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Thermomechanical Analysis of P91 Power Plant Components

by

Tadhg Farragher

B.E., University of Limerick, 2007.

A thesis submitted to the National University of Ireland, Galway
as fulfilment of the requirements for the Degree of Doctor of Philosophy

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Abstract

This thesis presents a combined experimental and computational study on the thermo-mechanical fatigue performance of welded P91 material to (i) characterise the thermomechanical behaviour, and (ii) predict the number of cycles to fatigue crack initiation, for power plant P91 material (including welded material) and components, operating under realistic loading conditions.

A programme of isothermal high temperature low cycle fatigue (HTLCF) strain- and stress-controlled and thermo-mechanical fatigue tests on service-aged (SA) P91 base metal, weld metal and welded specimens was conducted to characterise the thermo-mechanical performance of the P91 parent, weld metal and heat-affected zone materials. Novel cross-weld test specimens, as well as weld and parent metal specimens, were manufactured here from a specially-fabricated P91 weld repair header pipe, in collaboration with the plant operator.

A sequentially coupled thermomechanical simulation methodology, using realistic temperature and pressure loading histories, was developed within a general-purpose, non-linear, finite element code to predict and analyse the thermomechanical behaviour of P91 power plant components.

An anisothermal cyclic viscoplasticity material model for P91 parent material was developed and calibrated for both service-aged and as-new material, using the tests conducted within this thesis and published data. This constitutive model including isotropic softening, non-linear kinematic hardening and viscoplasticity (Norton power-law creep) terms to simulate the complex evolution of material hysteresis response. This calibrated material model formed the basis for a sequential thermo-mechanical methodology of power plant components, namely, a plain pipe
and a branched connection from a fossil fuel plant in Ireland. This methodology also includes a transient heat transfer phase with sequential cyclic thermo-mechanical phase to simulate (i) a plant start-up, and (ii) a load-following scenario, based on measured data to identify key damaging events caused during plant operation. The transient heat transfer model was calibrated and validated against the measured plant start-up cycle data. For the branched connection, a global-sub-modelling framework was employed for detailed mesh refinement to identify the local thermomechanical stress-strain response at critical locations.

A multiaxial, rain flow cycle counting methodology was developed for thermo-mechanical fatigue prediction using a critical-plane approach and applied to predict thermomechanical fatigue crack initiation at critical locations of the branched pipe sub-models. The results were shown to be consistent with plant operator experience and previously published experimental findings for similar plant connections.

A key contribution of the present thesis is the development of a methodology for identification of the high temperature cyclic viscoplasticity parameters for the weld material and heat-affected zone, including isotropic softening, non-linear kinematic hardening and viscoplasticity terms. Theoretical and finite element models of the parent, weld and cross-weld test specimens were employed for identification of the cyclic viscoplasticity material parameters, via direct comparison with the measured hysteresis evolutions for the parent, weld and cross-weld specimens.

The methodology developed, in direct collaboration with a power plant operator, provides a comprehensive framework for life assessment of existing, retrofitted and proposed new plant, for coal-fired and gas-fired operation. It will specifically allow assessment of the impact of more flexible plant operation to allow for renewable energy uptake and energy cost fluctuations and provides a framework
for extension to future ultra-supercritical operation scenarios. This work also provides a more rational basis for experimental thermomechanical fatigue characterisation of candidate materials, as well as, design of tests for welded connections under power plant loading conditions.
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Chapter 1

INTRODUCTION

1.1 General

Worldwide electricity consumption is expected to increase three fold over the next twenty years. Currently, in Ireland about 20% of our electricity consumption comes from renewable sources - mostly wind farms - one of the highest levels in the EU [CER 2011], as illustrated in Figure 1.1. In 2011 70% of electricity generation came from imported fuels of which 55% was natural gas, 15% was coal and 0.29% was oil. Peat as an indigenous source of fuel provided 8% of generation requirements. However, under the Kyoto Protocol, the European Union has committed itself to reducing the equivalent carbon dioxide emissions by 8% of the 1990 level by the end of 2012. In order to meet this objective, the member states have encouraged, financially and otherwise, the development of renewable energy, especially wind power. Around 75% of the world wind power capacity is installed in Europe. Locally, this results in some of the highest wind power penetration levels in the world [Estanqueiro et al., 2008]. The main driver of wind power development has been the so-called fixed feed-in tariff for wind power, which is used in Germany, Denmark, and Spain and Ireland [CER 2007/2008]. Such fixed feed-in tariffs reduce the financial risk for wind power investors because the power purchase price is basically fixed for at least ten to fifteen years. The Irish Government has set a national target to achieve 40% of electricity consumption from renewable sources by 2020 [CER, 2011]. The programme involves the issuance of around 3,200 MW of capacity to on-shore wind projects, with a further almost 800 MW of capacity to off-shore wind projects. By any standards, this will be a significant level of mainly
intermittent wind power [CER, 2011]. It therefore stands to reason that current and future power plants will need to become optimised to ensure maximum plant efficiency and system flexibility.

However, renewable energy is not without its drawbacks. Consistency of supply is of major concern for power plant operators. At present the industry stance on wind energy integration into the national grid favours constraining wind power output to ensure that the load gradient that must be supplemented by fossil fuel power plants is within their output capabilities in order to minimise the risk of plant damage due to a fluctuating response in wind output [Børre Eriksen et al, 2005]. Figure 1.2 depicts the fluctuating nature of wind power over a 24 hour period. This shows highlighted regions where inadequate interfacing between the wind power generation plant and the transmission and distribution networks have resulted in loss of wind power output. This drop-off in power output must be supplemented by an alternative power source, and in this case fossil fuel power is the primary supplemental source.
Theoretical power production calculations for an offshore wind turbine predicts that a 10% increase in wind speed results in a 30% increase in electrical output [Børre Eriksen et al., 2005]. Therefore even small changes in wind speed will require the fossil fuel power plant to respond in a manner to supplement the shortfall in wind energy. It can be expected that there will be an increased frequency in the number of plant start-ups which will be undoubtedly be in an irregular and unpredictable fashion. The ramifications these increased number of start-ups must be assessed to ensure the life cycle of the plant is not diminished.

![Figure 1.2. Wind power trips induced by faults in the Spanish network (MW) [Eriksen et al., 2005].](image)

Increased electrical power production efficiency can be achieved by increasing plant operating temperatures and pressures. Another important recent development in plant operating mode is load-following, whereby plant electrical output is adjusted
response to customer demand at a given time. In broad terms, the evaluation of power plant can be presented according to operational temperatures and pressures, moving from subcritical towards supercritical, as shown in Figure 1.3. This figure depicts improved plant efficiency with increasing plant pressure and temperature and the labels (e.g. 24/540/565) on each step define steam pressure, boiler temperature and superheated steam temperature.

![Figure 1.3. Evolution of plant steam temperatures and pressure. Adapted from Alstom Power Co, Technical report, [2004].](image)

However an increase in power plant operating temperatures and pressures will undoubtedly lead to an acceleration in degradation of material properties particularly given the overall high temperatures and long plant operating lives required (see Figure 1.3). To this end, more advanced materials are under development for design of the next generation of power plants.
Failures of pressurised steam power plant components can occur dramatically and violently. This is depicted in the photograph of Figure 1.4(a), which displays a 'bird-beak' type failure along a seam weld in a 2\textsuperscript{1/4}\text{Cr}-1\text{Mo} material header. Figure 1.4(b) shows a less severe failure, also due to creep in a header of the same material. Failures such as these are costly in terms of plant replacement or repair, but also in terms of plant down-time. Furthermore, danger to human life is a key factor in ensuring the safe and reliable operation of these components.

Figure 1.4. Photographs of two different catastrophic header failures due to creep where (a) depicts failure along a seam joint, and (b) shows a localised failure at a stub-to-header junction [Paterson and Wilson, 2012].
Another driving force for improved plant performance is privatisation and deregulation of the electricity markets which has become more prevalent in recent years. Deregulation allows privately owned power generation plants to compete with state held power producers within an open market, allowing the consumer to choose the provider. This form of market competition will undoubtedly increase the requirement to build new plant which will operate as cost effectively and as reliably as possible. One of the newer breed of power plant is that of the combined cycle power plants. These types of plant are favoured over conventional plant as they are more environmentally friendly, almost twice as efficient, require comparably low capital investment and require less maintenance [Kiameh, 2012].

An understanding the performance of candidate power plant materials for plant operation at higher temperatures and pressures is of the upmost importance. 9-12% Cr steels are an example of such candidate materials, owing to their excellent mechanical properties, and they are already commonly employed in power generation plant [Viswanthan et al., 2005]. Advancements in non-destructive and destructive testing techniques have allowed for improved knowledge of these materials. Traditionally these materials have been assessed in terms of creep behaviour, which is the typical deformation mechanism of a material subjected to a constant load at elevated temperature. However, the aforementioned variable and flexible loading, associated with more frequent plant hot and cold starts requires investigation of such candidate materials in terms of thermal fatigue, thermomechanical fatigue, creep-fatigue and creep behaviour within plant components. This latter aspect has increased the importance of design and assessment of materials and components for high temperature, pressurised plant with respect to thermomechanical fatigue (TMF) and creep-fatigue failure. Increasing
plant temperatures will increase creep strain accumulation and changing from a constant ‘base load’ operation cycle to a fluctuating, load-following cycle will result in an increased risk of thermal fatigue. Hence, detailed investigation of realistic plant operation is also needed.

1.2 Aims and Objectives

The key aims of this work are to:

- Develop a three-dimensional thermomechanical model for realistic power plant components, including transient heat transfer and anisothermal cyclic viscoplasticity material behaviour.
- Develop a high temperature low cycle fatigue test methodology for experimental characterisation of constitutive and failure behaviour of service-aged P91 power plant material under representative HTLCF and TMF conditions.
- Conduct HTLCF testing of welded P91 material to identify characteristic cyclic and failure behaviour of welded specimens.
- Constitutive characterise the behaviour of P91 weld material and heat affected zone (HAZ) material under HTLCF conditions.
- Develop a multiaxial TMF life prediction model for realistic thermomechanical cycling of real plant header geometries.

The key objectives are identified as follows:

- Calibrate and validate a finite element heat transfer model capable of capturing the complex thermal history of a power plant start-up.
- Develop and calibrate a constitutive material model capable of simulating the complex thermomechanical material behaviour of service-aged P91.
- Develop methodology to identify the material modelling constants for P91 HAZ material for a range of temperatures.
- To apply the calibrated constitutive model to a realistic header geometry subjected to realistic plant loading conditions.
- Identify key locations where premature failure is likely to occur and to make predictions as to the number of plant start-up cycles to failure.
- Commission and validate a HTLCF test rig at NUIG.
- Carry out a material test program for a range of temperatures, strain ranges, strain rates and hold period for service-aged base P91 and welded materials.
- Characterise the service-aged P91 constitutively and carry out the previously mentioned multi-stub header simulation using the identified service-aged P91 material parameters.

1.3 Scope of thesis

Chapter 2 presents a review of the literature in terms of phenomenological and physical behaviour of power plant materials at macroscopic and microscopic levels, with particular attention given to P91 material. Constitutive and numerical methods are reviewed in order to assess appropriate methods used in analysing power plant material. Computational and experimental analysis, as well as industrial reports of power plant components and component failures are reviewed in order to establish the current state of power plant component testing and assessment.

Chapter 3 describes the material testing carried out on service-aged P91 base and weldment materials. Isothermal strain controlled cyclic testing along with stress relaxation tests for a range of temperatures, strains and strain rates, are carried out to characterise the materials. Cyclic stress-controlled tests were carried out on the
service-aged P91 base material under non-zero mean stress conditions. Description of experimental set-up and procedures used at two different test facilities are given, along with commissioning and validation of the high temperature low cycle fatigue rig. Comparisons are made between the service-aged material test results and other P91 material test data from the literature. Isothermal fatigue tests at 400 °C, 500 °C and 600 °C are performed in collaboration with the University of Nottingham on the service-aged P91 material. Similar tests are carried out at NUI Galway on cross-weld and single (i.e. base and weld) material specimens.

Chapter 4 utilises existing P91 material data from the literature to calibrate and validate a viscoplastic constitutive material model. The calibrated model is then applied within a finite element (FE) model, to examine a thick-walled plain pipe subjected to simplified, yet representative, plant loading cycle. Furthermore measured plant data taken during a power plant start-up and load-following operation (which is inclusive of a plant trip and subsequent re-start) is applied to the same geometry to determine the cyclic response of a plain pipe operating under realistic loading conditions. The simulation is a sequential thermomechanical modelling technique, combining transient conduction convection analyses with elastic-plastic-creep mechanical analysis. The predicted stress-strain histories form the basis of a TMF life prediction methodology using the Coffin-Manson and Ostergren models. A representative simplified thermomechanical plant cycle is proposed which is shown to capture the salient TMF damage phases of the realistic (more complex) measured cycle. The representative cycle is potentially useful for laboratory (controlled condition) testing and simplified analyses, but also allows isolation of the key TMF damage aspects of the complex cycle.
Chapter 5 applies the TMF constitutive model and failure methodologies to the significantly more complex geometry of a multi-stub branched header section, and particular attention focused on plant start-up. A global-sub modelling technique is developed to facilitate comparatively coarse mesh global analyses with subsequent localised mesh refinement (using the sub-model) for further interrogation of the critical locations. A multi-axial critical-plane methodology is developed in conjunction with the Ostergren energy-based life prediction model for prediction of thermomechanical fatigue crack initiation locations, orientations and numbers of plant start-ups to failure.

In Chapter 6 a multiaxial, critical-plane rainflow cycle counting methodology is developed for application to the measured (stochastic) temperature histories from the plant header measurements. This, in particular allows calculations of the TMF damage due to the many small temperature fluctuations measured during attemperation events in the real (ESB) header at Lough Ree, which is the subject (case study) component of this thesis.

Chapter 7 presents the identification (calibration) and validation process for the temperature-dependent material parameters of the cyclic elastic-viscoplastic constitutive model for the service-aged P91 material, based on the tests presented in Chapter 3. The model incorporates combined isotropic and non-linear kinematic hardening (NLKH), power law creep with the so-called two-layer viscoplasticity material model. The newly identified material constants for the service-aged P91 material are implemented within the sequential transient thermomechanical model of the multi-stub branched header model. Comparisons are made with the material behaviour and finite element modelling results of Chapters 4 and 5 which utilised P91 material constants identified from P91 test data published in the literature.
Chapter 8 describes the methodology developed to characterise the constitutive behaviour of the weld and heat affected zone materials utilising the base material, weld material, cross-weld material test data from Chapter 3.

Chapter 9 reports the conclusions and major findings of the thesis and recommendations for future work.
Chapter 2

LITERATURE REVIEW

2.1 Introduction

Failure of power plant components is extremely costly in terms of both plant down time and component repair. Predictive techniques, both experimental and computational, facilitate understanding of material and component behaviour, which in turn allows for scheduling of planned maintenance and inspection activities, which are essential factors to the operation of efficient, reliable and cost effective electrical power plant. Power plant components need to withstand high temperature loading for long periods as well as undergoing the cyclic loading conditions associated with plant start-ups, shut-downs and thermal transients due to the temperature control process of the plant. Such power plant components, particularly power plant headers, have been extensively investigated in terms of steady-state high temperature operation; however relatively little is known about the behaviour of these headers operating under cyclic loading conditions (characterised by relatively frequent start-up cycles).

This literature review chapter is mainly concerned with highlighting the salient factors associated with failure of power plant components under realistic plant loading. This is addressed in a number of ways. Firstly, the relevant material behaviour is discussed both at macroscopic level and at microscopic level; the key mechanisms are explored, combined with experimental and numerical methodologies employed to elucidate characteristic material behaviour, with the majority of the attention given to material performance at elevated temperature. Secondly, the selection of an appropriate constitutive material model is described, with the view to
calibrating an anisothermal material model capable of capturing the identified phenomenological material behaviour. The calibrated material model will then be applied to a power plant component of choice, and will form the basis of a life prediction methodology in an attempt to predict critical locations within the plant component.

2.2 Microstructure of P91

The 9% chromium steels developed over recent years have attracted great interest from designers of fossil fuel plant pipework. Of particular interest is P91 ferritic-martensitic steel. P91 is a modified 9Cr–1Mo steel, where the addition of vanadium, niobium and nitrogen has markedly improved the rupture stress of the material [Ennis et al., 2002]. A typical P91 composition is given in Table 2.1, reported by Orlová et al., [1998].

Table 2.1. Chemical composition of P91 steel (in mass %), [Orlová et al., 1998].

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<tr>
<th></th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>Nb</th>
<th>N</th>
<th>Ni</th>
<th>Al</th>
<th>Cu</th>
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<td></td>
<td>0.09</td>
<td>0.2</td>
<td>0.56</td>
<td>0.021</td>
<td>0.009</td>
<td>8.36</td>
<td>0.86</td>
<td>0.2</td>
<td>0.06</td>
<td>0.065</td>
<td>0.47</td>
<td>0.007</td>
<td>0.05</td>
</tr>
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The martensitic grain structure of P91 produces high dislocation density causing a retardation of creep deformation. This is due to the presence of NbC and VC precipitates which improves the creep strength of P91 compared to the unmodified 9Cr–1Mo steel [Sanderson, 1981]. However, the dislocation density decreases during sustained exposure to elevated service temperatures [Ennis et al., 2002; Dubey et al., 2005].
Heat treatment of the modified 9Cr-1Mo steel is critical in order to produce the desired microstructural and mechanical properties. For P91, a normalization heat treatment at a minimum temperature of 1040 °C, followed by cooling in air, is sufficient to produce a fully martensitic structure [Swindeman et al., 2004]. However, the normalization temperature can have an effect on the tensile and creep-rupture properties of the steel [Totemeier, et al., 2006]. Figure 2.1 shows the effect of normalization temperature on the room-temperature strength and micro-hardness of modified 9Cr-1Mo steel as a function of time (where HV_{500} corresponds to an Vickers indentation load of 500g). It is evident that a normalization temperature below 925ºC has a significant effect on the strength and micro-hardness whereas between 925 and 1050ºC, the properties are relatively constant. The difference in properties is attributed to difference in microstructure after normalization. While the microstructure consisted of low dislocation density polygonized ferrite at normalization temperatures below 900ºC, it was entirely martensitic above 900ºC.

Figure 2.1 Variation of room temperature yield strength, tensile strength and micro-hardness with normalization temperature of a modified 9Cr-1Mo steel [Totemeier et al., 2006].
2.3 Creep

2.3.1 General
Creep is a permanent deformation that occurs primarily in metals stressed at high temperature over an extended period of time. Creep deformation usually becomes an issue when the metal is subjected to temperatures above $0.4T_m$, where $T_m$ is the absolute melting temperature in Kelvin of the material. Creep failure is determined to have occurred when a component exhibits excessive deformation that interferes with its function, or when the component fractures as a result of the creep process [Ashby and Jones, 2005].

Creep deformation can be decomposed into three different stages, for a material subjected to constant stress. After initial immediate extension which causes elastic strain, $\varepsilon_o$, the stages of creep deformation can be separated as follows:

(i) a decelerating creep region known as the primary creep region,

(ii) followed by a steady state creep region or secondary creep, and

(iii) accelerating creep region which ends in rupture or failure of the material.

These regions are shown schematically in Figure 2.2 as regions I, II and III, respectively, and are typical of tests conducted under constant loads. The primary creep stage, or the transient creep stage is characterised by a strain rate which is initially high but decreases as time and total strain increases; this behaviour is often described by the following:

$$\varepsilon = \beta t^{1/3}$$  \hspace{1cm} (2.1)

where $\beta$ is a constant and $t$ is time. As the strain rate decreases to a constant value, creep deformation enters the secondary or steady state creep stage. This region is described by the relation
\( \varepsilon = \kappa t, \) \hspace{1cm} (2.2)

where \( \kappa \) is the material constant called the minimum creep rate [Andrade, 1952]. The secondary stage persists for much longer than the other stages and therefore is more important in terms of plant design. The rate of steady-state creep is an increasing function of stress and is described by the following power-law relation;

\[ \dot{\varepsilon} = A\sigma^n \] \hspace{1cm} (2.3)

As time and total strain further increases, creep deformation enters a final phase where creep rate increases until rupture occurs. This regime of material damage is commonly referred to as tertiary creep.

Figure 2.2. Schematised uniaxial creep behaviour showing (i) primary creep (I), (ii) secondary or steady-state creep (II), and (iii) tertiary creep (III).
Various types of creep tests can be implemented to obtain useful creep behaviour data, which can then be used in component design. Figure 2.3a shows creep-rupture curves from creep-rupture tests [Hyde et al., 2006], performed on specimens under constant uniaxial loading. Creep-rupture tests requires strain measurement to be carried out for the duration of the test until failure occurs. It can be seen from Figure 2.3a that greater load amplitudes leads to a decreased time to rupture; it should be noted that increased temperatures will have a similar effect on rupture lives. Such tests provide a large range of information including time-to-rupture, strain at rupture, amount of primary and secondary creep, time to onset of secondary and tertiary creep, steady-state creep rate [Penny and Marriott, 1995]. Figure 2.3b depicts data from a stress rupture test [Hyde et al., 2006]. This test is a simpler type of test, where a specimen is subjected to a constant load and time to rupture recorded. Creep information available from this type of test is limited to time-to-rupture and strain-at-rupture.

Figure 2.3. (a) Tested and fitted uniaxial creep strain curves for P91 (Bar 257 grade material) from Hyde et al; and (b) Stress-rupture test data for the same material (Bar 257) and that of a higher grade of material A-369 [Hyde et al., 2006].
It may be necessary to record the microstructural changes of the material as the test progresses; in such a case the test would be stopped at specified intervals and the microstructural changes assessed. This method of testing is referred to as an interrupted creep test and was employed by Tabuchi et al. [2009] to determine the effect of creep deformation on the heat affected zone of a P91 weldment. It was observed that creep voids were initiating early in the life of the test and as the testing progressed these voids were coalescing to form cracks. Creep tests are usually performed for up to 1000 hrs.

Under a constant displacement loading scenario, creep deformation will cause a relaxation of stresses over time. These tests are particularly useful due to the brevity of test time required. If the stress relaxation time is known, then the creep constants can be determined using a curve fitting process. Conversely, the relaxation time can be determined using Norton’s power law creep equation.

At any time during the stress relaxation, if \( \sigma < \sigma_y \),

\[
\varepsilon^{\text{total}} = \varepsilon^{\text{elastic}} + \varepsilon^{\text{creep}}
\]  

(2.4)

Elastic behaviour is given by;

\[
\varepsilon^{\text{elastic}} = \frac{\sigma}{E}
\]  

(2.5)

Using the power law relation of Equation 2.3 for a constant strain rate at a constant temperature, and differentiating with respect to time and substituting in Equations 2.4 and 2.5 gives:

\[
\frac{1}{E} \frac{d\sigma}{dt} = -A\sigma^n
\]  

(2.6)
Then integrating both sides from $\sigma = \sigma_i$ at $t = 0$ to $\sigma = \sigma$ at $t = t$.

$$\frac{1}{\sigma^{n-1}} - \frac{1}{\sigma_i^{n-1}} = (n-1)AEt \quad (2.7)$$

For further details see Ashby and Jones [2005]. Figure 2.4 shows typical stress relaxation data for P91 material at two different temperatures. The associated fit using the Norton Power Law equation is also shown.

![Stress Relaxation Curves](image)

Figure 2.4 Stress relaxation curves for P91 for two different temperatures [Takahashi, 2008].

### 2.3.2 Creep damage
Creep damage models have been found to be effective in capturing the primary, secondary and tertiary stages of creep [Penny and Marriot, 1995]. One of the major advantages of these damage models is their ability to capture the tertiary stages of the creep curve, which can occupy a large fraction of the total life. Creep damage is based on a time-dependent accumulation of creep damage within the material under
creep loading conditions which will ultimately lead to failure. On a macroscopic level, damage due to creep is due to the growth of internal voids and cavities, along with change in the grain structure such as change in grain size, decrease in dislocation density and coarsening of precipitates. The Kachanov [1960] model, which was later modified by Rabotnov [1969], has been found to be an accurate model for predicting creep damage accumulation. These models are based on a power law relationship, incorporating a damage parameter \( \omega \). As damage accumulates over time, the strain rate at a point in the material also increases with time. Perrin and Hayhurst [1996] adopted a damage mechanics approach to propose creep constitutive equations for a 0.5Cr-0.5Mo steel, which were formulated as:

\[
\varepsilon_c = A \left( \frac{\sigma}{1 - \omega} \right)^n t^m 
\]

(2.8)

At time zero, \( \omega \) is equal to zero and, as time increases, the damage increases. At \( \omega=1 \) the material is said to have failed. The evolution \( \omega \) is described by the following equation:

\[
\dot{\omega} = \frac{M \sigma^\zeta}{(1+\varphi)(1-\omega)^\varphi} t^m 
\]

(2.9)

where \( \dot{\omega} \) is the damage rate and \( M, \varphi \) and \( \zeta \) are material constants, which can be determined from creep rupture data [Hyde et al., 2006].

Hyde et al., [1997] extended these equations to multiaxial forms, which is given as:

\[
\varepsilon_c = \frac{3}{2} A \left[ \frac{\sigma_{eq}}{1 - \omega} \right]^n \frac{S}{\sigma_{eq}} t^m 
\]

(2.10)
where $A'$, $n'$ and $m$ are material creep properties determined from experimental creep data, $S$ is the deviatoric stress and $\sigma_{eq}$ is the equivalent stress.

The creep rupture stress $\sigma_r$ is required for multi-axial creep behaviour and is calculated using the equivalent stress, $\sigma_{eq}$, and the maximum principal stress, $\sigma_1$, using the following equation:

$$\sigma_r = \eta\sigma_1 + (1-\eta)\sigma_{eq} \quad (2.11)$$

where $\eta$ is a material constant, which ranges from 0, for cases where the equivalent stress is dominant, to 1 for cases where the maximum principal stress dominates [Hyde et al., 2006]. Comparisons with test data have been made by Hyde and co-workers [2006] relating uniaxial creep curves to that of the damage model. It was found that there was good agreement between the test data and that of the model, as shown in Figure 2.5. The model was also validated multiaxially against stress rupture data from notched creep tests (see Hyde et al., 2006).

![Figure 2.5](image-url) Figure 2.5. Tested and fitted uniaxial creep strain curves of A-369 steel at 625 °C [Hyde et al., 2006].
2.3.3 Creep mechanisms
The physical mechanisms by which materials deform due to creep are dependent on the various combinations of material composition, grain structure, heat treatment, applied stress and temperature. Typically creep mechanisms are categorized into two general classes called diffusional flow creep and dislocation creep.

Diffusional creep is a thermally activated mechanism where atoms will move within a given material lattice or within the grain. Diffusional flow creep can occur at low stresses but requires high temperatures. Grain boundaries generally contain a high concentration of vacancies. Under an applied tensile stress, grain boundaries perpendicular to the stress direction will be in tension, thus assisting formation of more vacancies at the boundaries, whereas boundaries parallel to the applied loading direction will experience a compressive stress, which will prevent the formation of vacancies. This process is shown schematically in Figure 2.6. As a consequence, there will be a net migration of atoms near the boundaries in compression to the grain boundaries in tension. Continuous motion of these atoms and vacancies would eventually lead to permanent deformation of the grains in the direction of the applied stress [Ashby and Jones, 2005].

![Image of creep mechanisms](image)

Figure 2.6. Creep taking place by diffusion [Ashby and Jones, 2005] (a) atoms migrate to boundaries in tension from regions of compression (b) shows the change in grain dimensions after a period of time.
Generally speaking there are two types of diffusional flow creep, Nabarro-Herring creep and Coble creep. Creep strain-rates of both Nabarro-Herring and Coble creep are proportional to stress; however, Nabarro-Herring creep is inversely proportional to the square of the grain size while Coble creep is inversely proportional to the cube of the grain size [Nabarro, 2002]. Nabarro-Herring creep deformation occurs due to the motion of atoms and vacancies through the crystal lattice, whereas Coble creep deformation occurs due to motion of atoms and vacancies along the grain boundaries.

The rate of dislocation creep is limited by diffusional creep. Dislocation creep, requires the movement of dislocations and occurs at high stresses. The two classes of dislocation creep are: (i) dislocation glide creep, where dislocation move along slip planes and (ii) dislocation climb creep, where dislocations will ‘climb-over’ blockages, such as precipitates.

During high temperature creep, the deformation of individual grains may also lead to relative displacements between adjacent grains and this phenomenon is referred to as grain boundary sliding.

In power plant steam pipework applications, the P91 pipe material is subjected to temperatures approaching 0.5\(T_m\) of the material, while having to withstand service-life of up to and beyond 100,000 hours at relatively low applied stress levels.

P91 martensitic steels have been specifically designed to resist creep. This has been brought about through precipitate strengthening and the inherent creep resisting properties of a martensitic microstructure. However, there is an evolution in the microstructure of the material under sustained creep loading. The effect of the P91 microstructure from a creep stand point has received some attention in recent years and is discussed as follows; Figure 2.8 is the schematic representation of dispersal
precipitates states in martensitic 9-12%Cr steels [Gocmen et al., 1998]. The microstructure consists of martensitic blocks within former austenite grains, with each block being split into elongated subgrains (laths). Figure 2.7 is representative of P91 material post-heat treatment where the effect of tempering which causes the precipitates to mostly reside along the substructure interfaces and the coarse carbonitrides are non-uniformly distributed within the different subgrains.

Two types of precipitates are present in high chromium steels before creep: $M_23C_6$ carbides and MX carbonitrides [Maruyama et al., 2001], as shown in Figure 2.8. The $M_23C_6$ carbide particles formed during tempering are mainly located on the grain and subgrain boundaries, with $Cr_23C_6$ precipitates being the primary type of $M_23C_6$ precipitates [Maruyama et al., 2001; Orlová et al., 1998]. $M_23C_6$ precipitates inhibit the subgrain growth thus increasing the strength of the materials [Ennis et al., 1998]. However, $M_23C_6$ precipitates can only strengthen the steels at the early stage of creep due to their coarsening at the grain boundaries and subgrain boundaries after long time exposure at high temperatures, which gives rise to a reduction in creep strength [Yin and Faulkner, 2005]. Anderson et al. [2003] have found that the average size for the $M_23C_6$ carbides on the prior austenite grain boundaries increases after creep loading.

Another precipitate type present in 9-12%Cr steels is MX, where M denotes Niobium (Nb) or Vanadium (V); and X denotes Carbon (C) and/or Nitrogen (N), namely, carbides, nitrides and carbonitrides. Unlike $M_23C_6$ precipitates, MX precipitates form primarily in the subgrain interiors. MX precipitates are normally smaller than the $M_23C_6$ carbides [Orlová et al., 1998]. These MX particles are fine and distributed uniformly within the subgrains as well as on subgrain boundaries [Maruyama et al., 2001]. MX particles improve the creep strength of 9-12%Cr steels
[Maruyama et al., 2001; Yin and Faulkner, 2005] by two mechanisms: (i) MX particles themselves act as obstacles to dislocation motion, and (ii) they inhibit the recovery of the dislocation substructure [Maruyama et al., 2001].

Figure 2.7. Schematic representation of non-uniform precipitation states in tempered martensitic 9-12%Cr steels [Gocmen et al., 1998]

Figure 2.8. A schematised representation of precipitates in high chromium ferritic steel [Maruyama et al., 2001].

When sufficient time has passed, creep deformation will ultimately result in fracture. High temperature fracture can occur in three different modes – rupture, transgranular creep fracture, and intergranular creep fracture. Rupture is
accompanied by a high reduction in area and happens at high stress and high temperature. Transgranular creep on the other hand is analogous to low temperature ductile fracture where internal cracks and voids form at highly stressed locations of the microstructure and coalesce. The fracture is ductile and the appearance of the fracture surface is very similar to ductile fracture at lower temperature. The third type of high temperature fracture is intergranular creep which occurs at stress levels much less than those of rupture and transgranular creep fracture. In this fracture mode, cracks and voids form along grains boundaries and coalesce to fracture in a brittle manner. Cracks and voids are usually associated with inclusions or precipitates. It is believed that the presence of these particles along the grain boundaries prevents the rate of accommodation grain boundary sliding from matching the deformation rate of diffusional creep thereby allowing cracks and voids to form at the grain boundaries.

The effect of microstructural evolution of P91 material under service conditions was studied by Swindeman et al., [2003]. It was observed that the service-exposed material exhibited a coarsened substructure. The effect of service exposure on the P91 clearly illustrated in Figure 2.9, where creep tests revealed that the P91 materials which were subjected to prior service exposure exhibited higher creep strain rates than the unexposed material [Swindeman et al., 2003]. In addition to a loss in creep resistance, Swindeman et al., [2003] also reported that the service-exposed material exhibited a lowering in yield and ultimate tensile strength, relative to that of the unexposed P91 material.
Figure 2.9. Creep tests of service-exposed and unexposed P91 material [Swindeman et al., 2003].

2.3.4 Creep of welded components
The mechanical properties of the weld metal and heat-affected zone material of a P91 weldment are different than those of the base metal [Shankar et al., 2010]. The variations in hardness of a generic weldment are shown in Figure 2.10.

Figure 2.10 Schematised representation of variation in hardness of the base metal, weld metal and heat affected zones (HAZ).
Note that the minimum hardness of the weldment is found in the heat-affected zone; this type of material behaviour has been observed by Watanabe et al. [2006], Thomas Paul et al. [2007], and Das et al. [2008]. Investigation of P91 weldments found that creep rupture properties of the weld metal and HAZ are lower than those of the base metal at all temperatures, as shown in Figure 2.11. All the weldments failed in the weld metal at 550°C. At 600°C, fracture occurred in the weld metal at the higher stress and shorter time, however it should be noted that applied stresses above 100 MPa will lead to failures in the weld material, therefore it is necessary to conduct these tests at lower applied loads [Allen, 2012]. For rupture times greater than 4,000 h, the failure location shifted to the HAZ. At 650°C, though fracture occurred in the weld metal for rupture time less than 1,000 h, the failure location was in the HAZ for rupture times greater than 7,000 h [Watanabe et al., 2006].

Kimura et al., [2011] have performed a range of creep rupture tests for P91 weldments and base material. Creep tests were also performed at 550°C, 600°C, 650 °C, and 700°C. For temperatures above 550 °C the P91 weldment failed earlier than
the base material. In the case of 550 °C little difference in rupture time was found between the base material and the welded material for a rupture stress greater than 200 MPa.

2.3.5 Service Experience of Welded Sections
Modified P91 steel weldments require in depth investigation because they exhibit very low ductility in the as-welded condition [Shankar et al., 2010]. Preheating at elevated temperatures, and post weld heat treatment are necessary for all weldments, regardless of the geometry involved. However, there is a lack of understanding of the significant differences in welding of P91 steel versus the low-alloy steels, traditionally used in the power industry [Shibli, 2002]. P91 steel is less tolerant of temperature variations during welding and in post-weld heat treatment. This lack of understanding has resulted in several failures in the field. Problems that significantly impair the creep-rupture strength of P91 steels are over-tempering, under-tempering, and exposure to temperatures in the intercritical region. Furthermore the reduced creep strength of the Type IV region within the HAZ compared to the base metal has not been fully elucidated, either in terms of creep or creep-fatigue. These concerns along with factors such as weld-repair or service-induced residual stresses for components and sensitivity to transient or temperature events, require further investigation.

2.4 Fatigue
Fatigue is caused by periodic application of stresses due to mechanical or thermal loading. The metal subjected to fluctuating stress will fail at stresses much lower than those required to cause fracture in a single application of load [Fuchs and Stephens, 2001]. Fatigue can be categorized into two classes, (i) high cycle fatigue
and (ii) low cycle fatigue. Although, there is no distinctive limit between these two types of fatigue, the traditional approach is to classify failures occurring above 10,000 cycles as high-cycle fatigue and those occurring below that value as low-cycle fatigue. An important distinction between low-cycle fatigue and high-cycle fatigue is that in high-cycle fatigue most of the fatigue life is spent in crack initiation, whereas in low-cycle fatigue most of the life is spent in crack propagation, as cracks are found to initiate within 3 to 10% of the fatigue life [Fuchs and Stephens, 2001].

