4. CONCLUDING COMMENTS

Scientific progress is often based on the premise that careful observations should lead to theories capable of prediction. Yet, despite being widely supported for almost half a century, power-law concepts have not led to theories with verified predictive capabilities. Instead, reliance is still commonly placed on the use of parametric methods in order to obtain long-term stress-rupture estimates for high-temperature design calculations. Using these empirical procedures, protracted high-cost programmes involving large numbers of tests lasting up to 30,000 hours or more must then be completed in order to provide 100,000 hour data.

In contrast, numerous theoretical and practical benefits are gained by extending conventional approaches for creep data representation to include the ? methodology, which seeks to quantify the shape of individual creep curves and the variations in curve shape with changing stress and temperature. As now illustrated for polycrystalline copper, even straightforward curve shape analyses clarify the observed behaviour patterns in terms of the dislocation processes controlling creep strain accumulation and the damage processes shown to affect the tertiary acceleration and eventual failure. It is also evident that independently-measured 100,000 hour stress-rupture values reported for a range of commercial creep-resistant alloys can be predicted with reasonable accuracy by full ? analyses of the systematic shape variations of creep curves recorded in high-precision constant-stress tests lasting up to only 1000 hours or so[1, 5, 6], i.e. an extrapolation of two orders of magnitude in time compared with the three-fold maximum extrapolation achieved with traditional parametric methods. Finally, the theoretically justified ? methodology[2, 3] can be extended to quantify material behaviour under the complex non-steady stress-temperature conditions usually encountered in service applications[8], offering a computer-efficient approach suitable for use with the modern finite-element codes available for high-temperature design.

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By quantifying the overall curve shape, the ? methodology can be used to predict not just t_f values but any creep parameter. Yet, while long-term stress-rupture properties are available for many commercial creep-resistant alloys, allowing assessments to be made of the predictive capabilities of the ? relationships, (Figure 3), creep rate measurements have rarely been made for conditions approaching those expected in service. Indeed, because independently-measured low-stress ?$_m$ values were available for ½Cr½Mo¼V ferritic steel[7], samples from the same batch of material were obtained for initial evaluation of the ? concept[1]. As evident from Figure 4a, the ? predictions based on 1000 hour test data accurately matched the creep rates determined at stresses which would have given stress-rupture lives of well over 100,000 hours at 838K for the ½Cr½Mo¼V steel[7]. It should also be recognized that, given the accuracy of the t_f predictions shown for 2419[6] (Figure 3b), equivalent precision should be achieved when the same ? relationships are used to compute ?$_m$ values over comparable stress/temperature ranges (Figure 4b).

As with the logt_f/log? relationships (Figure 3), the predicted log ?$_m$/log? plots curve such that the stress exponent (n) increases continuously from low values at low stresses to high values as the applied stress is raised towards the UTS at the creep temperature (Figure 4). This gradual curvature then contradicts the common assumption that the log?$_m$/log? data for particle-strengthened alloys should be represented by successive regimes with n ? 1, n ? 4 and n > 4 as the dominant creep mechanism changes with increasing stress. Moreover, if different mechanisms controlled the creep properties (and therefore the creep fracture characteristics) within different stress/temperature regimes, analysing data collected in one mechanism regime could not predict the behaviour patterns displayed in another regime. The fact that ? analyses of short-term data allow long-term property values to be estimated with reasonable accuracy (Figures 3 and 4) therefore indicates that essentially the same dislocation processes are operative, irrespective of the n value recorded. Even so, the primary creep strains and the relative importance of the various processes governing the tertiary characteristics change as the test conditions are altered, resulting in systematic variations in creep curve shape which can be quantified through the ? relationships. The present analysis therefore supports the view[2, 3] that the gradual curvature of log?$_m$/log? plots over extended stress ranges (Figure 4) is a consequence not of changes in the dominant creep mechanism but of the variations in creep curve shape which occur with changing stress and temperature.

