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An analysis of modern creep lifing methodologies in the titanium alloy Ti6-4

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Abstract

Traditional creep lifing techniques based on power law equations have shown themselves to be extremely limited, particularly in the prediction of long term data based only on short term experiments. More recently, alternative approaches such as the Wilshire equations and hyperbolic tangent methods have been proposed which offer a new insight into the field, with the Wilshire equations in particular showing promise in predicting long term behaviour based on short term results. The aerospace industry however has different requirements to the power generation industry where the Wilshire equations have been thoroughly tested. The current work seeks to investigate the ability of these methods to extrapolate across temperatures and in the case of the Wilshire equations offer a capability to accurately describe the time to specific strains during creep deformation.

1. Introduction

The titanium alloy Ti6-4 (Ti-6Al-4V) has widespread applications due to its high strength to weight ratio and good fracture toughness properties. It also provides designers with flexibility due to the wide range of microstructures available which can be utilised to adapt mechanical properties. However, it has been in the aerospace industry where the most extensive use has been observed.

Widely used for low temperature applications at the front of the gas turbine engine, Ti6-4 displays evidence of conventional creep strain accumulation at temperatures in excess of 300°C. This should not be confused with strain accumulation at low temperatures in Ti6-4, loosely termed ‘cold creep’, which is a result of disparity between the critical resolved shear stress values (CRSS) for basal and prismatic slip in titanium [1-5].

At higher temperatures, however, the differences in the CRSS values no longer apply [6] and there is less preference for prismatic slip, making deformation more isotropic across grains. At these temperatures the alloy can now be considered to be exhibiting more conventional creep deformation.

The current work considers creep testing performed on constant stress tensile creep machines at temperatures ranging from 648-773K. At these temperatures, Ti6-4 displays what would be considered 'normal' creep curves, where a decaying primary stage is offset by an accelerating tertiary stage, leading to the observation of a minimum creep rate rather than the traditionally held view of a steady state region.

Since the early 1950s, however, research has focussed on relationships involving the minimum strain rate ($\dot{\varepsilon}_m$) and rupture time (t_f). The most commonly used relationship was proposed by Monkman and Grant [7] and suggests that the product of minimum strain rate and rupture time is constant (denoted M). As such, it has become common practice to analyse experimental trends using power law equations of the form

$$M/t_f = \dot{\varepsilon}_m = A\sigma^n \exp(-Q_C / RT) \qquad (1)$$

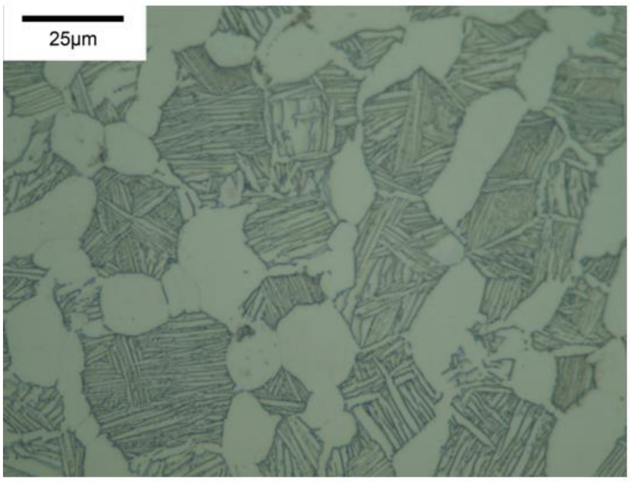
where $R=8.314 Jmol^{-1}K^{-1}$. With eq. (1), the parameter (A), the stress exponent (n), and the apparent activation energy for creep (Q_C) are themselves functions of stress and temperature. Traditionally, this issue of 'variable constants' has been addressed by invoking changes in creep processes under different stress and temperature regimes. Using such an approach, the dominant mechanism is usually identified by comparing measured and theoretical values of n and Q_c. Diffusion-controlled dislocation processes are considered to be dominant for $n > 4$ at high stresses, whereas at low stresses towards $n \cong 1$, a transition to diffusional creep mechanisms not involving dislocation movement occurs.

More recently, Wilshire[8-11] has proposed that available experimental evidence is difficult to reconcile with this approach and that more radical thinking is required to avoid the confusion of variable constants which makes prediction of long term behaviour based on short term data so difficult. As such the current research seeks to apply new, alternative methodologies to lifing the creep behaviour of Ti6-4.

Previous research [8-11] has demonstrated the capability of the Wilshire equations for extrapolation of short term data to predict long life behaviour in a range of alloys. This is clearly a significant development for the power generation sector where the reduction in lead time for new alloy implementation is a critical issue. However, the requirements of the aerospace industry, where Ti6-4 is widely used differ significantly from the power generation sector. Long term creep failure is no longer an issue due to the non-static nature of gas turbine engine components, which promote fatigue failures at comparatively short times. More important is the capability of any predictive method to accurately describe times to critical strains (for example to avoid blade extension and rubbing) and also to extrapolate to temperatures where previous data may not be available. The current paper seeks to investigate these two issues in the context of two of the more recently proposed creep lifing theories.

2. Material and experimental method

The Ti6-4 utilised in this programme was taken from a disc forged in the α+β temperature range. The microstructure, Figure 1, is shown to be typical of an α+β forged material and shows a bimodal microstructure with approximately 40% primary alpha phase, with an average alpha grain size of the order of 25μm. The remaining material appears as transformed product in a lath type structure. It is important to consider the crystallographic texture of the material, since strong textures can significantly affect deformation in these potentially highly anisotropic materials. Samples were removed from the tangential orientation of the disc and Figure 1 also shows a pole figure generated from material at the end of a sample testpiece. It can be seen that the texture is low, typical of a material heat treated above the beta transus, and that no significant orientation effects would be expected. It should also be noted that, considering recent research into titanium alloys [12], no evidence was found of significant macrozones of common orientation in the material.



