Recent progress in the high-cycle fatigue behaviour of γ-TiAl alloys

T.E.J. Edwards

Department of Materials Science and Metallurgy, University of Cambridge, Cambridge, UK

ABSTRACT
The high-cycle fatigue (HCF) properties of γ-TiAl (gamma titanium aluminide) alloys are reviewed, particularly with regards to the deformation mechanisms active in the near-threshold cyclic loading regime. By examining the influence of lamellar orientation and thickness on the HCF threshold, in addition to more conventional microstructural considerations such as the grain size or the volume fraction of lamellar colonies, factors to improve the γ-TiAl microstructure for HCF are assessed. Finally, experimental methods and loading strategies are surveyed to identify techniques for improving HCF testing of γ-TiAl alloys. In this, we consider both the conservatism of differing approaches and the possibility to measure with suitable resolution the local mechanical behaviour under HCF of the lamellar γ-TiAl microstructure.

Introduction
Gamma titanium aluminide (γ-TiAl) alloys are emerging as a viable lightweight replacement to nickel superalloys, for use in both the turbine and compressor stages of aero-engines [1,2], as well as in high performance automobiles [3] and the nuclear industry [4]. By mid-2017, the γ-TiAl alloy Ti–48Al–2Cr–2Nb (at.-%) has seen 6 years of commercial flight aboard the GEnx engines (General Electric, U.S.A.) [1]. Pratt & Whittey, with MTU, is also using the third-generation forgeable TiAl alloy in their new geared turbofan engine [1]. For almost three decades, much of the research into these alloys has focused on the low ductility and toughness that limits the tolerance to fatigue crack growth (FCG). This results in a steep slope of the Paris region for FCG when compared to conventional superalloys and steels. In turbines, high-cycle fatigue (HCF), where the number of cycles to failure, \( N_f \), is greater than \( 10^4 \), is due to aerodynamically induced vibrations at a component’s inherent resonant frequencies, as a result of changes in rotational speeds. This HCF therefore exists in addition to the steady state or low-cycle fatigue (LCF – \( N_f < 10^4 \)) loading that occurs upon engine start up and shut down. As a result of the low FCG tolerance, the most promising approach to fatigue lifing for HCF in γ-TiAl alloys has been to operate with stress intensity factors below the FCG threshold, \( \Delta K_{th} \), commonly at 50%[5] of the maximum peak cyclic loading stress at which the material can endure \( 10^7 \) cycles without failure, known as the run-out strength. Below this stress, the FCG rate should be vanishingly small (< \( 10^{-10} \) m cycle\(^{-1} \)).

For the next generation of titanium aluminide alloys, higher operational stresses and temperatures, to 750°C or higher, would enable a significant number of components in commercial aero-engines to be replaced by γ-TiAl alloys [2,6]. To direct the improvement of γ-TiAl alloys for HCF, a complete understanding of the mechanisms by which microstructures resist crack nucleation and growth is required. Similarly, an increased consideration of the complex HCF loading conditions that the component must withstand is required to optimise the texture of γ-TiAl alloy microstructures this top-to-bottom approach has been reviewed recently [7].

Several reviews of the fatigue properties of γ-TiAl alloys exist [8,9]. However, most of the understanding of FCG is focussed on plasticity in the bulk and at crack tips during LCF; HCF is treated using conventional approaches to fatigue life evaluation [10,11]. The present review aims to assess the potential for less conventional methods and considers progress in the understanding of the HCF behaviour of γ-TiAl alloys since ~ 2010. This review begins with a brief overview of conventional HCF characterisation of TiAl alloys in ‘Conventional, Macroscopic Analysis of the HCF Behaviour of γ-TiAl Alloys’ section, in which the effect of loading and test conditions on the HCF behaviour is discussed. Fatigue crack nucleation mechanisms are covered in ‘Fatigue Crack Nucleation Mechanisms in γ-TiAl Alloys’ section. Following this, considerations
Table 1. Characteristics of the three regimes of FCG, adapted from [10].

| Stage I: Near-threshold | Stage II: Paris regime | Stage III: High-growth rate |
|-------------------------|------------------------|---------------------------|
| Microscopic failure mode| Single shear mechanism | Duplex slip, eventually striations | Various static modes |
| Crack closure levels    | High                   | Low                        | –                        |
| Load ratio effects      | Large                  | Small                      | Large                    |
| Microstructural effects | Large                  | Small                      | Large                    |
| Environmental effects   | Large                  | Variable                   | Small                    |
| Near-tip plasticity     | \( r_c \leq d_g \)    | \( r_c \leq d_g \)          | \( r_c \gg d_g \)         |

\( r_c \) and \( d_g \) refer to the cyclic plastic zone size and the grain size, respectively.

Owing to their high Paris slope, \( \gamma \)-TiAl alloys are commonly operated in Stage I, where environmental and microstructural effects on the FCG rate are high. \( r_c \approx (1/\pi)(\Delta K/2\sigma_y)^2 \) for plane stress loading.

Conventional, macroscopic analysis of the HCF behaviour of \( \gamma \)-TiAl alloys

Damage-tolerant design

The Paris law of FCG [12] considers the rate of crack growth, \( \frac{da}{dN} \), as a function of the applied stress intensity factor, \( \Delta K = K_{\text{max}} - K_{\text{min}} \), according to

\[
\frac{da}{dN} = C_p \Delta K^n
\]

where \( C_p \) is a constant, measured crack growth coefficient and \( n \) is the dimensionless Paris exponent. On a logarithmic plot (Figure 1), a central, linear section of a sigmoidal FCG dataset exists, beyond the threshold stress intensity \( \Delta K_{\text{th}} \) and before the onset of unstable crack growth at \( K_{\text{IC}} \). The gradient of the straight line fit to this central region equates to the Paris exponent (Figure 1). The characteristics of the three regions of the Paris curve are shown in Table 1.

For many structural materials in the aero industry, a damage-tolerant design is favoured, by which the propagation of cracks occurs but at a sufficiently slow rate for the rate of crack growth to be measured. This requires the slope of the Paris regime of FCG to be sufficiently low (\( \sim 3–5 \)) that changes in the stress intensity factor do not cause a rapid increase in crack growth rate.

However, titanium aluminate alloys have a Paris exponent of \( n \sim 10–50 \) (see Figure 2), depending on the microstructural condition [13]. This was identified early on for second-generation TiAl alloys and complicates their use in the Paris regime, as a 1 MPa m\(^{1/2}\) increase in the stress intensity range can cause the crack growth rate to rise by a factor of over \( 10^2 \) (Figure 2). The reasons for and the effects of such a high Paris slope are often misunderstood. Indeed, though the Paris slope is often equated to that of engineering ceramics [14], the toughness of lamellar \( \gamma \)-TiAl alloys can reach 30 MPa m\(^{1/2}\) – part-way between the few MPa m\(^{1/2}\) of specific to lamellar \( \gamma \)-TiAl microstructures, such as the orientation of lamellar colonies (i.e. the packets of co-planar \( \alpha_2 \)-\( \text{Ti}_3\text{Al} \) and \( \gamma \)-TiAl lamellae that are each formed from a single high temperature disordered \( \alpha \)-TiAl grain) to the loading axis and the lamellar thickness, are discussed in ‘Lamellar TiAl-Specific Considerations for Crack Growth in HCF’ section. Finally, the possibility for the application of novel test methods to study the HCF behaviour of \( \gamma \)-TiAl alloys down to the nanometre scale, and with loading procedures that generate deformation fields typical of HCF conditions in industrial applications, will be covered in ‘Outlook for HCF Research on TiAl Alloys’ section.
ceramics and the \( \sim 200 \, \text{MPa} \, \text{m}^{1/2} \) that high strength steels can achieve. The ductility of \( \gamma \)-TiAl alloys is also greater than that of engineering ceramics, being of the order of 1% strain. To better understand the mechanisms behind crack growth in this intermetallic alloy, one should consider the modified Paris law according to Hojo et al. [15]. It describes the FCG behaviour of both ductile and very brittle materials by accounting for both the effect of the cyclic stress intensity range and the maximum applied stress intensity:

\[
\frac{\text{d}a}{\text{d}N} = C_p\Delta K^pK_\text{max}^q
\]

(2)

For common engineering metals, \( p \gg q \), whereas for toughened ceramics where some cyclic loading may be achieved, \( p \ll q \), such that the maximum loading stress is the most damaging aspect of fatigue [16]. Intermetallic materials, such as \( \gamma \)-TiAl alloys, lie somewhere in between, with \( p \sim q \), such that values of \( n = 50 \), obtained by fitting experimental data to the Paris law are gross over-approximations of the dependence of the FCG rate on the cyclic stress range itself. Filippini et al. [17] demonstrated that neither \( \Delta K \) nor \( K_\text{max} \) alone can fully describe the FCG properties of Ti–48Al–2Cr–2Nb (at. % henceforth). Similarly, tests at fixed \( \Delta K \) on the same alloy by Dahar et al. found the Paris slope \( n \) to increase with \( K_\text{max} \), due to static failure modes being activated at higher \( K_\text{max} \).

Recently, there has been renewed interest in the use of TiAl alloys in the Paris regime, driven by the wish to use the material in the higher temperature parts of the aero-engine [2]. This requires the material to withstand temperatures up to 850°C [6] and higher stresses. Hence, an improved understanding of, and potential operation within, the crack growth regime (i.e. Stage II) is essential. FCG data of region II fatigue of Ti–48Al–2Mn–2Nb [19,20] shows that both the crack growth prefactor, \( C_p \), and the Paris slope, \( n \), decrease significantly above 750°C (the latter to \( n = 4 \)), suggesting that once the ductile–brittle transition is exceeded, damage-tolerant design may be achievable if sufficient strength can be maintained. It has been reported that the Paris slope is lower when the lamellae in a TiAl lamellar colony are oriented parallel to the loading axis [21]. Recently, bulk single colony samples have emerged where the orientation of the lamellae can be controlled by the processing method [22], without seeding. This allows components to be produced where the lamellae lie parallel to the loading axis throughout and may result in a further renewed interest in lifing by damage tolerance; this is discussed further in ‘Fatigue of Polysynthetically Twinned Crystals’ section.

**Effect of the processing method on damage tolerance**

Early studies by Gloanec et al. [23] on powder metallurgy (PM) consolidation by hot isostatic pressing (HIPping) of \( \gamma \)-TiAl alloys revealed that the FCG rate in the Paris regime for the Ti–48Al–2Cr–2Nb alloy was approximately two orders of magnitude higher than the equivalent cast material, even after correction for crack closure. This was attributed to the PM processed material possessing a microstructure of fine equiaxed \( \gamma \) grains that was less able to resist FCG compared with the cast, HIPped and heat treated material where coarse lamellar colonies gave rise to a tortuous crack path [23]. Further analysis of this dataset by Zuo et al. [24] identified the effective Paris slope (\( m \) in ‘Crack Closure’ section) to be independent of stress ratio and similar between both cast and PM alloys, 8.11 and 8.39 respectively. Rather, it was the \( C_p \) prefactor of Paris regime crack growth that was the processing and microstructure-dependent variable [24] and caused the reduction in fatigue behaviour of the PM material. However, this effect was less marked at higher stress ratios where roughness-induced crack closure effects (see ‘Crack Closure’ section) are less significant.

