Evolution of a modern mechanical testing and design standard for high temperature materials

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The basis for high-temperature testing and analysis was established a century ago using test equipment that was available at the time. This fostered a tradition in research and engineering applications that has not changed fundamentally. The limitations of this approach are described, with emphasis on problems associated with predicting the remaining life of operating components. A new testing method and analysis philosophy is proposed based on measurement of current performance, with end of life unambiguously identified with limiting acceptable values of these current performance metrics. Instead of a single test to measure creep strength and fracture resistance, the new approach uses separate tests. For creep strength, the stress vs. creep rate curve determined at nearly constant state and covering five decades in creep rate in a one-day test is obtained from a high-precision stress relaxation test (SRT). The results for many alloys and materials often show a sigmoidal curve. Conventional long-time creep tests of steels have shown a similar effect. However, rather than recognising this as a fundamental rate-dependent phenomenon, it has been typically explained in terms of microstructural coarsening. For fracture resistance, a constant displacement rate test across a notch is used at the temperature of greatest sensitivity to high-temperature fracture. This latter test has been very successful in detecting embrittlement in superalloys. For more ductile alloys, a non-destructive measure of ductility based on the strain rate sensitivity determined from the SRT data is shown to have great potential. This introduces the concept of intrinsic ductility that can be determined without ambiguity over a wide range of stress and temperature for both new and serviced components.

Keywords: High-temperature alloys, Creep rate, Stress relaxation, Life prediction, Strain rate sensitivity, Embrittlement, Ductility, Design for creep

A brief history

Creep testing and analysis

Energy conversion systems based on steam turbines, gas turbines, jet engines and high-performance internal combustion engines provide the foundation for modern society. All of these machines have in common the use of metallic materials at temperatures where time-dependent or rate-dependent deformation and fracture processes must be considered in their design. The single-valued time-invariant strain associated with elastic or plastic design analysis in low temperature applications is not applicable, nor is there in most situations a unique value of fracture toughness that may be used as a limiting condition for part failure. In addition to the phenomenological complexities of time-dependent behaviour, there is now convincing evidence that the synergism associated with gaseous environmental interactions may have a major effect, in particular on high-temperature fracture.

The phenomenon of time-dependent deformation was referred to as slow stretch by Philips[1] and as viscous flow by Andrade[2] at the beginning of the twentieth century, and subsequently became known as creep. There were several seminal ideas in the Andrade work that have had a lasting impact on scientific studies and engineering dogma. The initial work was primarily on lead wires at room temperature (a high temperature relative to the melting point for lead). Andrade noted that after applying a fixed load, the rate of extension initially decreased then became constant for a time, but finally increased and continued increasing until failure. He recognised that as the wire stretched the load per unit area increased so he devised a scheme to compensate for this and maintain a constant stress on the wire. Andrade also recognised that the length of wire being experimented on at any time was increasing and thus used the concept of true strain. He derived a formula to describe the observed deformation:

$$l = l_0(1 + \beta t^{1/3})e^{kt} \quad (1)$$

where l and l_0 are the current and initial specimen lengths, t is the time, and ß and k are constants.

The initial transient strain (later to be called primary creep) was referred to as beta creep and followed a time to the one-third law, the viscous region (later to be called steady-state creep) was proportional to time and the accelerating strain region leading to fracture, which was not specifically treated

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by Andrade, later became known as tertiary creep. Much later, in a comprehensive study of creep in copper and aluminium, Wyatt[3] concluded that there are two types of transient creep in metals: at higher temperatures beta creep predominates as in Andrade's experiments, whereas at lower temperatures the strain is proportional to log (time) and the flow is referred to as alpha creep. It should also be noted that in some conditions the creep rate might increase continuously from zero time after load application. For example, high strength superalloys at very high temperatures may show only tertiary creep.

From this early work, subsequent studies diverged into two streams of investigation. The first sought understanding of creep deformation micromechanisms in pure metals and solid solution alloys in relatively short-term tests, accepted the concept of steady-state creep (although testing was more often conducted at constant load rather than constant stress), and often assumed implicitly that viscous flow was deformation path independent. This means that not only is there a steady creep rate associated with a given applied stress, but that this rate is obtained despite previous deformation at different stresses and temperatures. Although this might be a reasonable approximation for pure metals, it is manifestly wrong for most engineering alloys.[4]

The second stream of investigation concentrated on generating long-time creep data on engineering materials. By the late 1940s and into the 50s and 60s creep testing laboratories were being set-up in most industrialised nations. Often dozens and sometimes 100s of machines were used to generate creep and rupture data for design, alloy chemistry and processing optimisation, acceptance testing, and component life prediction. The testing was invariably at constant load, and data extracted included times for specific creep strains, minimum creep rates (although the term steady state was often used despite the fact that constant rates cannot be expected when the stress is changing) and time to failure (now referred to as rupture life). This latter measurement was of special significance because it became a basis for design against part failure and, later, as a basis for estimating remaining life of operating components. There thus emerged a framework for design against both creep deformation and fracture using a single mechanical test developed at the turn of the last century.

The issues of interest from a design basis are the role of primary creep, the validity of the concept of viscous steady-state creep, the dependence on temperature and the dependence on stress. These issues have been reviewed recently from both a mechanistic and phenomenological perspective in an ASM Handbook.[5] The treatment in the present paper is principally phenomenological. However, it is important to understand that the microstructure of metallic alloys evolves continuously during high temperature deformation. Changes in dislocation density and changes in hardening precipitate (e.g. type, size and distribution) as well as redistribution of alloying elements among phases and between grains and grain boundaries all occur during creep testing. In addition, damaging processes such as environmental attack, and grain boundary cavitation and cracking may develop progressively.