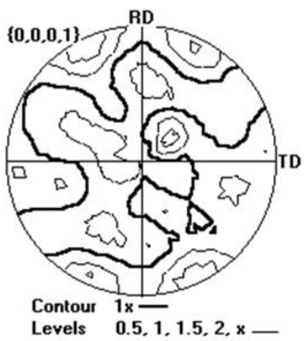


Figure 1: Microstructure and texture of tested Ti6-4

Creep tests were performed under tensile loading using high precision constant-stress machines [13]. Cylindrical specimens were utilised with a gauge length of 25.4mm and a diameter of 3.5mm. Tests were performed at 648, 673, 698, 723, 748K and subsequently at 773K. Some of these tests were allowed to run to failure, whereas others were removed from the test frame after specific strains were achieved as part of a larger programme of work. However, these tests still proved to be useful since detailed strain recording allowed for minimum creep rates to be calculated for these tests.

3. Results

Figures 2 and 3 show the experimental data generated during this programme, with lines added to show the trends of the data. Closer inspection of these graphs indicates that the stress exponent (n) from eqn (1) shows a great deal of variation, with values ranging from approximately 5 at low stress to 64 at the higher stress in the tests at 673K. Calculations also show that the value of the activation energy (Q_c) ranges from 376kJmol^{-1} when calculated for a constant stress of 100MPa to 193kJmol^{-1} when calculated at 600MPa. These values are consistent with those found in the published literature [14,15] but clearly this variability in both n and Q_c is unsatisfactory and makes extrapolation of short term data into long term properties extremely difficult, along with precluding a interpolative method for other temperatures. This is clearly highlighted in the work of Barboza et al [16] who estimate a value of Q_c of 415kJmol^{-1} at 291MPa but quote the traditionally held view of variations in n and Q_c being due to mechanism changes in the material under different stress/temperature regimes.

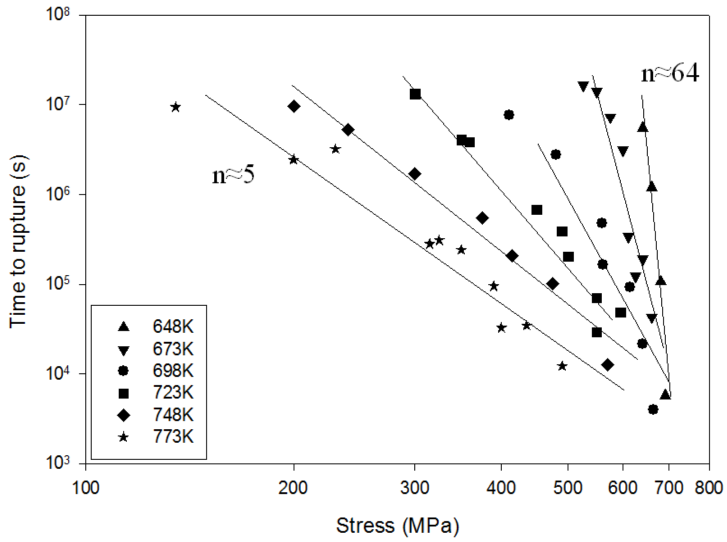


Figure 2: Rupture time as a function of stress in Ti6-4. Clearly evident is the wide variation in n value from low to high temperatures.

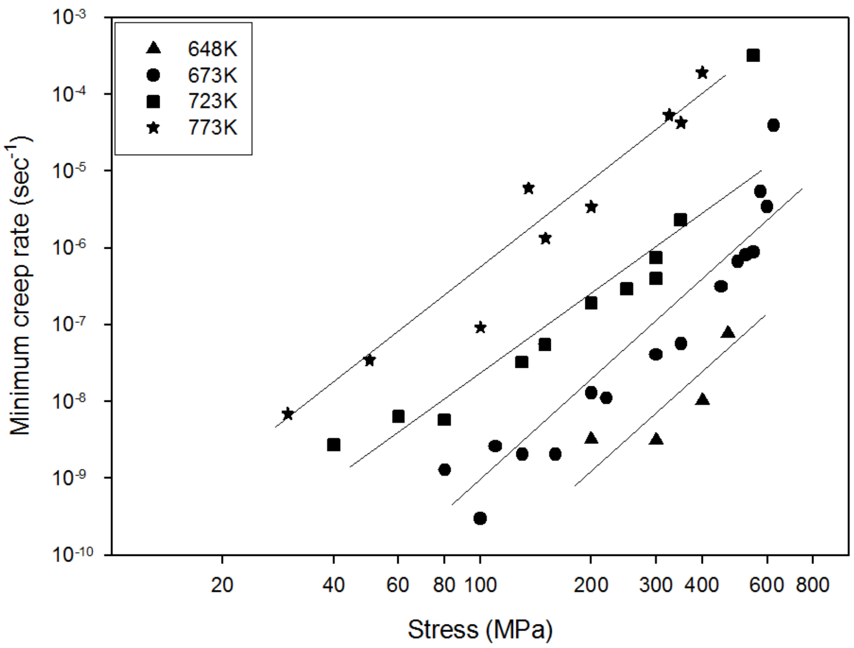


Figure 3: Minimum creep as a function of stress in Ti6-4. Creep curves in this material show that a minimum rate is reached rather than a 'steady state' phase.

At the core of the understanding provided by the Wilshire equations is an alternative method of calculating the apparent activation energy for the material (Q_c^*). The value of Q_c* can be calculated from the superimposition of different temperature data sets from log $\dot{\varepsilon}_m$/log σ and log t_f/log σ plots where the stress is normalised by the ultimate tensile stress at each tested temperature. Previous work for a range of materials [8-11] has shown the method to consistently provide more realistic and constant activation energy values, which compare favourably with the activation energy for lattice diffusion in the respective materials. For the current material, an activation energy of 250kJmol^{-1} is found to superimpose the data. Again, this is likely to be far closer to the activation energy for lattice diffusion in both the α and β phases of titanium than the Q_c value calculated traditionally.

Figures 4 and 5 show the dependency of both the temperature compensated creep rate and temperature compensated creep life respectively on the normalised stress (σ/σ_{TS}). The same calculation can be performed using the yield stress σ_Y rather than σ_{TS}. However, σ_{TS} is usually preferred since is becomes clear that $\dot{\varepsilon}_m \rightarrow \infty$ and $t_f \rightarrow 0$ as $(\sigma/\sigma_{TS}) \rightarrow 1$. By utilising σ_{TS} rather than $\sigma_{Y,}$ eqn (1) now becomes

$$M/t_f = \dot{\varepsilon}_m = A^*(\sigma/\sigma_{TS})^n \exp(-Q_C^*/RT) \quad (2)$$

where A ≠ A and $Q_c^* \neq Q_c$. The Q_c^* values in eqn (2) are then determined by the temperature dependencies of $\dot{\varepsilon}_m$ and t_f at constant (σ/σ_{TS}), whereas Q_c is calculated at constant σ with eqn (1). It is clearly evident from figures 4 and 5 that a value of $250kJmol^{-1}$ provides a good correlation of the different temperature data.

