

DAMAGE MODELLING OF LEADED FREE CUTTING STEEL UNDER HOT FORMING CONDITIONS

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ABSTRACT

In this thesis the influence of stress-state on ductile damage in free cutting steel under hot forming conditions is examined. The industrial motivation for the project focuses on edge cracking in hot rolling. A brief outline of the hot rolling process conditions is necessary to define the important parameters affecting edge cracking. Triaxiality was, thus, identified as the key parameter relating to damage under hot rolling conditions. Based on a detailed literature review, the appropriate testing and modelling methodology were identified for this body of work.

A high temperature, uniaxial tension test program was implemented to identify the effect of triaxiality on damage under hot forming conditions. Double notched bars with varying notch radii were utilised, thus inducing different stress triaxialities due to geometrical constraints. Based on the resultant stress-strain data the effect of triaxiality on ductility and the strain to failure was investigated. Subsequently, unbroken notches from tested double notched samples were sectioned and optically examined to reveal damage initiation sites. Interesting damage features were identified and correlated with sample geometry (i.e. triaxiality) and testing conditions.

Finite element analysis of the double notched samples revealed the effect of triaxiality on the local stress-state. The accuracy of the mechanical analysis from such simulations was improved by incorporating the thermal gradients induced during high temperature Gleeble tests. Three stress parameters were examined in relation to their effect on the experimentally observed damage; maximum principal stress, effective stress and hydrostatic stress. The maximum principal stress and equivalent stress were most clearly correlated to damage development under multiaxial conditions for this particular free cutting steel.

Based on the results of the stress-state investigation of the double notched samples, a multiaxial damage expression was developed that reproduced the experimentally observed damage characteristics. The new multiaxial damage model was calibrated using a combination of uniaxial and multiaxial stress-strain data and damage profiles. The model was shown to have good accuracy in predicting both the stress-strain data and the damage initiation sites as a function of geometry and damage conditions. Finally, an extensive range of temperature and strain rate conditions were simulated for all tested sample geometries, and an additional sample geometry, to fully understand how testing conditions affect damage characteristics and under what triaxialities this is prone to happen.

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NOMENCLATURE

$T_{rolling}$	Temperature of rolling (°C)
T_m	Material melting point (°C)
T	Temperature (°C)
T_0	Reference temperature (°C)
t_f	Time of failure (sec)
σ	Stress (MPa)
σ_1	Maximum principal stress (MPa)
σ_e	Equivalent stress (MPa)
σ_H	Hydrostatic stress (MPa)
σ_H / σ_e	Triaxiality
σ_A	Net stress (MPa)
ω	Damage
ω^*	Damage under multiaxial conditions
ω_0	Initial damage value
ω_{cr}	Critical damage value
ω_{gb}	Grain boundary damage
ω_{pi}	Plasticity induced damage

E	Young's modulus (MPa)
ε	Total uniaxial strain
ε_p	Plastic uniaxial strain
ε_e	Elastic uniaxial strain
ε_f	Strain to failure under uniaxial conditions
ε_f^*	Strain to failure under multiaxial conditions
$z_{\#}$	Material constants
$\alpha_{\#}$	Material constants
p	Hydrostatic pressure (MPa)
ρ	Absolute averaged dislocation density
$\bar{\rho}$	Normalised, average dislocation density
k	Initial yield stress (MPa)
K	Drag stress (MPa)
K_1	Material constant (1/ °C)
R	Isotropic strain hardening (MPa)
Q_x	Activation energy (J/mol)
β	Material coefficient
κ	Gas constant (J/(mol*K))

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CHAPTER 1

INTRODUCTION

SUMMARY

In this chapter the importance of steel manufacturing, particularly hot rolling, is outlined. Basic concepts and problems encountered in the hot rolling route are presented. The aims and objectives for this project are defined. Finally, the thesis structure and a breakdown of its content by chapter are detailed.

1.1 STEEL MANUFACTURING

Steel is one of the most versatile materials in the world and is integral in our everyday life. Steel is used in every single aspect of our lives, from transport and infrastructure to the consumer goods industry with products such as cans and tins that preserve our food and drinks (see Figure 1.1). Steel is also the most recycled material in the world with recovery rates of more than 70% in addition to by-products from steel manufacturing that can be reused (W. S. Association, 2012).