The development of additive PM processing methods for \( \gamma \)-TiAl alloys has also brought about renewed investigation of fatigue in region II. The application of electron beam melting (EBM) additive manufacturing, a patented advanced PM process [25], to second-generation Ti–48Al–2Cr–2Nb and third-generation Ti–(45–47)Al–8Nb–2Cr (TNB) has enabled near-net shape blade parts to be produced with minimal material wastage compared to cast parts [26]. EBM decreases the porosity and other typical defects compared to casting, and increases the lamellar colony content compared to conventional PM material. This was found to produce superior FCG characteristics, although only the \( \Delta K_{\text{th}} \) was higher (\( \sim 6 \, \text{MPa} \, \text{m}^{1/2} \) at a stress ratio \( R = 0.05 \) for EBM and 5 MPa \( \text{m}^{1/2} \) at \( R = 0.1 \) for PM) and the Paris slopes were similar.

**Stress-life approach**

The stress-life (S-N) data compiled by Larsen et al. [27] on second-generation \( \gamma \)-TiAl alloys indicates that across a range of microstructures, the maximum run-out strength (10\(^7\) cycles), \( \sigma_{\text{max,th}} \), remains above \( \sim 80\% \) of the ultimate tensile stress, \( \sigma_{\text{UTS}} \). Furthermore, the maximum stress of HCF loading for \( \gamma \)-TiAl alloys can exceed the initial yield stress of the material, irrespective of the stress ratio. This holds even for recent high niobium alloys [28]. Even at target operational temperatures, e.g. 700°C, the maximum run-out stress is at least 70% of the ultimate tensile strength, as a result of a decreased yield strength and increased ductility, at high temperatures, without any substantial change to the work hardening behaviour. This indicates that some degree of plastic cycling must occur in TiAl alloys cycled at the maximum run-out stress across the temperature range, as they are loaded above the elastic limit. This will be discussed further in ‘Towards a
Microscopic Model for HCF Loading of TiAl Alloys: a Focus on Plasticity’s section.

The emergence of gigacycle test methods [29,30] has meant that S-N measurements are no longer limited to $10^7$ cycles; $10^{10}$ is now commonly achievable. Testing of the alloy Ti–45Al–10Nb [31,32] revealed that by $10^8$ cycles at a stress ratio of 0.1, an endurance limit was reached, although this is not seen in samples tested at $R=0.5$ which lasted for $10^{10}$ cycles (Figure 3). Considering the nature of HCF by resonant vibrations in turbine blades which are already experiencing a steady-state centrifugal load, the apparent absence of an endurance limit even by $10^{10}$ cycles at high $R$ is prohibitive to conservative lifing to a stress well below a run-out value.

**FCG threshold**

The standard method for determining the FCG characteristics of γ-TiAl alloys has been to test at a frequency of 20 Hz and a stress ratio $R=0.1$, with progressive load shedding to measure the threshold stress intensity according to ASTM standard E647 [11]. Features of conventional fatigue threshold measurements commonly encountered on γ-TiAl alloys, such as stress ratio dependence and small crack effects, are discussed in the following subsections.

**Crack closure**

The dependence of the fatigue threshold on the stress ratio, $R = \sigma_{\text{min}}/\sigma_{\text{max}}$, for γ-TiAl alloys can be accounted for by considering the possibility for crack closure effects, first proposed by Elber [33]. For each stress intensity range $\Delta K$ an effective range $\Delta K_{\text{eff}}$ is considered in the Paris law treatment [10,34]:

$$\frac{da}{dN} = C_{\text{p}} \Delta K_{\text{eff}}^m$$

where $m=p+q$, $U$ is a material-dependent function that depends mainly on $R$ and $K_{\text{max}}$, and $K_{\text{op}}$ is the stress intensity factor at which a crack opens fully. $K_{\text{cl}}$ is the stress intensity at which the crack begins to make closure contact upon unloading and in practice $K_{\text{cl}} \approx K_{\text{op}}$ [10]; this occurs at a critical stress ratio $R = R_c$. The dependence of the fatigue crack threshold $\Delta K_{\text{th}}$ on $R$ resulting from crack closure can further be accounted for to produce an effective fatigue crack threshold, $\Delta K_{\text{th,eff}}$, which is independent of $R$ and a material property [35], according to

$$\Delta K_{\text{th,eff}} = K_{\text{cl}} + \Delta K_{\text{th,eff}}$$

for $R < R_{\text{cr}}$, $K_{\text{min}} < K_{\text{cl}}$ and $K_{\text{max}} = \frac{\Delta K_{\text{th}}}{1 - R}$

for $R \geq R_{\text{cr}}$, $K_{\text{min}} \geq K_{\text{cl}}$ and $\Delta K_{\text{th}} = \Delta K_{\text{th,eff}}$

Hénaff et al. [36] applied this model to a near-γ alloy. They demonstrated that the function $U$ was approximately constant across the $\Delta K$ range of testing in both air and vacuum at room temperature. The environment therefore interacts only weakly with the crack at this temperature and $K_{\text{op}}$ is constant. Similar results were obtained on lamellar and duplex structures [37]. The reasons for crack closure can be diverse [10] and are material dependent [16]. The extrinsic and intrinsic processes occurring at a crack tip loaded in fatigue for metals, intermetallics and ceramics are summarised in Figure 4. For γ-TiAl, roughness-induced closure appears to be a key mechanism for lamellar microstructures, where tortuous translamellar crack deflection results in a serrated fracture surface, particularly for larger colony sizes [18]. Roughness-induced crack closure does not occur in lamellar γ-TiAl alloys for stress ratios above $\sim 0.5$, and also in the cases of near-γ and duplex structures where smooth fracture surfaces cause only crack tip plasticity, not roughness, to bring the crack surfaces into contact sooner [36]. If fretting occurs alongside oxidation of the crack surfaces, then an improvement in the near-threshold fatigue properties is found at 800°C, which are better than those at both 25°C and 600°C [38,39].

**Small cracks**

Since the early 1990s, it has been recognised [40] that small cracks (<1 mm) are of particular concern in the fatigue of γ-TiAl alloys. Indeed, the concept of a threshold stress intensity range $\Delta K_{\text{th}}$ below which crack growth is vanishingly low (< $10^{-10}$ m cycle$^{-1}$) occurs only for long cracks, which obey linear elastic fracture mechanics (LEFM). However, short cracks have been observed to propagate in γ-TiAl alloys at stress intensity factors below the $\Delta K_{\text{th}}$ determined for long cracks, and at rates which are several orders of magnitude faster than predicted using stress states inferred from LEFM [40]. The conservative design limit for fatigue protection using $\Delta K_{\text{th}}$ is therefore not useful for short cracks. The reason for this effect has been demonstrated in a study of 25–500 μm surface cracks.
Figure 4. Mechanisms for FCG in intermetallics, compared with those present in metals and ceramics. Extrinsic toughening in intermetallics may be achieved by grains bridging cracks, such as bridging lamellae in lamellar $\gamma$-TiAl alloys. In ceramics and intermetallic composites, crack bridging may additionally involve strong and stiff fibres, or elongated metallic grains that undergo ductile rupture [16]. Ahead of the crack in $\gamma$-TiAl intermetallics, crack growth may occur by cleavage fracture along fatigue cycled slip planes and by coalescence of microcracks ahead of the crack tip; static fracture modes also occur. Microcrack/microvoid toughening occurs in the proximity of cracks. Adapted from [16] (reproduced with permission).

Figure 5. Small crack growth rates in TiAl alloy TNM-B1, with the minimum in FCG rate at the first microstructural barrier arrowed in red. Adapted from [46] (reproduced with permission).
To understand the size of flaws that might be generated to the production of TiAl components, it is necessary for design against fatigue failure due to flaws inherent in addition to understanding the material parameters forereportedtodominatetheHCFpropertiesof TiAl alloys. The effect of temperature on the micromechanisms of fatigue crack initiation and growth has been reviewed extensively elsewhere; however, little further progress has been made since towards understanding the effect of temperature on the microstructural dimension of lamellar colonies, that should be considered for component design and lifing. Indeed, above this stress intensity range, the small cracks grew rapidly, see the minimum in FCG rate arrowed in Figure 5. For most of the alloys investigated, this threshold was approximately half that determined from long crack growth that is usually used for lifing [5].

An alternative useful approach to accounting for the presence of small cracks in TiAl alloys is that of Filippini et al. [17,47], where the El-Haddad model modified by Tanaka et al. [48] was applied. This combines the stress-life and damage-tolerant (FCG) approaches to fatigue by the intermediate of an inherent minimum flaw size, which has previously been related to the concept of small cracks [10]. The intrinsic fatigue stress intensity range threshold was determined from compact-tension specimens and the maximum run-out stress values were determined from plain cylindrical specimens with and without EDM-introduced artificial defects, at room temperature and 700°C. Both the Ti–48Al–2Cr–2Nb and TNB alloys were found to obey the same Tanaka law (Figure 6), despite different strengths and colony sizes. A similar percentage of lamellar colonies were produced in both alloys through electron beam melt processing. This parameter, the volume fraction of lamellar colonies, was therefore reported to dominate the HCF properties of TiAl alloys.

**Object damage**

In addition to understanding the material parameters for design against fatigue failure due to flaws inherent to the production of TiAl components, it is necessary to understand the size of flaws that might be generated by object damage (foreign or domestic) during engine operation. There are relatively few published studies concerning object damage of γ-TiAl alloys. Early studies by Harding and Jones [49] employed quasi-static and low-speed drop indentations. Ballistic impact testing by Draper et al. [50,51] using a gas gun to project small steel spheres at aerofoil-shaped specimens at fixed stress intensity across the same temperature range of temperature appears to be complex. For instance, although the failure mechanism of brittle trans-inter-lamellar fracture is unchanged from 25 to 800°C, the alloy Ti–47.7Al–1.9Nb–0.9Mn + 1 vol.-% TiB2 has a higher threshold stress intensity factor in air at 25°C and 800°C than it does at 600°C [39] (Figure 7a). In a similar study [61] of FCG rates at fixed stress intensity across the same temperature range, an increase in temperature was found to reduce the measured reduction in fatigue strength [51]. Attempts to model this behaviour with a finite element model [52] identified that a lower amount of ductility occurred before crack formation in ballistic testing (1–2% plastic strain) compared with static indents (2–3% plastic strain).