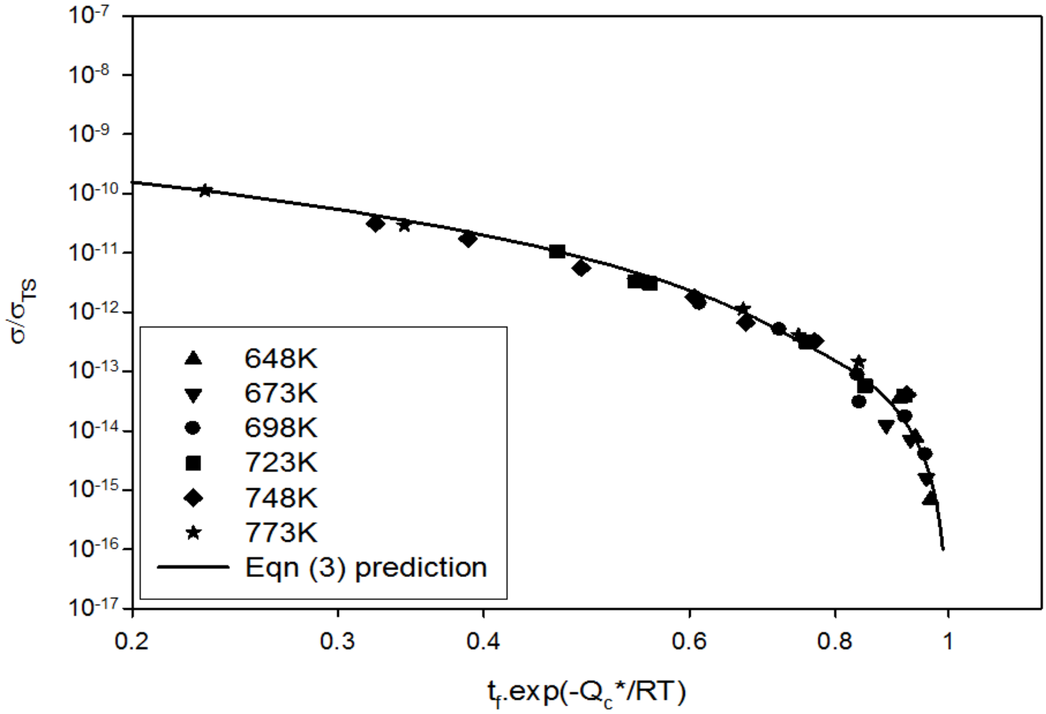


Figure 4: The dependency of the temperature compensated creep life on (σ/σ_{TS}) for Ti6-4 with $Q_c^*=250kJmol^{-1}$. The solid line indicates predictions made by eqn. (3).

Eqn (2) is beneficial since it avoids the large values of Q_c which have been previously reported and seem unrealistic in terms of creep processes. It also avoids the large variations in Q_c which plague this type of analysis. However, it does not avoid the variations seen in the creep exponent, n, which decreases significantly as the stress is lowered. As such, extrapolation of the data is still unreliable by power law type approaches [17].

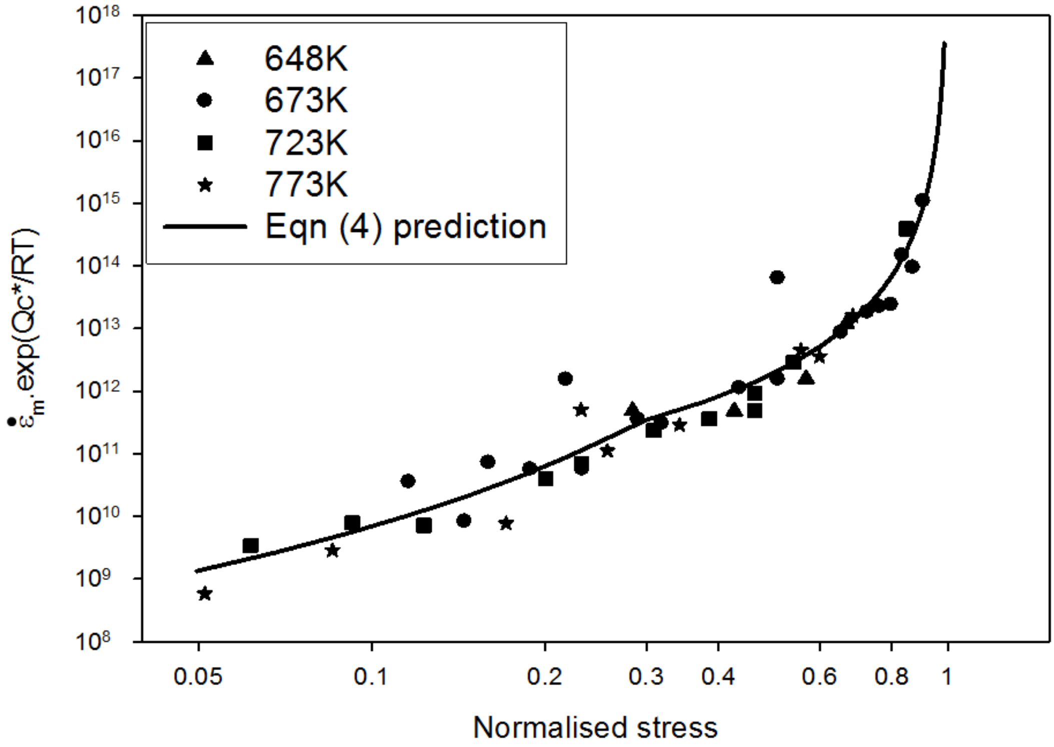


Figure 5: The dependency of the temperature compensated minimum creep rate on (σ/σ_{TS}) for Ti6-4 with Q_c^*=250kJmol^{-1}. The solid line indicates predictions made by eqn (4).