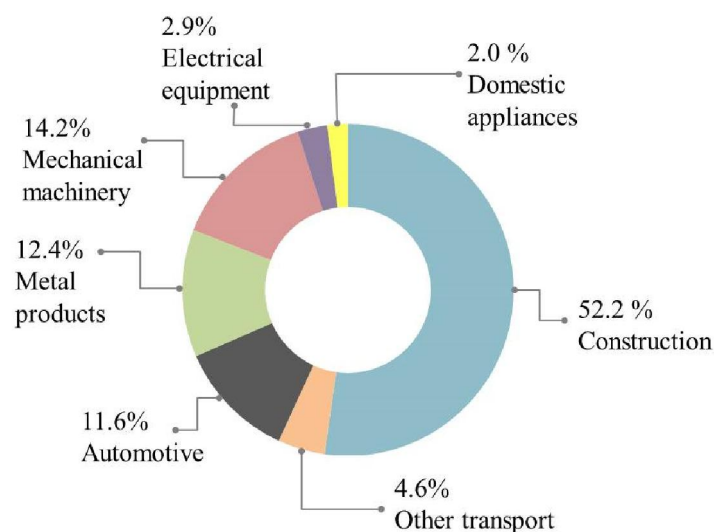


Figure 1.1: Breakdown of fractional steel consumption by industry in 2013 (W. Association, 2014a).

Demand for steel has always been high due to its significance in our lives and it is also predicted to play an important role in the future as the demands of modern life increase, particularly in developing countries. World crude steel production has already increased significantly in the past decade (see Figure 1.2). It is predicted that by 2050 70% of the world's population will become urban (W. S. Association, 2012). The demands from the housing and construction industries will have to be met by increasing steel manufacturing. For this reason steel manufacturers have to keep improving the manufacturing routes to increase production with minimal wastage.

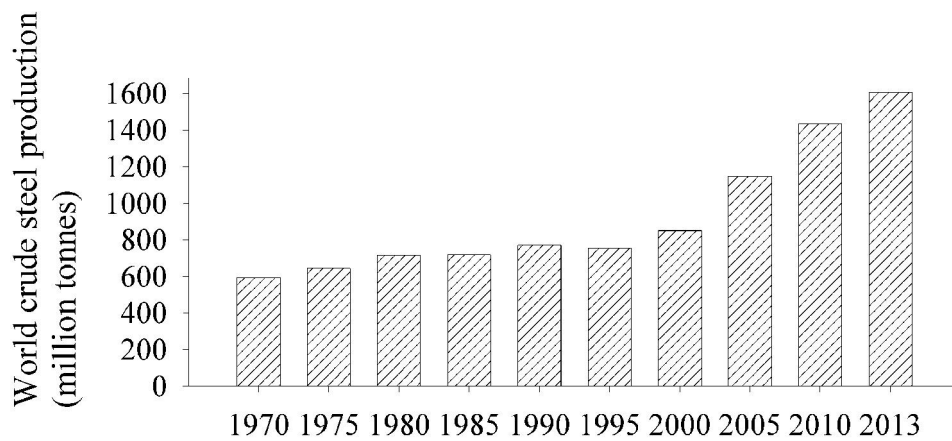


Figure 1.2: World crude steel production of steel from 1970 to 2013 (W. Association, 2014b).

There are multiple manufacturing routes for steel such as forging, extrusion, drawing and hot rolling. Hot rolling accounts for more than any other steel manufacturing process in terms of tonnage and is one of the oldest and most important processes in the steel industry (Roberts, 1983). In the following sections the basic concepts of hot rolling are outlined.