All these microstructural evolutionary features are dependent on temperature, stress and amount of deformation. The long-term creep test is thus unique among common mechanical property tests in that the properties being measured are changing as they are being measured. It has been argued that creep tests should be run for as long as possible to simulate operating conditions because high-temperature equipment is designed to run for tens of thousands of hours.[6] However, unless the operating conditions are similar, this argument is specious. There is no a priori reason why components destined for long-term service must be designed based on long-time constant load tests. For example, jet engines and industrial gas turbines used in cycling operation are subjected to multiple short-time creep with additional complexities of cyclic deformation, nonsteady stress and temperature, complex stresses, and aggressive environmental attack.

Nevertheless, much effort has been made in the last 60 years to develop means to extrapolate creep and rupture data to design lives for critical components. The earliest attempt to use a time/temperature parametric correlation to display and compare data for different alloys, and provide a basis for extrapolation, resulted in the widely used Larson–Miller (LM) parameter.[7] This has the form $P = T(C + \log t)$, where T is the temperature in degrees Kelvin, t is the time in hours (either to a specific creep strain or rupture) and C is a constant often taken to be 20. However, this and the many other suggested parameters are notoriously inaccurate. For example, for a data-set for a monocrystalline superalloy, CMSX-4, the optimised value of the constant in the parameter was found to vary from 13.7 to 41.2 depending on the test conditions.[8]

Although it is possible to develop an accurate empirical parametric representation for a specific data-set,[9,10] such approaches lack generality and their value is therefore limited. Extrapolation procedures may, therefore, be challenged because of their poor precision, as well as the fact that long-term predictions derived from them may not be applicable for most applications as discussed above.

Two additional important observations relative to the accuracy of data extrapolation for alloy steels have been described in Japanese work:[11] many alloys show a sigmoidal-shaped stress vs. rupture curve. Even at higher test temperatures, this phenomenon is only revealed for test times on the order of years. It was also shown that many different steels of different initial creep strengths tend to converge to similar creep strength levels at very long test times when plotted as stress vs. log time to rupture. The authors use the term inherent creep strength to describe this situation, reflecting a fully coarsened microstructure after very long-time exposures. These observations preclude accurate extrapolation of short-time data.

Remaining life prediction

The primary design of most high-temperature components involves the application of some proportion of the stress for failure in a given time (usually in the range of 20,000–100,000 h) calculated from extrapolated constant load stress–rupture tests. It might appear, therefore, that the most direct method of measuring remaining life after extended service would be to compute it based on the current remaining life of a sample taken from the component. Accordingly, methods are being developed to take miniature stress rupture samples from components and assess the remaining life based on extrapolation from such data.[12] There are a number of reasons that this approach is unsound:

- Component failure is often localised with little or no material degradation or damage remote from the failure.
- Cracks frequently initiate from the surface so that any post-exposure property measurement on material taken from the interior has limited value.
- Interactions with the operating gaseous environments, which may have profound effects on crack initiation and propagation, are generally ignored.

- The changing stress in a constant-load test is not normally accounted for in summing life fractions to predict remaining life.
- The sources of scatter in experimental property measurement (e.g. specimen alignment, temperature control, heating method, specimen geometry, and precision of stress and strain measurement), bear little connection to the sources of scatter in service.
- Time to rupture in an unnotched ductile alloy is principally a measure of deformation behaviour rather than fracture resistance.

The last item is implicit in the ASME boiler and pressure vessel code in which the design criterion is expressed in terms of both a creep rate and a rupture time. (The code specifies the lowest value of stress among: 100% of the average stress for a creep rate of 0. 01% in 1000 h, 67% of the average stress for rupture in 100,000 h and 80% of the minimum stress for rupture in 100,000 h). It is also consistent with the Monkman–Grant[13] relationship between minimum creep rate and rupture time. Such a design criterion is used to set stresses conservatively, and not explicitly to predict part life. We thus try to do for a used part what is not normally attempted for a new one.

The failure to establish a clear separation of a strength requirement from a failure criterion (an appropriate analogy might be between yield strength and fracture toughness for many low-temperature components) leads to a paradox. This may be stated as: when a component fails, the material of the component has a finite life, sometimes approaching the original design life, at the operating conditions of the component. It thus follows that a remaining life estimated from a sample taken from an operating component may bear no relation to the actual component remaining life.

Design for performance

Low-temperature design is dominated by properties that uniquely characterise the mechanical state in terms of stiffness (modulus), strength (yield) and fracture resistance (Kc). If service-induced changes in state occur (e.g. radiation hardening and embrittlement), these changes are monitored in terms of their effects on changes in the same short-time properties. Safe life is determined based on the current performance rather than the predicted failure. Thus, life management decisions are based on the same performance criteria as in the original design: there is no remaining life paradox in this case.

The basis for current methods of high-temperature design is different in that the objective is to incorporate time-dependent changes into the test methodology. The creep rates and rupture lives relate to the starting material only in terms of the specific deformation path being imposed, i.e. a variable strain rate, variable true stress test with an arbitrary interaction between creep deformation and fracture processes. This history is quite different from any real service history. To measure these properties, unlike in the low-temperature situation, we must change the structural and mechanical state. Even with the most complex test plan, we cannot emulate most service operations.

An alternative approach is to simplify the test methodology and develop tests to measure separately the high-temperature creep strength and fracture resistance, ideally to evaluate the current state in terms of these properties. The consequences of microstructural evolution and damage, induced in service or in laboratory simulations, can then be assessed using the same short-time tests. Design is then based on minimum acceptable performance levels. The question is whether such testing is possible at high temperatures.

In the 1970s, Hart was developing a plastic equation of state for aluminium at room temperature,[14] and subsequently supported the theory using a carefully controlled high-precision stress relaxation test or SRT.[15] This test could be analysed to produce a stress vs. strain rate curve covering at least five decades in strain rate in a few hours with minimal inelastic strain accumulation and little or no change in the plastic state (or hardness in Hart's terminology). This constancy of state was established by identical responses in repeat relaxation runs on the same specimen. If this test could be run at high temperature, it might provide the basis for a current creep strength analysis. Ideally, the test should not change the state of the material significantly in the process of measuring the critical property.

