The evolution of the Wilshire Equations for creep life prediction

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Abstract

In the past decade a new approach to predictive creep lifing has been developed, known as the Wilshire Equations. Having been applied to a range of power generation and aerospace materials, the understanding of material behaviour associated with the equations has developed significantly. With the equations based around the dominance of diffusion controlled dislocation movement for creep deformation under typical engineering stresses and behaviours, the predictions made are related to microstructural phenomena, such as the onset of yield. The current paper seeks to review the application and development of the Wilshire Equations, with suggestions for future research in the area.

Keywords

Creep, Wilshire Equations, Predictive lifing, Rupture

Introduction

At temperatures above approximately 0.4Tm (where Tm is the absolute melting point), materials experiencing an applied stress are considerably affected by time dependent strain accumulation known as creep deformation. Traditionally the creep behaviour of pure metals and alloys has been described as showing a decaying primary creep rate followed by a secondary or steady state phase which eventually gives way to an accelerating tertiary phase.

Based on this relationship between strain and time, a minimum strain rate can be defined for typical creep curves, for which power law equations of the form shown in eqn (1) describe the relationship with stress and temperature

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

where R=8.314Jmol^{-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. Based on this premise, the description of creep behaviour has been related to the diffusion controlled movement of dislocations when n≈4 or more. A decrease to n≈1 has then been widely attributed to a transition to diffusional mechanisms which do not involve dislocation movement. Theories have evolved to describe this directional flow of vacancies which occurs predominantly through the grain interiors at ~ 0.7T_m and above, termed Nabarro-Herring creep [1,2], and more widely along grain boundaries as temperatures are reduced towards 0.4T_m (Coble creep) [3]. In Nabarro-Herring creep therefore, n = 1, and $Q_C = Q_L$, where Q_L is the activation energy for lattice diffusion whereas in the Coble creep range, n = 1 and $Q_C = Q_{GB}$, where Q_{GB} is the activation energy for grain boundary diffusion.

Alongside this mechanism change, further complications occur when considering particle hardened alloys for which n values of 16 or greater are often calculated. The rapid variation in slope of typical log σ/log ε graphs, Figure 1, therefore prohibits accurate extrapolation techniques outside of the known data.

The requirement for accurate description of creep behaviour is critical in any high temperature operation, with creep acting as a single deformation (and potentially failure) mechanism, or alongside other damage mechanisms such as fatigue or oxidation. In static operations such as pipework carrying high temperature steam around power plants creep is the dominant mechanism and therefore the life limiting factor for plant operations. Typically power plants are designed to operate over extremely long periods of time, and therefore it is critical that design parameters are known, the most important of which is usually the allowable stresses which can be applied without causing failure within 250,000h. The requirement for these values however has significantly hampered development of the field however, since no method of accelerated testing has proved wholly reliable and no technique has proved capable of extrapolating short term data to a reliable long term prediction [4].

Methods such as the Larson-Miller method [5], Manson-Haferd [6] method and a range of other techniques have in the past proved notoriously inconsistent in a predictive capacity. Demonstrations of the difficulties associated with these techniques are wide ranging, although a particular example is that of the 9-12% Cr steels which were introduced in order to allow plant operation to be raised to over 870K (597°C). Based on testing at the National Institute for Materials Science (NIMS) in Japan, and other international establishments, the allowable strengths have been reduced progressively as the maximum test duration has increased beyond 30,000h. A number of modern approaches have been developed in recent years which have sought to address the issue with varying degrees of success (A detailed evaluation of these techniques is provided elsewhere [7]). However with the requirement for increased accuracy, the number of variables has also increased, along with the difficulty of the fitting procedure. The relationship between these approaches and the fundamental mechanisms governing creep behaviour also remains unclear. It is therefore apparent that further development is required within the field to provide a more standardised approach.

The Wilshire Equations

The Wilshire Equations were first proposed in 2007 by Wilshire and Battenbough [8] in the pursuit of more physically relevant representations of creep data prediction. As previously described, traditional theories predict that a mechanism change occurs between low stress ($\sigma<0.5\sigma_Y$) and high stress data, with vacancy diffusion dominating at low stresses and diffusion controlled dislocation movement. Utilising traditional approaches the transition between n≅1 and n≅4.5 occurred at $\sigma=0.5\sigma_Y$ (where σ_Y is the yield stress), Figure 1. The authors argue however, that this transition point is an artefact of the methodology being utilised, and that since no significant changes occur in microstructural behaviour at this point, the approach is limited. Indeed, the authors state that "In contrast to the idea of a distinct mechanism change, similar behaviour patterns are recorded in

stress change experiments carried out in the $n \cong 4.5$ and $n \cong 1$ regimes, indicating that the dislocation processes governing strain accumulation differ only in detail above and below $\sigma = 0.5\sigma_Y$."

