# LOW CYCLE FATIGUE BEHAVIOR OF HIGH TEMPERATURE TITANIUM ALLOYS AND ITS COMPOSITES - A REVIEW

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## ABSTRACT

Low cycle fatigue (LCF) behavior of high temperature titanium alloys and its composites designed for high temperature applications is reviewed .It is been observed that the fatigue life of titanium alloys strongly depends on the type of alloy, processing and microstructure. In general LCF life decreases with temperature, however it increases at intermediate temperature, the region of dynamic strain aging (DSA). LCF behavior of the composites depends on type of the composite and several other factors such as the nature of the matrix and that the second phase.

**Keywords: Low-cycle fatigue;** Titanium; Timetal 834; VT9; IMI 685; IMI 834; Coffin-Manson relationship; Metal Matrix Composites; Titanium aluminides

## 1. Introduction

High temperature titanium alloys are considered important material for aerospace industry because of their high specific strength. Titanium alloys have moderate density (4.5g/cm$^3$), high melting temperature (1668°C), good ductility, intermediate elastic modulus, excellent corrosion resistance and moderate tensile strength and a long history of successful aerospace applications.Their high specific strength provides fuel saving of about 40% as compared to many stainless steels without compromise on strength. Titanium alloys have progressively replaced aluminum, steel and to a certain extent even some super alloys. Titanium undergoes an allotropic transformation at 882°C from the β (bcc) to α (hcp) phase on cooling.This allotropic transformation provides an opportunity for development of different types of titanium alloys with various kinds of alloying additions. Table-1 gives temperature range and chemical compositions of high temperature Ti-alloys developed over the last 55years.[1-2] The more heat resistant alloys contain much less β phase than the alloy Ti-6Al-4V and are referred to as near α alloys. The α phase by virtue of its closed packed structure, far lower diffusivity than the BCC β phase, forms the major constituents (>95%) in the high temperature titanium alloys. The α phase is strengthened by elements which either stabilize it such as Al or which are neutral such as Zr and Sn to α phase.

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The addition of aluminium causes increase in tensile strength, creep strength, modulli and reduces the density. Tin is often used as solid solution strengthening element along with aluminum, however the use of Zr (> 5-6%) causes reduction in the ductility and creep strength. The oxygen content is usually kept fairly low (0.1-0.5%) in this class of alloys. Molybdenum has been used as prime β stabilizer in the alloys like IMI 685 and Timetal 834 as it increases the heat treatment response of these alloys. Another β stabilizer, Niobium, is added to improve surface stability at high temperatures. Silicon is an important element in high temperature titanium alloys as it increases strength at all temperatures and has a marked beneficial effect on creep resistance.[10]

Alloy Ti-6Al-4V accounts for almost 40-45% titanium alloys in usage. This alloy contains elements (Al) to stabilize and strengthen the α phase, together with V to stabilize and retain β phase on quenching from β /( α+β) phase field. These alloys are amenable to heat treatment and produce microstructure with different phase morphology. They have low hardenability, less toughness, good fabricability, and poor weldability[1]. Another α+β type alloy titanium alloy VT9 (GTM-900) is, designed for high temperature application as compressor disk in jet engines. The upper limit of service temperature for this alloy is around 450-500°C. Its β transus temperature is 970°C. It develops a wide variety of microstructures

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depending upon the solution treatment and the subsequent rate of cooling. The solution treatment in the α+β phase field gives rise to a two phase microstructure, consisting of primary α and transformed β.Heat treatment in the β phase field results in completely transformed β microstructure. The β phase transforms into α-platelets and thin layer of β is retained between the lath/platelets of α phase. The size of both α laths/platelets as well as retained β layer strongly depend on the cooling rate, following the solution treatment and increase with decrease in the rate of cooling.