For fracture resistance, the challenge is a little more difficult. We are seeking a test that is inexpensive, does not change the mechanical state significantly and can be conducted in a short time. A test which has shown considerable promise is a notched bar tensile test deformed under closed-loop displacement control (constant displacement rate test or CDR) across the notch. This test was originally developed to accelerate the onset of notch sensitivity in steels which might normally show notch sensitivity after stress–rupture testing for tens of thousands of hours.[16] For a notch strengthening material, the time to rupture is longer for a notched bar than a smooth bar at the same nominal stress. However, for some Cr–Mo–V rotor steels, it was found that after long times (the cross over time or COT), the notch specimen had a shorter life. A good correlation was found between this COT at 540C (1000F) and the extension at failure in the notched bar CDR tests at 650C (1200F). Clearly, this then became a valuable test to predict long time or low creep rate sensitivity to notches. It thus became a useful test to screen incoming heats and also predict useful safe operating lives. In the CDR test, a crack initiates at the notch root and propagates under control, which is independent of the machine compliance. The fracture resistance is related to the displacement at failure and the extent of unloading prior to fast fracture. If desired, the crack propagation rate can be computed in terms of fracture mechanics parameters. The CDR test does, in fact, change the state because a significant amount of plastic deformation may occur prior to crack initiation . However, it fulfils the other requirements, and has been shown to be repeatable, discriminating and appropriate for materials of limited ductility.

Although there are several available techniques for accelerated testing in traditional creep analysis,[17] and the SRT test has also been used with good success to predict long-time creep and rupture strength,[18] the emphasis should be on establishing a new philosophy of testing and design for high-temperature materials. Whereas the traditional approach to creep testing attempts to incorporate microstructural evolution and damage development during the test, the new approach, Design for Performance, measures the consequence of such changes. The objective in the latter case is to measure the critical properties with minimal change in the properties during measurement. Thus, the traditional approach might ask the question: 'what is the creep rupture strength for tests lasting 50,000 h?' The new approach would ask: 'what are the values of the creep strength and fracture resistance for new material and what are their values after 50,000 h service?' Design for Performance must ultimately be judged independently on its merits as a materials characterisation framework for direct use in materials selection, design and life prediction.

The new testing methods

In the original room temperature stress relaxation testing of Hart and Solomon,[15] the machine crosshead was fixed so that plastic strain occurred as a result of elastic strain replacement in both the specimen and machine. This required close control of the ambient temperature. It was found in testing at high temperature that it was necessary to use an extensometer to measure directly the component of strain replacing machine elasticity.[19] This reduced the importance of controlling the ambient temperature. In these tests, it was established that, even after multiple relaxation runs on a single specimen, a repeat test showed a variation in creep rate of less than a factor of two. Hence, we may conclude that the change in state during a SRT test is minimal. Most recently, in the results reported here, a closed-loop control on the specimen strain was used as well as high-speed digital data acquisition. The maximum creep strain that could be accumulated during the relaxation test was thus limited to the elastic strain on the specimen after loading. Machine compliance was not a factor.

Stress relaxation testing may be performed in a servohydraulic or electromechanical test system under closed-loop control. The specimen is equilibrated using standard testing procedures prior to stress relaxation testing (SRT) for creep strength evaluation or constant displacement rate testing (CDR) for fracture resistance evaluation. Several specimen configurations, including threaded or button-head cylindrical specimens and pin-loaded foil have been successfully employed. The most commonly used in the present tests were threaded cylindrical samples. They had a gauge length of 25.4 mm and diameter of 4.06 mm. Miniature specimens with a gauge of 25.4 × 1.9 mm have also been used, principally for evaluating the creep strength in thin sections of gas turbine blades, and are identified on the figures. The specimens for CDR tests, used with the same standard specimen dimensions, have a 60° notch with a root radius of 0.127 mm and a minimum diameter of 2.87 mm to produce an estimated stress concentration of 3.15.

Most laboratories can conduct these tests routinely at temperatures up to 1000C or 1100C. For testing at temperatures lower than 250C, for example, using polymers, an oven surrounding the specimen may be used. At temperatures higher than 1100C, for example, using ceramics, special water cooled grips and capacitance extensometers may be required.

The recommended practice for SRT and CDR testing is given below.

Stress relaxation test

For all SRT runs, the loading rate is 0.01% sec^{-1} under extensometer control to the set strain, hold at the set strain for 20 h, then unload and hold for 2 h at near zero stress. The stress vs. time is recorded as well as the strain and temperature. Sampling of data is normally at 1 s intervals. For actual plotting, the data are culled to provide a convenient plot on a logarithmic timescale. A typical sampling is: 1-s intervals for the first 300 s, 1-min intervals for the next hour and 10-min intervals for the remainder of the relaxation until unloading after 20 h (72,000 s). Zero time is set for relaxation at the instant the loading is stopped. During machine unloading, data may be sampled at 2-s intervals and then at 60-s intervals at close to zero stress up to the total test time of 22 h. This allows a modulus to be calculated, and any anelastic recovery measured. The stress may be recorded in MPa or psi, and temperature in Celsius or Fahrenheit.

The most frequent total strain levels used are 0.4, 0.8 and 1.3%. The former is designed just to exceed the elastic limit so that the relaxation response provides a measure of the current creep strength of the alloy. The latter is designed to include about 0.5 or 1% inelastic deformation and has been used to compute estimated times to specific creep strains.

Constant displacement rate test

The CDR tests are run to failure under extensometer displacement rate control at a rate of 0.25 mm h^{-1}. Controlling on the extensometer is important so that the displacement rate across the notch is constant. In this test, the end of the test is most important so that the unloading as the crack propagates may be followed. The tests are expected to have a maximum of about 5-h test duration. Data should be sampled sufficiently rapidly initially to get the loading curve, then occasionally, and finally, during crack propagation, most frequently.