### 2.4.1 Low cycle fatigue

Low cycle fatigue requires a material to undergo loading where each cycle will produce elastic strain and also some plastic strain [Coffin, 1977]. The low cycle fatigue process can be roughly divided into four stages: cyclic hardening/softening, crack propagation, and fracture [Hertzberg, 1996].

Fatigue failures occur in many different forms. Fluctuations of applied stresses or strains result in mechanical fatigue. Cyclic loads acting in association with high temperatures cause creep-fatigue. When the temperature of the cyclically loaded component also fluctuates, culminating in a synergetic damaging effect, this form of fatigue is known as thermomechanical fatigue. In power plant environments fatigue due to temperature fluctuations alone, i.e. thermal fatigue, can also in certain cases be identified. The majority of failures in power plant components can be attributed to one of the above fatigue processes. It should be noted that the magnitudes of the cyclic stresses and strains generated during power plant operation may be considerably lower than the maximum design stresses.

One characteristic of fatigue in metals is work hardening under reversed loading conditions. With continued cyclic loading, the rate of hardening progressively diminishes and a steady state of deformation, known as saturation, is
reached. Once saturation occurs, the variation of the resolved shear stress with the resolved shear strain is not altered by further load cycles and the stress-strain hysteresis loop develops a stable configuration [Fuchs and Stephens, 2001]. One of the key observations in terms of P91 material behaviour, is that the material cyclically softens [EBI and McEvily, 1984; Nagesha et al., 2002; Mannan and Valsan., 2005], not alone this, but it continuously softens, which creates a challenge in terms of material characterisation (i.e. isotropic hardening /softening constants) and that of fatigue life prediction. Unsymmetric cycles of stress between prescribed limits may cause an altogether different cyclic response. Kunz and Lukáš [2001] presented test data on 9%Cr-1%Mo steel which initially cyclically hardened under positive mean stress loading conditions before reverting to continuous cyclic softening. Experience from other material would indicate ratchetting in the direction of the mean stress [Kapoor, 1994].

The repeated loading leads to damage accumulation. After a saturation of energy accumulation has occurred, at which state no further slip processes can take place in the crystallite, small micro cracks form [Miller et al]. With an increasing number of load cycles these micro-cracks start to grow and additional regions undergo the process of micro-scale damage. As the crack grows to macroscopic size the stress intensity at the crack tip usually increases so that the crack growth is accelerated until the remaining material can no longer bear the stress and complete failure occurs. Three fundamental aspects related to fatigue crack initiation are:

- Significance of the free material surface,
- Irreversibility of cyclic slip,
- Environmental effects on micro-crack initiation.
Micro-cracks usually initiate at the free surface of material. The restraint on cyclic slip is lower than inside the material because of the free surface.

Mannan and Valsan investigated the high temperature LCF behaviour of 316L stainless steel and P91 and their welds. They observed that dynamic strain aging (DSA) increased the stress response and reduced the ductility and oxidation reduced fatigue life in the case of P91.

2.4.2 Creep-fatigue interaction
Prolonged ageing of a P91 alloy at elevated temperatures prior to testing has been found to reduce the LCF and creep–fatigue interaction lives [Okamura et al., 1999].

Creep-fatigue interaction or high temperature fatigue occurs at elevated temperatures, where viscous effects and stress relaxation affect the fatigue life. Creep-fatigue interaction of high-temperature alloys can reduce life in a nonlinear manner. Figure 2.12 shows that interaction for P22 steel is quite severe as compared to Type 304 and 316 stainless steels, and interaction for P91 is more severe than that for P22 [ASME 2005].

Figure 2.12 Creep-fatigue interaction for typical power plant piping materials [ASME, 2005]
As discussed earlier, some older plants now use, or plan to use, P91 as a replacement material and newer power plants use 9Cr martensitic steels for thick section components [Shibli, 2002]. This is considered useful from the standpoint of fatigue damage because thinner section components made of high strength P91 steel will be less prone to fatigue cracking during the cyclic operation, due to smaller temperature gradients through the wall thickness.

Takahashi [2006] has presented some results of creep-fatigue tests on P91 base metal and weldment, at 550 and 600°C, as shown in Figure 2.13. Figure 2.13 illustrates the effect of hold time on fatigue life, for the two different temperatures. It is clear that the hold period has little effect on the fatigue life of the base material at 550 °C, however at the higher temperature of 600 °C the decrease in fatigue life as a function of hold period is much more pronounced. Similar behaviour is observed for the weldment tests.

Figure 2.13 Creep-fatigue lives of P91 base metal and weldment specimens for various hold periods [Takahashi, 2006].
In the weldment specimens, failure always occurred in the fine-grained HAZ regardless of test temperature and test duration. The shortest test duration at 550°C was 600 h. No Type IV failure with equally short duration has been previously reported in pure creep tests within the same temperature range. This suggests that Type IV failure tends to take place more readily in creep-fatigue conditions in comparison to pure creep conditions.

Mannan et al. [2001] reported that with the introduction of a tensile hold, the fatigue life of P91 steel decreased rapidly with increasing hold time. A tensile hold of 1 h reduced life to approximately 25% of the continuous cycling life (i.e. no hold time). The reduction in life under creep-fatigue condition is attributed primarily to the reduction in strength of P91 steel due to microstructural degradation associated with the coarsening of precipitates and dislocation substructures. Shankar et al., [2006] studied the effect of a hold period at the peak strain in tension and as well as compression. Both types of hold test were found to lead to a reduction in fatigue life of P91 material, compared with those of pure fatigue tests. Compression hold was observed to be more damaging than tensile hold. It was also observed, that at low strain rates and long hold periods at high temperatures, oxidation was found to markedly influence fatigue life when compared with LCF tests carried out in an inert environment. Fournier et al. [2008] presented results of cyclic, strain-controlled and stress-controlled creep-fatigue tests on P91 steels at 550°C. The strain controlled tests were referred to as relaxation fatigue (RF) tests and the stress-controlled as (CF) tests. The RF and CF tests applied a hold period in tension. The material was tested for strain ranges of 0.6 %, 0.7 %, and 1 %, for hold periods of 10 min, 30 min, and 90 min. It was observed that the CF tests had shorter fatigue lives than the RF tests.
This was attributed to a greater amount of inelastic strain accumulation during the stress-controlled hold period.

**2.4.3 Thermomechanical fatigue**
Thermomechanical fatigue (TMF) is similar to high temperature low cycle fatigue (HT-LCF) in terms of mechanical strain loading; however, with TMF the temperature is also varied cyclically. TMF testing is categorized into two classes (i) in-phase (IP) where the maximum temperature coincides with peak tensile strain, as shown in Figure 2.14a, and (ii) out-of-phase (OP) where maximum temperature and maximum compressive strain coincide, as shown in Figure 2.14b.

![Figure 2.14 Schematized representation of temperature-strain loading for (a) TMF in-phase loading conditions, and (b) TMF out-of-phase loading conditions.](image)

The stress-strain behaviour of metallic materials in TMF has a characteristic loop response. This is depicted in Figure 2.15a and Figure 2.15b. Figure 2.15a depicts an IP stress-strain response with a resulting compressive mean stress, which is reported to influence in oxidation fatigue. In the case of TMF-OP, a tensile mean stress typically occurs (Figure 2.15b), which is deemed to be important in terms of fatigue crack growth.
Thermomechanical fatigue (TMF) type loading can be more damaging by more than an order of magnitude compared with isothermal LCF tests at maximum temperature, as illustrated in Figure 2.16 [Jaske, 1976] for 1010 low carbon steel. Figure 2.16 shows the isothermal fatigue data exhibiting much longer lives compared with that of the TMF lives. It can therefore be surmised that life prediction based on isothermal high temperature LCF data alone will be non-conservative. Hence, to obtain truly representative, high temperature fatigue behaviour for a material undergoing complex thermomechanical cycles, TMF testing is an essential requirement.
Another influence on fatigue life is that of oxidation damage. During mechanical straining, an oxide layer will crack and expose new clean surfaces. This clean metal will rapidly oxidize and the process repeats. IP loading is more likely to cause oxidation damage because an oxide layer can form in compression at the higher temperature and then rupture during the subsequent low temperature tensile reversal, where the oxide film is more brittle at the lower relative temperature. In OP TMF loading an oxide layer can also form on the materials surface during cycling. However the work of Shankar et al., [2012] observed that TMF-IP loading produced reduced fatigue lives compared with TMF-OP tests, when compared to tests carried out in an inert environment.

2.4.4 Thermomechanical fatigue of welded components
Inspection of in-service P91 components has shown that a significant number of failures occurs at weld sites [Brett, 1994]. A considerable amount of work has been carried out with regard to determining the creep behaviour of these weldments;
however relatively little work has been carried out in terms of TMF characterisation. Isothermal LCF tests has been carried out by Sandhya et al., [2010], showed that the TMF behaviour of the weld metal is similar to that of P91 base metal, both in terms of stress-strain response and that of softening behaviour. However the specimens were manufactured from a specially prepared weld pad. An alternative approach is to extract specimens from an ex-service (retired) component.

2.4.5 Microstructural behaviour due to cyclic material loading
High chrome steels which have been subjected to high temperature loading are known to undergo a microstructural evolution. This has been well documented in the works of EBI and McEvily, [1984]; Fournier et al, [2000]; Nagesha et al., [2002]. Studies carried out by Kimura et al., [2006] and Fournier et al., [2008a] reported that microstructural evolution take place much faster that under fatigue and creep fatigue type loading, compared with that of creep tests at the same temperature. The manifestation of this microstructural evolution came in the form of subgrain coarsening, where the boundaries between the martensitic laths and subgrain boundaries disappear [Fourier et al., 2008], which leads to a coarser microstructure. A decrease in dislocation density was also observed. Fournier and co-workers [2008a] went on to conclude that subgrain coarsening phenomenon depends on the applied plastic strain.

A common phenomenon with Cr-Mo steels is that of cyclic softening. The microstructural evolution which gives rise to this softening is summarised as follows: (i) reduction of dislocation density due to martensitic transformation, (ii) degradation of strength due to coarsening of precipitates [Shankar et al., 2006].
2.5 Material behaviour modelling

2.5.1 General
In order to simulate stress-strain behaviour of a material (computationally) it is essential that an appropriate material model be selected. The present section focuses on describing the development, implementation and use of phenomenological constitutive material models. These phenomenological models depict the macroscopic response of a given material. Development of these models allows for adequate simulation of the cyclic elastic-viscoplastic behaviour observed experimentally. The use of these models in complex geometries allows for prediction of stress-strain behaviour at a given material point, thus facilitating fatigue and rupture life calculations at locations of interest.

2.5.2 Cyclic plasticity
Once a material has exceeded its yield strength it is said to have undergone plastic deformation, from this point on any further deformation of the material requires a set of material deformation rules that will characterise the stress-strain response. These rules are discussed below:

2.5.3 Isotropic hardening
This particular hardening rule assumes that the strain hardening corresponds to an enlargement of the yield surface without a change on position of the yield surface in the stress space. The model assumes that the elastic domain expands equally in tension and compression during plastic flow [Lemaitre and Chaboche, 1990]. Figure 2.17 depicts a multiaxial isotropic hardening model with a von Mises yield surface where the effect of strain hardening has caused an increase in the radius of the yield surface in principal stress space. Also shown is the corresponding uniaxial stress-strain response for a cyclic uniaxial test for an isotropic hardening model [de Souza Neto et al., 2008]. The scalar $R$ is the change in the size of the yield surface,
commonly referred to as the drag stress. The evolution of isotropic hardening can be expressed by the following equation [Lemaitre and Chaboche, 1990]:

\[ \dot{R} = b(Q - R)\dot{p} \]  

(2.12)

where \( \dot{p} \) is the accumulated plastic strain, \( Q \) is the asymptotic value of \( R \) and \( b \) defines the rate at which saturation is reached.

Figure 2.17 Schematic representation of isotropic hardening on \( \pi \)-plane, and associated stress-strain response.

Under cyclic loading conditions material hardening is known to be a function of inelastic strain range. Progressive cycling will also have an evolutionary effect on the plastic strain. The cyclic hardening of a material refers to the decrease of the plastic strain range, with an associated increase in stress amplitude with increasing cycles [Dunne and Petrinic, 2005]. In general, hard materials have been found to cyclically soften whereas soft materials are found to cyclically harden [Fuchs and Stephens, 2001]. For example, cyclic hardening has been reported by Deshpande et al., [2010a; 2010] for a nickel-based refractory alloy and cyclic softening has been reported by Saad et al., [2011a], Koo et al., [2007], and Fournier et al [2006a], for 9Cr steels. Under constant amplitude cyclic strain range, cyclic hardening will have an effect of
increasing the elastic strain range [Chaboche, 2008]. Within the isotropic hardening model this behaviour is represented by an increase in size of the yield surface, i.e. an increase in the elastic limit, given by $\sigma_y + R$. The converse is true for a material which cyclically softens, where the material constant $Q$ can be assigned a negative value to account for the progressively decreasing size of the yield surface with each reversal in applied strain.

The isotropic softening model discussed so far has utilised material behaviour which will, after a certain number of cycles, settle to a stabilised state. However, some materials subjected to cyclic loading will not stabilise, and in the case of martensitic steels the material has been found to continuously soften until failure occurs. To account for this phenomenon Bernhart et al. [1999] developed a two-stage evolution equation, to accurately represent a continuously softening material response. The relation is given by;

$$R = Q_1p + Q_2(z)[1 - e^{(-lp)}]$$  \hspace{1cm} (2.13)

where $p$ is the cumulative plastic strain, $z$ is half the maximum cyclic plastic strain range reached at the current number of cycles. This two-stage approach can accurately capture the initial rapid decrease in stress and the subsequent secondary linear softening [Bernhart et al., 1999].

2.5.4Kinematic hardening
In kinematic hardening the size and shape of the yield surface remains the same but translates in the stress space as a rigid body [de Souza Neto et al., 2008]. Observations from material testing have shown that when a material is strained in one direction, its resistance to yielding in the opposite direction reduces [Lemaitre and Chaboche, 1990]. This behaviour is termed the Baushinger effect. Figure 2.18 is
a schematised representation of the kinematic hardening rule in stress space and the corresponding uniaxial model, having undergone initial tension followed by a compressive reversal, in which $\sigma_y$ represents elastic limit [de Souza Neto et al., 2008]. The kinematic hardening parameter, $\alpha$, better known as the backstress defines the instantaneous position of the loading surface [Lemaitre and Chaboche, 1994]. Unlike the isotropic hardening rule, the elastic region remains constant, both initially and during cyclic loading [Dunne and Petrinic, 2005].

$$f = f_y (\sigma - \alpha) - k$$  \hspace{1cm} (2.14)

Prager [1949] developed a kinematic hardening model to simulate material stress-strain response under cyclic loading conditions. A linear kinematic hardening rule is utilised which describes the translation of the yield surface, linearly relating it with the plastic strain.

Figure 2.18 Kinematic hardening behaviour and the Bauschinger effect, along with the associated uniaxial test behaviour, showing loading in one direction with decreased resistance to yielding in opposite direction.
where $a$ is the kinematic hardening variable known as the backstress given the position of the centre point of the yield surface and $k$ is the cyclic yield strength [Lemaitre and Chaboche, 1990].

The Prager model is given by the following equation:

$$ d\alpha_{ij} = cd\varepsilon_{ij}^p $$

where $\alpha_{ij}$ is the backstress tensor, $c$ is the hardening modulus taken from a uniaxial tensile stress-strain curve and $d\varepsilon_{ij}^p$ is the incremental effective plastic strain tensor.

For most material behaviour, the stress-strain response due to cyclic loading is a nonlinear type response. To address this, Frederick and Armstrong [1966] modified the Prager model with the addition of a recall term, also called the dynamic recovery term, which introduces non-linearity between the back stress $a$ and the actual plastic strain.

$$ \dot{\alpha} = \frac{2}{3} c \dot{\varepsilon}^p - \gamma \dot{\varepsilon}^p $$

where $\gamma$ is a material constant. The recall term employs a fading memory effect which generates the nonlinear response.

Another variation of Prager's rule is that of Ziegler's hardening rule, and is given as:

$$ d\alpha = (\sigma - \alpha) d\mu $$

where $d\mu$ is a constant determined from the hardening curve as $d\mu = \frac{c}{\sigma_0} d\varepsilon^p$.

The anisothermal linear form of the Ziegler kinematic hardening rule is given as follows:
\[
d\alpha = c \left( \frac{1}{\sigma_0} \right) (\sigma - \alpha) d\bar{\varepsilon}_p + \frac{1}{c} \dot{\alpha} \dot{\varepsilon} \tag{2.18}
\]

where \( c \) is the kinematic hardening modulus and \( \dot{c} \) is the rate of change of the hardening modulus with respect to temperature, \( d\bar{\varepsilon}_p \) is the equivalent plastic strain rate and \( \alpha \) is the backstress. In this model the initial size of the yield surface, \( \sigma_0 \), can be a function of temperature and remains constant.

**2.5.5 Combined isotropic-kinematic hardening model**

Observations from experimental material testing have revealed that the material stress-strain response is generally a combination of both isotropic and kinematic hardening i.e. the yield surfaces expands/contracts and translates simultaneously in the stress space [de Souza Neto et al., 2008]. In light of this, it is necessary to combine both of these hardening rules in order to accurately represent material behaviour subjected to cyclic loading. This behaviour is graphically represented in Figure 2.19 where the hysteresis loops show the stress-strain behaviour after one loading cycle (solid black line) and the stress-strain behaviour after some number of cycles (red dashed line), where due to isotropic hardening the size of the hysteresis loop has changed. The first cycle initially experiences a tensile strain where the stress linearly increases, until the yield stress (\( \sigma_y \)) is attained. The material then deforms kinematically due to the translation of the yield surface, until the strain reverses. Upon reversal of strain, the material response is initially elastic until the compressive stress reaches the yield surface, where plasticity begins again. With each reversal in strain the yield surface will also progressively expand or contract depending on the material, giving the isotropic hardening effects, which is depicted by the broken line in the stress-space [Dunne and Petrinic, 2005].
Figure 2.19 Combined isotropic and kinematic hardening, adapted from Dunne and Petrinic [2005].

An example of a non-linear kinematic hardening rule defining the translation of the yield surface in stress space is that of the anisothermal Ziegler evolution law for back stress which is given by the following equation:

$$d\alpha = c\left(\frac{1}{\sigma_0}(\sigma - \alpha)\dot{\epsilon}_p + \gamma \alpha \dot{\epsilon}_p + \frac{1}{c}\alpha \dot{\alpha}\right)$$  \hspace{1cm} (2.19)

where $\gamma$, is the relaxation term which decides the rate at which the hardening modulus $c$ decreases with increasing plastic deformation. The equation is comprised of three terms. The first is a linear term kinematic term based on the Prager model, the second is a recall term, which is based on the Fredrick-Armstrong recall term, which accounts for the fading memory effect of the deformation path, which introduces non-linearity to the Ziegler hardening law, and the final term is a temperature rate term. The isotropic component defining the size of the yield surface
\( \sigma_0 \), which is a function of equivalent plastic strain and temperature is given by the following equation [ABAQUS, 2010]:

\[
\sigma_0 = k + Q \left( 1 - e^{-b \varepsilon_p} \right)
\] (2.20)

where \( k \) is the yield surface size at zero plastic strain, and \( Q \) and \( b \) are isotropic hardening constants.

### 2.5.6 Uncoupled elastic-plasticity-creep

Discussions so far, have only accounted for modelling of time-independent plasticity, whereas cyclic loading at elevated temperatures will introduce time-dependent plasticity or creep effects. In order to simulate this material behaviour effectively, the inclusion of a creep equation, such as the Norton equation, needs to be combined with the isotropic and kinematic hardening models. These equations were combined in an uncoupled fashion by Shang et al. [2006] to predict the behaviour of superplastic forming dies. The predicted stress-strain response was used as a basis for life of the die. Deshpande et al., [2010a] also used an uncoupled elastic-plastic-creep model to predict cyclic creep-plastic and ratchetting behaviour in superplastic dies. It was found that life predictions generated from such predicted stress-strain data were broadly consistent with observed lives. However, in certain conditions, particularly for cyclic creep (ratchetting effect) and creep-plasticity interaction, the combination of the plasticity and creep equations gives unsatisfactory results when compared to experimental data (Krempl, 2000).

### 2.5.7 Two layer viscoplasticity model

The two-layer viscoplasticity model, also known as the two-dissipative mechanisms model, has been used in the modelling of exhaust manifolds and polymers. This model is in-built in ABAQUS commercial FE software and is capable of modelling materials in which significant time dependent as well as time independent plasticity
occurs. The original version of this phenomenological model was developed by Kichenin et al., [1996]. The two-layer viscoplasticity model is decomposed into an elastic-plastic network in parallel with an elastic-viscous network, as depicted in the rheological diagram of Figure 2.20. The contributions of elastic-plastic and elastic-viscous networks are apportioned by the user-specified parameter, $F$, which is defined by the following relationship:

$$F = \frac{K_v}{K_v + K_p}$$

(2.21)

where $K_v$ and $K_p$ are the elastic modulii of the elastic-viscous, and the elastic-plastic networks, respectively [ABAQUS, 2010].

![Rheological diagram of two-layer viscoplasticity model](image)

Figure 2.20. Rheological diagram of two-layer viscoplasticity model, adapted from Deshpande et al., [2010].

High values of the parameter $F$, provide a high contribution of elastic viscous network, and conversely low values of $F$, allow for large contributions from the elastic-plastic network [Kichenin et al., 1996]. In isothermal conditions, the total
strain is decomposed into elastic, $\varepsilon_e$, plastic, $\varepsilon_p$ and viscous, $\varepsilon_v$, strain components, given as follows:

$$\varepsilon = \varepsilon_e + (1 - F)\varepsilon_p + F\varepsilon_v$$  \hspace{1cm} (2.22)

The total stress of the model is divided into two different stresses, i.e., $\sigma_p$ and $\sigma_v$, which control the evolution of plasticity and viscous effects in each network, respectively.

Figiel and Günther [2008] utilised the two-layer viscoplasticity model to study the cyclic stress strain behaviour of SiC/Ti-6242 composite material. The model constants in the study were obtained by fitting the results of finite element simulations to quasi-static tensile and relaxation experimental tests carried out at various temperatures. The predicted FE results of the two-layer model showed good agreement with that of the measured experimental results; in addition to this the stress-strain behaviour of the two-layer model displayed comparable performance to that of the Chaboche model for the same monotonic test conditions, as depicted in Figure 2.21.
Charkaluk et al. [2002] studied the fatigue design of a cast iron exhaust manifold under thermomechanical loading conditions where two material models were applied to simulate the stress-strain response; one was a classical unified viscoplasticity model with linear kinematic hardening and the other one was the two-layer viscoplasticity model. The temperature material parameters were determined from isothermal strain controlled uniaxial tests and a stress relaxation test. The material parameters were then linearly extrapolated over a range of temperatures. Predicted regions of high stresses within the FE model of the exhaust manifold (shown in Figure 2.22(a)), showed good agreement with that of observed locations of cracking, (shown in Figure 2.22(b)).
In a similar approach Deshpande et al. [2010] determined the material constants for a high nickel-chromium alloy, XN40F, by conducting isothermal low cycle fatigue tests to determine the plasticity constants, $c$, $\gamma$, $Q$, and $b$, while the creep model constants $A$, $n$ were identified from isothermal stress relaxation tests carried out at various temperatures. The user-specified parameter, $F$, was identified by fitting the stress relaxation behaviour of the two-layer model to the relaxation test data. The calibrated material model was then used to simulate the stress-strain behaviour of a superplastic forming die under thermomechanical fatigue loading conditions. Further analysis by Deshpande et al., [2010b] utilised measured experimental data based on a representative thermomechanical cycle of a SPF die (see Figure 2.23(a) and (b)), which was compared against the predicted stress strain response of the die model. The results showed excellent agreement, as shown in Figure 2.23c, and form the basis of a life prediction methodology for this type of tooling. The model has also been used to simulate the polyelectrolyte material Nafion for under monotonic conditions [Solasi et al., 2008], where polymer materials exhibit significant time-
dependent deformation at relatively low temperatures. The model accurately captured a wide variety of stress-behaviour for a range of differing strain-rates.

Figure 2.23 Analysis of superplastic forming die, showing (a) contour plot of regions of localised inelastic strain, (b) loading conditions of representative cycle, and (c) comparison between measured stress-strain response and that of the predicted stress-strain response [Deshpande et al., 2010].

2.5.8 Unified Chaboche viscoplasticity model
The unified Chaboche viscoplasticity model has been used extensively [Koo and Lee, 2007; Aktaa and Schmitt, 2004; Hyde et al., 2010; Saad et al; 2011]

The viscoplastic flow rule is given as:

$$
\dot{\epsilon}_p = \left( \frac{f}{Z} \right)^n \ \text{sgn}(\sigma - \alpha)
$$

(2.23)
The yield criterion of the combined isotropic and kinematic hardening models, is given by the following equation:

\[
f = |\sigma - \alpha| - k - R
\]  
(2.25)

In the unified viscoplasticity model, the total stress can be decomposed into four parts, namely initial yield stress \(k\), drag stress \(R\), backstress \(\chi\) and viscous stress \(\sigma_v\), as given by the following equation:

\[
\sigma = \alpha + (R + k + \sigma_v) \text{sgn}(\sigma - \alpha)
\]
(2.26)

\[
\sigma_v = Z\dot{p}^{\nu/n}
\]
(2.27)

where \(Z\) and \(n\) are material constants, is the \(\sigma\) applied stress, \(f\) is the yield function; \(k\) is the initial cyclic yield stress, also known as the initial elastic limit, representing the initial size of the yield surface in a deviatoric plane; \(\alpha\) is the kinematic hardening parameter, also known as the back stress; and \(R\) is the isotropic hardening parameter, also known as the drag stress [Chaboche and Rousselier, 1983].

The unified Chaboche model has been successfully applied to P91 material by Saad et al., [2011], and, Koo and Kwon, [2011]. It was reported that the model is capable of accurately simulating the cyclic uniaxial stress-strain behaviour across a variety of strain ranges and temperatures. The model also accurately simulated the material softening behaviour effectively. However, the models ability to accurately capture strain rate effect has not yet been fully established. Another limitation to this model is that is does not distinguish between creep strain and plastic strain, which are useful when undertaking a creep-fatigue study and strain range partitioning analyses.
2.6 Branched pipe and header analysis

Work carried out by Shibli et al. [2007], reviews the high strength martensitic steel, T91. The study highlights the issues associated with the material and compares some of the perceived benefits with that of actual plant experience under creep-thermal fatigue conditions. It suggested that for thin-walled components the use of T91 or other 9Cr martensitic steel tubing is not suitable due to the high level of steam side oxidation. Brown [1994] used the finite element method to analyse the creep-fatigue life of boiler components. The resulting method was claimed to provide sufficient conservatism while still maintaining economically designed components, for combined-cycle, gas-fired stations subject to rapid start-up and cooling transients. The work also proposes the use of the finite element method for thermal, stress analysis aids in the correlation between operational events and damage mechanisms.

King et al. [1996] discussed recent experience in the condition assessment of economizer and superheater header components. The work went on to document regions of probable failure within these components and methods of assessment, with particular attention given to ligament failure. An example of such failure is shown in Figure 2.24.

![Figure 2.24 Ligament cracking at bore holes of economizer inlet header [King et al., 1996].](image)
Morgan et al., [1999] investigated ligament and stub-to-header cracking, by means of an acoustic emission monitoring (AEM) method, on an operational superheater plant header. The header had undergone some period of service before being inspected internally and externally via optical inspection. Using the optical inspection technique, it was found that substantial ligament cracking had occurred in the inner bore with the largest crack depth of magnitude 29 mm. Using the AEM system it was observed that the highest number of AE incidents occurred during incidents of temperature change, with the majority of incidents occurring at the stub-header weld intersection. The subsequent phase of testing involved the removal of the header to a laboratory environment where the temperature ramp rates could be controlled to a greater accuracy. The temperature ramp rates were increased to ensure growth of ligament bore cracking. Temperature variation and AE events were recorded as a function of time. A transient FE model of the region of the header under investigation was subjected to the same thermal loading conditions as the experimental header section. Inner bore crack was observed during cooling thermal transients and crack extension was measured, qualitatively, via an optical non-destructive testing inspection technique. It was found that the peak in AE activity (i.e. crack propagation signals) were coincident with peaks in tensile stresses above a certain threshold, in critical regions of the transient FE model. Kwon et al., [2006] also used a FE method to assess the remnant life of a service aged 2\text{\textperthousand}Cr-1Mo header. Inspection of the header revealed ligament cracking after 10800 h of service. Measured temperature data at the outside header surface and the stub pipe outer surface (which was taken to be an indication of the steam temperature entering the header) were utilised to calibrate a thermal FE model. The thermal history of this transient thermal FE model was then used to perform a sequential thermomechanical
elastic-plastic FE analysis. The FE analysis paid attention to thermally-driven stresses induced by plant start-up along with thermal transients that occur during steady-state operation. Critical regions of the header, where cracking is known to initiate, were analysed within the FE model via instantaneous axial stress profiles through the wall thickness. It was predicted that large axial compressive stresses occur at the edge of the bore during a heating transient, and the inverse stress-strain behaviour is predicted for a cooling transient.

The work by Paterson et al. [2002], presented a number of examples where component life monitoring was applied to in-service power plant components. This particular damage monitoring system was developed to provide plant operators with insight into quantification of damage and component life assessment. It was determined that heating transients were damaging to the weld toe region of the stub header interface, whereas cooling transients affect the inner bore holes. It is suggested that, headers are subject to the damage mechanisms of creep due to on-load (steady-state) temperature and pressure operation, whereas fatigue failures are due to start-up and shutdown. Rayner et al [2004] simulated creep behaviour of branched pipe intersections under steady-state operating conditions, for different pipe geometries using numerical and analytical methods. It was predicted that the inclusion of weld related material zones (i.e. weld and HAZ region) in a branched connection led to shortening of component life in terms of creep rupture stress, compared with FE simulation of a homogeneous component. It was also predicted that creep damage accumulated at geometric discontinuities. Kerezsi et al. [2000] developed an experimental test method to investigate crack growth in pressure vessels and piping materials. This method subjects a specimen to repeated thermal shocks, which is intended to simulate a plant component undergoing a thermal
transient which are associated with the plants operation. The material under investigation was a low carbon steel, grade AS 1548-1995, which is commonly utilised in power plant steam piping. Pre-notched specimens fabricated were preloaded in tension (which simulated internal steam pressure), and then heated to a temperature of 330 °C, before being quenched in a chemically controlled bath of room temperature water for a period of 7 s. This process was repeated for 500 cycles, before the specimen was removed and examined using an optical microscope. An estimate of crack initiation lifetime was defined as when a full-face crack was visible at the root of the notch. The results indicate that the primary stress or stress induced by preloading (which is a function of plant operating pressure) has little or no effect on crack initiation lifetime. However it was concluded that the internal steam pressure influences crack propagation. It should be noted that this approach has not taken account of cooling rate and at the influence of viscoplastic at the higher temperatures. In a similar approach, Smith et al. [1995] developed an experimental procedure using an induction coil and quenching system arrangement to reproduce high temperature cracks found in coal fired boiler tubes. It was proposed that the crack initiation was by an intergranular surface corrosion/thermal stress interaction mechanism and propagation was caused by thermal fluctuations due to the applied quenching. Samal et al., [2009], have used an on-line creep-fatigue monitoring system using measured data which includes thermal transients, pressure history and steam flow velocity, to calculate damage, corrosion rate, crack propagation, for a range of piping geometries. The monitoring system was calibrated against a transient FE model. Okrajni et al., [2008] have focused on developing a realistic header FE model, utilising an anisothermal material model with temperature dependent material
properties, in order to represent material behaviour under cyclic anisothermal condition similar in fashion to repeated plant start-up and shut-down.

Fournier et al. [2008] have discussed repeated start and stop operations and associated creep-fatigue type loading. Some work on the thermomechanical modelling of P91 in complex power plant geometries has been carried out by Okrajni et al. [2008], for example, including the effects of varying thermal and mechanical loads representative of plant start-up. A different approach taken by Mukhopadhyay et al [2000] utilises a system, again based on elastic stress analysis, to calculate creep damage and predict the life of an in-service 21/4Cr-1Mo power plant component under simplified loading conditions. Similarly, Samal et al [2009] extended this approach to a range of plant components and geometries. Header geometries under system loading were studied by Calonius et al., [2007]. This work examined an outlet header at a global and local level and highlighted the importance of modelling both the global geometry and local sub-sections of a header geometry.

Stub tube bore hole radial cracks, as shown in Figure 2.25, are common occurrences. By themselves, they are not a major concern, as they are generally self-limiting, but their potential coalescence into ligament cracks is a concern. As ligament cracks, they may grow into a through-wall crack with potential for major structural damage to the header. Importantly, Morgan et al., [1999] have highlighted difficulties with the repair of ligament cracks in creep damaged thick-walled Cr-Mo headers.
Figure 2.25 Photograph of ligament cracking in the inside surface of a header, at a bore hole intersection [Viswanathan, 2000].

2.7 Life prediction models

Life prediction and fatigue evaluation has become one of the major considerations in design of high temperature components. Therefore accurate life prediction techniques is fast receiving increased attention in terms of power plant component design. A number of damage models have been proposed to take account of the complex damage mechanisms involved in HT-LCF and creep-fatigue interactions.

2.7.1 Fatigue models

Liu et al., [2002] carried out thermomechanical fatigue life prediction using a strain energy based method (Ostergren) and an inelastic strain method (Coffin-Manson).

The Ostergren equation is as follows:

\[ N_f = C(\Delta \varepsilon_{in} \sigma_{max})^\beta \]  

(2.28)

where \( C \) and \( \beta \) are material constants, independent of temperature, \( \sigma_{max} \) is the maximum tensile stress in a given cycle, and \( \Delta \varepsilon_{in} \) and \( N_f \) are the inelastic strain range and number of cycles to failure respectively.
The Coffin-Manson equation is given as follows:

\[
\frac{\Delta \varepsilon_{in}}{2} = \varepsilon_f' \left(2N_f\right)^c
\]

(2.29)

where \( \varepsilon_f' \) and \( c \) are the fatigue ductility coefficient and fatigue ductility exponent respectively, which are temperature-dependent. Figure 2.26 shows the correlation between the predicted and observed life under TMF using Coffin-Manson (Figure 2.26a) and Ostergren (Figure 2.26b). Figure 2.26 suggests that the TMF fatigue life is dependent upon stress and strain rather than pure inelastic strain range and hence the Ostergren strain energy approach gave better fatigue life prediction than the Coffin-Manson strain method. In particular, the Coffin-Manson approach was shown to be significantly non-conservative for both TMF-IP and TMF-OP conditions.

Figure 2.26 Predicted and observed life under TMF by (a) the Ostergren and (b) Coffin-Manson methods [Liu et al., 2002].

Nagesha et al. [2002] predicted TMF life for 316L(N) stainless steel using an isothermal database and reasonable predictions were obtained by applying the Ostergren frequency modified damage function approach, which takes account of time dependent processes.
The modified Ostergren equation by Maier is given as follows:

\[ N_f = C(\Delta \varepsilon_{in} \sigma_{max})^\beta \times (\nu^*)^{(1-k)} \] (2.30)

where \( \nu^* \) is the effective frequency, which was calculated as \( \nu^* = \frac{1}{\tau + \Delta\tau} \), where \( \tau \) is the cycle time and \( \Delta\tau \) is the time per cycle during which creep damage occurs and \( k \) is a material constant identified from LCF data.