However, efficient and rapid quenching leads to transformation of the β phase into a single-phase structure consisting of martensite platelets with orthorhombic crystal structure (α").The commercial heat treatment of the alloy VT9 consists of solution treatment in α+β/ β phase field,cooled in air and stabilization at 530°C for 6hrs. The stabilization treatment is given to remove the internal stresses, if any, resulting from the solution treatment and the subsequent cooling. Alloy VT9, exhibits dynamic strain aging (DSA) at 450°C in the α+β solution treated and air cooled condition and at 400°C in the β solution treated and air cooled condition. Limited investigations have been carried out on LCF behavior of this alloy at room temperature and at elevated temperatures.[1,5,17-18]

Alloy IMI 685 belongs to near α-titanium alloy category as it contains relatively small proportion of β-phase. It has a nominal composition of Ti-6Al-5Zr-0.5Mo-0.25Si and is characterized by a predominantly α phase microstructure. The α+β /β transus temperature of this alloy is ~ 1020°C. Silicon atoms being21% smaller in size than titanium atom tend to segregate to dislocations and inhibit dislocation climb and thus improve creep resistance of the alloy. Addition of small amount of silicon~0.5(wt.%) to improve creep strength of titanium alloy is well established. Addition of Al in this alloy causes stabilization and strengthening of the α phase. The neutral element Zr dissolves readily in the α phase and enhances creep resistance at elevated temperature.[1]

Alloy Timetal 834 is a near □-titanium alloy designed for high temperature applications as compressor disk and blade of advanced gas turbines of jet engines, as a substitute of heavy nickel base super alloys. It is already in use for some applications where temperature excursions reach even up to 650°C. Because the service conditions of the structural components such as compressor discs and turbine blades involve monotonic as well as cyclic loading, a combination of both high creep strength and fatigue resistance is required for its satisfactory working. For optimum performance this alloy, a bimodal microstructure with ~ 14 vol % of primary □, in a lamellar matrix of transformed β, has been established.[3-4]Alloy Timetal 834, is weldable and can be used in both cast as well as wrought form. It possesses a good combination of creep, low cycle fatigue and crack propagation properties.[6,8,9]

Titanium aluminides, based on the compositions TiAl and $Ti_3Al$, are considered promising structural material for aerospace, in particular, for engine components to replace the heavy superalloys due to their low density, high specific strength and modulus retention, and excellent creep resistance. The potential of the aluminides is shown in Table-2, which compares these with conventional titanium alloys and superalloys. However, extreme brittleness of these intermetallics make their use impracticable,if not impossible. In the field of advanced material, one of the most significant development is titanium based metal-matrix-composites (MMCs). Metal -Matrix Composites (MMCs) are a class of materials that combine the high strength and stiffness of ceramic with the damage tolerance and toughness provided by a metal matrix.Titanium matrix composites are promising materials for development of new compressors in aeroengine turbines.

Titanium MMCs offer potential advantages for structural applications,where they combine high strength,high temperature capability, and oxidation resistance of titanium with increased stiffness provided by the ceramic reinforcement [19]. The MMCs can be divided in to two classes: the discontinuous or particulate reinforced MMCs and the continuous fiber reinforced MMCs. The potential reinforcing phases for titanium, include TiB, $TiB_2$, SiC, $Al_2O_3$ and TiC.[10,11,13-16]

## 2. Experimental Details

A standard procedure has been followed for performing the LCF tests. Materials have been subjected to specific heat treatments as per the requirements of the different grades of the alloys. Standard cylindrical samples were used to perform LCF tests. LCF tests were conducted at different temperatures under total strain control at different strain amplitudes. The temperature along the gauge length of a specimen was controlled and was monitored with a thermocouple placed in contact with the specimen. Tested specimens were examined under scanning electron microscope and transmission electron microscope.

## .3. Results and Discussion

### 3.1 α+β alloys

Singh P.N. etal[5] studied LCF behavior of an α+β type high temperature titanium alloy, VT9

(GTM-900), in total strain control mode at ambient temperature and 500°C in air, following solution treatment in the α+β phase field, air cooling and stabilization at 550°C for 6hours. The cyclic stress response curves for the LCF tests carried out at RT and 500°C showed continuous cyclic softening at RT and the degree of softening increased with increase in strain amplitude. On the other hand the stress response at 500°C was quite different. There was mild softening during the initial 100cycles, followed by stabilization and subsequent hardening till the failure of the specimen. It may also be seen that the degree of hardening increased with decrease in the strain amplitude.Variation of fatigue life with plastic strain amplitude at RT was examined by C-M relationship. There was dual slope behavior in the C-M plot both at RT as well as at 500°C and fatigue life was higher at the elevated temperature.

