The high-temperature multiaxial creep behaviour of alloy 800H tubes

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THE HIGH-TEMPERATURE MULTIAXIAL CREEP BEHAVIOUR
OF ALLOY 800H TUBES

by

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for the award of
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Supervised by:  Dr.R.C.Hurst
                Dr.T.E.Chung

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This work was carried out by the author at the Joint Research Centre
of the Commission of the European Communities, Petten, Netherlands,
while registered for the degree of Ph.D. with Loughborough University
of Technology.
# THE HIGH-TEMPERATURE MULTIAXIAL CREEP BEHAVIOUR OF ALLOY 800H TUBES

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DECLARATION

The contents of this thesis and the original aspects of the work are the sole responsibility of the author.
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ABSTRACT

THE HIGH TEMPERATURE MULTIAXIAL CREEP BEHAVIOUR OF ALLOY 800H TUBES

Author: A.S. McALLISTER

Literature concerning the development of mechanistic and empirical models of creep deformation and failure, under uniaxial and multiaxial stress systems are reviewed. The detailed metallurgy of the alloy is discussed and its specific creep behaviour in the experimental temperature range is reviewed from published experimental data, mostly obtained from the Alloy 800 Data Bank of the EEC Joint Research Centre at Petten.

Experimental work is subdivided into four areas,

i) Constant-stress and constant-load uniaxial solid bar experiments on specimens cut from the tube walls.
ii) Constant load experiments on tubular specimens.
iii) Constant internal pressure experiments on tubular specimens
iv) Combined internal pressure and axial load experiments on tubular specimens

Uniaxial constitutive equations are first derived from solid bar constant-stress results and the model checked, for time- and strain-hardening conditions, against the solid-bar constant-load results. These relationships are then incorporated into a multiaxial constitutive equation and the predicted axial and diametral strains compared with the internal pressure and combined internal pressure and axial load experimental results. The model is then updated to take account of possible material anisotropy and the predictions again compared to the experimental results.

Mechanistic models for failure of bar and tubular specimens are derived from considerations of specimen microstructure, high temperature oxidation and a knowledge of the real stress and strain-rate distributions in the specimen sections. Semi-empirical models are also derived for the multiaxial creep rupture criterion (MCRC) based on the rupture-lives of solid bar and tubular specimens and on the conclusions of the mechanistic failure model.

The main conclusions of the work are that the tertiary creep deformation behaviour and the creep fracture behaviour of the alloy, under multiaxial stress states, are controlled by a Tresca equivalent stress criterion.

KEY WORDS: Metallurgy
High Temperature
Creep
Alloy 800H
Tubes
Multiaxial Stress Systems
Mathematical Modelling
GLOSSARY OF TERMS

\( \sigma \) Direct stress
\( \tau \) Shear stress
\( \sigma' \) Deviatoric stress
\( \sigma_m \) Hydrostatic stress
\( \sigma_o \) Initial stress
\( T_{ij} \) Stress tensor
\( \sigma_{\text{eff}} \) Effective stress
\( \gamma \) Yield stress
\( \sigma_{\text{max}} \) Maximum principal stress
\( \varepsilon_{\text{tot}} \) Accumulated axial strain
\( \varepsilon \) Direct strain
\( \dot{\varepsilon} \) Direct strain-rate

( )\(_1, 2, 3\) Principal components
( )\(_i, j, k\) Principal components
( )\(_x, y, z\) Orthogonal components
( )\(_z, r\) Polar coordinates
( )\(_E\) Equivalent component
( )\(_g\) Generalised component
\( I_{1, 2, 3} \) Stress invariant
\( J_{1, 2, 3} \) Deviatoric stress invariant
DPN Diamond pyramid number
\( p \) Pressure
\( L \) Axial load
\( G \) Shear modulus

\( \phi(t) \) Function of time
\( t \) Time
\( \beta \) Andrade primary creep parameter
\( \nu \) Poissons ratio
\( k \) Boltzmann's constant
\( R \) Universal gas constant
\( \Omega \) Atomic volume
\( k^* \) Volterra integral operator
\( D \) Diffusion coefficient
\( D_v \) Lattice diffusion coefficient
\begin{align*}
D_b & \quad \text{Boundary diffusion coefficient} \\
b & \quad \text{Burger's vector} \\
Q & \quad \text{Activation energy} \\
Q_l & \quad \text{Activation energy for lattice diffusion} \\
Q_v & \quad \text{Activation energy for self diffusion} \\
Q_p & \quad \text{Activation energy for pipe diffusion} \\
Q_{ss} & \quad \text{Activation energy for steady-state creep} \\
\Delta G_F & \quad \text{Free energy of formation} \\
T & \quad \text{Absolute temperature} \\
T_m & \quad \text{Melting point temperature} \\
r & \quad \text{Radius} \\
R_o & \quad \text{Outside tube radius} \\
R_i & \quad \text{Inside tube radius} \\
R_m & \quad \text{Mid-wall tube radius} \\
t & \quad \text{Tube wall thickness} \\
w & \quad \text{Grain boundary width} \\
d & \quad \text{Grain size term} \\
n, m & \quad \text{Bailey's constants} \\
\xi & \quad \text{Kanter and Nadai viscosity term} \\
w & \quad \text{Rabotnov-Kachanov damage parameter} \\
\dot{w} & \quad \text{Rabotnov-Kachanov damage-rate parameter} \\
A, B; u, v & \quad \text{Rabotnov-Kachanov material constants} \\
P & \quad \text{Larson-Miller parameter} \\
C & \quad \text{Larson-Miller constant} \\
t_r & \quad \text{Rupture-time} \\
A, n & \quad \text{Norton power-law constants} \\
\theta_1, 2, 3, 4 & \quad \text{Theta-projection equation coefficients} \\
P, G, H & \quad \text{Hill's anisotropic constants} \\
R, P & \quad \text{Backofen and Duncombe's anisotropic constants} \\
\alpha, \beta, \gamma & \quad \text{Hayhurst's rupture criterion constants} \\
A, A', A'' & \quad \text{General constants} \\
A^*, A^{*o} & \quad \text{General constants} \\
B, C & \quad \text{General constants}
\end{align*}
1.0 INTRODUCTION

Incoloy 800 was first developed in the late 1940's as a low nickel replacement for the "Nichrome" electrical filament alloy arising out of the world nickel shortage at that time. The alloy was first designed to exhibit good corrosion and oxidation resistance in air and other aggressive environments and moderate mechanical strength at elevated temperatures, particularly under creep conditions.

In time, the alloy exceeded its initial design brief and has proven itself in such applications as furnace furniture components and chemical reaction circuitry, particularly in ethylene cracking, steam reformer and coal conversion plant. More recently, it has been proposed as a candidate material for the sodium coolant circuit of the liquid metal fast breeder reactor (LMFBR) project. The codes used in the design of components to operate in the creep regime are usually based on some manipulation of material data derived under uniaxial stress conditions. These codes frequently give rise to over conservative designs due to the lack of relevant information concerning the stress-state dependence of the alloy's creep deformation and rupture behaviour.

The object of this project is therefore to characterise the stress-state dependence of the creep deformation and rupture behaviour of Alloy 800H so that either the suitability of designs based on uniaxial data may be confirmed or that a more rigorous approach to designing under multiaxial stress conditions might be suggested.

To achieve such an objective requires three major areas of consideration: first, the characterisation of the uniaxial material properties; secondly, a means of conducting experiments under controlled multiaxial stress systems; and thirdly, the development of a mathematical model capable of predicting the creep deformation behaviour under multiaxial conditions based on the uniaxial material properties.

Uniaxial material properties are normally derived experimentally by testing small bar specimens, machined from bar-stock material of a similar composition to the component. Alternatively, however,
specimens may be machined directly from a component or even the component itself may be tested in uniaxial tension, if this is possible. If no possibility exists to conduct experimental programmes, the uniaxial data must then be taken from existing information sources such as published literature or data banks. Reasoned consideration of these alternatives backed up with experimental results forms the first major contribution to this work.

Many experimental techniques are available for the generation of multiaxial stress systems, for example the torsion or internal pressure testing of tubular specimens or the axial loading of notched bar specimens. None of these techniques are standard, and as such there is no guarantee as to the validity of the results these techniques may produce. For this reason, appropriate selection should be made of a technique which has direct bearing on the application to be modelled. In respect of this project, the internal pressure testing of tubes is of direct practical significance since, as previously indicated Alloy 800H is most commonly found as tubular components loaded by internal pressure. One further advantage of using tubular specimens is that later developments of this project envisage the characterisation of the mechanical behaviour of Alloy 800H in corrosive environments, for which the corrosive gases could be conveniently applied to the inside surfaces of the tube. Internal pressure alone, however, can only give rise to one particular multiaxial stress system. Variation of the principal stress ratio in a controlled manner, however, requires in its simplest form, only the superposition of an additional axial load in addition to the internal pressure.

The development of the multiaxial creep model first requires the characterisation of the uniaxial material properties. This often consists of deriving some satisfactory mathematical representation for the stress and temperature dependence of the creep curves. A number of approaches exist for the development of the multiaxial model itself and critical evaluation of these available techniques is required in order to include those elements within the model which are most relevant to the characteristics of the multiaxial experiments and the application to modelled.

The validity of modelling experimental and rupture behaviour based upon uniaxial material data and the derivation of the stress-
state dependence of these properties for the alloy must then be confirmed through a comparison of the multiaxial experimental results and the predictions of the model.
2.0 LITERATURE SURVEY

2.1 UNIAXIAL CREEP BEHAVIOUR

The creep behaviour of metals under uniaxial constant-load or constant-stress conditions may be idealised by the strain v. time curves shown in figs. 1. Normally, three main regions can be identified: a primary region of decreasing creep-rate; a secondary region of constant creep-rate; and a tertiary region of increasing creep-rate leading to fracture. Depending upon the test material and the precise stress and temperature conditions prevailing, any of these regions may be accentuated or absent. Additionally, as the stress level is reduced the constant-load curve tends towards the constant-stress curve as a result of the reduced strain-rates and the consequentially small change in true stress under constant-load conditions.

In general, uniaxial creep behaviour can be studied in two main ways, that is, either empirically or mechanistically.

The mechanistic approach attempts to model a material's creep behaviour by considering sub-microstructural physical mechanisms, while the empirical approach attempts to model a material's strain-rate behaviour as functions of stress, strain, time and temperature, this being usually based on the application of some form of curve fitting routine to experimental results.

There is, at this time, a growing trend to bring these two approaches closer together, either by attaching mechanistic significance to empirically derived constants or by developing mechanistic models for real engineering alloys. They are, however, seldom compatible.
2.1.1 Mechanistic approaches to creep deformation

Andrade (1) was the first to consider grain boundary sliding and lattice slip as the predominant mechanisms for creep deformation as a result of observations made from metallographic examination of specimens subjected to creep deformation.

McLean (2) followed this up in his work, where he attempted to relate the relative contributions of these mechanisms to the overall creep strain. His initial findings, based on microstructural measurements, revealed intra-crystalline slip to be responsible for about one-half and grain boundary sliding approximately 1/20th of the total measured creep strain. In this way he identified a missing creep component. Further investigation revealed (2) that this missing creep component was due to fine slip within the observed slip bands and to the polygonisation of dislocations, this leading to the formation of subgrains.

His approach was criticised on the grounds that it assumed the two components to act independently without interaction. This is unlikely since there must be some interaction in order to maintain compatibility of the grains at their boundaries.

Dorn (3), in his analysis of creep, investigated the dependence of creep strain on temperature, by using an analysis based on an Arrhenius function of the type

$$\varepsilon = f(t \exp(-Q/RT))$$  \hspace{1cm} (1)

By conducting experiments under constant-stress conditions over a small temperature range, he evaluated the activation energy for steady-state creep, $Q_{ss}$, and showed that it related well to the activation energy for self-diffusion, $Q_v$. By extending his temperature range, he discovered that the value of $Q_{ss}$ changed discretely with temperature, which he attributed to a change in creep mechanism.

Weertman (4), in an effort to identify the mechanisms suggested by Dorn, developed a theory based on two competing mechanisms. The first was the gradual reduction of available slip systems due to the interactions of dislocations with other dislocations and obstacles which he described as a hardening mechanism. The second mechanism was
a thermally activated mechanism in which dislocations could overcome these restrictions via non-conservative motion, allowing recovery of the structure.

He suggested two specific mechanisms for recovery. The first involved dislocation "cross-slip", where a screw component dislocation segment could change plane to another of the same family. The second, more important mechanism considered "climb" of edge dislocations out of plane, due to the presence of vacancies on adjacent planes. In this way he correlated the rate dependence of creep to the rate of vacancy diffusion in the lattice, or, as shown by Dorn, the activation energy for self diffusion, $Q_v$.

Weertman (5) quantified the climb mechanism in terms of the stress dependence of the steady-state creep-rate and showed that a dependence of the order

$$\dot{\varepsilon} \propto \sigma^{4-5}$$

was obtained.

Much work has been carried out in the field of dislocation creep mechanisms since the time of Weertman's initial observations, on materials ranging from pure metals to industrially relevant alloy systems. The results and conclusions of these studies are found to fall conveniently into three main groups: the behaviour of pure metals; solid solution alloys; and particle hardened alloys.

2.1.1.1 Pure metals

If measurements are taken to determine the steady-state creep rate of pure metals at constant temperature over a wide range of stresses, and these results presented as $\log \dot{\varepsilon} v. \log \sigma$, two slightly different behaviours are observed for temperatures near to $0.5T_m$ and $T_m$ (figs.2.&.3.)

In both cases the stress dependence of an "intermediate" stress range is found to be approximately equal to 5 as shown by Weertman. Transmission Electron Microscopy (TEM) studies in this region have shown the primary stage of creep to be associated with the
development of a dislocation sub-structure. During steady-state creep the sub-grains remain equiaxed and of a constant average grain diameter, which is inversely proportional to the applied stress (6). Further investigation of the sub-grain boundaries have shown them to be in fact in a condition of constant dynamic equilibrium, with migration of the boundaries occurring as a result of the annihilation of dislocations in the boundary and the pile-up of new dislocations from within the subgrains.

At "low" stresses, the stress dependence of the steady-state creep rate decreases to be approximately equal to 1. In this region a number of creep mechanisms have been proposed. Nabarro (7) and Herring (8) suggest that, at these "low" stresses, creep could occur by the diffusion of vacancies through the crystal lattice. They developed a model where stress directed diffusion of vacancies could occur under conditions of applied load (fig.4.). Whilst they specifically looked at diffusion through the lattice, Coble (9) examined the possibility of grain boundaries acting as preferential diffusion paths. Their results can be combined and summarised by the following equation

\[ \dot{\varepsilon} = \frac{14.3 \alpha \Omega}{kT} \frac{1}{d^2} \cdot D_v \left[ 1 + \frac{w \cdot D_b}{\pi d D_v} \right] \]  

Due to their grain size dependence, these mechanisms become increasingly less important as the grain size increases due to the long diffusion paths involved. Harper and Dorn (10)(11), however, identified a dislocation mechanism in this region where climb of vacancy saturated edge dislocations is controlled by the rate of diffusion of vacancies to and from the dislocation line.

TEM investigations of materials exhibiting Harper-Dorn creep have shown that no dislocation sub-structure is formed during primary creep. Instead uniform dislocation fields are observed, the density of which appear to be independent of applied stress.

At "high" stresses, breakdown of the power law creep behaviour is found, with the dislocations being able to overcome barriers and obstacles by glide. Only at temperatures near to 0.5T_m is an additional mechanism observed between the "intermediate" and "high" stress regions. This behaviour is referred to as "pipe diffusion"
(12) and occurs due to the diffusion of vacancies along the cores of the dislocations. Under these conditions the stress dependence of steady-state creep has been found to be approximately 7, the creep activation energy being controlled by $Q_p$, the activation energy for pipe diffusion, such that

$$Q_p = 0.6 Q_v$$

2.1.1.2 Solid solution alloys

As was seen for the pure metal type behaviour, the steady-state creep-rate is found to be controlled by the climb mechanism.

With solid solution alloys, however, solute atmospheres have been found to form around the dislocations which reduces the rate of the glide mechanism due to the dragging effect of the atmospheres. Under these "viscous-glide" conditions it has been shown both experimentally (13) and theoretically (14) that the stress dependence of strain-rate is approximately equal to 3, the activation energy for creep being controlled by $Q_i$, the activation energy for interdiffusion of the solute atoms.

TEM investigations have shown that under these conditions no formation of sub-grain networks occur, an essentially uniform distribution of dislocations being observed instead.

Plots of $\log \dot{\varepsilon}$ vs. $\log \sigma$ behaviour for solid solution alloys at around $0.5T_m$ (fig.5.) show identical behaviour to the pure metal type at "low" and "high" stresses with the "viscous-glide" mechanism dominating at "intermediate" stresses.

This mechanism can, however, be confined to quite narrow stress ranges with the possibilities of transitions to pure metal, climb controlled, creep behaviour at higher and lower stresses. At lower stresses, this transition occurs since climb and glide occur sequentially, the slower mechanism being rate controlling at any particular stress level. At higher stresses, transition may occur due to the stress level being sufficiently high to allow the dislocations to break free from their solute atmospheres.

It should be emphasised that this is an idealised solid solution
alloy behaviour, the specific ranges of each mechanism being
dependent upon such factors as the solute:solvent atom size ratio and
solute concentration (15)(16).

In practice, the results of the steady-state creep-rate v. stress
behaviour of both the pure metal and the alloy type materials, may
be conveniently described by a single general equation,

\[ \frac{A^* G b D}{kT} \left[ \frac{b}{d} \right]^p \left[ \frac{\sigma}{G} \right]^n \]  \( (4) \)

the values of \( p \) and \( n \) for each of the encountered creep mechanisms
being given as

<table>
<thead>
<tr>
<th>Mechanism</th>
<th>( p )</th>
<th>( n )</th>
</tr>
</thead>
<tbody>
<tr>
<td>Coble creep</td>
<td>3</td>
<td>1</td>
</tr>
<tr>
<td>Nabarro-Herring creep</td>
<td>2</td>
<td>1</td>
</tr>
<tr>
<td>Viscous drag</td>
<td>0</td>
<td>3</td>
</tr>
<tr>
<td>High temperature climb</td>
<td>0</td>
<td>5</td>
</tr>
<tr>
<td>Low temperature climb</td>
<td>0</td>
<td>7</td>
</tr>
<tr>
<td>Harper-Dorn creep</td>
<td>0</td>
<td>1</td>
</tr>
</tbody>
</table>
2.1.1.3 Particle hardened alloys

Particle hardened alloys have been found by many workers to yield abnormally high values for the stress exponent of creep-rate, \( n \).

\( (17)(18)(19) \)

Work to resolve this discrepancy, using stress drop techniques during creep testing, has shown that the strain-rate behaviour may be described by consideration of a modified or effective stress term.

\[ \sigma_{\text{eff}} = (\sigma - \sigma_0) \]

which when substituted in equation (4) gives

\[ \dot{\varepsilon} = A_0^* \frac{D \cdot G \cdot b}{kT} \left[ \frac{b}{d} \right]^p \left[ \frac{\sigma - \sigma_0}{G} \right]^{\eta_0} \]

The term \( \sigma_0 \) is referred to as a so called "friction stress" which arises from the resistance to dislocation glide and climb, provided by the precipitate network.

Experiments have shown that this mechanism, and therefore \( \sigma_0 \), has a strong stress and temperature dependence \( (20) \) and it is for this reason that the large values of the stress exponent, \( n \), are predicted by equation (4).

In practice, use of equation (5) gives fairly constant values of \( n_0 \) in the range 3-4 which then corresponds well with the long range back stress created by the dislocation structure \( (21) \).

After analysing all the possible material deformation mechanisms, Ashby \( (22) \) concluded that only six were totally independent, these being

i) Theoretical shear stress

ii) Dislocation glide (conservative)

iii) Dislocation creep (non-conservative)

iv) Lattice diffusion

v) Grain boundary diffusion

vi) Twinning

not all of these being relevant under creep conditions.
He recognised that if he normalised the constitutive equations for each of these mechanisms for stress and temperature and solved each pair simultaneously he could map out boundaries in stress and temperature space, within which a single mechanism would dominate the material's deformation behaviour (fig.6). Contours of steady-state strain-rate were then superimposed on these diagrams. Such "deformation maps" enable the predominant mechanism and steady-state deformation rate for a material to be readily determined for any given set of stress and temperature conditions.

More recent work by Ashby (23) has concentrated on the extension of these initial ideas to include the more detailed creep mechanisms associated with real alloy systems.

2.1.2 Empirical approaches to creep deformation

By far the most work in this field has been concentrated on the investigation of the primary and secondary creep behaviour, due to the conservative policy of designing up to the end of secondary creep.

Andrade (1) is generally acknowledged as one of the first authors to attempt to empirically fit creep data to an equation, using the form

\[ \dot{\varepsilon} = \dot{\varepsilon}_0 (1 + \beta t^{0.33}) e^{kt} \]  

Since then many other equation types have been proposed to relate the strain-rate behaviour of a material to stress, strain and time.

The equations that have received the most attention are those describing the stress dependence of the creep-rate in the secondary creep regime. The most commonly encountered equation

\[ \dot{\varepsilon} = A \sigma^n \]  

is generally acknowledged to be due to Norton (24) with typical values for "n" being in the range 3-7. It should be remembered, however, that the value of "n" is only constant if the creep
mechanism does not change over the stress range considered.

Kauzman (25) and Cottrell (26), among others, have shown that at high stress levels a better correlation of minimum strain-rate with stress was obtained with an exponential function

\[ \dot{\varepsilon} = A' \exp(B\sigma) \]  \(\text{(8)}\)

The existence of equations (7) and (8) led to the conclusion, for some time, that two separate mechanisms were operational at high and low stress levels. Garofalo (27), however, considered this to be unlikely and instead proposed an equation encompassing the whole stress range.

\[ \dot{\varepsilon} = A''(\sinh C\sigma)^n \]  \(\text{(9)}\)

Rather than considering only one region of the creep curve some authors have tried to describe the creep curve as a whole. This approach is important when significant strain contributions are made from more than one region.

Graham and Walles (28) used polynomial-type equations to fit their data and attached stress and temperature dependence to the coefficients. Unfortunately, polynomial equations can be unsatisfactory as they predict infinite strain-rates at zero time.

This problem is not encountered if creep strain v. time data is fitted to an exponential equation of the type proposed recently by Evans, Parker & Wilshire (29).

\[ \varepsilon = \Theta_1(1 - \exp(-\Theta_2t)) + \Theta_3(\exp(\Theta_4t)-1) \]  \(\text{(10)}\)

The authors believe that creep behaviour may be defined by two competing regions, one of decreasing and one of increasing creep rate, the occurrence of a linear steady-state region being only the consequence of a balancing of these two components (fig.7.). They have shown, that when the results of high-accuracy constant-stress creep curves are fitted with this equation, good stress dependence of the theta coefficients are found.

Rabotnov (30) and Kachanov (31) proposed a parametric creep deformation model based on an "abstract" measure of the current state
of the material deterioration, \( w \). They proposed a modified Norton power law relationship to describe the stress dependence of creep rate, modifying the stress term to include the damage parameter, \( w \),

\[
\dot{\varepsilon} = A \sigma^n/(1-w)^n
\]

and assuming the form of the damage-rate term to be

\[
\dot{w} = B \sigma^v/(1-w)^u
\]

where \( A, B, n, v \) & \( u \) are material constants.

This type of model has been adopted by a number of authors (32)(33) and quantified to describe material creep deformation under particular conditions. It also has the added advantage that constitutive equations developed in this way have a built-in failure criterion, with rupture being realised at values of \( w = 1.0 \)

### 2.1.3 Mechanisms of creep failure

Tertiary creep leading to creep fracture is characterised by intergranular cavitation and crack propagation mechanisms, the beginnings of which are traceable back to the primary and secondary stages of creep by use of metallographic examination and density measurement techniques on tested samples (34). The exact fracture behaviour of a material is dependent on a number of factors, such as temperature, applied stress and material composition.

Two main types of failure initiation are observed in practice, wedge type cracks originating from grain boundary triple points; and void formation along grain boundaries.

Wedge, or \( W \)-type, fracture behaviour is usually observed on boundaries normal to the applied stress axis and arises out of the high stress concentration at grain boundary triple points due to grain boundary sliding (35).

Grain boundary cavitation, or \( r \)-type fracture behaviour, was originally thought to be activated via a vacancy agglomeration mechanism (36). Theoretical analysis has since shown that extremely
high concentrations of vacancies are needed for stable void formation and that stress directed diffusion is away from transverse boundaries where these cavities are observed. A more recent theory (37), suggests that, like W-type, r-type voiding also occurs as a result of grain boundary sliding. In this case, sliding of a jogged or ledged grain boundary may open up a stable void which may then grow by vacancy agglomeration mechanisms. Davies & Wilshire (38) have shown that void formation does not occur on boundaries orientated normal to the stress axis since no sliding is observed at these boundaries.

The same mechanisms can be enhanced in the presence of grain boundary second phase precipitates due to delamination of the precipitate matrix interface during sliding, giving rise to a larger initial void size. This mechanism is particularly prevalent in stainless steels, where Cr$_{23}$C$_6$ precipitates are present at the grain boundaries.

It is often difficult to separate the individual effects of temperature and stress on the dominant fracture mechanism, but an increase in service temperature tends to favour a change in mechanism from W-type to r-type fracture whilst decreasing stress tends to favour void formation rather than triple point cracking.

With prolonged exposure to these mechanisms, eventual link up of the r- and W-type damage zones occurs, to form small grain boundary macro-cracks. Eventual rupture of the specimen may then occur by the propagation of these macro-cracks, or, if there is no substantial link up of the micro-damage zones to form macro-cracks, then failure may occur due to the reduction in the nett section area, until this can no longer support the applied load.
2.1.4 Life-time prediction

Design of high temperature plant eg. steam raising or nuclear, typically requires material property data of the order of 10-30 years' duration. Obviously, experiments of this length of time are unrealistic, a preferred route being the extrapolation of accelerated test data to these times. Two approaches are generally employed i.e. an extrapolation with respect to stress or with respect to temperature. A third technique aimed at predicting the remaining life of a component is also reviewed.

2.1.4.1 Stress extrapolation

For this technique experiments are conducted at the service temperature over a range of stresses which are higher than the design stress. The simplest form of extrapolation considers a straight line relationship between log (stress) and log (rupture-time). However, these extrapolations are frequently inaccurate and misleading since changes in the slope, such as the commonly encountered "knee", of the curves cannot be anticipated.

Evans, Parker & Wilshire (29) have argued that great care must be taken with accelerated testing techniques so that the true material trends are not disguised by experimental errors. On this basis they have conducted high-precision constant-stress experiments and then modelled these using equation (10). By deriving the stress dependence of these theta parameters and assuming a strain-to-failure value as a rupture criterion they extrapolated these equations to lower stresses and longer times, and in this way were able to predict the "knee" in the stress v. time-to-rupture curve.

Hayhurst et al (39) have conducted a thorough investigation into this theta projection technique with particular attention to a 1%Cr-0.5%Mo-0.25%V alloy. After initial curve fitting of their constant stress results they found that the scatter in the theta coefficients was too great to obtain a stress dependence with a satisfactory confidence limit. To resolve this dilemma, attempts were made to truncate and refit the data, this being most successful when a
truncation parameter of seven times the creep-rate at the half-life period was used.

By extrapolating to lower stresses in the same way as Evans et al, they too were able to predict the "knee" in the stress v. rupture time curve and obtain good agreement with available long-term constant-load data.

The success of the technique, in being able to accurately model long-term behaviour, must be largely due to the selection of an equation type capable of empirically modelling the observed strain v. time effects from two recognised sub-microstructural creep mechanisms, i.e. work hardening and recovery. The failure of previous techniques, particularly those concentrating on secondary creep, was that this region of secondary creep was only a consequence of the stress and temperature regime chosen for the experiments. At higher or lower stress levels this behaviour may not persist, extrapolation into these regions then being invalid.