The strain range partitioning (SRP) method was first presented by Manson et al., [1971], who successfully applied the method to predict high temperature low cycle fatigue life of a 21/4Cr-1Mo steel. The SRP method considers the cyclic inelastic strain to be comprised of both plastic strain and creep strain. However the method also assumes that the plastic and creep strains do not contribute to damage by the same amount. It was therefore proposed that the strain within a closed hysteresis loop be partitioned into four subcomponents, as shown in Figure 2.27 [Halford et al., 1972], where the total inelastic strain range is described by the following expression:

\[ \Delta \varepsilon_{in} = \Delta \varepsilon_{pp} + \Delta \varepsilon_{cc} + \Delta \varepsilon_{pc} \] (2.31)

where \( \Delta \varepsilon_{pp} \) = plastic strain reversed by plastic strain, \( \Delta \varepsilon_{cc} \) = creep strain reversed by creep, and \( \Delta \varepsilon_{pc} \) = tensile plasticity reversed by compressive creep (with the inverse being true for \( \Delta \varepsilon_{cp} \)). The various contributions to the damage can then be calculated by the following:

\[ d = \frac{1}{N_f} = \sum \frac{f_{ij}}{N_{ij}} = \frac{f_{pp}}{N_{pp}} + \frac{f_{cc}}{N_{cc}} + \frac{f_{pc}}{N_{pc}} \left( or \frac{f_{cp}}{N_{cp}} \right) \] (2.32)
The fractional strain $f_{ij}$, for each type of strain can be estimated from the following equation:

$$f_{ij} = \frac{\Delta \varepsilon_{ij}}{\Delta \varepsilon_{in}}$$  \hspace{1cm} (2.33)

The relationship given by $\Delta \varepsilon_{ij} = A(N_f)^{\gamma}$, which is the Coffin-Manson relation, is required for each component of strain. The number of cycles to failure is estimated by $N_f = 1/d$.

Shang et al., [2006] utilised the SRP method to predict failure of a superplastic forming die. The die material was characterised using isothermal fatigue and creep-fatigue tests to determine the fatigue constants. The plastic and creep strains predicted by an FE model were used to determine the relative contributions of the component strains. Bi-linear damage summation relations were then used to predict failure. It should be noted however that no comparisons were made with failures of real components.

Figure 2.27 Schematic of partitioning of the strain components into the four types of inelastic strain [Halford et al., 1972].

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2.7.2 Multiaxial life prediction methods

In general real components operate under complex loading scenarios and are often subjected to repeated multiaxial loadings. Another important factor with respect to multiaxial fatigue life prediction is that in HCF testing it has been shown that the planes on which maximum shear strain and stress occurs are coincidental. However they will typically not coincide under LCF conditions because of the non-linear stress-strain relationship [You et al., 1996].

The critical plane approach has been found to be effective when dealing with multiaxial fatigue problems because the critical plane concept is based on the fracture mode and the initiation mechanism of cracks growing on preferred planes. Therefore the critical plane approach can not only predict fatigue life, but also the crack orientation or failure plane [Fuchs and Stephens, 2001]. This has led to the approach being adopted to predict the multiaxial fatigue life under constant amplitude, in-phase and out-of-phase loading [Chen et al., 2011]. The critical plane approach requires knowledge of the component stress and strain histories. A chosen fatigue indicator parameter (FIP), is evaluated on all candidate planes, defined in 3D by the angles \( \theta \) and \( \theta_R \), as shown in Figure 2.28. The method adopted by Das et al. [1999], utilises a Mohr's circle type method, where a given candidate plane is calculated using direction cosines, which are given as follows:

\[
\begin{align*}
    n_y &= -\sin \theta \cos \theta_R \\
    n_x &= \sin \theta \sin \theta_R \\
    n_z &= \cos \theta
\end{align*}
\] (2.34)

The normal stress and strain are calculated for the candidate plane:

\[
\begin{align*}
    \sigma_n &= \sigma_x n_x^2 + \sigma_y n_y^2 + \sigma_z n_z^2 + 2 \tau_{xy} n_x n_y + 2 \tau_{xz} n_x n_z + 2 \tau_{yz} n_y n_z \\
    \varepsilon_n &= \varepsilon_x n_x^2 + \varepsilon_y n_y^2 + \varepsilon_z n_z^2 + \gamma_{xy} n_x n_y + \gamma_{xz} n_x n_z + \gamma_{yz} n_y n_z
\end{align*}
\] (2.35)

(2.36)
For a given time history stress strain response at a given material point the normal stresses and strain are calculated for all candidate planes. Similar methods have been proposed by Wang and Brown [1993].

The critical plane is the plane upon which the FIP is a maximum. However, the selection of a suitable FIP has caused some debate and some of the more prevalent FIP's are described below.

Figure 2.28 Orientation of the unit normal, $n$, to a candidate plane [Das et al., 1999].

Brown and Miller [1973] formulated an approach where the cyclic shear strain and the normal strain on the maximum shear plane are the governing damage parameters, where failure is defined by the plane containing the largest value of

$$\left( \frac{\Delta \gamma_{\text{max}}}{2} + s\Delta \epsilon_n \right).$$

The approach is described by:

$$C = \frac{\Delta \gamma_{\text{max}}}{2} + s\Delta \epsilon_n$$

(2.37)

where $\Delta \epsilon_{\text{max}}/2$ is the maximum shear strain amplitude, $s$ is a material constant and $\Delta \epsilon_n$ is the normal strain range on the $\Delta \gamma_{\text{max}}$ plane.
A more common model is that of the Fatemi-Socie model where the fatigue damage is governed by the maximum shear strain amplitude and the maximum normal stress acting on that plane. It is based on physical interpretation of fatigue cracking, where cracks are found to be usually irregular in shape. During cyclic shear loading, crack growth through the material is inhibited by the interlocking action and frictional forces of this irregular shape. This will lead to an increase in fatigue life. If a tensile stress perpendicular to the crack is sufficiently high to open the crack the interlocking and frictional forces no longer be effective, thus increasing the crack driving force and propagating the crack. This will lead to a decrease in fatigue life.

As discussed earlier, Das et al [1999] presented a methodology to determine the multiaxial fatigue life of a SAE 1045 steel notched shaft. The methodology proposed the use of a finite element model to predict the stress-strain history of the component at a given point of interest in the model. A Fatemi-Socie approach was adopted in order to predict failure, which is given as follows:

\[ \varepsilon_{n} \sigma_{t} = \left( \frac{\sigma'_{f}}{E} \right)^{2} \left( 2N_{f} \right)^{2b} + \varepsilon'_{f} \sigma'_{f} \left( 2N_{f} \right)^{b+c} \quad (2.38) \]

\[ \gamma \left( 1 + K \frac{\sigma_{\text{max}}}{\sigma_{y}} \right) = \frac{\tau'_{f}}{G} \left( 2N_{f} \right)^{b} + \gamma'_{f} \left( 2N_{f} \right)^{c} \quad (2.39) \]

where \( c \) and \( b \) are the shear fatigue ductility exponent and strength exponent, respectively, and \( \tau' \) and \( \gamma' \) are the shear fatigue strength and ductility coefficient respectively, and \( \gamma \) and \( \varepsilon_{n} \) are the Brown and Miller parameters.

### 2.7.3 Cycle counting methods

To assess fatigue damage from a variable stress or strain history, a method to count the individual cycles is needed in order to estimate the overall damage contribution.
to a particular material point. The rainflow method has become the standard for cycle counting. Rainflow counting was first developed by Matsuishi and Endo [1968]. The rainflow method accounts for the sequence of cycles by identifying closed hysteresis loops in the stress-strain response. There are many variations of the rainflow cycle counting technique; however, the two most prevalent are that of the three-point and four-point techniques given by Downing and Socie, [1982] and Amzallag et al., [1994], respectively. However multiaxial loading conditions introduce problems with the aforementioned techniques, due to the fact that the hysteresis loops are not as well defined for multiaxial stress-strain cases as it is for the uniaxial response. In light of this, Bannantine and Socie [1991] applied the traditional rainflow method to count cycle of shear strain on a given plane; however they did not address the possibility that the maximum normal stress may not coincide with maximum shear strain. Wang and Brown [1993] proposed a cycle counting method for a damage model estimation method. This routine counts cycles of relative equivalent strain, where a cycle is defined by a zero to maximum increase in the relative equivalent strain. An alternative to this method is that of Langlais et al. [2003], which counts multiple stress and strain histories or channels, where, using a primary strain channel defines the countable cycles which are linked to the associated amplitudes in the auxiliary channels, thus ensuring a reversal in strain is matched to its corresponding stress amplitude. The method has been shown to work for both uniaxial and multiaxial stress-strain histories.

2.8 Summary

The literature review chapter summarises the salient concepts, experimental techniques and numerical methodologies relevant to the thermomechanical
characterisation of high temperature power plant materials. A review of creep, strain-controlled low cycle fatigue, thermomechanical fatigue and creep-fatigue interaction has been carried out to provide a background for the research being presented in this thesis. The work in the literature can be summarised as:

- Characterisation and development of material models for, high temperature creep of P91.
- Creep damage mechanics modelling of P91 and weldments.
- High temperature creep analyses of power plant connections, including P91 headers.
- Characterisation, and development of material models for high temperature low cycle fatigue, and initial work on TMF, of P91 and some initial work on thermomechanical modelling of power plant headers.

However, clear gaps in the literature relate to the following aspects:

- Thermomechanical fatigue characterisation of (service-aged) P91 material.
- Development of transient thermomechanical models of realistic, complex power plant connections where failure is typically observed.
- Development and validation of anisothermal, cyclic viscoplasticity material models for new and service-aged P91, covering the temperature range from 20 °C to 600 °C.
- LCF characterisation of welded P91 material.
- Development of three dimensional TMF life prediction methods for realistic (measured) thermomechanical cycling of plant.

The present thesis will address these gaps.
Chapter 3

EXPERIMENTAL TESTING

3.1 Introduction

This chapter describes the experimental work carried out in order to characterise the high temperature low cycle fatigue (HTLCF) behaviour of service-aged (SA) P91 steel and weldment material. The results presented in this chapter have emanated from a collaborative test programme carried out between NUI Galway (NUIG) and the University of Nottingham (UoN). The high temperature test capability at UoN is well established for isothermal and anisothermal strain controlled cyclic testing. During the formulation of the service-aged material test plan, a programme of tests was agreed for the high temperature testing. Concomitantly the HTLCF facility at NUI Galway was commissioned, and the setup, commissioning and validation of this of test equipment is described here. Validation of the NUIG HTLCF rig was achieved by repeating some of the initial tests that were carried out at UoN to ensure the rig was achieving the correct stress-strain behaviour in terms of hysteresis loops, along with capture of the softening and failure behaviour of P91 material. Comparisons are made between the experimental test results of the SA P91 base material and published P91 material experimental test data, tested under similar conditions from the literature.

The material under investigation here was extracted from a subcritical power plant superheater outlet header after 35,168 hrs of service at Lough Ree power station, courtesy of ESB Energy International (ESBI). The header previously operated under a normal service temperature of between 460 °C and 485 °C. The number of plant start-ups and shut downs experienced by the header, based on
service history data supplied by ESBI, is summarised in Table 3.1. Prior to installation, the header was fabricated by a hot rolling process, normalised at 1050 °C for 0.5 h and tempered at a temperature of 765 °C for 1 h. During operation superheater headers are subjected to repeated thermal stresses, due to (i) plant start-up and shut-down and (ii) attemperation transients. Furthermore, steady state operation will induce creep deformation. Due to the frequency of start-ups and exposure to elevated service temperatures, it is expected that damage evolution and changes in material microstructure have occurred in the service-aged material. It can therefore be anticipated that such exposure to high temperature static and dynamic stresses will affect the material mechanical behaviour, e.g. see Ennis and Czyska-Filemonowicz [2002] and Swindeman et al., [2003]. Once the header had been removed from the plant, a certain portion was used to manufacture base material (BM) specimens. The remaining material was used to fabricate a full penetration girth weld on the header in order to simulate a field repair, from which welded specimens were fabricated. These newly fabricated SA BM and welded specimens were used to conduct a program of LCF testing, for complete characterisation of all the material regions found on a P91 superheater header.

Table 3.1. Plant start-up schedule for the service aged P91 material from ESBI.

<table>
<thead>
<tr>
<th>Start-ups</th>
<th></th>
<th>Cold</th>
<th>Warm</th>
<th>Hot</th>
</tr>
</thead>
<tbody>
<tr>
<td>Total</td>
<td>65</td>
<td>21</td>
<td>22</td>
<td>22</td>
</tr>
</tbody>
</table>

3.2 HTLCF experimental procedures

3.2.1 General
As previously mentioned, experimental testing was carried out at two different test facilities, namely University of Nottingham and NUI Galway. The work carried out
at these two test facilities require different test set-up methodologies. These methodologies are discussed in the following sections.

3.2.2 University of Nottingham TMF experimental set-up
The initial HTLCAF testing and subsequent TMF tests was carried out on Instron 8862 TMF test rig (hereafter referred to as the UoN test rig) situated in the Department of Mechanical, Materials & Manufacturing Engineering, University of Nottingham. The UoN test rig can be used to carry out a variety of fatigue tests, including isothermal HTLCAF, TMF and stress relaxation tests. The test rig is comprised of water cooled hydraulic grips, servo-electric screw driven actuator, load frame inclusive of load cell, and control system. This configuration is illustrated in Figure 3.1. The rig can apply forces up to 100 kN and has a maximum frequency of 1 Hz and minimum frequency of 1 micrometer/hour [Saad, 2012].

Figure 3.1. Photograph of TMF experimental rig set-up at University of Nottingham.
Figure 3.2 shows a close-up view of the specimen, radio frequency (RF) coil and extensometer arrangement on the UoN test rig. Strain is measured by a high temperature extensometer, where the extensometer legs are made from ceramic rods, with a gage length of 12.5 mm. The extensometer is held in place using a mounting kit which ensures that a force of no greater than 300 g is exerted on the extensometer ceramic legs [Deshpande, 2009a].

Figure 3.2. Photograph of UoN TMF test rig showing; service-aged P91 test specimen (under test), RF coil and high temperature extensometer.

The specimen is heated by an electro-magnetic force (EMF) coil, which creates a magnetic field around the individual coils. The interaction between the specimen
(which is positioned in the middle of the coil) and the magnetic field generated by the coil creates the heating effect. The coil itself is made of copper tubing which is manufactured by wrapping the tubing around a solid former to give the required coil shape. This type of heating system allows for rapid heating of specimens in order to ensure that the specimen gage length attains the set target test temperature and has a uniform temperature along its length. Spot welding of five K-type thermocouples to the specimen surface along the gage length (see Figure 3.3) provides the user with an indication of temperature gradient along the gage length during the calibration process. Typically a temperature gradient of ± 10 °C along the gage length is required. If necessary the individual coil pitch or number of coil turns on the RF coil may need to be adjusted in order to achieve a uniform temperature along the gage length. Thermal calibration is required for each new test specimen configuration and material. The finalised coil design for the service-aged P91 test specimens is depicted in Figure 3.4.

![Figure 3.3. Photograph of service-aged P91 specimen post thermal calibration with attached thermocouples.](Image)
For the service-aged P91 isothermal fatigue and TMF strain-controlled tests, the temperatures during the tests were measured by using two K-type thermocouples spot welded to the surface of the test specimen; one thermocouple acted as the main control, which indicates the true temperature for a test and another thermocouple was used to check the reading of the main control thermocouple before starting the test. Upon completion of the first three tests it was noticed that primary cracking was occurring where the main thermocouple had been welded to the gage length surface. Therefore to eradicate any potential of the attached thermocouple being the source of crack initiation (which could be contributory to premature fatigue failure), on the gage length, the position of the main control thermocouple was moved to the specimen shoulder. To ensure correct operation of the thermocouples, the two thermocouple wires must be less than 1 mm when welded to the surface of the specimen (see [Deshpande, 2009]).

Figure 3.4. Photograph of coil design for service-aged P91 HTLCF testing at UoN.
A detailed account of the pre-test set-up is given by Saad [2012], with the key points synopsised as follows:

1. Verification of the Young’s modulus is carried out at room temperature to ensure correct operation of the extensometer; a small load is applied to the specimen (low enough to ensure no plastic deformation occurs). Correct operation of the extensometer results in a reasonable smooth plot of linear data and the measured Young’s modulus were comparable to reference values.

2. The specimen is heated up to the target temperature, and allowed to remain at that temperature for approximately 5 minutes to allow the temperature to stabilise before testing commences.

3.2.3 NUI Galway experimental set-up
Additional HTLCF material testing was carried out on the Instron 8800 at NUI Galway (hereafter referred to as the NUIG test rig). The NUIG test rig is comprised of water cooled hydraulic pull-rods, servo-electric actuator, load frame inclusive of load cell, and control system. This configuration is illustrated in Figure 3.5. The rig is capable of carrying out axial strain-controlled, plastic strain-controlled or stress-controlled cyclic testing, and can host rectangular, cylindrical and tubular specimens. Heating of the specimen is achieved by a furnace and temperature control system, where a three-zone split furnace with embedded heating elements, gives a maximum and minimum specimen temperature of 1000 °C and 300 °C, respectively (depending on load string configuration). The temperature control system is comprised of three Eurotherm 3216 controllers in a master-slave configuration capable of following a single ramp to set-point. A high temperature axial gage extensometer, for use up to
1000 °C is utilised to control strain during strain-controlled tests and for data acquisition during stress-controlled tests. The extensometer gage length is 12.5 mm with a maximum and minimum travel of +20%, and -10% respectively. The extensometer legs are comprised of alumina chisel end rods and attaches to the specimen using ceramic wrap-around cord. The mounting arrangement of the high temperature extensometer is depicted in Figure 3.6.

Two software packages are utilised to carry out HTLCF tests, namely FT Console and LCF3. The former is used to control the position of the actuator prior to and during specimen installation. The FT Console software registers and maintains all the test rig settings, as well as setting safety limits to ensure that the position of the actuator, applied load and strain are within the user defined limits before, during and after testing. Safety features include a 'Specimen protect' function which will activate if a limit is tripped. The 'Specimen protect' function can also operate during heating of test specimen prior to testing. This function will allow the actuator to change position to ensure the specimen is not subjected to an excessive compressive load due to thermal expansion during heat-up. The latter (LCF3) software package registers and maintains all the fatigue test settings, and is used to create test methodologies and execute the test itself, as well data acquisition and determination of specimen failure.
Figure 3.5. Photograph of experimental rig set-up of the Instron 8800 high temperature low cycle fatigue rig in NUI Galway.

Other key features of LCF3 are, (i) automatic calculation of modulus within user specified limits, (ii) automatic adjustment of gage length at test temperature (i.e. compensation for thermal expansion of test material to ensure extensometer is zeroed prior to test commencing), (iii) support of sine, triangular and trapezoidal waveforms, (iv) ability to perform stress controlled run-out tests at up to 50Hz, and (v) the ability to stop and re-start a test at any point. The LCF3 software package complies with ASTM E606-04, BS 7270 (2006), and ISO 12106 (2003) standards.
Further details on the experimental procedure to conduct a HTLCF test on the NUIG test rig is given in Appendix A.

### 3.3 Experiments

The pre-service chemical composition of the service-aged P91 material is shown in Table 3.2. The chemical composition is assumed not to have changed during the period of service. Also included in Table 3.2 is the composition of two other P91 materials from the literature [Nagesha et al., 2002; Saad et al., 2010]. Table 3.2 shows the similarity in terms of chemical composition between the service aged P91 material and other P91 material from different casts.

The dimensions of the UoN and NUIG HTLCF test specimens are shown in Figure 3.7 and Figure 3.8, respectively. The test specimens were fabricated from the
longitudinal direction of the header i.e. parallel to the rolling direction. A series of isothermal low cycle fatigue tests on the service-aged P91 base material were conducted at room temperature, 400 °C, 500 °C and 600 °C under a number of controlled strain-range and strain-rate conditions, as summarised in Table 3.3. The welded specimen HTLCF tests carried out at 400 °C and 500 °C is presented in Table 3.4. The fatigue life, $N_f$, for each strain-controlled test was taken as the cycle number corresponding to a 30 % drop in stress range relative to the stress range of the 150th cycle.

Table 3.2. Chemical composition of service-aged P91 material and that of P91 material from the literature (wt%)

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>N</th>
<th>Cr</th>
<th>Mo</th>
<th>Nb</th>
<th>Cu</th>
<th>V</th>
<th>Al</th>
<th>P</th>
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<td>0.07</td>
<td>—</td>
<td>0.204</td>
<td>0.007</td>
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<td>8.72</td>
<td>0.9</td>
<td>0.08</td>
<td>—</td>
<td>0.22</td>
<td>—</td>
<td>0.012</td>
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<td>1.02</td>
<td>0.07</td>
<td>0.1</td>
<td>0.24</td>
<td>—</td>
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</table>

**Nagesha et al., [2002]
*Saad et al., [2010]
<table>
<thead>
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<th>Test No.</th>
<th>Temp (°C)</th>
<th>Strain Rate (%/s)</th>
<th>Strain Range (%)</th>
<th>Wave time (s)</th>
<th>Waveform</th>
<th>Waveform</th>
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<tr>
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<td></td>
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<td>0.8</td>
<td>48</td>
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<tr>
<td>7</td>
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</tr>
<tr>
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<td>80</td>
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</tr>
<tr>
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</tr>
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</tr>
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<tr>
<td>12</td>
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<td>48</td>
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</tr>
<tr>
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<td>Tension-hold (hold = 120 s)</td>
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<td></td>
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</table>

Table 3.3. Summary of isothermal P91 base material test program
### Table 3.4. Summary of isothermal welded specimen test program

<table>
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<tr>
<th>Test No.</th>
<th>Test type</th>
<th>Temp (°C)</th>
<th>Strain Rate (%/s)</th>
<th>Strain Range (%)</th>
<th>Wave time (s)</th>
<th>Waveform</th>
</tr>
</thead>
<tbody>
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<td>1.0</td>
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<td>Tension-hold (hold = 120 s)</td>
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<tr>
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<td>17</td>
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<td>Relaxation</td>
<td>1.0</td>
<td>10,020</td>
<td>Tension-hold (hold = 10,000 s)</td>
</tr>
</tbody>
</table>

### 3.3.1 Specimen design

Each fatigue rig has its own particular type of specimen design. The UoN test rig requires the specimen designs shown in Figure 3.7, which are cylindrical button end specimens. Isothermal LCF tests require a solid specimen (Figure 3.7a), whereas TMF tests require hollow specimens (Figure 3.7b) which permit additional cooling of the specimen under dynamic thermal conditions. The gage length of the specimen is 15 mm. A specialist contractor (Takumi Precision Ltd.) was employed to fabricate the UoN specimens. The NUI Galway test rig requires a shorter cylindrical specimen with threaded ends, as shown in Figure 3.8. These specimens were manufactured in-house at NUIG.
Figure 3.7. University of Nottingham test specimens for (a) isothermal LCF, and (b) thermomechanical fatigue tests (Note all dimensions are in millimetres).

Figure 3.8. NUI Galway HTLCF test specimen (Note all dimensions are in millimetres).
3.3.2 Fabrication of NUI Galway test specimens

In order to carry out HTLCF testing of the weld material, a weld must be fabricated which is located on the service-aged P91 header. Atlantic Projects Company Ltd. (APC), who specialise in fabrication of header welds and weld repairs were employed, to carry out this task. Figure 3.9 depicts the prepared header section prior to welding. The weld was fabricated by cutting the header in two, and machining a chamfered edge and a square edge. The header was then re-aligned and a (P91) backing plate was affixed to ensure deposition of weld material between the gap created by the two edges. The weld itself was generated by manual metal arc welding, whereby a P91 filler electrode coated in a protective flux is deposited circumferentially along the backing plate surface. Once an entire electrode has been deposited onto the surface, (which signifies one run or pass of welding), the slag (a brittle coating covering the weld metal) on the freshly generated weld run must be removed. This is carried out manually using an abrasive grinding disk. To ensure all slag is removed, some of the weld material is also removed, which ensures all possible inclusions have been eradicated. This welding and grinding process is repeated until the weld gap is entirely filled with weld material. The chemical composition of the weld filler wire is given in Table 3.5.

Figure 3.10 depicts the completed weld on the header section. It should be noted that a post weld heat treatment process was carried out in order to relieve residual stresses. This was performed by APC, where heating blankets were applied to the internal and external surfaces of the header which maintained the weld region at a constant temperature of 760 °C for a period of 80 minutes; after this the weld was allowed to cool at a rate of 50 °C/hour. A cross-section of the weldment was taken to assess to weld material in terms of the dimensional accuracy of the weld. It
can be seen from Figure 3.11 that the HAZ at the square edge is somewhat irregular in terms of straightness; this can be attributed to the nature of the multi-pass welding operation.

Figure 3.9. Photograph of prepared service-aged P91 header section prior to welded, courtesy of APC Ltd.

Figure 3.10. Photograph of completed weld of header section, courtesy of APC Ltd.
Table 3.5: Chemical composition (wt%) of the P91 weld electrode (filler) material.

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>N</th>
<th>Cr</th>
<th>Mo</th>
<th>Nb/Cb</th>
<th>V</th>
<th>Al</th>
<th>P</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>P91 electrode</td>
<td>0.09</td>
<td>0.6</td>
<td>0.2</td>
<td>0.04</td>
<td>9.0</td>
<td>1.1</td>
<td>0.05</td>
<td>0.2</td>
<td>-</td>
<td>-</td>
<td>0.8</td>
</tr>
</tbody>
</table>

Figure 3.11. Photograph of cross-section of the weldment showing heat affected zone regions and weld material region.

Upon visual inspection of the header cross-section of Figure 3.11, the header was cut into strips from which specimens could be manufactured. Figure 3.12 (a - c) is intended to represent a strip of material which has been cut from the header and also depicts the positions from which the test specimens were fabricated. Figure 3.12a shows the specimen positioned in the parent material region of the header section. Figure 3.12b shows the specimen positioned in the middle of the weld material, creating an all-weld test specimen. The use of the chamfered edge creates an extended gap, thus ensuring that the majority of the test specimen is comprised of weld material. Figure 3.12c illustrates the position of the cross-weld specimen, whereby the HAZ is positioned in the middle of the gage length, creating a cross-
weld specimen. The total width of the HAZ is 2.7 to 3 mm which is in similar to measurements given by [Shankar et al., 2011].

It should also be noted that the use of the square edge allows the HAZ to be aligned perpendicular to the longitudinal axis of the test specimen. Figure 3.13 shows a photograph of a fabricated service-aged P91 specimen.

Figure 3.12. Positions of different HTLCF fatigue specimen, showing (a) base material specimen, (b) weld material specimen, and (c) cross-weld specimen which is inclusive of the HAZ region. All dimensions in mm.
3.4 Service aged P91 base material experimental results

3.4.1 Strain range effect
Figures 3.14 (a)–(d) depict the measured cyclic softening of the SA P91 base material at four temperatures and for the three strain ranges studied, for a constant strain rate of 0.033 %/s. In these figures the peak tensile stress for each cycle is plotted. Material softening was observed for all strain-ranges and for all four temperatures. It was observed that the cyclic softening of the material is a function of both total strain range and temperature. It was found for all four temperatures the softening increased with increasing strain range, and the greatest amount of softening was coincidental with the largest application of total strain, and that a decreasing amount of softening occurred with decreasing temperature. The stress decreases continuously in three stages, initially at a rapid rate (stage 1), then a more gentle (almost linear) rate (stage 2), followed by a rapid reduction close to final failure (stage 3).

Figures 3.14 (e)–(h) show the evolution, with number of cycles, of the relationship between tensile stress amplitude and plastic strain amplitude for the same four temperatures. A power law relationship is commonly used to represent this relationship for cyclic analyses of components and materials, as follows:

\[
\frac{\Delta \sigma}{2} = K\left(\frac{\Delta \varepsilon}{2}\right)^n
\]

(3.1)
where $K'$ and $n'$ are the cyclic strength coefficient and cyclic strain hardening exponent, respectively. The temperature-dependent $K'$ and $n'$ values for Equation 3.1 were extracted for the half-life, (which in this case is taken to be half the number of cycles to specimen failure), and are tabulated in Table 3.6 alongside corresponding values for a P91 material from a different cast, from Nagesha et al. [2002], hereafter referred to as Material 1. The chemical composition of Material 1 is given in Table 3.2. At 500 °C the SA P91 coefficients $K'$ and $n'$ give a hardening cyclic stress response below $\Delta\varepsilon_{\text{in}} = 4.8 \%$, and at 600 °C the SA exhibits a hardened response below $\Delta\varepsilon_{\text{in}} = 1.2 \%$, i.e. the present service-aged material is found to be stronger (cyclically) than Material 1 from Nagesha et al. [2002] at lower inelastic strain ranges.

Table 3.6. Material constants at different temperatures for parent material.

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Service aged</th>
<th>Material 1**</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$K'$</td>
<td>$n'$</td>
</tr>
<tr>
<td>20</td>
<td>891.32</td>
<td>0.102</td>
</tr>
<tr>
<td>400</td>
<td>639.9</td>
<td>0.083</td>
</tr>
<tr>
<td>500</td>
<td>467.3</td>
<td>0.068</td>
</tr>
<tr>
<td>600</td>
<td>273.6</td>
<td>0.04</td>
</tr>
</tbody>
</table>

**Nagesha et al., [2002]

The stress-strain loops for the first ($N = 1$) cycle are shown in Figure 3.15. The effect on inelastic strain range is naturally dependent on temperature; hence, for example the inelastic strain range at 600 °C is about 1.5 times that at 20 °C.
Figure 3.14. (a)–(d) Cyclic softening curves for P91 base material for a constant strain rate of 0.033 %/s at different temperatures. (e)–(h) Evaluation of stress-plastic strain power law relationship for various numbers of cycle for P91 base material at different temperatures.
Figure 3.15. Measured initial \((N = 1)\) stress-strain loops for SA P91 base material, for different strain ranges and temperatures: (a) 20 °C, (b) 400 °C, (c) 500 °C, and (d) 600 °C.

Figure 3.16 shows a comparison between a HTLCF carried out on the UoN test rig and a test carried out at NUIG for the same test conditions. Figure 3.16a compares the stress-strain response at \(N = 1\) and Figure 3.16b shows the cyclic softening evolution of the maximum tensile stress with increasing number of cycles. As can be seen from Figure 3.16a there is a good agreement in the cyclic stress-strain response with the UoN exhibiting a marginally higher stress range. The cyclic softening behaviour of Figure 3.16b shows good agreement, with a qualitatively
similar softening evolution and number of cycles to failure. This result demonstrates that the newly-commissioned NUIG test rig is capable of capturing the cyclic stress-strain response of such materials.

Figure 3.16. Comparison between a UoN and NUIG HTLCF test on SA P91 base material at 500 °C for a strain range of 1% and a strain rate of 0.033%/s, showing (a) the stress-strain response for the first cycle, and (b) the evolution of the maximum tensile stress with number of cycles.

3.4.2 Strain rate effect

Figure 3.17 depicts the effect of strain rate on SA P91 base material for a strain range of 1% at three temperatures. A significant strain-rate effect is observed at 600 °C, but the effect is small at lower temperatures at lower temperatures. This observation supports the findings of Swindeman [1988], who reported strain rate effects only at temperatures above 500 °C for a P91 steel.
3.4.3 Comparison of SA P91 with material data published in the literature

Figure 3.18 (a, c and e) shows the measured stress-strain hysteresis ($N = 1$) loops of the present service-aged P91 base material compared with the P91 test data published, (hereafter referred to as Material 2) by Saad et al. [2010], across the range of temperatures at a 1% strain range. Figure 3.18 (b, d and f) shows the comparison after 100 cycles. For all temperatures, the $N = 1$ stress amplitude of the service-aged material is greater (~ 7% to 8%), which indicates that the SA P91 material cyclically harder than Material 2. In contrast, after 100 cycles, the service-aged and Material 2 (from Saad et al., [2010]) show similar stress ranges, except at 400 °C, where the service-aged material remains about 7% greater.
Figure 3.18. Comparison of evolutions of stress-strain hysteresis loops for service aged P91 and the P91 material (Material 2) from Saad et al., [2010], at a range of temperatures and 1% strain range.

Figure 3.19a shows the stress-strain response from cyclic tension-hold tests. These tests allow determination of stress relaxation behaviour. Figure 3.19b shows the stress response for the first quarter cycle. Again the effect of temperature is
considerable. The 600 °C test shows a significantly larger amount of stress relaxation than the 400 °C case. The degree of relaxation increases with increasing temperatures.

Figure 3.19. Stress relaxation data showing (a) tension hold strain controlled test, and, (b) relaxation of stress during the first quarter cycle for SA P91 base material, for a Δε = 1 % and a strain rate of 0.033 %/s.

3.4.4 Fatigue life results
Table 3.7 presents the fatigue lives of the service-aged P91 base material isothermal strain-controlled tests. Figure 3.20 shows fatigue life data in terms of the total strain amplitude plotted against number of reversals to failure. These results indicate that the fatigue life of the service-aged P91 material is heavily dependent on temperature between 20 °C and 500 °C. There is still a temperature effect between 500 °C and 600 °C, but at higher strain amplitudes (i.e. Δεₜ = 1 %) the effect of temperature is diminished.
Table 3.7. Isothermal base material fatigue lives for SA P91 base material.

<table>
<thead>
<tr>
<th>Test No.</th>
<th>Test type</th>
<th>Temp (°C)</th>
<th>Strain Rate (%/s)</th>
<th>Strain Range (%)</th>
<th>Inelastic strain Range (%)</th>
<th>Max. Tensile Stress (MPa)</th>
<th>N_f</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Base material</td>
<td>20</td>
<td>0.033</td>
<td>0.6</td>
<td>0.134</td>
<td>455</td>
<td>14281</td>
</tr>
<tr>
<td>2</td>
<td>Base material</td>
<td>400</td>
<td>0.033</td>
<td>0.8</td>
<td>0.257</td>
<td>478</td>
<td>5139</td>
</tr>
<tr>
<td>3</td>
<td>Base material</td>
<td>400</td>
<td>0.033</td>
<td>1.0</td>
<td>0.479</td>
<td>500</td>
<td>2677</td>
</tr>
<tr>
<td>4</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>0.6</td>
<td>0.193</td>
<td>365</td>
<td>10618</td>
</tr>
<tr>
<td>5</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>0.8</td>
<td>0.374</td>
<td>386</td>
<td>4532</td>
</tr>
<tr>
<td>6</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>1.0</td>
<td>0.562</td>
<td>406</td>
<td>2057</td>
</tr>
<tr>
<td>7</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>1.0</td>
<td>0.546</td>
<td>400</td>
<td>2022</td>
</tr>
<tr>
<td>8</td>
<td>Base material</td>
<td>500</td>
<td>0.025</td>
<td>1.0</td>
<td>0.547</td>
<td>405</td>
<td>1763</td>
</tr>
<tr>
<td>9</td>
<td>Base material</td>
<td>500</td>
<td>0.1</td>
<td>1.0</td>
<td>0.547</td>
<td>405</td>
<td>2120</td>
</tr>
<tr>
<td>10</td>
<td>Base material</td>
<td>500</td>
<td>0.1</td>
<td>1.0</td>
<td>0.575</td>
<td>384</td>
<td>1644</td>
</tr>
<tr>
<td>11</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>0.6</td>
<td>0.243</td>
<td>312</td>
<td>6832</td>
</tr>
<tr>
<td>12</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>0.8</td>
<td>0.431</td>
<td>313</td>
<td>2757</td>
</tr>
<tr>
<td>13</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>1.0</td>
<td>0.631</td>
<td>323</td>
<td>1610</td>
</tr>
<tr>
<td>14</td>
<td>Base material</td>
<td>500</td>
<td>0.033</td>
<td>1.0</td>
<td>0.616</td>
<td>334</td>
<td>1374</td>
</tr>
<tr>
<td>15</td>
<td>Base material</td>
<td>500</td>
<td>0.025</td>
<td>1.0</td>
<td>0.633</td>
<td>330.0</td>
<td>1413</td>
</tr>
<tr>
<td>16</td>
<td>Base material</td>
<td>500</td>
<td>0.1</td>
<td>1.0</td>
<td>0.615</td>
<td>339.8</td>
<td>1284</td>
</tr>
<tr>
<td>17</td>
<td>Base material</td>
<td>500</td>
<td>0.1</td>
<td>1.0</td>
<td>0.67</td>
<td>301.8</td>
<td>1440</td>
</tr>
<tr>
<td>18</td>
<td>Base material</td>
<td>600</td>
<td>0.033</td>
<td>0.6</td>
<td>0.322</td>
<td>230</td>
<td>2961</td>
</tr>
<tr>
<td>19</td>
<td>Base material</td>
<td>600</td>
<td>0.033</td>
<td>0.8</td>
<td>0.533</td>
<td>219.4</td>
<td>1827</td>
</tr>
<tr>
<td>20</td>
<td>Base material</td>
<td>600</td>
<td>0.033</td>
<td>1.0</td>
<td>0.648</td>
<td>279.7</td>
<td>1225</td>
</tr>
<tr>
<td>21</td>
<td>Base material</td>
<td>600</td>
<td>0.033</td>
<td>1.0</td>
<td>0.706</td>
<td>233.2</td>
<td>1244</td>
</tr>
<tr>
<td>22</td>
<td>Base material</td>
<td>600</td>
<td>0.025</td>
<td>1.0</td>
<td>0.752</td>
<td>202.7</td>
<td>1426</td>
</tr>
<tr>
<td>23</td>
<td>Base material</td>
<td>600</td>
<td>0.1</td>
<td>1.0</td>
<td>0.706</td>
<td>241.7</td>
<td>1331</td>
</tr>
<tr>
<td>24</td>
<td>Base material</td>
<td>600</td>
<td>0.1</td>
<td>1.0</td>
<td>0.782</td>
<td>221.6</td>
<td>980</td>
</tr>
</tbody>
</table>
In order to evaluate the fatigue life of the service-aged P91 base material a Coffin-Manson life prediction approach is adopted, which relates the number of reversals to failure, $2N_f$, to inelastic strain range, $\Delta \varepsilon_{in}$, as follows:

$$\frac{\Delta \varepsilon_{in}}{2} = \varepsilon'_f \left(2N_f\right)^c$$  \hspace{1cm} (3.2)

where $\varepsilon'_f$ and $c$ are the fatigue ductility coefficient and fatigue ductility exponent, respectively, which are temperature-dependent. Figure 3.27 shows the Coffin-Manson plots for the four temperatures. The material constants, obtained from the data in Figure 3.21, are tabulated in Table 3.8. The influence of temperature appears rather weak over the strain-life considered, given the scatter in the data. However the
following observations can be made: (i) the 500 °C life is lower than the 400 °C life, (ii) the 600 °C life is lower than both 400 °C and 500 °C lives for lower strain ranges, and (iii) at high strain ranges scatter makes distinctions difficult.