1.2 HOT ROLLING FUNDAMENTALS

Hot rolling is characterised by the presence of recrystallisation during deformation. For this reason the deformation temperature has to be above recrystallisation temperature, approximated by:

$$T_{rolling} = 0.4 \times T_m \quad 1.1$$

for steels (Javoronkov and Chaturvedi, 1981). For this project, the steel of interest is continuously cast into square billets of approximately 3 tonnes, 13 meters in length from an Electric Arc

Furnace (EAF) route which is based on 100% scrap and mainly cold charged. During the melting process provided by the EAF electrode/electric current, the charge is melted and ferrous alloys are added. Oxygen is also blown in with other additions such as limestone to form a slag. Sampling is performed to assess the composition and then the furnace is tilted in both directions to remove slag and pour steel into a ladle which then undergoes secondary steelmaking for further tuning of the chemical composition or performing vacuum degassing or powder injection. In an EAF route, a typical charge of iron is 150 tonnes and is converted to steel in a maximum of two hours.

The ladle is then transferred to the continuous casting machine where the liquid steel is poured in a tundish and cast at a controlled rate through careful solidification using copper moulds to form a continuous strand. The steel is drawn down into a series of rolls and water sprays which ensure that it is both rolled into shape and fully solidified at the same time. The final stage is to straighten and cut the product to the required length. Fully formed slabs, blooms and billets emerge from the end of this continuous process. These semi-finished products are then transported to the hot rolling mill bay area and are charged (cold/hot) to the primary reheating furnace (maximum temperature of 1250 °C) where they are submitted to a controlled heating profile using natural gas combustion.

During this process, oxidation takes place together with metallurgical phenomena such as grain coarsening and recrystallisation. For all steel grades produced through an EAF route, great care is taken to control residual elements such as copper (Cu) and tin (Sn). The as-cast billets are then further rolled into a continuous bar mill to various sections and bar sizes. This process is done in a single strand 24-stand continuous mill made of a compact rougher, intermediate, finishing and sizing stands. All products are fully tracked and 100% inspected. This process is outlined in Figure 1.3 below.

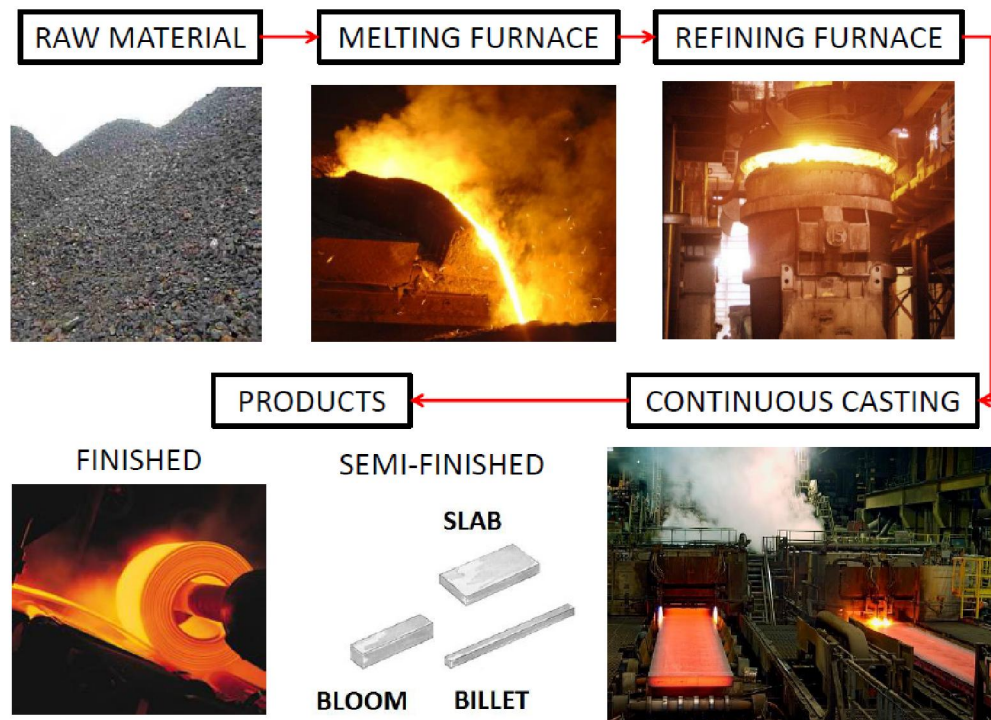


Figure 1.3: Schematic of the hot rolling process.