Since the data provided appears to indicate diffusion controlled movement of dislocations to be the dominant process in these creep experiments, it is natural to suggest an approach to representing creep behaviour which encompasses this assumption and attempts to align with observed micro scale phenomena.

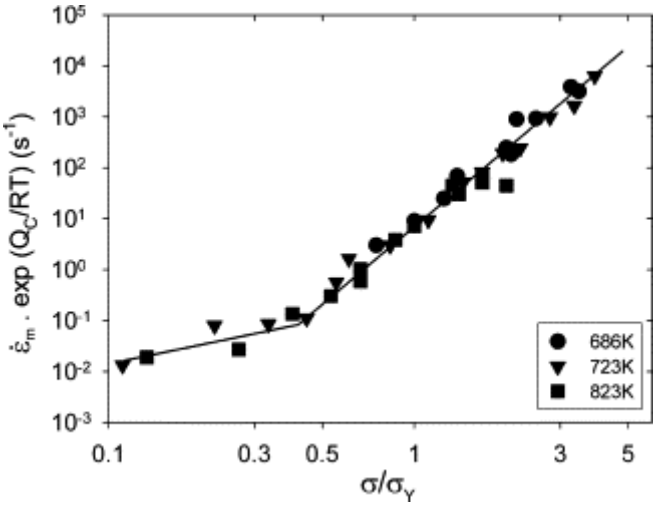

Figure 1: Traditional approach to determining creep behaviour, with n≅ 1 for $\sigma<0.5\sigma_Y$ and n≅ 4.5 for $\sigma>0.5\sigma_Y$.

Wilshire noted that a more robust approach can be adopted if it is considered that a transition occurs only at $\sigma=\sigma_Y$, which clearly aligns with observed behaviour. For stresses below the yield stress ($\sigma<\sigma_Y$) it is proposed that creep deformation occurs predominantly through the diffusion controlled movement of pre-existing dislocations within the structure under the action of a shear stress. However, above the yield stress ($\sigma>\sigma_Y$) new dislocations are continuously generated and this additional net movement of atoms contributes to a significant increase in creep rate than would be expected by simply extrapolating from sub-yield results.

The concept of this type of approach is not new. 'Region splitting' is the term given when a change in creep behaviour has been observed at certain stresses and is thought to be linked to a change in dominant creep mechanism. The concept is that activation energy in one stress region dominated by one creep mechanism cannot necessarily be extrapolated to another region dominated by a different creep mechanism. The earliest example of this was by Bernstein [9] who developed an equation with two different activation energies that dominated at different ends of the stress range examined for Zr and Zr alloys. Following this work, Foldyna *et al.* also implemented this technique for high chromium ferritic steels [10]. More recently, this technique has been consistently used by Kimrua *et al.* [11,12] amongst a number of other authors.

Based on these approaches, new relationships were proposed, which have become commonly known as the Wilshire Equations. These relationships quantify t_f and $\dot{\varepsilon}_m$ as

$$\left(\sigma / \sigma_{TS}\right) = \exp\left\{-k_1\left[t_f \exp(-Q_C^* / RT)\right]^u\right\} \quad (2)$$

and

$$\left(\sigma / \sigma_{TS}\right) = \exp\left\{-k_2\left[\dot{\varepsilon}_m \exp(Q_C^* / RT)\right]^v\right\} \quad (3)$$