More recently, Draper et al. have extended their study to third-generation γ-TiAl alloys [53]. The GMPX alloy Ti–45Al–5Nb–0.2C–0.2B displayed no improvement in the remnant fatigue strength after impact testing compared with older generation alloys such as Ti–48Al–2Cr–2Nb, hence calling into question the suitability of third-generation γ-TiAl alloys for higher strength aero-engine blade applications. Bache and Morgans [54] have undertaken simulated object damage of Ti–46Al–8Ta at much higher impact energies and using cube-shaped steel projectiles. Damage was often brittle, with surface chips ejected, and deep macroscopic cracks (>1 mm) surrounding the damage site. Further HCF of the damaged specimens at $R = 0.1$ with an increasing $\sigma_{\text{max}}$ until failure found that comparable HCF endurance strengths were achieved at 20°C and 650°C, although only a few samples were tested and up to a maximum of $10^5$ cycles. A plateau in fatigue strength was reached beyond an impact energy of $\sim 2.4$ J, which suggests there exists a maximum flaw size to engineer for, however, complete specimen failure occurred upon an impact energy of 9.8 J.

**Effect of temperature and environment**

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range on Ti–46.5Al–3Nb–2Cr–0.2W (Figure 7b), the mechanisms involved were thought to be associated with both increased ductility upon heating, and also increased environmental embrittlement in the crack region, which nominally decreases fatigue life. Growth rates are several orders of magnitude higher in air than in a vacuum [62]. However, up to 600°C, oxide formation is rarely reported, suggesting that oxidation-induced crack closure is not significant. An alternative mechanism for environmental embrittlement is by hydrogen uptake from water vapor. This occurs in iron aluminides [62], though the mechanism behind the temperature dependence of this is not well understood. Elsewhere, strain ageing effects in the 400–650°C range may reduce crack tip plasticity due to increased dislocation pinning by TiAl + VTi, an antise & vacancy pair, which has a negative strain-rate dependence [9].

Figure 7. Effect of temperature on FCG behaviour: (a) from [39] and (b) from [61] (reproduced with permission).

Fatigue crack nucleation mechanisms in γ-TiAl alloys

In situ SEM observations have found that cracks in γ-TiAl alloys initiate in fatigue during the latter stages of HCF life, generally within the final 10% of the lifetime [63]. However, acoustic emission measurements have found that even within the first loading cycle microcracking can initiate [64,65]. There is evidence [9,66] that the growth of fatigue cracks in γ-TiAl alloys operates by the formation of microcracks ahead of the growing crack. This can be seen as many crack initiation events that link up by more ductile-type translamellar rupture. Furthermore, the initiation of cracks has been observed in situ on electropolished surfaces, where no flaws were apparent, although laser ablation-formed micronotches elsewhere on the surface remained inactive [46]. From this evident complexity and preponderance of crack initiation in the HCF failure of γ-TiAl alloys, it is clear that understanding the mechanisms for crack nucleation is crucial to optimising the HCF properties.

The nearly flat S-N curve that is commonly reported for γ-TiAl alloys has been shown by Jha et al. [67] and elsewhere [68] to be characteristic of crack initiation both at the surfaces, and in the bulk. In itself, this distinction of flaw location is not novel. Weibull statistics have been used on brittle materials to demonstrate these two regimes [69]. The further consideration for γ-TiAl alloys is whether the mechanism of initiation is different. Although crack nucleation is common at internal inclusions and pores in ductile engineering alloys in the gigacycle regime [30], cracks in gigacycle fatigue tests of more brittle γ-TiAl alloys were also identified to be volume initiated [32]. This was in particular at γ/α2 interfaces and was attributed to plastic incompatibility.

At a surface, fatigue cracks may form by the extrusion of soft-mode oriented lamellae (where the angle of the lamellar interface to the loading axis, Φ, see Figure 8, lies between 15° and 75°) by plasticity in the γ-TiAl phase parallel to the lamellar interfaces, known as longitudinal plasticity [70,71]. This is an initiation mechanism commonly found in ductile systems [10] and requires the local critical resolved shear stress to be exceeded. In the bulk, this mechanism may be resisted by the constraint of the surrounding colonies [72]. Nevertheless, deformation twinning in the γ-TiAl phase can generate grain boundary decohesion at a free surface in equiaxed γ and lamellar microstructures [73,74]. Ahead of static and fatigue cracks in lamellar structures, transverse twinning in the γ-TiAl phase causes lamellar interfaces to debond [66,75].

Another crack initiation mechanism in the bulk is that defined by Cottrell [76], whereby in semi-brittle crystals, the intersection of slip planes can form an obstacle to slip, and a brittle-type crack nucleates at the obstacle. This occurs when the elastic
energy accumulated there by dislocation pile-up is sufficient to drive the formation of the free surfaces of a crack. This mechanism is thought to nucleate microcracks identified by TEM along the low energy \{111\}_γ cleavage planes at the intersection between the easy operating longitudinal slip systems and the Hall–Petch strengthened transverse slip systems in lamellar \(γ\)-TiAl microstructures \[77–79\]. In this sense, globular, non-lamellar, \(γ\) grains in nearly lamellar TiAl microstructures may initiate microcracks as the lack of lamellar interface strengthening permits relatively unimpeded slip on multiple planes. Hence significant pile-up forces at intersection obstacles may arise. This has not been extensively investigated.

A final crack initiation mechanism in lamellar structures is decohesion of lamellae oriented perpendicular to the loading axis (\(Φ = 90°\)) \[65,80\]. The scale of cracking achievable before any microstructural barrier is met, and the regularity of the occurrence of this initiation mechanism strongly suggests action should be taken to either strengthen the lamellar interfaces, reduce the colony size to limit the spatial extent of damage or avoid such colony orientations altogether.

Crack initiation in the latest generation high Nb alloys in both quasi-static and fatigue loading \[81,82\] also involves the retained ordered \(β_0\)-TiAl with brittle \(ω\)-Ti phase precipitates that concentrate stress. This becomes particularly problematic when the \(β_0\) phase covers large amounts of lamellar colony boundaries, such that extensive microcrack nucleation in \(β_0/ω\) may result in extended planes of coalescing flaws. Improved thermodynamic stabilisation of the \(β_0\) phase is required to avoid this.

**Lamellar TiAl-specific considerations for crack growth in HCF**

**Microstructural dimensions appropriate to HCF characterisation**

Classically, the \(γ\) grain size, or in the case of lamellar \(γ\)-TiAl alloys, the colony size, is the main microstructural dimension investigated towards optimising HCF properties. This is then supplemented by a consideration of the proportion of equiaxed \(γ\) grains versus lamellar colonies, the fatigue threshold generally increasing with decreasing equiaxed \(γ\) content \[83\] (Figure 9). The size of lamellar colonies influences the FCG rate as smaller colonies have a high density of colony boundaries. Hence cracks only propagate a small distance before being deflected at a boundary. However, as Dahar et al. \[18\] have pointed out, a large colony size maintains a cyclic crack tip plastic zone within the colony of initiation for as long as possible in the near-threshold regime. Hence, by not exceeding stage I type growth, surface roughness-induced crack closure is encouraged by the single-slip-like behaviour \[84\]. Therefore, although a finer microstructure limits initial crack propagation, once a short crack attains the size of several colonies, it acts as a long crack, where the presence of many weak colony boundary interfaces facilitates crack propagation \[37\]. It seems generally agreed that the optimum colony size is in the 50–100 \(\mu\)m range for several...
second- and third-generation $\gamma$-TiAl alloys \[85,86\]. In fact, according to Filippini et al. \[47\], the intrinsic effective threshold stress intensity range is a material property that cannot be much affected by grain size, as crack growth at the limiting rates for the application of $\gamma$-TiAl alloys is very much subgranular.

In lamellar $\gamma$-TiAl alloys, the microstructure is rarely considered in terms of the lamellar thickness with regards to HCF properties. It was generally thought that the lamellar thickness is related to lamellar colony size, being proportional to its square root \[87\]. Based on quasi-static toughness measurements alone, the influence of lamellar thickness is unclear: several studies \[88,89\] indicate that lamellar refinement increases fracture toughness, whilst another \[90\] found no such relationship to exist. This was possibly due to the variation of the colony size in these studies. However, a systematic study by Cao et al. \[91\] employing multistage cooling strategies demonstrated that the colony size and lamellar thickness can be varied independently, over several orders of magnitude. This method was adopted by Mine et al. \[92\] to produce polycrystalline single edge micronotched specimens with lamellar thicknesses of 0.91 μm and 3.8 μm, for a range of colony sizes. Materials with refined lamellae had lower FCG rates than those with thick lamellae. This was attributed to the high degree of microcracking in neighbouring lamellae ahead of the crack tip in the refined material. Once linked, such microcracking led to heavily roughened crack faces, and hence increased closure. In material containing thicker lamellae, there is not as much microcracking \[93\]. However, the wider bridging ligaments when microcracking does occur give rise to increased scatter in HCF behaviour compared with the refined lamellae condition. In contrast, Chan and Kim \[88,89\] found that refining the lamellae was associated with an increase in the resistance to translamellar microcracking. Plasticity upon fatigue loading generates transverse twins and slip bands that are known to be preferred propagation paths \[78\]. The CRSS of such transverse deformation mechanisms is dependent on lamellar thickness (Hall–Petche effect \[91\]) and composition; this may explain the above discrepancies. Compositional differences between the studies may have also influenced the ease of debonding lamellar interfaces. A decreasing Al content has been demonstrated to refine the lamellar thickness, as well as reducing the in-plane $\gamma$ domain length, for equivalent processing conditions \[94,95\]. Under LCF conditions, Umakoshi et al. \[70\] found the strengthening of this domain length refinement to resist longitudinal plasticity, which otherwise causes the formation of surface microextrusions of lamellae that act as microcrack nucleators, as evoked in ’Fatigue Crack Nucleation Mechanisms in $\gamma$-TiAl Alloys’ section.

Finally, long-term thermal exposure of Ti–45Al–2Mn–2Nb–0.8 vol.%TiB$_2$ at 700°C for periods of up to 10000 h \[96,97\] was found to improve the maximum fatigue strength at run-out by almost 100 MPa, but only beyond 6000 h of exposure. Appropriate variations in testpiece production methods showed that this apparent strengthening did not result from surface healing effects, but rather from stress relaxation in the bulk; the dominant microstructural change evident from imaging was simply transformation of the $\alpha_2$ phase to $\gamma$ and hence a thinning of the $\alpha_2$ lamellae. This suggests that increasing the volume fraction of $\gamma$ phase, presumably bringing this closer to the thermodynamic equilibrium, might improve the stress-life properties of lamellar $\gamma$-TiAl alloys.