The dual slope behavior was found to be associated with difference in the mode of deformation at low and high strain amplitudes. The higher fatigue life at 500°C than at RT may be attributed to higher ductility at 500°C. The continuous cyclic softening, increasing with strain amplitude, in the alloy VT9 at RT, may be attributed to high planarity of slip in the primary α phase and subsequent initiation of cracks in the planar slip band and their propagation. In contrast to expected increase in cyclic hardening with increase in strain amplitude, based on the concept of work hardening, there is opposite behavior in the cyclic stress response of the alloy VT9, at 500°C. The degree of hardening increases with decrease in the strain amplitude.

Since the LCF tests at 500°C also were conducted in air, it becomes essential to analyze the observed cyclic stress response at 500°C in terms of environmental influence because testing in air leads to enrichment of solid solution strengtheners like oxygen and nitrogen. Oxygen and nitrogen are known to increase the aluminum equivalent of titanium alloys. Nitrogen is twice as effective in increasing the aluminum equivalent as oxygen. Precipitation of $Ti_3Al$ phase occurs, once the equivalent concentration of aluminum exceeds the solubility limit. TEM examination of the sample tested in LCF at 500°C,revealed phase instability and precipitation of $Ti_3Al$ phase.[1,6,17-18]

### 3.2 Near α alloys

The near α alloy IMI 685 is an example of the composition developed to explore the opportunities of both β-forging and β-heat treatment. The α/β transus of this alloy is 1020°C. Quenching from 1050°C produces laths of martensite α' delineated by thin films of retained β between the α' platelets. Subsequent aging at 500-550°C reduces quenching stresses and causes some strengthening. Martensitic α' transforms to α and the microstructure comprises laths of α bounded by a fine dispersion of particles. IMI 685 was developed for high temperature strength and creep resistance for temperature up to 550°C. The heat treatment consists of solution treating at 1050°C ( β-range),oil quenching and aging for 24hrs at 550°C. β solution treatment followed by quenching, produces martensitic structures which are accompanied by small amounts of retained β-phase in the Mo containing alloys. The strength of these alloys is enhanced by dynamic strain aging in the temperature range ~ 427-577°C. This is considered to derive from and interaction between silicon and interstitials to produce a strong interaction with mobile dislocations. IMI 685 has high creep strength due to presence of silicon in solid solution in the alpha-phase.[1,12]

Nidhi Singh etal[3] studied the cyclic stress response and LCF behavior of the alloy 834 at its service temperature of 600°C in the bimodal microstructure condition with ~14 vol% primary α in the matrix of transformed β over a wide range of total strain amplitudes ($\Delta\varepsilon_p/2$) from ±0.75% to ±1.25%. The variation of average cyclic stress amplitude ($\Delta\sigma/2$) with number of cycles at room temperature and 600°C is showed that there was cyclic softening at both the temperatures, however,there was marked difference in the behavior of softening at the two temperatures. While at room temperature there was a continuous softening starting from the first cycle to the last cycle of failure particularly at higher strain amplitudes($\Delta\varepsilon_t/2 \geq \pm1\%$), at 600°C there was rapid softening during the 10 initial cycles and the rate of softening becomes relatively slower in the subsequent cycles. The cyclic stress remains stabilized for the major fraction of fatigue life at all the strain amplitudes. Further, both the rate as well as degree of softening was higher at room temperature than at 600°C at $\Delta\varepsilon_t/2 \geq \pm1\%$ and both of these increased with increase in strain amplitude at room temperature. On the other hand,there was a opposite trend at 600°C and the degree of softening decreased with increase in strain amplitude. The dependence of fatigue life on strain amplitude was analyzed using the Coffin-Manson (C-M) relationship between the plastic strain amplitude ($\Delta\varepsilon_p/2$) and number of reversals to failure ($2N_f$).