Calculation and presentation of results

The primary product of this high-precision short-time stress relaxation test is a plot of stress vs. creep rate. This normally covers at least five decades in creep rate in tests lasting less than one day. The components of strain can be considered to be elastic strain, ε_e, time-independent plastic strain, ε_p, (e.g. on loading) and time-dependent creep strain, ε_c. The creep strain can further be designated as nonrecoverable or recoverable (anelastic strain). In metals, this recoverable strain is usually only a small fraction of the total creep strain (less than 10%) and may be ignored. The total strain after loading is given as:

$$\varepsilon_t = \varepsilon_e + \varepsilon_p + \varepsilon_c = \text{constant} \tag{2}$$

Differentiating and remembering that the plastic strain on loading, by definition, is not time-dependent, and that the total strain is constant:

$$0 = \dot{\varepsilon}_e + 0 + \dot{\varepsilon}_c \tag{3}$$

Rearranging Eq.3:

$$\dot{\varepsilon}_c = -\dot{\varepsilon}_e = -\dot{o}/E \tag{4}$$

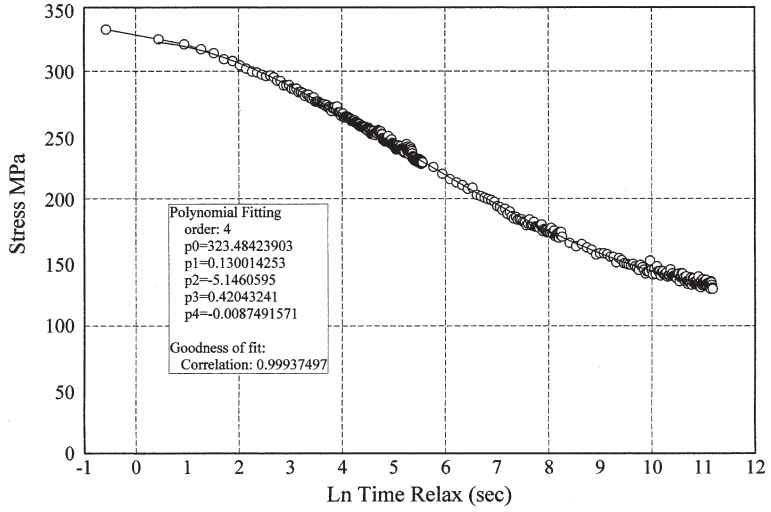
where ȍ is the stress rate and E is the elastic modulus.

Equation (4) shows how the creep rate is equal in magnitude to the elastic strain rate and can be calculated at any time during relaxation from the stress rate divided by the elastic modulus. The elastic modulus is taken as the average of measurements made on loading and unloading the specimen. Thus, the test is a self-programmed variable stress creep test and should be independent of the test machine characteristics.

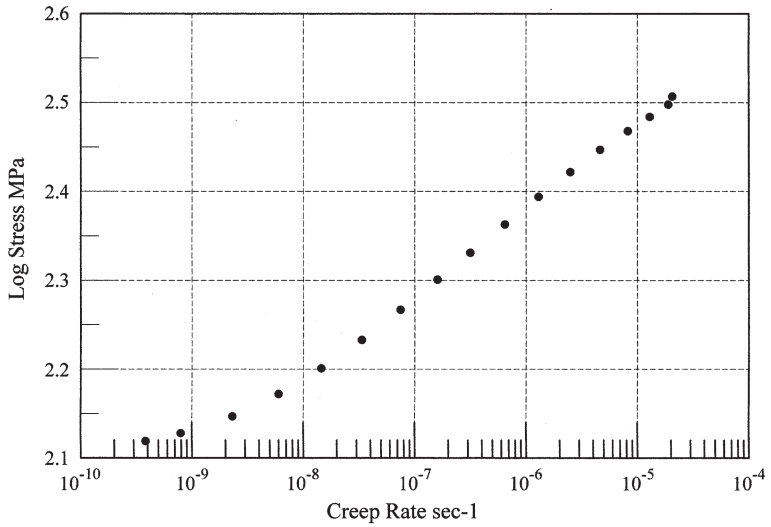
To calculate the stress rate, and hence the creep rate, it is necessary to differentiate the stress vs. time curve obtained from the experimental relaxation test as described above. A straightforward and accurate approach is to use a polynomial curve fit of a plot of stress vs. Ln time. It has been found that a third- or fourth-order polynomial generally gives a best fit to the experimental data. This natural logarithm of time function can be readily differentiated. The computed data are normally plotted as stress vs. creep rate on a log. log. plot.

Results and analysis

Figure 1 shows results of a typical SRT test on a 9% Cr T91 steel at 550C from 0.4% strain.[20] The fourth-order polynomial fit is differentiated using custom software integrated with the



1 Results of a typical SRT test on a 9% Cr T91 steel at 550C from 0.4% strain



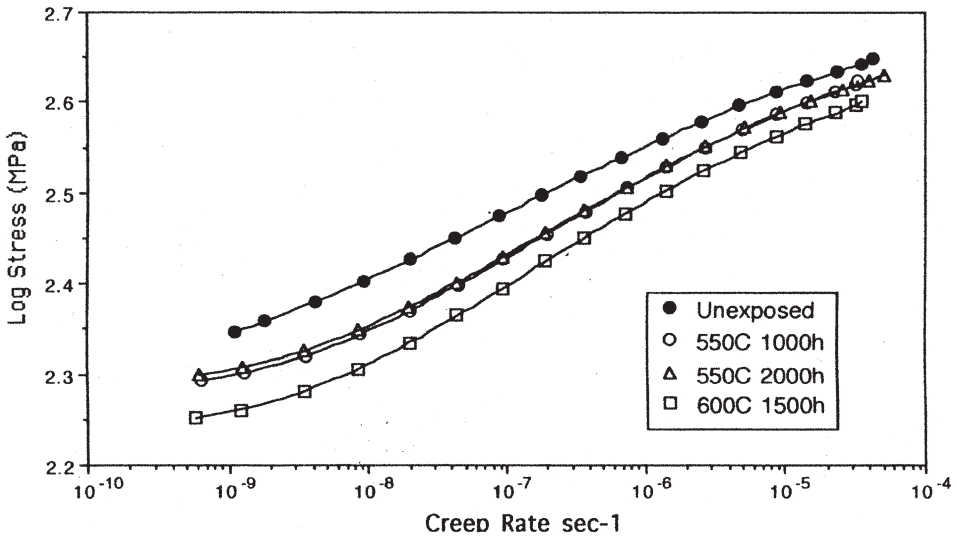
2 Logarithmic stress vs. creep rate curve for data from Fig. 1 for T91 at 550C

commercial graphics program PSI-Plot. The resulting graph is shown in Fig. 2 as log stress vs. log creep rate. This curve covers more than five decades in creep rate, and has a sigmoidal shape with the inflexion occurring at about 1×10^{-8} s^{-1}.

