Nickel-Base Superalloys: Current Status and Potential [and Discussion]

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Nickel-base superalloys: current status and potential

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The evolution of nickel-base superalloys has occurred over about 50 years through a combination of alloy and processing developments to satisfy quite different service requirements of various components of the gas turbine. There is now a good general understanding of the mechanisms leading to the unusual mechanical properties of these precipitation strengthened materials. Although the scope for further significant improvements in the behaviour of nickel-base superalloys appears to be limited, it is unlikely that their full potential is yet being achieved in engineering applications. Progress towards the development and validation of constitutive laws describing fully anisotropic deformation is described.

1. Introduction

The development of nickel-base superalloys has, almost entirely, been motivated by the requirement to improve the efficiency, reliability and operating life of gas turbines. There have been other peripheral applications but, at present about 90% of superalloys produced are used in gas turbines for a range of applications, including aerospace, electricity generation, gas/oil pumping and marine propulsion. The differing requirements in specific parts of the engine and the different operating conditions of the various types of gas turbine have led to the development of a wide range of nickel-base superalloys with individual balances of high-temperature creep resistance, corrosion resistance, yield strength and fracture toughness. However, all of these materials have evolved from the Ni–Al–Ti–Cr precipitation strengthened alloy, Nimonic 80A, developed by Pfeill and his colleagues at the Mond Nickel Company around 1940 in response to Whittle's need for a suitable turbine blade material for the first British gas turbine for aircraft propulsion (Betteridge & Shaw 1987; Sims 1984).

The principal characteristics of nickel-base superalloys largely derive from the precipitation of an $L1_2$ ordered intermetallic phase, γ' $Ni_3(Al,Ti)$, that is coherent with the face-centred-cubic γ-nickel solid solution matrix (Stoloff 1987). The development of viable superalloys has been achieved by a combination of compositional modifications that control aspects of the γ/γ' relationship (γ' volume fraction, γ' solvus, γ/γ' lattice mismatch), the use of more conventional alloying approaches to solid solution strengthening and corrosion resistance, and the introduction of a range of novel processing techniques (directional solidification, single crystal technology, powder processing, mechanical alloying, HIPping, etc.). A full review of superalloy technology is beyond the scope of this paper which will present a personal view relating to the most important recent developments and future requirements.

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2. Historical trends

The efficiency of operation of the gas turbine is largely determined by the combination of the temperature and pressure/volume of gas passing from the combustion chamber to the external environment to provide propulsion, or power to motivate other machinery. The materials available to be used in the turbine, particularly as turbine blades and discs, have to a large extent determined the operating conditions of gas turbines. The progressive improvement in the efficiencies of gas turbines has paralleled the increased temperature capabilities and strengths of successive generations of superalloys that have been specifically developed for these critical components, in particular turbine blades and discs. Indeed, much of the drive for superalloy substitutes, which is the major theme of the present meeting, derives from the same continuing requirement. We consider briefly the evolution of materials for turbine blade and turbine disc applications.

(*a*) *Turbine blades*

Materials for high-pressure turbine blades must be able to operate in the high-temperature gases emerging from the combustion chamber; they experience a combination of high temperatures and relatively low gas-bending and centrifugal stresses. In the half century since the development of the first precipitation strengthened superalloy (Nimonic 80) the temperature capability of nickel-base superalloys, as measured by the temperature at which a creep rupture life of 1000 h can be achieved with a tensile stress of 150 MPa, has progressively increased by about 7 K per year (figure 1*a*). The increase in the gas operating temperature has been much greater than this due to engineering innovations, such as blade and thermal barrier coatings, that allow the blades to operate in an environment in which the gas temperature exceeds the melting point of the alloys from which the blade is produced.

The strategies adopted in the development of turbine blade materials have depended on the service cycles for which specific engines were designed and where different failure mechanisms determine component life. However, there have been some common threads to this alloy development:

increasing γ' volume fraction (Al, Ti);
increasing γ' solution temperature (Co);
minimization of the γ/γ' lattice parameter mismatch;
solid solution strengthening (W, Mo, Ta, Re);
ductilizing additions (Hf, B).

Alloys with high chromium contents have been developed to give the enhanced oxidation/corrosion behaviour required for marine and industrial applications, but this has usually been achieved at the expense of mechanical performance. The development of reliable corrosion resistant coatings is rendering less important the need for alloys that are inherently oxidation/corrosion resistant.

As the alloys have evolved to give increased high-temperature strength, it has been necessary to develop new processing techniques to produce the turbine blades to acceptable tolerances and to maintain the required levels of ductility. The principal stages in this development can be summarized as follows.

(i) Forged blades were produced, and still are from some alloys, while there was a sufficiently wide heat treatment window, between the γ'-solvus and liquidus temperatures to allow reliable thermo-mechanical processing.