where, k_1, k_2, u and v are fitting constants and Q_c^* is calculated at constant (σ/σ_{TS}) rather than at constant σ as in the determination of Q_c with equation (1). In these cases, σ_Y and σ_{TS} are the yield and ultimate tensile stresses determined at high-strain-rates (~$10^3 s^{-1}$) at the creep temperatures for each batch of material investigated. Although the $\dot{\varepsilon}_m$ and t_f data sets are rationalized by normalizing σ by σ_Y or σ_{TS}, σ_{TS} is preferred because equations (2) and (3) cover the full stress range from $\sigma/\sigma_{TS} \rightarrow 1$ to $\sigma/\sigma_{TS} \rightarrow 0$, while it is straightforward to distinguish between results obtained above and below σ_Y. Also, eqns (2) and (3) have the obvious advantage that, with the sigmoidal ln $\dot{\varepsilon}_m$ and ln t_f against (σ/σ_{TS}) plots, $\dot{\varepsilon}_m \rightarrow \infty$ and $t_f \rightarrow 0$ when $(\sigma/\sigma_{TS}) \rightarrow 1$, while $\dot{\varepsilon}_m \rightarrow 0$ and $t_f \rightarrow \infty$ as $(\sigma/\sigma_{TS}) \rightarrow 0$.

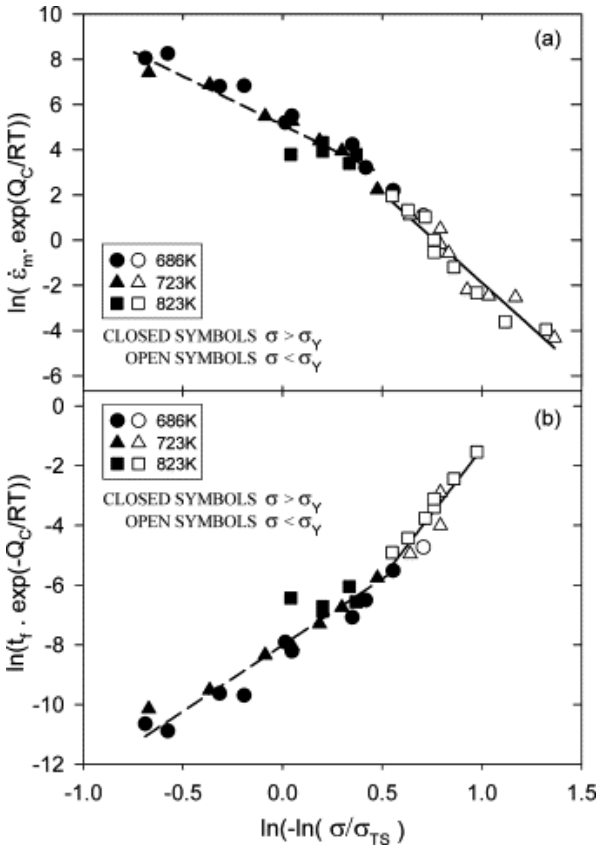

Figure 2: Transitions in behaviour of copper a) minimum strain rate data b) rupture life data

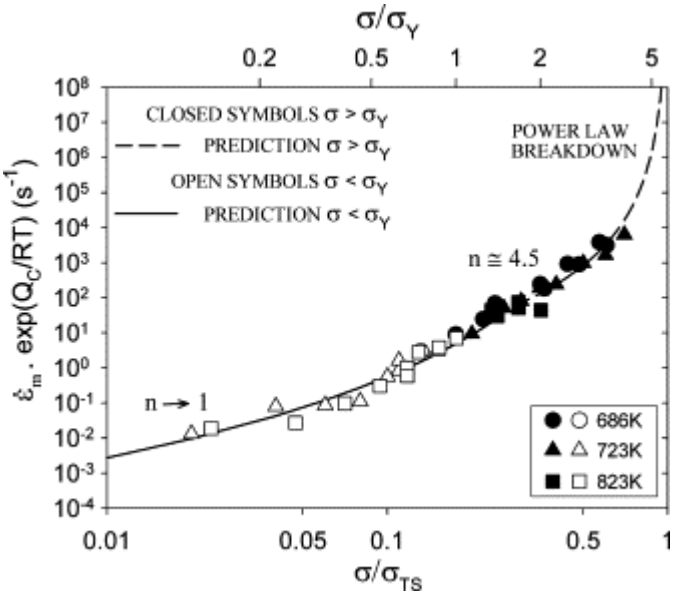

Figure 3: Predictions made by the Wilshire Equations of creep behaviour in copper.