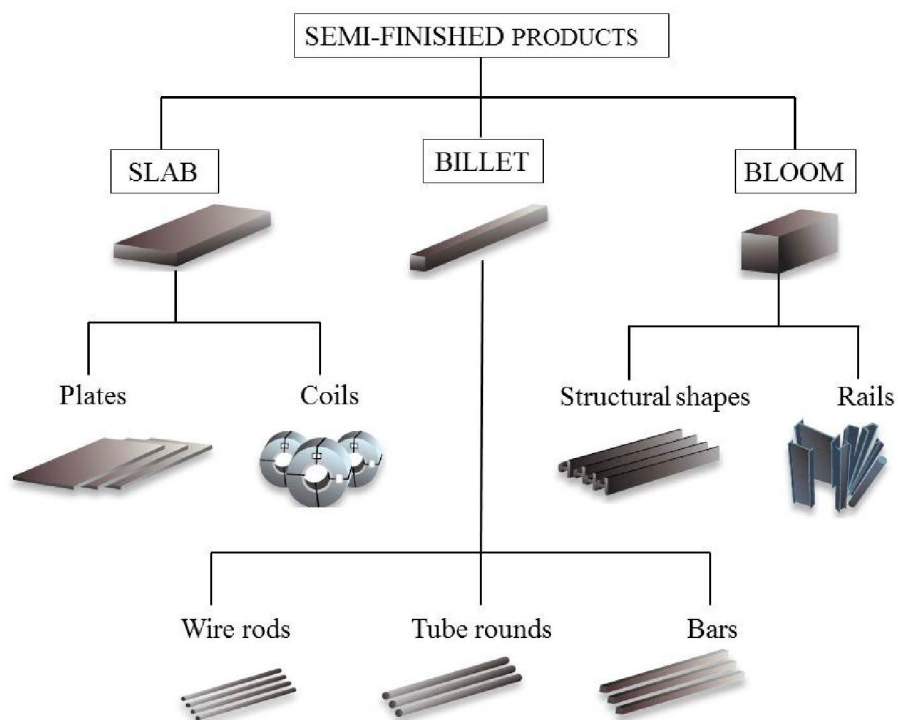


Figure 1.4: Categories of finished products delivered from slabs, blooms and billets.
Adapted from (W. Association, 2013).

Typically slabs would be shaped into flat products such as plates or strips (hot rolled coil) whilst blooms and billets would be shaped into long products like rods, bars and joists (see Figure 1.4). All of these hot rolled products can be further processed, for example via cold rolling or drawing to achieve the final product form. Hot rolled products are used abundantly in construction, transport and engineering (see Figure 1.5). Steel ropes, made of woven steel wires, are found in suspension bridges whilst steel plates or steel rods are used for structural support of the bridge 'floor'. Steel beams, columns and joists are used as supports in building foundations. Steel sheets can be used as wall claddings, false ceilings and roofs in modern buildings. Additionally steel housing frames, made from hot rolled beams and sheets, can provide an affordable and environmentally friendly solution to timber (W. Association, 2014a). Steel plates, sheets and sections are extensively used in ship hulls and superstructures. Cars also make use of steel sheets for the main body and steel sections for the sub-frame and safety cage, providing protection. These represent just a few examples of the many uses hot rolled steel products have in our daily lives.

Therefore, it becomes apparent how dependent we are on the use of steel in everyday life and by extension how important the hot rolling process really is to maintain quality of life. It follows that research into improving the product quality and efficiency, which would lead to cost reduction and improved environmental output, has a strong hold over global economies.