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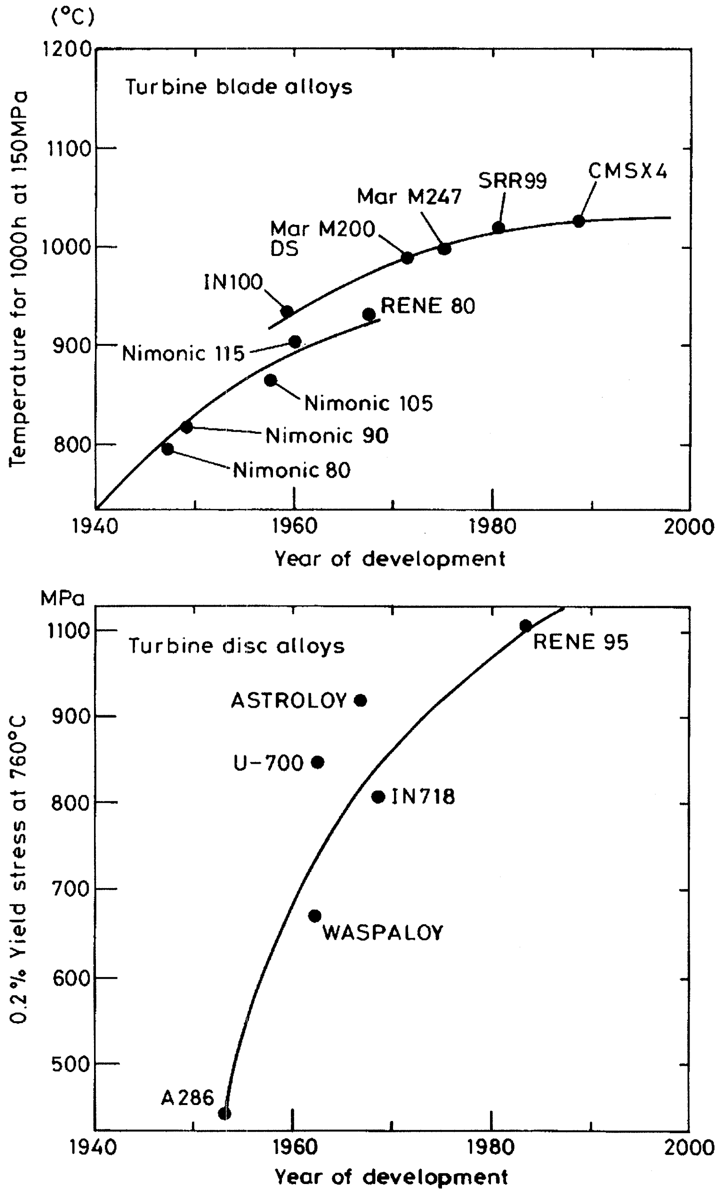

Figure 1. Trend in the improvement in superalloys for turbine blade and disc applications as a function of the year of introduction of the alloy. (*a*) Blade alloys temperature for 1000 h rupture life with 150 MPa stress. (*b*) Disc alloys. Yield stress at 850 °C.

(ii) Investment casting was introduced for alloys where forging was impracticable and where adequate ductility could be achieved. However, as the γ' volume fraction increased beyond about 50%, the ductilities became unacceptably low due to premature fracture at transverse grain boundaries.

(iii) Directional solidification, first used by Ver Snyder and co-workers by a modification of the investment casting process (see, for example, Ver Snyder *et al.* 1966), produced an elongated grain structure and a $\langle 001 \rangle$ crystal texture parallel to the solidification direction. The result has been an increase in creep ductility from less than 1% to more than 25% and considerable enhancement of the thermal fatigue resistance for alloys such as MarM 200.

(iv) Single crystal superalloys result from a fairly simple variant of the directional solidification process and is now state-of-the-art for advanced aerospace applications and is also being considered for large electricity generating plant.

It should be noted that processing changes have not always been in response to alloy development. Rather, alloy chemistry has often been adjusted to produce alloys tailored to the available processing technology. For example, the introduction of hafnium was to reduce the occurrence decohesion of longitudinal grain boundaries during directional solidification (Lund 1972). There is little or no inherent advantage of single crystal versions of the alloys produced by directional solidification; indeed, Piearcey *et al.* (1970) studied single crystal superalloys over twenty years ago. However, Gell *et al.* (1980) showed that by stripping the elements intended to modify the grain boundaries (C, B, Hf) it was possible to increase the liquidus temperature and allow more effective control of the γ' morphology through heat treatment. A wide range of superalloys specifically intended for use in the single crystal form has been and continues to be developed.

The most advanced single-crystal superalloy turbine blades are now operating at a homologous temperature $T/T_{\mathrm{m}} > 0.85$. Although further developments will certainly take place, the melting point of nickel provides a natural ceiling for the temperature capability of nickel-base superalloys. Consequently, there is limited scope for further large increments in temperature capability of this class of alloys.

(*b*) *Turbine disc alloys*

The turbine disc operates at considerably lower temperatures than do the blades (about 850 °C compared to 1150 °C for blades in current aero-engines). Consequently creep deformation is relatively insignificant. The principal design requirements, to reduce engine weight and increase rotational speed, both lead to high stresses on the disc and alloy development has been designed to increase the yield strength and to inhibit crack initiation and growth, particularly in fatigue conditions. Figure 1*b* indicates the progressive increase in yield strength and fatigue resistance of this class of alloy.

With increasing yield strength there has been an associated decrease in fracture toughness. Some amelioration of this effect has been obtained through control of grain size; attempts at producing duplex grain morphologies (*necklace grain structures*) appear to have been abandoned as being impractical (Jeal 1986). However, in current advanced disc alloys the critical defect size for brittle fracture at service stresses is about 30 μm (Sczercenie & Maurer 1987). Since various inclusions or clusters of precipitate particles can constitute such a critical defect this places demands on non-destructive examination that are currently unattainable. The consequences of disc failure, particularly in aero-engines, require a viable quality assurance procedure.