In connection with the mechanistic significance of the empirical equation type it is not unreasonable to consider the authors' tentative claims of correlation of some theta parameters with creep activation energies as valid.

Larson and Miller (40) proposed a parametric term \( P \) as a better parameter than rupture-time alone. \( P \) is a function of temperature and rupture-time (derived from an Arrhenius strain-rate v. temperature relationship) and is related to them by

\[
P = T_{c}(C + \log t_{c})
\]  

where \( C \) is a constant having a value in the range 15–23.

This has been shown to give remarkably linear correlations with rupture-time over wide ranges of stress, temperature and material types.

It should always be remembered that, although this technique offers a convenient method for lifetime prediction, it is based upon an arbitrary relationship which does not take account of the changes in material behaviour with change in stress and temperature. For the technique to be reliable, the parameter \( P \) should be insensitive to changes in temperature at constant-stress. Use of this approach for a 1/2%Cr-Mo-V ferritic steel in the temperature range 570 – 700°C has
been found, for example, to lead to optimistic lifetime predictions under service conditions (41).

2.1.4.2 Temperature extrapolation

This method requires the production of iso-stress data at temperatures above the nominal service temperatures and has been suggested as a preferred route, to the extrapolation of creep properties based on stress, by a number of workers (42)(43). The method correlates well with the Larson-Miller parametric plot of log (rupture-time) v. 1/(absolute temperature). However, even better correlation is found for some materials, notably for the low allow ferritic steels, by correlation with the Manson-Haferd (44) parameter log (rupture-time) v. absolute temperature.

Monkmann and Grant (45) have also shown that a correlation exists between log (strain-rate) and log (rupture-time) over a range of temperatures which appears to fit well for a number of materials, and in particular, the low alloy ferritic steels. Implicit in this relationship is that the creep rupture-time and deformation-rate should be controlled by the same mechanism.

One problem with this approach is that the results can be affected by oxidation effects such as section-loss and local grain boundary attack, these mechanisms generally becoming more severe with increasing temperature.
2.1.4.3 Life-fraction rule

Another frequently used life-time prediction technique is that using the life-fraction rule. This empirical rule states that

\[
\frac{t_s}{T_s} + \frac{t_t}{T_t} = 1
\]

(14)

where

- \(t_s\) is further life expected on changing from service to test conditions.
- \(t_t\) is length of exposure at service conditions
- \(T_s\) is the rupture-time of the virgin material under the test conditions
- \(T_t\) is the rupture-time of the virgin material under the service conditions

The technique requires that tests be conducted on material pre-exposed to service conditions. It therefore requires test specimens to be taken from plant components removed from service. Accelerated tests are then conducted on this material either on the basis of increased temperature or increased stress as discussed previously.

Work by Hart (46) and Cane & Williams (42) has shown that relatively good agreement with plant material rupture-lives can be obtained for ferritic materials using temperature accelerated test techniques whereas the accelerated stress technique leads to poor predictions. Additionally, they found that increased-temperature testing led to remaining life fractions which were relatively independent of the accelerated test duration. This is particularly advantageous since it means that relatively short tests can be conducted with confidence.

Although the life-fraction rule based on increased stress testing is not obeyed for ferritic steels, several authors (47)(48)(49) have found that it can hold for other materials such as austenitic alloys. The general conclusion to be drawn from these findings appears to be that the life-fraction rule holds for most materials for increased-temperature testing but that the validity of the increased-stress technique appears to be dependent upon the materials unstressed...
ageing behaviour. Where the creep strength of a material is reduced by unstressed thermal exposure then the life-fraction rule will not be obeyed by the increased stress approach. In the case of ferritic steels, this ageing influence relates to the thermally activated coarsening of precipitates.

One disadvantage of the technique lies in the fact that the life-fraction parametric expression contains a term for the "rupture-time of virgin material under service conditions", \( T_t \). Data for this parameter is rarely available and although estimates can be made, they generally lead to a loss in the predictive accuracy of the technique. The problem can be overcome, however, if test samples are removed from plant at two discrete time intervals and tested under accelerated conditions, in this way \( T_t \) can be eliminated from equation (14).
2.2 MULTIAXIAL CREEP ANALYSIS

Engineering components exposed only to uniaxial loading patterns are rarely found in practice. Instead one finds components exposed to combinations of tensile, bending and torsional loads, which interact to give complex multiaxial stress patterns. To predict material behaviour under these multiaxial systems requires the extension of the theoretical uniaxial equations to multiaxial conditions. Validation of these are then made by comparison with experimental results from test programmes designed to produce controlled multiaxial stress systems.

The following sections review the theoretical and experimental studies conducted in this field.

2.2.1 Deformation under multiaxial stress systems

2.2.1.1 Elastic stress analysis

Any general multiaxial loading system can be described in terms of direct stresses (stresses acting normal to a given plane) and shear stresses (stresses acting parallel to a given plane). If one considers a unit volume of material having a set of orthogonal directions x, y & z then three direct stresses and six shear stresses may be defined (fig.8.1). The notation here serves to define the plane and direction in which the stresses act (e.g. \( \tau_{xy} \) acts on the x-plane in the y-direction). It is found that the number of independent shear stresses can be reduced to three since \( \tau_{xy} \) and \( \tau_{yx} \) are equal in magnitude and said to be complementary. These stress components can be represented in matrix form, the system then being referred to as the "stress tensor".

\[
\begin{bmatrix}
\sigma_x & \tau_{yx} & \tau_{zx} \\
\tau_{xy} & \sigma_y & \tau_{zy} \\
\tau_{xz} & \tau_{yz} & \sigma_z
\end{bmatrix}
\]

The matrix has a further interesting property in its symmetry due to the complementary nature of the shear stresses.
If the previously referred to unit volume is rotated, there exists a unique position where the shear stresses on the cube faces reduce to zero. In this condition the faces are referred to as the "principal planes" and the stresses normal to these planes the "principal stresses" ($\sigma_1, \sigma_2, \sigma_3$). For this condition the stress tensor is written as

$$\begin{bmatrix}
\sigma_1 & 0 & 0 \\
0 & \sigma_2 & 0 \\
0 & 0 & \sigma_3
\end{bmatrix}$$

The magnitude of the principal stresses may be evaluated by the solution of the cubic equation

$$\sigma^3 - I_1 \sigma^2 + I_2 \sigma - I_3 = 0$$

(15)

where

$$I_1 = \sigma_x + \sigma_y + \sigma_z$$

$$I_2 = \begin{vmatrix}
\sigma_x & \tau_{yx} & \tau_{zx} \\
\tau_{xy} & \sigma_y & \tau_{zy} \\
\tau_{xz} & \tau_{yz} & \sigma_z
\end{vmatrix}$$

$$I_3 = \begin{vmatrix}
\sigma_x & \tau_{yx} \\
\tau_{xy} & \sigma_y
\end{vmatrix} + \begin{vmatrix}
\sigma_y & \tau_{zy} \\
\tau_{yz} & \sigma_z
\end{vmatrix} + \begin{vmatrix}
\sigma_z & \tau_{zx} \\
\tau_{xz} & \sigma_z
\end{vmatrix}$$

$I_1, I_2$ & $I_3$ are termed the "stress invariants" since their values are constant for a given stress system and independent of the original system of axes adopted.

The principal stresses themselves can be considered to comprise two components

i) a hydrostatic component $\sigma_m = (\sigma_i + \sigma_j + \sigma_k)/3$

ii) a deviatoric component $\sigma'_i = \sigma_i - \sigma_m$

Research carried out to evaluate the relative contributions of these two stress components to material deformation behaviour (50) has shown the hydrostatic component to be only responsible for a
change in volume, no change in shape occurring since no shear stresses exist on planes of any orientation. By contrast, it has also been shown, that the deviatoric components cannot give rise to a change in volume since their algebraic sum is zero. However, they do generate shear stresses on other planes and are therefore directly responsible for material deformation. It is often convenient, therefore, when considering material deformation behaviour to separate the principal stresses into their respective components.

By suitable substitution we can express equation (15) in terms of the stress deviators

\[ \sigma^3 - J_1 \sigma^2 + J_2 \sigma - J_3 = 0 \]  \hspace{1cm} (16)

where

\[ J_1 = 0 \]
\[ J_2 = (I_1^2 - 3I_2)/3 \]
\[ J_3 = (2I_1^3 - 9I_1I_2 + 27I_3)/27 \]

where \( J_1, J_2 \& J_3 \) are referred to as the "deviatoric stress invariants". An interesting expression is obtained when the expanded expressions for \( I_1 \& I_2 \) are inserted into the expression for \( J_2 \)

\[ J_2 = (1/6) \left[ \left( \sigma_1 - \sigma_2 \right)^2 + \left( \sigma_2 - \sigma_3 \right)^2 + \left( \sigma_3 - \sigma_1 \right)^2 \right] \]  \hspace{1cm} (17)

This expression will later be seen to be similar in form to the equivalent stress function proposed by von Mises (52).

2.2.1.2 Equivalent stress-strain relationships

The idea of an equivalent stress or strain was devised as a means of comparing stress or strain systems containing principal components of varying magnitudes. Two criteria are generally quoted in the literature and both are the extension of yield criteria, used to predict the yielding of material under multiaxial stress systems. The first, and the one most often quoted as the best representation of material deformation behaviour under elastic conditions (51), was developed by Maxwell, Huber, Henky & von Mises, although it is more
commonly accredited to von Mises (52).

The theory predicts that yielding occurs when the shear strain energy in a material exceeds a critical value equal to the shear strain energy at yield in uniaxial tension, or when

\[ y = (1/2)^{0.5} \left[ \left( \sigma_1 - \sigma_2 \right)^2 + \left( \sigma_2 - \sigma_3 \right)^2 + \left( \sigma_3 - \sigma_1 \right)^2 \right]^{0.5} \quad (18) \]

The locus described by this equation is a cylinder with axis normal to the octahedral plane. The section of the \( \sigma_1', \sigma_2' \) plane through this locus is shown in fig. 9. Von Mises postulated, that on further loading beyond the yield point, the cylindrical locus would expand proportionally. Based on this assumption the criterion may be expressed as a generalized equivalent stress criterion,

\[ \sigma_E = (1/2)^{0.5} \left[ \left( \sigma_1 - \sigma_2 \right)^2 + \left( \sigma_2 - \sigma_3 \right)^2 + \left( \sigma_3 - \sigma_1 \right)^2 \right]^{0.5} \quad (19) \]

the corresponding relationship considering strain being derived as

\[ \varepsilon_E = (2/9)^{0.5} \left[ \left( \varepsilon_1 - \varepsilon_2 \right)^2 + \left( \varepsilon_2 - \varepsilon_3 \right)^2 + \left( \varepsilon_3 - \varepsilon_1 \right)^2 \right]^{0.5} \quad (20) \]

such that the equivalent strain \( \varepsilon_E \) reduces to the uniaxial principal strain \( \varepsilon_1 \) under uniaxial loading conditions.

The second criterion, that due to Tresca (53), assumes that yielding occurs when the maximum shear stress exceeds a critical value, equal to the maximum shear stress in uniaxial tension.

\[ y = (\sigma_1 - \sigma_3) \quad (21) \]

The locus this criterion describes is a hexagonal cylinder inscribing the von Mises cylinder. Again the section of the \( \sigma_1', \sigma_2' \) plane through this locus is shown in fig. 9. With the same generalisation as for the von Mises criterion, the Tresca yield criterion may be extended for equivalent stress and strain.

\[ \sigma_E = (\sigma_1 - \sigma_3) \quad (22) \]

\[ \varepsilon_E = (2/3) (\varepsilon_1 - \varepsilon_3) \quad (23) \]
2.2.1.3 Stress and strain-rate relationships for creep

The earliest analyses of creep under multiaxial conditions were confined to steady-state or secondary creep behaviour.

Soderberg (54) developed a stress / strain-rate relationship under creep conditions based upon a number of assumptions. Among these were: the substitution of strain-rate for strain terms in the Levy - Mises flow rule

\[
\frac{\dot{\varepsilon}_1 - \dot{\varepsilon}_2}{\sigma_1 - \sigma_2} = \frac{\dot{\varepsilon}_2 - \dot{\varepsilon}_3}{\sigma_2 - \sigma_3} = \frac{\dot{\varepsilon}_3 - \dot{\varepsilon}_1}{\sigma_3 - \sigma_1}
\]

(24)

the coincidence of axes of principal stress and strain throughout the deformation; the constancy of volume during secondary creep;

\[
\dot{\varepsilon}_1 + \dot{\varepsilon}_2 + \dot{\varepsilon}_3 = 0
\]

(25)

initial material isotropy; and the power dependence of creep-rate on stress

\[
\dot{\varepsilon} = A\sigma^n
\]

(7)

By combining equations (24) and (25) he derived expressions for the principal creep-rates

\[
\begin{align*}
\dot{\varepsilon}_1 &= \frac{2C}{3}\left[\sigma_1 - 0.5(\sigma_2 + \sigma_3)\right] \\
\dot{\varepsilon}_2 &= \frac{2C}{3}\left[\sigma_2 - 0.5(\sigma_1 + \sigma_3)\right] \\
\dot{\varepsilon}_3 &= \frac{2C}{3}\left[\sigma_3 - 0.5(\sigma_2 + \sigma_3)\right]
\end{align*}
\]

(26a) (26b) (26c)

where, assuming the von Mises form for the equivalent stress and strain, \(C\) is defined as

\[
C = \frac{3}{2}\left[\frac{\dot{\varepsilon}\sigma}{E}\right]
\]

(27)
Thus
\[ \dot{\varepsilon}_1 = \left( \dot{\varepsilon}_E / \sigma_E \right) [\sigma_1 - 0.5(\sigma_2 + \sigma_3)] \quad (28a) \]
\[ \quad \text{etc.} \quad (28b) \]
\[ \quad \text{etc} \quad (28c) \]

He incorporated the creep aspect of the deformation by proposing an equivalent stress form of the Norton power law such that
\[ \dot{\varepsilon}_E = A \sigma_E^n \quad (29) \]
and substituted this into equation (28) giving
\[ \dot{\varepsilon}_1 = A \sigma_E^{(n-1)} [\sigma_1 - 0.5(\sigma_2 + \sigma_3)] \quad (30a) \]
\[ \quad \text{etc} \quad (30b) \]
\[ \quad \text{etc} \quad (30c) \]

Written in general form
\[ \dot{\varepsilon}_{ij} = f(J_2)^{(n-1)} S_{ij} \quad (31) \]

This result verifies that the deviatoric stress components are responsible for material deformation in creep as in plasticity.

Other authors, notably Marin (55) and Odqvist (56) derived almost identical solutions to that of Soderberg. Whilst accepting the strain-rate dependence of the function $J_2$, Bailey (57) considered the form of the deviatoric term to be inadequate in describing his results. Instead, he proposed a term of the form
\[ \dot{\varepsilon}_1 = A f(J_2)^m \left[ (\sigma_1 - \sigma_2)^{(n-2m)} - (\sigma_3 - \sigma_1)^{(n-2m)} \right] \quad (32) \]

Marin (58) has since correlated his results with both the Bailey and Soderberg models and reported little difference in the strain-rates predicted by both methods. He further suggested that the need for the added complexity in the Bailey equation may have arisen from the presence of initial anisotropy in the test material.

Many other authors have proposed theoretical relationships for stress and strain-rate behaviour under creep conditions. Kanter (59) and Nadai (60) suggest the inclusion of a viscosity term ($\dot{\varepsilon}$) in the
strain-rate equation. Difficulties are, however, realised in the
evaluation of the dependence of this parameter on stress state and
for this reason it has received little attention.

Rabotnov (61) generated his own flow rule based on the theory of
Volterra
\[ \gamma(\varepsilon_1) = (1+K^*).\sigma_1 \]  \hspace{1cm} (33)
where \( K^* \) is the Volterra integral operator.

This he incorporated with the von Mises equivalent stress and strain
relationships to successfully predict the multiaxial creep behaviour
of copper.

Belaiev (62) derived a relationship
\[ (\dot{\varepsilon}_{ij} - \dot{\varepsilon}) = \frac{(1+\gamma).((\sigma_{ij} - \sigma))}{2G} \]  \hspace{1cm} (34)
where \( \gamma \) is a function dependent upon creep strain history. Although
he incorporated the von Mises stress / strain criteria he considered
the volume change during creep to be elastic, being dependent on \( \gamma \)
(Poisson's ratio). This assumption can be shown to be in error from
density measurements made during creep (34).

Russanova (62) and Besseling (62) introduced the idea of relating
the multiaxial stress and strain-rate behaviour to microscopic
material mechanisms.

Russanova considered this in a rather general manner, evaluating
empirical relationships for three mechanism related observations:
lattice distortion; partial lattice recovery; and internal stress
relaxation.

Besseling derived more detailed relationships, by considering an
element of volume to be composed of sub-elements which showed
secondary creep and isotropic work hardening during creep
deformation.

Neither of these theories have been evaluated in relation to
experimental data and, as such, their validity remains questionable.
In addition, the complexity of the relationship derived by Besseling
could be considered to outweigh the mechanistic accuracy of the
analysis. Despite these criticisms, the model does attempt to link
the mechanistic and empirical approaches to multiaxial creep analysis, and as such is worthy of further attention.

Clough and Simmons (63) have examined the possibility of describing macroscopic flow and yield behaviour under multiaxial stress conditions on the basis of thermally activated dislocation motion. Due to the crystalline nature of materials, the plastic flow of single crystals is typically anisotropic due to dislocation glide being constrained to fixed crystallographic planes. On the macroscopic scale, however, deformation is observed to be isotropic due to the random orientation of the polycrystals.

The authors modelled this random orientation of the slip-planes in polycrystalline materials using a statistical randomness function and considered the material flow behaviour to be controlled by the rate at which dislocations were able to overcome barriers within the lattice. In order to generate a theoretical yield locus, yielding was considered to have occurred when the "multiaxial plastic power dissipation", \( \sigma \dot{\varepsilon} \), attained a prescribed critical value.

Their results demonstrates that under low temperature and/or high stress conditions the predicted yield locus resembles that of the Tresca yield criterion while at high temperatures and/or low stresses conditions a von Mises type yield locus is predicted.

Schwab (64) has more recently reported that although the predicted results are valid at these extreme conditions, as compared with experimental results, the model does not accurately predict material deformation behaviour under intermediate conditions of temperature and stress. He reports that this discrepancy arises from the inadequacy of the numerical solution technique, rather than from any inadequacy of the model.

Using a more appropriate technique, Schwab (64) predicts an extension of the von Mises criterion to lower temperature and/or higher stress conditions. He also considers material deformation behaviour under power and sinh law creep conditions and reports a further extension of the von Mises ellipse to lower temperature and/or higher stress conditions.

Prager (65) presents a derivation of a generalised plasticity stress / strain relationship based upon deviatoric stress terms, which he extends to creep conditions by incorporating a strain-rate term in place of the strain term.
For uniaxial conditions, he expresses strain-rate as a series of power functions of the stress deviators $S_{ij}$.

For multiaxial conditions, he replaces this with homogeneous expressions of the same order and making use of the Hamilton / Cayley relationship (66) he obtains

$$E = p(J_2, J_3^2).J_2 T + q(J_2, J_3^2).S$$

(35)

where $p$ and $q$ are polynomials in $J_2$ & $J_3$

$E$ is the strain tensor holding components of $E_{ij}$

and $T$ is the deviation of the square of the stress deviator $(S - (2/3).J_2.I)$

$$I = \begin{bmatrix}
1 & 0 & 0 \\
0 & 1 & 0 \\
0 & 0 & 1
\end{bmatrix}$$

Although this equation fulfils two of the assumptions of Soderberg's analysis, namely, the coincidence of axes of stress and strain and hydrostatic stress having no effect on material deformation, it does not fulfil the requirements of the Levy - Mises flow rule. Prager modified his relationship, therefore, to take this into account giving

$$E = f(J_2, J_3^2).[p(J_2, J_3^2).J_2 T + q(J_2, J_3^2).S]$$

(36)

He went on to consider how some commonly quoted plastic stress / strain relationships are in fact special cases of equation (36). With specific attention to Bailey's creep analysis, he showed that for the assumption of coincident axes of principal stress and strain to hold, the $(n-2m)$ term must have an odd value, although $m$ may be fractional. He correlated the $Af(J_2)^m$ term with $f(J_2, J_3^2)$ and showed that

$$[p(J_2, J_3^2).J_2 T + q(J_2, J_3^2).S]$$

reduced to

$$((\sigma_1 - \sigma_2)^{n-2m} - (\sigma_3 - \sigma_1)^{n-2m})$$
as a special case.

Johnson et al (67), in an effort to verify the correspondence of a von Mises equivalent stress criterion to material deformation behaviour under constant-stress primary creep conditions, called upon the theory of Ilyashim (68) which states that all points for all stress systems on a plot of log (octahedral stress) v. log (octahedral strain-rate) should lie on one common curve. A typical result of Johnson (fig.10.) shows a curve consisting of two nearly linear portions. Although each region correlated well with an equation of the form

\[ \varepsilon_{ij} = A(J_2)^p S_{ij} \phi(t) \]  

Johnson proposed a more general form to encompass the whole stress / strain range, namely

\[ \varepsilon_{ij} = \sum A(J_2)^p S_{ij} \phi(t) \]  

where \( \sum A(J_2)^p \) has the form

\[ A_1(J_2)^p_1 + A_2(J_2)^p_2 + \ldots + A_{(n-1)}(J_2)^p(n-1) + A_n(J_2)^p n \]

Johnson (66) also conducted an analysis in which he considered the effect of changing stress under multiaxial conditions. He proposed age-hardening and strain-hardening models as well as a combined-hardening theory, concluding that none satisfactorily represents the true material deformation behaviour under these changing stress conditions, even though the combined theory gave results nearest to the experiment.
2.2.1.4 The experimental realisation of controlled multiaxial stress systems

The choice of specimen geometry and loading pattern to produce multiaxial stress systems depends on a number of factors, such as

- the availability of material
- the relation of specimen geometry and loading to specific applications
- experimental difficulties involved in testing
- and the complexity of the stress / strain-rate analysis

Most of the work carried out in this field has utilized tubular geometries, loaded under internal pressure \(^{69}\)(\(^{70}\)). The reason for this is that these studies were designed to investigate the behaviour of industrially relevant piping systems under real service conditions, rather than to produce controlled multiaxial stress systems.

Some experimental programmes have, however, been conducted on thick \(^{71}\) and thin \(^{72}\) walled tubes under internal pressure, as well as combined internal pressure and axial load \(^{73}\), to investigate the more fundamental effects of stress ratio and equivalent stress, on deformation and rupture behaviour, in the positive principal stress quadrant (fig.11.).

Experimental problems do, however, arise from the testing of tubes under internal pressure, due to disturbances in the calculated stress profiles from end effects. An ingenious technique has recently been reported \(^{74}\), whereby this problem has been overcome by extending the tube out of the furnace and having the end features in the cold zone. Other problems are realised, however, due to the increased gauge-lengths of the specimens used in these experiments, in relation to the problems of machining such long gauge-lengths to the required dimensional tolerances.

An alternative, and widely used, mode of controlled multiaxial testing involves the application of a torsional load to create shear stresses in the specimen \(^{75}\) and hence a stress ratio in the second principal stress quadrant (fig.11.). Tubular specimens are usually
used here to minimise the effect of the variation in shear stress with specimen radius. The applied stress ratio can then be altered by the superposition of an axial load (76) or an internal pressure (77). Problems can, however, be encountered as a result of the onset of buckling instability which may limit the range of the stress ratios that can be applied (78).

Johnson et al (67) reported the use of flat cruciform specimens tested in balanced biaxial tension in an investigation into the effect of hydrostatic stress components on creep deformation. Some problems were encountered, however, concerning the satisfactory application of a uniform load over the specimen gauge-length.

Hayhurst (79) has also carried out experimental programmes using cruciform specimen geometries in order to determine the nature of the creep rupture locus in the tension-tension biaxial principal stress quadrant.

Hayhurst et al (80) and Hayhurst & Henderson (81) have conducted experimental programmes on axially loaded plates with central holes and notched bars respectively. The multiaxial stress systems were set up by the stress concentration effect of the notch and the hole and were evaluated by use of a finite element technique. These methods have the advantage of requiring relatively simple geometry specimens and less sophisticated test equipment than for the other techniques mentioned.

Many other geometries and loading patterns have been employed to produce multiaxial stress systems. These include the bending of thin-walled tubes (82) and the internal pressurization of spherical shells (83). However, these have been used largely to model specific applications.
2.2.1.5 Stress analysis of tubular geometries loaded under internal pressure and/or axial load

The most common technique for evaluating the principal stresses for tubes under internal pressure is the so called thin-wall formula. This analysis, based on the equilibrium of forces in the tube wall, assumes the radial stress component to be negligible as compared with the axial and tangential stress components, and predicts constant values of these through the wall thickness.

\[ \sigma_\theta = \frac{(pR_1)}{t} \]  
\[ \sigma_z = \frac{(pR_1)}{2t} \]

Another frequently used technique is the mid-wall theory. Here the mid-wall radius is substituted for the inside wall radius in the thin-wall formula.

\[ \sigma_\theta = \frac{(pR_m)}{t} \]  
\[ \sigma_z = \frac{(pR_m)}{2t} \]

Although these equations are completely empirical they have been found to give quite good correlation for tube stress rupture-times and are incorporated in many widely used design codes in the United States of America. This apparent correlation may be fortuitous due to the proximity of the skeletal and reference stress points to the mid-wall radius for tubular components. (see section 2.2.1.6)

The extension of this analysis to conditions of internal pressure and axial load presents no problem and can be handled by the superposition of the additional axial stress component on the original stress system set up by the pressure loading.

\[ \sigma_\theta = \frac{(pR_1)}{t} \]  
\[ \sigma_z = \frac{(pR_1)}{2t} + \frac{L}{(2\pi rt)} \]

Taira, Ohtani & Ishisaka (73) have conducted experimental and
theoretical analyses of thin-walled tubes stressed under both internal pressure and axial load. Using the von Mises equivalent stress criterion they conducted experiments at constant equivalent stress with a varying stress ratio ($\sigma_\theta/\sigma_z$).

Initial results showed considerable discrepancies when plotted as effective strain v. time, which the authors considered could be due to

i) the inadequacy of the von Mises equivalent stress function,

ii) the effect of initial material anisotropy,

iii) or the change of stress and stress ratio due to large deformations.

The authors considered option (iii) to be the largest single contributory factor to the discrepancies between the experimental and theoretical results and investigated its influence using a time iteration technique, taking account of changes in true stress due to specimen deformation. Theoretical equivalent strain v. time curves calculated in this way showed better correlation with the experimental results. 

By this same analysis, the authors also showed the variation of true stress and stress ratio with specimen deformation and the effect of initial applied stress ratio on this change in stress (fig.12.). Using the corrections for changing stress, theoretical axial and tangential creep strain v. time curves were calculated for the von Mises and Tresca criteria. The von Mises criterion gave the best correlation with the materials deformation behaviour under these conditions. Chubb and Bolton (84) have reported similar correlations with the von Mises criterion for their experimental work on thin-walled tubes under internal pressure and axial load.