**Fatigue of polysynthetically twinned crystals**

Polysynthetically twinned (PST) crystals were used to study the mechanical properties of lamellar $\gamma$-TiAl alloys, mostly in Japan in the 1990s and early 2000s \[70,98,99\]. Samples were grown using a floating zone technique with either seeding or initial narrowing of the floating zone to produce a single $\alpha$ phase grain that transforms to a single stack of $\gamma$-TiAl/$\alpha_2$-Ti$_3$Al lamellae, effectively a single lamellar colony, upon cooling. The majority of data collected concerned monotonic loading and strain-controlled LCF across the temperature range, as the study of plasticity mechanisms and the effect of the anisotropy caused by lamellar orientation to the loading axis, $\Phi$, were the focus. An extensive review by Umakoshi et al. \[70\] (Figure 10) nevertheless provides some insight as to optimum lamellar orientations for HCF testing: $\Phi = 0^\circ$ performs better than $\Phi = 45^\circ$, mainly due to the reduced strength of the $\Phi = 45^\circ$ orientation.

An alternative to growing bulk PST crystals is to produce micromechanical testpieces from individual

![Figure 10. Stress-life testing of PST crystals. From \[70\] (reproduced with permission).](image-url)
lamellar colonies by selective machining. This has been achieved on TiAl alloys by micro-electrodischarge machining (μEDM) [100] and focussed ion beam milling (FIB) [100–104]. It has enabled the critical resolved shear stress of deformation mechanisms to be measured that might not otherwise be individually activated, e.g. twinning of γ-TiAl. However, studies to date have been limited to monotonic testing between 25 °C [100–103] and 700 °C [104]. It is possible for in situ SEM measurements of stable crack growth to be made with micromechanical methods [105]. However, for meaningful FCG data to be achieved, the plastic zone should be considerably smaller than the testpiece dimensions, and though this is achievable for brittle ceramics [105], the plastic zone radius of γ-TiAl alloys in fatigue is several microns [106], which is at the limit for a micromechanical testpiece.

An intermediate between PST crystals (Figure 11a) and equiaxed, atextured polycrystalline microstructures is a columnar microstructure. For lamellar γ-TiAl alloys, there are two types of columnar-grained structures that can be developed that have a same Φ in all colonies [107]. The first have the lamellar planes of neighbouring colonies almost parallel to each other (Figure 11 b), the second only have equal tilt angles of the lamellar planes to the axis of the columnar growth (Figure 11 c). Multiple groups [108–111], mainly based in China, have pursued the directional solidification route on application-relevant alloy compositions to develop such columnar microstructures. Control of lamellar orientations was achieved by varying the extraction rate from a Bridgman-type furnace [108], and employing different α- or β-Ti [111] seeding techniques and multiple re-solidification [109]. Limited fatigue testing of such microstructures has been undertaken by both S-N [112] and compact-tension FCG [113] methods. For testing of Φ = 0° and Φ = 90° lamellar orientations, the closure-corrected fatigue threshold was identified to be equal in both cases, at ~ 5 MPa m^{1/2}, but the Paris slope was much lower in the Φ = 0° case (m ~ 5 vs. m ~ 32 for Φ = 90°); this is consistent with increased ductility and the impossibility for growth of the primary crack along lamellar interfaces in the Φ = 0° case.

The culmination of such directional solidification endeavours enabled Chen et al. [22,114] to produce macroscopic PST crystals of Ti–45Al–8Nb several tens of millimetres long. The lamellar orientation was determined by the extraction rate from the Bridgman furnace and not by prior seeding. The measured monotonic strength was high for the alloy system, and ductility reached 7% in the Φ = 0° orientation at room temperature. This method has potential for the mass production of effectively single crystal turbine blades, similarly to the single crystal nickel superalloy counterpart [114,115]. As expected, the creep strength was considerable [22]; however, data on the HCF behaviour are yet to appear.

Unfortunately, the optimum microstructure for high strength HCF properties, combined with a tolerance for FCG, is unclear. Within the context of preferentially oriented lamellar microstructures alone, there is no evidence as to whether columnar grains or PST (Figure 11) will perform better. From the absence of colony boundaries in PST crystals, their creep properties are likely to outperform their columnar-grained counterparts for the same composition and lamellar thickness. Failure by delamination at the free surfaces is known to occur from the machining of lamellar TiAl alloys [116], but is rarely investigated in the context of HCF. The varied combinations of mechanical loading – tension, torsion, bending – experienced by a component in fatigue may induce such delamination. Hence the possibility for a choice of orientations of the columnar grains could locally increase resistance to this failure mode, whereas PST crystals have no such flexibility (e.g. maintaining Φ = 0°, but changing the twist angle of the lamellae about the blade axis, Figure 11 c).

Fatigue cracking of specific colony orientations in polycrystals

Studies on PST crystals have highlighted the impact of lamellar orientation on the fatigue behaviour of lamellar TiAl alloys and the potential for the production of components from an optimum orientation. However, the majority of developmental and applied alloys are polycrystalline. Throughout the 1990s and 2000s, there was a considerable drive to develop compositions and processing methods that resulted in texture-free microstructures [117], where a growing fatigue crack is therefore likely to encounter all possible colony
orientations. The growth of PST crystals from arbitrary TiAl compositions to produce fatigue testpieces with a variety of lamellar orientations is a complex and onerous task. Further, it does not enable the study of changing lamellar orientations across colony boundaries on the FCG. Quite simply, PST crystals do not contain colony boundaries.

Over recent years, this has inspired efforts to establish the HCF properties of individual lamellar colonies in polycrystalline samples by the careful positioning of EDM or diamond wire pre-cracks [18, 63, 79, 118]. The measurement of $\Delta K_{th}$ of Ti–46Al–8Nb for lamellar orientations $\Phi$ of 0–75° at 650°C [118] determined the most FCG resistant orientation in the near-threshold regime to be $\Phi \sim 0°$ (Figure 12a). The colony orientations with the lowest threshold values were those where $\Phi = 40–65°$, i.e. the softest orientations in PST studies [98]. The secondary orientation, $\Psi$, of the lamellar planes with respect to the advancing crack front, had no marked effect: lamellar planes at $\Phi \sim 0°$ lying parallel to the crack front, or perpendicular, yielded equal $\Delta K_{th}$ values. Translamellar fracture was the dominant fracture mode for the majority of colony orientations, except for those close to $\Phi = 90°$, where interlamellar debonding occurred. Hence, understanding what causes the lamellae to crack transversally is paramount. The similarity between the U-shaped $\Delta K_{th}$ versus $\Phi$ in Figure 12(b) and that identified for the yield stress of PST crystals against $\Phi$ in compression (Figure 12a) suggests that the development of extensive longitudinal plasticity in the softest lamellar orientations causes the opening of microcracks where this intersects active transverse slip/twinning, i.e. the Cottrell mechanism detailed in 'Fatigue Crack Nucleation Mechanisms in $\gamma$-TiAl Alloys' section. In principle, transverse deformation mechanisms are less likely to cause microcrack formation at transverse–transverse intersections as the extent of plastic strain generated on these systems is lower [103]. This correlates well with the highest $\Delta K_{th}$ being at $\Phi \sim 0°$, where the Schmid factor is very low for longitudinal mechanisms. Furthermore, in situ SEM high temperature (750°C) HCF loading of Ti–45Al–8Nb–0.2 W–0.2 B–0.1Y found the development of longitudinal plasticity in soft-mode oriented colonies caused colony boundary cracking at the ends of lamellae [63].

In another study [79], the fatigue cracking mechanism of lamellar colonies of Ti–45Al–2Mn–2Nb–1B as a function of both $\Phi$ and $\Delta K$ were determined. At $\Delta K$ values in the sub-/near-threshold range ($<10$ MPa m$^{1/2}$), colonies with $0° < \Phi < 60°$ failed by translamellar fracture with relatively smooth fracture surfaces; only crack deviation along transverse slip bands and twins generated roughness. If instead the local $\Delta K$ was higher, secondary interlamellar cracking approximately parallel to the loading axis also occurred, resulting in increased roughness of the crack surfaces, which is generally recognised as beneficial towards reducing the effective $\Delta K$ by crack closure [10, 18]. Hence the intrinsic cohesion of lamellar interfaces may be sufficient to resist interlamellar debonding as a secondary crack mechanism in the sub-/near-threshold regime. This lack of interlamellar cracking leaves the material without the benefits of crack closure in the threshold regime of crack growth for the $0° < \Phi < 60°$ orientations. Again, the secondary orientation, $\Psi$,
was found to have no impact on the material fatigue properties. It should however be noted that fracture surfaces following gigacycle fatigue testing indicated that higher amounts of secondary cracking were generated in the testpieces loaded in excess of $10^7$ cycles [32], and therefore at lower stresses, suggesting that a lower cycling stress applied for a more extended period can eventually generate secondary cracking.

The $\Phi \sim 90^\circ$ orientation has generally been viewed as the weakest colony orientation [121]. In the study by Min et al. [63], the notched $\Phi \sim 90^\circ$ colonies failed rapidly once cracking initiated (at consistently above 93% of the total life). This was followed by a much slower crack propagation in the lower $\Phi$ colonies beyond. This steady propagation regime beyond the $\Phi \sim 90^\circ$ colony was identical to crack growth when cracking instead initiated in a colony with $\Phi$ closer to 0°. Debonding of the lamellar planes may be so easy that interlamellar cracking in $\Phi \sim 90^\circ$ colonies ahead of the crack tip occurs, followed by crack expansion in the reverse direction, towards the crack tip, to sever bridging ligaments [79,113]. In contrast, Yang et al. [118] found the $\Phi \sim 90^\circ$ orientation to have an intermediate $\Delta K_{th}$ between that of $\Phi \sim 0^\circ$ (best) and $40^\circ < \Phi < 65^\circ$ (worst) (Figure 12 b). It may be that the issue of $\Phi \sim 90^\circ$ colonies in [63] was overcome by the small colony size (70 μm, relative to 1 mm in [118]) such that the fatigue threshold was not exceeded after failure of a single $\Phi \sim 90^\circ$ colony.

Another study [18] on a textured cast billet of Ti–48Al–2Cr–2Nb instead yielded no noticeable effect of $\Phi$ on $\Delta K_{th}$ or the Paris slope across a range of stress ratios, for loading at $\Phi = 0^\circ$ and $90^\circ$. This may have been due to the testpieces not being sufficiently textured, whereby actual colony orientations varied within $\sim 40^\circ$ of the nominal $\Phi$.

**Outlook for HCF research on TiAl alloys**

*Towards a microscopic model for HCF loading of TiAl alloys: a focus on plasticity*

Within the field of TiAl alloy research, it is not clear what level of plasticity is optimal. Indeed, though there is a clear consensus that room temperature ductility should be raised, that is, increasing the elongation to failure, it is not clear how this influences the HCF life. This is not in the least because of conflicting interpretations of results, whereby almost identical electron images of interlamellar movement have been interpreted in different ways: microcracking [64,65,122], interlamellar sliding to form ledges [123] or longitudinal plasticity near the lamellar interface (see Figure 13). Evidently, these mechanisms also have a varying degree of impact on fatigue crack formation and growth, as has been explored in ‘Fatigue Crack Nucleation Mechanisms in $\gamma$-TiAl Alloys’ section.