Turbine discs are currently produced either by forging or by powder processing, although the former is more extensively used. In both cases quality is maintained through careful control of alloy cleanness at each stage of the process. The use of different secondary melting processes, on bar-stock initially produced by vacuum induction melting, is now standard procedure to control inclusion content. Vacuum arc refining (VAR), electro-slag refining (ESR), electron beam cold hearth refining (EBCHR) are all currently available as commercial processes producing

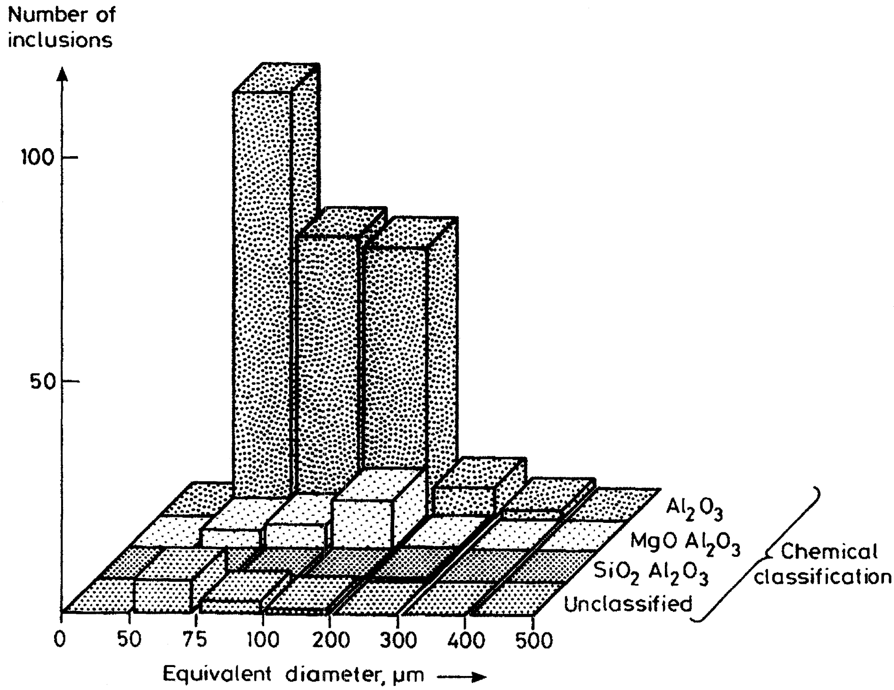

Figure 2. Size distribution of various types of inclusions found by analysis of the concentrated inclusions produced by electron beam button melting of the turbine blade alloy IN738LC (Chakravorty *et al.* 1987).

double, or triple, melted alloy (Patel & Siddal 1994). Because of the impracticability of characterizing the very low inclusion contents in these materials by conventional metallographic procedures or by mechanical tests on small specimens, novel approaches to evaluating very clean alloys are being developed. One such example is the electron beam button melting approach used by Quested & Hayes (1993) at the National Physical Laboratory which allows inclusions from 1 kg of alloy to be concentrated, identified and their size distributions determined (figure 2). A code-of-practice for use of the NPL approach to cleanness evaluation has been agreed by the leading UK producers of turbine disc materials (Quested, personal communication).

If replacement materials for superalloys are developed that allow a significant increment in turbine operating temperature, there will be a corresponding increase in the service temperature of the disc. This will require new disc alloys. There is certainly scope for the further development of superalloys for this purpose. However, alternative materials, such as structural intermetallics, could well supersede superalloys because of the attractions of weight savings through reduced density. Superalloys for discs are at a less mature stage of development than are those for blade applications.

3. Fundamentals of mechanical behaviour

There can be no doubt that the attractive mechanical properties of nickel-base superalloys derive directly from the disposition of a high volume fraction of the