Distinct changes are seen in properties such as initial extension and primary creep at the yield stress, indicating the nature of the microstructural changes occurring as elastic bonds are broken and dislocations form and clear evidence was shown of this in copper [8]. Plots of $\ln\left[t_f \exp(-Q_C^* / RT)\right]$ against $\ln\left[-\ln(\sigma / \sigma_{TS})\right]$ and $\ln[\varepsilon_m \exp(Q_c^* / RT)]$ against $\ln\left[-\ln(\sigma / \sigma_{TS})\right]$ were then utilised in order to derive the constants for copper. Interestingly, as shown by Figure 2, a transition or 'break point' occurs in the data as in Figure 1. However, analysis of the data shows that this transition now occurs at $\sigma \cong \sigma_Y$, rather than $\sigma \cong 0.5\sigma_Y$. Clearly this more adequately represents the changes in behaviour previously described at the yield point. The basis of these equations then allowed the authors to provide an excellent fit to the data shown in Figure 1, where the regimes where diffusional creep, dislocation creep and power law breakdown would traditionally be expected, are all predicted well by the Wilshire Equations, Figure 3.

The work on copper aided significantly in establishing a fundamental link between the Wilshire Equations and microstructural effects such as yield. However, follow up work [13-15] focussed on the potential applications to industry through extrapolation, particularly in terms of rupture time. Wilshire and Scharning attempted to apply the Wilshire Equations to industrially relevant steels such as 1Cr-0.5Mo, 1Cr-1Mo-0.25V, along with the 9 Cr series of alloys such as Grades 91, 92 and 122.

A common theme across the alloys is the application of an activation energy, Q_c^*, derived by correlation of data at constant (σ/σ_{TS}) rather than at constant σ as has previously been adopted by traditional approaches. Data is correlated using either plots of $\ln\left[t_f \exp(-Q_C^* / RT)\right]$ against $\ln\left[-\ln(\sigma / \sigma_{TS})\right]$ or $\ln[\varepsilon_m \exp(Q_c^* / RT)]$ against $\ln\left[-\ln(\sigma / \sigma_{TS})\right]$, with Q_c^* being optimised by best fit techniques. In each of the steel alloys described above, a value of 300kJmol^{-1} was derived for

Q_c*, approximately equal to the value of the activation energy for self diffusion in the iron lattice, further strengthening the relationship of the equations with observable microstructural phenomena.

Indeed, predictions made by Wilshire and Scharning [14] begin to demonstrate the potential of the Wilshire Equations for extrapolative purposes, Figure 4. As previously described in copper, data for 1Cr-0.5Mo steel indicates a break point at $0.8\sigma_{PS}$ ($\cong\sigma_Y$). The difference in predictive curves formed by using data either based on $\sigma>0.8\sigma_{PS}$, or $\sigma<0.8\sigma_{PS}$, is clear, as demonstrated in Figure 4, with an order of magnitude difference in predictions of data approaching 150,000h, indicating the importance of appropriate use of the region splitting methods previously described. Based on this type of approach it is evident that the predictive lines produced by applying the Wilshire Equations, which are based only on test data with a rupture life less than 30,000h, provide an excellent prediction of long term creep behaviour.

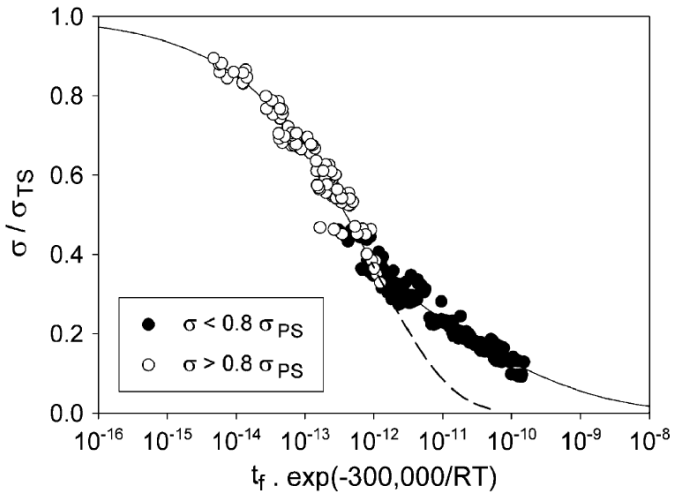

Figure 4: Predictions made by the Wilshire Equations of rupture behaviour of 1Cr-0.5Mo steel derived from data with t_f<30,000h.