The study of the HCF behaviour of $\gamma$-TiAl alloys has generally focused on plasticity in the region of a crack tip, rather than in the bulk of the material [66,106]. This is possibly because in common engineering alloys, achieving HCF loading to over $10^7$ cycles often means operating at stresses considerably below the yield stress, so that to exceed the local critical resolved shear stress for slip, stress concentration at a crack tip, a hard particle or a crystal boundary is required. In contrast, due to the lesser amount of plasticity necessary, the predominant mechanism is one of crack initiation. This behaviour can be modelled using a stable crack initiation approach with a slightly linear dependence on stress concentrators (Figure 13).

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**Figure 13.** SEM images of similar deformation features near lamellar interfaces, interpreted as (a) ledges at lamellar interfaces [123], i.e. interfacial sliding, (b) interlamellar cracking [122], i.e. debonding of the lamellar interface, and (c) longitudinal slip in the $\gamma$-TiAl lamellae, near the lamellar interface [124]. Above (b): AFM linescan across the feature reported as interlamellar cracking; above (c): scanning transmission electron microscopy (STEM) images of a slip step near an $a_2/\gamma$ interface. White arrows indicate the features of interest at the lamellar interfaces; black arrows are illustrative of AFM scan or cross-sectional imaging slice directions (reproduced with permission).
to the fourfold strength anisotropy of lamellar γ-TiAl alloys (Figure 12 a), at $\sigma_{\text{max,th}}$, the softest colonies may be loaded well above their plastic limit for longitudinal slip and twin systems. So-called microplasticity may therefore occur throughout such colonies, regardless of the existence of other stress concentrators. Evidence for this is reviewed in the next section. Hence, the HCF cycling of lamellar γ-TiAl alloys may in fact involve a variety of deformation mechanisms, that vary from colony to colony, some of which are characteristic of LCF cycling (i.e. bulk plasticity), whilst other colonies do not deform plastically.

Current common procedure is to carry out FCG testing of γ-TiAl alloys on stress relaxed, virgin material. Some data [65] has indicated that a single loading cycle below the general yield stress can have a significantly detrimental effect on the $\gamma$-TiAl behaviour of γ-TiAl alloys. In practice, a γ-TiAl component has to sustain HCF loading of an instantaneously imposed crack, as in the simulated object damage evoked in ‘Object Damage’ section, potentially towards the end of its life cycle. Despite this, no representative investigations have been made of the effect of such extended HCF loading of initially uncracked material on the FCG behaviour once a pre-crack has been introduced. This might be achieved, for example, by loading a lamellar γ-TiAl alloy for $10^5$ cycles at $\sigma_{\text{max,th}}$, then producing an FCG testpiece and performing a conventional FCG test to identify the effect on the fatigue threshold of this pre-conditioning of the material below the general yield stress.

Since the establishment of the continuous load shedding FCG measurement strategy standard ASTM E647 for crack threshold measurement [10,11], several further loading profiles have been proposed [125,126]. A range of the envisageable strategies is illustrated in Figure 14. ASTM E647 corresponds to case (b); others in [125,126] are (g, i, j) and one of the non-standard ones (d) has been applied to a γ-TiAl alloy in [127]. In selecting a strategy for FCG testing of γ-TiAl, some of the considerations are as follows. First, a constant $K_{\text{max}}$ (right-hand column) facilitates separation of the $p$ and $q$ exponents in Equation 2, where for intermetallics $p \sim q$. Depending on whether load is increased (c, g, k) or shed (a, e, i), the fatigue threshold is defined by the no-growth–growth transition, in either the forward or the reverse direction, respectively. This transition is defined in ASTM E647 to be at a growth rate of $10^{-10} \text{ m cycle}^{-1}$. Importantly, though, the size of the plastic zone ahead of the crack tip varies, being larger in decreasing $\Delta K$ loading than in increasing $\Delta K$ loading. In the loading strategies where a short loading period at a large stress intensity range, $\Delta K_0$, precedes FCG testing, the fatigue crack must now grow in a very large plastic zone.

There is no consensus as to which test strategies are most damaging or conservative, as this may be material dependent. It is reported in other engineering alloys [126] that prior cycling, even below the no-growth limit, can significantly reduce the fatigue life: load history can change the ranking of materials by FCG threshold. In a Ti alloy, for example, the strategy (i) produced a lower value for the fatigue threshold than (g) [125], although (g) was thought to be more representative of conditions during practical use. Lerch et al. [128] have investigated the behaviour of a γ-TiAl alloy Ti–46.5Al–2Cr–2Nb upon step increases in the applied cyclic stress during tests aimed at determining the maximum run-out stress. No evidence of coaxing was found, that is to say that there is apparently no load history in this alloy resulting from strengthening by strain ageing at lower stress levels. This does not, however, preclude overstress cycles (even a single cycle) from having a significant impact on the fatigue life.

For γ-TiAl alloys, although the shape of the crack tip plastic zone is known to be inconsistent with the predictions of continuum mechanics [106], the plastic zone size in application conditions is likely to be considerably different to that in a simple HCF test due to the additional underlying LCF or steady-state loading to the component. This LCF can lead to both strengthening by work hardening, or microcracking [66], in the crack process zone. The above FCG testing strategies (i–l) are able to reproduce the salient features of the superimposed LCF and HCF loadings, and may hence determine realistic or more conservative values for $\Delta K_{\text{th}}$. A more systematic approach, by comparison of the effect of the different loading strategies in Figure 14 on the measured fatigue thresholds, and an understanding of the underlying mechanistic reasons for this, may be necessary to gain confidence in the HCF lifing of γ-TiAl alloys.

Furthermore, there has been little work on the variable amplitude fatigue properties of γ-TiAl alloys [23]. This is possibly surprising as HCF in the proposed applications results from the relaxation of vibrations [5], where the amplitude varies. In general, however, variable amplitude fatigue is a loading regime for which the theory is also not well established [126].

**Advances in understanding the distribution of strain in fatigue**

Most of the studies on the fatigue of titanium aluminides where an effort has been made to seek a microstructural understanding of the deformation structures and mechanisms have used either scanning electron microscopy (SEM) to image flaw formation in situ [60,63] or transmission electron microscopy (TEM) to image dislocation and twin structures in samples extracted after testing. However, an image of the dislocation structure at one instance does not allow the plastic strain accumulated thus far in that region to be determined. Even techniques such as high-resolution
A selection of possible loading strategies for measuring the FCG rate \( da/dN \) as a function of \( \Delta K \), and hence determining the fatigue threshold where \( da/dN = 10^{-10} \) m cycle\(^{-1} \) (growth/no-growth transition). The FCG specimen is loaded at \( \Delta K_1 \) for an extended period (as per ASTM E647 [11]), followed by successive steps \( \Delta K_2, \Delta K_3 \) and so on, until an FCG rate of \( 10^{-10} \) m cycle\(^{-1} \) is either succeeded or exceeded, depending on whether \( \Delta K \) is being decreased (a, b) or increased (c, d), respectively. Constant \( K_{\text{max}} \) and constant \( R = K_{\text{min}}/K_{\text{max}} \) versions of each exist. Further, prior cycling at a large stress intensity range \( \Delta K_0 \) for a short period (e–h) may serve to generate a large plastic zone ahead of the crack tip within which the near-threshold crack must then grow. Finally, the large \( \Delta K_0 \) cycling may be applied between each step in \( \Delta K_1, \Delta K_2, \Delta K_3 \) (i–l) to remove history effects of the previous loading step by forcing the near-threshold crack to grow in the same sized large plastic zone at each step. Extended from [125] (reproduced with permission).
electron backscatter diffraction (HR-EBSD) [129] do not capture this as they measure only the elastic strain and rotation of the lattice, rather than the permanent, plastic strain associated with dislocation motion.

Work by Beran et al. [130] has enabled post-mortem TEM observations to be confirmed by in situ neutron diffraction during the cyclic loading of the testpiece. The volume fraction of deformation twins in the γ-TiAl phase was hence measured by both techniques; it rarely exceeded 0.025. Twinning as a plasticity mechanism in fatigue was found to be rapidly exhausted.

The past two decades has seen substantial progress in the field of digital image correlation (DIC) strain mapping [131], whereby the strain tensor for deformation in the surface plane is obtained by matching the positions of fiducial surface features before and after deformation. The strain developed between individual areas only 44 nm × 44 nm[132] in size can be measured by imaging in an SEM. Similarly to HR-EBSD, this technique is non-destructive and hence may be applied throughout the fatigue cycling of a testpiece. The first application of DIC strain mapping to TiAl alloys was made by Jiang et al. [133] on Ti–44Al–8Nb–1B loaded monotonically at room temperature to just below and above the yield stress to characterise microplasticity. Elsewhere, Niendorf et al. [134] cyclically loaded Ti–45Al–5Nb–0.2B–0.2C (TNB-V5) at 25°C and 700°C to successively higher loads until failure, with regular pauses at maximum load to image the surface for DIC strain mapping. Niendorf et al. demonstrated that the origin of final fracture within the gauge length upon LCF loading could be determined by DIC strain mapping after only 20% of the total fatigue life. Jiang et al. were able to show that in lamellar and nearly lamellar microstructures, plasticity first occurs at only ∼75% of the 0.2% proof stress, although in duplex microstructures this early onset was not seen; this was supported by post-mortem TEM imaging. However, both studies suffered from insufficient resolution in the strain maps to determine the exact location and distribution of plasticity relative to the lamellar microstructures. This was mainly due to their use of optical microscopy for image acquisition and insufficiently refined surface speckle patterns for higher resolution mapping.

More recent and extensive use of the DIC technique on TiAl alloys has been made by Filippini and co-workers [47,135–137]. They have measured the accumulation of plastic strain during monotonic and LCF loading of Ti–48Al–2Cr–2Nb testpieces with a duplex microstructure. Despite optical microscopy also being employed for pattern imaging, sufficient resolution was achieved to measure some local variations in the distribution of plastic strain between successive α2-Ti3Al and γ-TiAl lamellae, and the effect of lamellar orientation to the loading axis. Strain accumulation was identified in the vicinity of α2/γ lamellar interfaces in soft-mode oriented lamellar colonies (see Figure 15). This was consistent with DIC results obtained by Bode et al. [43] on an alloy of composition Ti–46.5Al–2.5V–1Cr. In both cases, lamellar orientations to the loading axis were obtained from the measurement of surface traces. However, soft-mode colonies might present a surface trace perpendicular to the loading axis and thereby appear to be hard-mode colonies, and vice versa. Attempts were made to assess the error in this by resolving strains along lamellae to calculate a minimum bound of confidence for the local plastic shear strain [135].