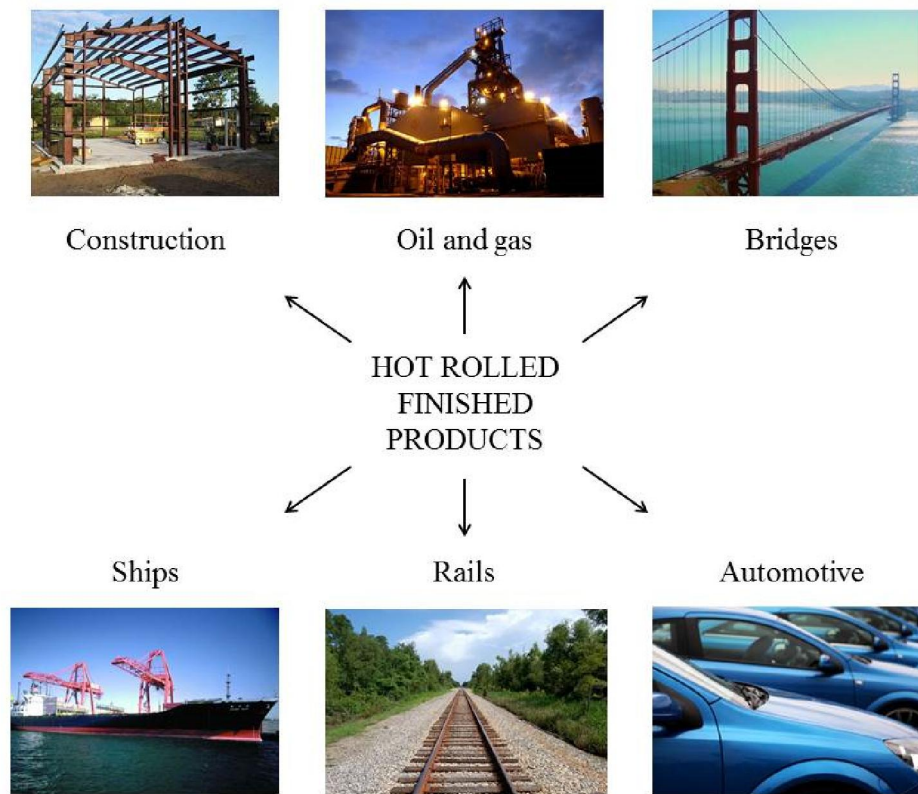


Figure 1.5: Examples of the applications of hot rolled finished products.

1.3 DEFECTS IN HOT ROLLED PRODUCTS

Despite hot rolling being a very popular and old metal forming method, defects are still frequently encountered during the processing route. Defects can arise from two main categories; defects from the cast ingot/continuous casting and defects from the rolling. The semi-finished rolled product has to go through the steelmaking, continuous casting and reheating all of which may introduce defects to the final product. These can often be internal but during the final stages of rolling they can become surface defects. Depending on the extent of the damage in a semi-finished hot rolled product, reconditioning may be necessary. This step, which can vary from surface grinding to downgrading or rejection of the product, is time consuming and expensive with costs amounting to as much as \$ 3,500,000 per year (Thomas et al., 2004).

1.3.1 DEFECTS FROM STEELMAKING AND/OR CONTINUOUS CASTING

An important aspect of steelmaking is the solidification of the molten steel when poured into the cast iron mould, the first step in the rolling process. As the steel solidifies, three separate zones are formed; (a) the outer zone populated by small grains, (b) the middle zone with larger columnar grains and (c) the inner zone of large equiaxed crystals (see Figure 1.6).

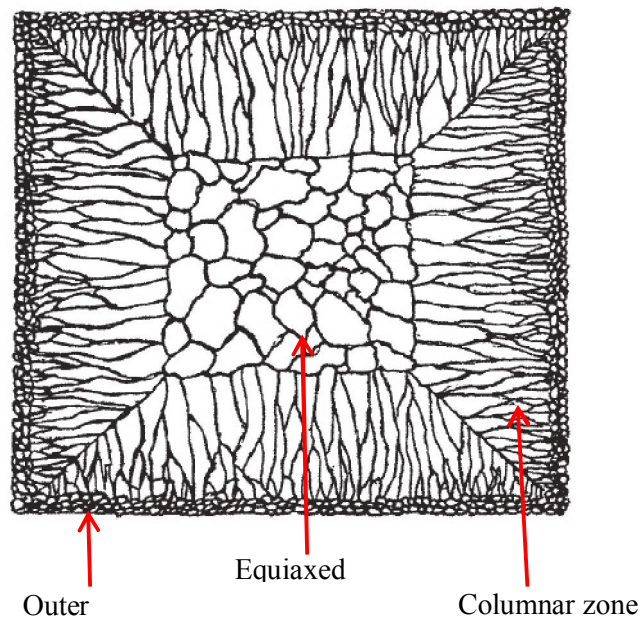


Figure 1.6: Solidification of ingot, modified from (Krauss, 2003).