With the extrapolation capability of the alloy well established, attention turned to alternative alloys such as HK40 [16], a super-austenitic steel, widely used for tubes or as static castings for applications such as retorts and supports, hangers and other parts for process and heat-treatment furnaces. These heat-resisting steels provide good creep strength together with oxidation and carburisation resistance at temperatures up to and exceeding 1000°C.

Efforts to compare the capability of the Wilshire Equations against long established techniques such as the Larson-Miller technique [5] were the focus of an initial publication on HK40, with predictions made using both techniques based on using only t_R<30,000h. Results for 100,000h strengths derived using the Wilshire Equations from these test results were lower than the values estimated by applying the Larson-Miller method to the full NIMS data set. Yet, in several recent studies employing a number of different analysis procedures to determine the allowable design stresses, the Larson-Miller equation consistently produced the highest long-term strength estimates [17,18]. Indeed, in

two investigations [19,20], it was necessary to discard short-term t_f measurements in order to avoid serious overestimation of the 100,000h strengths.

Later publications [21] reduced the amount of data required for accurate predictions to t_R<5000h, along with exploring a similar alloy, HP40, used in similar applications. Perhaps most interestingly for the alloys however, was the determination of a value of the apparent activation energy, Q_c*=100kJmol^{-1}, significantly lower than the value for self diffusion in the iron lattice. Whilst it should be considered that although HK40 and HP40 are nickel based, they do include high quantities of alloying elements, HK40 (25%Cr-20%Ni-0.4%C), HP40 (25%Cr-35%Ni), the value still remains remarkably low. Whilst no explanation was determined for the low value at this point Wilshire and Whittaker note the development of extensive cracking along the columnar grain boundaries.

The same authors investigated the behaviour of 2.25 Cr steels under creep loading [22], with these alloys acting as a mainstay of the power generation sector, having been extensively used for superheater and reheater tubing, as well as for headers and piping. Grade 22 (2.25Cr-1Mo) steel is a well established alloy which has seen widespread use in power plant, however, the series of alloys provides an excellent example of the issues associated with the implementation of new alloys into service. Figure 5 illustrates the stress rupture behaviour of Grade 22 steel at 600°C, alongside that of the more recently developed Grade 23 (2.25Cr-1.6W) and Grade 24 (2.25Cr-1Mo-0.3V) alloys. Clearly a marked improvement in stress rupture behaviour is evident in both alloys when compared with the Grade 22 data. However, with tests still currently ongoing in Grade 23 and particularly Grade 24, only limited confidence can be placed in 100,000h stresses for these alloys due to the lack of confidence in predictive techniques.

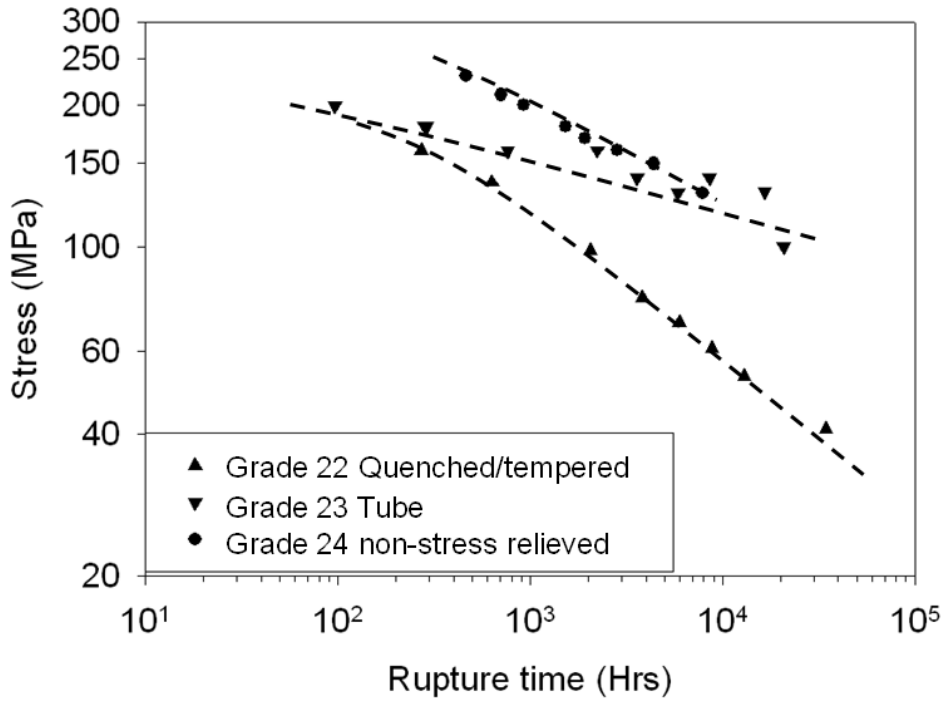

Figure 5: The stress dependence of the creep life at 873 K (600°C) for Gr. 22, 23, and 24 steels.