Although thin-walled tubes are convenient from an analytical point of view, the geometries of tubes used in practical applications often have $R_i/t$ ratios less than 10:1 which are too small for the assumption of negligible radial stress to hold and under these conditions a thick-walled analysis must be undertaken. Dyson & Henderson (85) have also reported a value of 10:1 as being a representative limit for the assumptions of a thin-wall analysis to be valid.
The classical analysis for thick-walled tubes under internal pressure is that due to Lamé \(^{(86)}\), where for a closed end cylinder the principal stresses at a radius, \(r\), are given in cylindrical polar coordinates as

\[
\sigma_\theta = \frac{pR_1^2}{(R_0^2-R_1^2)} \cdot \left[ \frac{R_0^2}{1+\frac{r^2}{R_0^2}} \right] \tag{45}
\]

\[
\sigma_r = \frac{pR_1^2}{(R_0^2-R_1^2)} \cdot \left[ \frac{R_0^2}{1-\frac{r^2}{R_0^2}} \right] \tag{46}
\]

\[
\sigma_z = \frac{pR_1^2}{(R_0^2-R_1^2)} \tag{47}
\]

Unfortunately these preceding analyses do not take into consideration the phenomena of redistribution of stresses during creep. The form of this redistribution can be evaluated for the steady-state creep regime by the incorporation of the Norton power-law into the thick-wall Lamé analysis to give the following differential equation

\[
d\sigma_r = \frac{D \cdot dr}{r^{(2+n)/n}} \tag{48}
\]

When solved for boundary conditions, this gives the stress distributions in the tube wall as a function of radial position, \(r\), and the Norton stress exponent, \(n\), (fig.13.).
The time to attain a steady-state stress distribution has been investigated by a number of authors. Taira, Koterazawa & Ohtani (87) have shown that full redistribution occurs within the first few hours after loading, while Johnson et al (71) and Bailey (57) have reported results of the order of $10^3 - 10^5$ hours. Additionally, Taira et al conducted X-ray analyses of their tube wall sections after exposure to determine the residual stress levels. By this method they reported good agreement with their theoretical analysis for stress redistribution.

Hayhurst and Henderson (81) have investigated the problem of redistribution times to steady-state creep stress distributions using circular and British Standard notched bar specimens. They conclude that redistribution times are high in structures where locally high values of $\sigma_1/\sigma_E$ exist. Material parameters and the form of the function of time term were also found to have an effect.

Unlike the preceding analyses, the analysis of thick-walled tubes exposed to axial and internal pressure loading in the presence of stress redistribution cannot be treated by the method of superposition since the additional axial load compromises the plane strain boundary condition for internal pressure loading

$$\sigma_z = 0$$

As a result, a closed form analytical solution of the system is
not possible. Finnie (88) has, however, derived a numerical solution by which the derived differential equation may be solved for steady-state distributions of radial, axial and tangential stresses.

The Lamé and creep redistributed thick-walled tube analyses considered till now have all assumed an infinitesimal theory of strain, whereby the strains have been considered small in relation to the initial specimen geometry and are therefore not considered to influence the initial stress system.

Although McGregor, Coffin & Fisher (89) developed equations for the plastic flow of thick-walled cylinders, based on the finite strain theory, it was Rimrott (90) who applied them to the creep regime. He presented an analysis predicting the stress distribution through the wall thickness of a thick-walled tube, based upon a known inside diametral strain.

This analysis was latter modified by King and Mackie (91) to consider a time-hardening model, equations being developed for the more practically useful case of a known outside rather than inside diametral strain.

2.2.1.6 Reference stress technique

The reference stress technique is an approximate method for reducing the effect of uncertainties in material data for the prediction of component creep behaviour under multiaxial and/or changing stress conditions, to a simple uniaxial test conducted at the reference stress.

Schulte (92) observed that for stress redistribution of beams, a cross-over point in the elastic and stationary creep stress distributions occurred. By conducting a uniaxial test at this stress he was able to make good approximation to the deflections of beams under bending.

Schulte had in fact recognised the existence of a "skeletal point" which incidentally happened to coincide with the reference stress point as a result of the geometry of his specimens. This is not a general identity for the reference stress point.

The reference stress concept was first suggested by
Soderberg (93), developed by Anderson (94) and later refined by Mackenzie (95). They reasoned that, by re-arranging the expression for the deflection of, say, a cantilever beam, the deflection could be expressed as a function of three independent terms

i) the Norton stress exponent, \( n \);

ii) the beam geometry; and

iii) the creep strain evaluated at the so-called reference stress.

\[
y = F(n).L^2 \cdot \dot{\varepsilon}_c(\sigma_0) \\
\]

(53)

The interesting observation was made that the Norton \( n \)-value term, was relatively insensitive to the value of, \( "n" \).

Sim (96) recognised that, by forcing the solution at two typical \( n \)-values for high temperature metals (eg. 3 and 9), the \( n \)-value term could be made even more insensitive to \( "n" \).

The main contribution of the work of Sim has been the extension of the idea of reference stress to conditions of varying stress. He has shown that complex component behaviour can be described by a single uniaxial test conducted with a loading history equivalent to that of the real component.

Fairbairn and Mackie (97) evaluated this idea for an internally pressurised thick-walled tube. They calculated the increasing stress history of the reference stress by a finite difference technique similar to that of Taira (73) and then conducted a uniaxial experiment under the same stress history conditions. In this way they reported good correlation between experiment and theory - certainly better than approximation of the creep behaviour, under varying stress conditions, using strain- or time-hardening models.
2.2.1.7 Computer techniques in creep

The rapid development of computer systems over the past decade offers great possibilities for the theoretical analysis of creep behaviour of materials. Specifically, it has permitted the development of two types of techniques, i.e. finite difference / iteration techniques and finite element techniques.

Finite difference techniques allow the rapid solution of such problem formulations as those of Taira (73) and Rimrott (90) where the specimen geometry is continually updated at each time increment or where the creep strain-rate is modified for each successive increment.

The finite element technique allows the solution of stresses and strains for complex-shaped structures. The structure is modelled by a mesh of elements which is then solved by considering the compatibility and equilibrium of each element with its neighbours as it deforms, according to a predefined deformation algorithm which is dependent on the element type selected. The technique was initially developed for solutions under static elastic conditions, but recent developments now allow analysis of a variety of material deformation mechanisms to be made such as plasticity and creep.

The creep module within the finite element package contains a combination of the iterative and finite element techniques to allow the prediction of stress and strain distributions after a given exposure time. The time interval used in such iterations has been found to have sensitive effects on the accuracy and efficiency of the eventual solution. As a result, most commercial systems have a built-in optimisation routine based upon a permissible increase in stress resulting from the strain accumulation over the previous time increment.

As already mentioned the primary advantage of the finite element aspect of the analysis is that it allows the analysis of complex-shaped components, taking into account the effects of stress concentrations on otherwise simple geometries.

Despite the fact that the finite element creep solution uses only the original elastic stiffness matrix, the solution times for structures of even small "wave-fronts" are long, due to the number of iterations required. In such cases, costs can often be of the order
of 100 times those of an equivalent elastic solution.

It has been possible, however, for the cases of some simplified geometries (eg. thin axisymmetric shells), to develop specialised packages for creep analysis (98). In such packages, the only difference is the method by which the stresses are evaluated, the creep routine remaining unchanged.

2.2.2 Creep rupture under multiaxial stress systems

Although uniaxial parametric extrapolation techniques are available for rupture-time in relation to stress, the representative controlling function of stress on rupture-time under multiaxial conditions is far more complex. This often requires the microstructural damage mechanisms within the material to be taken into consideration.

Siegfried (99) was the earliest author to take this approach. He proposed that the maximum deviatoric stress component governed transgranular fracture whilst the hydrostatic stress component seemed to be related to intergranular failure.

Little more work was done in this field until Johnson's (100) extensive study in the area. He initially set out to verify the work of Siegfried, but he found it to be generally invalid. Later, he developed his own model based on experiments using molybdenum steels, copper, aluminium and magnesium. He concluded, from rupture-time and metallographic results, that the maximum principal stress was the controlling function of stress on rupture-time for materials exhibiting a gradual build up of metallographic damage prior to failure, be it W- or r-type, whilst an octahedral function of stress tended to hold for materials exhibiting little or no cracking or cavitational damage prior to failure.

Browne, Lonsdale & Flewitt (101) have verified Johnson's findings through their work on two ferritic steels exposed to conditions of internal pressure and axial load. One steel, a 12%Cr-Mo-V-W alloy, failed in a ductile mode with little or no cavitation. This they found to correlate well with a function of octahedral stress. A second steel, a 1%Cr-Mo alloy, exhibited intergranular failure due to
grain boundary void formation; in this case the rupture-time correlating better with a function of maximum principal stress.

Browne et al, working on their ferritic steels, and Chubb and Bolton (84), working on 316 stainless steel, found that cavitation damage under multiaxial stress conditions occurred mainly on boundaries orientated perpendicular to the maximum principal stress direction.

Chubb and Bolton also investigated the observed variation of cavitation density with radial position for their torsional specimens. They considered the effects of distributions of stress and strain-rate and found best correlation with strain-rate, as have other authors (102)(103). When a superimposed axial stress was applied to the torsional stress system, the previously cavity-free centre portion of the bar showed a concentration greater than would have been expected for the same strain-rate under purely torsional conditions. They considered this to be evidence of a stress state dependence on cavitation nucleation and growth.

Another area which Johnson (100) investigated was the effect of hydrostatic stress component on rupture-time, for which he found no clear correlation. Browne, Lonsdale & Flewitt noticed, however, that concentrations of voids in the walls of their tubes showed a similar distribution to the hydrostatic stress component, being a maximum at the outside diameter. This suggested a correlation of hydrostatic stress with void nucleation and growth, and therefore rupture-time.

Other authors (104)(105) have also found that hydrostatic stress can have quite significant effects on the nucleation and growth of voids. Where creep failure is dominated by such mechanisms, then, the hydrostatic stress component would be expected to have some influence on the multiaxial creep rupture criterion.

Abo El Ata and Finnie (106), in their study of the effect of multiaxial stress state on the creep rupture behaviour of tubular components, suggested that creep crack initiation and propagation might be controlled by independent factors and that this might help in obtaining a more general model for creep rupture, as in the case of fatigue. They suggested that crack nucleation would be controlled by a von Mises type stress criterion, while crack propagation would be controlled by a maximum principal stress criterion.

From their results, they proposed that the ratio $\sigma_E/\sigma_{\text{max}}$ was a
useful parameter for predicting the creep rupture behaviour under multiaxial stress conditions. They found that low values of the parameter led to little crack nucleation, instead failure occurring by the rapid propagation of single cracks. For high values of the parameter, many small cracks were found to be nucleated, these propagating more slowly due to the reduced stress concentration associated with each crack.

Hayhurst (79) has conducted an investigation of the multiaxial fracture behaviour of an aluminium alloy using cruciform biaxial tensile specimens. He proposes that creep rupture behaviour of a material can be described by a linear combination of the maximum principal stress, the first principal stress invariant and the second deviatoric stress invariant.

\[ t = A[\alpha \sigma_1 + \beta I_1 + \sigma J_2^{0.5}]^{-\nu} \]  

(54)

By re-arranging this equation and normalising the parameters with respect to time and stress he derived

\[ T = [\alpha + \beta (a^2 + a + 1)^{0.5} + \gamma (a^2 - a + 1)^{0.5}]^{-\nu} \sigma_1 \]  

(55)

where \( a \) is the biaxial stress ratio and

\[ \alpha + \beta + \gamma = 1 \]

This is an entirely general multiaxial stress rupture criterion with hydrostatic, equivalent and maximum principal stress loci being described by values of \( \alpha, \beta \) & \( \gamma \) of \((0,1,0)\), \((0,0,1)\) and \((1,0,0)\) respectively. Values of the constants \( \alpha, \beta \) & \( \gamma \) can be evaluated by conducting experiments at three different stress ratios, \( a \), the generality of the equation allowing the description of mixed character rupture locii e.g. \( \alpha = 0.25 \); \( \gamma = 0.75 \).

Although many approaches have been proposed for the prediction of the multiaxial creep rupture behaviour, the validity of these can only be verified through a comparison of the model predictions and results obtained from controlled series of experiments conducted under multiaxial stress systems.
2.3 METALLURGY OF ALLOY 800

This section of the literature draws together and summarises the major research carried out to understand the metallurgy of Alloy 800 and its generic grades.

Alloy 800 is an iron - chromium - nickel ternary alloy with small alloying additions of carbon, titanium, aluminium and silicon. By reference to the specific compositions of the various available grades of Alloy 800 (table.1.) and the superposition of these compositions on the Fe-Cr-Ni ternary phase diagram (fig.14.) it can be seen that all grades are fully austenitic.

The alloy was for many years thought to be a solid solution strengthened alloy. However, today it is considered that following heat treatment it is a meta-stable alloy with strengthening being achieved at high temperatures by titanium carbide and chromium carbide dispersion hardening mechanisms and at lower temperatures (500 - 600°C) by gamma prime precipitation hardening.

2.3.1 Carbide precipitation

Titanium and chromium, both present in the alloy, have a strong affinity for carbon and compete to form their respective carbides. Chromium can form a range of carbides from Cr$_3$C$_2$ through Cr$_7$C$_3$ to Cr$_{23}$C$_6$ dependent upon the matrix activity of carbon. In Alloy 800 the very low concentration of carbon (< 0.1%) determines that only Cr$_{23}$C$_6$ is stable under these conditions. This is reinforced by reference to the Ellingham diagram for carbides (fig.15.) which indicates that Cr$_{23}$C$_6$ has the most negative free energy of formation, $\Delta G_f$, of the three carbides. However whether TiC or Cr$_{23}$C$_6$ should form preferentially under any set of conditions is governed by the thermodynamic and kinetic behaviours of the species within the system. Under equilibrium conditions, TiC is thermodynamically more stable, but tends to precipitate through an intragranular homogeneous mechanism, the kinetics of which are slower than the more typical heterogeneous precipitation of Cr$_{23}$C$_6$ on grain boundaries and upon dislocations, and for this reason Cr$_{23}$C$_6$ often forms preferentially.
The behaviour is further complicated by the specific effect of temperature. A clearer understanding can be achieved by reference to the carbide TTT curves for the alloy (fig.16.). At temperatures above 750°C the curves show that TiC is the most likely carbide to form. For this reason it can be considered that it is the titanium that controls the high temperature solubility and combination of carbon.

Degischer et al. (107) present an alternative result for the carbide stability diagram (fig.17.) which shows that no incubation time to carbide precipitation is needed and that both regimes are shifted to higher temperatures than proposed by Orr (108).

To fully understand the titanium-carbon-TiC behaviour at the heat treatment temperature it is best to examine the TiC solubility curves at 980°C and 1150°C. These curves indicate that the Ti:C ratio and the absolute carbon content are both controlling factors on the proportions of dissolved carbon and precipitated TiC. This has to some extent been disproved by Tavassoli and Colombe (109), who have shown that the specification of high Ti:C ratios in an effort to reduce the dissolved carbon content by promotion of TiC precipitation is not a sufficient condition for the elimination of Cr23C6 precipitation at the service temperature. A more specific indication of the effect of Ti:C ratio on the remaining dissolved carbon concentration is given in table.2. (taken from work by Orr (108)).

The absolute carbon content is also of significance in relation to the grain coarsening temperature of the alloy during heat treatment. Although TiC precipitation and prior cold work also have an effect, the variation of the grain coarsening temperature with carbon content has been shown to follow the trends indicated in fig.18. (108).

2.3.2 Heat treatment

The purpose of heat treatment of alloys is usually to achieve optimum and consistent mechanical properties for a material after fabrication. In Alloy 800 these properties are mainly controlled through the material's grain size. It is known that, for applications where improvements in creep behaviour are required, a large grain
size, and hence a reduced grain boundary surface area, is beneficial. Grain boundaries are detrimental since sliding of suitably orientated boundaries is thought to be a major high temperature deformation mechanism. In addition, grain boundaries act as sinks and preferred diffusion paths for vacancies, the agglomeration of which can influence creep fracture.

By contrast a material with a small grain size can show improvements in corrosion resistance. Alloy 800 is susceptible to Cr$_{23}$C$_6$ carbide precipitation at grain boundaries and dislocations and owing to the high chromium content of the precipitates a reduction in the matrix chromium concentration in the grain boundary adjacent regions consequently occurs. This can have serious consequences for the corrosion resistance of the alloy, since below a matrix concentration of approximately 12-13%, the protective chromia oxide is discontinuous in oxidising environments, resulting in a reduction in corrosion resistance. In a small grain sized material, rediffusion of chromium from within the grains to the depleted grain boundary adjacent regions is rapid since the diffusion paths are small. Consequently, the loss in corrosion resistance is markedly reduced.

In considering the heat treatment to be applied to any grade of Alloy 800 it is best to consider the specific conditions relevant to its use and the composition of the alloy. The effect of heat treatment on three commonly encountered compositions are as follows.

2.3.2.1 Alloy 800H minimum composition (0.06% C 0.15% Ti)

With this material, it is desired to produce a large grained material with a high concentration of dissolved carbon so that at service temperatures this will allow the heterogeneous precipitation of Cr$_{23}$C$_6$ chromium carbide on grain boundaries and dislocations. As already indicated, a large grain size material exhibits beneficial creep properties. A little grain boundary Cr$_{23}$C$_6$ precipitation can be beneficial in reducing grain boundary sliding.

These properties are achieved by the heat treatment of the material at 1150°C for 1 hour, followed by a water quenching to room temperature. Fig.18. shows that this temperature exceeds the grain
coarsening temperature for this carbon concentration. In addition, the small volume fraction of TiC, predicted by the solubility curves at this temperature (fig.19.), is insufficient to raise the grain coarsening temperature to above 1150°C.

Care should be taken, however, since results of work by Persson (123) have shown that there is no benefit in creep properties for an average grain size in excess of 2-3 ASTM (180-127 um). Rather it results in a reduction in creep rupture ductility.

2.3.2.2 800H maximum composition (0.1%C 0.59%Ti)

Due to the wide range of Alloy 800H composition specification it is possible at the top end of the composition range to realise a material with the above carbon and titanium concentrations.

With the same heat treatment cycle as before a higher volume fraction of TiC is retained at 1150°C (fig.19.). This, as explained, can raise the grain coarsening temperature by virtue of the TiC precipitates pinning the grain boundaries, resulting in a fine grained variety of 800H.

2.3.2.3 Alloy 800 nominal composition (0.02%C 0.3%Ti)

With Alloy 800 the intension is to produce a fine grained material with little dissolved carbon, in order to limit the extent of in-service Cr23C6 precipitation.

As can be seen, the heat treatment temperature of 980°C is below the grain coarsening temperature for this carbon concentration and no coarsening should therefore be expected in the short term (fig.18.). In addition the solubility curves indicate a relatively high retained volume fraction of TiC which will further raise the grain coarsening temperature. The result is therefore a fine grained material with a high concentration of TiC.
2.3.3 In-service ageing behaviour

2.3.3.1 Chromium carbide and titanium carbide

As already indicated, the in-service carbide precipitation is a function of alloy composition and heat treatment cycle in relation, specifically, to the concentration of carbon in solid solution. For Alloy 800H there is a high concentration of carbon in solution which is available for precipitation. By reference to fig.16, it can be seen that at service temperatures below 750°C $\text{Cr}_2\text{C}_6$ forms preferentially, this occurring almost entirely on grain boundaries and dislocations.

Despite the problems of carbide precipitation in relation to corrosion properties, many workers have attributed improvements in the time-independent and time-dependent mechanical properties to both $\text{Cr}_2\text{C}_6$ and TiC precipitation.

La Malfa (111) investigated a number of melts of general 800 and 800H compositions and the effects of various heat treatment cycles upon these. In many cases he was able to change the morphology and concentration of the $\text{Cr}_2\text{C}_6$ and TiC precipitates in the final microstructure. He concluded that fine intergranular $\text{Cr}_2\text{C}_6$ precipitates have a more beneficial effect on creep strength and ductility than larger ones. In addition, he found that, for conventional 800H grade compositions and heat treatment cycles, the volume fraction of ($\text{Cr}_2\text{C}_6$ + TiC) has no deleterious effect on creep ductility up to carbon concentrations of the order 0.1%.

Bassford and Rahoi (112) considered that above the gamma prime solvus the higher creep rupture properties were entirely due to $\text{Cr}_2\text{C}_6$ and TiC precipitation. In fact, Singhal and Martin (113) considered TiC to play the major role above 800°C.

Degischer et al (107) carried out an analysis of the ageing behaviour of Alloy 800 at 900°C, and their results are presented in fig.17. They found that during ageing, $\text{Cr}_2\text{C}_6$ and TiC precipitated in the austenite matrix with the number and distribution of precipitates being dependent on the ageing time. After 10 hours exposure, precipitation was fairly evenly distributed throughout the matrix, however, after 100 hours, intensification of the precipitation was noticed. TiC precipitation was found to be
predominantly intragranular in nature with the Cr$_{23}$C$_6$ being largely intergranular. These Cr$_{23}$C$_6$ precipitates were also found to have grown into the austenite grains due to the low lattice mismatch between the phases. After 1000 hours, the TiC precipitates were seen to coarsen and small amounts of sigma-phase and G-phase (Ti$_6$N$_{16}$Si$_{17}$) were reported.

Tambuyser and Franck (114) conducted a similar investigation into 800H at 1000°C, but also investigated the effect of stress on ageing behaviour. They showed that the upper limit for Cr$_{23}$C$_6$ precipitation was approximately 800°C and not 900°C as reported by Degischer. However, and more importantly, they showed that the same coarsening of TiC precipitates on dislocations occurred with ageing, and they identified the mechanism as Oswald ripening, a mechanism in which small precipitates dissolve and reprecipitate onto larger ones, leading to precipitate coarsening. They also showed that this mechanism was accelerated by applied stress.

2.3.3.2 Gamma prime

It was originally considered that at lower temperatures the alloy gained its strength by Cr$_{23}$C$_6$ carbide precipitation and solid solution strengthening. With the advent of high power electron microscopy, it was recognised that in the operating temperature range of 550°C-750°C some strengthening was produced by the precipitation and ageing of small gamma prime (Ni$_3$Al) precipitates. Further investigation of these precipitates by Tavassoli & Colombe (109) revealed that titanium could also form similar precipitates with a better strengthening effect than aluminium.

A number of authors also investigated the idea of optimizing concentrations of aluminium and titanium, temperature and ageing time for the maximum strengthening effect of gamma prime on 800H compositions. Persson & Egnel (115) and Cozar & Rouby (116) quote (Al+Ti) levels in the range 0.6-0.7% as a threshold concentration for precipitation while Bassford & Rahoi (112) conclude that a 0.45% titanium level gives optimum properties at 680°C after 1000 hours. Orr pointed out that exposure over 750°C resulted in rapid
precipitate coarsening and loss in mechanical properties.

Orr (108) and Persson (110), examining the effect of gamma prime on mechanical properties concluded that precipitation under creep conditions effectively accelerates the onset of secondary creep, giving rise to lower values of the Norton stress exponent. This, in turn results in a slight increase in creep rupture strength and life, but at a significant expense of the rupture ductility.

2.3.3.3 Sigma phase

Due to gamma prime precipitation and in particular, the precipitates of high nickel content, local depletion of the lattice nickel concentration can occur, resulting in a shift across the \( \gamma / (\gamma + \sigma) \) phase boundary (fig.14.) such that sigma phase can precipitate.

This sigma phase is a complex iron-chromium intermetallic phase which is commonly found in stainless steels (eg. 18/8 grades) after long service exposures. In these materials it is particularly detrimental as it usually forms at grain boundaries and is responsible for embrittlement of the alloy. Work by Orr (108) has shown that the kinetics of formation are fortunately slow, exposure times in excess of 25,000 hours at 600\(^\circ\)C-700\(^\circ\)C being necessary for its formation. He has also shown that the precipitates may form in acicular or spheroidal morphologies and he emphasized that in experiment, as opposed to service, neither had been shown to cause reduction in ductility or impact properties. This is most likely to be due to the small volume fractions present.
2.4 CREEP PROPERTIES OF ALLOY 800H

By reference to the previous metallurgical sections it can be seen that the metallurgy of Alloy 800 is particularly sensitive to small changes in chemical composition, heat treatment and in-service ageing time. The influence of these parameters on the creep behaviour of the alloy must therefore be evaluated and where possible quantified.

To examine these influences effectively requires a large volume of experimental data. This quantity of information is rarely available except for a few industrially relevant alloys. Fortunately, Alloy 800 is one such material. Experimental data has been collected on the alloy for over a decade, largely initiated through EEC coordinated research programmes. More interestingly, this data, along with other published data for the alloy, have been collected and stored in the "Alloy 800 Data Bank" at Petten.

This Data Bank is a computerised on-line materials data retrieval system allowing experimental results to be selected and displayed based on user defined parameter limits such as composition, test temperature, test environment, etc.

Using this Data Bank, data requests were first made for the full compositional range of Alloy 800H, based on search profile (1), shown in tables 3 & 4. Results of these searches are presented in stress rupture, elongation to rupture and minimum creep-rate form (figs.20 to 26).

A large number of stress rupture data were found at all temperatures fig.20, with the exception of 850°C where only three data points were retrieved, this result demonstrating the need for the generation of our own experimental data at this temperature. At the temperatures 800, 900 and 1000°C good linear correlations were found for the stress rupture results, however, at 700°C a much wider range of scatter was observed. At this temperature the alloy still retains some strengthening from gamma prime precipitation and ageing, the test results being more sensitive to the specific heat treatment cycle than at higher temperatures where strengthening is largely achieved from carbide precipitation and solid solution strengthening mechanisms.

A much greater degree of scatter was found for the elongation to

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rupture results. Such behaviour is quite common in creep elongation data, however, scatter was again found to be greatest for the 700°C series of data for the same reasons as described previously. It was also difficult to distinguish between results at 900 and 1000°C. In order to obtain a clearer idea of the trends of this data, linear regression analysis of the data was performed at each temperature (fig.27.). This figure shows a gradual decrease in rupture elongations with increasing temperature with a slight inversion at 1000°C for rupture-times of less than 500 hours.

The log (strain-rate) v. log (stress) results at 900 and 1000°C (fig.26.) were interesting in that the curves consisted of two linear portions, the change in gradient occurring at a strain-rate of \(10^{-4}\) h\(^{-1}\). Results at 700 and 800°C either did not exhibit this change in slope, or the data did not extend to low enough minimum strain-rates for it to be resolvable.

In an attempt to evaluate the influence of the carbon, aluminium and titanium concentrations on the mechanical behaviour of Alloy 800H the search profile was redefined to examine the elongation at rupture properties at 900°C in narrower compositional ranges. Search profile (2) is shown in tables.3.&.4.

Results of this search showed no conclusive trends with increasing aluminium and titanium concentrations (fig.28.). Additionally, for carbon concentrations, other than the 0.060-0.0699 range, results were too few in number for a realistic comparison to be made (fig.29.). On this point Orr (108) reports that increased aluminium and titanium concentrations result in increased rupture strengths but reduced rupture elongations. Additionally he reports that aluminium has a stronger influence in this direction than titanium. Pearce (117) has shown, during exposure to service environments, that aluminium and, to a lesser extent, titanium and silicon segregate to oxidised grain boundaries, the aluminium being found in the form of \(\text{Al}_2\text{O}_3\) in these grain boundary regions. It is possible therefore that alumina plays a role in weakening the grain boundaries resulting in a loss in ductility.

Increasing the carbon concentration is also reported to give an increase in creep rupture strength but a reduction in the ductility (118). This loss in ductility most likely arises from the increased \(\text{Cr}_2\text{C}_6\) precipitation at the austenite grain boundaries during high
temperature exposure.

The failure to resolve these influences of aluminium, titanium and carbon on the material creep properties from the Alloy 800 Data Bank data may well be due to the overshadowing influence of other less clearly defined variables such as the prior manufacturing and strain history of the material or the precise experimental procedure employed during mechanical testing.

Finally, the search profile was narrowed down still further in an attempt to recover data having the same chemical composition, heat treatment cycle, and test temperature as the experimental material search profile (3) (tables.3.&.4.). Unfortunately no data was found within this profile.