Figure 3.21. Inelastic strain-life data of service aged base material for all temperatures (20 °C to 600 °C).

Figure 3.22 shows that across all temperatures, for serviced aged material, the Coffin-Manson data fit agrees with the measured data to within a factor of less than 1.5.
Table 3.8. Material constants at different temperatures for P91 base material.

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Service aged</th>
<th>Material 1**</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\varepsilon'_f$</td>
<td>$c$</td>
</tr>
<tr>
<td>20</td>
<td>1.39</td>
<td>-0.74</td>
</tr>
<tr>
<td>400</td>
<td>0.41</td>
<td>-0.60</td>
</tr>
<tr>
<td>500</td>
<td>0.29</td>
<td>-0.57</td>
</tr>
<tr>
<td>600</td>
<td>8.11</td>
<td>-0.98</td>
</tr>
</tbody>
</table>

** Nagesha et al., [2002]

Figure 3.22. Coffin-Manson predicted lives versus observed lives.

To predict failure under thermomechanical type loading i.e. varying temperature conditions, an energy fatigue model is commonly adopted, such as the Ostergren fatigue model [Ostergren, 1967].
Figure 3.23 depicts the number of cycles to failure, $N_f$, versus the Ostergren fatigue parameter, $C$, (see Equation 2.28), i.e. the product of the inelastic strain range and maximum tensile stress, calculated at the half-life for each test. The material constants, obtained from the data in Figure 3.23, are tabulated in Table 3.9.

![Figure 3.23: Ostergren Fatigue Life Data](image)

Figure 3.23. Ostergren fatigue life data of service aged base material for all temperatures (20 °C to 600 °C) compared to that of a P91 material published in the literature, at 538 °C [EBI and McEvily, 1984].

Figure 3.23 also shows a comparison between the service-aged P91 for all temperatures and the P91 material HTLCF tests of EBI and McEvily [1984] (dashed line) at a temperature of 538 °C. For lower Ostergren parameters ($<\sim 1.5$) the EBI and McEvily material generally exhibits a lower number of cycles to failure than the SA P91 material, and the converse is true for Ostergren parameter values greater than...
1.5. However the influence of temperature remains an important factor, which is illustrated in Figure 3.24. The fitted trend generally underestimates fatigue lives below 500 °C (for an Ostergren parameter greater than 1), and for temperatures of 500 °C and over, the model tends to overestimate the fatigue lives. Figure 3.24 shows that the data fit generally agrees with the measured lives to within a factor of 2.

![Figure 3.24](image_url)

Figure 3.24. Observed fatigue life versus predicted fatigue life using the Ostergren fatigue life model.

Table 3.9. Comparison of Ostergren fatigue parameters, determined from the service-aged P91 material and the P91 material data of EBI and McEvily [1984].

<table>
<thead>
<tr>
<th>Material</th>
<th>$C$ (MPa. mm/mm)</th>
<th>$\beta$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Service-aged</td>
<td>5762</td>
<td>-2.14</td>
</tr>
<tr>
<td>EBI and McEvily, [1984]</td>
<td>4523</td>
<td>-1.6</td>
</tr>
</tbody>
</table>
3.4.5 Thermomechanical fatigue

The thermomechanical fatigue tests were carried out to characterise the behaviour of the SA P91 material under both variable temperature and strain conditions, where the temperature was cycled between 400 °C and 600 °C in both TMF-IP and TMF-OP tests. The tests were carried out under a constant strain rate of 0.025 %s⁻¹ and a strain range of 1 %. Figure 3.25 depicts the stress-strain response of the TMF-IP test at \( N = 1 \) and at the half-life. The initial cycle (\( N = 1 \)) exhibits a compressive mean stress of -60 MPa. After 509 cycles this effect is diminished to a value of -36 MPa due to material softening. Figure 3.26 shows the TMF-OP stress-strain response. In this case, the first cycle exhibits a positive mean stress of 73 MPa, which increases to a magnitude of 87 MPa after 244 cycles (half-life). This behaviour is anticipated to reduce the fatigue life of the material as a result of accentuated crack growth due the positive mean stress effects.

Figure 3.25. Measured TMF-IP cyclic stress-strain response at \( N = 1 \) and at the half-life with \( T_{\min} = 400 \) °C and \( T_{\max} = 600 \) °C for SA P91 base material.
Figure 3.26. TMF-OP cyclic stress-strain response at $N = 1$ and at the half-life $T_{\text{min}} = 400 \, ^\circ\text{C}$ and $T_{\text{max}} = 600 \, ^\circ\text{C}$ for SA P91 base material.

The cyclic stress evolution of the TMF-IP and TMF-OP tests is shown in Figure 3.27. The isothermal fatigue tests of $400 \, ^\circ\text{C}$ and $600 \, ^\circ\text{C}$ are also included in Figure 3.27, which correspond at the minimum and maximum temperatures applied during the TMF tests. In all cases, a large amount of softening is observed from the first cycle onwards. The IP and OP tests both show significantly higher stresses $>100 \, \text{MPa}$ than the isothermal tests. The IP and OP tests showed an almost identical stress amplitude evolution over the first $\sim 350$ cycles. However, the OP test exhibits a significantly shorter fatigue life than the IP test, and both the IP and OP tests exhibited shorter fatigue lives than the isothermal test. A possible reason for the reduced fatigue life of the OP compared with the IP and isothermal tests may be attributed to positive mean stress effects.
Figures 3.28 depicts the results from the Ostergren life prediction model (using the constants identified from the isothermal tests in Table 3.9) under TMF loading conditions. This figure indicates that the Ostergren model is conservative by a factor of 2, albeit with a limited number of TMF test results.

Figure 3.27. Measured cyclic stress response of TMF-IP, TMF-OP and isothermal fatigue tests conducted under a constant strain rate of 0.025 %/s (T_{min} = 400 °C, T_{max} = 600 °C for TMF).

Figure 3.28. Predicted TMF fatigue life using isothermally identified Ostergren constants, versus observed fatigue life.
3.4.6 Ratchetting
A series of stress-controlled ratchetting tests were carried for various temperatures and stress rates. Figure 3.29 depicts the results from stress-controlled tests for a variety of temperatures, for a constant stress rate of 50 MPa/s. The figure shows the typical evolution of ratchetting strain, which is defined as the maximum strain in each cycle as a function of cycle number. The tests shown in Figure 3.29 were carried out under tensile mean stress conditions, with a stress ratio of $R = -0.5$. It is clear from Figure 3.29 that accumulation of ratchet strain is heavily temperature dependent, where the 550 °C test exhibits a dramatic increase in ratchet strain rate compared with the tests at the lower temperatures.

In order to establish a test-end criterion, the tests were stopped when either (i) the ratchetting strain achieved 5% strain or (ii) 500 cycles was reached. This failure was necessary because of the limited stroke of the extensometer used. This test-end criterion is consistent with that used by Yaguchi and Takahashi [2005a].

Figure 3.29. Measured evolution of maximum tensile strain for various temperatures under stress-controlled ratchetting tests for SA P91 base material.
Figure 3.30 shows a comparison of the evolution of the maximum tensile strain for different stress rates at 400 °C. It is clear from Figure 3.30 that the material shows little evidence of stress rate sensitivity, for this particular set of loading conditions. The ratchet strain effectively shakes-down after the first 20 cycles.

Figure 3.30. Uniaxial ratchetting behaviour of SA P91 base material for various stress rates at 400 °C.

The effect of stress rate on ratchetting behaviour at the higher temperature of 550 °C is shown in Figure 3.31. These tests were carried out under the same loading conditions as the 400 °C tests of Figure 3.30. Figure 3.31 shows the sensitivity to stress rate at 550 °C, whereby an increased ratchet strain rate occurs with decreased stress rate. This behaviour is consistent with the observations of Yaguchi and Takahashi [2005a] for a P91 material at the same stress rate.
The ratchetting strain evolution can be partitioned into three stages: the primary, secondary and tertiary stages of deformation [Zhao and Xuan, 2011]. The primary stage exhibits a continuously decreasing ratchetting strain rate. The minimum ratchetting strain rate signifies the beginning of the steady-state secondary stage, where deformation progresses at a constant rate. The tertiary stage is defined by a continuous increase in ratchetting strain rate, where ratchetting deformation accelerates until final failure. In the case of the tests shown in Figure 3.31, the primary stage ratchetting occurs within the first six cycles for both tests, the 50 MPa/s test exhibits a 35 % longer secondary stage period compared with that of the 5 MPa/s test. It was also observed that the 5 MPa/s test had a secondary stage strain rate of 0.15 %/s, compared with a lower secondary strain rate of 0.012 %/s for the 50 MPa/s test.
MPa/s test. However, it is in the tertiary stage that the effect of stress rate is most evident.

Figure 3.32 shows the progress of the stress-strain response for the cyclic stress-controlled test (for selected cycles) at 550 °C and $R = -0.5$ with a strain rate of 50 MPa/s. The stress-strain loops of Figure 3.32 shows a change in the size and shape in each progressive loop with increasing cycle number. Figure 3.33 depicts the stress-strain evolution of a 550 °C $R=-0.5$ test at a stress rate of 5 MPa/s. Comparisons between Figure 3.32 and Figure 3.33 show the increased rate of strain after the 100th cycle. The evolution of inelastic strain accumulation is depicted in Figure 3.34. This figure shows that the inelastic strain accumulation is effectively the same up to about 90 cycles, after which there is a marked increase in the inelastic strain range in the 5 MPa/s test.

Figure 3.32. Stress-strain loops for selected cycles for 550 °C, for a controlled stress rate of 50 MPa/s and $R = -0.5$, for SA P91 base material.
Figure 3.33. Stress-strain loops for selected cycles for 550 °C, for a controlled stress rate of 5 MPa/s and R = -0.5, for SA P91 base material.

Figure 3.34. Evolution of inelastic strain accumulation at 550 °C and R = -0.5, for two different stress rates.
3.5 HTLCF of welded specimens

3.5.1 General
Testing of the weld material (WM) and cross-weld (CW) specimens were carried out for 400 °C and 500 °C, at a constant strain rate of 0.033 %/s. The overall objective of the cross-weld tests is to gain an understanding of the HAZ material in terms of stress-strain behaviour under HTLCF conditions. Tension-hold tests were carried out on the WM and CW specimens at a constant strain rate of 0.1 %/s, which allows for direct comparison of these tests with the BM tension-hold tests.

The HTLCF testing of the heat affected zone requires the specimen gage section to consist of weld, HAZ and base material, with the interface between the HAZ and the base material perpendicular to the specimen axis.

3.5.2 HTCLF behaviour
Figures 3.35 and 3.36 show the measured stress-strain loops for the first cycle for each of the WM and CW test specimens at 400 °C. Figure 3.37 shows the cyclic softening evolution with number of cycles for three different strain-ranges at 400 °C for the WM. Figure 3.38 shows the measured response of the CW specimen at 500 °C at $N = 1$, and Figure 3.39 shows the cyclic softening behaviour of the CW tests at 500 °C, for three different strain ranges. Some slight initial hardening is observed over the first 10 cycles followed by continuous softening. The 0.8 % and 1 % tests display similar softening behaviour over the first 100 cycles.

A comparison of the WM and CW first cycle responses at a strain range of 0.8 % at 400 °C with the corresponding BM results is shown in Figure 3.40a. It is clear that the WM response is significantly harder than that of the BM and CW tests. The CW tests are similar to the BM tests in terms of stress range, although the CW tests exhibit slightly larger inelastic strain range. Figures 3.40b, 3.40c and 3.40d show
corresponding comparisons for 0.6% and 500 °C, in all cases showing the same trend in terms of comparative WM behaviour.

Figures 3.41 and 3.42 show the stress versus plastic strain amplitude data extracted at half-life ($N_f/2$, where $N_f$ is defined in terms of 30 % load drop from the 150th cycle) from the test results for the WM and CW test specimens, for 400 °C and 500 °C, in comparison with the BM data. This highlights the cyclically harder response of the WM relative to the CW and BM specimens, particularly at 400 °C, at which the CW gives a response very similar to the BM; this, in turn, indicates that the BM dominates the response of these test specimens. At 500 °C the WM exhibits the hardest response; however, in this case, the CW tests show a significantly harder response than the BM tests, suggesting a more significant influence of the WM on the CW response at the higher temperature.

Equation 3.1 is used to represent the cyclic stress-plastic strain relationship for the different material tests. The temperature-dependent $K'$ and $n'$ values were extracted for the half-life, $N_f/2$, (where $N_f$ corresponds to 30 % load drop) and are presented in Table 3.10. It was found that the values of both coefficient $K'$ and exponent, $n'$, for the WM and CW tests across all temperatures were lower than for the BM material.

<table>
<thead>
<tr>
<th>$T$ (°C)</th>
<th>BM</th>
<th>WM</th>
<th>CW</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$K'$</td>
<td>$n'$</td>
<td>$K'$</td>
</tr>
<tr>
<td>400</td>
<td>639.9</td>
<td>0.083</td>
<td>836.5</td>
</tr>
<tr>
<td>500</td>
<td>467.3</td>
<td>0.068</td>
<td>589.3</td>
</tr>
</tbody>
</table>

Table 3.10. Cyclic material constants for BM, WM, and HAZ for 400 °C and 500 °C.
Figure 3.43 shows the measured stress relaxation behaviour of the three different types of specimen from the first quarter cycle of a creep-fatigue tensile hold test at a temperature of 500 °C. The peak tensile strain was held constant at 0.5 % for a period of 120 s.

Figure 3.35. Measured first cycle (N = 1) WM stress-strain response at 400 °C.
Figure 3.36. Measured first cycle \((N = 1)\) CW stress-strain response at 400 °C.

Figure 3.37. WM cyclic softening behaviour at 400 °C.
Figure 3.38. CW stress-strain response at $N = 1$ at a temperature of 500 °C for various strain ranges, at a constant strain rate of 0.033 %/s.

Figure 3.39. CW cyclic softening behaviour at 500 °C.
Stress (MPa) vs. Strain (%) at 400 °C

(a) Comparison of WM, BM, and CW at 400 °C.

(b) 400 °C plots showing stress and strain for WM, BM, and CW.
Figure 3.40. Comparison between BM, WM, and CW tests for various strain ranges at a constant strain rate of 0.033 %/s for 400 °C (Figure 3.40 (a to b)) and 500 °C (Figure 3.40 (c to d)).
Figure 3.41. Cyclic stress-plastic strain curves for BM, WM and CW tests at 400 °C at half-life.

Figure 3.42. Cyclic stress-plastic strain curves for BM, WM and CW tests at 500 °C at half-life.
3.5.3 Identification of the heat affected zone (HAZ)

The HAZ has been found to be comprised of narrow sub-regions of coarse and fine-grained microstructures. Numerous researchers have reported that these sub-regions have different creep strengths [Tabuchi et al., 2009; Watanabe et al., 2006; Kimura et al., 2011]. Samples for optical microscopy were extracted from tested specimens which were etched using Vilella’s reagent (5 ml HCl + 100 ml H₂O + 1 g Picric acid). Figure 3.44 shows cracking in the HAZ of the CW test specimens; cracking was also observed in the BM of the CW specimens.

Figure 3.43. Stress relaxation data for different specimen types.
Figure 3.44. Optical micrograph of (a) primary crack (occurring in the HAZ) and (b) magnified view of primary crack along with secondary (surface) cracking for cross-weld specimen tested at 1 % strain range at 500 °C.

Figure 3.45 shows a Vickers hardness profile measured across the sectioned CW test specimen, to help identify the HAZ dimensions. This also corroborates the higher hardness of the WM region, as demonstrated in the measured cyclic responses.

Figure 3.46 (a to d) shows micrographs of the different material zones annotated in Figure 3.45. Figure 3.46a and b show the grain structure of the WM and BM, respectively. The figures show that the grain sizes of both materials are similar. Figure 3.46c shows the fusion boundary between the WM and the fine-grain HAZ (FG-HAZ) and Figure 3.46d shows the coarse-grain HAZ (CG-HAZ), intercritical HAZ (IC-HAZ) and the BM regions on the CW specimen of Figure 3.45. The identification of these regions allowed estimation of the HAZ width and location. The HAZ cracks are seen to be consistent with Type IV cracking, i.e. located in the IC-HAZ region.
Figure 3.45. Hardness traverse across (post-test) CW specimen.
Figure 3.46. Micrographs taken at different locations across the CW specimen of Figure 3.45, showing (a) WM microstructure, (b) BM microstructure, (c) WM - fine grain HAZ fusion boundary and (d) coarse grain HAZ, inter-critical HAZ and BM boundaries.

Table 3.11 presents the number of cycles to failure of the WM and CW tests. Figure 3.47 (a and b) shows the strain-life behaviour of the BM, WM and CW tests. At both temperatures the BM has the largest lives. At 400 °C, the WM tests have the shortest fatigue lives. At 500 °C the WM and CW tests have the same lives at higher strain ranges and the CW is slightly lower at the lowest strain range.
Table 3.11. Failure lives of WM and CW tests

<table>
<thead>
<tr>
<th>Test No.</th>
<th>Test type</th>
<th>Temp (°C)</th>
<th>Strain Rate (%/s)</th>
<th>Strain Range (%)</th>
<th>Nf</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Cross-weld</td>
<td>400</td>
<td>0.033</td>
<td>0.6</td>
<td>5013</td>
</tr>
<tr>
<td>2</td>
<td>Cross-weld</td>
<td>400</td>
<td>0.033</td>
<td>0.8</td>
<td>2772</td>
</tr>
<tr>
<td>3</td>
<td>Cross-weld</td>
<td>400</td>
<td>0.033</td>
<td>1.0</td>
<td>-</td>
</tr>
<tr>
<td>4</td>
<td>Cross-weld</td>
<td>400</td>
<td>Relaxation</td>
<td>1.0</td>
<td>1411</td>
</tr>
<tr>
<td>5</td>
<td>All-weld</td>
<td>400</td>
<td>0.033</td>
<td>0.6</td>
<td>2730</td>
</tr>
<tr>
<td>6</td>
<td>All-weld</td>
<td>400</td>
<td>0.033</td>
<td>0.8</td>
<td>-</td>
</tr>
<tr>
<td>7</td>
<td>All-weld</td>
<td>400</td>
<td>0.033</td>
<td>1.0</td>
<td>1382</td>
</tr>
<tr>
<td>8</td>
<td>All-weld</td>
<td>400</td>
<td>Relaxation</td>
<td>1.0</td>
<td>-</td>
</tr>
<tr>
<td>9</td>
<td>All-weld</td>
<td>500</td>
<td>0.033</td>
<td>0.6</td>
<td>2524</td>
</tr>
<tr>
<td>10</td>
<td>Cross-weld</td>
<td>500</td>
<td>0.033</td>
<td>0.8</td>
<td>993</td>
</tr>
<tr>
<td>11</td>
<td>Cross-weld</td>
<td>500</td>
<td>0.033</td>
<td>1.0</td>
<td>1084</td>
</tr>
<tr>
<td>12</td>
<td>Cross-weld</td>
<td>500</td>
<td>Relaxation</td>
<td>1.0</td>
<td>1777</td>
</tr>
<tr>
<td>13</td>
<td>All-weld</td>
<td>500</td>
<td>0.033</td>
<td>0.6</td>
<td>3806</td>
</tr>
<tr>
<td>14</td>
<td>All-weld</td>
<td>500</td>
<td>0.033</td>
<td>0.8</td>
<td>1152</td>
</tr>
<tr>
<td>15</td>
<td>All-weld</td>
<td>500</td>
<td>0.033</td>
<td>1.0</td>
<td>1031</td>
</tr>
<tr>
<td>16</td>
<td>All-weld</td>
<td>500</td>
<td>Relaxation</td>
<td>1.0</td>
<td>-</td>
</tr>
</tbody>
</table>

Figure 3.47. Fatigue life comparison between BM, WM and CW tests at (a) 400 °C, and (b) 500 °C.

Figure 3.48 shows the effect of inelastic strain amplitude on fatigue life. The BM tests give the longest lives at both temperatures. At 400 °C (Figure 3.48a) the WM tests give the shortest lives (significantly). Again, at 500 °C, there is negligible
difference between the WM and CW test lives. The Coffin-Manson constants identified from Figure 3.48 (a and b) are presented in Table 3.12. The failure data for the Ostergren life prediction model is shown in Figure 3.49 and the identified failure constants are given in Table 3.13.

Figure 3.48. Coffin-Manson data for the welded and base material tests at (a) 400 °C and (b) 500 °C.

Figure 3.49. Failure data for WM and CW tests, based on the Ostergren failure parameter.
Table 3.12. Coffin-Manson constants for WM and CW tests, at 400 °C and 500 °C

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>εc'</th>
<th>c</th>
<th>εc'</th>
<th>c</th>
</tr>
</thead>
<tbody>
<tr>
<td>400</td>
<td>9539.9</td>
<td>-1.92</td>
<td>23.81</td>
<td>-1.092</td>
</tr>
<tr>
<td>500</td>
<td>2.7989</td>
<td>-0.914</td>
<td>0.6276</td>
<td>-0.72</td>
</tr>
</tbody>
</table>

Table 3.13. Ostergren constants for WM and CW tests, at 400 °C and 500 °C

<table>
<thead>
<tr>
<th>WM</th>
<th>CW</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>β</td>
</tr>
<tr>
<td>2235.4</td>
<td>-0.9363</td>
</tr>
</tbody>
</table>

The accuracy of the Coffin-Manson and Ostergren fatigue life prediction models is shown in Figure 3.50 and Figure 3.51, respectively. The predicted fits are clearly very good, based on the limited data generated.

Figure 3.50. Coffin-Manson predicted lives versus observed lives for CW and WM tests at 400 °C and 500 °C.
Figure 3.51. Predicted fatigue lives versus observed fatigue lives, using the Ostergren model for CW and WM tests at 400 °C and 500 °C.

3.6 Conclusions
This chapter has characterised the high temperature low cycle fatigue behaviour of a service-aged P91 base material from a fossil fuel power station, along with that of a P91 weld and cross-weld material. Comparisons with P91 material data from the literature are also presented. For the conditions examined, the key conclusions are as follows:

A programme of HTLCF and TMF tests have been carried out here on service-aged P91 base material. The majority of these tests were conducted in collaboration with the University of Nottingham at Nottingham, through funding provided by ESB Energy International and the SFI METCAM project. In parallel, a high temperature, strain-controlled low cycle fatigue test capability has been developed, commissioned and successfully validated in Mechanical Engineering at NUI Galway.
In addition, a weld-repair header connection has been commissioned in collaboration with ESB Energy International, to facilitate manufacture of weld material and cross-weld test specimens for low-cycle fatigue testing. Furthermore ratchetting (stress-controlled) tests have been conducted on the service-aged P91 base material.

- The service-aged material demonstrates similar cyclic softening trends to those observed for P91 material from the literature.
- The fatigue life evaluation using the Ostergren fatigue model suggests that the observed fatigue life of the service-aged material is similar to that of P91 material from the literature.
- At lower strain ranges, temperature effects on fatigue life are more pronounced with increasing temperature giving reduced life.
- TMF tests indicate that the effect of positive mean stress is detrimental to fatigue life.
- Ratchetting behaviour is heavily dependent on temperature and ratchet strain rate is sensitive to stress rate.
- The weld material is significantly cyclically harder than the BM.
- The fatigue lives of the WM and CW specimens are significantly lower than those of the BM tests.
- Type IV (IC-HAZ) cracking was observed in the tested cross-weld specimens
Chapter 4

THERMOMECHANICAL ANALYSIS OF A PRESSURISED PIPE UNDER PLANT CONDITIONS

4.1. Introduction

This chapter is concerned with the development of a methodology for thermomechanical analysis of high temperature, steam-pressurised P91 pipes in electrical power generation plant under realistic (measured) temperature and pressure cycles. In particular, these data encompass key thermal events, such as ‘load-following’ temperature variations and sudden, significant fluctuations in steam temperatures associated with attemperation events and ‘trips’ (sudden plant shutdown), likely to induce thermomechanical fatigue damage.

An anisothermal elastic-plastic-creep material model for cyclic behaviour of P91 is employed in the transient FE model to predict the stress-strain-temperature cycles and the associated strain-rates. The results permit characterisation of the behaviour of pressurised P91 pipes for identification of the thermomechanical loading histories relevant to such components, for realistic, customised testing. This type of capability is relevant to design and analysis with respect to the evolving nature of power plant operating cycles, e.g. associated with more flexible operation of fossil fuel plant. Different realistic loading histories for a high temperature, steam-pressurised P91 pipe, are investigated here, to predict the thermomechanical cyclic behaviour of P91 pipe. An anisothermal cyclic, elastic-plastic-creep material model is
developed, based on published experimental and material modelling data from the literature.

4.2. Methodology

4.2.1. Temperature and Loading Histories

Plant thermal cycles are complex due to the attemperation system, where cooling water is injected directly into the steam flow, to control the plant operating temperature. In order to investigate the envelope of possible thermomechanical conditions that might occur in a power plant pipe, three different plant cycles are considered here. The first, a simplified cycle (referred to hereafter as ‘Simplified cycle’), depicted in Figure 4.1, is intended to represent the steam temperature and assumed in-phase pressure. The simplified cycle is intended to schematise an actual cycle. The first part of this cycle, referred to hereafter as the major cycle, simulates a severe (cold) start-up to 500 °C and 17 MPa, followed by a short period of plant operation characterised by a hold period at 500 °C and 17 MPa and then a cool-down or ‘warm’ shut-down to 200 °C and 10 MPa. The minor cycle simulates a resumption to operating conditions, which again, is sustained for a short period, before cool down. There are three repeated minor cycles, corresponding to ‘warm’ start-up, hold at maximum temperature and pressure, followed by ‘warm’ shut-down. The steam header is housed in an enclosure which is assumed to heat to 325 °C after initial start-up, as shown in Figure 4.1. As described below, a transient heat transfer analysis is developed to predict the associated temperature distribution across the wall thickness of a pipe, for subsequent thermomechanical analysis, as also described below.
Figure 4.1. Assumed steam, pressure and enclosure air temperature histories for ‘simplified cycle’.

The second and third plant cycles are based on plant temperature and pressure data, as measured by ESB Energy International, for superheater steam headers from two different plants, with the same header geometry.

Figure 4.2 depicts the second plant cycle, corresponding to a cold-start, hereafter referred to as the ‘Cold start’ cycle. In this case, the external pipe temperature was measured using a thermocouple attached to the outside surface of the header and the steam temperature was inferred from a thermocouple attached to a stub pipe. In this case, it can be seen that the inside temperature (i) initially increases from ambient temperature, in a much less idealised way than the simplified cycle, over the first 14 hrs or so, to an intermediate state with a temperature which fluctuates about a mean value of approximately 300 °C, (ii) then increases more gradually, over the next 6 hrs, to a steady temperature of about 490 °C, (iii) stays approximately at this temperature for about 8 hrs and (iv) finally cools at a rate of 90 °C/hr to 50 °C within about 5 hrs. The external pipe temperature generally lags the steam temperature, with temperature differentials of up to 150 °C, e.g. during heat-up
after about six hours. One exception is between about 5.5 hrs and 14 hrs, when quite sudden reductions in steam temperature occur, leading to sudden temperature gradients across the wall thickness, with the outer surface in some cases being hotter by up to 100 °C. At the highest steady temperature condition, the outside is only about 10 °C cooler than the steam.

![Figure 4.2. Steam, pressure and enclosure air temperature histories for ‘cold start cycle’.](image)

Figure 4.3 shows the measured data for the third plant cycle to simulate a ‘load-following’ case, including a ‘trip’ and a subsequent restart. In this case, there is no data on external surface temperature.
Figure 4.3. Steam, pressure and enclosure air temperature histories for ‘load-following cycle’ (it is assumed that the enclosure air temperature is constant at 430 °C).

4.2.2. Thermomechanical pipe model

A key objective of the present chapter is the development of a methodology for predicting the thermomechanical temperature-stress-strain response of steam-pressurised power plant piping components under realistic pressure-temperature histories. This requires the development of (i) a transient heat transfer model which can predict the thermal gradients in such piping components, e.g. plain pipes, welded pipes, welded branched connections, and (ii) a sequential anisothermal cyclic, viscoplasticity mechanical model for the pipe material(s). Figure 4.4 shows the key steps taken in the development of a thermomechanical analysis of a plain pipe under realistic loading conditions.
Figure 4.4. Flowchart showing the key steps in the thermomechanical analysis of a plain pipe.

Attention is focussed on a plain pipe under closed-end conditions, with dimensions corresponding to the geometry of the fossil fuel burning power station superheater header.

Figure 4.5 shows a schematic of the geometry employed here. The inside diameter is 251.9 mm and the wall thickness is 36 mm.

Figure 4.5. Dimensions and thermomechanical loading conditions of P91 pipe.
The methodology developed here employs the commercial FE code ABAQUS, to predict the temperature-stress-strain response of the pipe subjected to the temperature-pressure histories defined above. An internal steam pressure is applied to the inside surface of the pipe \( p_i(t) \), along with an axial load \( F_{ax}(t) \) to simulate the closed end condition. The heat transfer problem is essentially a one-dimensional transient problem. However, the subsequent anisothermal, cyclic viscoplasticity analyses is less straightforward, being a non-linear axisymmetric problem.

### 4.2.3. Heat Transfer modelling

The thermal model employed for all three plant cycles is described as follows:

1. a time-dependent steam temperature \( T_i(t) \) is specified, based on measured data shown in Figures 4.1 to 4.3;
2. a time-dependent enclosure temperature \( T_\infty(t) \) is specified based on an assumed profile e.g. see Figs. 4.1 and 4.2, informed by industrial experience [ESB Energy International, 2011];
3. forced steam convection is modelled on the pipe inside surface;
4. transient heat transfer, with temperature-dependent conductivity and specific heat data, is modelled through the pipe wall;
5. natural convection is modelled on the outside surface of the pipe with the enclosure free-stream temperature, \( T_\infty(t) \).

The identification of conductivity and specific heat properties for the P91 material is relatively straight-forward and this data is presented in Table 4.1, as obtained from [ASME B31.1, 1995].
Table 4.1. Temperature dependent heat transfer properties

<table>
<thead>
<tr>
<th>Property</th>
<th>20 °C</th>
<th>200 °C</th>
<th>400 °C</th>
<th>600 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density (kg/m$^3$)</td>
<td>7860</td>
<td>7860</td>
<td>7860</td>
<td>7860</td>
</tr>
<tr>
<td>Conductivity (W/mK)</td>
<td>22.5</td>
<td>26.2</td>
<td>27.7</td>
<td>27.7</td>
</tr>
<tr>
<td>Specific heat capacity (J/kg K)</td>
<td>410</td>
<td>510</td>
<td>600</td>
<td>770</td>
</tr>
</tbody>
</table>

Similarly, the identification of steam-side convection coefficient is relatively straight-forward, based on the following equation

$$ Nu = \frac{hd}{k} \quad (4.1) $$

where $h$ is the steam-side forced convection coefficient, $k$ is thermal conductivity and $Nu$ is the Nusselt number, defined by the following empirical correlation for a smooth pipe with $0.5 < Pr < 1.5$ and $10^4 < Re < 5 \times 10^6$, taken from Holman [2002].

$$ Nu = 0.0214(Re^{0.8} - 100)Pr^{0.4} \quad (4.2) $$

with the Prandtl number, $Pr$, being the ratio of the viscous diffusivity to thermal diffusivity and the Reynolds number, $Re$, calculated for an assumed steam velocity of 15 m/s (recommended velocity to prevent significant pressure drop in long steam pipe lengths [Spirax Sarco, 2010]), as follows:

$$ Re = \frac{\rho ud}{\mu} \quad (4.3) $$
where $\rho$ is steam density, $u$ is steam velocity, $d$ is the internal pipe diameter and $\mu$ is the kinematic viscosity. The resulting $h_i$ values for the different plant cycles are tabulated in Table 4.2.

Table 4.2 Identified convection heat transfer coefficients.

<table>
<thead>
<tr>
<th>Plant cycle</th>
<th>$h_i$ (W/m$^2$.K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Simplified</td>
<td>3611</td>
</tr>
<tr>
<td>Cold-Start</td>
<td>3611</td>
</tr>
<tr>
<td>Load-following</td>
<td>2340</td>
</tr>
</tbody>
</table>

The identification of enclosure-side (outer pipe surface) convection coefficient, $h_o$, is not so straightforward. An inverse method was employed here based on trial and error estimation using correlation of predicted and measured outer pipe surface temperature time history as the criterion for identification. The identification process was applied to determine the value of $h_o$ that gave best correlation of the predicted outer surface temperature with the measured history for the ‘cold start cycle’ in Figure 4.2. The resulting value of $h_o = 1800$ W/m$^2$.K is used for the three plant cycle analyses.

**4.2.4. Cyclic Viscoplasticity Model**

Saad et al. [2010] have presented experimental isothermal and out-of-phase (OP) and in-phase (IP) TMF test data for P91, in the temperature range 400 °C to 600 °C, for a strain-range of 1% and at a strain-rate of $1\times10^{-3}$ s$^{-1}$. Saad et al. [2010] also identified and successfully implemented, for uniaxial conditions, a (Chaboche) unified cyclic viscoplasticity model for P91, based on the isothermal test data. In the present work, the two-layer viscoplasticity model [Kichenin and Dang Van, 1996] is implemented for the P91 material, under thermomechanical conditions. An alternative model using
the unified Chaboche model incorporating a Ohno and kinematic hardening model is also investigated and is presented in Appendix B.

In this analysis, a temperature-dependent version of the two-layer model is developed along with a combined non-linear kinematic and isotropic hardening cyclic plasticity model for the time-independent behaviour and Norton's power law to capture secondary creep.