The Wilshire Equations however do provide some cause for optimism. A range of publications [23-25] has investigated stress rupture behaviour in these alloys and shown the capability of the method to predict accurately 100,000h stresses based only on the use of 5000h data. Clearly these aggressive approaches to prediction demonstrate the stability of the equations with regards to predictive capacity. In the case where longer term data could be utilised (as in alternative techniques where t_f<30,000h data is employed) clearly further confidence could be placed in the predicted results.

The excellent results obtained are more impressive when the complex stress rupture behaviour of these alloys is examined in more detail. As opposed to previous materials considered using the Wilshire Equations, a second transition in behaviour (break point) was observed when attempting to derive the Wilshire equations, as shown in Figure 6.

Figure 6 is plotted by considering a single value of the activation energy for convenience, Q_c*=280kJmol^{-1}, using only data where tf<5000h. This enables the determination of three regions within the data which should then be considered separately. Howeverin order to provide more optimised fits the value of Q_c* is then allowed to vary within each individual region as previously described. Inclusion of shorter term higher temperature data aids significantly here in prediction of long term behaviour at lower temperatures. As indicated by the text on Figure 6, optimised fits are derived when Q_c*=280kJmol^{-1}for the high stress region, Q_c*=230kJmol^{-1} for the intermediate region and Q_c*=280kJmol^{-1} for the low stress region. Based on these values, excellent fits to the data are again provided, as shown in Figure 7, where the equations are produced using only t_f<5000h data.

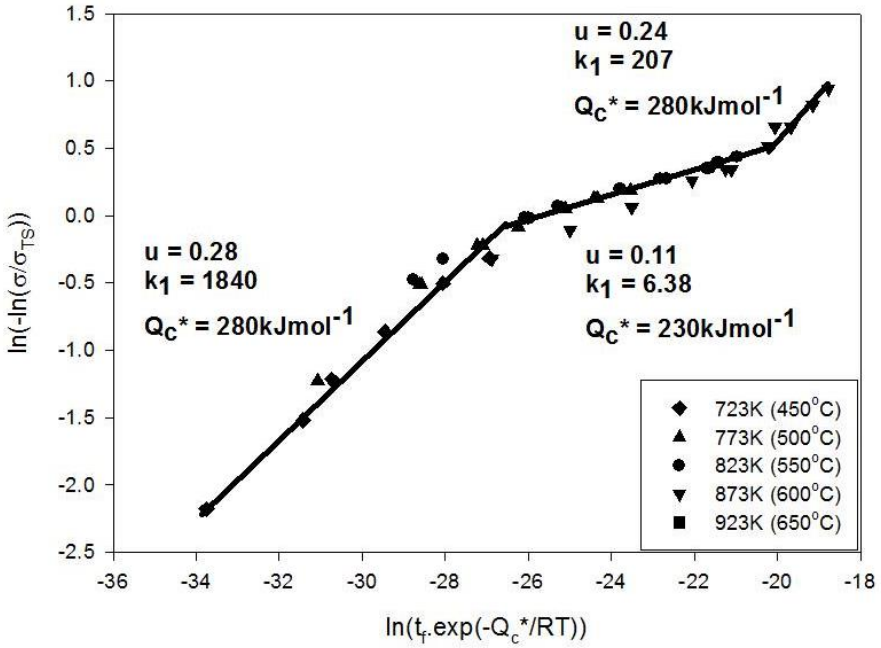

Figure 6: Adopting eqn.(2), the k_1, u and Q_c* values are determined by plotting of ln[t_f.exp(-Q_c^*/RT)] against ln[-ln(σ/σ_{TS})], for Gr.22 tube.