Examination of the test material composition (table.1.) shows it to have a relatively high concentration of carbon, aluminium and titanium in comparison to the Alloy 800H composition limits, typical alloy compositions being generally found towards the lower end of the compositional range. The reason for the absence of data within these limits may be simply due to the fact that material of this composition is rarely produced and therefore seldom available for testing, particularly at these high temperatures.
3.1 SMALL SPECIMEN EXPERIMENTS

It was initially planned to use the uniaxial solid bar results of Burgel (119) and other authors, from the Alloy 800 Data Bank at Petten, as the base line uniaxial data upon which to compare the uniaxial tube results. However, it soon became clear, from the literature and Data Bank surveys, that the creep properties of the test material 800H and especially the elongations to rupture, were particularly sensitive to small changes in composition, grain size and fabrication route. It therefore became necessary to manufacture and test small specimens cut from the walls of the tubes to establish the baseline data for comparison with the later tube experiments.

Preparation of the small specimens was conducted by first cutting rough square section bars from the previously solution treated tube material and then machining these to their near final dimensions. The end pieces were rough machined from larger stock 800H material and assembled and electron beam welded to the bars as shown in fig.30. The assembled specimens were then returned to the workshop for final machining of the gauge-length and end features.

3.1.1 Constant-load

These experiments were initially conducted in two Mayes 15:1 loading ratio lever arm creep machines. However, due to the small specimen diameters (2-4mm) difficulties were experienced in accurately setting up the experiments to the low loads desired and work was therefore transferred to two dead load testing frames where the loads could be set up more precisely.

Temperature control was achieved using a Eurotherm PID control system in association with a Mand 3000 watt 3-zone furnace. This control system was capable of giving temperature control to ± 1°C as measured. Work by Desvaux (120) at ERA has shown that for Pt/10%PtRh thermocouples operating at 850°C the output signal from the couple
falls off by about 30 micro volts (3°C) after about 1000 hours exposure. This error can be offset by recalibration of the thermocouples before each experiment. Unfortunately, no temperature calibration bath was available in this temperature range and instead the hot junction of the thermocouples were re-welded before every experiment and the thermocouples completely renewed after 2000 hours exposure. The loading train consisted of the specimen, screw threaded into two Nimonic 80A stub pull-rods by M10 screw threads, two Nimonic 80A ceramic pinned universal joints, and two longer Nimonic 80A pull rods which extended out of the furnace. (fig.31.)

Initially, a series of creep rupture tests (ST experiments) were performed at 900 and 1000°C on 4mm diameter and 30mm gauge-length specimens, from the Petten designation Batch I material, this having a composition different from the actual test programme material (Batch II). The compositions of these two materials are given in table.1. These experiments were initially intended as proving tests on the specimen design, however, the rupture-time and rupture-elongation results (figs.32.,&.33.) were recorded for comparison with the later results from the Batch II material. Later, specimens with extensometer profiles on the end pieces were machined from this Batch II material (fig.34.), and a series of continuous extensometry experiments (MS experiments) performed in an extended temperature range from 850 - 1000°C. The continuous strain v. time results for these are shown in figs.35.,36.&.37. An extensometry system (fig.38.), similar in design and utilising the same materials as for the TT experiment extensometry (see later) was used to measure the elongations of these specimens.

In order to investigate the effects of specimen diameter on creep behaviour, a single experiment (VMS experiment) was later conducted on a 2mm diameter, 10mm gauge-length specimen (fig.34.) at 42 MPa. This result is shown in fig.39. In order to test this specimen, a special light weight specimen string was constructed within the existing dead load testing frame. The specimen elongation was measured by the movement of the bottom pull rod relative to a fixed cross bar on the testing frame. This measuring system was chosen on the assumption that the contribution from the deflection of the specimen loading train to the total measured strain would be negligible in comparison with the specimen elongation. This
assumption was confirmed by measuring the specimen's elongation at fracture on removal from the rig and comparing this with the maximum strain recorded by the extensometry.

Initially, data acquisition of LVDT outputs and temperature were made using a four pen chart recorder. However, an increased measuring resolution was later achieved by use of a small digital voltmeter / scanner system in association with a Hewlett Packard HP 85 personal computer.

The specific test parameters associated with each of the above mentioned experiments are shown in table 5.

3.1.2 Constant-stress

Due to the continuously reducing cross section areas of the test specimens, as a result of axial deformation during creep, specimens in a constant load experiment are exposed to an increasing stress throughout their exposure. All the available mathematical analyses, on the other hand, are based on constant stress loading of specimens. For this reason, two shorter series of experiments were conducted under constant stress conditions (CS and ERA experiments) to obtain base line data for these mathematical analyses.

Due to the lack of facilities in the group, testing in this field was partly conducted on a machine of a sister department at the centre using 4mm diameter, 30mm gauge-length specimens (fig.34.) and then later at ERA Technology, Materials Division, Leatherhead, using 4mm diameter and 50mm gauge-length specimens (fig.34.).

Work at Petten was conducted on a 5 tonne ESH screw driven creep machine with an analogue constant stress computer to evaluate the change in applied load with time, while the equipment at ERA consisted of a conventional lever arm creep machine capable of reducing the load during the experiment via the movement of a motor driven jockey weight along the lever arm.

Initial results from the ESH machine showed some considerable irregularities in comparison to the previously obtained constant load data. To investigate this further two additional tests were conducted on the machine using the same CS specimen type but under constant
load conditions (CLB experiments).

The strain v. time results for the above mentioned experiments are shown in figs.40.&.41. and the experimental parameters in table.6..

3.2 TUBULAR SPECIMEN EXPERIMENTS

3.2.1 Axial load

The object of the experiment was to investigate the uniaxial creep behaviour of a tubular specimen geometry and to compare it with uniaxial solid bar data with a view to predicting tube deformation behaviour under combined axial and pressure loading conditions.

Tubular specimens were specially machined from as received 25mm outside and 4.5mm wall thickness, heat treated (1hour @ 1150°C and water quenched) hot formed seamless Alloy 800H tube. A 240mm length of the as-received and heat treated tube was taken and internally honed to a surface roughness (Ra value) of 0.1um. From this a 170 ± 0.05mm long gauge-length of outer diameter 28.7 ± 0.05mm and wall thickness 2.00 ± 0.01mm was machined with a concentricity of ± 0.005mm. A gauge-length having these proportions of diameter to wall thickness was chosen so that a "thin-walled" tube stress distribution could be approximated for mathematical modelling. At the extremities of the gauge-length, profiles were machined to aid the attachment of the extensometer. These specimens were then electron beam welded to two Nimonic 105 end plugs having internal M23 screw threads. These screw threads were chosen to give the best transfer of load to the tube walls without severe disruption of the loading lines (fig.42.).

The experiment was built within the frame of a 2 tonne and 10:1 ratio lever loading creep machine, in conjunction with a 220v, 4000watt vertically split furnace. Temperature control was realised by means of a 3-zone PID Eurotherm temperature control system which gave control to within ± 1°C (as measured) over the test duration.

The specimen train comprised of two Nimonic 105 pull rods extending out of the furnace zone with intentionally loosely cut screw threads to allow some degree of self alignment of the specimen
train on loading (fig.43.). Universal joints were used in the specimen train both above and below the furnace. Due to the uniqueness of the specimen geometry it was necessary to design and build a rather special extensometry system to measure the axial elongations of the specimen during the experiment.

Initially, a number of concepts were investigated, prompted by current work in the high temperature mechanical behaviour field. The initially most attractive ideas, such as the use of high temperature capacitance strain gauges (121) and proximity transducers (122), had to be ruled out due to the high experimental temperatures, these techniques only being usable up to 600-700°C.

The design adopted involved the use of alumina coaxial rod and tube actuators attached to the specimen by means of Nimonic 105 clamps. The extension of the specimen was measured using LVDT's situated outside the furnace (fig.43.). Nimonic 105 was chosen for the clamp material on account of its excellent creep and oxidation resistance within the experimental temperature range. The usual problems associated with the manufacture of components from this material, due to the materials poor machinability, were overcome by using a spark erosion technique.

Alumina was chosen for the actuators in preference to a metal such as Nimonic 105 for two reasons: first to reduce measurement errors due to the metals higher thermal expansion coefficient, this being necessary due to the long lengths of the extensometer actuators; and secondly, to reduce the heat conduction along the actuators which could influence the output from the LVDT's. In this respect, quartz might have been a first choice material due to its low coefficient of thermal expansion. However, problems were identified concerning a phase change which this material undergoes within the experimental temperature range. Unfortunately, some problems were initially encountered with the tube and rod slipping within the Nimonic clamps due to the difference in thermal expansion between the two materials. This was eventually overcome by drilling a small hole in the wall of the alumina tube and locating this on a pin in the extensometer clamp. The coaxial alumina rods were also more rigidly fixed by locating a C-clip within a specially machined groove to prevent the rod from slipping through the extensometer clamp.

Due to the long gauge-length of the specimen and the expected
large deformations during creep, special long stroke DC LVDT's were selected to monitor the relative movement of the actuators and to eliminate the need for periodic resetting of the transducers during the experiment.

The selection of a split-furnace design for the experiment was imperative as it allowed the assembly of the specimen train and extensometry before final transfer into the test rig. Initially, experiments were conducted at test temperatures of 900 and 1000°C, however, due to the persistent burn out of the furnaces at 1000°C, the experimental programme was modified to include 850°C instead of 1000°C.

In order to keep a constant heating up cycle for the series of experiments the temperature controllers were programmed to drive the furnaces up to 50°C below the experimental temperature and then held overnight with the last 50°C being achieved in manually adjusted steps the next morning. The load was then applied to the specimen in the early afternoon.

Strain v. time curves for these experiments conducted at 850 and 900°C are shown in figs.44 & 45, these showing decreasing rupture-strains with increasing rupture-life as would be expected. The parameters associated with these experiments and the experimental conditions are outlined in Table.7, and photographs of the failed specimens shown in fig.46.

3.2.2 Internal pressure

The objective of these tests was to conduct a series of tubular creep experiments with a repeatable biaxial stress ratio whilst varying the stress level in the tube walls by means of the internal pressure.

In order to conduct this and other internal pressure experiments at the Centre a special Tube Testing Facility (TTF) had previously been constructed consisting of four concrete safety cells, with one experimental rig located in each. The cells were designed to withstand eventual explosive failure of tubular specimens tested at internal pressures of up to 300 bars using chemically active gases
such as hydrogen/methane mixtures at 1000°C. For experiments conducted in this research project only inert argon pressures of between 50 & 80 bars were used with tube failures being characterised by slow pressure releases rather than explosive failures. Initially, a prototype pressurisation system (fig.47.) was designed, built and used for the early experiments. Once the initial development problems had been identified and rectified, a final design was agreed upon and each of the four safety cells were then equipped with identical systems (fig.48.)

The concept varied from other conventional designs in that it operated with a controlled gas flow. This design was chosen since future developments of the programme envisaged the use of chemically reactive gases, such as H₂/CH₄ mixtures, as the corrosive as well as pressurising medium, requiring a constant flow of this reactive supply gas.

The supply gas was stored in 200bar bottles in a gas station external to the laboratory and piped to the compressor cabinets and experiments on demand. The inlet pressure to the compressor was controlled by a pressure regulator on the front of the cabinet, while inlet, outlet and dump functions were operated from the master control panel (fig.49.) via electromagnetic valves. The pressure was monitored on the inlet side of the specimen by a high stability pressure transducer, the output signal being fed back to a Eurotherm PID set point pressure controller connected to a manostat valve positioner on the compressor. This compressor was then driven according to the out of balance signal between the set point and the actual pressure.

Since internal pressure was the medium by which the creep load was applied, initial load control was based on a permissible variation in applied load of ± 1% as prescribed by British Standard 3500 :part 1 :1969. However, due to the wall thickness of the tubes being only 2mm, small errors in the dimensions over the gauge-length were considered to have a significant effect on the stresses in the tubes. In order to minimise these errors it was considered that a narrower tolerance should be enforced on the applied pressure. Pressure control during these experiments was maintained to within ± 0.2% which was felt to be the limit of the control system considering the low pressures used during the experimental programme.
As a safety precaution, a secondary argon circulator circuit was also designed, this surrounding the furnace and specimen, in order to blanket the release of active gas on specimen failure and therefore prevent the risk of explosion. Since in this programme of experiments only argon was used as a pressurising medium, the argon blanket circuit was not utilised.

To control the safety aspects of the laboratory each experiment was fitted with its own Programmable Logic Controller (PLC) system programmed in assembly language to take appropriate actions based on pre-set experimental high and low limit alarms. Each of these experimental PLC units were in turn linked with a single master PLC system with responsibility for monitoring the general laboratory facilities such as cooling water supply, gas supply and gas leak detection systems (fig.50.). Typically, a facility or experimental failure would result in the issue of the relevant alarm signals and shut down the experiment according to a logical pre-programmed routine, leaving the facility in a safe condition for examination and rectification.

Data acquisition for the laboratory was based on DC voltage, and AC current lines to a Solartron data logger system. To increase the flexibility of the system a DEC Rainbow personal computer was used to process the raw data and give direct readings of load and strain.

The internal pressurisation rig consisted of a three zone vertically split furnace of internal diameter 75mm mounted on a support frame, the specimen being firmly located above the furnace by a cross bar support and then freely guided in a bearing below the furnace (fig.51.).

The specimens for these experiments were constructed in a similar manner to the axial tube experiment specimens except that no extensometer profiles were cut at the ends of the gauge-length. Special Alloy 800H end caps were then machined and thick walled (15x6mm) Alloy 800H pressure inlet pipes electron beam welded to these before in turn being electron beam welded to the machined specimen (fig.42.). Initially, only one line of weld and a small overlap of the specimen over the end cap was used. However, this resulted in a "hinging effect" of the tube at the weld, leading to local failure in this region. To overcome this problem, a larger overlap and three lines of electron beam weld were used to achieve a
fully built in end condition. Having this fully "encastre" end condition enabled good assumptions of the end-fixing conditions to be made for the subsequent theoretical analysis.

Since the closed ended pressurised tube is an example of a structure deforming by plane-strain, (the axial strain is zero), it was only necessary to measure the diametral strain during the experiment. To this end it was necessary to design and build an extensometer system capable of withstanding the high experimental temperatures.

Some previous work at the Centre (123) had investigated the possibilities of using gamma source radiography as a means of following the diametral expansion of the tubes. Initial results were particularly encouraging, however, the technique would have required a large development input and it was felt that this would have detracted somewhat from the main objectives of this project.

As for the axial experimental section, an actuator design was finally selected for the measurement of the diametral expansion of the tubes. Attempts were, however, made to make improvements on previously reported actuator designs where measuring datums were either taken on the furnace casing (124), which were considered to be inaccurate due to distortions of the furnace during the experiment, or for concepts where non-temperature controlled datums external to the furnace were used (125) leading to long non-compensated lengths of the actuators.

Initially a concept was tried consisting of three 8 x 5mm diameter alumina tubes inserted in the furnace wall at 120° to one another and referenced on the inside furnace liner. Inside these tubes, three 3mm diameter alumina rods were guided in ceramic bearings, each in contact with the specimen surface at its mid point, the relative motion of the ceramic tube and rod, and therefore the diametral expansion of the tube, being measured by DC LVDT's referenced on the outer ceramic tube.

A number of general problems were realised with this design concerning the instability of the tube liner as a reference measuring point, temperature effects on the LVDT output signal of the same order as the measured strains, and seizure of the alumina tubes in the furnace wall due to thermal distorsion of the furnace casing.

To overcome the main problem of a stable reference measuring
point it was decided to construct a Nimonic 105 ring within the furnace and to locate the alumina tubes to this by pinning them in the ring. This had the advantages that the non-common length of the ceramic tube and rod remained small (this being important from thermal expansion considerations) and that variations in movement of the reference ring due to temperature fluctuations were negligible due to the accurate temperature control within the furnace (+ 1°C). Next, larger clearance holes were drilled in the furnace wall and a bearing arrangement constructed on the outside of the furnace to support the tubes whilst at the same time guiding them away from the sides of the holes (fig.52.). Finally a temperature controlled water jacket system was constructed around each LVDT (fig.52.), and connected to a temperature controlled water bath. Temperature control was achieved to + 0.1°C throughout the experiment, which, due to a temperature effect of + 4μm/°C on the LVDT output signal gave a possible measurement resolution of approximately 1μm.

Unfortunately, attempts to assemble the three arm extensometer system failed due to difficulties in replacing the furnace front without breaking the ceramic actuator arms. It was envisaged that the system could be assembled if a horizontally, instead of vertically split furnace were used, however it was not possible to obtain a furnace of this design in the time available.

Due to this problem, it was decided to resort to a one arm extensometer concept making the measurement datum by pulling a modified Nimonic ring gently against one side of the specimen, (fig.53.) using a compression spring on the external bearing system. Initial proving experiments showed the technique to give reproducible results and this design was finally adopted for the experimental extensometer system.

Each tubular specimen was heated up to the test temperature over a period of 25 hours as previously described for the tensile tube experiment and once at the test temperature the pressure was ramped up to the experimental level over a few minutes.

Results of outside diameter strain v. time measured at the centre of the gauge-length are shown for 850 and 900° in figs.54. &.55. Visual examination of the specimens revealed that, in all cases, failure occurred local to the ends of the gauge-length (fig.56.), measured diametral strains at failure in these regions being
significantly larger than those measured at the centre of the gauge length (fig.57.).

Details of the parameters associated with each experiment are given in table.8.

3.2.3 Combined internal pressure and axial load

The object of these experiments was to conduct a series of tubular creep experiments with variable ratios of internal pressure and applied axial load whilst maintaining an inter experimental comparison by means of an equivalent stress relationship. The experiment was built within the frame of a Mand 10 tonne universal screw driven testing machine, located in one of the explosion proof testing cells of the Tube Testing Facility (fig.58.).

The rig consisted of a loading train with a hollow top pull rod through which a high pressure gas inlet pipe was run, which was welded to the specimen end cap. Due to the limited available length within the frame and the need to include a universal joint and an additional high precision load cell, the specimen was constructed as a "dead limb" off the main flow circuit, pressure being input from the top of the specimen.

The pressure was produced and controlled as for the internal pressure experiment and the axial load applied and controlled by means of a closed loop circuit between the high precision load cell, the Mand machine control console and the machine drive motor.

The specimens used for the experiments were constructed identically to the constant load axial tube specimens with two modifications. First, one end cap was modified to include the high pressure gas inlet pipe and secondly, a larger overlap of the specimen over the end cap and three lines of electron beam weld were used to achieve a fully "encastre" end condition, identical to that of the PT experiment specimen (fig.42.).

Following the work of Finnie (88) it can be shown that, under conditions of pure internal pressure, no axial elongation of the specimen is realised and a condition of plane-strain exists. For this reason it was not necessary to measure the axial elongation for the
tubular pressure specimen during the experiment. With the application of an additional axial load, as here, axial as well as diametral extensometry were necessary. To realise this combination it was first necessary to design a larger Nimonic 105 ring for the diametral extensometer to circumscribe the axial extensometer rods (fig.59.). Unfortunately this new design could not be made to fit within the existing 75mm diameter split-furnace and a second furnace of diameter 90mm was used instead.

The only remaining problem concerned the measurement of the same diameter throughout the experiment despite the axial motion of the specimen relative to the furnace and the extensometer. To overcome this problem, a lever system was designed with a 2:1 ratio such that the furnace and extensometry were driven down at half the rate of the specimen elongation (fig.60.). This effectively allowed the measurement of the mid-gauge-length diameter throughout the experiment, based on the assumption of an even distribution of strain along the gauge-length. This assumption was considered valid up to the onset of instability.

The experiment was assembled, by first connecting the pull rods and axial and diametral extensometry to the specimen, away from the rig and then transferring the assembled train to the experimental frame. Each experiment was heated up to the test temperature over a period of 25 hours as previously described for the tensile tube experiment.

Loading was always achieved by first ramping up the pressure over a few minutes and then applying the additional axial load, the precise loads being calculated for a given von Mises equivalent stress and hoop to axial stress ratio at the outside diameter, using the analysis according to Finnie (88).

Experiments were conducted at 900°C at equivalent stresses in the range 35-42 MPa and at hoop:axial stress ratios in the range 1.0-0.33. Results of hoop and axial strain v. time, measured at the centre of the gauge-length, are shown in figs.61.&.62. and the parameters associated with each experiment in table.9.&.10.. Photographs of the failed specimens are shown in fig.63..
3.3 SUPPORT INVESTIGATIONS

3.3.1 Microstructural analysis

3.3.1.1 Optical metallography

Optical microscopy was undertaken for a number of reasons, first to characterise the as-received and heat treated material structures, secondly to investigate the changes to these metallurgical structures after creep exposure and finally to analyse the distribution and nature of the creep induced metallurgical damage throughout the tube sections.

Longitudinal and transverse sections of each specimen were taken close to the fracture zone and polished samples examined in the etched and unetched condition. Etching of the alloy was achieved using an electrolytic technique of, 2.5v for 5s in 10% oxalic acid in water.

3.3.1.1.1 Macro-Metallography

Longitudinal sections of the MS and TT specimens at each of the test temperatures 850 and 900°C are shown in figs.64.to.67.. At both temperatures, the TT specimens exhibited W-type cracks at both the inside and outside surfaces of the tube, no clear trend being observed for preferential cracking at either surface. At 850°C internal W-type cracking was also observed, this being more prevalent in the TT specimen sections. At 900°C, however, this internal damage was only observed at the higher stresses. This increased internal W-type cracking at 850°C may have been due to the increased failure strains observed in the specimens at this temperature it is difficult to give a conclusive explanation since, for a valid comparison, the specimens would have to have been exposed to the same stresses and exposure times.

Comparison of the MS and TT sections at the same stress and temperature showed the length of the "damaged" portion of the gauge-length to be greater for the TT specimens. This may only be a
function of the longer gauge-length of these TT specimens.

Examination of the transverse and longitudinal sections taken from the 900°C PT specimens, near to the point of failure, show the type of creep damage to be entirely due to intergranular surface cracks, little or no internal void formation being observed. The number and distribution of the cracks in the transverse sections were seen to decrease with decreasing stress and therefore increasing rupture time.

Low stress experiments (A12 & A23) (figs.68.&.69.) show failure to have occurred at very low strains by the propagation of one or two long cracks, these apparently nucleating and propagating from the inside surface of the tube. Very few cracks other than those causing failure being observed around the tube circumference. At the higher stresses (A1 & A4) (figs.70.&.71.) much larger diametral strains were observed, fracture apparently being caused by a combination of crack propagation and ductile necking. Cracking was far more uniformly distributed around the tube circumference than at the lower stresses, however, no clear trend could be found for whether the cracking was predominantly observed on the inside or outside surfaces of the tubes.

Sections taken from the high and low stress PT experiments at 850°C (figs.72.&.73.) showed similar trends to those observed at 900°C. At 850°C, however, a predominant tendency for cracking at the inside surface of the tubes was apparent. Examination was also made of three transverse sections of thicker walled Alloy 800H tubes tested during previous research programmes at the Centre (figs.74.to.76.). In all cases crack nucleation and propagation were predominantly from the outside surface of the tube, the number of nucleated cracks, ie. the concentration of creep damage, being found to decrease with increasing rupture times.

For convenience of analysis, the MT experimental results were subdivided into two main groups, the first a series of fixed stress ratio experiments ($\sigma_0/\sigma_z = 1.0$) conducted at different equivalent stresses (figs.77.to.81.) and a second series conducted at a fixed equivalent stress and variable stress-ratio (figs.82.to.84.) and (fig.78.).
Examining the fixed stress-ratio series first, all specimens were found to fail in a pressure mode, surface cracking being predominantly from the outer surface of the tube with an increase in the diametral strain and the number of the surface cracks being observed with increasing applied stress (cf. B10 & IHMT) (figs.77 & 81).

Sections from the series of variable stress ratio experiments at approximately the same von Mises equivalent stress showed a transition in failure mode from longitudinal cracking and pressure failure at the higher stress ratios (hoop:axial = 1.0), to transverse cracking and axial failure at lower stress ratios (hoop:axial = 0.62) (figs.78 & 84).

3.3.1.1.2 Micro-Metallography

Longitudinal and transverse micro-sections taken from the as received and the as-received and solution-treated material showed no obvious differences in structure. Both exhibited a heavily twinned grain structure of ASTM austenite grain size 4 (figs.85 & 86). TiC precipitates were observed within the grains and at grain boundaries (fig.87a & 87b). These precipitates were frequently observed in the form of axially orientated "stringers" in the longitudinal sections and occasionally had a local grain refining influence on the structure even after heat treatment (fig.88).

Microstructural examination of the tubular specimens revealed a number of interesting results. The first observation was the apparent difference in nature of the cracks nucleated at the inside and outside surfaces of the tubes (figs.89 & 90). At the inside surface the cracks were intergranular; however, they were widely opened and showed no obvious signs of internal oxidation. At the outside the cracks were also intergranular but partially filled with oxide.

In specimens tested at low stresses, examination of the outside surface showed signs that the oxide had grown laterally and closed the opening to the crack (fig.91). In the higher stress experiments, however, this behaviour was not observed.

A section taken from the 900°C, 29.3 MPa, TT specimen,
(figs.92.&.93.), showed some evidence of r-type cracking at the grain boundaries at the inside surfaces of the tubes, these having linked up to form grain boundary cracks. This behaviour was not observed at higher stresses and lower temperatures.

Closer examination of the inside and outside surfaces of a number of specimens, revealed shallow recrystallised surface layers (figs.94.&.95.). These are probably the result of work-hardening of the surface layers during the specimen machining process, this work-hardening providing the necessary energy for recrystallisation during the subsequent high temperature exposure.

The previously mentioned "stringers" were also found to have an influence on the crack propagation behaviour in the longitudinal sections (fig.96.). Here, crack propagation was seen to be partially arrested due to the presence of the TiC "stringers". This micrograph also shows some evidence of the nucleation of voids around the TiC precipitates.

3.3.1.2 Scanning electron microscopy

Examination of fracture faces was undertaken on a JEOL JSM-35 scanning electron microscope with an accelerating voltage of 25 keV, to investigate the effects of specimen geometry, temperature and stress on fracture mode and mechanism. Only the MS and TT specimen fracture faces were examined since only these specimens gave complete fracture faces which could be examined.

Generally the fracture faces showed an intergranular appearance, the sharpness of the facetted nature of the surface being dependent upon applied stress and test temperature.

Comparison of results of fracture faces for TT experiments at 900°C showed a decrease in the facetted appearance from low to high stress (figs.97.&.98.) with the high stress result showing a much deeper topography. Similar results were also observed between high and low stress MS specimen fracture faces (figs.99.&.100.).

Examination of the TT results of nominally 1000 hours duration at 850, 900 and 1000°C (figs.101.,.97.&.102.) showed a decrease in the facetted nature of the surface with decreasing temperature.
Comparison of MS and TT surfaces from specimens tested under equivalent stress and temperature condition showed essentially similar appearances (figs.103 & 104).

One further fractograph, taken from the 29.8 Mpa, 900°C MS specimen, showed evidence of internal oxidation of the grain boundary network resulting from a surface crack (fig.105).

3.3.1.3 Transmission electron microscopy

Due to some discrepancies in the literature concerning the time, temperature precipitation behaviour of MC and M₂₃C₆ type carbides, transmission electron microscopy studies were made on a few aged samples to investigate the effects of aging time and temperature on the as-received and solution-treated structures.