$$\Delta\varepsilon_p/2 = \varepsilon'_f\,(2N_f)^c \qquad [1]$$

Where $\varepsilon'_f$ and c are the fatigue ductility coefficient and fatigue ductility exponent, respectively. The C-M plot showed dual slope for the test conducted at room temperature, however, there was only a single slope for the tests conducted at 600°C. The variation of fatigue life with elastic strain amplitude ($\Delta\varepsilon_e/2$) was analyzed using the Basquin equation:

$$\Delta\varepsilon_e/2 = \sigma'_f/E(2N_f)^b \qquad [2]$$

Where $\sigma'_f$ and b are fatigue strength coefficient and exponent,respectively and E is the modulus of elasticity.It results showed that transition fatigue life at 600°C is significantly higher than that at room temperature. TEM study of LCF samples tested at 600°C showed that the dislocations in the planar slip bands were mostly unpaired,it might be inferred that there would have been partial recovery of order in the sheared $Ti_3Al$ precipitates at 600°C and this could have led to hardening. Further, since LCF tests were carried out in laboratory air, absorption of oxygen and nitrogen from the atmosphere would have caused solid solution strengthening,up to the depth of their absorption. The enrichment of oxygen and nitrogen in this alloy above the Al-equivalent would have also led to precipitation of $Ti_3Al$ in the matrix. The stability of cyclic stress at 600°C may thus be attributed to the balance between the various softening and hardening effects resulting from the different processes referred to above. TEM analysis showed that the spacing between the slip bands decreases with increase in strain amplitudes. It can be seen that at 600°C, the applied cyclic strain is distributed in relatively more number of slip bands than that at room temperature and hence the cyclic strain in the individual slip band at 600°C is relatively less than that at room temperature. The elimination of bilinearity from the C-M plots at 600°C might be associated with increased homogeneity of deformation,combined with partial recovery of ordering in the sheared $Ti_3Al$ precipitates and solid solution strengthening due to oxygen and nitrogen absorption thus the process of fatigue crack initiation,even at low strain amplitudes,is delayed and fatigue life is increased. This results in elimination of dual slope from the C-M relationship at 600°C.

K.V.Sai Srindh[1] showed that for alloy Timetal 834, the cyclic stress decreased from the first cycle and continued to decrease up to the last cycle to cause fracture. Thus there was continuous softening at room temperature. In general the degree as well as the rate of softening increased with the total applied strain amplitude. The cyclic stress response at 400°C was found to be completely different from that at RT and there was cyclic hardening followed by initial softening for a few cycles at $\Delta\varepsilon_t/2 \leq \pm 1\%$. However, there was continuous cyclic hardening at higher strain amplitude, ($\Delta\varepsilon_t/2 \geq \pm 1\%$). Further, it may be noted that the number of cycles up to which the initial softening occurs, decreases with increase in the total strain amplitude ($\Delta\varepsilon_t/2$) from ±0.65% to ± 1.0%.The degree as well as the rate of cyclic hardening was found to increase with the total strain amplitude. It was found that at 400°C, the cyclic stress reaches to a peak value and starts decreasing during the subsequent cycles. Further the rate of fall of the cyclic stress, during the later stage of cycling, leading to failure, was much higher than that at RT. The variation of fatigue life in terms of the number of reversals to failure ($2N_f$) with,total elastic and plastic strain components for the different test temperatures showed that bilinearity in case of LCF test at RT, with respect to elastic,plastic and the total strain amplitudes. However, only linear behavior is revealed from all the plots for the tests carried out at the higher test temperatures of 400°C and 600°C. From the C-M plots of the alloy 834, it was found that the LCF resistance was highest at RT and decreased with increase in test temperature.