The general equations for the two-layer viscoplasticity model are given below:

\[
\varepsilon^e_v = \frac{1 + \nu}{K_v} \sigma_v - \frac{\nu}{K_v} tr(\sigma_v)I \quad (4.4)
\]

\[
\varepsilon^e_p = \frac{1 + \nu}{K_p} \sigma_p - \frac{\nu}{K_p} tr(\sigma_p)I \quad (4.5)
\]

where \(K_v\) and \(K_p\) are the elastic constants of the elastic-viscous and elastic-plastic networks respectively, \(\nu\) is Poisson's ratio, \(I\) is unit tensor of second-rank, \(tr\) expresses the trace of a tensor, and \(\sigma_v\) and \(\sigma_p\) are the stress tensors for the elastic-viscous and elastic-plastic networks respectively, defined as:

\[
\sigma_v = K_v : (\varepsilon - \varepsilon_v) \quad (4.6)
\]

\[
\sigma_p = K_p : (\varepsilon - \varepsilon_p) \quad (4.7)
\]

\[
\sigma = \sigma_p + \sigma_v \quad (4.8)
\]
where $K_v$ and $K_p$ which have been previously defined in Chapter 2 as the elastic tensors of the elastic-viscous and elastic-plastic networks respectively. The yield surface is defined by:

$$f(\sigma_p - \alpha) = \sigma^0$$

(4.9)

where $\sigma^0$ is the current size of the yield surface and $f(\sigma_p - \alpha)$ is the equivalent (von Mises) stress, with the back stress $\alpha$, defined as:

$$f(\sigma_p - \alpha) = \sqrt{\frac{3}{2}(S_p - \alpha^{dev}) : (S_p - \alpha^{dev})}$$

(4.10)

where $\alpha^{dev}$ is the deviatoric part of the back stress and $S_p$ is the deviatoric (plastic) stress tensor. The Ziegler kinematic hardening law is used to simulate the translation of the yield surface in the stress space through the back stress, $\alpha$ [Ziegler, 1959]. The plastic flow rule is:

$$\dot{\epsilon}_p = \frac{\partial f(\sigma_p - \alpha)}{\partial \sigma_p} \dot{p}$$

(4.11)

where $\dot{p}$ is the equivalent plastic strain rate defined as follows:

$$\dot{p} = \sqrt{\frac{2}{3} \dot{\epsilon}_p : \dot{\epsilon}_p}$$

(4.12)
The cyclic hardening/softening behaviour of the material is captured using an exponential law where the size of the yield surface \( \sigma^0 \) can be defined as a function of equivalent plastic strain, \( p \):

\[
\sigma^0 = k + Q_\infty (1 - \exp(-bp))
\]

(4.13)

Here, \( k \) is the size of the yield surface at zero plastic strain, and \( Q_\infty \) and \( b \) are material constants identified from cyclic stress-strain loops, where \( Q \) is the asymptotic value of the stabilised loop and \( b \) controls the speed of stabilisation.

The evolution of the kinematic hardening component, is defined according to the generalized Ziegler rule for the anisothermal case, has been defined in Chapter 2, Equation 2.19. The steady state (secondary) creep behaviour is modelled using the Norton power law, given by the following:

\[
\dot{\varepsilon}_v = \frac{3}{2} A [f(\sigma_v)]^n \frac{S_v}{f(\sigma_v)}
\]

(4.14)

where \( A \) and \( n \) are temperature-dependent material constants, \( \dot{\varepsilon}_v \) is the viscous creep strain rate, and \( S_v \) is the deviatoric stress tensor.

As implemented in ABAQUS, a user-defined, temperature-dependent, parameter \( F \) is introduced to define the relative magnitude of \( K_v \), the elastic modulus of the elastic-viscous network, and \( K_p + K_v \), the total instantaneous elastic modulus (assumed equal to \( E \), Young's modulus), this has been discussed in detail in Chapter 2, Equation 2.21.

The coefficients \( c_i \) and \( \gamma_i \) \( (i = 1, 2) \) corresponding to short strain-range and long strain-range (Frederick-Armstrong) non-linear kinematic hardening (NLKH)
constants, are identified from the following integrated version of the non-linear
kinematic hardening equation, see Lemaitre and Chaboche [1990]:

$$\frac{\Delta \sigma}{2} - k = \frac{c}{\gamma'} \tanh \left( \gamma' \frac{\Delta \varepsilon_p}{2} \right)$$  \hspace{1cm} (4.15)

A critical issue in the development of cyclic viscoplasticity models for
anisothermal conditions, e.g. TMF, is the identification of the large number of
constants required. Figures 4.6(a) and 4.6(b) depicts the correlation between the
adjusted monotonic data of Okrajni et al., [2008], and the material model of Equation
4.15, for 20 °C and 223 °C, respectively, where yield strength is subtracted from the
curve in order to obtain the plastic behaviour and the material model is then fitted by
varying the $c/\gamma'$ ratio, from which the constants $c$ and $\gamma'$. Figures 4.7(a) and 4.7(b)
depicts the hysteresis loops for $N=1$ for 20 °C and 223 °C, respectively. For the
subsequent multiaxial modelling analysis, the plasticity constants $c$ and $\gamma'$ for
temperature below 400 °C were determined using the published monotonic data
published by Okrajni et al., [2008].

Figure 4.8 presents uniaxial finite element results compared with test data from
Saad et al. [2010]. For the temperature range 400 °C to 600 °C, the constants $k, C_i, \gamma_i,
Q, b, A$ and $n$ for the two-layer model have been derived directly (algebraically) from
the corresponding unified (Chaboche) model constants presented by Saad et al.
[2010]. The elastic constants have been obtained from ASME B31.1, [1995]. Then,
the remaining two-layer constant, $f$, has been identified by comparing the predicted
two-layer stress-strain loops to the measured and predicted loops of Saad et al.,
[2010], for each temperature. Table 4.3 shows the identified P91 material constants
for the two-layer model.
Figure 4.6. Identification of coefficients $c$ and $\gamma$ for P91 material from monotonic data published by Okrajni et al., [2008], for (a) 20 °C, and (b) 223 °C.

Figure 4.7. Predicted hysteresis loops for (a) 20 °C, and (b) 223 °C, using coefficients determined from the monotonic experimental test data published by Okrajni et al., [2008].
Figure 4.8. Uniaxial hysteresis loops from two-layer viscoplasticity model compared against test data from Saad et al., [2010]. For the initial cycle (N=1) Fig. 4.7 (a), (c) and (e), and the corresponding test after completion of 100 cycles (N=100), labelled Fig. 4.7 (b), (d) and (f).
Table 4.3. Identified temperature-dependent NLKH and viscoplastic constants.

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>k (MPa)</th>
<th>E (MPa)</th>
<th>q (MPa)</th>
<th>b (MPa·hr⁻¹)</th>
<th>A (MPa·hr⁻¹)</th>
<th>n (MPa)</th>
<th>C₁ (MPa)</th>
<th>γ₁</th>
<th>C₂ (MPa)</th>
<th>γ₂</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>166</td>
<td>213000</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>374999</td>
<td>790</td>
<td>0</td>
<td>0</td>
<td></td>
</tr>
<tr>
<td>223</td>
<td>159</td>
<td>198791</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>363500</td>
<td>791</td>
<td>0</td>
<td>0</td>
<td></td>
</tr>
<tr>
<td>400</td>
<td>95</td>
<td>183938</td>
<td>-55</td>
<td>0.45</td>
<td>0.000134</td>
<td>2.25</td>
<td>352500</td>
<td>2350</td>
<td>48600</td>
<td>405</td>
</tr>
<tr>
<td>500</td>
<td>90</td>
<td>165882</td>
<td>-60</td>
<td>0.6</td>
<td>1.62E-05</td>
<td>2.54</td>
<td>215872</td>
<td>2191</td>
<td>48235</td>
<td>460.7</td>
</tr>
<tr>
<td>600</td>
<td>85</td>
<td>140700</td>
<td>-75.4</td>
<td>1</td>
<td>6.31E-06</td>
<td>2.7</td>
<td>106860</td>
<td>2055</td>
<td>31159</td>
<td>463</td>
</tr>
</tbody>
</table>

4.3. Results

4.3.1. Material model

Figures 4.8 (a, c and e) show the predicted isothermal stress-strain hysteresis initial
(N = 1) and stabilised (N = 100) loops for 1% strain range at a strain rate of 0.1% s⁻¹
and for a range of temperatures from 400 °C to 600 °C. These loops give good
correlation with the predicted and measured data of Saad et al. [2010] for T ≥ 400 °C.
The figures show the monotonic tensile curve, as well as the first cycle (N = 1) and
softening behaviour of the material after 100 cycles. Regarding the initial hardening
between the first quarter cycle and the end of the first cycle, this is simply the result
of shakedown of the kinematic hardening model, which takes about one cycle to
shake down to a stable loop and which pre-dominates over the isotropic softening
within the first cycle. This occurrence is also exhibited in the simulated loops
generated by Koo et al. [2011] under similar loading conditions. It is also noted that
this effect is more pronounced at lower temperatures. Figure 4.9 shows the predicted
effect of strain-rate on the stress strain response at 600 °C, showing significant
reduction in stabilised stress range from Δσ = 637 MPa at a strain rate of 1×10⁻³ s⁻¹ to
Δσ = 498 MPa at a strain rate of 1×10⁻⁴ s⁻¹. Figure 4.9 reflects the ability of the two
layer model to capture material strain rate sensitivity based on identification of the viscoplasticity constants from stress relaxation tests, an approach validated by Deshpande et al. [2010].

Figure 4.9. Predicted effect of strain rate on isothermal uniaxial hysteresis loops.

4.3.2. Simplified cycle

Figure 4.10 shows the predicted thermal histories at the inside and outer surfaces of the pipe. Points A to H are picked in order to cross-reference the predicted thermal response history with the stresses and strains. The average temperature difference between the inside and outside pipe surfaces is approximately 54 °C. The inside surface of the pipe is on average 25 °C colder than the bulk steam temperature.
From the model, the predicted critical location within the pipe for stresses and strains is the inner pipe surface and the dominant stress component is hoop stress. The hoop stress-strain response at the inner pipe surface is shown in Figure 4.11. A maximum compressive stress of -130 MPa is predicted at the inner surface during the first heating cycle (point B), due to the surrounding cooler material restricting the (hotter) inner pipe surface from expansion. When the inner pipe surface has been exposed to a continuous steam temperature of 500 °C, the hoop stress settles to 32 MPa (points D and E). Cooling of the steam, and hence inside surface of the pipe, induces a severe tensile stress (point F) of 170 MPa at the inside surface, due to the surrounding (hotter) material preventing contraction of the inside surface material. The stress range for the major cycle is 303 MPa with a mean tensile stress of 22 MPa. The minor cycle produces a stress range of 247 MPa and a mean tensile stress of 47 MPa, and during prolonged exposure to the elevated steam temperature the hoop stress settles to -43 MPa at point H.
Figure 4.11. Predicted hoop stress-strain response at inner pipe surface for 'Simplified cycle' of Figure 4.1.

Figure 4.12 illustrates the predicted transient hoop stress response. It is clear that the critical variations in stress are associated with (i) heat-up (e.g. points A to C), (ii) cool down (e.g. points C to F), and (iii) subsequent heat-up and cool-down phases. Two important observations here are (i) that cool-down is more detrimental than heat-up and (ii) heat-up from ambient (point B) is more detrimental than from 200 °C (point G). Figure 4.13 shows the predicted creep and plastic (hoop) strains during the thermomechanical loading. Clearly, both compressive creep and plastic strain are predicted during the initial heat-up from ambient, followed by creep relaxation (from C to E). During cool-down, it is clear that relatively severe tensile plastic and creep hoop strain is predicted (point F), leading to an inelastic hysteresis cycle (Figure 4.11), followed by sudden compressive creep strain due to heat-up from 200 °C (F to G). Overall, the creep strain is predicted to be pre-dominant, with
predicted inelastic strain ranges of 0.038 % for the major cycle and 0.015 % for the minor cycle.

Figure 4.12. Predicted hoop stress response at the inner pipe surface for 'Simplified cycle' of Figure 4.1.

Figure 4.13. Predicted evolution of creep and plastic hoop strains for 'Simplified cycle' of Figure 4.1.
4.3.3. Cold start cycle(s)

Figure 4.14 shows the predicted thermal histories at the inside and outer surfaces of the pipe for the cold start cycle of Figure 4.2. The average temperature difference between the inside and outside pipe surfaces is approximately 20 °C. The inside surface of the pipe is on average 15 °C colder than the bulk steam temperature, with a maximum temperature difference of 113 °C after 7.2 hours. It is clear that, although the inside pipe surface is generally hotter, at certain times during the attemperation process, the outside surface is up to 80 °C or 90 °C hotter.

![Figure 4.14](image)

**Figure 4.14.** Predicted FE thermal histories for inside and outside pipe surface for N=1, for 'Cold start cycle' of Figure 4.2.

Figures 4.15 and 4.16 show the predicted hoop stress-strain response for the cold start cycle for \( N = 1 \) (i.e for one such cycle) and the corresponding predicted evolutions of creep and plastic hoop strains. In general, the predicted pattern is similar to that of the simplified cycle of Figure 4.11. The maximum predicted compressive hoop stress is slightly greater (-154 MPa, as opposed to -130 MPa for the simplified cycle). Again, slightly higher peak tensile stresses are predicted for the attemperation process temperature fluctuations than for the simplified cycle. Figure 4.15 shows peak tensile hoop stresses of about 175 MPa at point G in the cycle and
195 MPa at point H in the cycle, despite the apparently significantly smaller temperature gradient in this cycle. Another important point to note from Figure 4.15 is the larger number of smaller elastic cycles, many of which have tensile mean (hoop) stresses at the inside surface. A larger stress range of 349 MPa is predicted for the ‘cold start’ cycle (Fig. 4.15), as opposed to 300 MPa for the simplified cycle (Fig 4.11), both cases with tensile mean stresses of greater than 30 MPa. Figure 4.17 shows four consecutive cycles for the ‘cold start’ plant cycle, which shows an almost identical pattern to the $N = 1$ cycle, thus resulting in shakedown to cyclic creep behaviour. Overall, the predicted inelastic strain range (and accumulated strain within each cycle) is greater than that of the simplified minor cycle, which is unexpected given the presence of a hold period at about 300 $^\circ$C for the former and the much longer time period for heating to 500 $^\circ$C.

![Figure 4.15](image)

Figure 4.15. Predicted hoop stress-strain response at inside surface of pipe at $N = 1$ for 'Cold start cycle' of Figure 4.2.
Figure 4.16. Predicted evolution of creep and plastic hoop strains for 'Cold start cycle' of Figure 4.2.

Figure 4.17. Predicted hoop stress-strain response at inside surface of pipe for first four cycles for 'Cold start cycle' of Figure 4.2.
4.3.4. Load following cycle

Figure 4.18 shows the predicted temperature-time histories at the inside and outer surfaces of the pipe for the load-following cycle.

![Temperature-time history graph](image)

Figure 4.18. Predicted thermal histories at inside and outside surfaces of pipe for 'Load following cycle' of Figure 4.3. Inside surface temperature drops to almost 158 °C at the ‘trip’.

The average temperature difference across the pipe wall is approximately 35 °C. The inside surface of the pipe is on average 20 °C colder than the bulk steam temperature. After 80 hours a trip has been modelled, characterised by the sudden drop in inside and outside surface temperatures, based on the recorded input data.

Figures 4.19 to 4.21 shows the predicted hoop stress-strain response near the inside surface for the ‘load-following’ plant cycle. It is clear that the ‘trip’ (sudden loss of power and hence steam temperature and pressure) leads to a sudden tensile stress peak of 280 MPa (point B). The equally sudden resumption of power, as well steam temperature and pressure, leads to an equally sudden change to a compressive (hoop) stress of -72 MPa. The resulting predicted inelastic strain range is 0.042 %; the predicted hoop stress range is 352 MPa, with a mean stress of 104 MPa.
Figure 4.19. Predicted hoop stress-strain response at inner pipe surface for ‘trip’ part of 'Load following cycle' of Figure 4.3.

Figure 4.20. Predicted sudden ‘trip’-induced transient hoop stress response in 'load following cycle' of Figure 4.3.
Figure 4.21. Predicted time history of creep and plastic hoop strains for ‘trip’-induced transients in 'load following cycle' of Figure 4.3.

Table 4.4 shows a comparison of the plant cycle results with a marked difference between the load-following case and the other two cycles.

Table 4.4. Comparison of results from the three plant cycles

<table>
<thead>
<tr>
<th>Cycle</th>
<th>Mean Hoop Stress (MPa)</th>
<th>Max. Hoop tensile Stress (MPa)</th>
<th>Hoop Stress Range (MPa)</th>
<th>Inelastic Hoop Strain Range (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Simplified (Major cycle)</td>
<td>30</td>
<td>170</td>
<td>300</td>
<td>0.038</td>
</tr>
<tr>
<td>Simplified (Minor cycle)</td>
<td>47</td>
<td>170</td>
<td>247</td>
<td>0.015</td>
</tr>
<tr>
<td>Cold Start</td>
<td>30</td>
<td>195</td>
<td>349</td>
<td>0.03</td>
</tr>
<tr>
<td>Load-Following</td>
<td>104</td>
<td>280</td>
<td>352</td>
<td>0.042</td>
</tr>
</tbody>
</table>
4.4. Summary and Conclusions

The key summary and conclusion points are as follows:

1. A two-layer viscoplasticity material model has been developed for P91, based on previously published uniaxial, cyclic, isothermal test data from the literature. This material model has been shown to have the capability to model combined cyclic elastic-plastic and creep behaviour, as well as non-linear kinematic hardening/softening and strain-rate effects. For the P91 material studied here, experimentally-observed cyclic softening behaviour was modelled.

2. A finite element based, sequential thermomechanical methodology for pressurised steam pipes was developed for power plant piping components, incorporating (i) a transient heat transfer method taking as input the steam temperature-time histories, and (ii) a multiaxial implementation of the P91 cyclic viscoplasticity material model, taking as input the pressure-time histories.

3. Realistic power plant steam and header temperature and pressure time-history data was employed for (i) inverse identification of external convection heat transfer coefficient and (ii) thermomechanical analysis for characterisation of realistic thermomechanical histories in P91 pipe components.

4. The plant cycle analyses included a cold-start cycle, a load-following cycle, incorporating a ‘trip’ and a simplified cycle for comparison purposes.

5. Large stress ranges of more than 247 MPa with tensile mean stresses were predicted for all three plant cycles.

6. The cold-start cycle produced a total inelastic strain of 0.03 %, whereas the simplified cycle predictions revealed cyclic inelastic strains of 0.038 % and
0.015% for the major and minor cycles respectively, with compressive (hoop) creep strains resulting from inside pipe surface heat-up and tensile (hoop) stresses and inelastic strains being driven by inside surface cool-down. A large number of smaller (tensile) stress cycles resulted from thermal differentials associated with the attemperation process for the realistic cold-start cycle.

7. This analysis allows for extension of the methodology developed here to more complex piping components, such as branched headers, welded connections etc., and will develop life prediction methods.
Chapter 5

DEVELOPMENT OF LIFE ASSESSMENT PROCEDURES FOR POWER PLANT HEADERS OPERATED UNDER FLEXIBLE LOADING SCENARIOS

5.1. Introduction

A finite element methodology for thermomechanical fatigue analysis of a subcritical power plant superheater outlet header under realistic loading conditions is presented. The methodology consists of (i) a transient heat transfer model, (ii) a sequential anisothermal cyclic viscoplasticity model for the material and header, implemented within a global-sub-modelling framework, and (iii) a multiaxial, critical-plane implementation of the Ostergren fatigue indicator parameter. The methodology will permit identification of the local thermomechanical stress-strain response at critical locations and prediction of fatigue life and cracking orientation within the complex connection for transient, anisothermal, cyclic elastic-plastic-creep material behaviour. As in the previous chapter, the measured plant data, in the form of steam and pipe temperature transients and steam pressure data, are employed to identify heat transfer constants and validate the predicted thermal response, with particular attention given to plant start-up and attemperation effects. A representative test cycle is also presented, with a view to capturing the salient thermomechanical cyclic damage of the realistic cycle.

In this chapter, the previously discussed methodology of Chapter 4, is applied to a realistic section of a power plant header. Again, this methodology employs, a transient heat transfer model, which is developed to predict the associated
temperature distribution across the wall thicknesses of the header geometry, and the subsequent thermomechanical analysis which utilises temperature dependent material properties in order to capture material behaviour when subjected to plant thermal transients during start-up. In order to accurately predict the thermomechanical cyclic behaviour of a P91 steam header with multiple stubs, mesh refinement of the global model via a submodeling technique will allow for interrogation of identified critical locations and allow for a life prediction methodology to be applied. A subsequent representative cycle is then presented which is based upon a simplified version of the realistic cycle. The representative cycle analysis aims to be utilised as an aid to plant operators whereby salient features of a plant operating cycle can be used as a means of predicting components life.

A multiaxial critical plane (CP) methodology is implemented using the Ostergren thermomechanical fatigue parameter as a fatigue indicator parameter (FIP) to predict number of cycles to crack initiation. The CP approach is predicated on the observation of nucleation of fatigue cracks on preferential planes.

5.2. Methodology

5.2.1. General

As previously mentioned, a key component of the methodology is the ability to predict the thermomechanical temperature-stress-strain response of header sections under realistic loading histories. The methodology developed here employs a finite-element (FE) code to predict the thermal gradients in such piping components, and the stress-strain response of the header subjected to realistic thermal gradients coupled with pressure histories. An internal steam pressure is applied to the inside surfaces of the header $p_i(t)$, along with axial loads $F_{ax}(t)$, applied to the thick walled
section and to the individual stub pipes, which simulate the closed end condition. Figure 5.1 shows the as-constructed header geometry in end-view (Fig. 5.1(a)), stub-header junction sectional close-up (Fig. 5.1(b)), and a partial view of the overall header geometry (Fig 5.1(c)). This figure gives the reader an indication of the intricacy and scale of one of these superheater headers. Where four equally spaced stub pipes intersecting the thick-walled header section, the exact arrangement and inclination of the stub pipes are detailed in Figure 5.1a. At the stub-header intersection the weld dimensions vary due to the intersection of two cylinders of differing diameters, the dimensions of which are shown in the sectional view Figure 5.1b. Figure 5.1c depicts a sample region (highlighted by the dashed red line) that is the subject of analysis of this chapter.

A global submodeling technique is employed in order to deliver an efficient but accurate solution to this complex three-dimensional thermomechanical problem.
Figure 5.1. The as-constructed drawing of the header geometry showing (a) the end-view, (b) cross-section of the stub-header junction, and (c) a partial view of the overall header.

5.2.2. Temperature and loading histories

Figure 5.2 shows a set of temperature-time histories, consisting of (i) steam temperature, $T_i$, during a ‘cold’ plant start-up, (ii) outer wall temperature, $T_o$, from a steam header during the same time period, and (iii) an estimated enclosure temperature-time history, for the building in which the header is situated. This cycle is hereafter referred to as the ‘realistic’ cycle. The steam temperature, $T_i$, was
inferred from a thermocouple attached to a stub pipe; the outer wall temperature, $T_o$, was measured using a thermocouple attached to the outside surface of the thick walled section of the header. As indicated in Fig. 5.2, the measured data are only available for the first 20 hours; beyond this, the steam, outside header and enclosure temperatures are assumed to cool at a fixed rate.

![Figure 5.2. Realistic cold start cycle, based on measured plant data.](image)

It can be seen that the steam temperature (blue solid line in Fig. 5.2) follows the following pattern:

(i) $T_i$ initially increases from ambient temperature to 398 °C over 2.4 h; $T_i$ remains for a further 8.3 h at an intermediate temperature, which fluctuates about a mean value of approximately 300 °C. (These fluctuations are the result of the attemperation process, where cooling water is injected directly into the steam flow, to control the steam operating temperature).

(ii) $T_i$ increases gradually, over the next 3 h, to a steady temperature of about 490 °C.

(iii) $T_i$ stays approximately at this temperature for about 5.8 h.
(iv) \( T_i \) returns to ambient temperature at a rate of 13°C/h over 36 h.

The outer wall temperature, \( T_{ow} \), generally lags the steam temperature, \( T_i \), with temperature gradients of up to 150 °C, for example, during heat-up (after about six hours). An exception is between about 5.5 h and 14 h, when quite sudden reductions in steam temperature occur, leading to large temperature gradients through the wall thickness, with a temperature differential between steam temperature and outer wall of up to 100 °C. At the highest steady temperature (490 °C), the wall temperature is about 10 °C cooler than the steam. Two times \( H_1 \) and \( C_1 \) are annotated in Figure 5.2, these times correspond to cases when thermal gradients with a hotter inside surface (heating transient, \( H_1 = 1.974 \) hrs) and a cooler inside surface (cooling transient, \( C_1 = 7.434 \) hrs) occur. The steam header is housed in an enclosure which is heated over time,; the enclosure temperature, \( T_\infty \) (indicated by the dashed line in Figure 5.2) is initially at ambient temperature and increases to 325 °C after initial start-up, and then, following a dwell, lagging that of the steam, ascends to about 475°C, still lagging the steam temperature, this lag between the steam temperature and \( T_\infty \) is deemed to cause significant thermal gradient between the inner and outer surfaces of the header during start-up.

A simplified representative cycle (referred to hereafter as the ‘representative’ cycle), is presented in Figure 5.3, which captures the salient features of the realistic cycle, but without the complex thermal transients shown in Figure 5.2. The first part of the representative cycle simulates a start-up to 400 °C and includes one contrived attemperation event. The steam temperature, \( T_i \) increases to 500 °C followed by a linear cooling rate, as in the realistic cycle.

Figure 5.4 shows the steam pressure cycle for both the realistic and representative cycles. The pressure increases from zero to 17 MPa linearly over the
first 4.4 h, the pressure is then held constant for a further 16.8 hrs during which time
the plant temperatures eventually stabilises. When the plant enters its cool-down
phase the pressure decreases to zero pressure over 1.4 hrs as the plant shuts-down. In
order to conduct a comparison between times $H_1$ and $C_1$ of the realistic cycle, times
$H_2$ and $C_2$ are annotated in Figure 5.3; these times corresponding to a heating
transient ($H_1 = 1.974$ hrs) and a cooling transient ($C_1 = 7.4$ hrs) which mirror the
transients of the realistic cycle.

Figure 5.3. Representative cold-start cycle for plant header conditions, designed to
mimic salient thermomechanical damaging plant conditions.
5.2.3. Thermal model

The thermal model employed to simulate the plant cycle is described as follows:

1. A time-dependent steam temperature $T_i(t)$ is specified, based on measured data, (see e.g. Figs. 5.2 and 5.3.);

2. A time-dependent enclosure temperature $T_{\infty}(t)$ is specified based on an assumed profile (e.g. Fig. 5.2), informed by industrial experience, or, where available, outer wall temperature measurement, $T_o$, is used;

3. Natural convection is modelled on the header external surfaces;

4. Forced steam convection is modelled on the header internal surfaces;

Transient heat transfer, with temperature-dependent conductivity and specific heat data, is modelled through the header walls.

The thermal properties of P91 is presented in Table 4.1, from ASME B31.1 [1995]. The determination of steam-side convection coefficients is described in Chapter 4,
Section 4.2.3 using Equations 4.1 to 4.3. As with the plain pipe analysis, the steam velocity is taken to be 15 m/s, the recommended velocity to prevent significant pressure drop in long steam pipe length [Spirax Sarco, 2010]. The resulting $h_i$ values for the header surfaces are tabulated in Table 5.1 An inverse method is employed to determine the enclosure-side (outer header surface) convection coefficient, $h_o$, based on correlating the predicted and measured outer pipe surface temperature, $T_o$. The value of $h_o$ was chosen to give the best correlation between predicted and measured outer surface temperature, $T_o$, seen in Figure 5.2. The resulting value of $h_o = 1800$ W/m$^2$.K is used for the external surfaces.

Table 5.1. Calculated steam-side convection heat transfer coefficients.

<table>
<thead>
<tr>
<th>Header region (Internal surfaces)</th>
<th>$h_i$ (W/m$^2$.K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Stub pipes</td>
<td>4786</td>
</tr>
<tr>
<td>Thick-walled pipe</td>
<td>3611</td>
</tr>
</tbody>
</table>

5.2.4. Cyclic viscoplasticity model

Isothermal, in-phase (IP) TMF and out-of-phase (OP) TMF test data for P91 are presented in Saad et al., [2010] for the temperature range 400 °C to 600 °C, a strain-range of 1% and a strain-rate of $1 \times 10^{-3}$ s$^{-1}$. A comparison for in-phase and out-of-phase thermomechanical loading for the model is shown in Figure 5.5a and 5.5b, for the isothermally-calibrated material model. The model predictions are compared to data from Saad et al., [2010]. The data corresponds to a temperature variation between 400 °C and 500 °C and a strain rate of 0.1 % s$^{-1}$. 
Figure 5.5. Comparison of predicted and measured ([Saad et al., 2011]) uniaxial, anisothermal cyclic stress-strain curves for P91 material for TMF-IP (a) and TMF-OP (b) between 400 °C and 500 °C.

In this chapter, we consider a welded branch connection of a power plant header. Previous work, e.g. [Rayner et al., 2005], has carried out steady-state creep and creep damage analyses of welded branched connections similar to that of the present study, using different Norton and creep damage material parameters for the three weld regions (weld, HAZ and parent metal) in an branched pipe connection. However, as limited data are available for high temperature, cyclic viscoplasticity characterisation of P91 weld metal and heat-affected zones, the current work assumes that the same material parameters can be used to represent weld metal, heat affected zones (HAZ) and base metal regions.

5.2.5. Global and sub-modelling

A key challenge in the thermomechanical analysis of power plant components is the need to capture complex geometries, including stress concentrations and to provide an accurate representation of time-dependent and possibly non-linear thermal and mechanical response, in practical solution times. To this end, a sub-modelling
approach to the thermomechanical analysis is adopted here. Sub-modelling is used to study a local part of a model with a refined mesh based on interpolation of the solution from an initial, relatively coarse, global model [ABAQUS, 2010]. The global and local submodel geometry for a header with stub-pipe attachments is shown in Figure 5.6. The global model contains 25,356 eight noded linear brick elements with reduced integration (C3D8R). A detailed mesh refinement study was then carried out to establish a converged sub-model mesh for the analysis; these analyses established that the stresses were converged to within 3%. The dependence of stress with respect to mesh density is graphically illustrated in Figure 5.7. This degree of convergence required 78,276 C3D8R elements with particular refinement focused on the inner bore and the weld region ensuring that maximum integration point depth from the free surface were no greater than 150 µm at these locations. Sectional views of the critical regions of the mesh, both global and sub-model, are provided in Figure 5.8(a) for the sub-model and Figure 5.8(b) for the global model. Figure 5.8(b) also shows the key header dimensions. The lines A-A´, B-B´, C-C´, and D-D´, indicated in the figure identify locations for data analysis to be presented in Section 5.3. The first point along path C-C´, in particular, identifies the inner crotch corner, where the stub intersects the main header.
Figure 5.6. Depiction of submodel position relative to global model.

Figure 5.7. Mesh refinement study carried at (a) inner bore, and (b) weld toe.
Figure 5.8. Detail of mesh refinement in (a) inner bore region of sub-model and (b) global model. Also shown are header through-thickness paths for sampling of stress distributions.

The steady-state conduction solution for temperature distribution across a thick-walled cylinder is given by the following equation:

\[
T = \frac{T_b - T_a}{\ln \frac{b}{a}} \ln r + \frac{T_i ln b - T_a ln a}{\ln \frac{b}{a}} \tag{5.1}
\]

where \( T_a \) and \( T_b \) are the external and internal header surfaces temperatures respectively, and \( a \) and \( b \) denote the internal and external radii, respectively, of the cylinder. It has been shown [Jaeger, 1945] that, using thick-cylinder (elastic) stress analysis (Lame’s equations), the radial distributions of axial, radial and hoop thermal stresses, respectively, for elastic material behaviour are as follows:
\[ \sigma_z = \frac{E(T)\alpha(T)(T_a - T_b)}{2(1-v)\ln \frac{b}{a}} \left[ 1 - \frac{2a^2}{b^2 - a^2} \ln \frac{b}{a} - 2\ln \frac{b}{r} \right] \] (5.2)

\[ \sigma_y = \frac{E(T)\alpha(T)(T_a - T_b)}{2(1-v)\ln \frac{b}{a}} \left[ \frac{a^2}{b^2 - a^2} \left( \frac{r^2}{b^2} - 1 \right) \ln \frac{b}{a} - \ln \frac{b}{r} \right] \] (5.3)

\[ \sigma_\theta = \frac{E(T)\alpha(T)(T_a - T_b)}{2(1-v)\ln \frac{b}{a}} \left[ 1 - \frac{a^2}{b^2 - a^2} \left( \frac{r^2}{b^2} + 1 \right) \ln \frac{b}{a} - \ln \frac{b}{r} \right] \] (5.4)

where the Young’s modulus, \( E(T) \) and coefficient of thermal expansion \( \alpha(T) \) are dependent on temperature.

**5.2.6. Representation of multiaxial fatigue damage**

The critical plane approach, e.g. [Das and Sivakumar, 1999], is a method to determine fatigue damage under multiaxial conditions. A chosen fatigue indicator parameter, FIP, is evaluated on all candidate planes, defined in 3D by the angles \( \theta \) and \( \theta_R \), as shown in Figure 5.9(a). The critical plane is the plane which maximises the FIP. The method can thus predict location, orientation and life for crack initiation under multiaxial conditions. Here the FIP used is the Ostergren parameter, \( O \), such that \( \Delta O = \Delta \varepsilon_{in} \sigma_{max} \). Here the relevant stress is the stress normal to the plane, as indicated in Figure 5.9(b) and the strain is the inelastic normal strain range. The procedure is described as follows:

1. The stress-strain-time data predicted for a sample candidate material point, using the sub-model of the header, are obtained from the finite-element model.
2. For a given candidate plane, defined by \( \theta \) and \( \theta_R \), direction cosines are defined, using Equation 2.34.
3. The normal stress and strain are calculated for the candidate plane, using Equations 2.35 and 2.36, respectively.

4. Steps 2 to 3 are repeated for a given time history response at a given material point for the complete range of candidate planes (0 ≤ θ < 180° and 0 ≤ θ_R < 180°).

5. The maximum normal inelastic strain range Δε_{in} and the maximum tensile stress σ_{max}, associated with that strain range, on each particular plane are calculated. This allows for calculation of the Ostergren parameter given as Δε_{in}σ_{max}. The plane with the maximum value of O is identified as the critical plane for that material point.

6. The number of cycles to crack initiation for that material point is then predicted using the Ostergren fatigue life model, defined by Equation 2.28.

The value of the Ostergren parameters, C and β, were obtained from the published test data of EBI and McEvily, [1984] as illustrated in Figure 5.10. The constants C and β are found to be 4.5×10^3 and −1.6, respectively (for stress in units of MPa).

Figure 5.9. (a) Graphical definition of θ and θ_R for a typical material point of interest, (b) normal stress vector to identified critical plane.
Figure 5.10. Calibration of Ostergren TMF parameter from the published test data of EBI and McEvlly [1984] at a temperature of 538 °C.

5.3. Results and discussion

5.3.1. Thermal model

The thermal model was validated by comparing the measured thermal histories (Figure 5.2) to the predicted thermal history from the FE model. The points selected on the model correspond to the positions from which the measured data were taken in the plant. Figure 5.11(a) shows the measured and predicted temperature data on the outside surface of the stub pipe at point $T_1$, (approximately 150 mm away from the body of the header, see Fig. 5.8). Figure 5.11(b) shows measured and predicted data at point $T_2$ on the outside surface of the header (a distance 100 mm normal to the plane of the stub pipes, see Fig. 5.8). It is clear that the FE model captures the general trends of the measured transients (maximum difference about 35 °C). These results provide confidence in the predicted temperature fields to be used in the thermomechanical modelling.
Figure 5.11. Comparison between (a) measured outside stub surface temperature and the predicted surface temperatures of the stub outside surface and (b) measured outside header surface temperature and the predicted header outside surface temperature.

Note that, as a first approximation, the temperature of the outside surface of the stub (from point \( T_1 \)) can be taken to equal to the steam temperature, due to the thin wall thickness of the stub, which in turn is close to the inside wall temperature of the header. Hence the difference in temperature between Figures 5.11(a) and (b) reflect the instantaneous thermal gradient through the wall of the steam header, which is a key driver for the thermal stresses which are the subject matter of the present study.