Specimens were prepared by first cutting a 700um slice of material from the specimen using a 300um diamond saw. This slice was then mounted in a special holder and ground down to about 400um using 260-600 grit paper. The sample was then turned and the procedure repeated on the second face until 120-100um thick. The specimen was further prepared by a special technique developed at the Centre by Helbach (126) which allows the preparation of a number of fields from a single specimen. The 100um sliver was held in a low-turbulence double-stream electrolytic cell and reduced to about 30um. The advantage of this technique is that the sample section is reduced from both sides, the sample faces remaining perfectly plane and parallel to one another throughout the preparation. The sample was then framed with a protective film since the field lines in the gap tended to attack the edges of the sample preferentially at this stage in the preparation. With the protective frame in position the sample was further thinned until the corners of the sliver just penetrated giving a potential for the preparation of 4 foils from a single specimen. These corners were then cut from the sliver and mounted separately in small copper cages which were in turn mounted in the TEM specimen mount.

Five specimens were investigated with the following treatments in order to investigate the specific carbide precipitation behaviour of
this experimental batch of Alloy 800H.

i) as received

ii) as received and solution treated at 1150°C for 1 hour and water quenched

iii) as for ii) and aged 885 hours at 1000°C

iv) as for ii) and aged 1013 hours at 900°C

v as for ii) and aged 900 hours at 850°C

The as received material structure showed a medium dislocation density with local concentrations adjacent to grain boundaries and around matrix MC type carbides (fig.106.&.107.). Grain boundaries were decorated with M23C6 carbides (fig.108.).

On solution treatment at 1150°C for 1 hour the average dislocation density was seen to be reduced with some early stages of network formation being observed (fig.109.). Again some intensification of the dislocation density was seen around matrix MC type carbides. M23C6 carbides were no longer visible at grain boundaries, however, some coherent MC carbides were observed instead (fig.110.).

On aging at 1000°C for 885 hours the structure showed a lower dislocation density than in the heat treated condition. Large M23C6 carbides were observed at grain boundaries while many small ones (200-300 A) were visible within the matrix (fig.111.&.112.). Closer examination revealed these to be largely precipitated on short dislocation lines, these dislocations being effectively pinned by the carbides (fig.113.). MC carbides were again observed in the matrix as well as at grain boundaries, some of which exhibited coherency with the matrix (fig.114.&.115.).

Specimens aged at 850 and 900°C for nominally 1000 hours were found to show essentially similar structures. Large MC carbides were observed within the matrix, the localised dislocation network around these being decorated by M23C6 carbides (fig.116.). The only significant difference between the 850 and 900°C behaviour was the size of the matrix M23C6 precipitates; these being of the order 2000A at 900°C (fig.116) and 3000A at 850°C (fig.117.).

Grain boundary MC and M23C6 carbides were both observed, these often being found in association with one another with the M23C6 carbide growing around the MC carbides (fig.118.)
3.3.2 Finite element analysis

Finite element analysis of the tubular specimens was undertaken to investigate the reasons for the preferential failure near to the ends of the gauge-length. Work by Murakami and Iwatsuki (127) has previously modelled local bulging of pressurised tubes, by means of a finite difference mathematical approach and has shown that the position of this local bulging is sensitive to the tube aspect ratio and the end fixing condition.

The present specimen was modelled by a mesh of four node quadrilateral isoparametric elements as shown in fig.119. with four elements through the wall thickness to model the non-linear stress profile with radial position. Additionally a smaller element size was used in the area of interest, at the end of the gauge-length, with an element "grow-ratio" being used in the direction of the centre of the gauge-length.

Initially an elastic stress analysis of the structure was performed for conditions of internal pressure and internal pressure plus axial load, producing hoop/axial stress-ratios of approximately 2:1, 1:1 and 1:2. Plots of centre element hoop stress v. axial position were made for the outside and inside wall elements (fig.120.). Results show the hoop stress to be a maximum at the inside surface as predicted by the Lamé analysis. It is also found that a stress concentration occurs near to the end of the gauge-length for both the hoop and axial stress profiles, this concentration factor being a maximum at the outside surface of the tube, increasing in magnitude and moving towards the end of the gauge-length with increasing axial stress component. Investigation of the axial stress profiles for the inside and outside wall elements also show sharp stress concentrations, these being positive at the outside and negative at the inside surfaces (fig.121.). This form of stress profile is characteristic of a superimposed bending moment and may have been induced by the local hoop strain from the hoop stress concentration.

Having set up the base line behaviour of the tube under elastic stress conditions the analysis was developed to consider the behaviour under creep conditions. Unfortunately the cost of such creep finite element analyses were found to be of the order of 1000x
the cost of the equivalent elastic run. For this reason it was not possible to investigate the development of the stress system for long creep exposure times, the maximum allowable time being only 1 hour. Despite this restriction some interesting results were obtained.

For the internal pressure condition almost complete redistribution of the hoop and axial stress profiles were observed within the first hour of exposure (figs.122. & 123.), this serving to verify the assumption of considering full redistribution of the stress distribution from time zero in the Finnie analyses. Superimposed on this figure are the predicted stresses based on the thin-wall elastic analysis and the thick-wall redistributed creep analysis. Comparison of these theoretical stresses with those predicted by the finite element analysis show good correlation. However, in the elastic analysis, and particularly in the redistributed creep analysis, the finite element runs predict a reduction in the stresses towards the centre of the gauge-length. It is also found that the position of the hoop stress concentration peak moves towards the centre of the gauge-length on redistribution and decreases in magnitude compared with the elastic stress profile fig.122.. With the addition of an axial stress component, the redistributed hoop and axial stress concentrations are found to increase and move towards the end of the gauge-length (fig.124.), as found for the elastic analysis. This movement of the stress concentration towards the end of the gauge-length corresponds with the observation that the local bulge also moves further towards the end of the gauge-length with increasing axial stress.

Plots of hoop and axial stress profiles through the wall thickness at two positions, first adjacent to the stress concentration and secondly mid way along the gauge-length are shown in fig.126.. These show that elastic and redistributed hoop stress profiles are highest near to the stress concentration than elsewhere in the gauge-length. Examination of the same profiles for axial stress, however, show a rotation of the redistributed stresses at the concentration with respect to the rest of the gauge-length. This rotation was also seen to increase with exposure time from 0.5 hours to 1 hour exposure and it was therefore considered that this resulted from an induced bending moment in the tube due to the local hoop stress concentration and its associated local bulge.
This evidence of local stress concentrations adjacent to the ends of the gauge-length and their influence upon the tube deformation behaviour is used later to help explain the axial and hoop strain v. time results obtained from the MT experiments and to explain the preferential internal or external surface cracking in the bulged regions of the tubes.

3.3.3 Characterisation of anisotropy

By the nature of the production route of the tubes, it was likely that the material would contain an inherent degree of anisotropy between the axial, circumferential and radial directions. In order to quantify this variation in properties with direction, two experimental approaches were followed. The first consisted of Vicker's Diamond Pyramid (VDP) hardness testing of prepared faces normal to each of the principal tube directions. Two specimen preparations were used, a conventional 800 grit and 6um diamond polishing treatment and an 800 grit preparation followed by electropolishing.

Secondly, a series of small compression tests were conducted on an ESH 10 tonne hydraulic testing machine under ambient temperature conditions. Small 4mm cube specimens were cut from the tube wall and accurately machined to their final dimensions before being tested in one of their three principal directions. Stress v. strain plots were made for each direction and the results compared on the basis of the intercept of the linear elastic and plastic regions of the stress strain curve. Obviously it would have been more beneficial to conduct experiments under creep conditions and in the tension rather than the compression mode, in order to characterise the material anisotropy, however, this was obviously impossible due to dimensions and geometry of the source material.

Results of the hardness tests and the compression test stress v. strain curves are given in table.11 and fig.127. respectively. Hardness results from the specimens which were mechanically polished were found to be significantly higher than those which were electropolished. This result arises from the induced surface stresses
from the mechanical polishing technique, these having a large effect on the Vickers hardness readings due to the small depth of penetration of the indentor. For this reason the electropolished specimen results were taken as the most representative of the two sets of hardness results.

Comparison of these hardness results and the small compression test tangent-intercept results (table 11.), show good agreement, the radial direction exhibiting the highest yield stress in both cases, followed by the hoop and then the axial directions.
4.0 DISCUSSION

4.1 COMPARISON OF UNIAXIAL RESULTS WITH PUBLISHED ALLOY 800H DATA

In order to put the experimental results obtained in this project into context with other data obtained on Alloy 800H, comparison is made with the least squares best fit lines to the Data Bank results presented in the literature survey.

As previously mentioned in the experimental sections, two batches of Alloy 800H tubing of significantly different composition were initially tested using small solid-bar specimens machined from the tube walls. Results of both these series of experiments are first compared with each other and then in relation to the Data Bank results obtained using search profile (1) (tables.3 & 4). Comparison is based on three material properties: Creep rupture-life; creep rupture-elongation; and minimum creep-rate.

Creep rupture-life data at 900 and 1000°C shows the ST (Batch I) material to give longer rupture-lives than the MS (Batch II) material (fig.128). In comparison to the least squares fit of the Data Bank results, both the MS and ST results showed longer rupture-lives at 900°C. At 1000°C, however, the ST results correlated well with the Data Bank curves, the MS results exhibiting much lower rupture-lives. Unfortunately, no correlation of the 850°C results could be made due to the lack of available results from the Data Bank at this temperature.

Creep rupture-elongation data showed higher strains to failure for the ST as compared to the MS material (fig.129). In both cases, however, the results were much lower than those suggested by the Data Bank, particularly at the longer times. Once again no correlation was possible between the experimental and Data Bank results at 850°C.

Since the ST experiments were conducted without extensometry, no minimum creep-rate data could be derived for these experiments. Comparison was, however, possible between the MS and Data Bank minimum creep-rate results. This comparison (fig.130) shows quite good correlation at 900 and 1000°C but again no comparison was possible at 850°C.
The lower rupture-life results recorded for the MS cf. the ST material can be explained if rupture-strain is considered to be the controlling influence on rupture-life. These reduced rupture-strains could be accounted for through a consideration of the compositional differences of the two batches of material. The MS batch II material exhibits higher carbon, aluminium and titanium concentrations as shown in table 1., these manifesting themselves in the form of:

- first, an increased volume fraction of TiC (fig.19.),
- secondly, an increased concentration of dissolved carbon and therefore an increased tendency for grain boundary Cr$_{23}$C$_6$ precipitation and
- finally, due to the increased aluminium concentration, an increased weakening of the grain boundaries due to Al$_2$O$_3$ precipitation as a result of the segregation of aluminium to these regions.

As already shown (figs.28.&.29.), the experimental results do not show a particularly good correlation to the Data Bank lines. It should be remembered, however, that these best fit curves, particularly those of rupture-strain, were made on data sets having a wide degree of scatter (fig.24.). This scatter originates due to the wide compositional search profile used to obtain these results from the data bank (search profile [1] (tables.3.&.4.)) and almost certainly, due to variations in the testing procedure used to obtain these results. With particular reference to fig.24. It can be seen that a number of low strain results exist, these showing much better correlation to the experimental data.
4.2 COMPARISON OF UNIAXIAL BAR AND TUBE RESULTS

As indicated in the experimental section two forms of constant load experiments were performed; those on tubular specimens and those on small solid-bar specimens cut from the tube walls. It might have been expected that these experiments would produce identical creep results, however this was not found in practice.

A comparison of the small solid-bar specimen (MS) and tubular specimen (TT) constant load results at 850 & 900°C are shown in figs.131.&132.. Comparison is also made for the 900°C, 42MPa condition between the VMS, MS and TT results (fig.39.)

Initial examination shows the strain v. time curves to be dominated by a region of increasing strain-rate with little or no evidence of an initial region of decreasing strain-rate. This feature is felt to be largely due to the high-stress, high-temperature regime in which the experiments were performed, although other factors such as the high C, Al & Ti levels of the material and previous processing history may also have significant influences.

Additionally some differences in the rupture-lives and failure-strains were observed between the MS and TT results and to investigate this further plots of log (stress) v. log (rupture-time) and rupture-elongation v. log (rupture-time) were plotted (figs.133.&134.).

Two general points are concluded from these results: first, at 900°C, the TT specimen results show increased rupture-times with respect to the MS results, whereas, at 850°C the rupture-times of both specimen types were similar; second, at high stresses the rupture-strains from the MS specimens were greater than for the TT specimens, whereas, at lower applied stresses little difference in the rupture-strains between the two specimen types was observed. These general observations are summarised in figs.135.&136..

Roedig, Penkalla et al (74), working on Alloy 800H at 950°C have reported tensile creep results for tube and bar specimens similar to those obtained here at 900°C. It should be noted, however, that their bar specimens were machined from bar stock material from the same material cast and not from the tube material itself.
The authors proposed no specific explanation for these differences but did suggest that the differences may have originated from some surface related mechanism such as oxidation.

To investigate further the divergence of the MS and TT results, a number of potential influencing factors were examined.

1) The effect of the differences in specimen geometries on the measured strains is considered in BS.3500 where a proportionality ratio of specimen gauge-length to diameter is set out, \( L_0 = 5d \), in order that results obtained from solid-bar specimens of different dimensions may be comparable. No such guide lines are available for the comparison of tubular and solid-bar geometries.

Increasing specimen gauge-length, for a fixed cross-section area will result in lower measured axial strains. This will have a more significant effect on the rupture-strains rather than the strains measured throughout the experiment, since the source of this difference relates to the length of an "active deformation-zone" or zones in relation to the length of the gauge-length. This "active deformation-zone" is most significant in relation to the latter stages of the experiment where local necking occurs.

The length of the "active deformation-zone" is primarily dependent upon the the specimen diameter and not the gauge-length itself. There is also the problem of correlating equivalent diameters of solid-bar and tubular sections in relation to predicting the length of the "active deformation-zone" in each specimen type.

With due consideration of the above points it is estimated that the the TT specimen exhibits a smaller ratio of the "active deformation-zone" to the gauge-length than the MS specimen and that this accounts for the smaller rupture-strains observed for the TT specimens at the high applied stresses at both temperatures.

The similarity of the rupture-strains at lower stresses can be explained by the reduced necking behaviour of the specimens, as seen in the longitudinal metallographic sections (figs.64.to.67.), there being little or no contribution to the measured strains from an "active deformation-zone" in this region.

To verify this assumption of reduced axial rupture-strains in the case of the TT experiments, due to the increased specimen gauge
length, comparison was made of the strains to failure based on two measurements: axial strain and reduction-in-area. Since these reduction-in-area strains are not influenced by the specimen gauge-length it might be expected that the differences in the rupture-strains for the two specimen types, measured on this basis would be less than those based on the axial strains. If true, this would support the theory of the lower rupture-strains for the TT specimens being due to the difference in gauge-lengths of the specimens.

Results of reduction-in-area strains for the MS and TT specimens (fig.137.) do in fact show this improved correlation compared to the axial strains, lending support to the theory.

2) Consideration may also be made of the possibility that the actual state of stress operating in the tube and bar specimens might be different. The intention of these experiments was to generate constant-load data under uniaxial stress conditions. Conventional tensile specimen geometries do in fact give rise to such uniaxial stress systems, however, a tendency towards a multiaxial stress state, and in particular one of plane-strain can occur if the specimen geometries constrain the material deformation and flow behaviour in one direction.

Two examples of such constraints are found for axially loaded sheet metal and large diameter solid-bar tensile specimens. For the sheet metal specimen the large width-to-thickness ratio of the specimens effectively restricts the material deformation in the width direction, this having a retarding effect on the axial deformation. The large diameter bar specimen behaves in much the same way, the large diameter having a restricting on the diametral contraction and therefore the axial strain.

Because the MS specimen is of such a small diameter the stress system present must be one of uniaxial tension. The tube, however, shows similarities with both the above examples of plane-strain stress systems. First, the tube can be considered as a wide sheet metal specimen bent round into a cylindrical geometry and joined at its edges. Second, if the critical tube dimension is taken to be the mid-wall diameter, then it has similarities with the large diameter bar tensile specimen. Clearly neither of these analogies are entirely fulfilled, however, it is conceivable that the axially loaded tubes
do exhibit some tendency towards a plane-strain stress system in comparison to the small solid-bar MS specimens. This tendency can account for the reduced axial strain-rates encountered from the commencement of loading for the TT specimens, when compared to the MS specimens.

3) The effect of an oxidising environment can also be considered as a contributory factor to the differences in the MS and TT behaviours, this being suggested by the apparent temperature dependence of the stress rupture results (fig.133.). Borggreen and Huntley (128) have reported significant reductions in load-bearing section-area due to oxidation during testing of ferritic solid-bar specimens and Cane (129) has shown that specimen diameters of 50mm and greater are required before testing in air becomes equivalent to that in vacuum, for a 1/2%Cr-Mo-V steel at 675°C (fig.138.). A similar result to these was found for the comparison of MS and VMS specimen creep curves for experiments conducted at 900°C and 42 MPa. Although Cane has presented this figure as evidence for an oxidation influence on strain-rate and rupture-life, the results can also be described by a change in stress system towards plane-strain with increasing specimen diameter as discussed previously.

Taking the dimensions of the MS and TT specimens and considering only the outside surface of the tube to be oxidised (the specimen was electron beam welded under vacuum) it can be shown that the ratio of the surface-area per unit volume for the MS specimen as compared to the TT specimen is 1.0:0.54. Unfortunately, evidence from weight-gain curves, measured oxide thickness data for Alloy 800H (130) (table.12.) and metallographic sections (fig.90.) reveal that the oxide scale thickness is only of the order of 5 to 20 microns. This is insufficient to explain the appreciable increases in stress necessary to account for the differences in behaviour of the two specimen types.

4) Another feature of oxidation behaviour of Alloy 800H, is its susceptibility to internal oxidation along grain boundaries adjacent to the oxidised metal surface. Metallographic evidence of this surface grain boundary attack has been found for the MS and TT specimens (figs.91.&.64.to.67.). Similar findings have been reported
by Wilshire and Evans (131) for Nimonic 105 creep tested in air at 850°C. In order to quantify this phenomenon and particularly the depth of penetration, reference was made to the work of Baxter (130) on the oxidation behaviour of 800H under un-stressed conditions (table.12.). He found typical oxide penetration depths after 150 hours at 900°C to be only of the order of 15um. McMahon & Coffin (132) have, however, shown that increased oxidation depths can be encountered under conditions of applied stress.

Schultz & Rahmel (133) have shown that grain-boundary oxidation in Alloy 800H under creep conditions can be initiated by strain-induced cracking of the surface oxide layer. Further they observed that this cracking behaviour was strain-rate dependent, strain-rates of $10^{-6}s^{-1}$ resulting in the early failure of the oxide layer adjacent to the matrix grain boundaries. This rupture of the oxide allowed the penetration of oxygen and the subsequent oxidation of grain boundary carbides. By contrast at strain-rates of $10^{-8}s^{-1}$ initial cracking of the oxide took much longer, sealing of the surface being observed at these lower strain-rates due to the lateral growth of the oxide film. Despite this sealing action, preferential cracking of the oxide at these repaired sites was found to occur on further strain.

Collins (134) has reported the failure of Alloy 800H tubular components in service under an oxidising environment by a similar stress-corrosion mechanism. He proposes that, after initial rupture of the oxide-scale, subsequent grain boundary penetration of the oxygen rather than diffusion occurs, with the gradual oxidation of grain boundary carbides. The oxidation of these grain boundary species leads to a reduction in the grain boundary strength, cracking of the boundaries occurring with further strain.

Grain-boundary cracks were, however, observed at the inside as well as the outside surfaces of the tube (figs.65.&67.), those at the inside showing no obvious signs of oxidation. This apparent absence of oxidation at the inside surface is expected since the specimens were electron beam welded under vacuum.

Oxidation cannot, however, be ruled out completely as a crack initiation mechanism at the inside surface since selective oxidation of grain boundaries and grain boundary carbides can still occur at low partial pressures of oxygen, certainly lower than those encountered during the electron-beam welding process. Much work has
been carried out to investigate the influences of low partial pressures of oxygen on the high temperature mechanical properties of materials (135)(136)(137). Results tend to show that significant loss in mechanical properties can occur at these low ppm oxygen levels, and although there is conflicting evidence for the precise mechanisms which are operating they tend to be related to selective oxidation of grain boundaries and grain boundary precipitates, this being enhanced compared to the high oxygen partial pressure environments due to the absence of continuous protective oxide films.

5) Another potential influencing factor may have been non-axiality of loading due to mis-alignment of the testing machine and specimen screw threads. These effects could be largely reduced by the use of universal joints in the specimen loading train.

This misalignment of the specimen and the loading axis would result in the production of a bending moment, and therefore a bending stress. This bending stress is algebraically additive to the applied axial load such that there is an increase in the stress on one side of the specimen and a corresponding decrease on the opposite side.

It is likely, however, that the actual eccentricity in loading of the MS specimen is smaller than that for the TT specimen, since the MS specimen was tested in a "dead-load" machine, whereas the TT specimen was tested in a conventional lever-arm machine. The dead load machine finds its own axis due to gravity, whilst the lever-machine has its loading line dictated by the top and bottom fixtures of the frame.

A superimposed bending stress could have two possible effects on the specimen's creep behaviour. First, the bending stresses may be relieved early in the experiment by differential creep straining in the specimen. In which case consideration must be given to the possible effects of this initial straining on the subsequent creep behaviour of the specimen. Secondly, the bending stresses may not be relieved or, at most, only partially relieved throughout the experiment. In this case, consideration must be given to the effect of this differential stress system on the subsequent creep behaviour.

Very few experimental investigations have been undertaken to evaluate the influence of bending on creep behaviour. Hayhurst (138) reports that where bending stresses are not relieved then significant
reductions in rupture-life can be expected. However, in cases where the stresses are relieved during testing he noticed that rupture-lives were not so greatly reduced. Bressers (139) also reports similar results, finding reductions in creep strains and rupture-lives and increases in the secondary creep rate with testing under conditions of a superimposed bending stress. By contrast, however, Loveday (140) reports increases in creep rupture-lives of approximately 25% for Nimonic 80 tested in tension with a superimposed bending moment.

The general conclusion that can be drawn from the above is that a superimposed bending moment will have an influence on the creep deformation and rupture behaviour of the specimens. This will generally give rise to decreases in rupture-life and ductility, however, this result is not conclusive and the precise behaviour is likely to be sensitive to such variables as specimen geometry and material, degree of misalignment and the applied stress level.

It is likely that the MS and TT results are both influenced to some degree by bending stresses due to misalignment in loading, however, due to the differences in specimen geometry, and the designs of the machines used for both types of experiment it is difficult to conclude to what degree each are influenced. In fact the consistent differences in the the two series of results rather suggest that bending may not be playing a significant role since one might expect bending to manifest itself in a random way upon the results.

In conclusion, therefore, it appears that there are two fundamental differences in the MS and TT results: first, the TT experiments predict lower rupture-strains than the MS experiments and second, the TT experiments predict lower creep-rates than the MS experiments.

It is believed that the principal influences on the strain-rate differences are stress-state related due to the tendency towards a plane-strain stress system in the TT specimens as discussed in point 2. This might also be expected to give reduced rupture-strains for the TT specimens, however, other factors are also influential in this respect. For example the increased gauge-length of the TT compared to the MS specimens will result in lower rupture-strains (point 1) and although it is not clear to what degree the results
will be affected, the oxidation related surface cracking, observed in metallographic sections, (point 4), will have the effect of reducing rupture-life and rupture-strain.

These results demonstrate that great care must be taken in the selection of a representative specimen design from the point of view of understanding the stress systems which may be operating. In certain applications where a process is to be modelled, and the stress system associated with this process is well characterised, e.g. the plane-strain rolling of sheet metal, experimental data derived under plane-strain testing conditions can be used directly to characterise the material behaviour during the process. Alternatively, testing may be conducted under uniaxial stress conditions and this data incorporated in a mathematical model to predict the behaviour under plane-strain, or other stress system, conditions.

Since this second alternative is the one which is to be adopted in this research programme, it is important to recognise that the experimental data should be derived under uniaxial stress conditions and for this reason a small solid-bar specimen design has been adopted for the generation of the constant-stress base-line data necessary for the generation of the mathematical model of deformation and rupture behaviour under multiaxial stress conditions.
4.3 CREEP DEFORMATION MODELLING

As a first step to the development of a multiaxial creep model for tubular components, the uniaxial creep properties of the material are characterised under constant-stress conditions.

The validity of a changing uniaxial stress model is then examined using the derived uniaxial constitutive equation and comparison of the model results made with the experimental constant-load results.

This changing stress model is then extended to consider multiaxial stress conditions with the inclusion of such variables as equivalent stress criteria, principal stress-ratio and material anisotropy. A comparison of the predictions of this model with the PT and MT experimental results is then made in order to determine the controlling stress criteria on the multiaxial creep deformation behaviour.

4.3.1 Uniaxial

The first series of constant-stress experiments (CS experiments) conducted on the machine of a sister department at the Centre appeared to show variable stress dependence when strain was plotted against time (fig.40.). Comparison of these results with the MS results exhibited further discrepancies (fig.139.), the constant-stress results showing much higher initial creep-rates than the MS constant-load results.

To investigate this further, comparison was made between the CS results and a series of constant-load experimental results conducted on the same machine and utilising the same specimen design (CLB experiments) (fig.40.). Correlation here, although not ideal, was significantly improved, with the CLB results showing higher initial creep-rates than the CS results, as would normally be expected from a comparison of constant-load and constant-stress creep curves.

Subsequent investigation revealed that these CLB and CS specimens had been mistakenly prepared from an earlier batch of material (Batch I) rather than from the actual experimental material (Batch II), the earlier batch having a different chemical composition (table.1.).
Earlier constant-load creep-rupture experiments on this Batch I material (ST experiments) had shown larger values of rupture elongations than for the actual experimental material (Batch II) (fig.129.), this corresponding well with the rupture-strains observed for the MS and CLB experiments.

Additional examination of the CS specimens after failure did show some evidence of multiple necking between each of the points on the gauge-length where thermocouples had been attached. Since the reduction-in-load calculation for these constant-stress experiments had been based upon a theory of uniform deformation then it is likely that some error in this calculation had occurred. Due to these irregularities no attempt was made to model the CLB specimen constant-load behaviour from the CS constant-stress results.

Instead, a second series of constant-stress results were conducted at ERA Technology Ltd., Leatherhead, on specimens machined from the Batch II material. Results of these, presented in (fig.140.), show, in all cases, the ERA constant-stress curves lying below the corresponding MS constant-load results. These strain v. time results do not, however, exhibit the classical, linear steady-state region normally associated with constant-stress creep curves. By contrast the curves showed a gradually increasing creep-rate with time from loading, this possibly being a consequence of the high-temperature, high-stress regime in which the experiments were conducted, the stress and temperature being too high for strain-hardening mechanisms to be operative.

Initial attempts were made to predict the MS constant-load behaviour from the ERA constant-stress results using a Norton power-law constitutive equation, the equation constants being evaluated from a double logarithmic plot of minimum strain-rate and applied stress.

\[ \dot{\varepsilon} = 2.5 \times 10^{-24} \cdot \sigma^{10.191} \]  

(56)

This constitutive equation was incorporated into a time-step integration model in order to take account of the changing true-stress in the constant-load experiments. The stain over the time-step being calculated by the integration of equation (56) using a 4th order Runga-Kutta numerical integration routine with a modification
on the step-size to reduce the accumulated error. The true-stress for
the following integration step was then recalculated using equation
(57), based on the accumulated axial strain, \( \varepsilon_{\text{tot}} \):

\[
\sigma = \sigma_o (1 + \varepsilon_{\text{tot}})
\]  

(57)

This procedure was repeated until the strain-rate predicted by
equation (56) became infinitely large. Allowing the strain-rate to
become so large frequently gave problems in respect to the failure of
the computer programme. It was found, however, that by redefining the
failure criterion to be equivalent to 100% strain, that the programme
behaved normally, this change in criterion having no influence on the
calculated results. This failure criterion was referred to as the
"instability failure criterion" and a flow diagram for this model is
shown in Appendix 1.

In this way model constant-load strain v. time curves were
derived and the results compared with the experimental MS constant-
load results.