All DIC studies thus far on γ-TiAl alloys have used etching or sand blasting to produce the speckle pattern. To date, no systematic assessment exists of the effects of etching on the surface strains measured by DIC. However, these patterning methods may have preferentially weakened the lamellar interfaces and the colony boundaries and thereby caused strain to accumulate in these regions more so than on a flat, polished surface.

In short, digital image correlation strain mapping can measure the accumulation of local plasticity and damage during cyclic loading of γ-TiAl alloys. If applied using a high temperature stable nano-scale overlay speckle pattern [138,139], combined with SEM imaging [140,141], it has the opportunity to map plastic flow to the scale of individual γ lamellar domains, at aero-engine operational temperatures, without artificially weakening microstructural features. This would

![Figure 15](image-url). High-resolution digital image correlation strain map for a lamellar grain captured from a sample cyclically loaded (ex situ) in compression at a maximum nominal stress of 390 MPa and R = 0.05. Axial (vertical here) strain accumulates inside the defined lamellar platelets which are composed by the γ-TiAl phase. From [136] (reproduced with permission).
Figure 16. Atom probe tomography reconstruction of the $\beta_0/\gamma$ interface in Ti–43Al–4Nb–1Mo–0.1B–0.75C. Note the peak in C, arrowed in red, near the interface; variations in the C content at the interface may change the cohesion strength of the $\beta_0/\gamma$ interface. Adapted from [147] (reproduced with permission).

enable the location of slip/twin deformation during HCF, as well as crack initiation, to be studied as already achieved in other metallurgical systems [142].

A remark about alloy compositions

The present review has not focused on alloy compositions. One may note, as in [5], that conventional Ti alloy development has seen many incremental improvements in composition, and hence successive material replacements, without much change to processing. In contrast, for $\gamma$-TiAl alloys, most of the currently preferred and applied chemistries were developed over 20 years ago, e.g. Ti–48Al–2Cr–2Nb. The development of alloy compositions targeted at improving the strength and reliability of TiAl alloys under HCF loading at operational temperatures is being hindered by a lack of clarity as to optimal microstructures. Recognising that microstructural parameters and strengthening mechanisms beneficial to single colony (PST) components are not necessarily optimal for polycrystalline structures, and vice versa, might facilitate more directed compositional development.

An example of how the composition might be tailored for improved HCF properties is to consider secondary cracking by interlamellar debonding, which is a potential toughening mechanism that might be exploited in colonies where translamellar cracking prevails, e.g. in $\Phi = 0^\circ$ PST and columnar-grained components. It was mentioned in ‘Fatigue Cracking of Specific Colony Orientations in Polycrystals’ section that this mechanism does not operate in Ti–45Al–2Mn–2Nb–1B at stress intensity factors close to the threshold value, yet it is in this regime that TiAl alloys are to be operated. Furthermore, in untextured polycrystalline alloys, interlamellar debonding in $\Phi \sim 90^\circ$ colonies can lead to crack initiation; absorption of mechanical energy by interlamellar debonding is therefore undesirable in such microstructures. Measurements of the lamellar cohesion energy, its dependence on Al content and other alloying elements, and an understanding of the mechanisms therefore are scarce. Difficulty arises because accurate compositional measurements with lamellar scale resolution, e.g. by STEM-EDX [143], are difficult given the light elements present (Al, C, B) [144]. On the other hand, atom probe tomography (APT) studies of lamellar TiAl [145–148] have the potential to quantify the partitioning of elements to interfaces [149]. The accumulation of C in Ti–43Al–4Nb–1Mo–0.1B–0.75C (TNM) at a $\beta_0/\gamma$ interface was recently measured [147,150] (Figure 16). The lamellar cohesion strength also depends on the composition of the matrix, which defines the local stresses associated with semi-coherency at the interface. Direct microscale measurement of the interfacial strength is possible [151], however, current methods can generate spurious results, stemming from the sample preparation process [152]. Density functional theory modelling of interfacial energies for differing $\gamma/\gamma$ interface types exists [153], however, a more systematic analysis of the effect of a range of alloying elements on the interlamellar misfit stresses is lacking. In short, if $\gamma$-TiAl alloy compositions are to be tailored for HCF performance, an improved understanding of how solutes influence the interlamellar cohesion strength would be beneficial, and is achievable with current methods.

Summary

A summary of the conclusions drawn from the present review with regards to improving the lamellar $\gamma$-TiAl microstructure for HCF, depending on the morphology and orientation of the lamellar colonies, is given in Table 2.

| Table 2. Summary of the microstructural properties to optimise for improved HCF behaviour of fully or nearly lamellar $\gamma$-TiAl alloys as a function of lamellar orientation, $\Phi$. |
|----------------------------------|---------------|-----------|---------------|---|---|---|
| Microstructure                  | Polysynthetically twinned | Directionally solidified (columnar) | Polycrystalline atextured |
| $\Phi$                          | $0^\circ$ | $45^\circ$ | $90^\circ$ | $0^\circ$ | $45^\circ$ | $90^\circ$ | N/A |
| Optimal lamellar orientation    | ++       | N/A       | N/A         | +++   | ++      | N/A         | N/A |
| Optimal colony size             | N/A      | N/A       | N/A         | N/A   | N/A     | N/A         | N/A |
| Optimal lamellar thickness      | N/A      | N/A       | N/A         | N/A   | N/A     | N/A         | N/A |
| Plasticity (mainly longitudinal)| ++       | --        | +           | +     | --      | +           | --  |
| Interlamellar debonding         | +        | --        | +           | +     | +       | --           | --  |

+: Mechanism to be promoted; --: mechanism to be inhibited or avoided.
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Disclosure statement

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ORCID

T.E.J. Edwards  http://orcid.org/0000-0002-3089-0062

References

[1] Bewlay BP, Nag S, Suzuki A, et al. TiAl alloys in commercial aircraft engines. Mater High Temp. 2016;33(4–5):549–559.
[2] Goold I. Rolls-Royce advances toward ultrafan. Report, 2014.
[3] Tetsui T, Miura Y. Heat-resistant cast TiAl alloy for passenger vehicle turbochargers. Tech Rev. 2002;39(1):1–5.
[4] Zhu H, Wei T, Carr D, et al. Assessment of titanium aluminide alloys for high-temperature nuclear structural applications. J Miner Met Mater Soc. 2012;64(12):1418–1424.
[5] Rugg D, Dixon M, Burrows J. High-temperature application of titanium alloys in gas turbines. Material life cycle opportunities and threats – an industrial perspective. Mater High Temp. 2016;63(4–5):536–541.
[6] Smarsly W, Esslinger J, Clemens H. Status of titanium aluminide for aero engine applications. In: IWTA’16, 2016, Tokyo, Japan.
[7] Kim YW, Kim SL. Advances in gammaalloy materials–processes–application technology: Successes, dilemmas, and future. JOM. 2018;70(4):553–560. http://dx.doi.org/10.1007/s11837-018-2747-x
[8] Hénaff G, Gloanec A-L. Fatigue properties of TiAl alloys. Intermetallics. 2005;13(5):543–558.
[9] Appel F, Paul JDH, Oehring M. Gamma titanium aluminide alloys: science and technology. Weinheim: Wiley-VCH; 2011.
[10] Suresh S. Fatigue of materials. Cambridge: Cambridge University Press; 1998.
[11] Standard test method for measurement of fatigue crack growth rates. Standard, ASTM International, West Conshohocken, PA, 2015.
[12] Paris P, Erdogan F. A critical analysis of crack propagation laws. J Basic Eng. 1963;85(4):528–533.
[13] Kruzic JJ, Campbell JP, McKelvey AL. In: Kim Y-W, Dimiduk DM, Loretto MH, editors. Gamma titanium aluminides 1999. TMS; 1999, San Diego.
[14] Ritchie RO, Daussardt RH. Cyclic fatigue of ceramics. J Ceramic Soc Japan. 1991;99(1154):1047–1062.
[15] Hojo M, Tanaka K, Gustafsson CG, et al. Effect of stress ratio on near-threshold propagation of delamination fatigue cracks in unidirectional CFRP. Compos Sci Technol. 1987;29(4):273–292.
[16] Ritchie RO, Gilbert CJ, McNaney JM. Mechanics and mechanisms of fatigue damage and crack growth in advanced materials. Int J Solid Struct. 2000;37(1):311–329.
[17] Filippini M, Beretta S, Patriarca L, et al. Fatigue sensitivity to small defects of a gamma-titanium aluminide alloy. J ASTM Int. 2012;9(5):1–12.
[18] Dahar MS, Seifi SM, Bewlay BP, et al. Effects of test orientation on fracture and fatigue crack growth behavior of third generation as-cast Ti–48Al–2Nb–2Cr. Intermetals. 2015;57:73–82.
[19] Mabru C, Bertheau D, Pautrot S, et al. Influence of temperature and environment on fatigue crack propagation in a TiAl-based alloy. Eng Fract Mech. 1999;64(1):23–47.
[20] Feng RC, Rui ZY, Zuo Y, et al. Influence of temperature on fatigue crack propagation in TiAl alloys. Strength Mater. 2014;46(3):417–421.
[21] Bowen P, Chave RA, James AW. Cyclic crack growth in titanium aluminides. Mater Sci Eng A. 1995;192–193: 443–456. 3rd International Conference on High Temperature Intermetallics.
[22] Chen G, Peng Y, Zheng G, et al. Polysynthetic twinned TiAl single crystals for high-temperature applications. Nat Mater. 2016;15(8):876–881.
[23] Gloanec AL, Hénaff G, Bertheau D, et al. Fatigue crack growth behaviour of a gamma-titanium-aluminide alloy prepared by casting and powder metallurgy. Scr Mater. 2003;49(9):825–830.
[24] Zuo YR, Rui ZY, Feng RC, et al. Influence of microstructure and stress ratio on fatigue crack propagation in TiAl alloy. Adv Mat Res. 2014;881:1330–1333.
[25] Andersson LE, Larsson M. Device and arrangement for producing a three-dimensional object. EP Patent 1,296,788. 2005.
[26] Biamino S, Penna A, Ackelid U, et al. Electron beam melting of Ti–48Al–2Cr–2Nb alloy: microstructure and mechanical properties investigation. Intermetallics. 2011;19(6):776–781.
[27] Larsen JM, Rosenberger AH, Worth BD, et al. Assuring reliability of gamma titanium aluminides in long-term service. Minerals, metals and materials society/AIME, gamma titanium aluminides 1999 (USA). 1999. p. 463–472.
[28] Zhou Y, Wang JQ, Zhang B, et al. High-temperature fatigue property of Ti6Al4VNb alloy with the fully lamellar microstructure. Intermetallics. 2012;24:7–14.
[29] Xue HQ, Tao H, Montembault F, et al. Development of a three-point bending fatigue testing methodology at 20 kHz frequency. Int J Fatigue. 2007;29(9–11):2085–2093.
[30] Bathias C. Fatigue of materials and structures. Hoboken (NJ): John Wiley & Sons; 2013.
[31] Xue H, Tao H, Bayraktar E, et al. High-cycle fatigue of a TiAl alloy in three-point bending test. J Mech Strength. 2008;30(1):112–116.
[32] Bayraktar E, Bathias C, Hongquan X, et al. On the giga cycle fatigue behaviour of two-phase (α2 + γ) TiAl alloy. Int J Fatigue. 2004;26(12):1263–1275.
[33] Elber W. The significance of fatigue crack closure. West Conshohocken (PA): ASTM International; 1971.
[34] Halford GR. Fatigue and durability of structural materials. West Conshohocken (PA): ASM International; 2006.
[35] Schmidt RA, Paris PC. Threshold for fatigue crack propagation and the effects of load ratio and frequency. West Conshohocken (PA): ASTM International; 1973.
[36] Hénaff G, Cohen S-A, Mabru C, et al. The role of crack closure in fatigue crack propagation: behaviour of a TiAl-based alloy. Metall Mater Trans A. 1996;34(9):1449–1454.