3.1 The Wilshire equations

A significant difference between traditional power law type approaches and the more recently developed 'Wilshire equations' is that the latter begins with the assumption that $t_f \rightarrow 0$ as $\sigma \rightarrow \sigma_{TS}$, while $t_f \rightarrow \infty$ as $\sigma \rightarrow 0$. Naturally, this basis in reality immediately provides confidence. The applied stress is shown to be related to the rupture time, t_f, by the equation

$$(\sigma/\sigma_{TS}) = \exp\{-k_1 [t_f. \exp(-Q_c^*/RT)]^u\} \qquad (3)$$

where the parameters Q_c^*, k_1 and u are derivable from a reasonably comprehensive set of creep rupture data, as is available in this programme of work. A similar equation then describes the stress and temperature dependence of the minimum creep rate ($\dot{\varepsilon}_m$) as

$$(\sigma/\sigma_{TS}) = \exp\{-k_2 [\dot{\varepsilon}_m . \exp(Q_c^*/RT)]^v\} \quad (4)$$

where the coefficients k_2 and v, are also easily computed from suitable data sets.
In order to derive the constants for eqn (3), it is clearly necessary to employ the value of Q_c^* previously derived for eqn (2). By then plotting ln [t_f . exp ($-Q_c^*$/RT)] against ln [-ln (σ/σ_{TS})] the constants u and k_1 are easily determined from the gradient and intercept, figure 6. It can be seen a 'break' occurs in the data, leading to two distinct straight lines. By least squares analysis, it can be shown that the derived constants are u=0.4 and k_1=29700 for high stresses and u=0.273 and k_1=757 for low stresses.

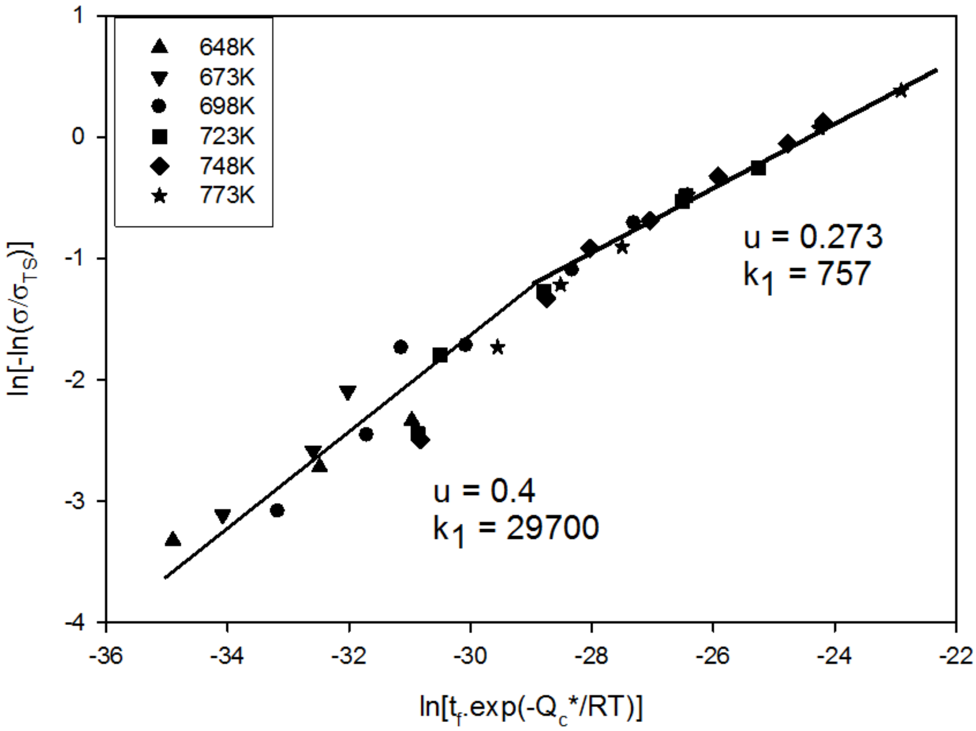


Figure 6: Plot of ln[t_f.exp(-Q_c^*/RT)] vs ln[-ln(σ/σ_{TS})] for Q_c^*=250kJmol^{-1} allowing for calculation of u and k_1 values in eqn (3) for high and low stress regimes.

As previously described, advances in lifing approaches to consider creep-fatigue interactions have allowed the extension of traditional alloys towards temperature regimes for which they would not have been originally considered, hence improving engine efficiency. In order to demonstrate the capabilities of the Wilshire equations, all calculations have been based only on data produced at temperatures up to 475°C. The derived Wilshire equations have then been used to extrapolate predictions to the higher temperature of 500°C to provide an example of the potential effectiveness of the equations when compared with the 500°C data subsequently produced.

A similar analysis is performed in order to derive the constants for eqn (4), figure 7. It is encouraging that a similar break occurs in the data, and that it also appears at a similar value of stress. In this case the derived constants are v=-0.31 and k_2=1097 for high stresses and v=-0.234 and k_2=166 for low stresses. The predictions made by the Wilshire equations method is shown in figure 8, and it can be seen that excellent interpolation of the lower temperature results occurs, along with good extrapolation to the higher temperature results at 500°C.