These model results, based on the Norton power-law, showed very
poor correlation with the MS experimental results (fig.14.).
Although the theoretical constant-load curves modelled the initial
creep-rates quite well, they were unable to predict the higher strain
behaviour. This result is not particularly surprising, since the
minimum creep-rate in no way models the whole constant-stress creep
curve and cannot, therefore, be expected to be a good basis for the
prediction of material deformation behaviour under changing stress
conditions.

A modification to the technique was therefore made, whereby, the
strain dependence of the strain-rate was taken into consideration by
deriving functions of the Norton constants, \( A \) and \( n \), in terms of
strain. Values of strain-rate for the ERA constant-stress creep
curves were taken at successive 1% strain intervals (table 13.) and
values of the Norton constants \( A \) and \( n \) evaluated on each occasion.
The strain-dependence of these constants were then derived and the
equations presented below.
\[
\log_{10} A = 0.940 \varepsilon - 19.636 \quad (58)
\]
\[
n = -0.566 \varepsilon + 7.94 \quad (59)
\]

Using these equations constant-load creep curves were predicted in the same manner as for the fixed value Norton power-law analysis using the Runge-Kutta integration routine. Results of this analysis (fig.142.) show a much improved correlation with the MS constant-load experimental data, than previously found using the unmodified Norton power-law theory. Best correlation with the curve shapes, was obtained for the 42, 32 and 29.8MPa results. The instability failure criterion, however, continued to predict rupture-lives far in excess of the experimental results particularly at the low stresses. This result can be explained by the fact that such a criterion is essentially only valid for ductile materials. It does not take into account the build-up of creep damage due to internal cavitation or surface cracking, these mechanisms being particularly prevalent at the lower stresses, leading to premature failure. If the predominant damage mechanisms could be modelled as functions of time, strain or strain-rate then it would be possible to include these as a component of the modelling programme. Of course one complication is that the constant-stress data is also influenced by these same mechanisms and therefore some correction of the data must first be made before a valid extrapolation can be realised.

As a final improvement on the prediction of the constant-load results from the constant-stress results, use was made of the recently published "Theta-Projection-Technique" (29). The method was used essentially as a curve fitting routine on the constant-stress results, values of the theta constants in equation (10) being derived using a non-linear regression analysis.

\[
\varepsilon = \Theta_1 (1 - \exp(-\Theta_2 t)) + \Theta_3 (\exp(\Theta_4 t - 1)) \quad (10)
\]

The stress dependence of these constants were then derived by linear regression of semi-logarithmic plots of the theta coefficients v. stress (fig.143.).

Due to the non-standard shape of the experimental constant-stress
curves and, in particular, the absence of a distinct region of primary creep the behaviour of the constants $\theta_1$ and $\theta_2$ were found to be particularly unstable. Due to the dominance of the tertiary portion of the creep curves it was considered a valid assumption to neglect the primary creep constants $\theta_1$ and $\theta_2$ and to model the materials deformation behaviour using the tertiary component of equation (10)

$$\varepsilon = \theta_3 \exp(\theta_4 t - 1)$$

(60)

the stress dependence of the constants again being derived from semi-log plots of theta v. stress as described above.

Initial analysis was conducted after only three constant-stress experiments had been conducted at stresses of 42, 38 and 36 MPa, the stress dependence of the $\theta_3$ and $\theta_4$ constants being found to be

$$\log_{10} \theta_3 = 0.2050\sigma - 7.205$$

(61)

$$\log_{10} \theta_4 = -0.0530\sigma - 0.329$$

(62)

When these equations were used in the modelling programme to predict the constant-load experimental results it was found that the 30 and 32 MPa constant-load results predicted identical rupture lives (fig.144.). By further reducing the initial stress these rupture lives remained essentially unchanged. The reasons for this behaviour were found to be related to the stress dependence of the theta values derived from the regression analysis. By considering equation (60) and its second derivative

$$\varepsilon'' = \theta_3 \theta_4^2 \exp(\theta_4 t)$$

(63)

it can be seen that $\theta_4$ is strongly related to the curvature and $\theta_3$ to the extent of the tertiary creep curve.

Examination of fig.143 shows $\theta_3$ to be approximately three orders of magnitude greater than $\theta_4$ at 42 MPa. However, with the extrapolation of these functions to lower stresses $\theta_4$ becomes progressively more dominant due to the negative gradient of the $\theta_4$ function and the steep positive gradient of the function of $\theta_3$. 

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This limit on rupture time under constant-load conditions has been identified both experimentally and theoretically by Evans et al (29) and Hayhurst et al (39) for ferritic low alloy steels using the same Theta-projection modelling technique. In this case, however, the limit was found to be of the order of several thousands of hours. In the light of this information and since constant-load data is known to exist for Alloy 800H up to 100,000 hours, it was concluded that the prediction of such a low limit on rupture-life could only be due to the non-representative nature of the functions of $\Theta_3$ and $\Theta_4$ used in the model.

To investigate this further, use was made of Hayhurst's approach in which improved correlation is claimed for the stress dependence of the theta values when the initial curve fitting is performed on truncated experimental data. Hayhurst et al (39) originally used a cut-off based on a constant multiple of the minimum creep-rate. This was found to be unnecessarily complicated and a modified criterion was used instead, where a cut-off criterion of 95% of the rupture-life was taken. This was suggested since the latter portions of the constant-stress curves are essentially performed under non-constant-stress conditions due to the local necking behaviour of the specimens. Additionally, Hayhurst reports that, these latter portions of the curves have quite significant effects on the theta constants.

Application of this 95% truncation technique to the determination of the theta coefficients from the ERA constant-stress results was found to have two effects: first it resulted in improved global correlation coefficients for the curve fitting routine; and second it gave improved correlation coefficients for the stress-dependence of the theta coefficients. The actual stress-dependence of the theta constants were, however, found to remain essentially constant.

In order to investigate these findings more fully, additional constant-stress experiments were conducted to extend the stress range of the data. Owing to limitations of time, only two experiments were performed at stresses of 34 and 70MPa. The 34MPa result appeared to fit well with the general shape of the other experimental curves (fig.41.). However, the 70MPa result exhibited a significant portion of primary creep which could not be explained other than through inaccuracies in the reduction in load mechanism of the machine, due to the high strain-rates.
Due to the large primary creep component observed for the 70 MPa experiment, this result was neglected as being unrepresentative of the other creep curves and modelling of the creep behaviour was, instead, only based on the 34, 36, 38 and 42 MPa results. As for the previous results the $\Theta_3$ and $\Theta_4$ coefficients were obtained for the full creep curves and for the curves truncated to 95% of their rupture-lives. By reference to fig.145., it was found that with the inclusion of the 34MPa curve, the 36 Mpa result appeared to be at odds with the other three. Without prejudging the situation, linear regression of the data was conducted for the four conditions listed below:

1) full curves, including 36 MPa result  
2) full curves, excluding 36 MPa result  
3) 95% of curves, including 36 MPa result  
4) 95% of curves, excluding 36 MPa result

The derived stress functions of $\Theta_3$ and $\Theta_4$ for each of these cases were then inserted into the tertiary strain theta function (equ.(60)) and a comparison made between the experimental and predicted constant-stress curves (figs.146.&.147.). Examination of this figure shows the best correlation to be obtained for case iv), the stress functions of the theta coefficients corresponding to this case being given by

$$\text{Log}_{10} \Theta_3 = 0.132\sigma - 4.264$$  (64)

$$\text{Log}_{10} \Theta_4 = -0.012\sigma - 1.975$$  (65)

Using these equations, constant-load creep curves were evaluated using the previously described modelling routine. Two variants of the procedure were used based on strain and time-hardening algorithms. A comparison of the results obtained from these and the experimental MS constant-load results are shown in fig.148., where it can be seen that the 30 and 32 MPa results no longer predict the same rupture lives as found for the previous analysis using equations (61) and (62).

Comparison of the strain-dependent Norton power-law model
(fig.142.) with the strain- and time-hardening Theta model curves (fig.148.) show both techniques to represent the MS experimental results quite well. The Theta models, however, predicted creep curves with more curvature than the strain-dependent Norton model, this perhaps being due to some inadequacy in the extrapolation of the Norton model to strain-levels outside the experimental range. In all cases the strain-hardening algorithm predicted lower strains for a fixed time period and longer instability rupture-times, than for the time-hardening case, the differences in the two approaches being accentuated at lower stresses. No clear conclusion could be drawn in respect of the most representative algorithm for the constant-load MS results.

As previously discussed in the section on the comparison of the MS and TT constant load results, increases in specimen gauge-lengths for fixed gauge-length cross-sectional areas, do lead to lower measured rupture-strains. Unfortunately, the results here are, to some extent, influenced by this factor since the ERA constant-stress results were performed on 50mm gauge-length and 4mm diameter specimens as compared to the MS experiments which were conducted on 30mm gauge-length and 4mm diameter specimens. This influence would be expected to give rise to modelled curves with reduced curvatures at higher strains, compared to the MS experimental data. It is difficult, however, to estimate the degree to which the model curves are influenced by this effect.

Fortunately the sensitivity of the constant-stress results, to the gauge length affect rupture strains is either partially or even completely reduced due to the 95% truncation in rupture-life which was used when the theta parameters are fitted to the constant-stress curves.

In summary, therefore, it has been found that the Norton power-law constitutive equation is unsuitable for the description of the ERA constant-stress creep curves, use of this equation in the constant-load model also resulting in poor description of the MS results.

The two other constitutive equation types investigated, the strain-dependent Norton model and the Theta model both represented the ERA constant-stress data satisfactorily.

In respect of the constant-load modelling, time- and strain-
hardening variants of the Theta model were taken, as well as the strain-dependent Norton model, all three giving reasonable representation of the MS creep curves. The strain-dependent Norton model, however, predicted constant-load curves with significantly reduced curvatures as compared to both the strain- and time-hardening Theta models, this perhaps being in some way due to the inadequacy of the extrapolation of this model to strain limits beyond those encountered in the ERA experiments.

For these reasons the Theta models were considered the most representative of the constant-stress and constant-load material behaviours for the planned use in multiaxial modelling.
4.3.2 Multiaxial

4.3.2.1 Internal pressure

To develop the uniaxial model to multiaxial conditions a number of factors must be considered.

i) the selection of a multiaxial stress criterion,

ii) the selection of a multiaxial stress v. strain-rate relationship,

and iii) the representation of the stress system in the tubes due to axial and internal pressure loading.

Each of these were examined separately and the procedure and assumptions made in making a selection of alternatives are discussed.

When considering multiaxial stress systems it is necessary to represent the equivalence of stress systems having different principal components. As already described in the literature two common theories of stress and strain equivalence are generally quoted, those according to Tresca (53) and to von Mises (52). Previous work on multiaxial creep deformation studies has found that functions of the von Mises criterion give the best correlation with experimental results (67). Although the Tresca criterion is not normally associated as being characteristic of creep deformation, both models were retained for incorporation in the modelling routine and comparison with the experimental results.

Although the equivalence of stress systems can be evaluated in this way, to evaluate the individual components of principal strain-rate it is necessary to derive stress v. strain-rate relationships representative of the flow behaviour of the material. As indicated in the literature, a number of approaches have been proposed to this end. For this modelling exercise, however, it was decided to adopt the approach used by Soderberg (54), (equations (28)).

This approach was chosen for its simplicity, the relationship containing only one material parameter in the original derivation, that of the Norton stress exponent, "n". Additionally, the relationship has the advantage that the Theta-projection constitutive
equation type, and a representation of material anisotropy can be conveniently included into its form.

Finally, consideration was made of the representation of the stress system in the tube walls. Due to the dimensions of the tubes used in the experimental programme, having an \( R_o/R_i \) ratio of approximately 1.15, some doubt was expressed as to the applicability of the thin-wall tube stress formulae and its assumption of negligible variation of stress through the wall thickness of the tubes. In this respect it was considered that redistributed creep stress solutions for thick-walled tubes under internal pressure or internal pressure and axial load would be more representative.

Two limitations were identified with respect to these approaches, first, the problem of the Norton power-law being unrepresentative of the uniaxial constant-stress experimental results and associated with this the difficulty of including the Theta-projection equations into these stress distribution equations. Secondly, it was considered that an updating routine would have to be included in the model to take account of the change in true-stress due to specimen deformation, based on the earlier findings of Taira et al. An improved representation of the material behaviour could have been achieved by the incorporation of the strain-dependent Norton power-law constitutive equation. A consideration of changing true-stress could also have been achieved if a 3-dimensional relationship could have been found or developed, relating true-stress and the principal strain components \( \epsilon_r \), \( \epsilon_z \) and \( \Theta \).

After an extensive literature search in this area no 3-dimensional relationship could be found and it was felt that the development of such a relationship for ourselves was beyond the analytical scope of the project.

The literature survey did, however, reveal a biaxial plane-stress relationship for the variation of true-stress with strain (73).

\[
\sigma_\theta = \sigma_\theta_0 (1 + \epsilon_\theta)/(1 + \epsilon_r) \quad \text{(66)}
\]

\[
\sigma_z = \sigma_z_{\text{applied}} (1 + \epsilon_z) + 0.5 \sigma_\theta \quad \text{(67)}
\]

Due to these restrictions it was felt that certainly a good approximation of the stress system could be obtained using the biaxial thin-wall equations where, at least the true-stress update
Results of the mid-specimen diametral strain v. time curves at 850 and 900°C are shown in figs.54 & 55. These show a general trend of increasing rupture-life and reduced rupture-strain with decreasing pressure, with the 850°C results exhibiting larger strains to failure than the 900°C results. Further comparison of these results, with the uniaxial constant-load tubular results (figs.44 & 45.), show the diametral strains to be much lower than the axial strains for corresponding rupture-times. This result arises partly due to the hoop strain being only one of the principal strain components, the equivalent strain being a function of all three components. Additionally, the closed ended pressurised tube is an example of a plane-strain stress system, and as discussed earlier, under these conditions reduced strains are measured due to the constraint on the material flow behaviour.

It had originally been intended to conduct modelling exercises on both the 850 and 900°C series of results. This was, however, not possible due to the lack of facilities and available time to conduct a second series of base line, constant-stress, small specimen, experiments at 850°C. Modelling was therefore only undertaken for the 900°C results.

Modelling of the internal pressure (PT) experiment results was made using the Runge-Kutta time-integration routine used for the uniaxial constant-load modelling. The Theta constitutive equation was first inserted into the Soderberg principal strain-rate equations (equations (28)) and these numerically integrated to give the principal strains corresponding to the integration step. Two variants were used where the equivalent stress was evaluated using the von Mises and the Tresca criteria.

After each integration step the hoop and axial stresses were updated using equations (66) & (67) based on the accumulated principal strains. Unfortunately, due to the complexity of the analysis, relating to the simultaneous consideration of three principal strain components, modelling was only conducted for the case of a time-hardening algorithm. Later modifications were made to the Soderberg, Tresca and von Mises equations within the model to investigate the influence of the material anisotropy on the deformation behaviour. A flow diagram for this model is shown in
Appendix 2.

Results of the von Mises and Tresca criteria model strain v. time curves and their comparison with the PT experimental results are presented in fig. 149. These show that the Tresca criterion model predicts creep curves having significantly larger creep-rates than those predicted by the von Mises model. The curvatures of these model curves are, however, seen to be reduced with respect to the experimental curves, this being the main reason for the poor correlation. This increased curvature of the experimental curves may be due to an increase in the net-section-area stress, as compared to the model, arising from the development of surface cracks as seen in the metallographic sections (figs. 68 to 71). The only conclusion that can be drawn is that the von Mises model correlates better with the lower-pressure results and the Tresca with the higher-pressure results.

4.3.2.2 Combined internal pressure and axial load

Results of the combined internal pressure and axial loaded MT experiments, are presented in figs. 61 & 62. The experimental programme was divided into two sections, the first a series of controlled stress-ratio experiments at a hoop/axial stress ratio of 1.0 with varying equivalent stress and the second a series of constant equivalent stress experiments at varying stress-ratios.

At the commencement of the programme a Finnie redistributed creep analysis (88) was used to calculate the experimental values of applied axial load and internal pressure for each experiment. These values were calculated based on an assumed von Mises equivalent stress and stress-ratio acting at the outside surface of the tube and a Norton power-law constitutive equation with an estimated n-value of 6. Due to the later realisation of the important influence on the results of the changing stress system due to specimen deformation, this calculation method was changed, the stress system corresponding to each experiment being recalculated using the thin-wall tube formulae, based on the actual experimental pressures and loads. Due to this recalculation, the constant equivalent stress levels chosen
for the variable stress-ratio series of experiments, were no longer found to be constant.

On initial examination of the results of the 1.0 stress-ratio series of experiments (fig.61.), no clear trends could be found, both in respect of the stress dependence of the rupture lives or the predominance of the axial or diametral strains. To investigate possible reasons for this behaviour, examination was made of the ratio of the maximum rupture-strain (this generally corresponding to a local bulge towards the end of the gauge-length) against the maximum measured strain at the centre of the gauge-length as measured by the diametral extensometry.

For experiments B10, B4 and 1HMT this ratio was found to be of the order of between 10:1 and 20:1, while for the other two experiments, B6 and B7 much smaller ratios were found of between 2:1 and 3:1. These results suggest that the measured diametral strains in experiments B7 and B6 may have been influenced by the local bulging of the specimen with the result that an artificially high diametral strain would have been measured. Although this explanation can account for an increase in the measured diametral strains in these experiments, it cannot account for the reduction in rupture-lives also observed.

Although the influence of the local bulge on the diametral strains appears to have affected the measured hoop strain at the centre of the gauge-length for these two experiments, it is also possible that the axial strains in all of the experiments might have been influenced by the local bulging behaviour of the tubes. Although no quantitative evidence can be brought to substantiate this point, this bulging could either be imagined to cause an increase in the axial strain due to a reduced cross-section area, or a reduction in the axial strain due to a contraction of the gauge length resulting from the bulge. Examination of the results where the diametral strains were not influenced by the local bulging, B10, B4 and 1HMT, show slightly increased axial strains compared to the diametral strains. This might be evidence for an increase in the axial strain due to specimen bulging.

Results of the MT variable stress-ratio series of experiments, B6, B9, B16 and B5 are shown in figs.62. These results show a
reduction in the rupture-life with decreasing hoop/axial stress-ratio. This may have been due to the slight increase in equivalent stress with decreasing stress-ratio, as found from the recalculation of the stress system using the thin wall formulae. There also appeared to be an increase in axial strain with decreasing stress ratio, however, in this respect the results were out of sequence, the B5 and B9 results apparently being inverted. The hoop strain curves for all experiments were very similar, that is, with the exception of the B9 result which showed a relatively high creep strain.

Examination of the ratios of the local bulge strain to the mid-specimen gauge-length strain, showed all specimens to exhibit a similar ratio of between 2:1 and 3:1. This result might have suggested an influence of the measured diametral strains from the local bulge, as suggested for the constant stress-ratio MT experimental results. This was felt unlikely, however, the result being explained by the reduced bulge strains due to the increased axial component of the stress system.

Modelling of the axial and hoop strains were undertaken using the model developed for correlation with the PT experimental results, solutions being obtained for the Tresca and von Mises equivalent stress criteria. Results of these predictions in comparison to the experimental results are shown in figs.150 to.153.

Both the von Mises and the Tresca criteria predict very similar results for the 1.0 stress-ratio series of experiments, the Tresca model predicting slightly higher strain-rates at all stresses (fig.150. &.151.). Both models predict higher initial strain-rates than those found in the experiments.

More interesting were the predictions observed for the variable stress-ratio series of experiments, since at stress-ratios other than 1.0 an increased resolution between the predictions of the von Mises and Tresca criteria was realised. The von Mises model predicted an increase in the axial strain-rate and a corresponding decrease in the hoop strain-rate with decreasing stress-ratio (fig.152.). By contrast the Tresca model predicted a more significant increase in the axial stain-rate and a corresponding increase in the hoop strain-rate (fig.153.).

It is difficult to draw a positive conclusion as to the best correlation of the experimental hoop and axial strains with the
Tresca and von Mises deformation criteria. By consideration of experiments B6, B9 and B16 the hoop strains show a trend of increasing strain-rate with decreasing stress-ratio this correlating best with the Tresca criterion model. Alternatively, consideration of the hoop strain results for experiments B6 and B5 shows a trend of decreasing strain-rate with decreasing stress-ratio, this correlating better with the von Mises criterion model.

4.3.2.3 Multiaxial summary

Comparison of the PT experimental results with the Tresca and von Mises models, showed the von Mises model to give the best correlation with the low pressure experiments A23 and the Tresca criterion to give an improved correlation with the high stress experiments A1 and A4. Metallographic examination of these high and low stress specimens (figs.68 & 70.) show a higher concentration of creep crack "damage" at the high stresses compared to the low stresses. This trend was also observed for the thick-walled internally pressurised tube specimens taken from a previous research programme at the Centre (141).

The results from this programme correspond well with the observations of Johnson (100), who concludes that where a build-up of creep damage is observed prior to failure, a function of the maximum principal stress (equivalent to the Tresca criterion in the first principal stress quadrant) controls both the tertiary creep-strain and the creep-rupture behaviour. Where no build up in metallographic "damage" is observed he concludes that a function of octahedral stress controls the tertiary creep-strain and creep-rupture behaviour.

The MT variable stress-ratio results as already indicated, show some conflicting results. If it is assumed, however, that experiment B5 is a "rogue" result, a reasonably good correlation with the Tresca criterion models is obtained. Additionally since these MT experiments were performed in a relatively high stress range, this correlation then fits with the observations of the PT model and experiments and the metallographic evidence of Johnson (100).
A transition in the deformation and rupture behaviour between high and low stress regimes, has also been suggested by other investigators. Schwab (64) and Clough and Simmons (63) have shown a trend of octahedral (von Mises) deformation dependence at low stresses and maximum principal stress at high stresses, from mathematical models based on dislocation creep mechanisms.

Henderson (142) has concluded that the von Mises stress criterion correlates best with long term, 3000 hours, and therefore low stress torsional test data produced on a different batch of Alloy 800H at 800°C.

4.3.3 Influence of anisotropy

One additional aspect which these equations did not previously take into account, was the anisotropic behaviour of the material. Consideration of an anisotropic influence was prompted by the fact that the experimental programme was conducted on a wrought tubular geometry material, which it was considered might exhibit anisotropic creep properties in each of the principal tube directions resulting from its manufacturing history.

In an attempt to derive some quantitative measure of this anisotropy, two sets of experiments were initiated: first, a series of Vicker's Diamond Pyramid hardness measurements, and second, a series of ambient temperature compression experiments. Ideally these characterisation experiments would have been best conducted in the tensile mode and at the experimental temperature, however, this was not possible due to the restricting dimensions of the as-recieved tubes.

The anisotropic deformation behaviour of materials has been investigated by Hill (143). To analyse these systems he proposes a modified form of the von Mises yield criterion

$$F(\sigma_z - \sigma_\theta)^2 + G(\sigma_\theta - \sigma_r)^2 + H(\sigma_r - \sigma_z)^2 = 1 \quad (68)$$

where $F, G$ and $H$ are constant functions of the material's yield strengths in the principal directions.
This theory was later modified by Backofen (144) and Duncombe (145) to consider a generalised yield stress $\sigma_g$ having the value of the uniaxial yield stress in the axial direction, such that

$$\sigma_g^2 = [R(\sigma_\theta - \sigma_r)^2 + RP(\sigma_z - \sigma_\theta)^2 + P(\sigma_r - \sigma_z)^2]/P(R+1) \quad (69)$$

Here the number of anisotropic constants are reduced from 3 to 2 and the values of the anisotropic coefficients $R$ and $P$ are evaluated from the relationships

$$R = \frac{(Y_{r} + Y_{z})^2 + (Y_{r} Y_{\theta})^2 - (Y_{\theta} Y_{z})^2}{(Y_{z} Y_{\theta})^2 + (Y_{r} Y_{\theta})^2 - (Y_{r} Y_{z})^2} \quad (70)$$

$$P = \frac{(Y_{r} Y_{z})^2 + (Y_{r} Y_{\theta})^2 - (Y_{\theta} Y_{z})^2}{(Y_{z} Y_{\theta})^2 + (Y_{r} Y_{\theta})^2 - (Y_{r} Y_{z})^2} \quad (71)$$

When these anisotropic considerations are taken into account the principal creep-rate equations derived by Soderberg (54) can be modified to give

$$\begin{bmatrix}
\dot{\varepsilon}_r \\
\dot{\varepsilon}_\theta \\
\dot{\varepsilon}_z
\end{bmatrix} = \frac{\dot{\varepsilon}_g}{P(R+1)\sigma_g} \begin{bmatrix}
-(R+P) & -R & -P \\
-R & R(R+1) & -RP \\
-P & -RP & P(R+1)
\end{bmatrix} \begin{bmatrix}
\sigma_r \\
\sigma_\theta \\
\sigma_z
\end{bmatrix} \quad (72)$$

To evaluate the anisotropic form of the Tresca equivalent stress function, Hill assumes that the only necessary intersection of the von Mises and Tresca equivalent stress loci occurs for the two uniaxial loading cases. Based on this assumption the anisotropic Tresca equivalent stress criterion can be described by the equations

$$\sigma_g = \sigma_z \quad (73)$$

$$\sigma_g = \sigma_\theta \left[\frac{R(R+1)}{P(R+1)}\right]^{0.5} \quad (74)$$

In order to evaluate the values of $R$ and $P$ for the material, values of the Vicker's hardness results in each of the principal
material directions were initially inserted in place of the material yield stress on the basis of the relationship between Vicker's hardness and yield stress (146)

\[ \sigma_0 = DPN \cdot k \]

\[ \sigma_0 = \frac{DPN}{3} \cdot (0.1)^{n-2.981} \]

where \( n \) is the material work-hardening coefficient having a value of between 2.0 and 2.5 and \( \sigma_0 \) is the 0.2% proof stress

Evaluation of the constant \( k \) for a value of \( n = 2.5 \) (this corresponding to a fully solution-treated material condition) gives a value of approximately 1.0 and therefore the hardness values were used directly in the evaluation of the anisotropic constants \( R \) and \( P \) to give

\[ R = 1.185 \]

\[ P = 1.235 \]

Care should be taken, however, since the value of \( k \) is acutely sensitive to the assumed value of \( n \), a value of \( n = 2.1 \) giving \( k = 2.5 \).

As a second approach, values of the compression, tangent intercept yield stress were taken directly from the small compression tests in each of the principal tube directions to calculate the anisotropic constants \( R \) and \( P \). The values of the yield stress obtained by this technique were much larger than the values obtained from the Vicker's hardness experiments although the relative magnitudes in each direction were similar (table.11). This discrepancy may have been due to one of two factors, either the inadequacy of the relationship between the DPN and the material yield stress or the raising of the apparent yield stress in the small compression experiments due to "barrelling" of the specimens. Quoted values for the room temperature yield stress in compression and tension for Alloy 800H in the hot worked and solution treated
condition are 278 MPa and 276 MPa respectively \(^{(147)}\). On this basis it was considered that the compression yield stresses could be taken as representative of the materials behaviour in tension and that the small compression specimen results gave the most representative measure of the material yield point. Due to the yield stresses derived from these compression experiments being nearer to the quoted yield stress values for the material these were used to evaluate the anisotropic parameters R and P for use in the anisotropic deformation model. The values of R and P calculated from the compression yield stresses are

\[
R = 1.330 \\
P = 1.349
\]

Results of the anisotropic modelling for the PT and MT series of experiments are shown in figs.154.to.158.

In all cases the anisotropic model predicted strain v. time curves of increased strain-rate compared to the equivalent isotropic model curves. Generally speaking the anisotropic models did not offer an improved correlation to the experimental data compared to that of the isotropic models except for the case of the constant stress-ratio MT series of experiments (fig.156.) where the correlation of the anisotropic von Mises model was satisfactory.