The detailed FE thermal analysis provides a more accurate representation of these thermal gradients. For this analysis, two times from the thermal cycle of Figure 5.2 were selected (\( H_1 \) and \( C_1 \)), as preliminary analysis of the stress strain loops indicated that maximum stresses occurred these times. The temperature distribution at time \( H_1 \) and \( C_1 \) are depicted in Figure 5.12(a) and (c), respectively. In both cases there are clearly significant instantaneous thermal gradients, particularly in the regions of the stub-header intersections and the regions between the stubs (ligament regions). The stub pipes show a uniform wall temperature due to the small wall
thickness. The predicted thermal profiles of paths A-A´ to D-D´ (annotated in Figure 5.8(b)) are shown in Figure 5.12(b) at time $H_1$ indicating that the largest thermal gradient is along path A-A´ (blue line), with $\Delta T = 117$ °C ($T_i = 323$ °C; $T_o = 215$ °C). Figure 5.12(c) depicts the temperature distributions at time $C_1$. Figure 5.12(d) shows the temperature distribution with paths A-A´, B-B´ and D-D´ showing thermal gradients of approximately 75 °C, while path C-C´ has a significantly lower gradient of 21 °C.

Figure 5.12. (a) Predicted global model temperature distribution and (b) thermal profiles taken from the model along paths A-A´ to D-D´ at time $H_1$. (c) Predicted global model temperature distribution and (d) thermal profiles taken from the model along paths A-A´ to D-D´ at time $C_1$. 
5.3.2. Thermomechanical elastic analysis

Figure 5.13(a) shows the FE-predicted steady-state temperature distribution in a thick-walled cylinder with the same dimensions as the header of the present analysis, which is heated internally producing a thermal gradient, $\Delta T$, of 111.5 °C. Figure 5.13(b) shows the associated hoop, radial and axial stress distributions for elastic material behaviour. Clearly, the FE-predicted thermal stresses are identical with the analytical solution, using Equations 5.2 to 5.4. It can be anticipated that this plain cylinder thermal stress solutions will prevail remote from the junction regions of the header model.

(a) Steady-state temperature distribution in plain pipe with thermal gradient, (b) comparison of FE-predicted and theoretical radial, hoop and axial stress distributions corresponding to steady-state temperature distribution of (a).

The global header model was first analysed using a linear elastic material model in conjunction with the realistic temperature cycle of Figure 5.2. Only thermal loading is considered; internal pressure and end loads are ignored. Two points in time were selected (times $H_1$ and $C_1$), as preliminary analysis of the stress strain loops indicated...
that maximum stresses occurred at these times as annotated in Figure 5.2. The stress distributions along paths A-A', B-B', C-C' and D-D' in Figure 5.8(b) are first examined. The temperature distribution at time $H_1$ is depicted in Figure 5.12(b) corresponding to the predicted thermal profiles of paths A-A' to D-D' (annotated in Figure 5.8(b)), which indicate that the temperature distribution through the pipe wall is linear with the largest thermal gradient along path A (blue line), with $\Delta T = 117 \, ^\circ C$ ($T_i = 323 \, ^\circ C; T_o = 215 \, ^\circ C$). Figure 5.14(c) depicts the temperature distribution at time $C_1$. Figure 5.14(d) shows the instantaneous temperature distribution with paths A-A', B-B' and D-D' showing linear thermal gradients of approximately 75 $^\circ C$, while path C-C' has a significantly lower gradient of 21 $^\circ C$. In both cases (Fig. 5.14a, Fig. 5.14c) there is clearly significant instantaneous thermal gradients, particularly in the regions of the stub-header intersections and in the ligament regions. The stub pipes show a uniform temperature due to the small wall thickness.

In the first instance only thermal loading is considered, where internal pressure and end loads are ignored. The stress distributions along paths A-A', B-B', C-C' and D-D' in Figure 5.8(b) were examined first. Particular attention is focussed on times $H_1 = 1.974$ hrs and $C_1 = 7.434$ hrs, annotated in Figure 5.2, corresponding to hotter or cooler inside surface, respectively. Figure 5.14 shows the predicted stress distributions at time $H_1$. Figure 5.14(a)-(c) show the associated predicted stress distributions (defined relative to the header axes) at time $H_1$. The analysis predicts radial stresses for paths A-A', B-B' and D-D' which are close to zero, as expected for a relatively thin walled cylinder under thermal loading. For path C-C', which is parallel to the axis of the stub, the header radial stress is the axial stress for the stub (see Fig. 5.8(b)) and, in this case, is compressive, increasing approximately linearly from zero at the inner surface to a maximum value of 135 MPa. The axial stress for
paths A-A', B-B' and D-D' (Figure 5.14b) are also not strongly affected by the presence of the stub, with a large compressive stress at the inside surface (approximately -200 MPa) which increases almost linearly to a tensile stress of approximately 300 MPa on the outer surface. The header axial stress for path C-C' corresponds to the stub hoop stress and is seen to be significantly larger (more compressive) at the inner surface than those of paths A-A', B-B' and D-D', with a corresponding lower tensile stress at the outer surface. Figure 5.14(c) shows the header hoop stress distributions for the four paths. Again, the hoop stress at paths A-A', B-B' and D-D' are similar and unaffected by the presence of the stub, with a compressive stress on the inner surface, increasing almost linearly to a tensile stress on the outside surface. The hoop stress for path C-C' corresponds to the stub radial stress for path C-C' and is close to zero.

Figures 5.14(d)-(f) show the FE-predicted stress distributions at time C₁ (when the inside surface is cooler than the outside surface). The radial stresses in Fig. 5.14(d) show significant stress only along path C-C' — zero at the inner surface (as expected) increasing to a tensile stress of 115 MPa at the outer surface. The axial stress profiles of Fig. 5.14(e) are the inverse of the profiles of time H₁, with tensile stresses on the inside surface and, in particular, for path C-C' a large tensile stress of 256 MPa at the inside surface. The hoop stress profiles of Fig. 5.14(f) show tensile stresses on the inner surface of the header with the exception of path C where the radial stress is zero (as for Fig. 5.14c).
Figure 5.14. Predicted stress distributions along paths A-A’ to D-D’ from elastic analyses for realistic thermal cycle at time H₁ (Fig. 5.14a to 5.14c) and C₁ (Fig. 5.14d to 5.14f).

A subsequent analysis was then carried out to determine the effects of applying an internal pressure. Figure 5.15 depicts model behaviour at times H₁ (Figs. 5.15(a)-(c)) and C₁ (Figs. 5.15 (d)-(f)) for an elastic analysis where the realistic thermal loading history of Fig. 5.2 and the pressure history of Fig 5.4. are applied to the model. Figure 5.15(a) shows the effect of the applied internal pressure at the internal surface of the header where paths A-A’ to D-D’ experience a compressive stress in excess -17 MPa in contrast with the radial stresses of Figure 5.14(b) which experiences zero stress on the internal surface due to the absence of pressure in the simulation. Path C-C’ demonstrates the same general trend as Figure 5.14(b); however there is a reduction in compressive stress at the outer surface of 74 MPa, due to the superimposed compressive stress from the internal pressure. The axial
stress profiles of Figure 5.15(b) show an increase of approximately 50 MPa in the tensile stress at the outer surface for paths A-A', B-B' and D-D' and a 106 MPa increase for path C-C', when compared to those of Fig 5.15(c). Figure 5.15(c) shows a decrease in hoop stress on the internal surface of the header along paths A-A', B and D-D' due to effect of the tensile stress introduced by the application of the internal pressure and an increase in tensile stress on the outer surface of the header for the same paths, when compared to the elastic analysis results of Fig. 5.14. However an increase in compressive stress of 48 MPa is predicted at the internal surface of path C-C' which is actually the radial stress of the stub pipe. Figures 5.15(d-f) depict model behaviour during the cooling transient of C1. The radial stresses of Fig. 5.15(d) again generate a compressive stress on the internal header surface and path C-C' displays comparably large tensile stress on the outer surface to that of the thermal elastic analysis. There is an increase in axial stress on the across the header wall thickness for all paths. Similar behaviour is also predicted for hoop stress profiles where the model predicts an increase in tensile stress through the wall thickness which is particularly evident at the inner header surface.
Figure 5.15. Predicted stress distributions from elastic analysis subjected to a realistic thermal and pressure loading histories, along paths A-A' to D-D', at time H₁ (Figs. 5.15a to 5.15c) and C₁ (Fig. 5.15d to Fig. 5.15f).

5.3.3. Thermomechanical elastic-viscoplastic analysis

Figure 5.16 shows stress profiles obtained from the global model for a full non-linear elastic-viscoplastic analysis, i.e. incorporating the two layer constitutive model with Norton creep. Again, the realistic temperature cycle of Figure 5.2 is employed and the steam pressure cycle of Figure 5.4 is also included in the analysis.
Figure 5.16. Predicted stress distributions along paths A-A' to D-D' from elastic-viscoplastic analyses for realistic cycle at time H₁ (a to c) and C₁ (d to f).

Figures 5.16 (a) - (c) show the component stresses for paths A-A' to D-D' at time H₁. The radial stresses (Figure 5.16(a)) show the influence of the internal pressure of 17 MPa on the inside header surface. Path C-C' which is positioned along the inside surface of the stub pipe predict the axial stresses of the stub pipe which in this instance are predominantly compressive in nature, similar to those of Figure 5.14(a), but with evident stress reduction near the outside surface due to the effects of the internal pressure. The predicted axial stress distributions of Figure 5.16(b) are very similar to those of Fig. 5.14(b) and Fig. 5.15(b), with large tensile stresses of up to 300 MPa on the outer surface along paths A-A’ to D-D’ and large compressive stresses above 250 MPa on the inside surface. The predicted hoop stresses of Figure 5.16(c) are again similar to those of the thermal-pressure elastic analysis of Figure 14(c).
Figures 5.16(d) to 5.16(f) show the predicted elastic-viscoplastic stress distributions along the sample paths at time $C_1$. In general, the predicted results are, at least qualitatively, almost an exact inverse of the distributions for time $H_1$, which can be seen by comparing Figures 5.16(a) and 5.16(d), for example, and similarly Figures 5.16(b) to 5.16(e) and Figures 5.16(c) to 5.16(f). The radial stress of path $C-C'$ (Figure 5.16(d)) predicts comparably large stress on the outer header surface.

5.3.4. Global-sub-model elastic-viscoplastic analysis
Figure 5.17 shows von Mises stress distributions obtained from the global model for a full elastic-viscoplastic analysis. The realistic temperature cycle of Figure 5.2 is employed and the steam pressure cycle of Figure 5.4 is also included. Figure 5.17(a) shows significant stress concentrations (grey regions in figure) at time $H_1$ at the inside surface intersection corners (inner crotch corner). This observation supports the work of Nakoneczny and Schultz [1995], where regions around the stub pipe penetration with the thick-walled header body experience large thermal stresses due to rapid changes in the steam temperature where ligament cracking initiates at sites between bore holes. Figure 5.17(b) show the von Mises stress contour plot at time $C_1$. Figure 5.17(b) shows that the predicted highest stress sites are located at the stub pipes inner bore region of the junction with the header.

Based on the results of the global modelling analysis four points of interest were selected for further analysis. These points, as illustrated in Fig. 5.18, are the weld toe saddle and crotch positions, at which tensile stress have been predicted during plant heat up ($H_1$) and at the inner bore saddle and crotch corners, which are localised 'hot spots' during plant cooling transients ($C_1$). This observation supports the work of Patterson et al. [2002] where regions around the stub pipe penetration
with the thick-walled header body experience large thermal stresses due to rapid changes in the steam temperature where ligament cracking initiates at sites between bore holes.

Figure 5.17. Predicted global model von Mises stress distributions using elastic-viscoplastic material model at times (a) $H_1$ and (b) $C_1$.

Figure 5.18. Position of areas of interest within the submodel.
Figure 5.19 shows the predicted von Mises and component stress distributions at time $H_1$. The stress levels in Figure 5.19(a) may be compared directly with those in the global model (Figure 5.17(a)). Figure 5.19(a) shows a significant region of high stress concentration on the outside surface of the header and stub adjacent to the junction region and weldment. It may be noted that the peak stresses are approximately 24% higher than that of the global model (Figure 5.17(a)), indicating the importance of the sub-model in providing an accurate stress prediction. It is clear from Figures 5.19(b) to 5.19(c) that this is a biaxial tensile stress (tensile hoop and axial stresses, with comparatively smaller radial stresses). There is also a stress hot-spot on the inside surface of the header-stub junction, at the crotch corner, but this is a compressive state. The largest tensile stresses on the outside surface of the header are approximately 346 MPa at the weld toe. Figure 5.20 shows the predicted sub-model stress distributions corresponding to time $C_1$ (cooling transient). In this case, the peak stresses are predicted on the inside surface of the header, along the stub-header intersection line, extending from the crotch corner to the saddle position, with the peak tensile (hoop) stress, of up to 281 MPa, predicted at the saddle position. At the inside crotch corner, a tensile axial stress concentration of 254 MPa is predicted. Hence, biaxial tensile stresses are predicted along this header inner surface edge of the header-stub intersection, leading to tensile plasticity along this edge.
Figure 5.19. Predicted sub-model stress distributions at time $H_1$: (a) von Mises stress, (b) radial stress, (c) hoop stress, (d) axial stress.

Figure 5.20. Predicted sub-model stress distributions at time $C_1$: (a) von Mises stress, (b) radial stress, (c) hoop stress, (d) axial stress.
Figure 5.21 shows the predicted thermomechanical, stress-strain hysteresis loops for each of the three normal stress components for the realistic thermal cycle. Figures 5.21(a), and 5.21(b) provide the results for the inner bore saddle and crotch, respectively. The trends are remarkably similar, with the peak tensile and compressive stresses (hoop for the saddle and axial for the crotch) corresponding, respectively, to times $C_1$ and $H_1$. Significant stress-ranges of 443 MPa and 505 MPa are predicted for the saddle and crotch positions, respectively, with tensile stress peaks of 280 MPa and 249 MPa. For the inner bore positions, the thermomechanical loops are clearly out-of-phase, i.e. TMF-OP (annotated in Fig. 5.21(a) and Fig 5.21(b)). The predicted strain ranges for the saddle and crotch are approximately 0.3% and 0.4%, respectively.

Figure 5.21. FE-predicted (sub-model) thermomechanical stress-strain responses at various locations on header-tube junction for realistic plant cycle (Fig 5.2): (a) Inner bore - saddle position, (b) Inner bore - crotch position, (c) Weld toe - saddle position, (d) Weld toe - crotch position.
The stresses for the weld toe saddle and crotch are provided in Figs 5.21(c) and 5.21(d), respectively. The weld toe TMF hysteresis loops are predominantly tensile with tensile mean stresses, the dominant stresses being hoop for the saddle and axial for the crotch. These loops are in-phase, i.e. TMF-IP (annotated in Fig. 5.21(c) and Fig. 5.21(d)), with the peak tensile stress/strain corresponding to time $H_1$ and the peak compressive stress/strain to $C_1$. The strain ranges are significantly smaller than those of the weld toe, approximately 0.12% and 0.2%, respectively, for the saddle and crotch and the stress ranges are also smaller than for the weld toe, i.e. about 400 MPa and 500 MPa, respectively, with peak tensile stresses of 275 MPa and 340 MPa, respectively. The prediction of a large tensile stress in the axial direction is consistent with the results of Kwon et al. [2006], pertaining to observed ligament cracking in steam headers.

Figure 5.22 present the corresponding TMF hysteresis loops for the representative thermal cycle, for the same four locations. Comparison with the results in Fig. 5.21 reveals that the predicted stress-strain responses are quantitatively very similar, in terms of stress and strain magnitudes and ranges, in all three directions to those of the realistic cycle.
Figure 5.22. FE-predicted (sub-model) thermomechanical stress-strain responses at various locations on header-tube junction for representative plant cycle (Figure 5.3): (a) Inner bore - saddle position, (b) Inner bore - crotch position, (c) Weld toe - saddle position, (d) Weld toe - crotch position.

5.3.5. Thermomechanical fatigue life prediction
A TMF fatigue life analysis was carried out using the approach discussed in Section 5.2.6. Table 5.2 shows the predicted time to fatigue crack initiation (FCI), based on a position of 150 µm from the surface (the location of the closest numerical integration point to the surface in the FE model) for both the realistic and representative cycles. The critical-plane Ostergren approach has been applied based on maximum values of the stress and inelastic strain ranges over the complete cycle, i.e. neglecting the damage contributions from smaller cycles associated with temperature fluctuations. It is clear from Table 5.2 that the representative cycle gives life predictions very close to those of the realistic cycle. The use of this simplified cycle isolates the key
damaging thermal events and provides guidance for scheduling of planned maintenance of power plant.

Table 5.2. Results of multiaxial critical-plane analyses for selected locations of interest on representative TMF header simulation showing crack orientation and cycles to fatigue crack indication for both realistic plant cycle (Fig. 5.2) and representative plant cycle (Fig. 5.3).

<table>
<thead>
<tr>
<th>Region</th>
<th>Location</th>
<th>Angle $\theta$</th>
<th>Angle $\theta_r$</th>
<th>FCI Realistic (cycles)</th>
<th>FCI Realistic (years*)</th>
<th>FCI Representative (cycles)</th>
<th>FCI Representative (years*)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weld toe</td>
<td>Saddle</td>
<td>90</td>
<td>90</td>
<td>3508</td>
<td>67.5</td>
<td>3624</td>
<td>69.7</td>
</tr>
<tr>
<td></td>
<td>Crotch</td>
<td>0</td>
<td>0</td>
<td>3032</td>
<td>58.3</td>
<td>3093</td>
<td>59.5</td>
</tr>
<tr>
<td>Bore</td>
<td>Saddle</td>
<td>90</td>
<td>90</td>
<td>1998</td>
<td>38.4</td>
<td>2178</td>
<td>41.9</td>
</tr>
<tr>
<td></td>
<td>Crotch</td>
<td>0</td>
<td>0</td>
<td>1910</td>
<td>36.7</td>
<td>1954</td>
<td>37.6</td>
</tr>
</tbody>
</table>

*based on assumption of one cold start cycle per week.

The crotch position on the inside (bore) of the header is predicted to be the critical location for crack initiation, with an initiation life of almost $2 \times 10^3$ cycles predicted. The predicted orientation of the crack initiation plane indicates that cracking is perpendicular to the longitudinal axis of the header. Assuming one cold start-up per week, which is becoming typical for a fossil-fuel burning plant, with increased use of renewable energy, this corresponds to a crack initiation life of 36.7 years (realistic cycle). For fatigue crack initiation at the outside surface, which is more easily detected, the crotch position at the weld toe, is also the predicted FCI location, within about $3 \times 10^3$ cycles (corresponding to 58.3 years), which is about 60% longer than for the inside crotch position. Clearly, cracking of the inside surface can be generally considered more critical, due to the possibilities for steam pressure enhancement of crack growth, additional corrosion effects, tensile stresses on the inside surfaces of pressurised cylinders and the difficulty in detection of such cracks, without plant shut-down.
The predicted results are consistent with industrial observations. For example, ligament bore cracking of a 40 to 45 mm thick economiser header has been observed after $8 \times 10^2$ to $1 \times 10^3$ start cycles [Patterson et al., 2002]. Here, ligament bore cracking is defined as the initiation of longitudinal cracks in the vicinity of the stub pipe-header intersection and the internal header surface interface. These cracks will grow from the header ID toward the OD and between stub pipe holes [Nakonezny and Schultz, 1995]. It is important to note that the frequency of these cycles was lower than that examined here, so that significant creep damage in addition to LCF is likely to occur. In Viswanathan, [2000] weld toe crack initiation was reported, followed by radial and circumferential growth, leading to complete fracture of the tube from the header. This is consistent with the observations of Patterson et al. [2002] where stub weld fracture led to catastrophic failure.

5.4. Conclusions

A detailed study of the thermomechanical behaviour of a realistic steam header in a fossil fuel power plant has been presented, incorporating multiaxial fatigue damage prediction. The key conclusions are as follows:

- The global submodeling technique employed in these analyses has been found to be an effective method for interrogation of the identified critical locations for the FE header model subjected to thermomechanical loading conditions.

- High tensile stresses are predicted (i) on the inside surface during cooling transients and (ii) on the outside surface at the weld toe during heating transients. The cooling transient stresses are shown to be more detrimental, leading to predicted crack initiation (using the Ostergren fatigue indicator
parameter) within about 1,900 cold start cycles. Outside surface cracking is predicted to occur about 60% later than inside surface cracking.

- The predicted behaviour at the inner bore corresponds to out-of-phase temperature-strain cyclic conditions, whereas at the weld toe corresponds to in-phase temperature-strain conditions.

- Predicted cracking directions are consistent with previously reported in-service cracking.

- A simplified, representative cycle, designed to represent the salient features of the significantly more complex realistic cycle, is shown to give almost identical predicted fatigue life and very similar thermomechanical cyclic conditions, at the hot-spots of the header, to the realistic cycle.
Chapter 6

MULTIAXIAL RAINFLOW CYCLE COUNTING
THERMOMECHANICAL FATIGUE METHOD

6.1 General

The previous chapter assessed critical locations of the FE header model in terms of time to fatigue crack initiation, using the maximum stress and maximum inelastic strain range on a given plane, whilst ignoring the smaller fluctuations in stress and strain which are generated by the attemperation control process. It is these minor cycles combined with the overall stress-strain response produced by the global start-up which is the focus of this chapter. In order to determine the effect of the minor cycle on fatigue life, a method of capturing and counting the individual minor cycles is required. To perform this task, a rainflow cycle counting (RCC) technique is adopted which fits within the multiaxial critical plane framework of Chapter 5.

The RCC technique counts and quantifies, in terms of range, the individual cycles of a given stress, strain or load history, allowing summation of the individual contributions of load etc. toward (in this case) fatigue crack initiation. As a means of describing the rainflow cycle process, the arbitrary load time history of Figure 6.1a is re-oriented from the time-axis being horizontal to the time-axis pointing vertically downwards, as shown in Figure 6.1b. An effective means of describing the operation of RCC is to visualise rain flowing off a pagoda roof top, as represented by the dashed lines in Figure 6.1b, where the flow of water will move horizontally and downwards until reaching the outer most point of the roof, thus defining half a reversal. This process is repeated with the flow of water moving in the opposite
horizontal direction. However, there are some rules associated with the rainflow cycle counting technique, as follows:

1. The history being counted must be rearranged to begin with the highest peak or lowest valley.

2. Starting with highest peak move down to the next reversal. The rainflow falls off and will continue to fall unless the magnitude of the peak below is greater than or equal to, the peak from which it came, or a previous rainflow is encountered. The max peak to the outer most point is considered as half a cycle.

3. The same process must be repeated for the next reversal until the end is reached. This procedure must also be repeated for any parts that were not used in the previous counts.

4. Each cycle must be counted only once.

The application of these set of rules is best illustrated using Figure 6.1b, where the largest peak is found at point A; therefore the RCC can begin at this point. From point A move downwards to the next peak at point B. The falling rain from point B, creates point B', before proceeding to the peak at point D, the path of which is represented by the dashed red line in Figure 6.1b. This has created half a reversal (A-D). To define the remainder of the reversal; start at point D and move downwards to point E. The rain falls off point E and hits E'. The rain continues to run downwards until it reaches point I. Point I is equal to A and there are no further peaks of greater magnitude. Therefore the cycle A-D-I is counted as one cycle.

Rule 3 states that any parts of the load history that were not used must also be counted. From Figure 6.1b it can be seen that there is a cycle created by points B, C
and B'. This must be counted creating BCB', and similarly for E, H, and E' and F, G and F', creating cycles EHE' and FGF', respectively. It should be noted that every part of the load history is counted only once. The full cycles generated are depicted in Figure 6.1c, as A-D-I, B-C-B', E-H-E' and F-G-F'.

Figure 6.1. Example of rainflow cycle counting (a) load history, (b) graphical representation of rainflow cycle counting, (c) extracted cycles generated by RC counting method. [Adapted from Fuchs and Stephens, [2001]].
6.2 Rainflow cycle counting algorithms

The implementation of a rainflow cycle algorithm is required to calculate fatigue damage from long, complex loading histories. There are two main types of RCC algorithm; namely the three-point algorithm developed by Downing and Socie [1982], and the four-point algorithm formulated by Amzallag et al., [1994]. The three-point algorithm is relatively straight-forward to implement and is the algorithm of choice within this work. It is discussed as follows.

The three-point method is a vector-based method where a one-dimensional array is used to store the stress, strain or load history in question, until a cycle is identified which will then be eliminated from the array [Downing and Socie, 1982]. This method compares the range under consideration, $X$, with the previous range adjacent to $X$, which is designated as $Y$. The algorithm itself is given as follows;

\[
\text{while There is more data do} \\
\quad \text{Read the next peak or valley} \\
\quad \text{if there is less than three points do} \\
\quad \quad \text{Form ranges $X$ and $Y$} \\
\quad \quad \text{if $X \geq Y$ do} \\
\quad \quad \quad \text{Count range $Y$} \\
\quad \quad \quad \text{Discard the peak and valley of $Y$} \\
\quad \quad \text{end if} \\
\quad \quad \text{end if} \\
\quad \text{end if} \\
\text{end while}
\]
As a means of demonstrating the operation of the three-point algorithm, an arbitrary strain history is generated (see Figure 6.2) which is intended to represent a variable amplitude strain history prior to being submitted to the three-point RCC algorithm. It should be noted that the data in Figure 6.2 has been rearranged to begin with the largest peak in strain.

![Diagram of RCC algorithm](image)

Figure 6.2. Variable amplitude strain history prior to cycle extraction by the RCC technique. [Adapted from Downing and Socie, 1982].

The operation of the three-point RCC algorithm is illustrated in Figure 6.3 (a to l), which shows the counting of the individual cycles of the strain history of Figure 6.2. Figure 6.3a depicts the start of the RCC process, represented by point A. A new point is read in (point B) as shown is Figure 6.3b. As mentioned previously the algorithm compares the scalar quantities $X$ and $Y$ in order to identify a countable range. With this in mind, the absolute value between points A and B of Figure 6.3b defines $X$ ($X=\text{abs}(B-A)$), however there is no previous range adjacent to $X$ (the range
that defines \( Y \), therefore a new peak or valley must be read in, as shown in Figure 6.3c. Now comparing the new value of \( X \), which is generated by the absolute value of the difference between point \( C \) and point \( B \) (\( X=\text{abs}(C-B) \)), and \( Y \), which is the absolute value of the difference between point \( B \) and point \( A \) (\( X=\text{abs}(C-B) \)), indicates that \( X \) is equal to \( Y \), therefore range \( Y \) can be counted and it's points discarded, as illustrated in Figure 6.3d. Figure 6.3d also defines a new starting point at \( C \), however, there is insufficient data to define \( X \) and \( Y \), therefore a new peak or valley must be read in, as depicted in Figure 6.3e. \( X \) is now defined but \( Y \) remains undetermined and so the algorithm reads in another peak or valley, as illustrated in Figure 6.3f. Comparing ranges \( X \) and \( Y \), (defined as \( \text{abs}(D-E) \) and \( \text{abs}((D-C) \), respectively), reveals \( X \) is less than \( Y \) and therefore cannot be counted and another peak or valley must be read in, as shown in Figure 6.3g. Now the value of \( X \) is greater the \( Y \), (where \( \text{abs}(F-E)>\text{abs}(E-D) \)), therefore range \( Y \) may be counted and it's points discarded (see Figure 6.3h). With points \( E \) and \( D \) removed, \( Y \) is undetermined, and so a new peak or valley is read in, in this case point \( G \) (see Figure 6.3i). \( X \) will now become the range generated by the points \( G \) to \( F \) and \( Y \) by the points \( F \) to \( C \). When \( X \) and \( Y \) are compared, it is found that \( X \) is less than \( Y \), therefore \( Y \) cannot be counted and requires a new peak to be introduced. Figure 6.3j indicates that range \( X \) (\( \text{abs}(H-G) \)) is still less than range \( Y \) (\( \text{abs}(G-F) \)), therefore another peak or valley must be read into the algorithm, as shown in Figure 6.3k. The reintroduction of point \( A \) signifies the algorithm has moved through the whole history and forms the final loop, where range \( X \) defined by \( \text{abs}(A-H) \) and range \( Y \) defined by \( \text{abs}(F-G) \) are compared, revealing \( X \) to be greater than \( Y \), which is then counted as a range and it's points discarded. The remaining cycle \( C-F-A \) is processed \( X \) is equal to \( Y \), therefore the
cycle is counted. The history is now exhausted. The counted cycle are tabulated in Table 6.1.

Table 6.1. Extracted cycles from Figure 6.2 using the three-point algorithm

<table>
<thead>
<tr>
<th>Cycle number</th>
<th>Cycle range</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>AB</td>
</tr>
<tr>
<td>2</td>
<td>ED</td>
</tr>
<tr>
<td>3</td>
<td>HG</td>
</tr>
<tr>
<td>4</td>
<td>FC</td>
</tr>
</tbody>
</table>
Figure 6.3 (a to l). Depiction of the step-by-step operation of the three-point rainflow cycle counting algorithm. [Adapted from Downing and Socie, 1982].

Typically rainflow cycle counting (RCC) methods count the cycles of either a single stress or strain history. However the fatigue life model utilised in this work requires the counting of cycles in strain along with the associated cycle in stress. This requires a modified RCC method.
The RCC algorithm developed by Langlais et al., [2003], has the ability to count the hysteresis loops from a variable history for multiple channels. This approach is based on the three-point algorithm of Downing and Socie [1982]. The following methodology describes the implementation of the Langlais multiaxial rainflow cycle counting technique within the context of the multiaxial critical plane framework.

6.3 Methodology

6.3.1 Overview

Figure 6.4 depicts the critical plane methodology combined with the rainflow cycle counting technique. The steps involved in the process are labelled in Figure 6.1 and are described in detail further on in this section.

![Flow-chart of multiaxial critical plane life prediction methodology incorporating rainflow cycle counting.](image)

Figure 6.4. Flow-chart of multiaxial critical plane life prediction methodology incorporating rainflow cycle counting.
6.3.2 Steps 1 and 2 of RCC method

Figure 6.5 (a and b) shows the component stress and strain histories for a material point from the finite element simulation. These stresses and strains are read into a critical plane computer (MATLAB) program and are converted (using Equations 2.35 and 2.36) to a normal stress and normal strain history, as shown in Figure 6.6, for the candidate plane in question, in this case $\theta = 0^\circ$ and $\theta_R = 90^\circ$. These histories are then passed to the rainflow cycle counting routine for cycle extraction. It should be noted that $\sigma_{13}$, $\sigma_{13}$, $\varepsilon_{13}^{\text{in}}$, and $\varepsilon_{23}^{\text{in}}$, were also processed; however they are not shown here as they are of negligible magnitude.

Figure 6.5a Stress histories and inelastic strain histories generated at the weld toe, crotch position.
Figure 6.5b. Stress histories and the inelastic strain histories generated at the weld toe, crotch position.

Figure 6.6 depicts the normal stress and normal inelastic strain histories for a given plane (in this case, $\theta = 0^\circ$ and $\theta_R = 90^\circ$). It should be noted that the peaks in stress are not coincidental with peak in strain. The algorithm developed by Langlais et al., [2003] will identify a peak in strain and will then identify the corresponding peak in peak in stress, even if these peaks in stress and strain are not coincidental.
Figure 6.6. Normal stress and normal inelastic strain history for a given plane.

6.3.3 Steps 3 and 4
The algorithm is implemented using a stack of data, stk$_i$ where $i$ is the index of the topmost entry on the stack. Each stk$_i$ represents a sample that includes the strain and stress. In this case stk$_i$(j) represents the strain channel of the $i$-th member of the stack.

X denotes the range under consideration (i.e. the range created by the two topmost points on the stack) and Y the previous range.
Multiaxial rainflow cycle counting algorithm

1. Read a new data point, New
2. Push New onto stack, stk_i
3. **while** There are more data **do**
   4. Read a new data point, New
   5. **if** There is more than 1 point on the stack **then**
      6. **if** (New - stk_i(j)) ((stk_i-1(j)-stk_i(j)) ≥ 0) **{If New is in the same direction}** **then**
         7. **for** each auxiliary channel, **do**
         8. stk_i-2(j^max) = max (stk_i-2(j^max), stk_i-1(j^max))
         9. stk_i-2(j^min) = max (stk_i-2(j^min), stk_i-1(j^min))
      10. **end for**
      11. Remove stk_i from the stack
   12. **end if**
   13. **end if**
   14. Push New onto stack, stk_i
   15. **if** There are more than 2 points on the stack **then**
      16. X=| stk_i(j) - stk_{i-1}(j) |
      17. Y=| stk_{i-1}(j) - stk_{i-2}(j) |
      18. **while** X ≥ Y and there are more than 2 data points on the stack **do**
         19. Range Y is a completed cycle; compute damage
         20. **for** each auxiliary channel, **do**
         21. stk_i-3(j^max) = max(stk_i-3(j^max), stk_i-2(j^max), stk_i-1(j^max))
         22. stk_i-3(j^min) = min(stk_i-3(j^min), stk_i-2(j^min), stk_i-1(j^min))
      23. **end for**
      24. Remove stk_{i-1} and stk_{i-2} from the stack
      25. **if** There are more than 2 points on the stack **then**
      26. X=| stk_i(j) - stk_{i-1}(j) |
      27. Y=| stk_{i-1}(j) - stk_{i-2}(j) |
      28. **end if**
      29. **end while**
   30. **end while**
The three-point RCC algorithm requires that the data be rearranged to start with either the maximum peak or valley. In the case of this analysis the maximum peak in stress and strain is conveniently located at the beginning of the stress-strain histories. The algorithm output a series of strain ranges and their associated stresses which are shown in Table 6.2. A maximum strain range is found on cycle 17 of Table 6.2 which is associated with its corresponding stress magnitude of 390 MPa.

Table 6.2. depicts the strain ranges and their associated maximum tensile stresses extracted by the multiaxial RCC technique for the stress-strain history of Figure 6.6.

<table>
<thead>
<tr>
<th>Cycle number</th>
<th>Δε_in</th>
<th>σ_T (MPa)</th>
<th>FCI Parameter (MPa. mm/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.000002</td>
<td>88.4</td>
<td>1.77E-04</td>
</tr>
<tr>
<td>2</td>
<td>0.000143</td>
<td>98.1</td>
<td>1.40E-02</td>
</tr>
<tr>
<td>3</td>
<td>0.000179</td>
<td>43.1</td>
<td>7.71E-03</td>
</tr>
<tr>
<td>4</td>
<td>0.000171</td>
<td>40.3</td>
<td>6.89E-03</td>
</tr>
<tr>
<td>5</td>
<td>0.000232</td>
<td>72</td>
<td>1.67E-02</td>
</tr>
<tr>
<td>6</td>
<td>0.000435</td>
<td>155</td>
<td>6.74E-02</td>
</tr>
<tr>
<td>7</td>
<td>0.000361</td>
<td>73.5</td>
<td>2.65E-02</td>
</tr>
<tr>
<td>8</td>
<td>0.000576</td>
<td>175.5</td>
<td>1.01E-01</td>
</tr>
<tr>
<td>9</td>
<td>0.000077</td>
<td>62</td>
<td>4.77E-03</td>
</tr>
<tr>
<td>10</td>
<td>0.000396</td>
<td>161</td>
<td>6.38E-02</td>
</tr>
<tr>
<td>11</td>
<td>0.000027</td>
<td>34</td>
<td>9.18E-04</td>
</tr>
<tr>
<td>12</td>
<td>0.000144</td>
<td>146.3</td>
<td>2.11E-02</td>
</tr>
<tr>
<td>13</td>
<td>0.000062</td>
<td>37.1</td>
<td>2.30E-03</td>
</tr>
<tr>
<td>14</td>
<td>0.000018</td>
<td>36.7</td>
<td>3.61E-04</td>
</tr>
<tr>
<td>15</td>
<td>0.000916</td>
<td>131.2</td>
<td>1.20E-01</td>
</tr>
<tr>
<td>16</td>
<td>0.000359</td>
<td>98.2</td>
<td>3.53E-02</td>
</tr>
<tr>
<td>17</td>
<td>0.003273</td>
<td>390.2</td>
<td>1.28</td>
</tr>
</tbody>
</table>
The Palmgren-Miner (Miner's Rule) linear cumulative fatigue damage theory is used to determine the total damage by summing the individual cycles extracted by the RCC process, as follows:

\[ D = \sum_{i=1}^{k} \frac{n_i}{N_{f_i}} \]  

(6.1)

where \( D \) is the overall damage, when \( D=1 \) failure has occurred, \( n_i \) is the number of minor cycles and \( N_{f_i} \) is the number of cycles to failure.