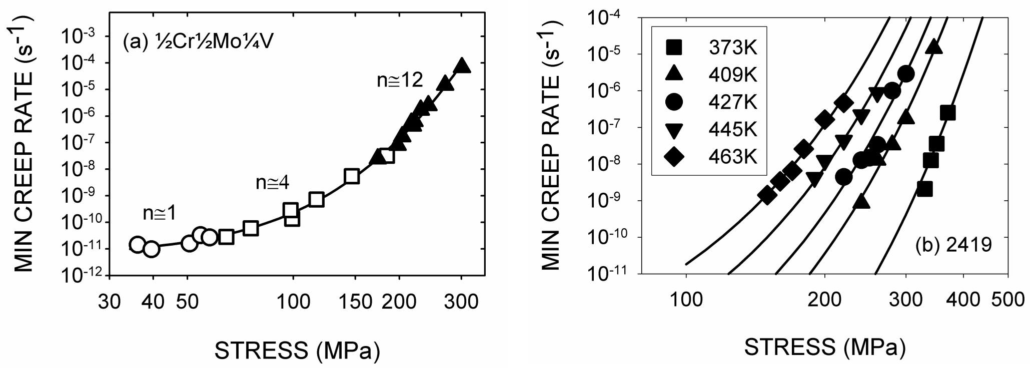

Figure 4 (a) The stress dependences of the minimum creep rates (solid lines) predicted by ? analysis of short-term constant-stress creep curves (a) for a ½Cr½Mo¼V steel at 868K[1] and (b) for 2419[6] at 373 to 463K (full symbols). The predicted behaviour for the ½Cr½Mo¼V steel[1] is compared with measured long-term data for the same batch of material[7] (open symbols).

3. ANALYSIS OF CREEP CURVE SHAPES

Although most academic studies seek to describe a set of creep curves by reporting only the $?_m$, t_f and $?_f$ values, the results presented in Figures 1 and 2 demonstrate that information relevant to identification of the detailed processes governing primary and tertiary behaviour can be gained from even a basic examination of curve shape changes. In fact, all features of the behaviour patterns now reported for polycrystalline copper are incorporated in the ? methodology[1-3], which offers both theoretical and practical advantages by quantifying creep curve shapes.

During creep of pure metals, dislocations experience a resistance to continued movement which depends on the local dislocation configurations. The primary creep rate then decays because the strengths of the barriers opposing dislocation movement increase as the dislocation density increases and the dislocation distribution becomes less uniform with increasing primary strain. These basic processes do not seem to be altered by the presence of fine dispersions of precipitates or insoluble particles, which simply impose additional barriers to dislocation movement, increasing the creep resistance. However, with pure metals and particle-strengthened alloys, the primary creep rate rarely decays to a ‘steady state’ value. Instead, a minimum rate occurs when the decaying primary rate is offset by the tertiary acceleration associated with one or more damage mechanisms, such as cavitation, neck formation and microstructural instability. Modelling of the processes controlling strain accumulation and damage evolution then supports a quantitative description of the variation in creep strain with time as[2, 3]

$$? \; ? \; ?_1(1 \, ? \, \exp(? \, ?_2 t)) \; ? \; ?_3(\exp(?_4 t) \; ? \; 1) \qquad (1)$$

where $?_1$ and $?_3$ scale the primary and tertiary strains, while $?_2$ and $?_4$ are rate parameters governing the curvatures of the primary and tertiary stages respectively.

Non-linear least-squares curve-fitting routines are available[2] to compute the best values of the four ? parameters in eqn. 1 for curves showing clearly-defined primary and tertiary stages. In addition to offering a good description of individual curves, the ? values then vary systematically with stress and temperature to represent the variations in curve shape with changing test conditions. In this way, for copper[3] as well as various creep-resistant alloys[1, 5, 6], analyses of high-precision constant-stress creep curves having maximum creep lives of only 1000 hours or so have been shown to predict independently-measured stress-rupture properties over stress ranges giving t_f values up to 100,000 hours or more, as illustrated in Figure 3 for a 1CrMoV rotor steel[5] and for the aluminium alloy, 2419[6].