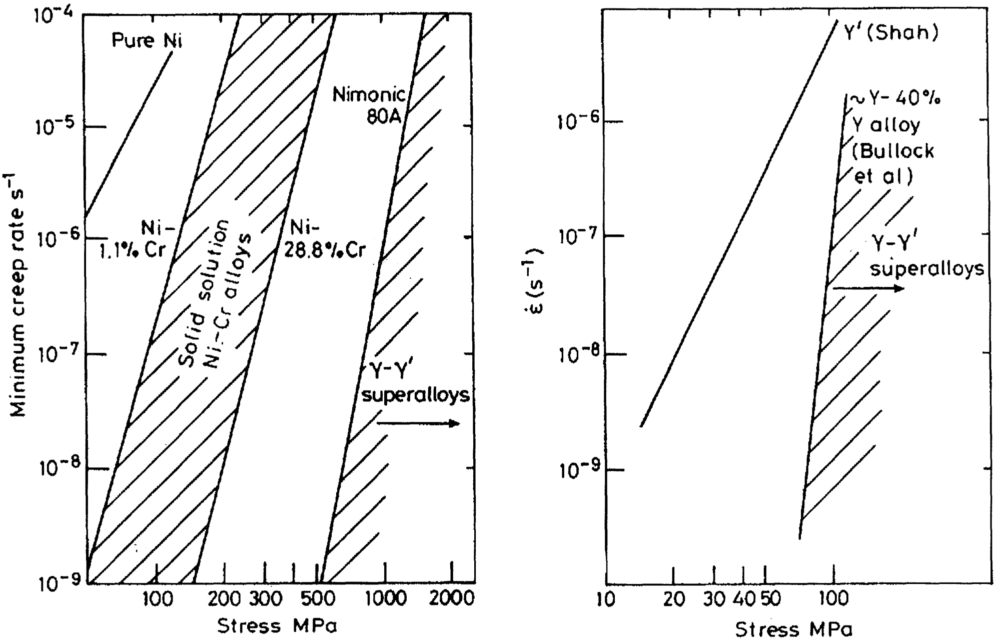

Figure 3. Comparison of the creep behaviour of (*a*) solid solution Ni–Cr alloys, (*b*) γ' Ni_3(Al,Ti) intermetallic compounds and (*c*) γ/γ' nickel-base superalloys.

coherently dispersed γ' precipitate (up to 70% by volume) in the γ-nickel matrix. However, in spite of extensive fundamental studies there is still no clear consensus on the mechanisms controlling the engineering performance in service conditions. Here we focus on a limited number of aspects of the mechanical behaviour of these materials that have been thought to be of particular importance.

Superalloys with high volume fractions of γ' show a similar anomalous rise in yield strength with increasing temperature as is exhibited by the monolithic $L1_2$ intermetallic phase (Stoloff 1987). Detailed analysis of this phenomenon has been carried out by a large number of authors, notably Paidar *et al.* (1984), and there is now a general consensus that it a consequence of thermally activated cross-slip onto cube planes that produces sessile dislocation segments that inhibit dislocation glide on octahedral planes. Hirsch (1992) has recently considered the detailed dislocation interaction that are occurring. Such considerations are likely to be important in service conditions where the yield stress is attained, or at least approached. However, design stresses will almost invariably be below the threshold for time-independent yield and γ' cutting is unlikely to be a significant factor during service.

In the sub-yield creep regime, relevant to turbine blades, it is clear that the creep performance of the duplex γ/γ' alloys is significantly superior either to (Ni,Cr) solid solution matrix or to the γ' Ni_3(Al,Ti) precipitated phase (figure 3). Consequently, the γ' behaviour cannot be taken as a limit to the alloy performance. Rather, the coexistence of the two phases requires the operation of a radically different deformation mechanism than would occur in either individual phase. Indeed at stress levels below those required for γ' shearing it is likely that dislocation activity is largely restricted to the γ matrix and this is supported

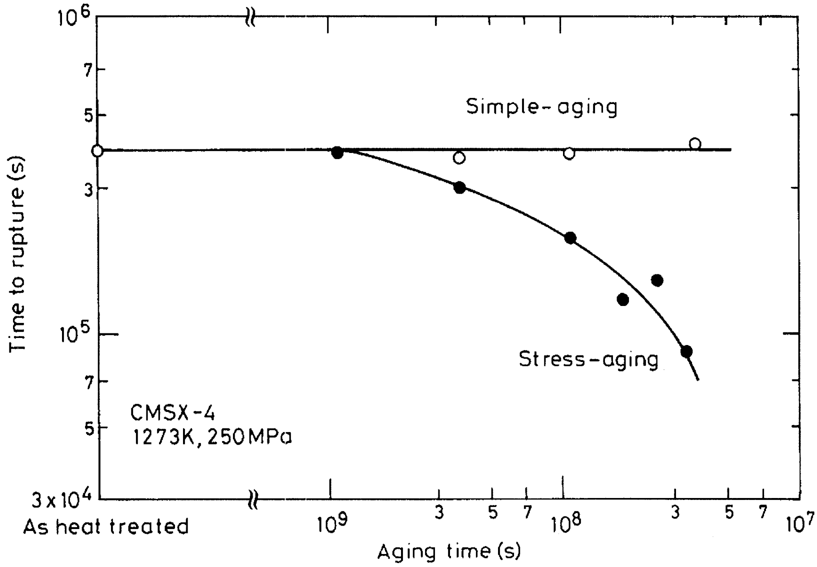

Figure 4. Comparison of the creep behaviour of the single crystal superalloy CMSX4 in two microstructural forms: (*a*) regular γ' morphology produced by commercial solution and ageing heat treatment and (*b*) rafted γ' morphology produced by heat treatment under stress (Kondo *et al.* 1994).

by transmission electron microscopy of creep deformed material (Henderson & McLean 1983). Dyson, McLean and co-workers (e.g. Ion *et al.* 1986; Dyson & McLean 1990) have developed a model of creep deformation that considers the dispersed particles to inhibit glide in the matrix: deformation occurs at a rate largely determined by dislocation climb and dislocations generated are mostly mobile leading to an increase in creep rate with accumulated plastic strain instead of the normal work-hardening exhibited by single-phase metals. This successfully accounts for a range of observations relating to the creep behaviour of superalloys that are not compatible with earlier models:

(i) Creep in both tension and compression exhibits a progressively increasing creep rate, rather than a steady state deformation rate.

(ii) Plastic prestrain of superalloys increases the creep rate relative to the unstrained material, rather than leading to strain hardening.

(iii) There is little difference in creep curves of superalloys produced under constant load and constant stress conditions, indicating the dominance of an intrinsic strain softening mechanism over the effect of increased stress due to reduction in cross-sectional area during tensile deformation.