This approach utilises a major cycle generated by the global plant start-up \( N_{maj} \) and the smaller cycles given by \( N_{min} \)

\[ \frac{1}{N_f} = \frac{1}{N_{maj}} + \frac{n}{N_{min}} \]  

(6.2)

The extracted cycles listed in Table 6.2 are used as an input to Equation 6.2, where the minor cycle FCI parameters are converted to FCI cycle (\( N_{minor} \)), counted and added, in this case Table 6.2, lines 1 to 16 are considered as minor cycles, and line 17 as the major cycle.

6.3.4 Steps 5 and 6
The fatigue crack initiation damage on the particular candidate plane in question is stored, and a new candidate plane is indexed (step 5a) and the code is re-run. Upon calculation of FCI damage on all candidate planes, the plane containing the largest FCI is determined as the critical plane.
6.4 Results

Figure 6.7 (a to d) depicts the fatigue crack initiation parameter for all directions normal to the fatigue crack plane, for the various locations of interest. Figure 6.7a illustrates the predicted FCI parameters at the weld toe, crotch position. It is predicted that fatigue crack initiation will occur after 2552 cycles or 49.1 years, which, when compared to the lives predicted by the multiaxial critical plane (MCP) method of Chapter 5, reveals a ~1 % shortening in life at this location. The crack direction however remains the same as the MCP method, where Figure 6.7a depicts the maximum FCI parameter occurring on the plane shown in Figure 6.8a, where the stress and strain normal directions are in the axial direction (θ = 0° or 180°). Figure 6.7b illustrates the predicted FCI parameters at the weld toe, saddle position. Again, a reduction in fatigue life similar to that of the weld toe crotch position of ~1 % is predicted when compared to the same position using the MCP method. Crack direction again remains the same as the MCP method (θ = 90°, and θR = 90°).

Upon analysis of the individual minor cycles, it was found that the contribution of the minor cycles towards increasing the FCI was low. For this reason the change in fatigue lives between the RCC method and MCP method are similar. It should be noted however that Miner's rule does not take account of sequencing effects. A more suitable method perhaps is that proposed by Marco and Starkey [1954], which proposed a fatigue concept based on infrequent large cycle within normally small loading fluctuations.

Figure 6.7 (c and d) illustrates the predicted FCI parameters at the inner bore, crotch and saddle positions, respectively. Similar behaviour in terms reduction in fatigue life at both points is observed when compared against the MCP method.
Comparisons between Figure 6.7a (Weld toe, crotch position) and Figure 6.7c (inner bore, crotch position) show the effects of a multi-axial stress state at the weld toe, where a relatively large FCI parameters are found in wide range of direction (ie between $\theta = \sim 50^\circ$ to $180^\circ$, and $\theta_R = \sim 70^\circ$ to $130^\circ$). Similar behaviour is predicted between the weld toe, and the inner bore saddle positions. The direction of cracking is presented in Figure 6.8.
Figure 6.7. Fatigue crack initiation parameter for all candidate planes for positions including: (a) weld toe crotch corner, (b) weld toe saddle position, (c) inner bore crotch position, and (d) inner bore saddle position.
Figure 6.8. Illustration showing orientation of fatigue crack plane relative to the submodel geometry for locations of interest including (a) weld toe, crotch position, (b) inner bore crotch position, (c) weld toe, saddle position, and (d) inner bore, saddle position.

Table 6.3 depicts the lives predicted by the multiaxial critical plane RCC method where the inner bore crotch position remains the location where failure is likely to occur first.
Table 6.3. Results of multiaxial critical-plane rainflow cycle counting analyses for selected locations of interest on realistic TMF header simulation for the realistic cold-start cycle (see Fig. 5.2).

<table>
<thead>
<tr>
<th>Region</th>
<th>Location</th>
<th>Angle $\theta$</th>
<th>Angle $\theta_g$</th>
<th>Fatigue damage</th>
<th>FCI Realistic (cycles)</th>
<th>FCI Realistic (years*)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weld toe</td>
<td>Saddle</td>
<td>90</td>
<td>90</td>
<td>1.18</td>
<td>3473</td>
<td>66.7</td>
</tr>
<tr>
<td></td>
<td>Crotch</td>
<td>0</td>
<td>90</td>
<td>1.30</td>
<td>2959</td>
<td>56.9</td>
</tr>
<tr>
<td>Bore</td>
<td>Saddle</td>
<td>90</td>
<td>90</td>
<td>1.68</td>
<td>1972</td>
<td>37.9</td>
</tr>
<tr>
<td></td>
<td>Crotch</td>
<td>0</td>
<td>90</td>
<td>1.77</td>
<td>1816</td>
<td>34.9</td>
</tr>
</tbody>
</table>

*based on assumption of one cold start-up cycle per week.

6.5. Conclusion

A rainflow cycle counting algorithm was successfully incorporated within a multiaxial critical plane fatigue methodology, which is capable of assessing thermomechanical stress-strain behaviour. The key findings are as follows:

- The rainflow cycle counting procedure predicts a decrease in fatigue life in all locations investigated. It was found that all predicted fatigue crack directions remain the same as those predicted by the multiaxial critical plane method of Chapter 5.

- The multiaxial critical plane rainflow cycle counting fatigue method predicts shorter lives than those predicted by the critical plane method without the RCC, by approximately 1%. This is due to the comparably small inelastic strain ranges of the minor cycles.
Chapter 7

THERMOMECHANICAL ANALYSIS OF A SERVICE-AGED P91 SUPERHEATER HEADER

7.1 General

This chapter presents a sequential thermomechanical analysis, including transient heat transfer and subsequent cyclic elastic-viscoplastic analysis, of the power plant header of Chapter 5, for the service-aged P91 material. The cold start cycle is used for these analyses.

The material constants identification and calibration methodology for the service-aged P91 material is firstly described with particular, focus given to determination of the isotropic and non-linear kinematic hardening constants, and the creep constants required to implement an elastic-viscoplastic model. The procedure followed is based on the published work of Deshpande et al., [2010].

The identified constants are then utilised within the complex transient thermomechanical simulation of the realistic header model described in Chapter 5 in order to predict the stress-strain behaviour of the header at critical locations. The previously described multi-axial critical-plane life prediction methodology of Chapters 5 and 6 is utilised in order to predict the number of start-up cycles to fatigue crack initiation for the previously identified critical locations on the header.
7.2 Identification of material constitutive constants

7.2.1 Non-linear kinematic hardening constants identification

The material constant $k$, is identified by selecting the linear portion of the true stress and plastic strain loop at the half life, as shown in Figure 7.1. A value for $2k$ is identified at the different strain ranges at a given temperature and the mean of these individual values is taken to be the final $k$ [Lemaitre and Chaboche, 1990]. The non-linear kinematic hardening (NLKH) coefficients $C$ and $\gamma$ are identified by fitting Equation 4.15 which reflects the relationship between the cyclic stress-strain to the measured cyclic stress-strain data [Lemaitre and Chaboche, 1990]. The ratio $C/\gamma$ is determined as the asymptotic value of $\Delta \sigma /2-k$ versus the inelastic strain amplitude, which is taken to be the saturated value of the backstress.

Figure 7.1 Identification of material constant $k$ from half-life stress-plastic strain loops, for different applied strain ranges at 500 °C.

Figure 7.2 shows the identification of $C$ and $\gamma$ for the service-aged P91 material at 20 °C, 400 °C, 500 °C, and 600 °C for a strain rate of 0.033 %/s based on the half-life stress-plastic strain curves. Figure 7.3 depicts the stress-strain response at the
half-life using the non-linear kinematic hardening model compared against the experimental data, for two different strain ranges, and for various temperatures.

Figure 7.2. Identification of material constants $C$ and $\gamma$ for service-aged P91 for a constant strain rate of 0.033 %/s for all temperatures, based on the half-life cyclic stress-plastic strain curves.

Figure 7.3 (a to h) depicts a comparison between the NLKH model and the experimental stress-strain response at the half-life. Figure 7.3 (a to b) shows the NLKH models performance compared with the corresponding test results at two different strain ranges, at 20 °C, for a constant strain rate of 0.033 %/s. Similarly,
Figure 7.2 (c - d), compares the respective results at 400 °C, 500 °C (Figure 7.2 (e - f)), and 600 °C (Figure 7.2 (g - h)).

![Graphs of stress-strain response at different temperatures.](image)

Figure 7.3. Comparison of predicted and measured stress-strain response at the half-life using the non-linear kinematic hardening model, for two different strain ranges,
for 20 °C (Figure 7.2 (a - b)), 400 °C (Figure 7.2 (c - d)), 500 °C (Figure 7.2 (e - f)), and 600 °C (Figure 7.2 (g - h)), for a constant strain rate of 0.033 %/s.

7.2.2 Determination and validation of isotropic hardening constants
Due to the fact that all of the service-aged P91 LCF test results exhibited continuous cyclic softening, a negative isotropic term, $Q$, is used in order to capture the cyclic softening behaviour. Parameter $Q$ is typically identified by the difference in stress amplitude ($\Delta\sigma/2$) between the initial and stabilized loops. However, in this study the half-life is utilised due to the continuous softening behaviour of the SA P91 during testing. Therefore, the constant $Q$ was taken to be the stress difference between the first and the half-life cycle. The evolution of isotropic softening is defined as follows:

$$ R = Q(1 - e^{-bp}) $$ (7.1)

The change in parameter $p$ is equal to $2\Delta\varepsilon_p$ and its value increases monotonically with cycles. The constant $b$ was determined by fitting Equation 7.1 to the test data, as shown in Figure 7.4. This figure shows the fit for two different strain ranges for temperatures from 20 °C to 600 °C.

The NLKH model was then combined with the isotropic hardening model. Figure 7.5 depicts a comparison between the test results and the combined isotropic-non linear kinematic hardening model for $N=1$, for two different strain ranges, for temperatures 20 °C to 600 °C. In general, the model is seen to predict the cyclic stress-strain and softening behaviour of the SA P91 material reasonably well.

Figure 7.6 depicts the combined isotropic-NLKH models ability to capture the softening effects of the service-aged P91 material across the various temperatures and strain ranges.
Figure 7.4. Determination of isotropic hardening parameters for various temperatures and strain ranges, for a strain rate = 0.0333 %/s.
Figure 7.5. Combined isotropic and non-linear kinematic hardening stress-strain loops at N=1, for two different strain ranges, for 20 °C (Figure 7.3 (a - b)), 400 °C (Figure 7.3 (c - d)), 500 °C (Figure 7.3 (e - f)), and 600 °C (Figure 7.3 (g - h)), for a constant strain rate of 0.033 %/s.
Figure 7.6. Combined isotropic and non-linear kinematic hardening cyclic softening curves, for two different strain ranges, for 20 °C (Figure 7.4 (a - b)), 400 °C (Figure 7.4 (c - d)), 500 °C (Figure 7.4 (e - f)), and 600 °C (Figure 7.4 (g - h)), for a constant strain rate of 0.033 %/s.
7.2.3 Determination of creep constants
Cyclic tension hold tests were carried out for temperatures greater than and equal to 400 °C. A representation of a tension-hold load cycle is depicted in Figure 7.7. The first quarter cycle of a given cyclic tension-hold test can be considered to be stress relaxation test during which the specimen was held in tension for a period of 120 s.

![Figure 7.7. Representation of tension-hold strain-controlled test conditions.](image)

An initial set of values for creep constants $A$ and $n$ were identified by plotting the experimental test data in the form log stress rate against log stress, as depicted in Figure 7.8 for a range of temperatures. Equation 2.3 was then fitted to the experimental data and the creep constants $A$ and $n$ were iteratively adjusted in order to achieve a good fit to the data, as shown in Figure 7.9. The final creep constants are given in Table 7.1.
In order to determine the user-specified ratio, $F$, a uniaxial stress relaxation simulation using the two-layer viscoplasticity model was carried out in ABAQUS, utilising all the previously identified isotropic, kinematic, and creep parameters. The value of $F$ is adjusted by trial and error to achieve a good fit the experimental stress relaxation data. The agreement between the test results and the two-layer model is shown in Figure 7.10, at $N = 1$, for temperatures between 20 °C and 600 °C. Figure 7.11 shows a comparison between the measured TMF data, for a temperature range of 400 °C – 600 °C, compared against the two-layer viscoplasticity model for in-phase, under the same temperature conditions, and Figure 7.12 shows the out-of-
phase comparison, for a temperature range of 600 °C to 400 °C. Both figures illustrate good agreement between the model and the experimental data.

Figure 7.9. Validation of analytical model (power law creep) model and the two-layer viscoplasticity model for 400 °C to 600 °C.

Table 7.1. Identified parameters for service-aged P91 material across a range of temperatures

<table>
<thead>
<tr>
<th>T  (°C)</th>
<th>E  (MPa)</th>
<th>k  (MPa)</th>
<th>Q  (MPa)</th>
<th>b</th>
<th>C  (MPa)</th>
<th>γ</th>
<th>A  (MPa° hr⁻¹)</th>
<th>n</th>
<th>F</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>210213</td>
<td>300</td>
<td>-75.47</td>
<td>0.32</td>
<td>399500</td>
<td>1463.4</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>400</td>
<td>184230</td>
<td>205</td>
<td>-72.42</td>
<td>0.283</td>
<td>220290</td>
<td>1112.6</td>
<td>5.19E-58</td>
<td>19.8</td>
<td>0.07</td>
</tr>
<tr>
<td>500</td>
<td>168220</td>
<td>175</td>
<td>-96.855</td>
<td>0.5</td>
<td>136000</td>
<td>763</td>
<td>4.28E-40</td>
<td>13.55</td>
<td>0.1</td>
</tr>
<tr>
<td>600</td>
<td>147840</td>
<td>155</td>
<td>-103.15</td>
<td>0.583</td>
<td>132900</td>
<td>1020.7</td>
<td>4.1E-21</td>
<td>6.625</td>
<td>0.11</td>
</tr>
</tbody>
</table>
Figure 7.10. Validation of the two-layer viscoplasticity model at $N = 1$ for various temperatures.
Figure 7.11. Stress-strain response of the two-layer viscoplasticity model compared against the measured experimental data under IP TMF loading conditions, at $N = 1$, for a temperature range of $400 \, ^\circ C - 600 \, ^\circ C$.

Figure 7.12. Stress-strain response of the two-layer viscoplasticity model compared against the measured experimental data under OP TMF loading conditions, at $N = 1$, for a temperature range of $600 \, ^\circ C - 400 \, ^\circ C$. 
7.3. Thermomechanical elastic-viscoplastic analysis of a superheater outlet header

The realistic anisothermal header analysis methodology of Chapter 5 is applied in this chapter with the material constants for the service-aged P91 material. The FE simulation carried using these service-aged P91 constants will be called the “service-aged analysis”, and the simulation of in Chapter 5 will be referred to as the “comparative P91 material analysis” hereafter.

Figure 7.13 shows a comparison between the predicted stress history for the comparative material analysis and the service-aged analysis, at the inner bore crotch position. It can be seen that the general trend of both the stress histories is very similar.

![Figure 7.13](image.png)

Figure 7.13. Comparison of FE predicted axial stress histories of the service-aged P91 material and the comparative P91 material (utilised in the FE header analyses in Chapter 5) at the inner bore crotch position, over the first 80,000 s.
Figure 7.14 shows a comparison between the mechanical axial strain histories of both the service-aged P91 and the comparative P91 material (utilised in Chapter 5) analyses. These strain histories show a similar trend to the stress history of Figure 7.13.

Figure 7.14. Comparison of FE predicted mechanical axial strain histories at inner bore crotch position for the comparative P91 material (utilised in the FE header analyses in Chapter 5) and service-aged analyses, over the first 80,000 s.

Figure 7.15 compares the stress-strain responses of the comparative P91 material and service-aged analyses at the inner bore crotch position. There is a clear difference in terms of inelastic strain range between the two analyses, where the comparative P91 material analysis predicts an inelastic strain of 0.065 %, compared with a predicted service-aged value of 0.042 %.
Figure 7.15. Comparison of FE predicted stress-strain response of service-aged and comparative P91 materials at the inner bore crotch position.

Figures 7.16a to 7.16d show the predicted von Mises and component stress distributions at time $H_1$ (heating transient) for the service-aged analysis. Direct comparison can be made between the contour plots of Figure 7.16 (a to d) and (the comparative P91 material analysis contour plots of) Figure 5.19 (a to d) in Chapter 5, which captures the instantaneous stress state during a heating transient at time point $H_1$. Similarly Figure 7.17 (a to d) can be compared to Figure 5.20, corresponding to
the cooling transient $C_1$. The calculated FCI parameters for the SA analysis are presented in Table 7.1, at the inner bore crotch position.

Table 7.1. Fatigue crack indication (Ostergren) parameters for the inner bore crotch position.

<table>
<thead>
<tr>
<th>Cycle number</th>
<th>$\Delta \varepsilon_{in}$</th>
<th>$\sigma_T$ (MPa)</th>
<th>FCI Parameter (MPa. mm/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>2.3E-05</td>
<td>131</td>
<td>0.00301</td>
</tr>
<tr>
<td>2</td>
<td>9.9E-05</td>
<td>98.5</td>
<td>0.00975</td>
</tr>
<tr>
<td>3</td>
<td>0.0001</td>
<td>93.1</td>
<td>0.0094</td>
</tr>
<tr>
<td>4</td>
<td>0.00036</td>
<td>156.3</td>
<td>0.02376</td>
</tr>
<tr>
<td>5</td>
<td>0.0002</td>
<td>294</td>
<td>0.05998</td>
</tr>
<tr>
<td>6</td>
<td>0.00019</td>
<td>73.5</td>
<td>0.01374</td>
</tr>
<tr>
<td>7</td>
<td>0.00038</td>
<td>301</td>
<td>0.11348</td>
</tr>
<tr>
<td>8</td>
<td>1.1E-05</td>
<td>145</td>
<td>0.0016</td>
</tr>
<tr>
<td>9</td>
<td>0.00019</td>
<td>172</td>
<td>0.0297</td>
</tr>
<tr>
<td>10</td>
<td>0.00034</td>
<td>289.1</td>
<td>0.04163</td>
</tr>
<tr>
<td>11</td>
<td>6.2E-05</td>
<td>75.3</td>
<td>0.00467</td>
</tr>
<tr>
<td>12</td>
<td>1.8E-05</td>
<td>72.5</td>
<td>0.00131</td>
</tr>
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<td>13</td>
<td>0.00092</td>
<td>261.2</td>
<td>0.10866</td>
</tr>
<tr>
<td>14</td>
<td>0.0003</td>
<td>167.9</td>
<td>0.05037</td>
</tr>
<tr>
<td>15</td>
<td>0.0029</td>
<td>343.2</td>
<td>0.77963</td>
</tr>
</tbody>
</table>

The von Mises contour plot of Figure 7.16a shows very similar stress behaviour in terms of locations of stress concentrations and in terms of stress magnitudes to that of Figure 5.20a. However, the radial stress contours depicted in
Figure 7.16b shows a stress concentration of in excess 200 MPa on the inside surface of the stub pipe. This tensile stress hot spot is directly adjacent to a compressive stress concentration of magnitude -184 MPa. These stress concentration are more pronounced than those of the corresponding contour plot from the comparative P91 material analysis (Figure 5.20b) which give much lower stresses at these locations. Other areas on the submodel geometry of Figure 7.16b predict very similar stresses to that of Figure 5.20b. The same is true for the hoop stresses of Figure 7.16c, showing large tensile stresses in the area behind the weld toe and supporting the observation that there is a biaxial tensile state, tensile hoop and axial stresses in the header (Figures 7.16b to 7.16c). The largest stress concentrations are found in the axial stress component, which is tensile and of magnitude 339 MPa at the weld toe; these stresses are slightly lower than that predicted by the comparative P91 material analysis.

Figure 7.17a to 7.17d shows the predicted sub-model distributions of stresses at time $C_1$ (cooling transient). As with the comparative P91 material analysis, the majority of the tensile stress concentrations are predicted on the inside surface of the header. An exception to this is the radial stress distribution of Figure 7.17, where again there is a notable stress concentration on the inside surface of the stub-pipe just above the top of the weld. This stress concentration has a magnitude in excess of 200 MPa. It is also located in the same position as the radial stress hot spot of Figure 7.16b, which occurs during a heating transient. This observation would therefore imply that this relatively large tensile stress concentration is present during heating as well as cooling transients. This stress hot spot was not predicted for the comparative P91 material analysis. The hoop stress of Figure 7.17c show a peak tensile stress at the inner bore saddle position. This is consistent with the
comparative P91 material analysis. However the peak tensile (hoop) stress, of Figure 7.17c show a higher magnitude of 296 MPa compared to 281 MPa from the comparative P91 material analysis. At the inside crotch corner, a tensile axial stress concentration of 260 MPa is predicted. This is similar to the comparative P91 material analysis prediction of biaxial tensile stresses along the inner surface edge of the header-stub intersection.

Figure 7.16. Predicted sub-model stress distributions at time H1: (a) von Mises stress, (b) radial stress, (c) hoop stress, (d) axial stress.
Figure 7.17. Predicted sub-model stress distributions at time $C_1$: (a) von Mises stress, (b) radial stress, (c) hoop stress, (d) axial stress.

The same four hot-spot locations that were selected in Chapter 5 (i.e. the inner bore crotch and saddle points along with the weld toe crotch and saddle) were again selected in the service-aged analysis for life prediction analysis. These points, are illustrated in Figure 5.18. Figures 7.18a, 7.18b, 7.18c and 7.18d show the predicted stress-strain loops radial, axial and hoop stresses for these respective points. The critical-plane Ostergren rainflow cycle counting method of Chapter 6 is applied to predict the number of cycles to fatigue crack initiation at these locations.

The stress-strain response in the hoop direction at the inner bore saddle position (Figure 7.18a) predicts a comparably larger stress range of 488 MPa, than the comparative P91 material analysis, which gave 443 MPa (see Figure 5.21). The
maximum and minimum stresses are coincidental with time points \( H_1 \) and \( C_1 \), consistent with the comparative P91 material analyses. Using the life prediction methodology of Chapter 6, the predicted number of cycles to FCI at this location is 2852 cycles. This is significantly longer FCI life than predicted when compared with the comparative P91 material analysis. This extension in predicted FCI life is primarily due to the determination of fatigue life constants from the HTLCF whereas previously the fatigue life constants were extracted from a very limited amount of test data. Significant stress-ranges 530 MPa are predicted for the inner bore crotch position, with tensile stress peaks of 285 MPa. The predicted strain ranges for the saddle and crotch positions are approximately 0.25% and 0.3%, respectively. The cycles to FCI at the crotch position is predicted to be 2735 cycles to failure, which again is longer than the comparative P91 material analysis. For the inner bore positions, the thermomechanical loops are out-of-phase, i.e. TMF-OP, which is consistent with the comparative P91 material analysis. The weld toe TMF hysteresis loops are predominantly tensile with tensile mean stresses and these are in-phase loops, i.e. TMF-IP, which is again consistent with the previous study in Chapter 5. The strain ranges are similar to those than those of the inner bore, approximately 0.25% and 0.3%, for the saddle and crotch positions respectively. The stress ranges are about 400 MPa and 450 MPa, respectively, with the peak tensile stresses of 295 MPa and also 340 MPa, respectively. Comparably larger numbers of cycles to FCI are predicted for the weld toe compared with the inner bore. At the weld toe saddle FCI is predicted to occur after 5988 cycles and after 5658 cycles at the weld toe crotch position. Both of these are significantly longer than the comparative P91 material analysis.
Figure 7.18. Stress-strain responses at various locations; (a) Inner bore-saddle position, (b) Inner bore-crotch position, (c) Weld toe-saddle position, (d) Weld toe-crotch position.

Table 7.2. Results of the multiaxial critical-plane rainflow cycle counting analyses for selected locations of interest on TMF header simulation showing crack orientation and cycles to fatigue crack initiation for the realistic plant cycle compared against the comparative P91 material predictions.

<table>
<thead>
<tr>
<th>Region</th>
<th>Location</th>
<th>Angle ( \theta )</th>
<th>Angle ( \theta_R )</th>
<th>FCI Comparative material (cycles)</th>
<th>FCI Comparative material (years(^*))</th>
<th>FCI Service-aged P91 (cycles)</th>
<th>FCI Service-aged P91 (years(^*))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weld</td>
<td>Saddle</td>
<td>90</td>
<td>90</td>
<td>(3.5 \times 10^3) (3.0 \times 10^3)</td>
<td>66.7</td>
<td>5988</td>
<td>115.2</td>
</tr>
<tr>
<td></td>
<td>Crotch</td>
<td>0</td>
<td>90</td>
<td></td>
<td>56.9</td>
<td>5658</td>
<td>108.8</td>
</tr>
<tr>
<td>Bore</td>
<td>Saddle</td>
<td>90</td>
<td>90</td>
<td>1972</td>
<td>37.9</td>
<td>2852</td>
<td>54.8</td>
</tr>
<tr>
<td></td>
<td>Crotch</td>
<td>0</td>
<td>90</td>
<td>1878</td>
<td>36.1</td>
<td>2735</td>
<td>52.6</td>
</tr>
</tbody>
</table>
7.4. Conclusions

A study of the thermomechanical behaviour of a realistic steam header using material constants determined from service-aged P91 material, incorporating multiaxial fatigue damage rainflow cycle counting life prediction has been carried out in this chapter. The key conclusions are as follows:

- The overall trends of the model which were presented in Chapter 5 are the same, in the sense that high tensile stresses are predicted (i) on the inside surface during cooling transients and on the outside surface at the weld toe during heating transients. The predicted behaviour at the inner bore corresponds to out-of-phase temperature-strain cyclic conditions (TMF-OP), whereas conditions at the weld toe correspond to in-phase temperature-cyclic strain (TMF-IP). Predicted cracking directions are the same as those presented in Chapter 5.

- The SA FE header analysis predicts longer fatigue lives at the critical locations compared with those from the comparative P91 material FE header analysis of Chapter 6. This extension in life can be attributed to the following:

- The predicted fatigue lives for the service-aged analysis predicts longer lives compared with the comparative P91 material, this was attributed to smaller inelastic strain ranges.
Chapter 8

CONSTITUTIVE CHARACTERISATION OF P91 WELD AND HEAT AFFECTED ZONE MATERIAL

8.1. Introduction
This chapter is particularly interested in understanding the constitutive behaviour of P91 weld material (WM) and heat affected zone (HAZ) material under high temperature, cyclic loading conditions.

In order to predict the material response in these regions a series of finite element (FE) models were utilised incorporating the two-layer viscoplasticity material model to simulate the cyclic stress-strain behaviour of the BM, WM and HAZ at various temperatures and strain ranges. Further details on this model can be found in Chapter 2, Section 2.5.7. The methodology to identify the temperature dependent material parameters for a cyclic viscoplasticity material model which incorporates non-linear kinematic hardening (NLKH), isotropic softening and strain-rate (creep) effects is described in the following sections.

8.2. Constitutive behaviour and finite element modelling.
8.2.1 General
Figures 8.1a to 8.1c show the FE models developed for the BM, WM and CW specimens, respectively. The constitutive parameters for the BM and WM specimens were identified first, from the corresponding single-material tests, and validated via comparison of the FE and measured responses. These identified BM and WM
parameters were then utilised within a multi-material CW simulation (Figure 8.1c) to identify the constitutive behaviour of the HAZ material (treated as a single material zone). The width of the HAZ is taken to be 2.7 mm, as identified optical microscopy (see Figure 3.46 for example). The overall height of the model is 12.5 mm which corresponds to the distance between the extensometer legs.

![Figure 8.1. Representation of FE models showing (a) BM single-material model, (b) WM single-material model, and (c) multi-material CW model.](image_url)

The identification procedure undertaken to identify all the material constants for all the material zones is given in the following sections:

**8.2.2 Step 1 - Estimation of NLKH parameters**

The cyclic yield stress $k$ is identified from the linear portion of the true stress versus plastic strain loop at the half life, averaging across the different strain ranges at each
temperature following the approach of Lemaitre and Chaboche [1990]. The NLKH coefficients $C$, $\gamma$ are identified using Equation 4.15 which describes the relationship between the measured cyclic stress and plastic strain amplitude [Lemaitre and Chaboche, 1990]. The ratio $C/\gamma$ is determined as the asymptotic value of $\Delta\sigma/2-k$ versus the inelastic strain amplitude, which is taken to be the value of the backstress at the half-life. For example, Figure 8.2 shows the identification of $C$ and $\gamma$ for the WM material at 400 °C and 500 °C. The identified NLKH constants for the WM are presented in Table 8.1. Figure 8.3 shows the comparison of predicted NLKH stress-strain responses and the measured data at the half-life, at 500 °C, for two strain ranges. The NLKH constants for the BM at 400 °C and 500 °C are identified in the same manner and are presented in Table 7.1.

Table 8.1. Cyclic viscoplasticity material parameters for WM.

<table>
<thead>
<tr>
<th>$T$</th>
<th>$k$</th>
<th>$E$</th>
<th>$\gamma$</th>
<th>$C$</th>
<th>$Q$</th>
<th>$b$</th>
<th>$A$</th>
<th>$n$</th>
<th>$f$</th>
</tr>
</thead>
<tbody>
<tr>
<td>(°C)</td>
<td>(MPa)</td>
<td>(MPa)</td>
<td>(MPa)</td>
<td>(MPa)</td>
<td>(MPa)</td>
<td>(MPa)</td>
<td>(MPa)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>400</td>
<td>235</td>
<td>202040</td>
<td>925.2</td>
<td>288600</td>
<td>-111</td>
<td>0.88</td>
<td>2.3E-60</td>
<td>20.6</td>
<td>0.05</td>
</tr>
<tr>
<td>500</td>
<td>185</td>
<td>195375</td>
<td>889.5</td>
<td>205690</td>
<td>-122</td>
<td>1.1</td>
<td>1.77E-59</td>
<td>20.2</td>
<td>0.05</td>
</tr>
</tbody>
</table>

Figure 8.2. Identification of $C$ and $\gamma$ for WM at 400 °C and 500 °C.
Once the WM and BM NLKH parameters have been identified and validated using the single material models, a CW FE model was employed to simulate the stress-strain response of the CW tests. In the first instance the material constants $C$ and $\gamma$ of the HAZ are assumed to equal those of the BM. The constant $k$ was taken to be a value $\sim 4\%$ lower than the corresponding value of $k$ for the BM. This was based on the observation that the results of Figure 3.45 show that hardness within the HAZ is $4\%$ lower than that of the BM. The Young’s modulus of the HAZ was determined using a three-spring model with the three in-series springs representing the three material zones, as follows:

$$k_T x_T = k_{BM} x_{BM} + k_{HAZ} x_{HAZ} + k_{WM} x_{WM}$$

(8.1)

where $k_T, k_{BM}, k_{HAZ}, k_{WM}$ are the material stiffnesses and $x_T, x_{BM}, x_{HAZ}, x_{WM}$ are the lengths of each material zone. With the estimated constants $E$ and $k$, the global strain-controlled response of the CW FE model was compared with the half-life test data to iteratively identify $C$ and $\gamma$. Figure 8.4 depicts a comparison between the resulting NLKH response for the CW FE model and the corresponding CW test for two strain ranges at a temperature of $500^\circ C$. 

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Figure 8.3. Comparison of half-life experimental and NLKH WM responses at 500 °C.

Figure 8.4. Comparison of half-life experimental and NLKH CW responses at 500 °C.
8.2.3 Step 2 - Identification of isotropic hardening parameters

The cyclic softening curves presented in Figure 3.37 illustrate substantial softening behaviour occurring during testing of the WM specimens. The constant $Q$ was taken to be the stress difference between the first initial cycle and that taken at the half-life. The constant $b$ is determined by fitting Equation 7.1 to the test data, as shown in Figure 8.5. This figure shows the correlation for the BM and WM for a strain range of 0.8 % at 500 °C. Good agreement is observed between the model and the measured data from the initial cycle to the half-life, after which the model and the test behaviours begin to diverge. The identified isotropic material constants for the BM and WM are given in Table 7.1 and 8.1, respectively.

![Figure 8.5. Comparison of predicted isotropic and measured softening behaviour for WM and BM at 500 °C.](image)

The HAZ isotropic parameters were again (as for the NLKH parameters) obtained by iterative comparison of the CW FE model response and the measured
response, with the WM and BM values taken as input. The identified value of \( Q \) is significantly lower than that of the BM and WM, indicating that the HAZ does not cyclically soften as much as WM or BM. Figure 8.6 shows the comparison between the resulting CW model response (combined isotropic-NLKH) and the test data at \( N = 1 \). Figure 8.7 shows a comparison between the predicted cyclic softening behaviour of the combined isotropic-NLKH CW model with measured cyclic softening behaviour of the CW test at 500 °C at \( \Delta \varepsilon = 0.8 \% \).

Figure 8.6. Comparison of predicted (combined isotropic-NLKH) and measured CW response at 500 °C for \( N = 1 \).
8.2.4 Step 3 - Identification of creep constants

The creep constants $A$ and $n$ for the WM and BM were identified from the stress relaxation experimental data in the form of $\log(\dot{\sigma})$ versus $\log \sigma$, using the Norton equation for steady-state creep (Equation 2.3), which assumes elastic-creep material behaviour only. The identified constants were compared with the theoretical Norton creep relaxation response to ensure accuracy. The WM and BM constants were then utilised within the CW FE elastic-creep model under stress-relaxation conditions to facilitate identification of the HAZ creep constants. The resulting CW FE predicted stress relaxation behaviour is compared with the test data in Figure 8.8 and the identified creep constants are for the WM and HAZ are given in Tables 8.1 and 8.2, respectively, and the previously identified creep constants for the BM are given in Table 7.1.

Figure 8.7. Comparison of softening behaviour predicted by the two-layer viscoplasticity CW model and measured softening behaviour test at 500 °C at $\Delta \varepsilon = 0.8 \%$.
Figure 8.9 shows the validation of the resulting identified cyclic viscoplasticity parameters for the WM (Table 8.1), using the two-layer material model, for the first cycle at 500 °C. Figures 8.10 and 8.11 show the sample results for validation of the identified HAZ cyclic viscoplasticity parameters (Table 8.2), for \( N = 1 \) and at half-life against the measured data. The quality of comparison was equivalent for at 400 °C and for other strain-ranges.

A single material model utilising the identified HAZ modelling constants was employed to predict the stress-strain behaviour of the HAZ material. Figure 8.12 shows the measured stress-strain of the WM, BM and CW at \( N=1 \) compared against the predicted response of the single material HAZ model at 500 °C for \( \Delta \varepsilon = 0.8 \% \). The figure shows that the predicted response of the HAZ model is 31 % softer than the measured WM material response and 14 % softer than the BM. A sensitivity study on the effect of HAZ length in the FE CW specimen model established negligible effect on predicted stress-strain response of the CW specimen, varying the HAZ length between 2.5 mm and 3 mm, covering the approximate measured range of values.

![Stress relaxation behaviour of CW test](image)

Figure 8.8. Stress relaxation behaviour of CW test, compared against the stress relaxation response of the CW FE model at 500 °C.
Table 8.2. Cyclic material parameters for HAZ material.

<p>| | | | | | | | | |</p>
<table>
<thead>
<tr>
<th></th>
<th></th>
<th></th>
<th></th>
<th></th>
<th></th>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>$T$ (°C)</td>
<td>$k$ (MPa)</td>
<td>$E$ (MPa)</td>
<td>$\gamma$</td>
<td>$C$ (MPa)</td>
<td>$Q$ (MPa)</td>
<td>$b$</td>
<td>$A$ (MPa s$^{-1}$)</td>
<td>$n$</td>
</tr>
<tr>
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<td>173191</td>
<td>729</td>
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<td>500</td>
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<td>121076</td>
<td>-47</td>
<td>0.31</td>
<td>1.97E-71</td>
<td>25.56</td>
</tr>
</tbody>
</table>

Figure 8.9. Validation of cyclic viscoplasticity model for WM at $N = 1$ and at the half-life, at a temperature of 500 °C for $\Delta \varepsilon = 0.8\%$. 
Figure 8.10. Comparison of FE-predicted (cyclic viscoplasticity) and measured CW response at $N = 1$ at 500 °C for two strain ranges.