It may be noted that there was dual slope in strain life plot at RT . The low fatigue life in the region of low strain amplitudes was essentially due to localization of the plastic strain in a few widely spaced slip bands. Such localization of plastic strain resulting from shearing of ordered $\alpha_2$ precipitates enhances the process of fatigue crack initiation and ultimately reduces the fatigue life. On the other hand it was found that there was only single slope in case of the test temperature of 400°C and 600°C. Potozky et.al[7] have observed that mode of deformation in the alloy 834 during LCF, remains planar slip from RT to 600°C and it becomes wavy at 650°C. Thus there should not be any difference in the process of deformation of the alloy 834 during LCF testing at RT, 400°C and 600°C. The lower LCF life at higher temperatures in the alloy 834 was associated with detrimental influence of oxygen enrichment in enhancing the the process of fatigue crack initiation and propagation.

LCF test at elevated temperature is known to be influenced by time dependent processes like creep, oxidation, metallurgical instabilities and dynamic strain aging (DSA).Occurrence of DSA during total strain controlled fatigue test is manifested in the form of serrated plastic flow in the stress-strain hysteresis loops, increased cyclic work hardening and reduced plastic strain range. Further, DSA promotes localization of plastic flow, planarity of slip and produces widely spaced slip bands. Impingement of slip bands at grain boundaries leads to increased grain boundary decohesion, and consequent reduction in fatigue life.[1]

### 3.3 Titanium Aluminides

Praveen etal[11] have studied the effects of alloying on ductility and LCF behavior of Ti-based aluminides. A number of aluminides have been identified in the Ti-Al system. However, for high temperature applications, aluminides like $Ti_3Al$ and

 

TiAl have received considerable attention because of their improved performance over the conventional high temperature titaniuma alloys.

These aluminides have ordered structure and possess attractive properties like low density, high specific strength and stiffness, creep resistance and oxidation resistance at elevated temperatures.$Ti_3Al$ exhibits very limited ductility below 500°C and hence, is not amenable to plastic strain amplitude controlled fatigue tests at low temperature. The possibility of combining the $Ti_3Al$ with either α or β phase has been explored by many investigators. The alloys containing β stabilizers in the range 10-12% such as Ti-24Al-11Nb and Ti-25NbAl-10Nb-2Mo-2Ta have two phase microstructure of $\alpha_2$ and β. The category of alloys containing 14-17% β stabilizers like Ti-24Al-(14-15)Nb, Ti-24Al-10Nb-3V-1Mo and Ti-24Al-17Nb and its derivatives may contain, depending up on the heat treatment, three phases viz $\alpha_2$, β and O(orthorhombic) (a distorted form of $\alpha_{2)}$ based on $Ti_2AlNb$.The next class developed more recently, again consists of a two-phase structure of the O and β phase. The fatigue life of these class of aluminides is mainly influenced by the microstructure and alloying elements present in the system.

Further for improving the ductility, addition of alloying element like V, Cr, Mn etc., have been made. These addition have led to development of complex alloys with general composition Ti-(46-49)Al-(0-4)Cr-Mn,V-(0.5-2)Nb,W,Mo-(0-1)Si,B,C. The LCF behavior of a  titanium aluminide alloy has been studied in fully lamellar (FL) and nearly lamellar (NL) conditions at two different temperatures, 650°C and 800°C, without and with hold period. It was observed that there was mild cyclic hardening at 650°C at higher strain amplitudes whereas there was cyclic stability at 800°C for both the microstructures. While tensile hold was found to increase the fatigue life, the compressive hold as well as the tensile/compressive hold had detrimental effect on the fatigue life.[13-16]