(*a*) *Microstructural influences on mechanical behaviour*

An interesting microstructural feature of single crystal superalloys is the directional coarsening of the γ' particles that occurs during high-temperature ageing (greater than 1000 °C) in the presence of a small stress (Tien & Copely 1971). Plausible explanations have been given of the effect due to stress gradients generated as a result of differences in elastic and lattice constants of the γ and γ' phases. The rafted microstructures that develop in most commercial single crystal superalloys under tensile stresses have often been cited as the reason for their un-

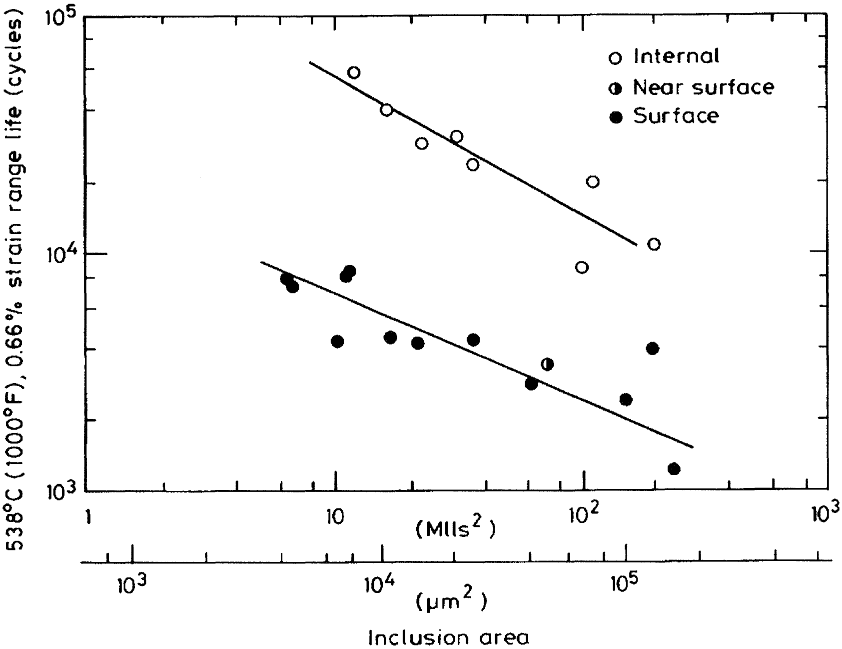

Figure 5. Variation in low cycle fatigue life with increasing volume fraction of inclusions in Rene 95 (Shamblen 1983).

expectedly good creep performance at temperatures in excess of about 1000 °C. However, it is now quite clear that the rafted γ' morphology is quite detrimental to the low and intermediate temperature creep behaviour (Caron *et al.* 1988; Kondo 1994). Figure 4, from the work of Kondo *et al.* (1994) clearly indicate that pre-rafting of the γ' reduces the rupture life relative to both the original material and that subject to conventional heat treatment. The benefits of a rafted γ' is now open to question particularly in the variable stress and temperature conditions likely to be experienced in service.

The importance of inclusions in controlling the low-cycle-fatigue life of turbine disc alloys has been demonstrated unambiguously in a number of studies. Shamblen (1993) for example, has deliberately added inclusions of known size and concentration to the alloy RENE95 and has shown a progressive decrease in cycles to failure with increasing inclusion volume fraction (figure 5). Pineau (1990) has paid particular attention to the problems of characterizing the fracture behaviour of materials with very low concentrations of defects. Mechanical testing of such materials must be carried out on a sufficiently large volume of material to ensure a high probability of the occurrence of defects characteristic of the component of interest.

4. Engineering considerations

The implementation of computer-aided design methods, to replace the traditional design codes, depends on there being a sufficiently sophisticated representation of the material behaviour. Whereas previous designs of gas turbine blades were based on simple measures of material performance, such as stress rupture life and minimum creep rate, in the future attempts will be made to numerically simulate the performance of a component in likely service cycles that will inevitably involve multiaxial stresses and variable stresses and temperatures. It is unrealistic to collect an experimental database to cover all possible options, particularly in the case of anisotropic materials such as single crystal superalloys. It is necessary to devise a reliable approach to the extrapolation/interpolation of a restricted database through the use of appropriate constitutive equations that can be incorporated in the design calculations. Several empirical approaches (Graham & Walles 1955; Evans & Wilshire 1987) have been successful in representing uniaxial creep databases, but these are difficult to extend to variable and multiaxial loading conditions.

The mechanisms of creep deformation of superalloys, described in the previous section, have been translated into such a set of constitutive equations using the general formalism of continuum damage mechanics. Here, the uniaxial creep rate $\dot{\epsilon}$ is expressed as a function of state variables (or damage parameters) that represent the current condition of the material, in particular of the structural and microstructural features that control the strength of the alloy. For the isotropic form of the model appropriate to superalloys, an acceptable fit of creep data can be obtained by using two state variables; S is a dimensionless internal stress that increases to a steady-state value of S_{ss} leading to primary creep and $\omega = (\rho - \rho_{\mathrm{i}})/\rho_{\mathrm{i}}$ represents the increasing density of mobile dislocations ρ (ρ_{i} is the initial value). The isotropic creep behaviour is represented by the following set of three equations involving the two variables S, ω and four constants $\dot{\epsilon}_{\mathrm{i}}$, where $\dot{\epsilon}$ has dimensions of s^{-1} and H, S_{ss}, C are all dimensionless (Ion *et al.* 1986; Dyson & McLean 1990):

$$\dot{\epsilon} = \dot{\epsilon}_{\mathrm{i}}(1 - S)(1 + \omega), \quad \dot{S} = H\dot{\epsilon}_{\mathrm{i}}(1 - S/S_{\mathrm{ss}}), \quad \dot{\omega} = C\dot{\epsilon}. \tag{4.1}$$