Figure 8.11. Comparison of FE-predicted (cyclic viscoplasticity) and measured CW response at half-life (softened) 500 °C.
Figure 8.12. Comparison between predicted HAZ stress-strain response and measured BM, WM and responses at $N = 1$.

8.3. Conclusions

The high temperature low cycle fatigue behaviour of P91 weld metal (WM) and weld joints (cross-weld) has been characterised using strain-controlled tests at 400 °C and 500 °C. The cyclic behaviour of the weld material was shown to be significantly harder and stiffer in terms of stress-strain response than both the base material and the cross-weld test specimens. The cross-weld tests exhibited a cyclic stress-strain response which was similar to that of the base material. All specimen types exhibited cyclic softening but the degree of softening exhibited by the CW specimens was lower than that of the base and all-weld tests. IC-HAZ (Type IV) cracking was the primary mode of failure for the cross-weld tests. Cyclic viscoplasticity modelling,
including isotropic softening and non-linear kinematic hardening, was employed to characterise the cyclic viscoplasticity behaviour of the WM and HAZ at 400 °C and 500 °C for finite element modelling and predictive response and failure analyses of high temperature welded power plant.
Chapter 9
CONCLUSIONS AND FUTURE WORK

9.1 General

During the investigation of P91 power plant components under thermomechanical loading conditions, the following key activities were undertaken:

1. A high temperature low cycle fatigue (HTLCF) rig was commissioned at NUI Galway and validated against the measured data taken on the thermomechanical fatigue rig at University of Nottingham.

2. A programme of HTLCF tests were carried out at University of Nottingham on service-aged P91 material across a ranges of temperatures, strain rates and strain ranges. HTLCF tests were also conducted to determine the effect of a tensile hold period during cyclic testing.

3. A series of tests were carried at NUI Galway to determine material behaviour of service-aged P91 material at room temperature. Stress-controlled tests were conducted to characterise the SA P91 base material under non-zero (tensile) mean stress conditions across a range of temperatures and stress rates.

4. Welded specimens were manufactured from a specially commissioned weld on a service-aged P91 header. HTLCF strain-controlled tests were carried out (at NUI Galway) on weld material (WM) and cross-weld (CW) specimens at 400 °C and 500 °C, across a variety of strain ranges, along with tension-hold cyclic strain-controlled HTLCF tests.

5. To elucidate the behaviour of P91 piping and header systems under realistic plant temperature and pressure conditions, a finite element methodology was
developed where (i) measured thermal plant data was utilised to calibrate a thermal finite element model, and (ii) a sequential thermomechanical analysis, utilising material constants identified from P91 experimental test data published the literature, was developed to simulate the behaviour of a plain pipe under simplified yet representative and realistic plant loading conditions.

6. The methodology developed to assess the behaviour of the P91 plain pipe was then applied to the more complex geometry of a multi-stub header section. Again a thermal FE model was developed and validated against the measured thermal data from an in-service ESB superheater header, the thermal model was taken as input to a sequential thermomechanical model. The same P91 material constants that were used in the plain pipe model were again utilised. In a similar approach to the plain pipe analysis a representative yet simplified version of the plant start-up thermal history was utilised to capture the effects of the salient thermal transients. In a separate model, the measured realistic thermal history was also taken as input to a sequential thermomechanical model. The contour plots from both the representative and realistic header analyses were examined with a view to identifying stress "hot-spots" at key regions of the header. Upon identification of these locations of high stress a submodel was developed which allowed for increased mesh refinement in these regions. The submodel used displacements from the global (header) model as an input. The predicted stress and strain histories of the submodels were used within a multiaxial critical plane fatigue life approach incorporating the Ostergren fatigue life model to predict number of cycles to fatigue crack initiation (FCI).
7. A rainflow cycle counting technique was incorporated into the multi-axial critical plane methodology, which takes account of these attemperoration driven minor stress and strain cycles. Their effect on the number of cycles to FCI was quantified using Miner's damage summation rule.

8. Material constants were identified from the service-aged P91 HTLCF tests and implemented with the realistic header analysis. The same methodology as before was followed, where the sequential thermomechanical analysis generated stress-inelastic strain histories at the locations of interest. Then using the multi-axial critical plane rainflow cycle counting technique the number of cycles to FCI were predicted.

9. Material constants were identified from P91 weld and HAZ material via for two temperatures.

9.2 Experimental testing and material characterisation

9.2.1 Strain controlled testing of service-aged P91 base material

For the strain controlled tests the range of temperatures investigated include; 20 °C, 400 °C, 500 °C and 600 °C. For all loading and temperature conditions the service-aged material exhibited continuous cyclic material softening. Comparisons between the cyclic stress-strain response of the service-aged P91 material and that of a comparative P91 are made. The results show the service-aged material to give a slightly hardened cyclic stress-strain response compared with that of the comparative P91 material. It was also found that this effect diminished following cycling loading. Ratchetting behaviour is heavily dependent on temperature and ratchet strain rate is
sensitive to stress rate. The TMF tests indicate that the effect of positive mean stress is detrimental to fatigue life.

The fatigue life evaluation using the Ostergren fatigue model suggests that the observed fatigue life of the service-aged material is similar to that of comparative P91 material from the literature.

9.2.2 Testing of P91 welded material.
The weld material exhibits a larger cyclic stress range compared with that of the cyclic stress response of base material test under the same loading conditions. The cyclic behaviour of the weld material was found to be significantly harder and stiffer in terms of stress-strain response than both the base material and the cross-weld test specimens. The stress-strain behaviour of the cross-weld tests exhibited a similar stress-strain response to that of the base material. The fatigue lives of the cross-weld tests is lower than their all-weld and base material counterparts, and the weld material exhibits a fatigue live which is greater than the cross-weld but lower than the base material.

9.2.3 Constitutive behaviour of P91 service-aged material
It is difficult to make comparisons between the constitutive behaviour of the comparative P91 material and service-aged materials due to the fact that the material constants calibration process is different. The comparative P91 material constants which were obtained from the literature, were calibrated for the Unified Chaboche model following the method outlined by Hyde et al., [2009], whereas the service-aged material constants were identified via a process detailed by Deshpande and co-
workers [2010] which applies to the two-layer viscoplasticity model. However for the service-aged material model, the plasticity modulus $C$ is similar to that of the comparative P91 material across all temperatures, the rate term $\gamma$, is on average lower at temperatures including and above 400 °C, but lower at room temperature. The Young's Modulus $E$ is effectively identical to that of the comparative P91 material, however the softening behaviour defined by $Q$ and $b$ is greater. $Q$ for the service-aged material is on average 30 % lower than that of the comparative P91 material (e.g. for 600 °C the service-aged value of $Q = -103.15$ MPa, whereas the comparative P91 material value $Q = -75.4$ MPa); however the rate term $b$ is greater for the comparative P91 material, implying a greater rate of softening. The user defined constant $f$, is greater than that for the comparative P91 material, for 400 °C to 600 °C cases which implies the cyclic stress-strain behaviour is more affected by the creep behaviour of the material. The result of this is evident in the stress-strain loops of Chapter 7 where the stress-strain response is adequately captured in terms of inelastic strain range and stress range, but the kinematic hardening behaviour is less so.

9.3 Thermomechanical simulation of P91 power plant components

9.3.1 Axisymmetric plain pipe
A two-layer viscoplasticity model has been successfully developed for P91 material based on material modelling constants taken from the literature. A finite element based, sequential thermomechanical methodology for pressurised steam pipes was developed, incorporating (i) a transient heat transfer method taking as input the steam temperature-time histories, and (ii) a multiaxial implementation of the P91 cyclic two-layer viscoplasticity material model, taking as input the pressure-time histories. Three different types of thermal histories were utilised to order to determine the most
detrimental types of thermal loading history. The first loading history was a simplified or representative type loading history which was designed in order to capture the characteristic behaviour of the plain pipe model under thermomechanical conditions. The resulting thermomechanical stress-strain response predicted compressive stresses at the inner bore during plant heating, which is associated with the start-up process and tensile stresses occurring during cooling transients. Two realistic cycles were also utilised; (i) recorded data for a plant cold-start, and, (ii) a load following history which was inclusive of a plant trip. The cold-start cycle produced a total inelastic strain of 0.03 %, whereas the simplified cycle predictions revealed cyclic inelastic strains of 0.038 % and 0.015% for the major and minor cycles respectively, with compressive (hoop) creep strains resulting from inside pipe surface heat-up and tensile (hoop) stresses and inelastic strains being driven by inside surface cool-down. A large number of smaller (tensile) stress cycles resulted from thermal differentials associated with the attemperation process for the realistic cold-start cycle. The load-following cycle predicts a large tensile mean stress of 104 MPa and an inelastic stain range 0.042 %. This finding indicates that a plant trip is the most damaging occurrence within the two realistic cycles.

9.3.2 Multi-stub header using comparative P91 material constants
The modelling of the multi-stub header model predicts large tensile stresses on the outer surface, weld toe region of the header during heating transients, and on the inside surface inner bore region during cooling transients. The inner bore crotch position of the header is the location where crack initiation is predicted to occur first. After 1,900 cold-start cycles a crack of 100 µm is predicted. The cracking directions predicted by the multi-axial critical plane methodology are consistent with cracking directions reported in the literature, where a crack initiation in the inner bore and
weld toe crotch positions will travel in the hoop direction of the header. The crack
direction at the inner bore and weld toe saddle positions is predicted to be in the axial
direction of the header. The use of a representative cycle which has been designed to
represent the salient features of the realistic (cold-start) thermal history has been
found to give similar thermomechanical response to that of the (Cold-start) loading
history. The representative cycle predicts almost identical lives to that of the realistic
cycle, thus vindicating the use of such representative cycle in the design of TMF
testing and as a means of rapidly simulating particular loading events occurring in a
power plant.

The critical plane rainflow cycle counting technique predicts slightly lower
cycles to failure than the critical plane methodology at the weld toe region. This is
primarily due to the most damaging transient (point $H_1$) which produces the largest
tensile stress and inelastic strain range during the initial cycle; the remaining
attemperation cycles cause little inelastic deformation and therefore contribute little
to the damage accumulation in that region. The inner bore however which is mostly
affected by cooling transients (which predominantly occur during the attemperation
process) accumulates slightly more damage than the weld toe.

### 9.3.3 Multi-stub header using service-aged material constants

Simulation of the multi-stub header utilising the service-aged material constants
again produced a complex stress-strain response at the previously identified critical
locations. The direction of crack orientation remains the same as for the comparative
P91 material however there is some difference in terms of fatigue life. The SA FE
header analysis predicts longer fatigue lives at the critical locations compared with
those from the comparative P91 material FE header analysis of Chapter 6. The
predicted fatigue lives for the service-aged analysis predicts longer lives compared with the comparative P91 material, this was attributed to smaller inelastic strain ranges.

9.3.4 High temperature, low cycle fatigue characterisation of P91 weld and heat affected zone material

The high temperature low cycle fatigue behaviour of P91 weld metal (WM) and weld joints (cross-weld) has been characterised using strain-controlled tests at 400 °C and 500 °C. The cyclic behaviour of the weld material was shown to be significantly harder and stiffer in terms of stress-strain response than both the base material and the cross-weld test specimens. The cross-weld tests exhibited a cyclic stress-strain response which was similar to that of the base material. Cyclic viscoplasticity modelling was employed to characterise the cyclic viscoplasticity behaviour of the WM and HAZ at 400 °C and 500 °C.

The fatigue lives of the WM and CW specimens are significantly lower than those of the BM tests.

9.4 Future work

Future work should incorporate weld and HAZ regions in the multi-stub header model as reports from industry indicate that the weld-base material interface around the stub pipe-header junction to be a known location of failure. The current FE analysis of the multi-stub header (which treats the header as being entirely comprised of base material) predicts higher number of cycles to fatigue crack initiation at the weld-toe regions compared with those of the inner bore. The uniaxial strain-controlled HTLCF tests of the welded specimens have given lower fatigue lives, and
different cyclic stress-strain behaviour compared with the base material. To this end, the inclusion of a multi-material weld region in the header models has the potential to impact the number of cycles to fatigue crack initiation at the weld-toe in two ways, (i) the relatively hard (cyclic) stress-strain response of the weld material compared to the base material may give rise to stress singularities, (ii) the fatigue constants identified from the high temperature low cycle fatigue tests of the weld and cross-weld tests will lead to a reduced life to FCI. It is proposed that the material constants for these regions be determined in a similar manner to those presented in Chapter 8.

The effect of oxidation on P91 material has been ignored in the present thesis, however the literature has shown that oxidation plays a key role in the fatigue life of the material. Sandhya et al [2010] have observed a dramatic increase in number of cycles to failure in HTLCF tests conducted in liquid sodium environment compared to those carried out in air. This was attributed to the oxidation-inhibiting nature of the liquid sodium, which retarded the onset of oxidation-induced crack initiation. Neu and Sehitoglu [1989a; 1989b] investigated the oxidation behaviour of carbon steel under TMF and creep conditions. It was reported that oxidation-initiated crack nucleation and propagation was a damaging mechanism during isothermal fatigue and TMF-OP conditions. It is therefore recommended that high temperature oxidation be explored with a view to including its effects in fatigue prediction models and methodologies.

It is also recommended that other constitutive material models with higher simulation capability like that of the unified Chaboche, hyperbolic sine or Ohno-Wang models should be investigated. The hyperbolic sine model in particular takes account of the key mechanisms of inelastic deformation allowing for greater accuracy across a wide range of strain rates and stress regimes. This model may be
particularly applicable to predicting the stress-strain behaviour of a material under realistic loading conditions, where low cooling and heating rates (thus giving rise to low strain rates) are known to occur. The commonly-utilised Fredrick-Armstrong kinematic hardening model has been shown (in the literature) to successfully capture the Baushinger effect during cyclic strain-controlled simulation. However, the Fredrick-Armstrong model and that of the similar Ziegler model have a tendency to over-predict both the uniaxial and multi-axial ratchetting behaviour of a material. The Ohno-Wang (II and III) constitutive models have been reported to predict the uniaxial and multiaxial ratchetting with better accuracy. In the current work, the realistic header analysis predicts significant tensile mean stresses at the locations of interest (i.e. inner bore saddle and crotch, and weld-toe crotch positions) and therefore future simulations of this geometry and applied loading conditions should utilise the more suitable Ohno-Wang hardening model.

Physically-based models have also received attention in recent years. Physically-based models take account the alterations to the material microstructure when subjected to load. In particular, these models have been applied using P91 material constants, which were identified by experimental methods to observe and quantify subgrain and lath coarsening and dislocations annihilation. These models have been found to capture the stress-strain response, cyclic softening behaviour and the microstructural changes with reasonable accuracy (see Fournier et al., [2011] and Sauzay, [2009] for further details). In addition the key advantage of these models is that they may be used to extrapolate with greater accuracy than the aforementioned phenomenological models.

Another alternative to the conventional phenomenological constitutive models is that of the crystal plasticity type. Crystal plasticity models have the ability to
capture localised stress fields that would not ordinarily be captured by phenomenological models. These models also have an enhanced ability to predict crack initiation sites and the orientation of crack growth, which is of particular use in predicting fatigue crack initiation in the bore hole of the header where cracks are known to occur.

The effects of system loading on the header should also be considered. The present work focused on a localised header section of a header geometry subjected to an internal steam pressure and neglects the fact that this geometry is only a section of a much larger header, (over 12 m in length), which contains many of rows of stub pipes. The effect of these pipes may subject the header to additional torsional loads. Real headers are also known to bend due to thermal gradients occurring during operation. Therefore it is suggested that future work take account of system loading on TMF behaviour which would reflect the movement of the entire header section.
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APPENDIX A

NUIG HTLCF Test Rig - Experimental set-up

A.1. General

The following appendix describes the experimental procedure to carry out a HTLCF test on the NUIG test rig.

NB. all activities carried out during specimen installation and during the pre-test heat-up operation are carried out in the displacement channel setting in FT Console software and with 'Specimen protect' function turned on. Any deviation from the aforementioned software settings will be explicitly stated in the proceeding set-up methodology.

A.2. Initial start-up of test equipment

Switch on hydraulic ring main, set the system pressure to 206 bar and the flow rate to 20 l/s, on the wall mounted ring main user interface (Note the flow rate may need to be increased depending on the use of other lab equipment). Switch on Instron controller. Start-up PC and open FT Console software. By default, the software will start-up in the displacement channel with 'Specimen protect' turned on. Move into 'Medium pressure' mode by pressing the 'I' button on the console located adjacent to the hand held position controller, then select 'High pressure' mode by pressing the 'II' button, on the same console. If necessary move the cross-head (bottom surface) to a distance 1150 mm above the top of the platen. This is done by loosening the clamp knob (located on the front vertical surface of the platen), and then, while the clamp...
knob is fully is open, opening either the raise or lower knob to achieve the correct height. Then close the raise/lower knob and close the clamp knob.

A.2.1 Load string pre-tension

1. Screw on pull-rod stud, insert bearing plate and wedge washers, and screw on pull-rod (Note: the pull-rod with the graduated dial around its base is the upper pull-rod). The arrangement should be the same as that shown in Figure A.1. During screwing-on of the pull-rod, ensure the following: (i) the wedge washers are extended horizontally to ensure the net thickness of the washers is as narrow as possible and, (ii) the pull-rod studs do not loosen during the screwing-on process. Perform the same procedure for the upper pull-rod.

2. Insert the pre-tension specimen and tighten into pull-rods using the specimen nuts and pull-rod nuts (annotated in Figure A.1), set an upper limit of 85 kN in the load channel, turn off 'Specimen protect'. Then extend the load string (in tension) using the hand held controller in the displacement channel to 84 kN (Warning: Do not exceed this load value).

3. Move the wedge washers horizontally in order to increase the combined thickness of the wedge washers, this allows removal of any remaining slack in the load string while in tension. Tighten the locking bolt to the side of the wedge washer, and tighten down the locking nut.

4. Repeat step 3 for the upper pull-rod.

5. Un-load the load string and turn on 'Specimen protect', loosen the pull-rod nut on both pull-rods and remove the pre-tension specimen.

Both pull-rods should now be tight and should not loosen.
Figure A.1. Illustration of load string assembly on NUIG test rig [Instron, 2012].
A.2.2 Alignment

Alignment of the load string must be carried out each time the pull rods are installed.

1. Slacken the six M10 fixing nuts on the underside of the upper pull-rod (see Figure A.1). Adjust the angularity adjustment scales (located to the side base of the upper pull-rod) so that the upper and lower scales align at zero in the forward position. Re-tighten the six M10 fixing bolts to a torque of 20 Nm, using a torque wrench.

2. Attach a dial gauge to the bottom pull rod. Position the pointer of the dial gauge on the surface of the upper pull-rod, as shown in Figure A.2. Zero the dial gauge.

3. Using the hand held controller move the lower pull-rod (down-wards) by a distance 50 mm. Record the displacement (reading from the dial gauge) and the direction of the displacement (taken as being either; a movement away from or toward the user). Note displacement magnitudes should then be multiplied by a factor of 2 to convert displacement/50 mm to displacement/100 mm. This is due to the fact that travel in the actuator is approximately 50 mm, and the error must be read in mm/100mm

4. Repeat three times to ensure repeatability.

5. Rotate the dial gauge through 90° about the axis of the pull-rod, and repeat steps 2 to 4, to determine misalignment on the new plane.

6. By measuring the displacement magnitude and direction on two different plane it is now possible to represent the misalignment between the two pull-rods graphically, as shown in the example case of Figure A.3. This figure indicates a misalignment displacement of 0.04 mm (the reading taken on the
front face of the pull-rod) over a 100 mm distance. When checked on a plane 90° to the front face it was found that the a misalignment error of 0.05 mm. The combination of the misalignment in both directions produces a total misalignment of 0.07 mm. Figure A.3 also indicates the direction of the misalignment. In this case the bottom of the pull-rod is in a direction 36 ° relative to the left-right axis. Refer to drawing F3117-011 in the Instron manual for further details if necessary.

7. Calculate the angle of misalignment and magnitude of misalignment.

8. To correct the calculated misalignment;
   - Slacken the six M10 bolts on the under-side of the upper pull-rod and rotate the alignment adjusters (still set at zero/zero) so that they lie in the direction of the calculated angle of misalignment (in this case 36°).
   - Holding the upper ring in position, rotate the lower ring until the calculated value (0.7 mm/m) aligns with the zero on the upper ring.
   - Now holding the lower ring, rotate the upper ring until the calculated value (0.7 mm/m) aligns with the same value on the lower ring.
   - Re-tighten the M10 bolts to 20 Nm.

9. Repeat steps 1-7 to ensure alignment error is less than 0.2 mm/m.
Figure A.2. Load string alignment configuration [Instron, 2012]
A.2.2.3 Specimen installation

To insert the specimen in the test rig:

1. Screw specimen nut onto specimen ensuring bottom of the specimen protrudes beyond the guide of the inner nut (This creates to 2 -3 mm preloading gap detailed in Figure A.1). Screw on the pull-rod nut, then slide the other pull-rod nut (facing in the opposite direction) over the other end of the specimen. Screw on the other specimen nut. This has created the specimen-nut assembly.

2. Insert specimen-nut assembly into lower pull-rod first, ensuring the base of the specimen is in contact with the bottom of the recess of the pull-rod and the guide are not taking any load.

3. Hand tighten down the lower outer nut.

4. Move the actuator up so that the flat top of the specimen comes in contact with flat portion of the recess on the pull-rod. A small compressive load of no more than 0.3 kN should be applied.

5. Hand tighten the upper pull-rod nut.
6. Tighten top pull-rod nut with torque wrench to 120 Nm, ensuring that the pull rod is also held. Repeat the process for bottom pull rod.

7. Set and arm displacement limits: set to ±1.2 mm absolute movement.

8. Set and arm load limits: typically ±25 kN (this value is dependent on material stiffness, applied strain range, test temperature and associated strain rate effects).

**A.2.2.4 Calibrating and attaching extensometer**

To calibrate the extensometer:

1. Plug in extensometer. Allow 30 mins to elapse before calibrating the extensometer.
2. Position extensometer legs against the gauge setting tool (see Figure A.4) to separate the extensometer legs to the correct distance. Holding the tool in place, press the “Start” (calibration) button within the calibration wizard.

To attach the extensometer

1. Starting with the upper attachment cord. Insert and tighten one end of the cord into the upper hole on the extensometer. Loop the un-attached end of the cord around the back of the specimen and insert the cord into the other upper hole ensuring that the spring loaded aluminium connection (which is at the ends of each cord) is pulled back with sufficient force to ensure the upper portion of the extensometer is held in place, then tighten down the grub screw.
2. Repeat step 1 for the lower cord. The attached extensometer should appear as it does in Figure A.5.
3. Check alignment of the extensometer by sight to ensure that legs of the extensometer are sitting on the gauge length of the specimen and that the legs are in line with the longitudinal axis of the specimen.

4. Adjust the width of the legs while the extensometer is mounted on the specimen so that the strain output value is returned to that observed when using the gauge length setting tool; the strain reading can be read directly off the live display on the desktop. Re-check to ensure that the extensometer legs are still aligned along the longitudinal axis gauge length.

Figure A.4. Positioning of extensometer legs against the gauge length setting tool, before subsequent calibration and balancing of extensometer [Instron, 2012]
Figure A.5. Attached extensometer and test specimen arrangement [Instron, 2012].

A.2.2.5 Positioning and start-up of the furnace

It is important that the test specimen be positioned in the middle of the working region of the furnace to ensure a uniform temperature along the axial direction of the specimen.

1. Swing in the furnace and check the height to ensure that the extensometer in unimpeded. To adjust the height:
   - Lock the main movement handles. Undo side grub screws on support collar. Move the collar to desired height and retighten collar. There are also bolts on the collar which (when the collar is locked in position) allows the loose support section of the furnace to be forced upward.
• When the desired height is achieved lock the two movement handles. Close one half of the furnace around the load string and lock in position.

2. Insert required rock wool on the furnace, ensuring that the extensometer is covered.

3. Swing the other half of furnace into position.

4. Loosely pack more insulating wool around the extensometer if necessary, pack rock wool in gaps in the furnace as necessary.

5. Close furnace clasps.

6. Switch on the cooling fan.

7. Ensure all wires and hoses are away from the furnace surfaces.

8. Turn on cooling water and measure flow rate.

9. Set the target temperature and press furnace heating button.

Note: In order to achieve the desired specimen temperature, a thermal calibration exercise was undertaken whereby a base material P91 specimen was drilled to accept a number of high temperature K type stainless steel sheathed thermo-couples, which are rated to 1100 °C. The specimen was then installed as per the usual procedure (outlined above) with the thermo-couples inserted in the holes in the specimen. Different temperatures were set at the controller and the temperature was allowed to stabilise. The stabilisation soak times varied due to the fact that the PID control system was found to produce a stepped application of heating. The soak times required are detailed in Table A.1, along with the associated target test temperatures and controller temperature reading. Note that the set point temperatures
required in order to obtain the target temperature at the specimen are significantly higher than that of set point.

Table A.1. Temperature setting required to achieve desired temperature at the specimen.

<table>
<thead>
<tr>
<th>Target Temp. (°C)</th>
<th>Controller temperature reading (°C)</th>
<th>Thermocouple reading (°C)</th>
<th>Soak time (hrs)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Zone 1 (Top)</td>
<td>Zone 2 (Middle)</td>
<td>Zone 3 (Bottom)</td>
</tr>
<tr>
<td>200</td>
<td>276</td>
<td>276</td>
<td>276</td>
</tr>
<tr>
<td>300</td>
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<td>394</td>
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<tr>
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<td>715</td>
<td>715</td>
<td>715</td>
</tr>
</tbody>
</table>

To run a test, first open the LCF3 software. In the test methodology section, enter specimen and test parameters i.e. cross-sectional area of the gage length, frequency of data logging, number of points to be recorded per stress-strain cycle, preliminary Young's modulus, waveform, strain range, strain rate, specimen failure criterion due to a specified drop in load from a certain point in the test. Once this data has been supplied, save and store this methodology and move into the 'Test' section of the software. Select the methodology which has been created, then start test. Figure A.6 depicts a screen-shot from a typical HTLCF test for service-aged P91 material after 339 cycles. The softening effects are clearly visible on the lower right plot. After 339 cycles the tensile stress has softened by 38 MPa.
Figure A.6. Screen shot of LCF data acquisition software.
Appendix B

A cyclic viscoplasticity model for service-aged P91 base material

B.1 General
The kinematic hardening models proposed by Fredrick-Armstrong and Ziegler tend to over-predict ratchet strain [Abdel-Karim and Ohno, 2000; Chen et al., 2003].

The present section investigates the hardening model proposed by Ohno and Wang (model II). This hardening model is implemented within the framework of the unified Chaboche viscoplasticity model, in MATLAB code. This constitutive model will hereafter be referred to as the Ohno-Wang (OW) model. The service-aged P91 base material test data at 500 °C and 600 °C is utilised to identify material modelling constants. The modelling results are validated against the cyclic strain-controlled test data at 500 °C and 600 °C, and ratchetting test data carried out at 550 °C. The model performance is compared with that of the two-layer viscoplasticity model which utilises a Ziegler hardening law. This model will hereafter be referred to as the two-layer model.

Figure B.1a shows a micrograph of the as-received SA P91 based material. The microscopy specimens were mounted and polished, and then etched using Villella’s reagent. Examination of the microstructure revealed a number of creep voids (highlighted by the red arrows). The formation of these voids most probably occurred during the materials prior period of service, during which time the material was exposed elevated service temperatures and pressures. The scanning electron beam micrograph of Figure B.1b shows a detailed view of the SA P91 base materials microstructure. Precipitates are visible along the edges of the grain boundaries. Figure B.1c is a magnified view of the highlighted region of Figure B.1b. The figure
shows (in the region bounded by the red box) a grain with a chain of precipitates along its boundary.
Figure B.1. (a) Optical micrograph of as-received SA P91 microstructure, (b) Scanning electron microscope image of the as-received SA P91 base material microstructure, (c) magnified view of Figure B.1b, showing a chain of precipitates along the edge of a grain boundary.

B.2 Constitutive material modelling

B.2.1 Hardening model
As previously mentioned, the Ohno and Wang [1993] non-linear kinematic hardening model (models II) is incorporated within the unified Chaboche viscoplasticity model. The unified Chaboche model is described in Chapter 3, Section 2.5.8. The OW hardening model is described as follows:

The backstress is assumed to be composed of $N$ number of components:

$$\chi = \sum_{i=1}^{N} \chi_i$$  \hspace{1cm} (B.1)
It is assumed that the dynamic recovery of $\chi_i$ becomes significantly non-linear as $\chi_i$ approaches the surface $f_i = 0$

$$\dot{\chi}_i = \xi_i \left[ \frac{2}{3} r_i \dot{\varepsilon}_m - \left\{ \frac{J(\chi_i)}{r_i} \right\} \left( \dot{\varepsilon}_m + \frac{\chi_i}{J(\chi_i)} \right) \right]$$  \hspace{1cm} (B.2)

The McCauley bracket $\langle \cdot \rangle$ operates as $\langle x \rangle = 0$ when $x < 0$ and $\langle x \rangle = x$ when $x \geq 0$.

$$f_i = \left\{ J(\chi_i) \right\}^2 - r_i^2$$  \hspace{1cm} (B.3)

where $\zeta_i(i = 1, 2)$, $r_i(i = 1, 2)$ and $c$ are material constants.

**B 2.2 Determination of material parameters**

The material parameter identification methodology proposed by Abdel-Karim and Ohno [2000] is employed here to identify the kinematic hardening constants, $r$ and $\zeta$, for the SA P91 base material at 500 °C and 600 °C. The parameters are identified from the tensile curve of the first quarter cycle, of the hysteresis loop for a constant strain rate. The evolution of the backstress is assumed to be comprised of multi-linear parts which have corners [Jiang and Sehitoglu, 1996]. Thus, the kinematic hardening parameters $r$ and $\zeta$, can be related to the coordinates of these corners, $\alpha(i)$ and $\varepsilon_p(i)$, (as represented by Figure B.2), using the following equations:

$$\zeta_i = \frac{1}{\varepsilon_p(i)}$$  \hspace{1cm} (B.4)
\[ r_i = \left( \frac{\alpha_{(i)} - \alpha_{(i-1)}}{\varepsilon_{p(i)} - \varepsilon_{p(i-1)}} - \frac{\alpha_{(i+1)} - \alpha_{(i)}}{\varepsilon_{p(i+1)} - \varepsilon_{p(i)}} \right) \varepsilon_{p(i)} \]  

(B.5)

where \( \alpha(0) \) and \( \varepsilon_{p(0)} = 0 \).

The parameter \( r_2 \) were then iteratively adjusted to obtain an improved fit against the measured data.

![Graph showing the change of backstress under tensile loading.](image)

Figure B.2. Change of backstress under tensile loading. (adapted from Abdel-Karim and Ohno, [2000]).

The same calibration methodology proposed by Hyde et al., [2010] and Saad et al., [2012] is adopted here to identify the creep constants. Figure B.3 shows the identification of creep constants \( Z \) and \( n \), at 600 °C.
Figure B.3. Determination of creep constant $Z$ and $n$ for SA P91 base material at 600 °C.

The isotropic hardening constants $Q$ and $b$ are the same as those presented in Chapter 7 for the various temperatures. Table B.1 presents the identified material constants for the OW model at 500 °C and 600 °C.

Table B.1. Identified material constants for OW model, at 500 °C and 600 °C.

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>$k$ (MPa)</th>
<th>$Z$ (MPa s$^{1/n}$)</th>
<th>$n$</th>
<th>$r_1$</th>
<th>$\zeta_1$ (MPa)</th>
<th>$r_2$</th>
<th>$\zeta_2$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>500</td>
<td>185</td>
<td>1975</td>
<td>2.47</td>
<td>75</td>
<td>2035</td>
<td>79</td>
<td>659</td>
</tr>
<tr>
<td>600</td>
<td>155</td>
<td>1892</td>
<td>2.56</td>
<td>55</td>
<td>2000</td>
<td>57.7</td>
<td>485.5</td>
</tr>
</tbody>
</table>

A.3 Results and validation

Figures B.4 (a, c and e) shows the validation of the OW model at 500 °C at $N=1$, at a constant strain rate of 0.033 %/s, for various strain ranges. Figure B.4 (b, d, and f) shows a comparison between the measured experimental data at the half-life with that of the material model for the corresponding number of cycles, at 500 °C.
Figure B.5 (a – e) shows a comparison between the predicted stress-strain response of the OW model, and the corresponding experimental test data at 600 °C. Figure B.3 (a, c and e) compares the predicted stress-strain response to the experimental data at $N = 1$, and Figure B.3 (b, d, and f) compares the cyclic responses at the half-life.

For both temperatures the predicted stress-strain response at the half-life exhibits a slightly hardener response compared with the experimental data. The half-life prediction, over the non-linear portion of the curve does not accurately fit with the experimental data. The addition of additional backstress terms to the model, may improve the prediction.
Figure B.4. Validation of the OW model for various strain ranges at (i) $N = 1$ (Figure B.4 (a, c and e)), and (ii) at the half-life (Figure B.4 (b, d and f)), at 500°C, for a constant strain rate of 0.033 %/s.
Figure B.5. Validation of the OW model for various strain ranges at (i) \(N = 1\) (Figure A5 (a, c and e)), and (ii) at the half-life (Figure B.5 (b, d and f)), at 600°C, for a constant strain rate of 0.033 %/s.
Figure B.6 shows the OW models ability to capture strain rate effect. Figure B.6 (a and b) shows the predicted stress-strain response of the model at 500 °C and 600 °C, respectively, at a constant strain rate of 0.025 %/s, compared with the measured experimental data at the same temperatures and strain rate. Figure B.6 (c and d) again shows predicted stress-strain response of the model at 500 °C and 600 °C, respectively, but at a higher strain rate of 0.1 %/s. The model is shown to exhibit adequate strain rate sensitivity, albeit for limited data.

Figure B.6. Comparison of OW model against the measured experimental data at \( N = 1 \) at (i) 500 °C and 600 °C at a constant strain rate of 0.025 %/s, and (ii) at 500 °C and 600 °C at a constant strain rate of 0.1 %/s.
Figure B.7a shows a comparison between the predicted stress-total strain response OW model and two-layer model compared with the measured experimental data at 500 °C at a constant strain rate of 0.033 %/s. Figure B.7b shows the OW and two-layer models ability to capture stress-inelastic strain behaviour. The stress-strain responses predicted by both models give good agreement with the measured data in terms of stress-total strain (Figure B.7a) and stress-inelastic strain (Figure B.7b) behaviour.

![Figure B.7a](image1.png)  
![Figure B.7b](image2.png)

Figure B.7. Predicted response of the OW and two-layer models compared with the measured experimental data for (i) stress-total strain response and (ii) stress-plastic strain response, at $N = 1$ for a temperature of 500 °C.

Figure B.8 shows the ratchet strain predicted by the OW model at a stress rate of 5 MPa/s and a R-ratio = -0.5, with a maximum applied stress of 400 MPa, at a temperature of 550 °C. The figure shows the models response the measured test data and the predicted response of the two-layer model. Figure B.9 shows a similar
comparison for the higher stress rate of 50 MPa/s. It is clear from Figures B.8 and B.9 the OW model gives improved accuracy compared with the two-layer model, however in both cases the OW model overestimates the ratchet strain.

Figure B.8. Evolution of ratchetting strain predicted by the two-layer model, and the OW model (II), at a stress rate of 5 MPa/s, compared with the measured experimental data at 550 °C.

Figure B.9. Evolution of ratchetting strain predicted by the two-layer model, and the OW model (II), at a stress rate of 50 MPa/s, compared with the measured experimental data at 550 °C.
B.4 Conclusions

The developed constitutive material model for cyclic viscoplasticity is presented for high temperature cycle fatigue behaviour of SA P91 base material. For the conditions examined, the key conclusions are as follows:

- The OW model is found to adequately capture the cyclic strain-controlled stress-strain response for various strain ranges. It is also found to capture the softening behaviour of the material with acceptable accuracy.

- The OW model was found to predict the strain rate behaviour adequately, for both temperatures.

- The OW model exhibited improved predictions for the ratchetting behaviour compared with the predicted response two-layer (Ziegler hardening) model. The OW model tends to overestimate the ratchetting strain for both stress rates.