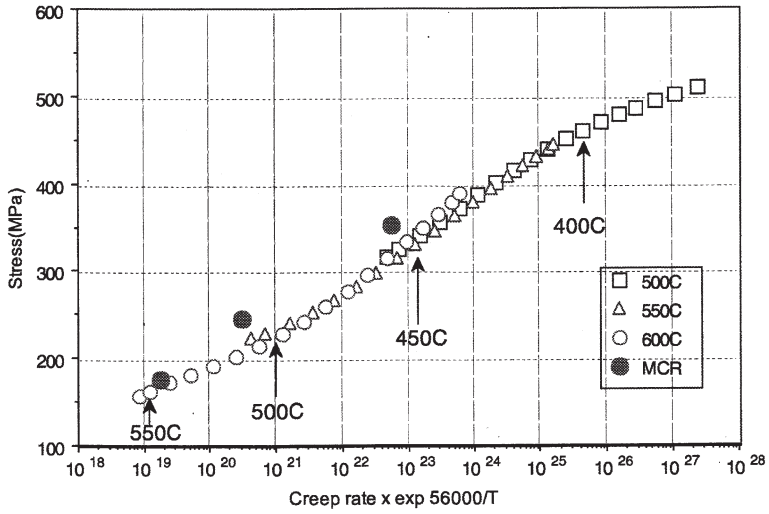
Creep rate curves for Cr–Mo–V steel at 550C and 0.4% strain for four starting conditions are compared in Fig. 3. The curves have the same sigmoidal shape, but are shifted to lower stress levels with increased thermal exposures.[21] This implies that the shape is stress or strain rate controlled and not significantly influenced by any precipitate coarsening. The inflexion point again occurs at about 1×10^{-8} s^{-1}. It is interesting to note that, assuming 20% strain in rupture tests at 550C, this creep rate would give an approximate life of 6000 h. This suggests we are dealing with exactly the same phenomenon as in long-term creep tests, and that this is strictly rate dependent rather than a response to a microstructural coarsening resulting from long-term exposure.[11,22] Sigmoidal curves have been observed in SRT tests in many alloys and even in SRT testing of polymers[23] and ceramics.[24]

As with several other studies, including Incoloy 800H and an aluminium alloy,[25] it was found that the creep rate data for the steel at three temperatures could be superimposed with good precision by horizontal translation of the individual creep rate plots. This means that the curves may be parameterised by an exponential temperature function. This was optimised by taking a horizontal section at log stress = 2.5 on the data for 0.4% strain, and plotting log creep rate vs. reciprocal temperature. This indicated an activation energy of 465 kJ mol^{-1} and allowed the parameter plot of Fig. 4 to be constructed. This master curve may be used, based on three one day tests, to compare processes, heat treatments, service exposures, etc. It may also be used as a basis for creep design. As an example, possible design points at different temperatures are identified on the figure at creep rates of 3×10^{-11} s^{-1}. This is a rate corresponding to 1% in 100,000 h. This format is also convenient for finite-element analysis. Also included for comparison in this figure are minimum creep rate data.

Multiple tests on the nickel based alloy Waspaloy from 0.4% strain are shown in Fig. 5. The right column depicts duplicate tests at three temperatures on Standard Waspaloy (SW), and the left column indicates three results for powder metallurgy (PM Waspaloy). Apart from a slightly higher creep strength at 600C for the PM, these data indicate comparable



3 Creep rate curves for Cr–Mo–V steel at 550C for various prior exposures at the indicated conditions showing similar sigmoidal behaviour for all exposure conditions



4 Stress vs. creep-rate/temperature parameter plot for Cr–Mo–V steel using just three one-day tests. Possible design points at four temperatures are indicated. Minimum creep rate data are included for comparison

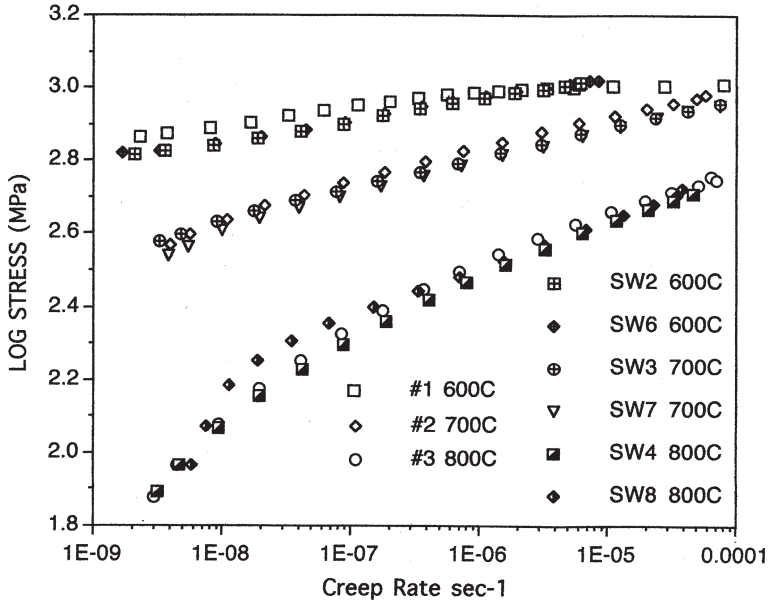
strengths over five decades in creep rates and a temperature range of 200C in less than two weeks of test time.

