operating up to 600°C. The technology is now being established where alloys are tailored by thermo-mechanical processing and microstructural developments to achieve a particular combination of properties. In the future there is likely to be an increasing use of a particular alloy in a number of microstructural conditions, where more account can be taken of the specific requirements be they creep, fatigue, fracture toughness or stability.

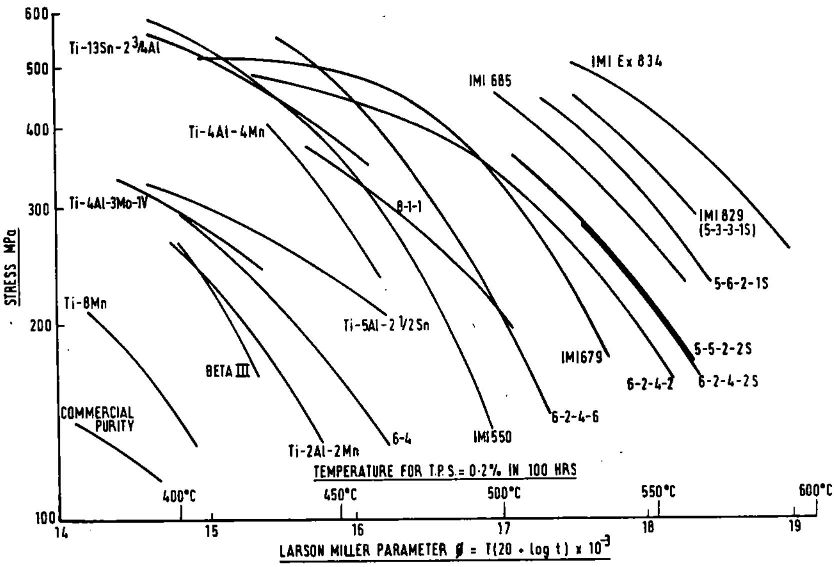

Fig.6 : Creep Resistant Alloy Development. Larson-Miller Plot Showing Improvements Achieved over Last 30 Years.

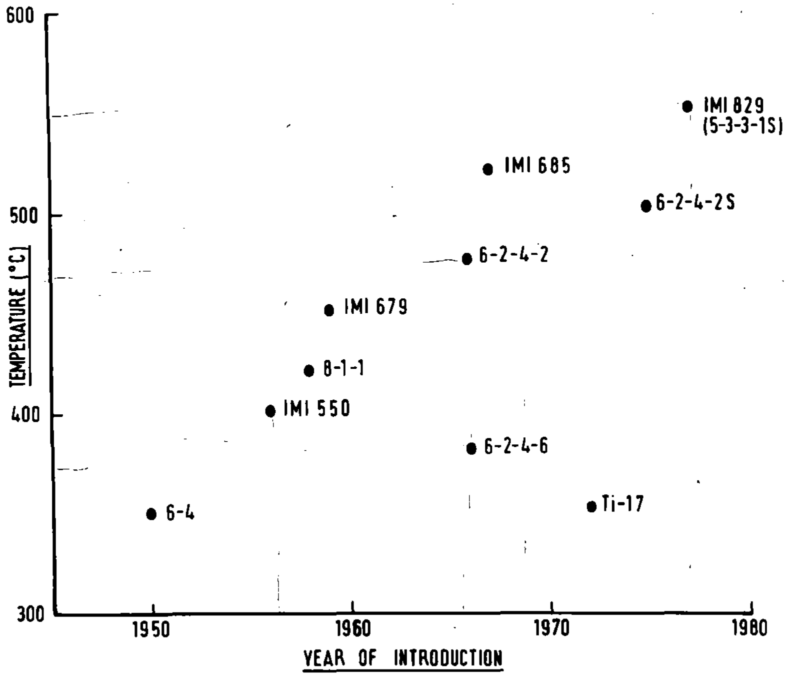

Fig.7 : High Temperature Alloys at Present in Production Aero-engines: Year of Development and Temperature Capability Under Optimum Stress Conditions.

Future Developments

A temperature in the range 600-650°C may represent a barrier for future developments in titanium alloys. The prolonged exposures to temperature demanded by the aero-engine designers throws into question the long-term surface and metallurgical stability of conventional alloys. Of particular concern is the surface oxidation and the resultant loss of surface integrity. While conventional titanium alloys can be developed that will have the requisite mechanical properties to operate at these high temperatures none have sufficient inherent oxidation resistance to overcome the problem. The choice therefore exists of :

a) Developing coatings for conventional and new types of alloys
b) Developing new inherently oxidation resistant alloys

Coatings

The combination of coatings with existing titanium alloys has many attractions. Work in both the US and UK has tended to concentrate on the use of noble metal coatings e.g. Pt and Au using the ion plating technique. In a US Air Force programme[33] it was found that 1 μm thick ion plating provided good surface adherence without degrading fatigue properties. The effect of ion plated coatings for long-term oxidation protection is demonstrated on 6-2-4-2 for times up to 500 hours and temperatures up to 700°C. (Table 4).