The current methodology appears to be consistent with micro mechanical phenomena. The seemingly complex behaviour of these alloys can be rationalized in a manner consistent with inferred microstructural behaviour (based on elongation and reduction in area measurements) and elements of traditional creep theory. Above the yield stress, creep takes place through the generation and movement of new dislocations, formed at appropriate sources, due to the fact that the yield stress of the material is exceeded. When the applied stresses fall below the yield stress, the creep lives therefore become longer than expected by direct extrapolation of results when $\sigma>\sigma_Y$ because grain deformation is restricted so that creep occurs only by grain boundary zone deformation when $\sigma<\sigma_Y$. A second break occurs in Figure 6 which is a consequence of the bainite regions in the initial ferrite/bainite microstructure degrading to ferrite and molybdenum carbide particles in long-term tests, with very coarse carbides along the grain boundaries. As a result, the creep rates are faster and the creep lives are significantly shorter in tests of long duration at 873 K (600 °C) and above that which would be expected by extrapolation of the property sets recorded for the intermediate stress data when the bainitic microstructures are present. This interpretation of the creep behaviour of the Gr.22 steels is also consistent with the changes in Q_c^* (Table II). Thus, when $\sigma>\sigma_Y$, new dislocations are created during the initial specimen extension, with creep occurring largely as a result of these dislocations moving within the grains. Under these circumstances, $Q_c^* \cong 280 kJmol^{-1}$, equivalent to that for self diffusion in bainitic matrices. When $\sigma<\sigma_Y$, so that creep takes place within the grain boundary zones rather than in the grains, $Q_c^* \cong 230 kJmol^{-1}$, representing diffusion along dislocations and grain boundaries. However, with the marked loss of creep strength in the Gr.22 tube when $\sigma<0.2\sigma_{TS}$, $Q_c^* \cong 280 kJmol^{-1}$, suggesting that diffusion occurs largely within the grains as the creep resistance falls when the bainitic regions transform to ferrite and coarse molybdenum carbide particles in long-term tests at high temperatures.

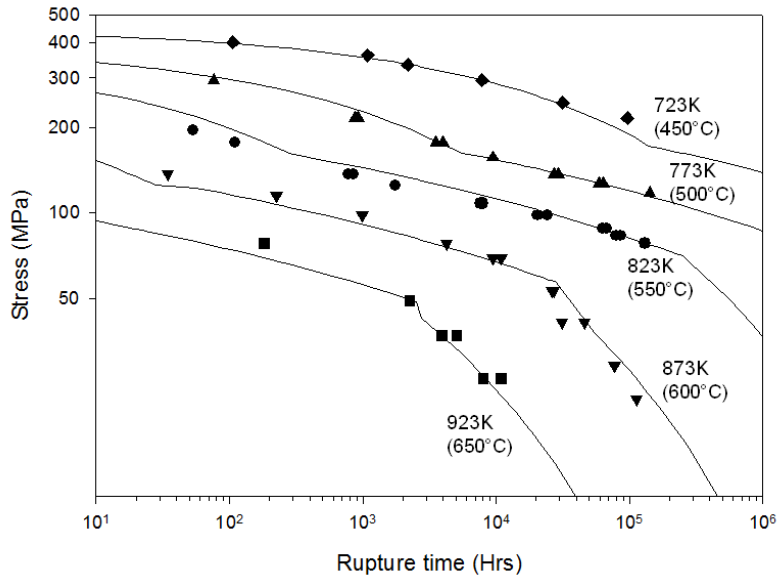

Figure 7: The stress dependence of the creep life at 723-923K (450-650°C) for 2.25Cr-1Mo steel tube (T22). The lines are drawn after analysing the results according to eqn. (2).

The basis of a non-constant activation energy also allowed for more appropriate predictions in alternative alloys. In particular Type 316H stainless steel showed significant differences in the value of Q_c^* above and below the yield stress of the material (250kJmol^{-1} above the yield stress, 150kJmol^{-1} below the yield stress) [26]. As previously described, the authors state that the value of 250kJmol^{-1} when $\sigma>\sigma_Y$ coincides with the value of self diffusion within the austenitic grains. However, when $\sigma<\sigma_Y$ ($<\sigma_{PS}$), Q_c^*=150 kJmol^{-1} the value is more relatable to the activation energy for diffusion along grain boundaries and related dislocations near to the grain boundaries. Again the measured values of Q_c^* quoted above can be related to measurable test parameters such as elongation at failure, where the authors indicate that strain at failure decreases as the stress falls below the yield stress, since the contribution of grain deformation to overall creep rate decreases significantly, in line with the previous assertions made.