Extensive studies of the cast nickel based alloy IN738 have been made. Figure 6 compares standard and miniature specimen data from both 0.4 and 0.8% strain at 850C. There is no significant effect of test section size on creep strength. However, there is a large effect on creep strength of sampling location in a gas turbine blade. For example, Fig. 7 shows that the creep strength at 850C in a blade shank is comparable to replicate tests from a cast slab, but the thin leading edge has a much lower creep strength. This is presumably due to the finer grain size in the thin section, which is actually of concern since this is the region subjected to the highest temperature.

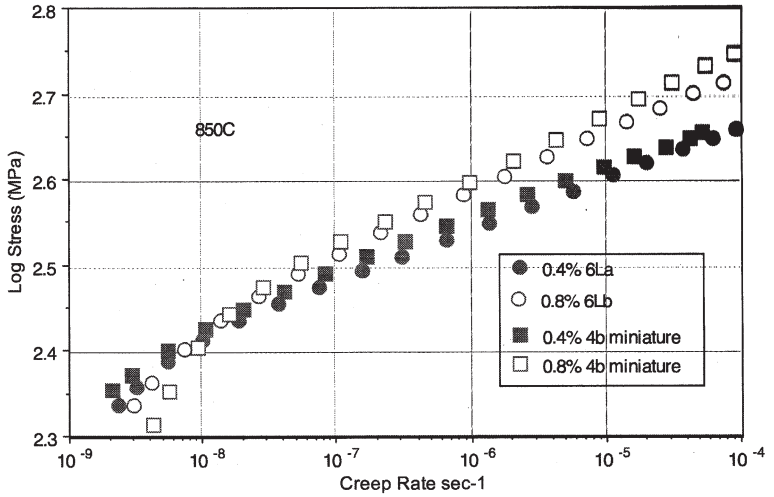
Most tested superalloys could be readily parameterised. Figure 8, for example, shows the results for IN738, including duplicate tests at 900C. Note, in this case the data are plotted on a linear stress scale so that possible design stresses may be readily shown. The indicated stresses at 800 and 850C correspond to the equivalent of 1% strain in 100,000 h.

A number of alloys have been tested before and after various exposures in CDR tests. In general, ferritic steels showed a progressive softening due to microstructural coarsening. Although this would result in reduced rupture life in conventional tests, there is no decrease in fracture resistance. The test was also very useful for studying weldment properties by locating the notch in the base metal, heat affected zone or weld metal. For example, Fig. 9 indicates in tests at 510C, only the weld metal showed fast fracture compared with controlled unloading to zero stress for the other conditions.

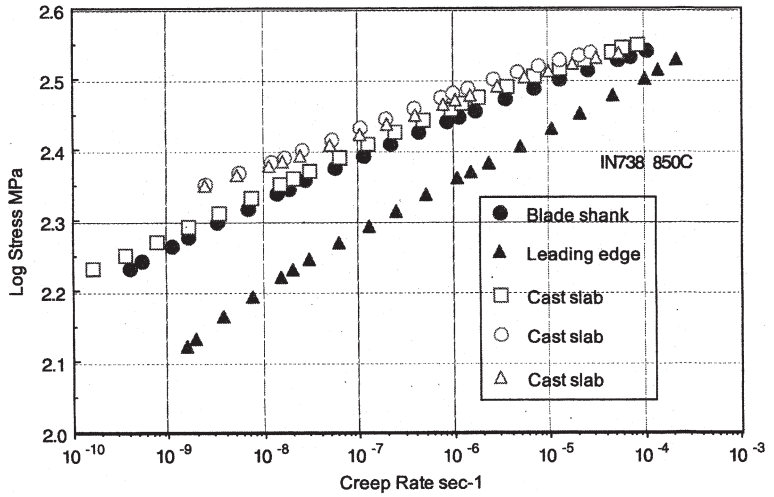
For superalloys, the situation was quite different in that exposure at high temperatures in air led to severe intergranular embrittlement due to oxygen penetration. It was found that the embrittlement could be readily detected using the standard miniature SRT specimens tested under CDR conditions, i.e. constant displacement rate across the gauge section. Embrittlement was progressive, with decreasing displacement at failure and fast fracture after partial unloading, with increasing exposure. The role of grain boundaries is demonstrated in Fig. 10 showing the effect of exposure in air at



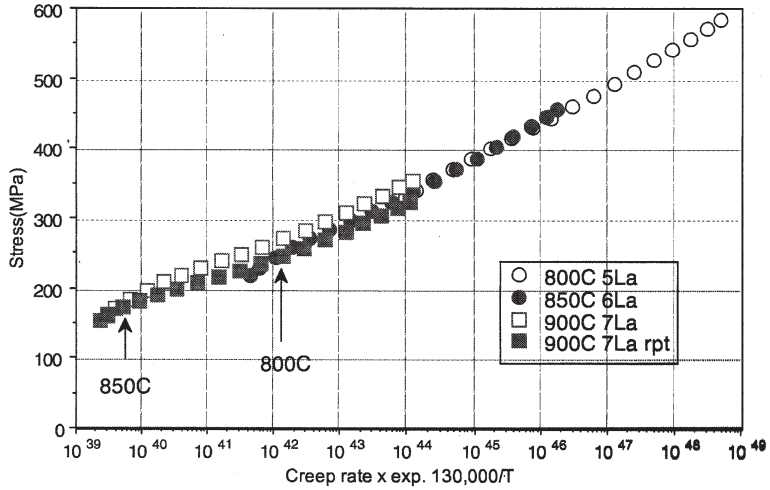
5 Multiple tests on both standard and powder metallurgy Waspaloy from 0.4% strain at three temperatures