In conclusion it should be stressed that there is no assurance that the values of the anisotropic parameters R and P, used to model the experimental strain v. time curves, are representative of the anisotropic properties of the material. For example, the observance of longitudinally orientated TiC "stringers" in the tubes (fig.88.) may potentially give rise to tensile mode anisotropy. Whether these "stringers" contribute in the same way to the measured anistropy in the compression mode is unconfirmed.

If the characterisation of material anisotropy used here is representative of the material's tensile-mode mechanical properties then the experimental results suggest that an anisotropic von Mises criterion may be more representative of the material's creep deformation behaviour than the isotropic Tresca criterion concluded earlier.
4.4 CHARACTERISATION OF CREEP RUPTURE BEHAVIOUR

4.4.1 Establishment of the multiaxial creep rupture criterion (MCRC)

In order to establish an initial estimate of the multiaxial creep rupture criterion (MCRC) for the alloy, plots of applied stress v. log (rupture-time) were made for the TT and MS results.

A series of curves were overlayed for the PT results based on a number of different representative stress functions as shown in figs.159 & 160. The stress functions chosen were:

i) Thin-wall hoop stress
ii) Thin-wall equivalent stress
iii) Maximum redistributed thick-wall hoop stress at the outside diameter
iv) Redistributed thick-wall hoop stress at reference point
v) Maximum redistributed thick-wall equivalent stress at the inside diameter

At first sight these results show that the MS and TT curves asymptote to a thin-wall formula, maximum principal (hoop) stress function at low stresses. Whether this is a valid correlation is not clear since, alternatively, it could be considered that the uniaxial results asymptote to a line between the thick-wall redistributed maximum principal and maximum equivalent stress theories, which if valid, would suggest a mixed character rupture criterion. The validity of a comparison of the curves at low stresses and long times, and the divergence of the curves at high stresses and short times, arises from the differences in the variation of true-stress with time for the two loading conditions. Taira (73) has shown that the greatest acceleration of true-stress with time occurs for a biaxial stress ratio of 2.0 as in the internal pressure experiments (fig.12.). This explains the reduced rupture-times of the PT specimens for a given initial stress compared to the MS and TT specimens.

Due to the different rates of increase in true-stress for the two loading conditions, and due to the poor resolution between the different criteria, an alternative approach to determining the MCRC...
was investigated based on the rupture-time being evaluated by the "instability failure criterion" as used in the deformation modelling programme. Fig.161. shows the positive-positive principal stress quadrant with TT, PT and MT results plotted for rupture times of 99.9, 112 and 172.3 hours. The series of stress-rupture results from experiments conducted at fixed stress-ratios (hoop/axial) of 0, 1.0 and 2.0 were first interpolated to evaluate the equivalent stresses to give the above rupture-lives, these lives being chosen so as to include other single MT experimental results having these rupture-lives but performed at other stress-ratios. The principal stresses were then evaluated from the equivalent stress and stress-ratio and these plotted as points on axes of hoop and axial stress. In order to establish a true correlation of this data based on the increase in true-stress during the experiments, the locii of initial applied stress to cause rupture in 100 hours were plotted for the Tresca and von Mises criteria. These locii were evaluated using a modified deformation modelling programme. One of the inputs to this programme was equivalent stress and a value of this was, therefore, estimated and an iteration routine created based on the comparison of the instability rupture-life with the desired 100 hour life.

A comparison of the experimental rupture data with the Tresca and von Mises rupture ellipses, shows best correlation to be given by the Tresca criterion. Correlation over the full range of stress ratios was, however, not ideal with lower initial stresses being predicted for stress-ratios less than 0.9 and higher stresses for ratios greater than 0.9 (fig.161.). Although the resolution of the results in comparison to the two theories was improved by this representation it was felt that an improved resolution could be found.

To this end, both theory and results were plotted on axes of rupture-time v. stress-ratio (fig.162.). The theoretical lines were calculated using the deformation modelling programme for an equivalent applied stress of 38 MPa and the rupture-life results of experiments conducted at 38 MPa (as calculated by the Finnie analysis) superimposed upon these. As can be seen a much clearer correlation with the Tresca criterion is found than for the previous representations of these results and theories.

Although the general trends of this plot suggest that the Tresca criterion is the appropriate MCRC, some corrections to the
theoretical 38 MPa Tresca line on fig.162. were felt necessary due to the slight variation of the experimental equivalent stress from 38 MPa resulting from the recalculation of the equivalent stress of the MT results based on the thin-wall formulae. These corrections to the theoretical line are shown as the blue squares in fig.162. Despite this recalculation the rupture-life results still retain a strong correlation with the Tresca criterion. Interestingly, this recalculation reveals that, for stress-ratios less than approximately 0.7, the experimental rupture-life results are under-estimated by the theory, while for ratios greater than approximately 0.7 they are over-estimated. This transition in rupture behaviour correlates particularly well with the change in fracture mode of the tubes, from one of circumferential cracking and axial failure at ratios less than 0.7 to one of longitudinal cracking and tangential failure at ratios greater than 0.7.

This transition suggests that the rupture results may have been influenced by the specimen geometry, since the hoop and axial strain v. time curves calculated from both theories suggest that this transition should occur close to a stress-ratio of 1.0. Further evidence for an influence of specimen geometry on the transition in rupture behaviour comes from the earlier metallographic observation of longitudinally orientated TiC "stringer" precipitates in the tube walls. These could have contributed to a relative de-stabilising influence on longitudinally orientated cracks compared to circumferentially orientated cracks.

Such geometrical and/or metallurgical influences on crack behaviour were not expected to be identified by such an approach since the model did not attempt to represent the contribution of a metallurgical damage mechanism to the multiaxial loading conditions, the only inclusion of such an influence being through the modified stress dependence of the uniaxial constant-stress base-line data.

Although the TiC "stringers" could have influenced the cracking behaviour of the material it is also possible that they influenced the transition in failure mode due to their contribution to material anisotropy. Results of the anisotropic deformation modelling showed no such trends, however, it has already been suggested that the anisotropic characterisation experiments may not have generated a representative measure of the material anisotropy under tensile...
loading conditions. Had these results, in fact revealed a difference in the hoop and axial direction material properties, then it is conceivable that the shifted transition in failure mode would have been revealed by the anisotropic deformation modelling routine.

Results of the anisotropic deformation modelling exercise, did show some signs of limited correlation with the anisotropic form of the von Mises criterion. Since Johnson (100) has suggested that both tertiary creep deformation and rupture-life are controlled by the same stress criterion, it was decided to investigate the correlation of the anisotropic von Mises criterion to the rupture behaviour.

A plot of this anisotropic locus shows little if any improvement upon the isotropic von Mises criterion and certainly no improvement over the isotropic Tresca criterion (fig.163.). This result suggests that any correlation of the deformation behaviour with the anisotropic von Mises criterion is probably only fortuitous.
4.4.2 Influence of fracture mechanism

The previous discussion has concentrated on an empirical approach to the prediction of the representative controlling factors on creep rupture. In order to examine if these results have a more fundamental mechanistic basis two further approaches were followed. The first, investigates the controlling influences on the observed oxidation crack nucleation mechanism while the second, based on the conclusions of the first, investigates the tendency for preferential inside or outside surface cracking based on the variation of stress and strain-rate in the tube walls. These profiles were calculated using the Finnie thick-wall redistributed creep analysis.

Schultz and Ramel (133) have conducted detailed studies of the oxidation enhanced crack nucleation behaviour of Alloy 800H. Using an acoustic emission technique they have shown that there is a strain-rate dependence of crack nucleation. At low strain-rates cracking of the oxide film occurs adjacent to matrix grain boundaries, cracking occurring here due to stress concentrations associated with the sliding of grain boundaries. Rupture of this oxide film results in the oxidation of the grain boundary and the grain boundary carbides, however, as a consequence of the low strain-rate, sealing of the ruptured oxide film can occur, so arresting this internal oxidation. On further strain, rupture is again found to occur at these same sites, so increasing the depth of internal grain boundary oxidation. As a result of this successive rupturing and sealing of the oxide layer, at only a few deeply oxidised grain boundary sites, eventual specimen rupture may be realised by the unstable propagation of single cracks.

At higher strain-rates the same strain-dependent rupture of the oxide is found to occur. At these higher strain-rates, however, sealing of the oxide film is not observed due to the lateral growth-rate of the oxide being lower than the deformation strain-rate. This absence of oxide sealing and the associated increased strain-rate results in an increased number of nucleated surface cracks. The conclusion from their work is that crack nucleation behaviour is controlled by the specimen strain, the crack density being a function of the strain-rate.
These results correlate well with the present experimental observations. For example, metallographic sections taken from the low pressure, low strain-rate PT experiments, eg. A23PT, show failure to have occurred through unstable propagation of single cracks, few other cracks being observed in the section (fig.69.). By contrast at higher pressures and higher strain-rates eg. AIPT, many more surface cracks were observed (fig.71.), these cracks being found to exhibit stable propagation behaviour, due to the high strains reducing the crack-tip stress concentration factor.

During the modelling of the creep strain v. time behaviour of the tubes, thin-walled tube formulae were used to calculate the stresses in the tube walls from the applied axial loads and internal pressures. These equations model a constant stress-state through the tube wall and, as such, predict only a simplified result. To try to understand the influence of the real tube wall stress and strain-rate distributions on the crack nucleation and propagation behaviour, distributions of these were calculated using the Finnie thick-wall redistributed stress analysis (88) for each of the experimental cases.

This analysis holds one disadvantage in that it assumes a Norton power-law constitutive equation. Despite this limitation, the stress and strain-rate distributions in the tube walls can be modelled using the Norton coefficients derived from the ERA constant stress results as:-

\[ A = 2.50E-24 \]
\[ n = 10.19 \]

These principal and equivalent stress and strain-rate distributions are shown in figs.164.&.165.

The stress distributions show an increasing axial stress and a decreasing hoop stress, with decreasing hoop/axial stress-ratio as would be expected. The axial stress is always observed to be maximum at the outside surface of the tube, while the hoop stress is a maximum at the outside for high stress-ratios, changing to become a maximum at the inside at lower stress-ratios. At all stress-ratios the equivalent stress distribution is found to remain essentially
fixed, being slightly higher in value at the inside.

Metallographic examination of the PT and MT specimens shows cracking to be predominantly nucleated at the inside surfaces of the tubes for the PT specimens (stress-ratio 2.0) (figs.68.to.73.) and at the outside surfaces for all MT specimen stress-ratios (figs.77.to.84.).

A correlation of crack nucleation with principal stress is unable to offer a satisfactory correlation with the observed crack nucleation behaviour, outside surface cracking being predicted for both high and low stress-ratios. A correlation with equivalent stress is also considered unlikely due to its small variation in magnitude across the tube wall and due to the fact that this function of stress remains unchanged with changing stress-ratio.

A consideration can also be made of the relationship between the principal strain-rate distributions and the crack initiation behaviour. In the case of the internal pressure experiment the hoop strain-rate is the predominant principal strain-rate, this being a maximum at the inside surface of the tube. This correlates well with the observance of preferential cracking at the inside surface of the tubes for these internal pressure experiments.

With decreasing stress-ratio, the hoop strain-rate distribution is found to decrease in magnitude, the axial strain-rate becoming the predominant strain-rate component for stress-ratios less than 0.83. However, these axial strain-rate distributions predict a constant value of strain-rate through the tube wall and as such cannot account for preferential external surface cracking at the lower stress-ratios.

Although the calculation method for the stress distribution takes account of the stress gradient in the tube wall these stress and strain-rate distributions are somewhat idealised since the metallographic sections were taken from the local bulge regions of the tube. In these regions the stress profiles were known to be significantly modified with respect to those calculated by the Finnie analysis.

Finite element analysis of the tubular specimens under conditions of internal pressure and internal pressure and axial load has shown that these local failures near to the ends of the gauge-length are due to the presence of a hoop stress concentration in this region.
This local bulging has two significant effects on the as-calculated Finnie strain-rate distributions. First, it gives rise to a proportional increase in the internal and external hoop strain-rates and secondly, it induces an axial bending moment. This bending moment results in an increase in the axial strain-rate at the outside surface, and a decrease at the inside surface, of the tube, this modification helping to explain the tendency for external cracking under conditions of a dominant axial strain-rate.

Abo el Ata and Finnie (106) also observed surface cracking of their tubes when loaded under conditions of axial load and internal pressure. They concluded that crack initiation was controlled by a function of the von Mises equivalent stress, while a function of maximum principal stress appeared to control crack propagation. Based on these assumptions they suggested that the ratio $\sigma_E/\sigma_{\text{MAX}}$ might be a useful parameter for the characterisation of the fracture behaviour of their tubes under multiaxial stress conditions. Values of this ratio greater than 1.0, they argued, corresponded to the nucleation of many small cracks having low propagation-rates. Values of the ratio less than 1.0, however, corresponded to fewer nucleated cracks, these having an increased propagation-rate.

Values of this parameter were calculated and compared for the outside and inside surfaces of the tube, at the stress-ratios 2.0, 1.0 and 0.62.

<table>
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<tr>
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<td>&lt;1</td>
</tr>
<tr>
<td>1.0</td>
<td>&gt;1</td>
<td>=1</td>
</tr>
<tr>
<td>0.62</td>
<td>=1</td>
<td>&lt;1</td>
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</table>

The assumption was then made that preferential cracking would occur at the position in the tube wall where the value of the parameter $\sigma_E/\sigma_1$ was a maximum, this corresponding to the position of highest crack nucleation. Analysis on this basis does not correlate particularly well with the experimental observations, preferential cracking being predicted at the outside surface of the tube for each stress-ratio.
The conclusion from this analysis is that principal strain-rate and not principal stress, controls crack nucleation, this correlating with the strain-rate dependence of oxidation cracking proposed by Schultz and Ramel (132). Metallographic evidence shows crack propagation behaviour to be of a stable nature at high stresses and unstable at low stresses. Since stable crack growth correlates best with a crack opening displacement mechanism, due to the high crack-tip strains reducing the crack-tip stress concentration, it may be postulated that crack propagation behaviour at high stresses is principal strain-rate controlled. Conversely unstable crack propagation, as observed at low stresses, generally correlates best with a function of the crack-tip stress concentration factor and therefore a function of maximum principal stress.
5.0 CONCLUSIONS

- High temperature creep results obtained from solid bar specimens (MS) of Alloy 800H under constant-load conditions showed some considerable differences when compared with those obtained from tubular specimens (TT) of the same material, the tubes exhibiting lower creep-rates and lower rupture-strains than for the comparable solid bar specimen experiments. These differences were principally felt to be due to differences in stress-state in the two specimen types, however, the differences in specimen gauge length and oxidation induced grain boundary cracking were also considered to be influential but in the latter periods of the tests.

- The strain levels measured in these uniaxial constant-load experiments were too high for the assumption to be made that these experiments were conducted under near constant-stress conditions. Since no convenient technique exists for correcting constant-load data to constant-stress conditions, base line constant-stress data was derived from special constant-stress experiments. Both the Norton power-law, modified to consider the strain dependence of the creep-rate, and the Theta-projection equations gave good representations of the whole of these constant-stress creep curves. By contrast, correlations with the unmodified Norton power-law were particularly unsatisfactory.

- The uniaxial deformation model considering the update in true-stress, based on a constancy of volume criterion, predicted reasonable correlation with the (MS) constant-load experimental results using both the modified Norton power-law and Theta-projection constitutive equations. The Theta-projection model predicted creep curves with less curvature than the experimental results, but more than predicted by the Norton power-law model. For this reason, and because there was some doubt concerning the use of the modified Norton power-law above the stress and strain ranges for which the equations had originally been derived, the Theta-projection model was adopted as the base for the multiaxial modelling.
- The multiaxial deformation model, based on a thin-wall biaxial stress system with a true-stress update calculated on constancy of volume, gave a reasonable description of the creep curves acquired from the internal pressure (PT) and the combined internal pressure and axial load (MT) experiments. The curvature of the model results were less than the experimental results, particularly at high stresses due to the further reduction in section area, and the consequential increase in true-stress, arising from the observed grain boundary cracking.

- Comparison of the (PT) and (MT) experimental results with the predictions of the multiaxial model suggest that, at high stresses (>38 MPa at 900°C), the multiaxial creep deformation behaviour is controlled by the Tresca equivalent stress criterion. By contrast, results of PT experiments conducted at lower stress levels suggest that there may be a transition to a von Mises stress criterion in this region.

- Although some limited correlation was found between an anisotropic form of the von Mises stress criterion model and the MT variable stress-ratio series of experiments, this may only have been fortuitous, there being no guarantee that the experimental techniques used to characterise the material anisotropy produced results representative of the material's creep properties.

- Multiaxial creep rupture behaviour showed a strong correlation with the Tresca equivalent stress criterion over the range of experimental stress-ratios (hoop:axial = 0.33 to 2.0). Additionally, there was some evidence of an influence of specimen geometry on the creep rupture results. First, the transition from axial to hoop failure modes being shifted to lower stress ratios than expected (i.e., 0.7 as compared to 1.0) and secondly, the discontinuity in the rupture time results at this transition stress-ratio of 0.7.
Metallographic examination shows evidence of the build-up of surface cracking before failure at high stresses and the absence of cracking before failure at lower stresses. This correlates well with Johnson's observations and theory which states that, the evidence of "metallographic-damage" in sections taken from failed specimens corresponds to a function of maximum principal stress controlling tertiary creep deformation behaviour. By contrast, the evidence of no "metallographic-damage" corresponds to a function of the von Mises stress criterion controlling tertiary creep deformation.

Examination of the detailed mechanisms associated with the surface cracking behaviour, and the distribution of these cracks in the tube sections, were found to correlate better with functions of strain-rate rather than stress.
6.0 REFERENCES

1) Andrade E.N. Da.C.

2) McLean D.
   J. Inst. Metals, 1951, 80, 507;

3) Dorn J.E.

4) Weertman J.

5) Weertman J.

6) Bird J.E., Mukerjee A.K. & Dorn J.E
   Quantitative Relation Between Properties and Microstructure,
   (eds. Brandon D.G. & Rosen A.), p. 225

7) Nabbarro F.R.N.
   Report on Conf. on Strengths of Solids.

8) Herring C.

9) Coble R.L.

10) Harper J. & Dorn J.E.

    Acta. Metall., 1958, 6, 509

12) Robinson S.L. & Sherby O.D.

13) Weertman J.

14) Yavari P., Mohammed F.A. & Langdon T.G.

15) Yavari P., Millar D.A. & Langdon T.G.

16) Oikawa H.
    Creep and Fracture of Engineering Materials and Structures (Part 1),
    ( eds. Wilshire B. & Owen D.R.G.)
17) Burt H., Dennison J.P. & Wilshire B.  

18) Williams K.R. & McLauchlin I.R.  

19) Wilcox B.A. & Clauer A.H.  
Trans. AIME., 1965,236,570

20) McLauchlin I.R.  
Creep strength in Steels and High Temperature Alloys  

21) Parker J.D. & Wilshire B.  

22) Ashby M.F.  

23) Frost H.F. & Ashby M.F.  

24) Norton F.H.  
Trans. ASME., 1939,61,239.

25) Kauzman W  
Trans. AIME., 1941,143,57.

26) Cottrell A.H. & Aytekin.V.  

27) Garofalo F.  

28) Graham A. & Walles K.F.A.  
JISI, 1955,193,105

29) Evans R.W., Parker J.D. & Wilshire B.  
Recent Advances in Creep and Fracture of Engineering Materials,  
(Eds. Wilshire B. & Owen D.R.J.)  

30) Rabotnov Yu.M.  
Creep Properties in Structural Members  
(English trans. ed. Leckie F.A.), Ch.6.,  

31) Kachanov L.M.  
The Theory of Creep  
(English trans. ed. Kennedy A.J.),Ch.IX,X,  

32) Odqvist F.K.G.  
The Mathematical Theory of Creep and Creep Rupture  

33) Hayhurst D.R.  
34) Boettner R.C. & Robertson W.D.
35) Chang H.C. & Grant N.J.
    Trans. AIME, 1956,206,544.
37) McLean D.
38) Davies P.W. & Wilshire B.
40) Larson F.R. & Miller J.
41) Cane B.J. & Williams K.R.
42) Cane B.J. & Williams K.R.
43) Woodford D.A.
    Proc. Int. Conf. of Creep and Fatigue in Elevated
    Temperature Applications, Philadelphia & Sheffield,
    1974,179.1-180.1
44) Manson S.S. & Haferd A.M.
45) Monkman F.C. & Grant N.J.
    Proc. ASTM, 1956,56,593.
46) Hart R.V.
47) Randall P.N.
    J. Basic Eng., 1962,84,239.
48) Montandon R. & Kirchner F.
49) Hart R.V.
    CEGB Report SSD/NE/N 189 (1979)
50) Nishihara M., Tanaka K. & Muramutsu T.
51) Taylor G.I. & Quinney H.
    Phil. Trans. Roy. Soc., 1931,A230,323
52) von Mises R.

53) Tresca H.

54) Soderberg C.R.
Trans. ASME, 1936, 58, 733.

55) Marin J.

56) Odqvist F.G.

57) Bailey R.W.

58) Marin J.

59) Kanter J.J.

60) Nadai A.

61) Rabotnov Yu.M.
Vestnik Moskov. Univ., 1948, 10, 81.

62) Johnson A.E.

63) Clough R.B. & Simmons J.A.

64) Schwab P.R.

65) Prager W.
J. Appl. Physics, 1945, 16, 837.

66) Bocher M.
Introduction to Higher Algebra Macmillan, New York, 1907.

67) Johnson A.E.

68) Ilyashim A.A.
Prikladnaia Matematika i Mekhanica, 1945, 9.

69) Conf. on Creep Behaviour of Piping
70) Coleman M.C., Parker J.D., Walters D.J. & Williams J.A.
3rd Int. conf. on Mechanical Behaviour of Materials

71) Johnson A.E., Henderson J. & Khan B.

72) Kennedy C.R., Harms W.O. & Douglas D.A.
J. Basic Engineering, 1959,599.

73) Taira S., Ohtani R. & Ishisaka A.

74) Rodig M., Penkalla H.J. et.al.
KFA Julich GmbH COST 501 Report-Feb 1985
ISSN 0360-0885)

75) Johnson A.E, Henderson J. & Khan B.

76) Johnson A.E. & Frost N.E.

77) Johnson A.E. & Khan B.

78) Finnie I.

79) Hayhurst D.R.

80) Hayhurst D.R., Dimmer P.R. & Chernuka M.W.

81) Hayhurst D.R. & Henderson J.

82) Fairbairn J.

83) Johnson A.E.,Henderson J. & Khan B.

84) Chubb E.J. & Bolton C.J.
Int. conf. on Engineering Creep
I. Mech. E., 1980

85) Dyson B.F. & Henderson J.
Measurement of high temperature mechanical properties of materials,
(eds. Loveday M.S., Day M.F. & Dyson B.F.),

86) Lamé G.
Bachelier, Paris, 1852.
87) Taira S., Koterazawa R. & Ohtani R. 

88) Finnie I. 
J. Basic Engineering, 1960,689.

89) McGregor C.W., Coffin L.F. & Fisher J.C. 

90) Rimmrott F.P.J. 

91) King R.H. & Mackie W.W. 

92) Schulte G.A. 
Proc. ASTM., 1960,60,895.

93) Soderberg C.R. 
Trans. ASME, 1941,63,737.

94) Anderson R.G. 

95) Mackenzie A.C. 

96) Sim R.G. 

97) Fairbairn J. & Mackie W.W. 
Symp. of Creep in Structures, (Ed. Hult J.) 
Gothenburg, 1970.

98) Goodman A.M. 
CEGB Report RD/B/N2584, 1973

99) Siegfried W. 

100) Johnson A.E. & Henderson J. 
Complex-Stress Creep, Relaxation and Fracture of 

101) Browne R.J., Lonsdale D. & Flewitt P.E.J. 
Creep and Fracture of Engineering Materials and Structures, 
(Eds. Wilshire B. & Owen D.R.J.) 

102) Dyson B.F. & McLean D. 
Metal Sci., 1977,11,37.

103) Cane B.J. 
Metal Sci., 1978,12,102.

104) Cocks A.C.F. & Ashby M.F. 
105) Cane B.J.

106) Abo El Ata & Finnie I.
    Symp. of Creep in Structures, (ed. Hult J.)
    Gothenberg, 1970.

107) Degischer H.P. & Aigner H.
    Proc. Int. coloquium on Stainless Steels,
    St.Etienne, 1982

108) Orr J.
    Alloy 800

109) Tavassoli A.A. & Colombe G.

110) Persson N.G.
    Alloy 800

111) La Malfa U. & Quaranta S.
    Alloy 800

112) Bassford T.H. & Rahoi D.W.
    Alloy 800

113) Singhal L.K. & Martin J.W.

114) Tambuyser P. & Franck F.

115) Persson N.G. & Egnell L.

116) Cozar R. & Rouby M.
    Alloy 800

117) Pearce R.J.
    A Status Review of Alloy 800
    (ed. Pugh M.S.)

118) Asbury F.E. & Willoughby G.
    A Status Review of Alloy 800
    (ed. Pugh M.S.)

119) Burgel R.
    Doktor-Ingenieur thesis
    University of Hannover, 1981.
120) Desvaux M.P.E.

121) Coleman M.C., Fidler R. & Williams J.A.
Conf. on Multiaxial Creep Testing Techniques CERL/ERA, Leatherhead, UK.

122) Walters D.J.

123) Pullen D.

124) Clay B.D.
J. Mat. Sci., 1974,9,1275.

125) Tucker J.T., Coulter E.E. & Kooistra L.F.

126) Helbach P.

127) Murakami S. & Iwatsuki S.

128) Borggeen K. & Huntley P.

129) Cane B.J.

130) Baxter D.J.

131) Evans R.W. & Wilshire B.

132) McMahon C.J. & Coffin L.F.

133) Schultz M. & Rahmel A.

134) Collins P.A.
135) Pandey M.C. & Taplin D.M.R.
Proc. first Irish Durability & Fracture Conf.
Parsons Press, Dublin, 1983.

136) Cook R.H.
A Status Review of Alloy 800
(ed. Pugh M.S.)

137) Hosoi Y. & Abe S.

138) Hayhurst D.R.

139) Bressers J., Fenske E. & De Cat R.

140) Loveday M.S.
Measurement of High Temperature Material Properties,
(ed. Loveday M.S., Day M.F. & Dyson B.F.),

141) Hurst R.C.
Proceedings of the 4th Int. Conf. on the Mechanical Behaviour
of Materials, (ed. Carlsson J. & Ohlson N.G.)

142) Henderson J.

143) Hill R.
The Mathematical Theory of Plasticity,

144) Backofen W.A.
Deformation Processing,

145) Duncombe E., Friedrich E.M. & Guilinger W.H.
Nucl. Tech., 1971,12,194.

146) Dieter G.
Mechanical Metallurgy

147) Henry Wiggin Co. Ltd.
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Strain

Time / Hours
fig. 17. MS specimen strain vs. time results at 1000°C.

STRAIN

0.00 1.00 2.00 3.00 4.00 5.00 6.00

TIME / HOURS

Stress in MPa

16

12
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Diametral Strain

Rupture Time / Hours

Bulge Strain 850°C
Bulge Strain 900°C
Centre Strain 850°C
Centre Strain 900°C
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<td>B10MT</td>
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Stress in MPa
Fig. 62. Axial and diametral strain v. time results for the variable stress-ratio series of MT experiments.