The model has been extended by Ghosh *et al.* (1990) to represent anisotropic creep of single crystals by considering creep deformation to be restricted to specific slip systems and computing the total strain resulting from each shear displacement. A set of equations equivalent to equations (4.1), but expressed in terms of shear, rather than tensile, strains γ^k is required for each family of slip systems. Then the total displacement ϵ_{ij} from all N components of shear on the allowed system $(n_1 n_2 n_3)\langle b_1 b_2 b_3\rangle$ is given by

$$\epsilon_{ij} = \sum_{k=1}^{N} \gamma^k b_i^k n_j^k. \tag{4.2}$$

Here i, j represent the cube directions and k identifies one of the slip systems being considered: γ^k is the amount of shear on that system. An arbitrary crystal direction x will transform to a new orientation X, with a strain in that direction of $(\bar{X} - \bar{x})/\bar{x}$, where

$$\begin{bmatrix} X_1 \\ X_2 \\ X_3 \end{bmatrix} \begin{bmatrix} 1 + \epsilon_{11} & \epsilon_{12} & \epsilon_{13} \\ \epsilon_{21} & 1 + \epsilon_{22} & \epsilon_{23} \\ \epsilon_{31} & \epsilon_{32} & 1 + \epsilon_{33} \end{bmatrix} \begin{bmatrix} x_1 \\ x_2 \\ x_3 \end{bmatrix}. \tag{4.3}$$

In the analysis dislocation activity is taken to occur on two families of slip systems, $\{111\}\langle\bar{1}10\rangle$ and $\{100\}\langle 011\rangle$. There is considerable evidence of the occurrence of octahedral slip during creep, but only occasional observations of cube slip (Caron *et al.* 1988). However, elements of octahedral to cube cross-slip are also thought to occur.

For the simple axial orientations that are symmetrically disposed to the active slip systems there is no change in orientation during deformation and the tensile and shear formulations are totally equivalent being two different mathemetical representations of the same physical model. The model can accurately represent individual creep curves, as can other more empirical approaches. However, the anisotropic model is also capable of representing (i) change in magnitude and order of creep anisotropy with stress and/or temperature; (ii) crystal rotations during creep; (iii) strain in any direction; (iv) change in material shape during deformation.

When combined with a representation of elastic anisotropy, a wide range of strain and load controlled types of deformation, such as stress relaxation and low cycle fatigue can also be simulated (Pan *et al.* 1993).

The success of such a model can only be assessed through searching validation and, as has been indicated above, comparison of measured and model-calculated creep curves does not constitute an adequate test. Here we indicate three quite different experimental validations of the predictions of the model.

(*a*) *Low cycle fatigue*

Figure 6 shows successive stress–strain loops for a strain controlled low-cycle fatigue test on SRR99 at 950 °C with a fixed tensile strain range between 0 and 0.75%. Here the loading is axial along the ⟨001⟩ direction. Model simulation has been carried out assuming that creep, as described by the parameters derived by analysis of a database of constant stress creep tests, is the dominant deformation mechanism; the only additional information required is the value of Young's modulus at the deformation temperature in order to account for elements of stress relaxation. Both measured and calculated behaviour show that individual cycles exhibit essentially elastic response, but there is a progressive shakedown due to stress relaxation with the maximum tensile stress decreasing from about 700 to 360 MPa and a compressive stress developing from 0 to about 300 MPa. Bearing in mind the relatively high stresses and strain rates in the LCF compared with those in the creep database, the prediction is remarkably accurate (Pan *et al.* 1994).

(*b*) *Shape changes*

A series of creep specimens of SRR99 tested to fracture with tensile loads along complex crystallographic directions have been examined to characterize the change in cross-sectional shape that develops along the necked portion of the fractured test pieces which were originally cylindrical: these measurements have been compared with computer simulations particular attention being paid to the crystallographic directions along which the maximum and minimum diameters develop (Pan *et al.* 1994). Figure 7 shows the different predictions of the shape development that would occur due either to octahedral or to cube shear deformation occurring alone; the experimental measurements for the example shown are clearly consistent with cube slip being dominant for that particular orientation. As expected from Schmid factor considerations orientations close to ⟨001⟩