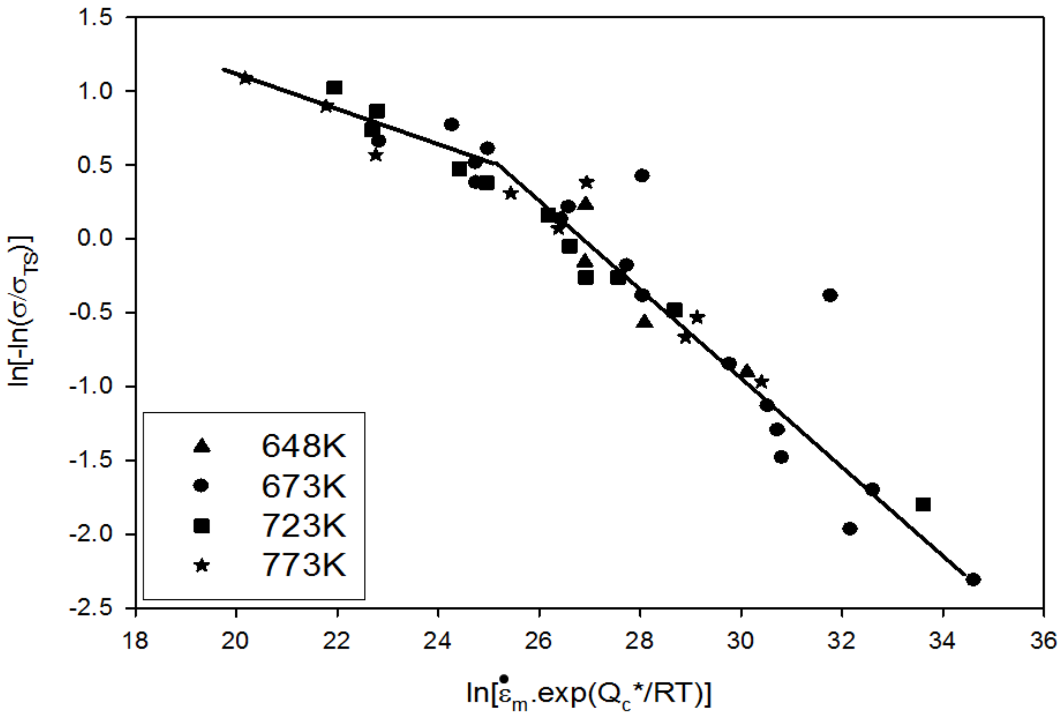


Figure 7: Plot of ln[ε_m.exp(-Q_c^*/RT)] vs ln[-ln(σ/σ_{TS})] for Q_c^*=250kJmol^{-1} allowing for calculation of v and k_2 values for high and low stress regimes.

Furthermore the times to pre-defined strains can also be calculated from the equation

$$(\sigma/\sigma_{TS}) = \exp \{-k_3 [t_\varepsilon . \exp (-Q_c^*,/RT)]^w\} \quad (5)$$

In the current work, this equation becomes extremely significant because of the requirements of the aerospace industry. Due to its good formability, reasonable temperature capability and availability Ti6-4 is utilised extensively in the front part of the gas turbine engine. A typical example will be the use in the early stages of the IP compressor as a blade material, before increasing temperatures lead to the requirement for more appropriate near α titanium alloys. As such, creep deformation at intermediate temperatures will be a significant concern due to the requirement for tip clearance of the blades from the surrounding casing.

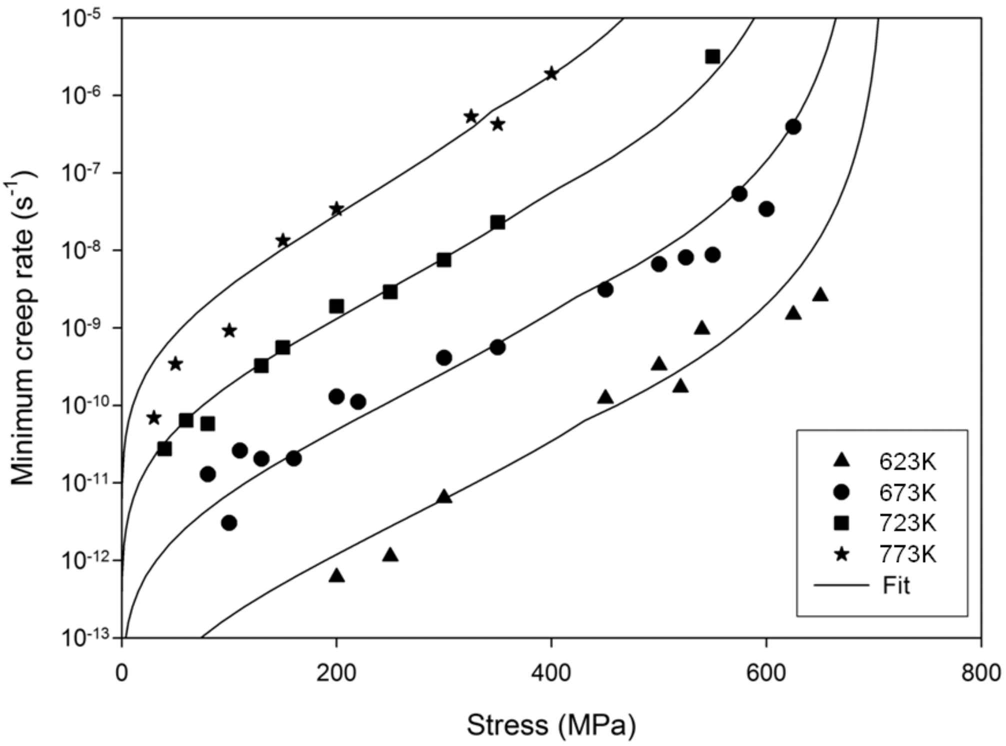


Figure 8: Predictions of the minimum creep rate as a function of stress. Predictions made using the Wilshire equations, eqn (4).

Clearly these requirements will differ for different engines and manufacturers, but as a demonstration of the technique a critical strain of 1% has been chosen in the current work. For this strain the values of the parameters have been optimised at k_3= 660 and w=0.23. The fits provided by the Wilshire equation, Figure 9, again are based only on temperatures up to and including 475°C, with predictions made for the 500°C data.