Ion Plated Coating	Temperature °C	Weight Gain mg $cm^{-2}h^{-1}$
None	590	6.9×10^{-2}
Gold	430	2.2×10^{-4}
Gold	480	2.6×10^{-3}
Platinum	590	1.2×10^{-3}
Tungsten/platinum	650	3.3×10^{-4}
Tungsten/platinum	700	1.7×10^{-3}

Table 4 : Effect of Coating on Oxidation(Weight gain) of Ti-6-2-4-2 in Air for 500 Hours(33)

Other advantages have been found to ensue from the use of oxidation resistant coatings including improvements in fretting resistance(34) and creep performance(35) in conventional alloys.

While laboratory results on ion plated coatings show considerable potential further development is required particularly in terms of improved coating adherency and the avoidance of coating defects which can lead to premature local oxidation and spalling.

New Alloys

Perhaps the greatest effort has been expended in studies of the titanium aluminides. One of the first reported studies was carried out by Winter et al(36) on the Ti_3Al + Nb system and reported in the First International Titanium Conference in 1968.

Subsequent work on both TiAl(γ) and $Ti_3Al(\alpha_2)$ systems has been extensive and is well reported in the literature. The major advantage of the aluminide systems is seen in their lightness and extremely good oxidation resistance up to very high temperatures (~ 900°C). Cyclic and isothermal oxidation weight gain curves are shown in Fig.8 for TiAl, Ti_3Al, and a ternary Ti-Al-Nb(37). An important feature of oxidation behaviour not revealed in Fig.8 is the propensity of the scales to spall upon cooling to room temperature. In isothermal tests the scales on all three alloys remained adherent during actual high temperature exposure. Under conditions of cyclic oxidation, however, extensive spalling occurred on Ti_3Al exposed at 900 and 1000°C and on TiAl exposed at 1000°C. In contrast the scale found during cyclic exposure of Ti-Nb-Al specimens was adherent at both 900 and 1000°C.

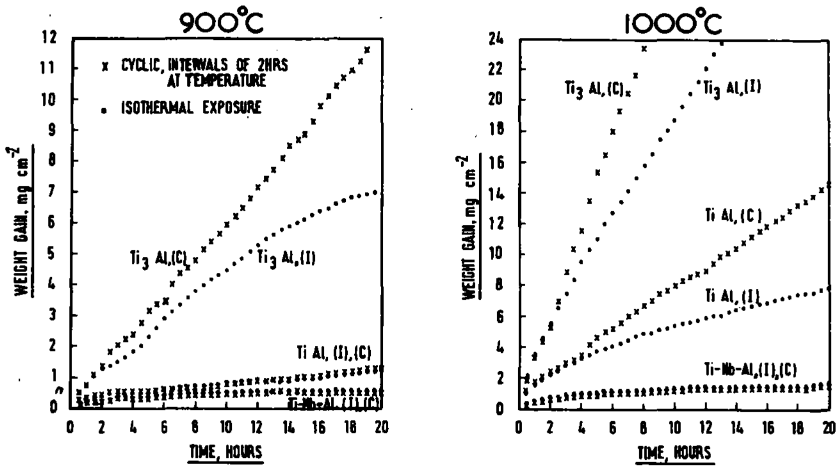

Fig.8 : Isothermal(I) and Cyclic(C) Oxidation Plots of Weight Gain(Δ m) versus Time(t) for Binary Ti-Al and Ternary Ti-Nb-Al Alloy Compositions[37]

While the oxidation resistance of the aluminides is a major attraction, the alloys are difficult to process particularly TiAl and exhibit poor low temperature strength and limited ductility. Much effort has been directed towards improving the mechanical properties of these alloys by compositional modifications, some examples are detailed in Table 5[38].

Alloy Designation	Temperature °C	0.2%YS MPa	UTS MPa	Elong %	RinA %
α_2	25	-	552	-	0.1
Quaternary α_2	25	796	920	0.4	1.1
α_2	700	413	521	2.7	4.0
Quaternary α_2	700	544	696	4.4	7.0
γ	25	335	450	0.9	1.6
Ternary γ	25	642	790	0.8	1.3
γ	700	300	360	1.4	2.2
Ternary γ	700	537	807	2.9	4.7

Table 5 : Room and Elevated Tensile Properties of Titanium Aluminides

Work on the aluminide systems continue particularly on the Ti_3Al + Nb system which offers a good combination of properties together with ease of fabrication. However, little information has been published in the open literature in the last few years.