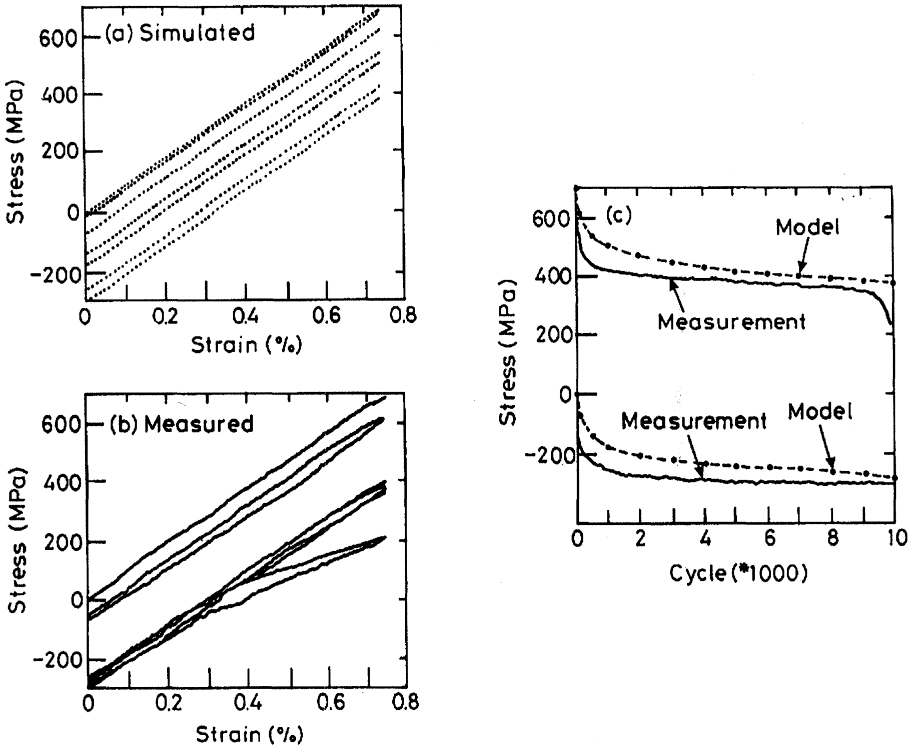

Figure 6. Simulated and measured low cycle fatigue behaviour for strain controlled cycling between 0 and 0.75% uniaxial tensile strain at 950 °C for ⟨001⟩ SRR99. (*a*) Simulated and (*b*) measured stress–strain curves for selected cycles, and (*c*) comparison of predicted and measured maximum and minimum stresses as a function of cycle number. (Pan *et al.* 1994.)

and ⟨111⟩ exhibit octahedral and cube slip dominance respectively, and give good agreement with the model predictions.

(*c*) *Orientation changes*

The same specimens used in the shape change measurements have been sectioned and the changes in orientation along the specimen length have been determined by using electron backscatter patterns produced in scanning electron microscopy (Dingley 1984). This technique has a high spatial resolution allowing local orientations to be determined on a scale of less than 1 μm. The results show an average drift of orientation that is largely consistent with the model simulation (figure 8) indicating dominant octahedral and cube shear near ⟨001⟩ and ⟨111⟩ orientations respectively, as was observed in the shape change experiments. However, deformation is highly heterogeneous leading to a growing spread of orientations, on a scale of a few micrometres, with increasing creep strain. High deformation is particularly associated with defects such as solidification porosity which in SRR99 aligns along the ⟨001⟩ solidification direction.

The design of discs using material with low fracture toughness is subject to the quite different problem of accounting for the possible presence of defects that can constitute, or grow to, a critical crack leading to brittle failure. Here, a probabilistic approach must be taken when, as is projected to be the case in the next generation of disc alloys, there is a sparse distribution of inclusions. Pineau

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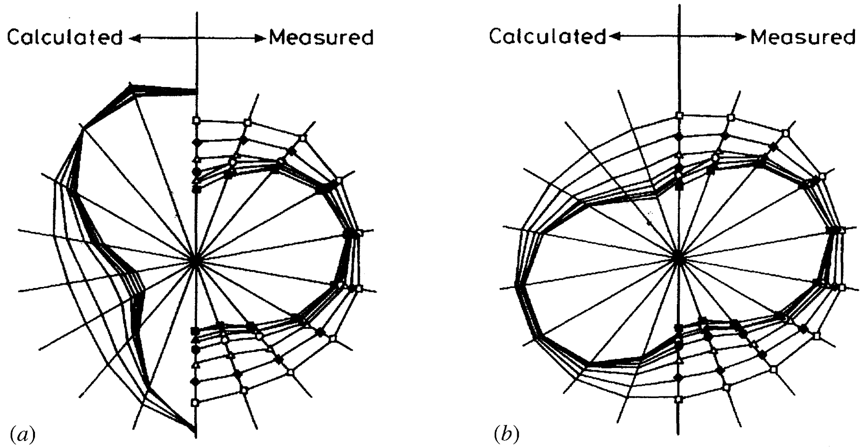

Figure 7. Comparison of the measured and predicted creep specimen cross-sections, shown as two halves of the same polar plot, at various values of reduction in area for a specimen of SRR99 of complex orientation after creep testing to fracture at 850 °C/450 MPa. (*a*) Octahedral slip and (*b*) cube slip. (Pan *et al.* 1995.)

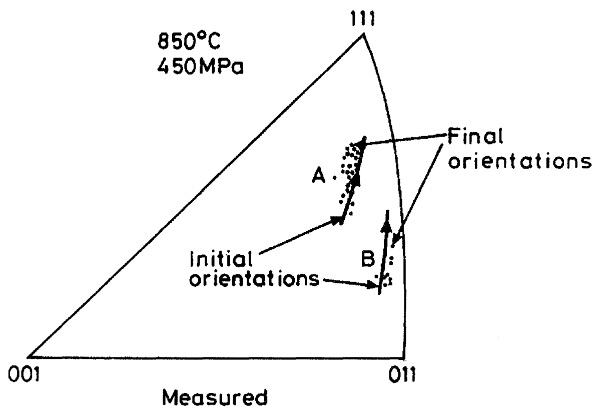

Figure 8. Change in orientation with creep strain for a specimen of SRR99 after tensile deformation at 950 °C/300 MPa showing model predictions and measurements using electron backscatter patterns.