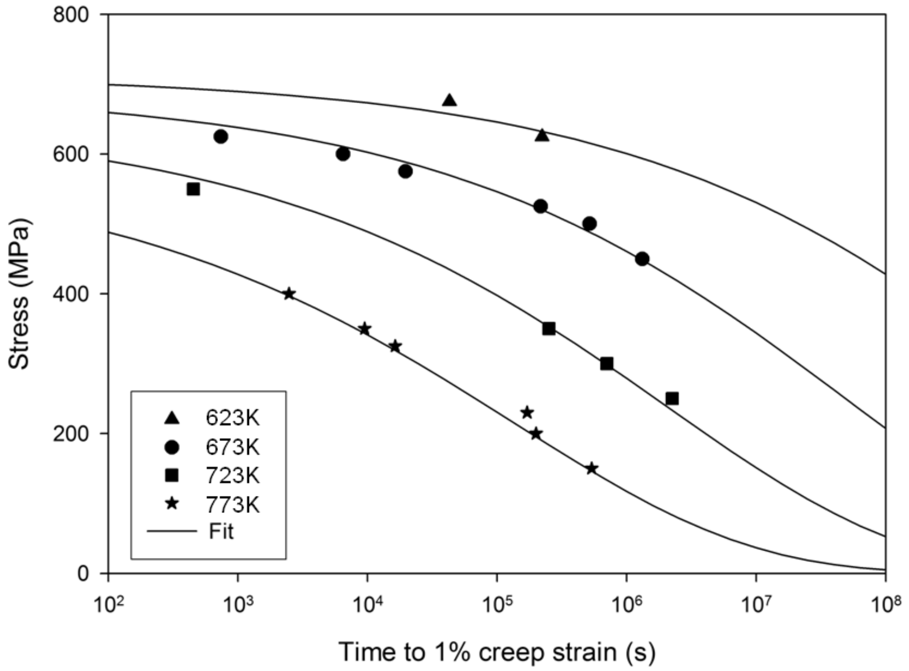


Figure 9. Predictions of the time to 1% strain as a function of stress. Predictions made using the Wilshire equations, eqn (5).

3.2 The Hyperbolic Tangent method

Developed by Rolls-Royce plc in the mid-1990s in support of high performance component lifing, this method also recognises that σ_{TS} is the highest stress which can be applied at the creep temperature, i.e. $t_f \rightarrow 0$ as $\sigma \rightarrow \sigma_{TS}$, while $t_f \rightarrow \infty$ as $\sigma \rightarrow 0$ [18].

The Hyperbolic Tangent method offers a similar approach to the Wilshire equations in that it concentrates on rupture data rather than attempting to fit equations to individual creep curves and then express the derived constants as functions of stress and temperature as in methods such as the theta projection method [13,19]. It recognises instead that the rupture data at any temperature can be represented by a hyperbolic tangent function

$$\sigma = \sigma_{TS}/2 \{1 - \tanh [k \ln (t/t_i)]\} \quad (6)$$

The time co-ordinate and slope at the inflection point of each hyperbolic tangent (t_i and k, as shown in Figure 10) are determined using experimental data at each temperature and are then expressed as functions of temperature as

$$t_i = \exp[t_i1 + t_i2\,(T_M-T)^2 + t_i3\,(T_M-T)^4 + t_i4\,(T_M-T)^6] \qquad (7)$$

$$k = \exp\,[k1+k2(T-T_N)^2+k3(T-T_N)^3+k4(T-T_N)^4+k5(T-T_N)^5] \qquad (8)$$

The emphasis with the Hyperbolic Tangent method is on sensible extrapolation outside the range of experimental measurements, which is essential for analysing components that experience a wide range of temperatures and stresses during cyclic loading. Upper and lower temperature limits (T_M and T_N) are defined, between which the expected trends in material behaviour are reproduced.

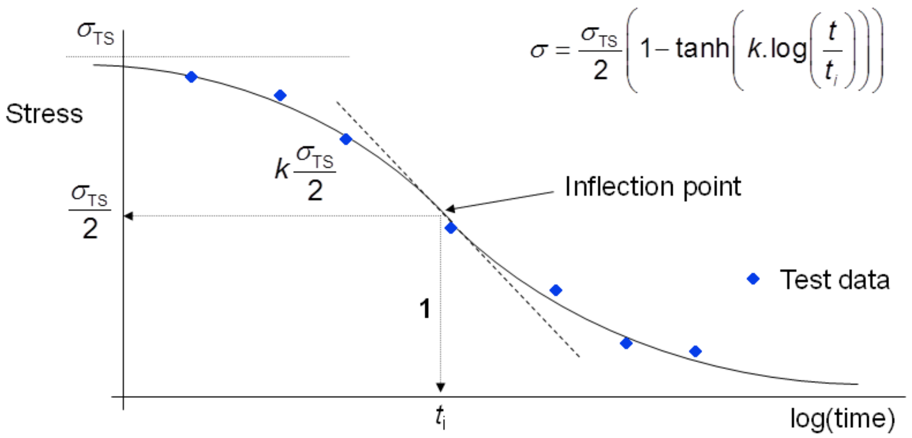


Figure 10. Hyperbolic tangent equation used to describe stress rupture behaviour.

T_M is set either to the material's melting temperature or 250°C above the highest experimental temperature, whichever is the lower. T_N is assumed to be 0.4 T_M (in Kelvin), approximating the lowest temperature at which creep will occur when T_M is the melting temperature. Clearly the definition of this temperature in a material such as Ti6-4, which shows strain accumulation at low temperatures, is of vital importance, since there will be a distinct mechanism change between creep at low and high temperatures. In the current work, this method is extremely applicable, operating as it does over relatively limited temperature and rupture time regimes. It is possible to extend the method to predict minimum creep rates or times to particular strains, although this element of the method is also widely used to predict minimum creep rates and times to particular strains although this is not considered further in this paper.

Figure 11 shows the predictions made by both the Wilshire equation and hyperbolic tangent methods with both sets of predictions again based only on data up to 475°C. It can be seen that both methods provide extremely accurate methods of predicting rupture times in Ti6-4.