[37] Worth BD, Larsen JM, Balsone SJ, et al. Mechanisms of ambient temperature fatigue crack growth in Ti–46.5Al–3Nb–2Cr–0.2W. Metall Mater Trans A. 1997;28(13):825–835.

[38] Soboyejo WO, Deffeyes JE, Aswath PB. Investigation of room- and elevated-temperature fatigue crack growth in Ti–48Al. Mater Sci Eng A. 1991;138(1):95–101.

[39] McKelvey AL, Rao KTV, Ritchie RO. On the anomalous temperature dependence of fatigue-crack growth in γ-based titanium aluminides. Scr Mater. 1997;37(11):1797–1803.

[40] Campbell JP, Kruzic JJ, Lillicbridge S, et al. On the growth of small fatigue cracks in γ-based titanium aluminides. Scr Mater. 1997;37(5):707–712.

[41] Kruzic JJ, Campbell JP, Ritchie RO. On the fatigue behavior of γ-based titanium aluminides: role of small cracks. Acta Mater. 1999;47(3):801–816.

[42] Chan KS, Shih DS. Fatigue and fracture behavior of a fine-grained lamellar TiAl alloy. Metall Trans A. 1997;28(1):79–90.

[43] Bode B, Wessel W, Brueckner-Foit A, et al. Local deformation at micro-notches and crack initiation in an intermetallic γ-TiAl-alloy. Fatigue Fract Eng Mater Struct. 2016;39(2):227–237.

[44] Wessel W. Mikrostrukturelle Untersuchungen der Rissinitierung und -ausbreitung in intermetallischen TiAl-Legierungen unter zyklischer und quasistatischer Belastung [PhD thesis]. 2012.

[45] Wessel W, Mildner J, Pitz P, et al. Micronotches for studying growth of small cracks. Fatigue Fract Eng Mater Struct. 2015;38(6):673–680.

[46] Wessel W, Zeismann F, Brueckner-Foit A. Short fatigue cracks in intermetallic γ-TiAl-alloys. Fatigue Fract Eng Mater Struct. 2015;38(12):1507–1518.

[47] Filippini M, Beretta S, Patriarca L. Effect of the microstructure on the deformation and fatigue damage in a gamma-TiAl produced by additive manufacturing. In: Won Kim Y, Smarsly W, Dennis Dimiduk JL, editors. Structural intermetallics. Warrendale (PA): TMS; 1997. p. 555.

[48] Hénaff G, Odemer G, Tonneau-Morel A. Environmentally-assisted fatigue crack growth mechanisms in advanced materials for aerospace applications. Int J Fatigue. 2007;29(9–11):1927–1940.

[49] Min Z, Xi-ping S, Long Y, et al. In situ observation of fatigue crack initiation and propagation behavior of a high-Nb TiAl alloy at 750°C. Mater Sci Eng A. 2015;622:30–36.

[50] Botten R, Wu X, Hu D, et al. The significance of acoustic emission during stressing of TiAl-based alloys. Part I: Detection of cracking during loading up in tension. Acta Mater. 2001;49(10):1687–1691.

[51] Wu X, Hu D, Botten R, et al. The significance of acoustic emission during stressing of TiAl-based alloys. Part II: Influence of cracks induced by pre-stressing on the fatigue life. Acta Mater. 2001;49(10):1693–1699.

[52] Lebensohn R, Uhlenhut H, Hartig C, et al. The role of mechanical twinning on microcrack nucleation and crack propagation in a near-γ TiAl alloy. Intermetallics. 2004;12(12):1317–1323.

[53] Jha SK, Larsen JM, Rosenberger AH. The role of competing mechanisms in the fatigue life variability of a nearly fully-lamellar γ-TiAl based alloy. Acta Mater. 2005;53(5):1293–1304.

[54] Han B, Wan WJ, Zhu CL, et al. Effect of casting defects on high cycle fatigue life of fully lamellar TiAl alloy. Mater Res Innov. 2015;19(Suppl. 4):S142–S146.

[55] Weibull W 1997 A statistical theory of strength of materials. Proceedings of the Royal Swedish Institute, Stockholm. Engineering Research No. 151; 1939.

[56] Umakoshi Y, Yasuda HY, Nakano T. Plastic anisotropy and fatigue of TiAl PST crystals: a review. Intermetallics. 1996;4(Suppl. 1):S65–S75.

[57] Lebensohn R, Uhlenhut H, Hartig C, et al. Plastic flow of γ-TiAl-based polysynthetically twinned crystals: micromechanical modeling and experimental validation. Acta Mater. 1998;46(13):4701–4709.

[58] Kad BK, Asaro RJ. Apparent Hall–Petch effects in polycrystalline lamellar TiAl. Philos Mag A. 1997;75(1):87–104.
[73] Simkin BA, Ng BC, Crimp MA, et al. Crack opening due to deformation twin shear at grain boundaries in near-γ TiAl. Intermetallics. 2007;15(1):55–60.

[74] Edwards TEJ, Di Gioacchino F, Muñoz-Moreno R, et al. The interaction of borides and longitudinal twinning in polycrystalline TiAl alloys. Acta Mater. 2017;140(Suppl. C):305–316.

[75] Lu YH, Zhang YG, Qiao LJ, et al. In-situ TEM study of fracture mechanisms of polycrystallinely twinned (PST) crystals of TiAl alloys. Mater Sci Eng A. 2000;289(1):91–98.

[76] Cottrell AH. Theory of brittle fracture in steel and similar metals. Trans Metall Soc AIME. 1958;212:192–203.

[77] Huang ZW, Bowden P. Persistent microslip bands in the lamellar TiAl structure subjected to room temperature fatigue. Scr Mater. 2001;45(8):931–937.

[78] Ding R, Li H, Hu D, et al. Features of fracture surface in a fully lamellar TiAl-based alloy. Intermetallics. 2015;58:36–42.

[79] Yang J, Li H, Hu D, et al. Microstructural characterisation of fatigue crack growth fracture surfaces of lamellar Ti45Al2Mn2Nb1B. Intermetallics. 2014;45:89–95.

[80] Filippini M, Beretta S, Patriarca L, et al. Defect tolerance of a gamma titanium aluminide alloy. Procedia Eng. 2011;10:5677–5682.

[81] Lettner T, Schlaffer M, Mayer S, et al. Fracture and R curve behavior of an intermetallic β-stabilized TiAl alloy with different nearly lamellar microstructures. Intermetallics. 2014;53:1–9.

[82] Yang Z-J, Sun H-L, Huang Z-W, et al. Fatigue properties of a medium-strength γ-TiAl alloy with different surface conditions. Rare Metals. 2016;35(1):93–99.

[83] Campbell JP, Ritchie RO, Venkateswara Rao KT. The effect of microstructure on fracture toughness and fatigue crack growth behavior in γ-titanium aluminide based intermetallics. Metall Mater Trans A. 1999;30(3):563–577.

[84] Ritchie RO, Suresh S. Some considerations on fatigue crack closure at near-threshold stress intensities due to fracture surface morphology. Metall Trans A. 1982;13(5):937–940.

[85] Filippini M, Schloffer M, Crist E 2017 Panel discussion: microstructure-defects-life. In: GAT’17 at TMS 2017. 2017, San Diego, CA.

[86] Clemens H, Mayer S. Design, processing, microstructure, properties, and applications of advanced intermetallic TiAl alloys. Adv Eng Mater. 2013;15(4):191–215.

[87] Dimiduk DM, Hazzledine PM, Parthasarathy TA, et al. The role of grain size and selected microstructural parameters in strengthening fully lamellar TiAl alloys. Metall Mater Trans A. 1998;29(1):37–47.

[88] Kim Y-W. Effects of microstructure on the deformation and fracture of γ-TiAl alloys. Mater Sci Eng A. 1995;192–193:519–533. Part 2.

[89] Chan KS, Kim Y-W. Influence of microstructure on crack-tip micromechanics and fracture behaviors of a two-phase TiAl alloy. Metall Trans A. 1992;23(6):1663–1677.

[90] Rogers NJ, Crofts PD, Jones IP, et al. Microstructure toughness relationships in fully lamellar γ-based titanium aluminides. Mater Sci Eng A. 1995;192–193:379–386.

[91] Cao G, Fu L, Lin J, et al. The relationships of microstructure and properties of a fully lamellar TiAl alloy. Intermetallics. 2000;8(5–6):647–653.

[92] Mine Y, Takashima K, Bowden P. Effect of lamellar spacing on fatigue crack growth behaviour of a TiAl-based aluminide with lamellar microstructure. Mater Sci Eng A. 2012;532:13–20.

[93] Lin D, Chen M. Interaction between deformation process and lamellar interfaces in polysynthetically twinned (PST) crystals of TiAl. In: Kim Y-W, Wagner R, Yamaguchi M, editors. Gamma titanium aluminides. Warrendale (PA): Minerals, Metals and Materials Society; 1995. p. 323–330.

[94] Umakoshi Y, Yasuda HY, Nakano T. Plastic anisotropy and fracture behavior of cyclically deformed TiAl polysynthetically twinned crystals. Mater Sci Eng A. 1995;192–193:511–517.

[95] Umakoshi Y, Nakano T. Plastic behaviour of TiAl crystals containing a single set of lamellae at high temperatures. ISIJ Int. 1992;32(12):1339–1347.

[96] Huang ZW, Hu W. Thermal stability of an intermediate strength fully lamellar Ti–45Al–2Mn–2Nb–0.8 vol.% TiB2 alloy. Intermetallics. 2014;54:49–55.

[97] Huang ZW, Hu W. On the role of thermal exposure on the stress controlled fatigue behaviour of an intermediate strength γ-TiAl based alloy. Mater Sci Eng A. 2015;636:77–90.