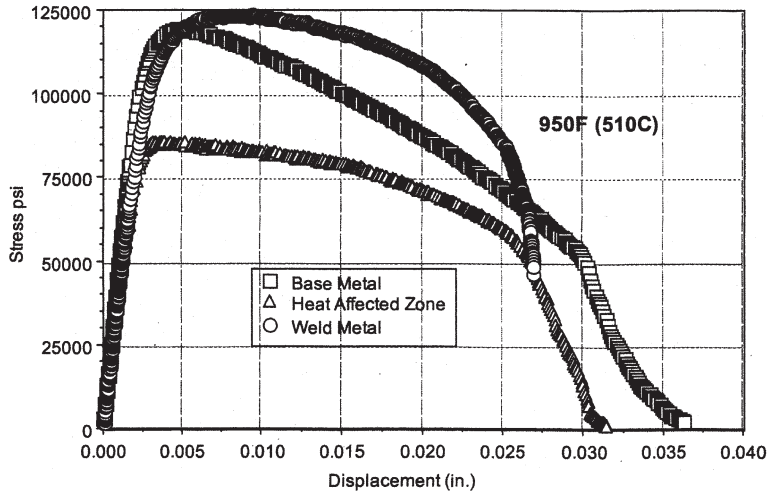
6 Stress vs. creep rate from two strain levels for IN738 showing no significant effect of test section size



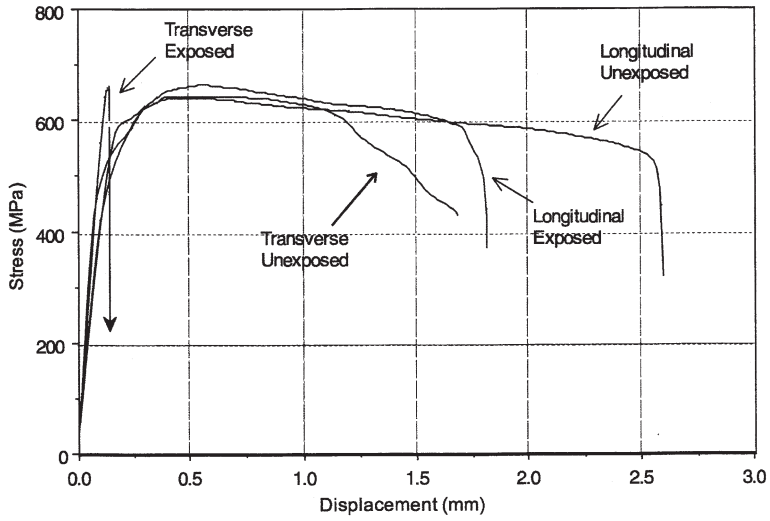
7 Stress vs. creep rate for different microstructures in IN738



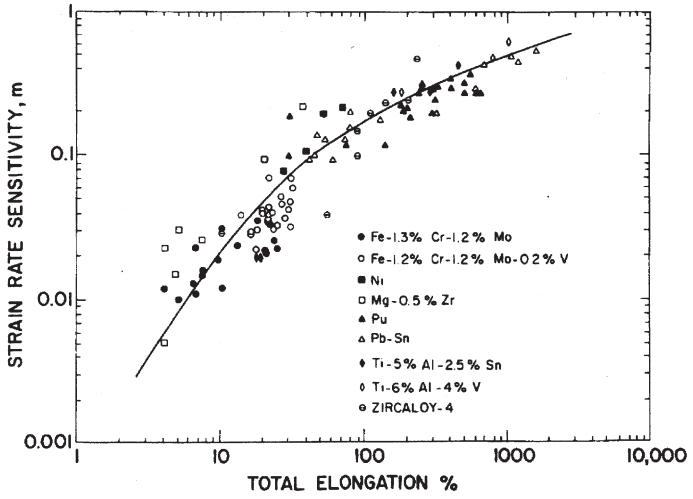
8 Stress vs. parameter plot for IN738, including duplicate tests at 900C, showing possible design points at 800C and 850C



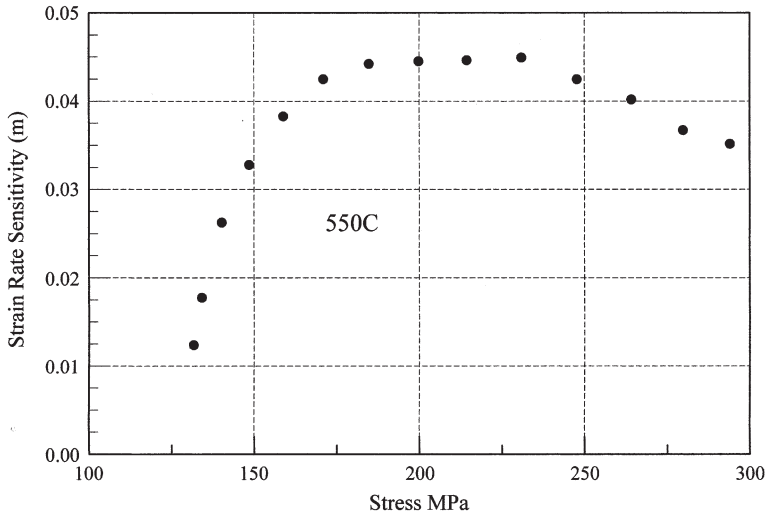
9 CDR tests on notched specimens from different weldment locations in12% Cr steel at 510C (1 mm = 0.04 in., 690 MPa = 100,000 psi)



10 Effect of air exposure at 1000C for 24 h and grain boundaries on unnotched miniature specimens of directionally solidified IN738 tested in CDR mode at 800C. Data points omitted for clarity



11 Correlation between strain rate sensitivity and total elongation for a variety of alloys[26]



12 Strain rate sensitivity as a function of stress for alloy T91 at 550C showing how the intrinsic ductility varies with stress

1000C for 24 h on the CDR response at 800C on specimens taken either parallel or perpendicular to the grain orientation in directionally solidified GTD111. The transverse specimen is fully embrittled across the 1.9-mm cross section, whereas the longitudinal specimen shows only a modest reduction in displacement at failure. This latter is probably due to local transverse grain boundary segments in the longitudinal orientation. The kinetics, phenomenology, mechanisms and broad implications for high-temperature fracture of gaseous embrittlement in superalloys have been summarised elsewhere.[26]

Common methods of measuring ductility such as uniform elongation, elongation at fracture, or true strain at fracture based on the reduction in area, may involve extensive strain in ductile alloys. This means that the property is changing as it is being measured. It is also complex in the sense that stable uniform deformation may be followed by unstable neck development and/or crack propagation. For these reasons, ductility does not appear directly in design procedures.