### 3.4 Titanium based Meta Matrix Composites (TMCs)

Advanced near-term high temperature composite materials are essential key material for successful development of the next generation of aerospace structural, propulsion and power generation systems. Titanium metal matrix composites (TMC] have been found to possess better high fatigue life and high temperature stability than the monolithic alloys. Table-3 shows comparison of mechanical properties of titanium and potential ceramic reinforcement phases.[19] Many applications have been identified for the use of titanium matrix composites (TMCs) in aircraft and aerospace vehicles. Integrally bladed compressor rings, shafts, ducts, fan components and structural rods are but a few of the components that have been proposed. For the national aerospace plane (NASP), titanium-based composites are envisioned for the aircraft skin, internal structure and cool engine parts. Another potential application is for components of the exhaust nozzle of the high-speed civil transport (HSCT] aircraft. In recent years, both continuously-reinforced and discontinuously-reinforced titanium alloy metal matrix composites (TMCs) have been developed that offer dramatic improvements in maximum temperature capability in regions where only the nickel-based alloys were used. Titanium matrix composite materials provide an opportunity to extend the operating temperature range of the monolithic titanium alloys in advanced aerospace products. Density of TMCs is only half that of the competing nickel-based alloys [20]. Presently some of the more promising materials are fiber reinforced titanium alloys: these generally consist of a conventional titanium alloy such as Ti-6-4 (Ti--6Al4V, all wt%); Ti-15-3 (Ti- 15V-3 Cr-3 Sn-3 Al, all wt%); IMI 834 (Ti-5.8Al-4Sn-3.5 Zr-0.7 Nb-0.5 Mo-0.35 Si-0.06 C, all wt%); and Ti /β 21s (Ti-15 Mo-3 Al-2.7 Nb-0.2 Si, all wt%), reinforced with continuous silicon carbide (SiC) fibers [21].

Since most TMCs are exposed to engineering service, their structural integrity is often limited by their mechanical performance under cyclic loads. Thus, in order to ensure proper design and safe use of the TMCs, the cyclic fatigue processes must be comprehensively characterized and their effects incorporated in life-prediction procedures. Several studies have been conducted to both understand and document the fatigue life evaluation and fatigue crack propagation characteristics of both the continuously-reinforced and discontinuously-reinforced titanium alloy metal matrices. A particulate can have either a spherical, cubic, platelet or random morphology, with a particle size of approximately 1±100 m. However, particulate reinforced MMCs also have the advantage of being isotropic, cheaper to manufacture, and amenable to subsequent processing and component forming operations compared to fiber reinforced MMCs [19].

Low-cycle fatigue failure in titanium metal matrix composites is caused by two separate damage mechanisms: fatigue crack growth in the Ti matrix and fiber breakage, interfacial debonding, microscopic voids at locations of the reinforcement in the matrix [20, 22, 24]. The fatigue life depends on the type of reinforcement used. Mechanical and cyclic stability of the intrinsic microstructural features coupled with an ability of the composite microstructure to distribute the cyclic strain over the entire microstructural volume, are two important factors controlling the cyclic response, cyclic strain resistance and stability of the TMC during fully reversed total strain amplitude-controlled cyclic deformation.

Srivatsan et al [20] observed that the Ti-6Al-4V/TiB composite showed evidence of only softening till the failure. Softening was more pronounced at the higher cyclic plastic strain amplitudes and concomitant higher response stress than that at the lower cyclic plastic strain amplitudes was resultant lower response stress.

The progressive decrease in stress-carrying capability of the Ti-6A1-4V/TiB composite microstructure is attributed to concurrent and mutually competitive influences of the formation of a number of microscopic cracks, their growth through the composite matrix and eventual coalescence to form one or more macroscopic cracks, and resultant growth of one or more of the macroscopic cracks along with the microscopic cracks through the Ti-6A1-4V/TiB composite matrix [20].At high-cyclic strain amplitude resultant short fatigue life than at the lower strain amplitudes and early crack growth was due to formation of numerous microscopic cracks surrounded by randomly distributed voids of varying sizes.

LCF life and damage mechanism of unidirectional SCS6 fiber at room temperature and high temperature has been studied [22,23]. Jeng et al [22] studied the LCF behavior of different titanium alloys reinforced with unidirectional SCS-6 fibers at stress levels in the range 470-1400MPa.Fatigue lives of the SCS-6/Ti6Al4V and SCS-6/Ti-25Al-10Nb-3V-1Mo are quite similar and a deviation from linear relationship of fatigue life and stress was observed at stress level greater than 1300MPa. The major damage mechanisms found are matrix cracking, interfacial debonding, fiber bridging and random fiber failure in the wake of the crack tip [22].