[98] Inui H, Oh MH, Nakamura A, et al. Room-temperature tensile deformation of polysynthetically twinned (PST) crystals of TiAl. Acta Metall Mater. 1992;40(11):3095–3104.

[99] Kim HY, Maruyama K. Parallel twinning during creep deformation in soft orientation PST crystal of TiAl alloy. Acta Mater. 2001;49(14):2635–2643.

[100] Porter WJ, Uchic MD, John R, et al. Compression property determination of a gamma titanium aluminide alloy using micro-specimens. Scr Mater. 2009;61(7):678–681.

[101] Rester M, Fischer FD, Kirchlechner C, et al. Deformation mechanisms in micron-sized PST TiAl compression samples: experiment and model. Acta Mater. 2011;59(9):3410–3421.

[102] Palomares-García AJ, Pérez-Prado MT, Molina-Aldareguia JM. Effect of lamellar orientation on the strength and operating deformation mechanisms of fully lamellar TiAl alloys determined by micropillar compression. Acta Mater. 2017;123:102–114.

[103] Edwards TEJ, Di Gioacchino F, Muñoz-Moreno R, et al. Deformation of lamellar TiAl alloys by longitudinal twinning. Scr Mater. 2016;118:46–50.

[104] Edwards TEJ, Di Gioacchino F, Mohanty G, et al. Longitudinal twinning in a TiAl alloy at high temperature by in situ microcompression. Acta Mater. 2017;148:202–215.

[105] Liu S, Wheeler JM, Howie PR, et al. Measuring the fracture resistance of hard coatings. Appl Phys Lett. 2013;102(17):171907.

[106] Appel F, Christoph U, Wagner R. An electron microscope study of deformation and crack propagation in (α2 + γ) titanium aluminides. Philos Mag A. 1995;72(2):341–360.

[107] Yamaguchi M, Johnson DR, Lee HI, et al. Directional solidification of TiAl-based alloys. Intermetallics. 2000;8(5):511–517.

[108] Li X, Fan J, Su Y, et al. Lamellar orientation and growth direction of α phase in directionally solidified Ti–46Al–0.5W–0.5Si alloy. Intermetallics. 2012;27:38–45.
[109] Ding XF, Lin JP, Zhang LQ, et al. Microstructural control of TiAl–Nb alloys by directional solidification. Acta Mater. 2012;60(2):498–506.

[110] Zhang L, Lin J, He J, et al. Influence of thermal stabilization treatment on the subsequent microstructure development during directional solidification of a Ti–46Al–5Nb alloy. Intermetallics. 2015;63:67–72.

[111] Liu T, Luo L, Su Y, et al. Lamellar orientation control of Ti–47Al–0.5W–0.5Si by directional solidification using β seeding technique. Intermetallics. 2016;73:1–4.

[112] Wan W, Han B, Han W, et al. High-cycle fatigue behavior of TiAl alloy containing preferentially oriented lamellar microstructures. J Aeronaut Mater. 2016;36(1):87–92.

[113] Shiota H, Tokaji K, Ohta Y. Influence of lamellar orientation on fatigue crack propagation behavior in titanium aluminide TiAl. Mater Sci Eng A. 1998;243(1):169–175.

[114] Schütze M 2004 Recent advances in the understanding of the halogen effect in the oxidation of intermetallic titanium aluminides. Proceedings, Electrochemical Society. vol. PV 2004–16. 2004. p. 1–15, Honolulu, HI.

[115] Schweizer FA, Duhl DN. Single crystal nickel superalloy. Patent. 1980.

[116] Clemens H, Kestler H, Eberhardt N, et al. In: Kim Y-W, Dimiduk DM, Loretto MF, editors. Gamma titanium aluminides. San Diego: TMS; 1999. p. 209.

[117] Hu D. Role of boron in TiAl alloy development: a review. Rare Metals. 2015;35(1):1–14.

[118] Yang J, Li H, Hu D, et al. Lamellar orientation effect on fatigue crack propagation threshold in coarse grained Ti46Al8Nb. Mater Sci Technol. 2016;95:647–662.

[119] Fuhrer T, Nakamura A, Hosomi M, et al. Deformation of polycrystallinely twinned crystals of TiAl with a nearly stoichiometric composition. Philos Mag A. 1990;61(4):591–606.

[120] Nomura M, Kim M-C, Vitek V, et al. In: Kim Y-W, Dimiduk DM, Lorenz MH, editors. Gamma titanium aluminides. San Diego: TMS; 1999. p. 67.

[121] Beretta S, Filippini M, Pasquerio PLG, et al. Fatigue properties and design criteria of a gamma titanium aluminide alloy. Key Eng Mater. 2011;465:531–534.

[122] Hu D, Huang A, Jiang H, et al. Pre-yielding and pre-yield cracking in TiAl-based alloys. Intermetallics. 2006;14(1):82–90.

[123] Muñoz- Moreno R, Pérez- Prado MT, Llorca J, et al. Effect of stress level on the high temperature deformation and fracture mechanisms of Ti–45Al–2Nb–2Mn–0.8 vol. pct TiB2: an in situ experimental study. Metall Mater Trans A. 2013;44(4):1887–1896.

[124] Edwards TEJ, Di Gioacchino F, Clegg WJ. An experimental study of polycrystalline plasticity in lamellar titanium aluminide. unpublished data.

[125] Lenets YN, Nicholas T. Load history dependence of fatigue crack growth thresholds for a Ti–alloy. Eng Fract Mech. 1998;60(2):187–203.

[126] Nicholas T. High cycle fatigue: a mechanics of materials perspective. New York: Elsevier; 2006.

[127] Pippan R, Hageneder P, Knabl W, et al. Fatigue threshold and crack propagation in γ-TiAl sheets. Intermetallics. 2001;9(1):89–96.

[128] Lerch BA, Draper SL, Pereira JM. Conducting high-cycle fatigue strength step tests on gamma TiAl. Metall Mater Trans. 2002;33(12):A3871–A3874.

[129] Wilkinson AJ, Meaden G, Dingley DJ. High resolution mapping of strains and rotations using electron backscatter diffraction. Mater Sci Technol. 2006; 22(11):1271–1278.

[130] Beran P, Heiczko M, Kruml T, et al. Complex investigation of deformation twinning in γ-TiAl by TEM and neutron diffraction. J Mech Phys Solids. 2016;95:647–662.

[131] Sutton MA, Orteu JJ, Schreier H. Image correlation for shape, motion and deformation measurements: basic concepts, theory and applications. New York: Springer Science & Business Media; 2009.

[132] Orozco-Caballero A, Lunt D, Robson JD, et al. How magnesium accommodates local deformation incompatibility: a high-resolution digital image correlation study. Acta Mater. 2017;133:367–379.

[133] Jiang H, Garcia-Pastor FA, Hu D, et al. Characterization of microplasticity in TiAl-based alloys. Acta Mater. 2009;57(5):1357–1366.

[134] Niendorf T, Burs C, Canadinc D, et al. Early detection of crack initiation sites in TiAl alloys during low-cycle fatigue at high temperatures utilizing digital image correlation. Int J Mater Res. 2009;100(4):603–608.

[135] Patriarca L, Filippini M, Beretta S. Digital image correlation-based analysis of strain accumulation on a duplex γ-TiAl. Intermetallics. 2016;75:42–50.

[136] Patriarca L, İçöz C, Filippini M 2014 Microscopic analysis of fatigue damage accumulation in TiAl intermetallics. ASME Turbo Expo 2014: turbine technical conference and exposition. Vol. 7A. 2014, Düsseldorf.

[137] İçöz C, Patriarca L, Filippini M, et al. Strain accumulation in TiAl intermetallics via high-resolution digital image correlation (DIC). Procedia Eng. 2014;74:443–448.

[138] Di Gioacchino F, Quinta da Fonseca J. Plastic strain mapping with sub-micron resolution using digital image correlation. Exp Mech. 2012;53(5):743–754.

[139] Edwards TEJ, Di Gioacchino F, Springbett HP. Stable speckle patterns for nano-scale strain mapping up to 700°C. Exp Mech. 2017;57(9):1469–1482.

[140] Sutton MA, Li N, Joy DC, et al. Scanning electron microscopy for quantitative small and large deformation measurements. Part I: SEM imaging at magnifications from 200 to 10,000. Exp Mech. 2007;47(6):775–787.

[141] Sutton MA, Li N, Garcia D, et al. Scanning electron microscopy for quantitative small and large deformation measurements. Part II: experimental validation for magnifications from 200 to 10,000. Exp Mech. 2007;47(6):789–804.

[142] Stinville JC, Echlin MP, Callahan PG, et al. Measurement of strain localization resulting from monotonic and cyclic loading at 650°C in nickel based superalloys. Exp Mech. 2017;57(8):1289–1309.

[143] Pfullmann T, Beaven PA. On the relationship between lattice parameters and composition of the γ-TiAl phase. Scr Metall Mater. 1993;28(3):275-280.

[144] Zhang XF, Yang Q, De Jonghe LC, et al. Energy dispersive spectroscopy analysis of aluminium segregation in silicon carbide grain boundaries. J Microsc. 2002;207(1):58–68.

[145] Huguet A, Menand A. Atom-probe determination of interstitial element concentration in two-phase
and single-phase TiAl-based alloys. Appl Surf Sci. 1994;76:191–197.

[146] Menand A, Zapolsky-Tatarenko H, Nérac-Partaix A. Atom-probe investigations of TiAl alloys. Mater Sci Eng A. 1998;250(1):55–64.

[147] Klein T, Mendez-Martin F, Schachermayer M. 2015 Distribution of alloying elements within the constituent phases of a C-containing γ-TiAl based alloy studied by atom probe tomography. MRS proceedings. vol. 1760. 2015, San Francisco, CA.

[148] Lefebvre W, Loiseau A, Menand A. Field evaporation behaviour in the γ phase in Ti–Al during analysis in the tomographic atom probe. Ultramicroscopy. 2002;92(2):77–87.

[149] Klein T, Clemens H, Mayer S. Advancement of compositional and microstructural design of intermetallic γ-TiAl based alloys determined by atom probe tomography. Materials. 2016;9(9):755.

[150] Klein T, Usategui L, Rashkova B, et al. Mechanical behavior and related microstructural aspects of a nano-lamellar TiAl alloy at elevated temperatures. Acta Mater. 2017;128:440–450.

[151] Sernicola G, Giovannini T, Patel P, et al. In situ stable crack growth at the micron scale. Nat Commun. 2017;8(1):108.

[152] Xiao Y, Wehrs J, Ma H, et al. Investigation of the deformation behavior of aluminum micropillars produced by focused ion beam machining using Ga and Xe ions. Scr Mater. 2017;127:191–194.

[153] Yoo MH, Fu CL. Physical constants, deformation twinning, and microcracking of titanium aluminides. Metall Mater Trans A. 1998;29(1):49–63.