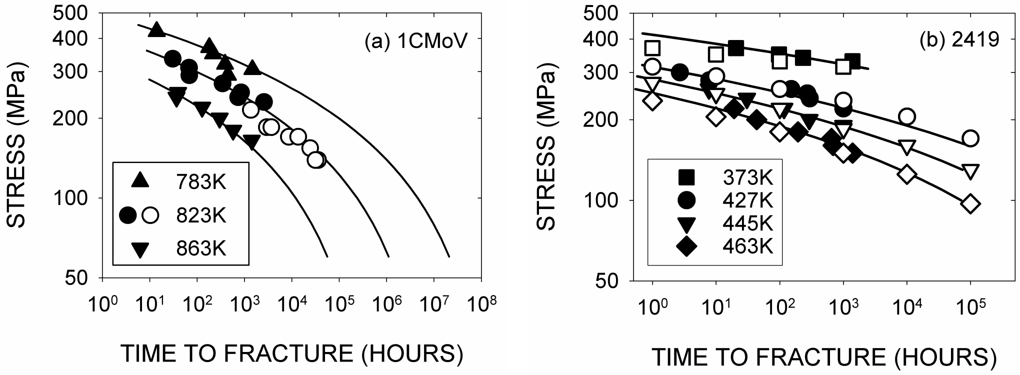

Figure 3. Long-term stress rupture behaviour (solid lines) predicted by ? analysis of short-term constant-stress creep curves (full symbols), compared with independently-measured long-term data (open symbols) obtained for constant-load tests completed (a) for a 1CrMoV steel at 783 to 863K[5] and (b) for the aluminium alloy, 2419 at 373 to 463K[6].

The variations in $?_p$ and $?_f$ with stress for copper (Figure 2a) can be linked to the magnitudes of the rapid-strain-rate yield stress ($?_y$) and the UTS value ($?_{TS}$), where both $?_y$ and $?_{TS}$ are determined at the creep temperature. In high-strain-rate tests (? ? 6.5x10^{-4}s^{-1}) for copper at 723K, $?_y$? 25MPa and $?_{TS}$? 80MPa. The primary strains are then large when rapid multiplication of dislocations takes place on loading a creep test at ? > $?_y$, whereas low $?_p$ values are found when relative little increase in dislocation density occurs on loading when ? < $?_y$ (Figure 2a).

As with the primary creep behaviour, the relative importance of the various processes contributing to the tertiary acceleration may differ above and below $?_y$. Under constant-stress conditions, the decrease in uniform cross-sectional area accompanying the increase in tensile creep strain does not affect the rates of creep strain accumulation, whereas neck formation causes a creep rate acceleration. Over the stress range now covered for copper, no evidence was found to suggest that the creep behaviour patterns were affected by grain growth or recrystallization so, apart from necking, the process most likely to influence the tertiary acceleration is intergranular damage development. In fact, under all conditions studied, failure occurred by the nucleation, growth and link-up of grain boundary cavities and cracks. However, when the primary strains were high at stresses exceeding $?_y$, (Figure 2a) the tertiary stage began with the onset of necking, with the cavities and cracks forming predominantly within the necked region of the testpieces. Conversely, cavitation caused both the tertiary acceleration and eventual failure when necking became progressively less evident as the applied stress was reduced below $?_y$.

Because the magnitude of the primary strain varies with stress (Figure 2a), the recorded $?_m$ value can also depend on whether constant-load or constant-stress procedures are employed. As illustrated in Figure 1a, the constant-load $?_m$ value progressively exceeds the constant-stress value as $?_p$ increases with increasing stress above $?_y$ (Figure 2a). Hence, using constant-stress machines, n ? 4.5 over the full stress range now studied for copper, whereas the equivalent constant-load plot indicates that n ? 4.5 in the lower stress range (when $?_p$ is small), but with n increasing as the applied stress is raised above $?_y$ (Figure 2b). The adoption of constant-load test procedures must then contribute significantly to the 'power law breakdown' region where n > 5 when $?_y$ < ? < $?_{TS}$. Moreover, with most creep data obtained using constant-load equipment, using creep properties derived from short-term constant-load tests can also introduce errors when projecting high-stress data to predict low-stress behaviour patterns, (Figure 2b).