In the studies of tension-compression fatigue behavior of a unidirectional titanium-matrix composite at elevated temperature of 427°C by Kraabel et al [23] can be categorized schematically into three regions on a fatigue life diagram. Region I involving the high stress/low cycle portion of the diagram controlled by the fiber fracture. Region II, at intermediate stress levels, consists of damage mechanisms which are controlled by matrix fatigue. Two separate damage mechanisms were observed in this region: Region IIa consisted of short matrix cracks, and Region IIb of long matrix cracks. Region III involved low stress/high cycle portion of the fatigue life diagram with no damage up to three million cycles at stress levels around the fatigue limit of the matrix material [23].

The nature and properties of the interface play a critical role in the performance and failure of composites. The interfacial properties and origin of interfacial failure were influenced significantly by the fiber surface chemistry, matrix alloy composition, microstructure and residual stresses at the interface [25]. Several investigations have been conducted to understand the fatigue crack propagation resistance and mechanisms on different continuous SiC fiber reinforced titanium alloy composites at room temperature and high temperature. The fatigue crack growth resistance may vary depending on the volume fraction of reinforcement.

At low fiber volume fraction the transverse fatigue crack growth resistance of MMC is improved than at high volume fraction [26]. This is due to increase in crack tip radius by interaction between the holes from debonding interface and cracks, while the decrease in growth is due to the coalescence of these voids and microcracks. Transverse crack growth in unidirectional MMCs can be essentially of two types.

The principal damage mechanism at high temperature is crack bridging and fiber/matrix debonding proposed by Zheng etal [27], in their studies on the fatigue crack growth behavior on SiC (monolithic) fiber reinforced Timetal-21S composite at elevated temperature of about 500 and 600°C, at different loading frequencies [27]. There is no influence on the transgranular fracture mode of matrix, but the increase in temperature causes decrease in the crack growth rate. This is mainly due to enhancement of propagation of the bridging fiber/matrix interfaces by decrease in frictional shear stress.

## 4. Conclusions

1. Titanium alloy VT9 (GTM-900), exhibits continuous softening during LCF test at room temperature, increasing with increase in strain amplitude. Cyclic hardening at 500°C at low strain amplitude is essentially due to solid solution strengthening and precipitation of $Ti_3Al$ phase due to increase in the aluminum equivalent because of enrichment of oxygen and nitrogen from air.
2. Alloy IMI 685 possesses characteristic features like upper temperature limit of 550°C, high specific strength at ambient and equivalent temperature, excellent creep resistance at

temperature up to 520°C, adequate low cycle fatigue strength, stability at high temperature and stress levels, good weldability and ease of working.

3. Timetal 834, exhibits continuous cyclic softening during LCF test at room temperature. However, at 400°C it exhibits cyclic hardening following cyclic softening during the initial cycles and again exhibits softening in the later stage during macro crack propagation. The degree and rate of cyclic softening/hardening strongly depend on the strain amplitude and test temperature. C-M relationship showed a bilinearity at room temperature. It showed only linear relationship at 400°C and 600°C over the entire range of strain amplitude investigated.
4. Alloy Timetal 834, in the ( α+β ) Solution Treated-Air Cooled-Aged condition, exhibits rapid cyclic softening in the initial stage (<10% of fatigue) and saturation of stress for the major fraction of fatigue life (>90%), at 600°C .At 600°C, the initial softening occurs due to planarity of slip, shearing of the ordered $Ti_3Al$ precipitates and coarsening of silicide's. The saturation in cyclic stress occurs from the balance between the softening and hardening effects. Strengthening occurs due to partial recovery of ordering in the disordered $Ti_3Al$ precipitates, sheared by leading dislocations; solid solution strengthening of the matrix by oxygen and nitrogen enrichment and consequent precipitation of $Ti_3Al$ precipitates. Thus fatigue life at 600°C is enhanced and is not reduced even at low strain amplitudes unlike that at room temperature.

5. The improvement in ductility by alloying OF titanium aluminides has been brought out by
6. introduction of a more ductile phase. Much attention has been paid for development of Ti based aluminides because of their higher specific strength. Titanium aluminides, TiAl (γ) has higher potential for high temperature applications than its counter part $Ti_3Al$ ($\alpha_2$).

7. TMCs have been found to possess higher fatigue life and higher temperature stability than the monolithic titanium alloys. Several mechanism play role in the failure of TMCs..

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