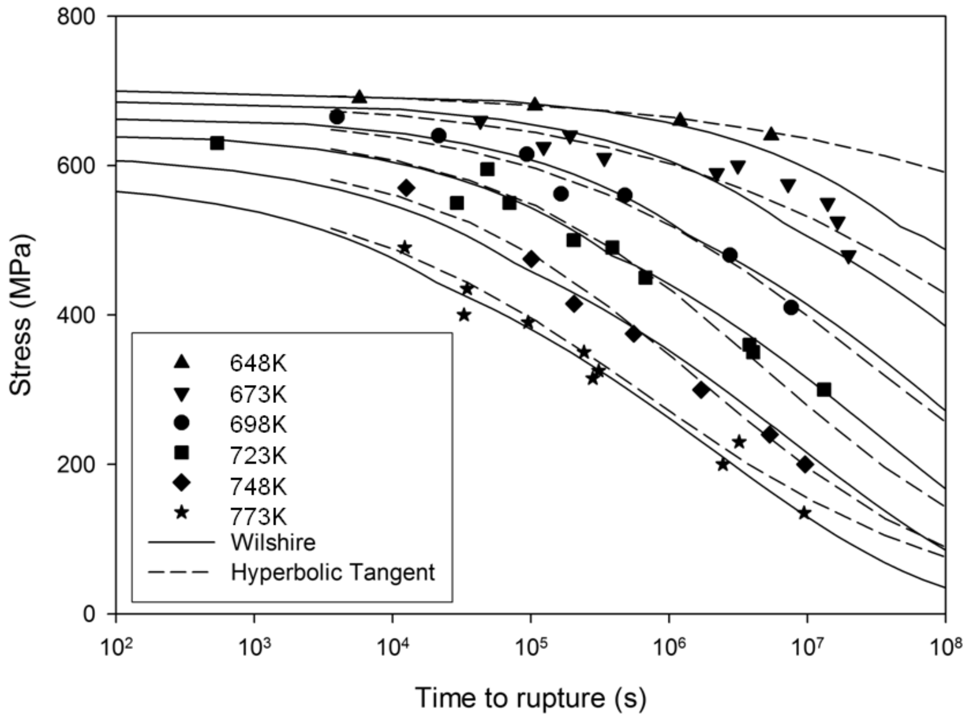


Figure 11: Predictions of the rupture time as a function of stress. Predictions made using the Hyperbolic tangent method (dashed lines) and the Wilshire equations, eqn. (3) (solid lines).

4. Discussion

The ability to extrapolate creep test data outside of the boundaries of original test data is clearly of great importance to design engineers, since alloys are often utilised at temperatures beyond their original design envelope due to advances in lifing methodologies. The fact that both methods, the Wilshire equations and the hyperbolic tangent approach, seem to show excellent correlation with the test data indicates that they would be well suited to such an application.

Previous assessments using the Wilshire equations have also shown the ability to extrapolate accurately towards longer lives. This is extremely useful in creep dominated applications such as power plant steels. However, for aerospace applications, titanium alloys are usually employed in components where the life is curtailed by fatigue rather than creep. The ability to define adequate creep laws which can be extrapolated to conditions outside of the original dataset, however, is of paramount importance in modelling the stress-strain characteristics of the material, particularly around stress concentrations where fatigue cracks may initiate. As such, both of these methodologies show themselves to be able to characterise creep behaviour over the relevant temperatures and creep lives, with the hyperbolic tangent showing slightly improved fits.

Both methods detailed here rely on the normalisation of the stress by the UTS, and it is clear that the resultant activation energy (Q_c^*), which is calculated using the Wilshire equation method is a value which is significantly closer to the activation energy for lattice diffusion in the material than the range of values produced by traditional approaches. However, as illustrated by figures 4 and 5, extrapolation of data using a traditional power law type method is still difficult due the varying n value seen in both rupture time and minimum strain rate graphs. The lines in figures 4 and 5 represent solutions to the Wilshire equations derived by the methods detailed previously. It can be seen the predictions made are extremely accurate and overcome this issue of variable constants.

As discussed, a clear break occurs in the ln [t_f . exp ($-Q_c^*/RT$)] against ln [-ln (σ/σ_{TS})] graph used to derive the Wilshire equation for rupture times (eqn. (3)), and a similar break occurs in the graph used to derive the minimum strain rate equation (eqn. (4)) and also the time to 1% strain (eqn. (5)). The results from the rupture time graph are encouraging, with the break in the data occurring at 90-100% of the yield stress, σ_Y. The reason for this change in behaviour is clear. For stresses below σ_Y, dislocation creep will be the main source of strain accumulation, with the movement of pre-existing dislocations being the dominant factor. However, for stresses above $\sigma_{Y,}$ new dislocations will be generated during the initial strain on loading. Since titanium alloys, including Ti6-4, usually show a low rate of strain hardening, indicating minimal dislocation interaction for stresses above yield, only a subtle change in the behaviour above and below yield is observed, compared with other alloy systems [20]. Evidence for the change in behaviour can be clearly seen in figure 12, which shows the extent of the strain on loading (ε_0) for each series of tests. The horizontal axis shows the applied stress normalised by the yield stress at each temperature and it is clear that values of ε_0 increase rapidly above σ_Y, as further dislocations are generated. By considering this behaviour, the shape of the predicted curves can be rationalised. For stresses above the yield stress, the predicted curve shows quite a steep decline as the stress decreases. However, as the yield stress is reached, dislocations are no longer generated during the initial strain. Creep now relies primarily on existing dislocations and, as such, predicted lives tend to be longer, hence the shallower slope in the curve derived for low stresses.

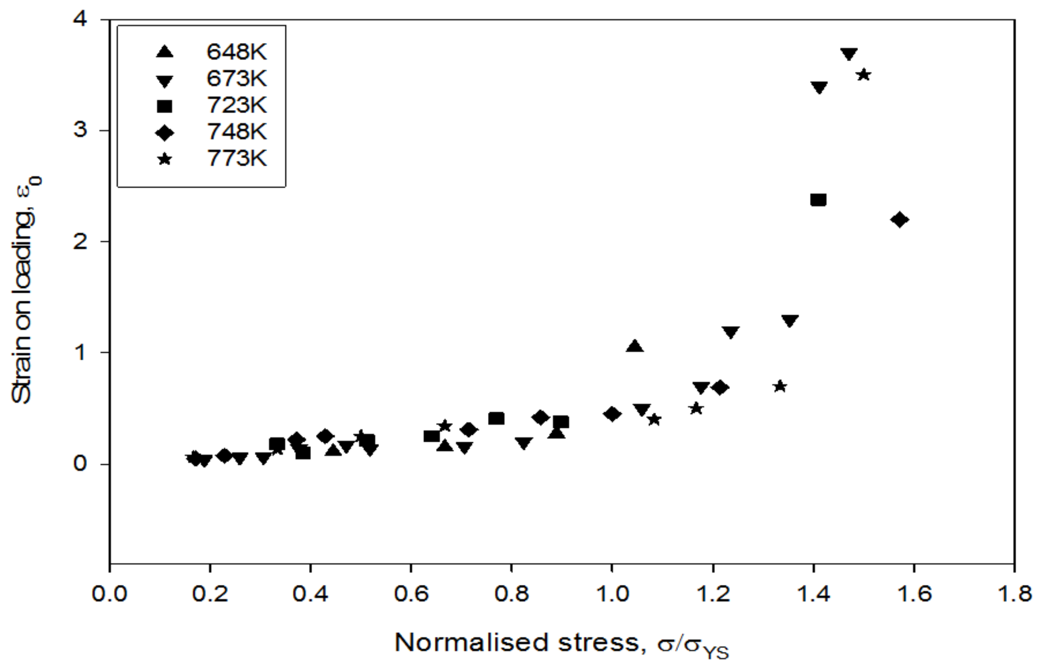


Figure 12: Strain on loading as a function of the stress normalised by yield stress. It is clear that above the yield stress new dislocations are generated at the onset of plasticity.