A relatively new development in the area of high temperature titanium alloys is in rapid solidification processing (RSP). The process is seen as offering the opportunity of developing novel alloy compositions with associated beneficial microstructures and property improvements. Of particular attraction is the potential to develop precipitation strengthened alloys. The state of the art has been the subject of a recent review by Sastry et al[39] who have been particularly active in attempts to develop precipitation strengthened alloys using B, C and rare earth element additions. The use of such alloys holds particular attractions in terms of high temperature creep resistant titanium alloys. However, the work is in its infancy and will represent, if successful, the alloying approach for the 1990's or later.

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References

(1) F.A. Crossley, W.F. Carew : Trans. AIME 209, (1957) 43
(2) H.W. Rosenberg : Science Technology and Fabrication of Titanium Pergamon Press, (1970) 851
(3) R.A. Wood, R.J. Favor : Titanium Alloys Handbook. MCIC-HB-02 Battelle Metals and Ceramic Information Centre (1972)
(4) Anon. Mod.Metals 10, (1954) 94
(5) G.T. Harris, H.C. Child and A.C. Dalton : JIM 88, (1959) 112
(6) J.B. Gurney, V.C. Petersen, E.J. Dulis : Metal Progress, 96 (1969) 21
(7) E. Bohanke : Titanium, Science and Technology, Plenum Press (1973) 1993
(8) W.P. Fentiman et al : Science, Technology and Application of Titanium Pergamon Press, (1970) 987
(9) M.R Winstone, W.R. Rees : J. of Less Common Metals, 39 (1975) 205
(10) M. Kehoe, R.W. Broomfield : Titanium, Science and Technology, Plenum Press (1973) 2167
(11) N.E. Paton, M.W. Mahoney : Met.Trans 7A (1976) 1685
(12) R.G. Broadwell, M.L. Greenlee : TMCA Special Report No.24 (1966)
(13) P.A. Blenkinsop, D.F. Neal, R.E. Goosey : Titanium and Titanium Alloys, Plenum Press (1976) 2003
(14) S.R. Seagle, H.W. Bomberger : Science, Technology and Fabrication of Titanium, Pergamon Press (1970) 1001
(15) A.W. Hodge, D.J. Maykuth : Defense Metals Inf. Centre (DMIC 230) Battelle (1968)
(16) L. Seraphin et al : Mat. et Tech.No. 4/5 (May 1973) 156
(17) D. Eylon et al : Met.Trans 7A (1976) 1817
(18) H.L. Gegel, S. Fujishiro : Titanium, Science and Technology, Plenum Press (1973) 2179
(19) M.J. Godden, W.N. Roberts : Ibid 2207
(20) C.G. Rhodes, N.E. Paton, M.W. Mahoney : J.of Metals 30, No.1(1978) 54
(21) C.G. Rhodes, M.W. Mahoney, N.E. Paton : Proceedings of 5th Int.Ti Conference Munich (1984)
(22) J.A. Hall, C.M. Pierce et al : Mater.Sci.Eng. 9 (1972) 197
(23) D.F. Neal, P.A. Blenkinsop : Acta Met. 24 (1976) 59
(24) D.F. Neal : Forging and Properties of Aerospace Materials, Metals Society, (1978) 199
(25) D.F. Neal, P.A. Blenkinsop : Titanium 80 Science and Technology AMS-AIME, (1980) 1287
(26) P.J. Postans, R.H. Jeal : Ibid 441
(27) C.C. Chen, J.E. Coyne : Ibid 1197
(28) J.A. Hall : TMS-AIME Conference Philadelphia (October 1983)
(29) P.J. Bania, J.A. Hall : Proceedings of 5th Int.Ti Conference Munich (1984)
(30) S.R. Seagle, G.S Hall, H.B. Bomberger : Met.Eng.Quart. (Feb 1975) 48

(31) D.F. Neal, P.A. Blenkinsop : To be published in Proceedings of 5th Int.Ti Conference Munich (1984)
(32) A. Vassel, F.H. Froes, J.P. Herteman et al : Ibid
(33) D. Eylon, S. Fujishiro, F.H. Froes : To be published in Review of High Temperature Materials and Processes (1984)
(34) D. Eylon, R.K. Betts, S. Fujishiro : Thin Solid Films 73 (1980) 323
(35) S. Fujishiro, D. Eylon : Titanium 80 Science and Technology TMS-AIME (1980) 1175
(36) F. Prinzbach, H. Winter : The Science, Technology and Application of Titanium, Pergamon Press (1970) 875
(37) M.W. Kearns : The Production and Applications of Less Common Metals Hangzhou China - Metals Society, London (1982)
(38) P.A. Martin, H.A. Lipsitt et al : Titanium 80 Science and Technology TMS-AIME (1980) 1245
(39) S.M.L. Sastry, T.C. Peng et al : J. of Metals 35 (1983) 21