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<th>EQUIVALENT STRESS</th>
</tr>
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<tbody>
<tr>
<td>B6MT</td>
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<td>37.20</td>
</tr>
<tr>
<td>B9MT</td>
<td>0.83</td>
<td>37.88</td>
</tr>
<tr>
<td>B5MT</td>
<td>0.71</td>
<td>38.48</td>
</tr>
<tr>
<td>B16MT</td>
<td>0.62</td>
<td>38.94</td>
</tr>
</tbody>
</table>

Stress in MPa
fig.63. Failed specimens from the MT series of experiments.
fig. 64. Longitudinal macro-sections of MS specimens tested at 850°C.
fig.65. Longitudinal macro-sections of TT specimens tested at 850°C.
fig 66. Longitudinal macro-sections of MS specimens tested at 900°C.
fig. 67. Longitudinal macro-sections of TT specimens tested at 900°C.
fig. 68. Transverse and longitudinal macro-sections taken from specimen Al2PT.
fig. 69. Transverse and longitudinal macro-sections taken from specimen A23PT.
fig. 70. Transverse and longitudinal macro-sections taken from specimen A01PT.
fig. 71. Transverse and longitudinal macro-sections taken from specimen A04PT.
fig. 72. Transverse and longitudinal macro-sections taken from specimen Al4PT.
fig.73. Transverse and longitudinal macro-sections taken from specimen B13PT.
fig.74. Transverse macro-section of a thick-walled tube specimen taken from previous work at the centre (141).
fig. 75. Transverse macro-section of a thick-walled tube specimen taken from previous work at the centre (141).
fig. 76. Transverse macro-section of a thick-walled tube specimen taken from previous work at the centre (141).
fig. 77. Transverse and longitudinal macro-sections taken from specimen B10MT.
fig. 78. Transverse and longitudinal macro-sections taken from specimen B06MT.
fig. 79. Transverse and longitudinal macro-sections taken from specimen B04MT.
fig. 80. Transverse and longitudinal macro-sections taken from specimen B07MT.
fig. 81. Transverse and longitudinal macro-sections taken from specimen 1HMT.
fig. 82. Transverse and longitudinal macro-sections taken from specimen B09MT.
fig. 83. Transverse and longitudinal macro-sections taken from specimen B05MT.
fig. 84. Transverse and longitudinal macro-sections taken from specimen B16MT.
fig.85. Longitudinal metallographic section of solution-treated material (x200).

fig.86. Transverse metallographic section of solution-treated material (x200).
figs. 87a) & b). Grain boundary and matrix TiC precipitates in solution-treated material (x500).
Fig. 88. Longitudinal metallographic section showing axially orientated TiC "stringers" in solution-treated material (x200).
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fig. 90. Typical surface cracks nucleated at the outside surface of the tubular specimens (x500).
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fig. 92. r-type surface cracking at grain boundaries as observed in specimen A03TT (x500).

fig. 93. r-type surface cracking at grain boundaries as observed in specimen A03TT (x500).
Fig. 94. Inside surface recrystallisation caused by the machining operation as observed in specimen A23PT (x500).

Fig. 95. Outside surface recrystallisation caused by the machining operation as observed in specimen A04PT (x200).
fig. 96. Influence of TiC "stringers" on crack propagation as observed in specimen A03TT (x200).
fig. 97. SEM fractograph from experiment A03TT conducted at 900°C and 29.3 MPa.

fig. 98. SEM fractograph from experiment A21TT conducted at 900°C and 42.0 MPa.
fig. 99. SEM fractograph from experiment 03MS conducted at 900°C and 29.8 MPa.

fig. 100. SEM fractograph from experiment 09MS conducted at 900°C and 42.0 MPa.
fig. 101. SEM fractograph from experiment Al9TT conducted at 850°C and 40.0 MPa.
fig.102. SEM fractograph from experiment A20TT conducted at 1000°C and 12 MPa.
fig.103. SEM fractograph from experiment 15MS conducted at 850°C and 48 MPa.

fig.104. SEM fractograph from experiment A17TT conducted at 850°C and 48 MPa.
fig.105. SEM fractograph from experiment 03MS showing oxidised grain boundary surface crack.
fig.106. Local concentration of dislocations adjacent to grain boundaries in the as-recieved material.

fig.107. Local concentration of dislocations adjacent to intragranular TiC precipitates in the as-recieved material.

fig.108. Grain boundary Cr$_{23}$C$_6$ precipitates in as-recieved material.
fig.109. Early stages of network formation in the as-recieved and solution treated material.

fig.110. Grain boundary TiC precipitates in the as-recieved and solution treated condition.
fig.111. Large grain boundary and small intragranular $\text{Cr}_2\text{C}_6$ in material aged for 885 hours at 1000$^\circ$C.

fig.112. Small intragranular $\text{Cr}_2\text{C}_6$ precipitates (200–300Å) in material aged for 885 hours at 1000$^\circ$C.

fig.113. Dislocations pinned by small $\text{Cr}_2\text{C}_6$ precipitates in material aged for 885 hours at 1000$^\circ$C.
fig.114. Intragranular TiC precipitates in material aged for 885 hours at 1000°C.

fig.115. Grain boundary TiC precipitates in material aged for 885 hours at 1000°C.
fig. 116. Intragranular TiC precipitates surrounded by Cr$_{23}$C$_6$ heterogeneously precipitated on the dislocation structure in material aged for 1013 hours at 900°C.

fig. 117. Intragranular Cr$_{23}$C$_6$ precipitates in material aged for 900 hours at 850°C.

fig. 118. Grain boundary TiC surrounded by Cr$_{23}$C$_6$ precipitate in material aged for 1013 hours at 900°C.
fig.119. Finite element model and mesh
fig. 120: Elastic hoop stress distributions along gauge length for three hoop:axial stress-ratios.
fig. 121. Elastic axial stress distributions along gauge length for three hoop:axial stress-ratios.
The finite element and finite element mesh distributions along the specimen gauge length, as calculated by

Fig. 1.22: Variation of the elastic and creep redistributed hoop stress.
fig. 123. Variation of the elastic and creep-redistributed axial stress distributions along the specimen gauge length, as calculated by the finite element and Finnie analyses for a pressurised (55 Bar) tube tested at 900°C.
fig.124. Variation of the elastic and creep redistributed hoop stress distributions along the specimen gauge length, as calculated by the finite element and Finnie analyses for an axially loaded (2.94 kN) and pressurised (60.7 Bar) tube tested at 900°C.
fig. 125. Variation of the elastic and creep-redistributed axial stress distributions along the specimen gauge length, as calculated by the finite element and Finnie analyses for an axially loaded (2.94 kN) and pressurised (60.7 Bar) tube tested at 900°C.
fig.126. Variation of the elastic and creep-redistributed hoop and axial stress distributions through the wall thickness, at the stress concentration and at the centre of the gauge length, as calculated by the finite element and Finnie analyses for a pressurised (55 Bar) tube tested at 900°C.
fig. 2. Stress vs. strain curves for small compression specimens tested in the axial, radial and tangential directions.
fig. 128. Comparison of ST, MS and Data Bank search profile (I) creep rupture data.
fig. 129. Comparison of ST, MS and Data Bank search profile (1) creep elongation data.
fig. 130. Comparison of ST, MS and Data Bank search profile (1) minimum strain-rate data.
fig. 131: Comparison of MS and TT strain v. time results at 850°C.
fig. 132. Comparison of MS and TT strain vs. time results at 900°C.
APPLIED STRESS / MPa

Rupture time / Hours

1000°C  900°C  850°C

MS and II stress rupture results at 850, 900 and 1000°C.
Fig. 134. MS and TT elongation to rupture results at 850, 900 and 1000°C.
fig. 135. Idealised MS and TT strain v. time results for high and low stresses at 850°C.

fig. 136. Idealised MS and TT strain v. time results for high and low stresses at 900°C.
fig.13. MS and TT axial strain and reduction in area 'w', applied stress and strain at 900°C.
fig. 138. Effect of specimen diameter on the strain v. time behaviour of a 1/2%Cr-Mo-V steel creep tested in air (After Cane (129)).
Fig. 139. Comparison between CS constant-stress and MS constant-load experimental results.

Stress in MPa

0.00 1.00 2.00 3.00 4.00 $E+2$

TIME / HOURS

0.00 1.00 2.00 3.00

STRAIN

42 38 38 36

MS CS
Experimental results.

Fig. 140: Comparison between ERA, constant-stress and MP constant-load.
predicted constant-load results using the Norton power-law model.

Figure 14.4: Comparison between NS constant-load experimental results and...

**Model**

**Experimental**
36, 38 and 42 MPa.

Stress for the BPA constant stress experiments conducted at

FIG. 143. Semi-logarithmic plots of theta coefficients applied

STRESS / MPa

3.00

3.50

4.00

4.50

5.00

E+1

STRESS / MPa

10^{-4}

10^{-3}

10^{-2}

10^{-1}

10^{0}

10^{1}

10^{2}

THETA COEFFICIENTS
fig. 144. Comparison between MS constant-load experimental results and predicted constant-load results using the time and strain hardening Theta-projection models based on equations (61) & (62).
3.4, 3.6, 3.8 and 4.2 MPa.

Stresses for the EMA constant-stress experiments conducted at

Fig. 14.5. Semi-logarithmic plots of Theta coefficients vs. applied

Theta 4
Theta 1
Theta 2
Theta 3

STRESS / MPa

3.00
3.50
4.00
4.50
5.00
6.00

E+1

100%
95%

Fig. 14.6. Comparison between FEA constant-stress experimental results and the predicted constant-stress results using the Theta-100% extrapolation model.
Figure 14.7: Comparison between FEA constant-stress experimenatal results and the predicted constant-stress results using the Thea projection model.
fig. 1.6. Comparison between MS constant-load experimental results and predicted constant-load results using the time and strain hardening Theta-projection models based on equations (64) & (65).

<table>
<thead>
<tr>
<th>Stress in MPa</th>
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<tbody>
<tr>
<td>29.8</td>
</tr>
<tr>
<td>32</td>
</tr>
<tr>
<td>36</td>
</tr>
<tr>
<td>38</td>
</tr>
<tr>
<td>42</td>
</tr>
</tbody>
</table>

**Axes:**
- **Y-axis:** Strain
- **X-axis:** Time / Hours

**Graph Key:**
- **EXPERIMENT**
- **STRAIN HARDENING**
- **TIME HARDENING**
hardening models.

The predicted results using von Mises and Tresca time

comparison between the 900°C PR experimental results and

Figure 14.9.
<table>
<thead>
<tr>
<th>CODE</th>
<th>STRESS RATIO</th>
<th>EQUIVALENT STRESS</th>
</tr>
</thead>
<tbody>
<tr>
<td>△ B7MT</td>
<td>1.00</td>
<td>34.28</td>
</tr>
<tr>
<td>○ IHMT</td>
<td>1.00</td>
<td>34.31</td>
</tr>
<tr>
<td>● B4MT</td>
<td>1.00</td>
<td>35.25</td>
</tr>
<tr>
<td>□ B6MT</td>
<td>1.00</td>
<td>37.20</td>
</tr>
<tr>
<td>■ B10MT</td>
<td>1.00</td>
<td>41.12</td>
</tr>
</tbody>
</table>

- **Hoop strain**
- **Axial strain**
Fig. 150. Comparison between the 900°C MT 1.0 constant stress-ratio experimental results and the predicted results using the von Mises time-hardening model.
<table>
<thead>
<tr>
<th>CODE</th>
<th>STRESS RATIO</th>
<th>EQUIVALENT STRESS</th>
</tr>
</thead>
<tbody>
<tr>
<td>△ B7MT</td>
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<td>34.28</td>
</tr>
<tr>
<td>○ B1HMT</td>
<td>1.00</td>
<td>34.31</td>
</tr>
<tr>
<td>● B4MT</td>
<td>1.00</td>
<td>35.25</td>
</tr>
<tr>
<td>□ B6MT</td>
<td>1.00</td>
<td>37.20</td>
</tr>
<tr>
<td>■ B10MT</td>
<td>1.00</td>
<td>41.12</td>
</tr>
</tbody>
</table>

Hoop strain

Axial strain
Figure 1.12. Comparison between the 900°C M1:0 constant stress ratio triangular time-hardening model and experimental results using the 900°C M1:0 constant stress ratio.
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<tr>
<th>Equivalent Stress Ratio</th>
<th>Code</th>
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<td>0.1</td>
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<tr>
<td>0.8</td>
<td>●</td>
</tr>
<tr>
<td>1.0</td>
<td>□</td>
</tr>
<tr>
<td>0.62</td>
<td>■</td>
</tr>
</tbody>
</table>

**Hoop strain**

**Axial strain**
von Mises time-hardening model.

Experimental results and the predicted results using the

Fig. 15.2. Comparison between the 900°C MTS variable stress ratio

AXIAL / DIA. STRAIN

TIME / HOURS

DIAMETRAL STRAIN

AXIAL STRAIN
<table>
<thead>
<tr>
<th>Stress Equivalent Code</th>
<th>Ratio</th>
<th>Stress</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.720</td>
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<td>15.0</td>
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<tr>
<td>0.70</td>
<td>0.80</td>
<td>13.5</td>
</tr>
<tr>
<td>0.78</td>
<td>0.98</td>
<td>12.0</td>
</tr>
<tr>
<td>1.00</td>
<td>0.94</td>
<td>10.5</td>
</tr>
</tbody>
</table>

- Hoop stress
- Axial stress
Fig. 15. Comparison between the 900°C M1 variable stress ratio and experimental results using the isotropic time-hardening model.
<table>
<thead>
<tr>
<th>CODE</th>
<th>STRESS RATIO</th>
<th>EQUIVALENT STRESS</th>
</tr>
</thead>
<tbody>
<tr>
<td>△ B7MT</td>
<td>1.00</td>
<td>34.28</td>
</tr>
<tr>
<td>○ 1HMT</td>
<td>1.00</td>
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<td>35.25</td>
</tr>
<tr>
<td>☐ B6MT</td>
<td>1.00</td>
<td>37.20</td>
</tr>
<tr>
<td>■ B10MT</td>
<td>1.00</td>
<td>41.12</td>
</tr>
</tbody>
</table>

- **Hoop strain**
- **Axial strain**
fig. 154. Comparison between the 900°C PT experimental results and the predicted results using anisotropic von Mises and Tresca time-hardening models.
anisotropic von Mises time-hardening model.
Experimental results and the predicted results using the
Fig. 15. Comparison between the 900°C M.T. 1.0 constant stress ratio
<table>
<thead>
<tr>
<th>CODE</th>
<th>STRESS RATIO</th>
<th>EQUIVALENT STRESS</th>
</tr>
</thead>
<tbody>
<tr>
<td>△ B7MT</td>
<td>1.00</td>
<td>34.28</td>
</tr>
<tr>
<td>○ IHMT</td>
<td>1.00</td>
<td>34.31</td>
</tr>
<tr>
<td>● B4MT</td>
<td>1.00</td>
<td>35.25</td>
</tr>
<tr>
<td>□ B6MT</td>
<td>1.00</td>
<td>37.20</td>
</tr>
<tr>
<td>■ B10MT</td>
<td>1.00</td>
<td>41.12</td>
</tr>
</tbody>
</table>

---

- **Hoop strain**
- **Axial strain**
Anisotropic fracture time-hardening model.
Experimental results and the predicted results using the
Fig. 1.56. Comparison between the 900°C MT1.0 constant stress ratio
<table>
<thead>
<tr>
<th>CODE</th>
<th>STRESS RATIO</th>
<th>EQUIVALENT STRESS</th>
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<tbody>
<tr>
<td>B6MT</td>
<td>1.00</td>
<td>37.20</td>
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<tr>
<td>B9MT</td>
<td>0.83</td>
<td>37.88</td>
</tr>
<tr>
<td>B5MT</td>
<td>0.71</td>
<td>38.48</td>
</tr>
<tr>
<td>B16MT</td>
<td>0.62</td>
<td>38.94</td>
</tr>
</tbody>
</table>

Diagram:
- Hoop strain
- Axial strain
Comparison between the 90°C MP variable stress ratio experimental results and the predicted results using the anisotropic von Mises time-hardening model.
CODE | STRESS RATIO | EQUIVALENT STRESS
---|-------------|-----------------|
B6MT | 1.00        | 37.20           |
B9MT | 0.83        | 37.88           |
B5MT | 0.71        | 38.48           |
B16MT| 0.62        | 38.94           |

Hoop strain

Axial strain
antisotropic true stress-hardening model. Experimental results and the predicted results using the comparison between the 900°C MTR variable stress ratio.
for the listed stress functions.

Figure 1.59. Comparison between the 850°C uniaxial MS and TT creep rupture results calculated in

\[ \text{STRESS / MPa} \]

\[ \begin{array}{cccccc}
 3.00 & 3.50 & 4.00 & 4.50 & 5.00 & 5.50 \\
 3.00 & 4.00 & 5.00 & 6.00 & 7.00 & 8.00 \\
\end{array} \]

\[ \text{RUPTURE TIME / HOURS} \]

\[ \begin{array}{cccccc}
 10^1 & 10^2 & 10^3 & 10^4 & 10^5 & 10^6 \\
 10^1 & 10^2 & 10^3 & 10^4 & 10^5 & 10^6 \\
\end{array} \]

- MS
- TT Thin-Wall Hoop Stress
- Thick-Wall Equivalent Stress
- Thick-Wall Redistributed Equivalent Stress
- Reference Stress
for the fitted stress functions.

Figure 160. Comparison between the 900°C uniaxial MS and TT creep rupture results calculated.
correlation of these criteria to the experimental results.

stress-ratio for the von Mises and Tresca criteria and the comparison between the variation of rupture-time with

Figure 16.2. Comparison between the variation of rupture-time with

\[\frac{\text{rupture time}}{\text{hours}}\]

\[\frac{\text{stress ratio (hoop/axial)}}{\text{stress ratio (hoop/axial)}}\]

- Von Mises
- Tresca
- Corrected Tresca
- Experimental Data
fig. 6.3. Comparison between the isotropic Tresca and von Mises 100 hour rupture loci and the TR, PT, and MT experimental results.
fig. 164. Stress distributions in the tube walls as calculated by the Finnie thick-wall redistributed creep analysis.
fig.165. Strain-rate distributions in the tube walls as calculated by the Finnie thick-wall redistributed creep analysis.
Table 1. Compositions of experimental batches of Alloy 800H in relation to the alloy's compositional limits.

<table>
<thead>
<tr>
<th>ELEMENT</th>
<th>ALLOY 800</th>
<th>ALLOY 800H</th>
<th>BATCH I</th>
<th>BATCH II</th>
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<tr>
<td>Ni</td>
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<td>30.0–35.0</td>
<td>30.77</td>
<td>31.50</td>
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<tr>
<td>Cr</td>
<td>19.0–23.0</td>
<td>19.0–23.0</td>
<td>19.94</td>
<td>19.98</td>
</tr>
<tr>
<td>Fe</td>
<td>Bal</td>
<td>Bal</td>
<td>Bal</td>
<td>Bal</td>
</tr>
<tr>
<td>C</td>
<td>0.1 max</td>
<td>0.05–0.10</td>
<td>0.06</td>
<td>0.08</td>
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<tr>
<td>Mn</td>
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<td>1.5 max</td>
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<td>0.76</td>
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<tr>
<td>S</td>
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<td>0.015 max</td>
<td>0.004</td>
<td>0.006</td>
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<tr>
<td>Si</td>
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<td>1.0 max</td>
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<td>0.29</td>
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<td>Cu</td>
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<td>0.75 max</td>
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<tr>
<td>Al</td>
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<td>0.15–0.60</td>
<td>0.28</td>
<td>0.39</td>
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<tr>
<td>Ti</td>
<td>0.15–0.60</td>
<td>0.15–0.60</td>
<td>0.39</td>
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Table 2. Influence of absolute carbon concentration as well as Ti:C ratio on the % dissolved carbon at 980°C.

<table>
<thead>
<tr>
<th>% CARBON IN STEEL</th>
<th>% DISSOLVED CARBON AT 980°C</th>
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<tbody>
<tr>
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<td>Ti:C = 12</td>
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<tr>
<td>0.010</td>
<td>0.010</td>
</tr>
<tr>
<td>0.015</td>
<td>0.014</td>
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<tr>
<td>0.020</td>
<td>0.011</td>
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<tr>
<td>0.025</td>
<td>0.010</td>
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<tr>
<td>0.030</td>
<td>0.008</td>
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Table 3. Data Bank search profile code

<table>
<thead>
<tr>
<th>Property</th>
<th>1) creep rupture</th>
<th>2) elongation to rupture</th>
<th>3) minimum creep rate</th>
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<tr>
<td>Composition</td>
<td>carbon</td>
<td>aluminium</td>
<td>titanium</td>
</tr>
<tr>
<td></td>
<td>0.05-0.10</td>
<td>0.15-0.60</td>
<td>0.15-0.60</td>
</tr>
<tr>
<td></td>
<td>0.05-0.10</td>
<td>0.20-0.39</td>
<td>0.20-0.39</td>
</tr>
<tr>
<td></td>
<td>0.05-0.10</td>
<td>0.20-0.39</td>
<td>0.40-0.59</td>
</tr>
<tr>
<td></td>
<td>0.05-0.059</td>
<td>0.20-0.39</td>
<td>0.30-0.49</td>
</tr>
<tr>
<td></td>
<td>0.06-0.069</td>
<td>0.20-0.39</td>
<td>0.30-0.49</td>
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<tr>
<td></td>
<td>0.07-0.079</td>
<td>0.20-0.39</td>
<td>0.30-0.49</td>
</tr>
<tr>
<td></td>
<td>0.08-0.089</td>
<td>0.20-0.39</td>
<td>0.30-0.49</td>
</tr>
<tr>
<td></td>
<td>0.075-0.085</td>
<td>0.35-0.45</td>
<td>0.50-0.60</td>
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</tbody>
</table>

<table>
<thead>
<tr>
<th>Temperature</th>
<th>1) 700 ± 5°C</th>
<th>2) 800 ± 5°C</th>
<th>3) 850 ± 5°C</th>
<th>4) 900 ± 5°C</th>
<th>5) 1000 ± 5°C</th>
</tr>
</thead>
</table>

<p>| Heat-treatment temperature | 1) 1080 - 1200°C | 2) 1120 - 1160°C          |</p>
<table>
<thead>
<tr>
<th>SEARCH PROFILE</th>
<th>PROPERTY</th>
<th>COMPOSITION</th>
<th>TEMPERATURE</th>
<th>HEAT TREATMENT</th>
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<td></td>
<td>1 2 3 4 5</td>
<td>1</td>
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<td>1 2 3 4 5</td>
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Table 5. Small solid-bar constant-load experiment results.

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<td>++</td>
</tr>
<tr>
<td>B18MT</td>
<td>900</td>
<td>67.20</td>
<td>1086</td>
<td>60.0</td>
<td>++</td>
<td>++</td>
<td>++</td>
</tr>
<tr>
<td>B19MT</td>
<td>900</td>
<td>63.85</td>
<td>2288</td>
<td>126.0</td>
<td>++</td>
<td>++</td>
<td>++</td>
</tr>
</tbody>
</table>
Table II. Results of Hardness and Compression Experiments.

<table>
<thead>
<tr>
<th>Direction</th>
<th>Hardness of Electropolished Samples (DPN)</th>
<th>Hardness of Mechanically Polished Samples (DPN)</th>
<th>Tangent Intercept Yield stress in compression (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HOOP</td>
<td>130</td>
<td>158.3</td>
<td>338</td>
</tr>
<tr>
<td>AXIAL</td>
<td>128.8</td>
<td>159.0</td>
<td>337</td>
</tr>
<tr>
<td>RADIAL</td>
<td>136</td>
<td>165.3</td>
<td>365</td>
</tr>
</tbody>
</table>
Table 12. Surface and grain boundary oxidation depth data under stress-free conditions. (after Baxter (130))

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Time (h)</th>
<th>Oxide Scale Thickness (um)</th>
<th>Oxidation Depth (um)</th>
</tr>
</thead>
<tbody>
<tr>
<td>850</td>
<td>144</td>
<td>3</td>
<td>10</td>
</tr>
<tr>
<td>900</td>
<td>144</td>
<td>5</td>
<td>15</td>
</tr>
<tr>
<td>950</td>
<td>144</td>
<td>8</td>
<td>35</td>
</tr>
<tr>
<td>1000</td>
<td>5</td>
<td>4</td>
<td>10</td>
</tr>
<tr>
<td>1000</td>
<td>20</td>
<td>6</td>
<td>10</td>
</tr>
<tr>
<td>1000</td>
<td>65</td>
<td>8-10</td>
<td>45</td>
</tr>
<tr>
<td>1000</td>
<td>144</td>
<td>10-12</td>
<td>60</td>
</tr>
<tr>
<td>1000</td>
<td>336</td>
<td>12-15</td>
<td>70</td>
</tr>
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</table>
### Table 13. Variation of strain-rate with strain used in the derivation of the strain dependent Norton power-law coefficients.

<table>
<thead>
<tr>
<th>STRAIN (MPa)</th>
<th>34</th>
<th>36</th>
<th>38</th>
<th>42</th>
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</thead>
<tbody>
<tr>
<td>Minimum</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.01</td>
<td>1.0E-8</td>
<td>8.4E-9</td>
<td>1.4E-8</td>
<td>8.7E-8</td>
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<tr>
<td>0.02</td>
<td>3.9E-8</td>
<td>4.3E-8</td>
<td>6.7E-8</td>
<td>1.7E-7</td>
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<tr>
<td>0.03</td>
<td>5.3E-8</td>
<td>6.3E-8</td>
<td>1.0E-7</td>
<td>2.1E-7</td>
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<td>0.04</td>
<td>5.8E-8</td>
<td>8.2E-8</td>
<td>1.1E-7</td>
<td>2.3E-7</td>
</tr>
<tr>
<td>0.05</td>
<td>6.6E-8</td>
<td>1.1E-7</td>
<td>1.4E-7</td>
<td>2.3E-7</td>
</tr>
<tr>
<td>0.06</td>
<td>8.8E-8</td>
<td>1.1E-7</td>
<td>1.5E-7</td>
<td>2.4E-7</td>
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<tr>
<td>0.07</td>
<td>9.7E-8</td>
<td>1.3E-7</td>
<td>1.6E-7</td>
<td>2.5E-7</td>
</tr>
<tr>
<td>0.08</td>
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<td>1.5E-7</td>
<td>1.7E-7</td>
<td>2.6E-7</td>
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<tr>
<td>0.09</td>
<td>1.2E-7</td>
<td>1.9E-7</td>
<td>2.7E-7</td>
<td></td>
</tr>
</tbody>
</table>
9.0 APPENDICES

Appendix 1. Uniaxial constant-load deformation modelling programme.

START

INPUT AXIAL STRESS

CALCULATE AXIAL STRAIN BY THE INTEGRATION OF EQUATION (56) OR (60) USING A 4th ORDER RUNGE-KUTTA ROUTINE

IS STRAIN < 30%

UPDATE STRESS BASED ON EQUATION (57)

OUTPUT AXIAL STRAIN v. TIME CURVE

STOP
Appendix 2. Multiaxial deformation modelling programme.

START

INPUT HOOP AND AXIAL STRESSES

SELECT EQUIVALENT STRESS CRITERION

CALCULATE EQUIVALENT STRESS AND STRESS RATIO

CALCULATE THE EXTERNAL APPLIED LOAD

CALCULATE PRINCIPAL STRAINS FROM ISOTROPIC OR ANISOTROPIC SODERBERG EQUATIONS (28) OR (72) USING RUNGE-KUTTA NUMERICAL INTEGRATION ROUTINE AND EQUATION (60) AS THE CREEP CONSTITUTIVE EQUATION

ARE ANY PRINCIPAL STRAINS > 30%?

UPDATE STRESSES BASED ON PRINCIPAL STRAINS [EQUATIONS (66) & (67)]

RE-CALCULATE EQUIVALENT STRESS FROM ISOTROPIC OR ANISOTROPIC EQUATIONS

OUTPUT HOOP, AXIAL AND EQUIVALENT STRAIN v. TIME CURVES

STOP