Furthermore, it is encouraging that a similar break is found in deriving the minimum strain rate (and time to strain) equation, figure 7, and that the break occurs at a similar value of stress, again indicating the effect of the onset of plasticity in the material. It is the fact that in applications of the Wilshire equations, physical effects such as the yield stress, or microstructure degradation [21,22] produce marked transitions in the predicted curves that helps to provide confidence in the method. Whereas in the current material, the hyperbolic tangent method produces excellent predictive curves, its ability to extrapolate results to longer lives still remains relatively untested.

It is clear that the although the Hyperbolic Tangent method is well utilised for creep life prediction of aerospace alloys [18], the Wilshire Equations may offer a valuable alternative technique which is able to provide more fundamental understanding of the alloys in question. The derivation of an activation energy related to that for lattice self diffusion, along with a change in behaviour at the material yield stress imply that the method relates well to observed experimental phenomena. The method has also been shown to predict minimum strain rates and times to particular strains accurately using simplistic approaches which are consistent with the approach for deriving rupture times.

Extrapolation to conditions outside of the initial dataset is clearly a test of any model, and the current paper seeks to demonstrate the capability of both approaches for a realistic industrial issue. Increased confidence in the ability of creep-fatigue lifing often leads to the employment of materials at temperatures in excess of those they were originally considered for. As such relevant test data at higher temperatures may be scarce or non-existent. However, both the Hyperbolic Tangent and Wilshire Equation methods are shown to extrapolate well to a higher temperature with the Wilshire Equations also capable of similar extrapolations for both minimum strain rate and time to 1% strain. Clearly this has added relevance for designers who may wish to remove the alloy from service at a specified amount of creep deformation to prevent contact between different components.

5. Conclusions

- Power law equations of the type detailed in eqn (1) are extremely limited in their applicability to materials such as Ti6-4. More realistic predictive methods such as the Hyperbolic Tangent method or the Wilshire equations should be employed.

- Both the Hyperbolic Tangent method and the Wilshire Equations show excellent results when attempting to accurately life creep tests over specified temperature and stress ranges in Ti6-4, and are capable of extrapolation to a higher temperature. The Wilshire equations also show excellent predictions of the minimum creep rate and time to 1% strain behaviour as a function of stress.
- The Wilshire equations remove uncertainty in the values of the so-called 'power law' constants, n and Q_c, predicting a lower value for the activation energy of 250kJmol^{-1} and a change in behaviour above and below the yield stress, associated with the generation of new dislocations on load application.
- Creep life approaches such as those described in the current paper should be employed in preference to more traditional power law based techniques due to improved accuracy and more realistic predictive curve shapes which provide more intuitive UTS and zero stress creep test predictions.

6. Acknowledgements

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References

[1] WJ Evans, MR Bache, *Int. J. Fatigue*, **16, (**1994) p443-452.

[2] MR Bache , *Int. J. Fatigue*, **25**, (2003) p1079-1087.

[3] V Hasija, S Ghosh, MJ Mills, DS Joseph, *Acta Materialia*, **51**, (2003) p4533-4549.

[4] FPE Dunne, D Rugg, A Walker, *Int. J. Plasticity*, **23**, (2007) p1061-1083.

[5] V Sinha, MJ Mills, JC Williams, *Met. Mat. Trans. A*, **37**, (2006) p2015-2026.

[6] PG Partridge, *Met. Revs.*, **12,** (1967) p169-194

[7] FC Monkman, NJ Grant, *Proc. ASTM* , **56**, (1956) p593-620

[8] B Wilshire, AJ Battenbough, *Mater. Sci. Eng. A*, **443**, (2007) p156-166.

[9] B Wilshire, H Burt, N Lavery, *Mater. Sci. Forum*, **519-521**, (2006) p1041-1046.

[10] B Wilshire, PJ Scharning, *Scripta Mater*., **56**, (2007) p701-704

[11] B Wilshire, PJ Scharning, *Scripta Mater*., **56**, (2007) p1023-1026

[12] D Rugg, M Dixon, FPE Dunne, *J. Strain Anal. For Eng. Design*, **42**, (2007) p269-279.

[13] R W Evans, B Wilshire "Creep of Metals and Alloys" Publ. The Institute of Metals, London, 1985

[14] T Seshacharyulu, SC Medeiros, JT Morgan, JC Malas, WG Frazier, YVRK Prasad, *Scripta Mater*, **41**, (1999) p283–288.

[15] P Rangaswamy, H Choo, MB Prime, MAM Bourke1, JM Larsen "THERMEC 2000 International Conference on Processing & Manufacturing of Advanced Materials, December 4-8, 2000, Las Vegas, USA

[16] MJR Barboza, EAC Perez, MM Medeiros, DAP Reis, MCA Nono, F Piorino Neto, CRM Silva, *Mat. Sci. Eng A*, , **428,** (2006) p319–326

[17] B Wilshire, MT Whittaker, Acta Materialia, **57**, (2009) p4115-4124

[18] SJ Williams, MR Bache, B Wilshire, *Mat. Sci. Tech.*, **26**, (2010) p1332-1337

[19] RW Evans and B Wilshire: 'Introduction to Creep': 1993, London, The Institute of Materials.

[20] M Whittaker, M Evans, B Wilshire, *Mat. Sci. Eng. A*, **552** (2012) p145-150.

[21] B Wilshire, PJ Scharning, *Mat. High Temp.*, **25**, (2008) p55-65

[22] MT Whittaker, B Wilshire, *Met. Mat. Trans. A*, (2012) *In Press*