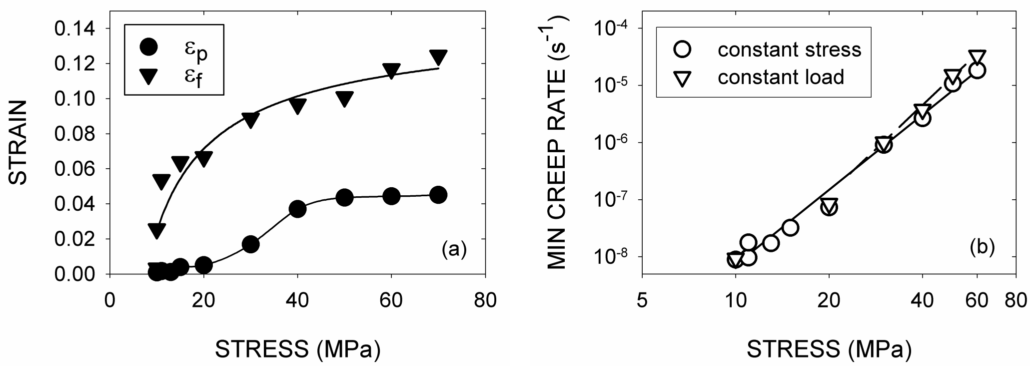

Figure 2 (a) The variations of the primary creep strain ($?_p$) and the total creep strain to failure ($?_f$) with stress for constant-stress tests carried out at 723K for copper. (b) The stress dependences of the minimum creep rates recorded in constant-stress tests at 723K for copper, together with the corresponding data computed for constant-load test conditions.

2. VARIATIONS IN CREEP CURVE SHAPE

The creep and creep fracture properties of metals and alloys are generally determined under uniaxial tensile stresses, so the theoretical benefits of curve shape analysis are now illustrated by reference to tensile creep data obtained for pure polycrystalline copper.

The basic information derived from a tensile creep test is a record of the variation in creep strain (?) with time (t), with 'normal' creep strain/time curves usually observed (Figure 1a). In seeking to identify the processes governing creep strain accumulation, it is then commonly assumed that the minimum or secondary creep rate ($?_m$) is the most significant parameter, with the creep properties of a material at a fixed creep temperature often described using Norton's Law[4], as $?_m$? $?^n$. However, over extended stress ranges, standard log$?_m$/log? plots curve such that the stress exponent (n) decreases with decreasing applied stress (?). In turn, because the creep life (t_f) frequently depends inversely on $?_m$, the corresponding logt_f/log? plots also curve, with this curvature causing the problems inherent in parametric extrapolation exercises.

A flaw in these conventional approaches is revealed simply by re-plotting ?/t curves (Figure 1a) to show the changes in creep rate (?) as a function of strain or time, (Figure 1b). In general, $?_m$ is a minimum rather than a 'steady state' value, with $?_m$ reached when the decaying primary rate is offset by the tertiary acceleration. Moreover, the curve shapes change as the test conditions are altered (Figure 1a), but the curve shape also depends on whether constant-load or constant true-stress test methods are used. With materials displaying low total creep strains to failure ($?_f$), virtually identical curves will be displayed with constant-load and constant-stress procedures. In contrast, when the creep ductilities are high (say, $?_f > 0.05$), the continuous increase in stress which occurs as the specimen cross-sectional area decreases with increasing creep strain can cause a constant-load creep curve to be distorted considerably compared with the corresponding constant-stress curve. Thus, the constant-stress curves for copper in Figure 1 show that, as with most other materials, the total primary strain ($?_p$) usually decreases and the tertiary stage becomes a more dominant feature with decreasing stress at a fixed creep temperature. Consequently, compared with constant-stress curves, constant-load test methods cause greater creep curve distortions when significant primary strains are exhibited at high stresses, with lower distortions found with the tertiary-dominated curves observed at low stresses (Figure 1a).