Ductility (defined either as elongation to fracture or reduction in cross-sectional area at fracture) is generally closely linked with strain hardening at temperatures below about 0.3Tm, where Tm is the melting temperature, and with strain rate hardening at temperatures above about 0.4Tm. In fact, a good correlation has been found at high temperature between strain rate sensitivity (m = dlog stress/dlog creep-rate) and elongation at fracture[27] for elongations between about 4 and 2000% (Fig. 11). The values of m were taken at close to a constant mechanical state, i.e. using stress increments or decrements and not pseudo-steady states, i.e. minimum creep rate data. For ductile materials at high temperature, the strain rate dependence provides a measure of the growth rate of a plastic instability or neck.[14] A material with a low rate sensitivity will be less stable in the sense that a local stress inhomogeneity will cause a high local increase in strain rate and local plastic instability. Conversely, a material with a high rate sensitivity will resist the growth of a local instability. The higher the rate sensitivity the greater is the total elongation so that as the rate sensitivity approaches unity deformation becomes superplastic.[28] A comprehensive assessment of all theories of tensile instability concluded that an assumption of negligible strain hardening was fully consistent with the data correlation shown in Fig. 11.[29]

Although indications are that the strain rate sensitivity may provide a useful current measure of the fracture resistance for ductile materials, no comprehensive study has yet been made. However, Fig. 12 shows how the differential

of Fig. 2 indicates a low value of m of 0.01 at a stress of 130 MPa. From Fig. 11, this corresponds to an elongation at fracture of about 7% compared with a maximum of about 15% (m = .044). This concept of a current value of ductility, in the same sense as a current value of creep strength or fracture resistance, says that for any mechanical state, at a given stress and temperature, there is an instantaneous value of ductility given be the appropriate differential of the log stress vs. log creep rate curve taken from the SRT test analysis. Unlike destructive methods of measuring ductility, the value of m taken over five orders of magnitude in strain rate for each constant high temperature is independent of specimen geometry and test machine. It enables measurements in ranges of stress and creep rates relevant to equipment operating conditions. It also eliminates the large sources of scatter in traditional destructive measurements of ductility. It remains to be determined whether this non-destructive measurement can be used as an alternative to the CDR test in assessing embrittlement in a variety of materials.

Concluding comments

The traditional approach to high-temperature mechanical testing and analysis, and component design, has evolved around one-hundred-year-old technology and very basic test machines. Millions of creep tests have been conducted in laboratories around the world using vast quantities of electricity over long periods. The output has led often to sterile research and inefficient design. The 'Remaining Life Paradox' precludes reliable life prediction, and has resulted in very limited advances in component life prediction. Nevertheless, there are enormous vested interests in this technology, both in universities and industry.

Design for Performance has been promoted for about 30 years. The work has been supported by many original equipment manufacturers and gas turbine users. Nevertheless, it has not yet achieved widespread application. It requires the use of one or more universal testing machines with appropriate extensometry and skilful testing. Analysis should be within the capabilities of most graduate engineers.

In previous publications, the author has shown how the stress vs. creep rate response may be used to compare with traditional long-term creep–rupture data. In this paper, Design for Performance is presented as a stand-alone framework for alloy selection, optimisation, design and life prediction. The concept involves setting limiting current values of creep strength and fracture resistance. The stress vs. creep rate plot generated from the SRT test provides design stress levels for specific temperatures and creep rates in just a few one-day tests. End of life in terms of creep strength is then determined when a miniature sample taken from the component no longer has the required creep strength. Figure 3 shows how progressive degradation in creep strength is expected to occur with an increased exposure for a low alloy steel.

The fracture resistance at the most vulnerable temperatures in the current structural state may be determined from the displacement at failure and degree of unloading prior to fast fracture taken from the CDR test. Minimum acceptable values can then be set by the designer for end of life.

Alternatively, in a promising approach that has seen limited attention, the strain rate sensitivity (m), which provides a uniquely current measure of ductility, may be determined directly from the stress vs. creep rate analysis. Thus, ductility at high temperatures is a definable non-destructive property of an alloy for a given microstructure and depends only on stress and temperature. Minimum acceptable levels of (m) at critical machine locations operating at specific stresses and temperatures may also be set to define end of safe life unambiguously.

Finally, it is worth reiterating that the sigmoidal plot of log stress vs. log rupture time is most likely an intrinsic creep rate effect leading to lower creep rates than expected for extrapolation of short-time data. It is not, as is widely believed, a consequence of microstructural coarsening during the test. This same effect is observed in SRT tests with or without coarsening in low alloy steels and in many other alloys. It also occurs in materials with covalent and ionic bonding.

Acknowledgements

Many people and companies have supported this work. However, I wish particularly to thank David Stiles, then of ARCO Alaska, who provided the initial support when this was just an idea, Philip Dowson of Elliott Corporation who provided extended financial support and eventually transferred the technology to his Company, and Tom Gibbons for his continued interest and encouragement. In addition, colleagues at GE provided stimulation and guidance in the development years. Edward Hart prompted my interest and commitment to the SRT test and to rigorous thinking in general. Robert Goldhoff and Louis Coffin were prime mentors in high temperature deformation and fracture, and Roger Bricknell worked creatively with me on environmental embrittlement. Experimental details for SRT and CDR were developed by Donald Van Steele for metals, ceramics and polymers. The original software for SRT analysis was written by Kyle Amberge for the Mac., and recently upgraded and modernised by Paul Woodford for the PC.

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