Alternative publications by a range of authors have also developed confidence in the Wilshire Equations by applying them to materials which would traditionally be considered as difficult to predict. The work by Abdallah et al [27] is informative with regards to the intermetallic titanium aluminide alloy Ti-45Al-2Mn-2Nb. The titanium aluminide materials are generally regarded as potential successors to polycrystalline nickel alloys in a number of engineering applications, but suffer from an inherent lack of room temperature ductility which has limited their use in service. The work by Abdallah et al showed that creep ductilities at intermediate to high temperatures (600-750°C) far exceeded the extremely low values obtained in tensile tests at room temperature. Interestingly however, the intermetallic nature of the alloy did not affect the capability of the Wilshire Equations to affect an accurate prediction across a range of temperatures and stresses in the material. Further work by Abdallah et al [28], and also Harrison et al [29] has also been critical in the development of the Wilshire Equations for the aerospace industry, since application of the third Wilshire Equation (Eqn. 4) is investigated with regards to prediction of times to specific strain values. In each case methods are developed by which prediction of the entire creep curve is possible.

$$(\sigma/\sigma_{TS}) = \exp\{-k_3[t_\varepsilon \exp(-Q_c^*/RT)]^w\} \qquad (4)$$

Further applications of the Wilshire Equations to the titanium alloy Ti6-4 [30] and the nickel superalloy Waspaloy [31] further enhance the reputation of the equations, with the nickel superalloys in particular showing n values of 16 or higher, and hence sitting in the ‘power law breakdown’ region which is poorly described by traditional creep theories based on power law type equations. Interestingly Ti6-4, Waspaloy and 316H stainless steel provide an insight into the creep behaviour as described by the Wilshire Equations, and in particular the variations in the value of Q_c^* which occur above and below the yield stress. With Ti6-4 showing no change in Q_c^* value either side of the yield stress, Waspaloy showing a difference of approximately 60kJmol^{-1}, and 316H differing by 100kJmol^{-1}, it is natural to look for underlying mechanisms which may produce this effect. Since, as described previously by Wilshire [8], at reasonable stress and temperature values, creep occurs by the diffusion controlled movement of dislocations, it is reasonable to expect the dislocation

structure to play a significant role in the effect. It is therefore interesting to relate the difference in Q_c* values to the rate of strain hardening in these alloys, a simple but effective measure of dislocation interaction. Titanium alloys show an extremely low rate of strain hardening, whereas Waspaloy shows an increased rate. In type 316H stainless steel, with yield occurring at $\sigma \cong 0.35\sigma_Y$, the rate of strain hardening is particularly high. It is therefore undeniable that the change in activation activity predicted by the Wilshire equations approach is related to micro-mechanical behaviour, and in particular to dislocation activity and interaction.

As such, recent studies have sought to investigate this effect in Waspaloy [31], and the following conclusions were drawn. Below yield creep takes place through the movement of dislocations controlled by diffusive climb around precipitates, whereas above yield dislocation movement is limited by forest hardening. The change in apparent activation energy, Q_c^*, is directly related to the amount of strain hardening in an alloy brought about by high dislocation densities generated at stresses above yield. It is clear that further studies of this type will play an important role in the development of the Wilshire Equations, which offer a novel and important methodology towards a long standing problem of predictive lifing of creep behaviour

Conclusions

With no appropriate resolution to the issue of prediction of long term creep behaviour based on short term data, the Wilshire Equations offer an alternative approach to power law based equations and other techniques. The technique is based upon the assumption that only dislocation creep plays a significant role in the creep behaviour of metals and alloys for engineering stresses appropriate to most industrial application, and therefore no mechanism change occurs which would prevent the established predictive capacity of the equations. Instead, changes in behaviour are related to observable microstructural phenomena such as the yield stress, and overaging of structures which become unstable under a combination of high temperatures and long exposure times. Whilst this ‘region splitting’ has been previously attempted within the field the relationship with fundamental behaviour provides added confidence in the current method. With microscopic studies now proving supportive of the findings of Wilshire Equation analyses, future studies should seek to further develop understanding of the role of fundamental mechanisms in order to provide added confidence in the methodology.

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