(1990) has considered this problem in some detail showing the effect of component size on the design performance using such materials. The practical difficulty is in determining the probability distribution of defects when they are too dilute to be characterized by metallographic or conventional NDT techniques. Advanced methods, such as the EBBM approach developed at NPL, can be useful in providing a ranking of the general cleanness of various materials, but they can only sample a small fraction of the alloys from which components are manufactured.

Because of the catastrophic implications of a disc failure, it is essential to develop a quality assurance and design strategy that will guarantee the component integrity. Since the resolution of available techniques for non-destructive evaluation is inadequate, attention has been focused on assuring material quality through control of the entire processing cycle. This has led to growing interest in modelling the critical parts of the process, such as VAR, ESR and EBBM, to identify the optimum values of the process parameters required to achieve the desired microstructure.

5. Conclusions

The development of nickel-base superalloys has been one of the major successes of modern metallurgy. Over a period of 50 years or so very significant improvements have been achieved by a combination of alloy and process development, and these have rapidly been introduced into service. However, the scope for further substantial improvements in material properties, particularly for turbine blade alloys, is limited; here the most advanced materials are used in single crystal form at greater than $0.8T_m$.

By contrast, full exploitation of these materials has not yet been achieved. Two particular areas have been identified where substantial improvements in the engineering applications can be expected and which are at an early stage of development:

(i) Design procedures will be developed to make use of the constitutive laws accounting for the anisotropic mechanical behaviour single crystal superalloys. These should be capable of simulating component performance in realistic service conditions incorporating multiaxial stresses and variable loading conditions.

(ii) Control of the defect density in turbine disc alloys is required in order to achieve the loading conditions expected in future gas turbine designs. This is most likely to be achieved through process control informed by detailed process modelling.

Several colleagues have contributed significantly to the work reviewed in this paper; these include Dr M. B. Henderson, Dr R. N. Ghosh, Dr L. M. Pan and Dr B. A. Shollock. I also thank the Defence Research Agency for support of much of the work.

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Discussion

G. A. Webster (*Imperial College, London, UK*). I would like to comment on Professor McLean's remarks that alloy development is nearly complete, some process development is possible and that the biggest scope is for engineering exploitation. I agree with these views. Single crystals are being used for turbine blade applications and Professor McLean showed how behaviour in specific crystallographic orientations could be predicted for uni-axial stressing. The presence of thermal gradients in cooled blades will introduce a multi-axial stress state in these components and thorough understanding of the response of single crystals to multi-axial stress is needed before their full potential can be exploited. Also, there may be scope for deliberately introducing compressive residual stresses in engine discs to reduce the influence of small critical effects in high strength disc alloys.

M. McLean. I fully agree with Professor Webster's initial remarks. Too much emphasis has been given in the past on the problems of extrapolating short-term data to long times; this is not a serious problem for aerospace applications. Rather, there is a need to predict multi-axial and cyclic loading responses from uniaxial data. I believe that the state-variable approach is beginning to allow us to do so. I am less sure of the viability of deliberately imposing compressive residual stresses; I would be concerned that they would relax during high-temperature service.

F. R. N. Nabarro (*CSIR, Pretoria, South Africa*). To use an analogy introduced by our chairman many years ago, the dislocations are moving in narrow channels, so flow is lamellar rather than turbulent. Only one family of dislocations is active, and these do not obstruct one another.

M. McLean. This observation must be basically correct. One might debate if a

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single set of dislocations will propagate along the cube-oriented channels between γ particles; perhaps there will be activity on two octahedral planes. However, clearly there will be little potential for work hardening.

A. Cottrell (*University of Cambrdige, UK*). Professor McLean said that nickel is near its limit of development as a high-temperature material. This must be so, since the working temperature of the nickel superalloys is so near the melting point of nickel. As well as its excellent mechanical properties, i.e. high-temperature creep strength and high fracture toughness, the nickel superalloy has another major property which is so good that it is often taken for granted and overlooked: its excellent oxidation resistance. I think that there is, and will be, great difficulty in moving beyond nickel to more refractory materials, because so few of them have good oxidation resistance at really high temperatures. If it were not for this we might now be using niobium or molybdenum blades in advanced gas turbines. The same problems are likely to apply generally to intermetallics and carbide ceramics. Probably the only way forward, beyond the nickel superalloys, for the very highest temperatures, is to develop refractory oxides into useful engineering materials.

M. McLean. In terms of oxidation resistance, there is a compromise course of coating the nickel-base superalloys and this is already extensively used in practice. Ceramic coatings are used for their thermal-barrier qualities, as well as for oxidation resistance. I agree that oxide ceramics offer the potential of intrinsic strength and oxidation resistance. If their toughness can be improved to a sufficient level they will certainly find an application. It must also be borne in mind that ceramic turbine blades are unlikely to be cooled, as are nickel superalloy blades. Consequently, the comparison must be between the material operating temperature in the case of the ceramic, with gas temperature for the superalloy, which may be about 300 K in excess of the metal temperature.