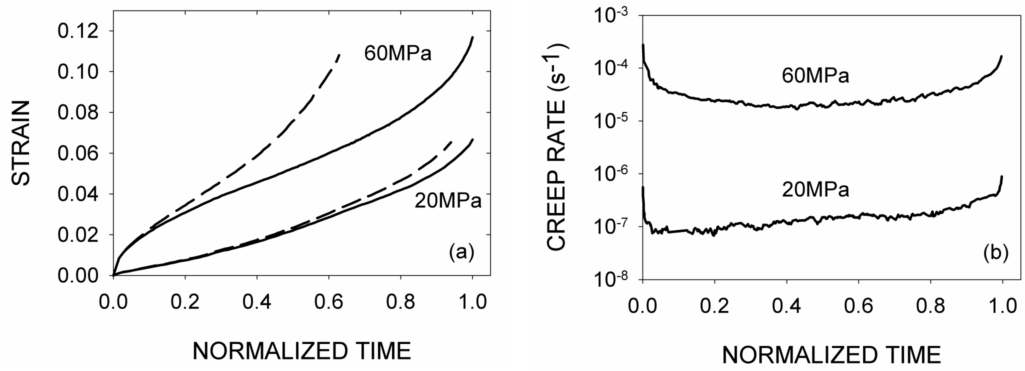

Figure 1 (a) The variations in creep strain (?) with normalized time (t/t_f) for constant-stress tests at 723K for copper at 60 and 20MPa (solid lines), together with the corresponding curves computed for constant-load conditions (broken curves). (b) The variations in creep rate (?) with normalized time for these constant-stress tests. At 60MPa, $t_f = 3.5 \times 10^3$s, whereas $t_f = 4.2 \times 10^5$s at 20MPa.

ACQUISITION AND ANALYSIS OF SHORT-TERM CREEP CURVES FOR PREDICTION OF LONG-TERM ENGINEERING DESIGN DATA

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ABSTRACT

Using empirical parametric methods, the provision of long-term stress-rupture estimates for any new commercial alloy is a protracted and expensive task, so innovative approaches are needed for rapid low-cost prediction of engineering design data. The selected extrapolation procedure must then be accurate and, preferably, theoretically sound. One proposed option is the ? methodology, which quantifies the shape of individual creep curves and the variations in curve shape with changing stress and temperature. Evidence is therefore provided to demonstrate that, for polycrystalline copper, curve shape analysis explains the observed behaviour patterns in terms of the deformation and damage processes controlling strain accumulation and fracture. In addition, with considerable accuracy, the ? relationships derived from high-precision constant-stress creep curves lasting up to only 1000 hours are shown to predict independently-measured 100,000 hour data for a range of creep-resistant alloys.

1 INTRODUCTION

For design of high-temperature components and structures, an accurate knowledge is required of the stresses which the relevant engineering materials can sustain without creep fracture occurring within the planned design life. However, because of the long operational lives usually specified, acquisition of the necessary stress-rupture data represents a protracted and expensive task, even when various parametric methods are adopted to estimate long-term behaviour. Particularly with engineering steels, the experimentally-determined stress-rupture properties are often characterised by high degrees of scatter, even though stringent national and international standards are defined for test procedures which should give acceptable levels of accuracy and reproducibility. These broad scatter bands, together with the unknown curvatures of the empirical parametric plots, limit extrapolation to only about three times the longest reliable test data available.

Using these widely-adopted parametric procedures, large numbers of tests varying in duration from a few hundred to many thousands of hours must then be completed to provide the design data for any new material. Yet, while a variety of high-performance materials are now being introduced to meet the increases in operational temperatures needed for improved plant efficiencies, over recent decades, many large-scale creep testing laboratories have been closed within the UK and elsewhere. As a matter of some urgency, attention should therefore be focussed on the development and validation of innovative methods for rapid low-cost prediction of design properties. Furthermore, in order to take advantage of the powerful computer-based design codes currently available, creep as well as creep rupture behaviour must be quantified. In seeking to achieve these goals, confidence in the accuracy of the long-term predictions would then be enhanced by the extrapolation procedures being based on a sound theoretical understanding of the processes controlling strain accumulation and damage evolution during creep. To this end, the present report considers the ? methodology, which allows extended extrapolation through analyses of short-term data sets describing the variations in creep curve shape with changing stress and temperature[1-3].