The size, shape and distribution of those grains depend on the chemical composition of the molten steel and the cooling rates applied. The chemical composition is inhomogeneous in the ingot due to the different melting points of the alloying elements such as sulphur, phosphorous and carbon, which segregate to the non-solidified areas of the steel accordingly. This chemical inhomogeneity, termed liquation, is greatest along the axis of the ingot. Additionally, non-metallic inclusions such as manganese sulphides (MnS) which are products of steel deoxidation, and other impurities such as slag can be present in the ingot and greatly influence the mechanical properties of the steel giving rise to more defects.

During solidification, the outer layer of the steel solidifies first, reducing its volume, thus leaving an empty volume at the top central part of the ingot. This cavity is called the shrinkage pipe and can either be positioned in the central area of the ingot or be dispersed as porosity. A good quality ingot would contain the shrinkage pipe with the help of a device that limits its length, called hot

top. In steel ingots where the shrinkage pipe has dispersed as porosity, blow holes are created below the surface. If these are present in sufficient numbers near the surface they can cause surface defects such as seams and cracks later on. The temperature and pouring rates during ingot casting also affect the quality.

Finally, discontinuities on the inner surface of the moulds can cause surface defects like cracks and scabs. Cracks are formed due to rough inner mould surfaces that interfere with the shrinkage pipe of the solidified surface layer. Additionally scabs are formed due to splashing of the molten steel at the lower part of the mould, thus, leaving drops on the walls that can oxidise and not melt during casting. These defects can be avoided if the mould walls are serviced properly. Some of the resulting defects from continuous casting are shown in Figure 1.7 below.

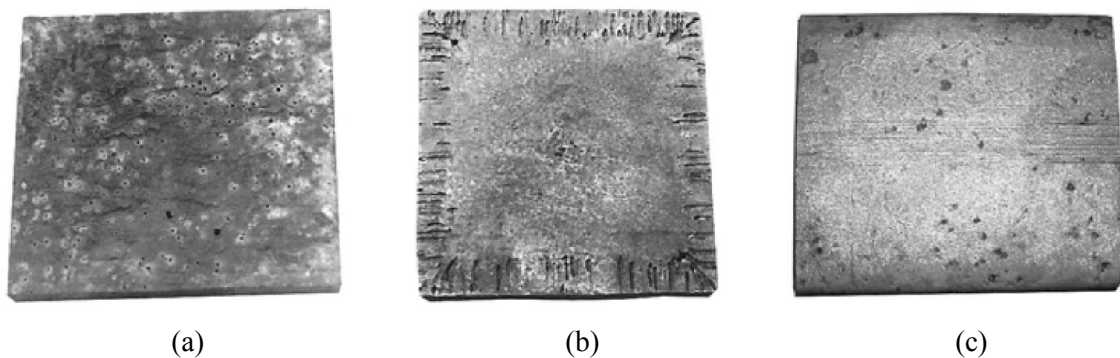


Figure 1.7: Surface defects of a continuous cast semi-finished product (CCS); (a) surface blowholes, (b) subsurface blowholes and (c) slag inclusions (Sychkov et al., 2006).

Surface oxidation of the semi-finished product prior to entering the finishing rolling mills is also a very important factor that can contribute to creating surface defects. Surface oxidation can occur due to the high temperature and environment humidity. In all cases the oxide layer has to be removed using high-pressure water sprays before entering the finishing rolls (Min et al., 2012).

1.3.2 DEFECTS FROM ROLLING

During hot rolling, recrystallization takes place transforming the grain structure of the as-cast metal to an equiaxed fine grained structure (see Figure 1.8), thus, affecting the final mechanical properties of the product.

The thermo-mechanical conditions during hot rolling play a key role in the development of the final microstructure of the finished product and, thus, its mechanical properties. These process parameters include the complex triaxiality regime, the stress-strain paths, deformation rates and

heat loss from contact with the rolls (Farrugia, 2013). These parameters, as defined by the manufacturing process, influence the microstructural properties depending on the initial microstructure which has been defined during the steelmaking and continuous casting. For example, the size and distribution of MnS inclusions which are generated during continuous casting as influenced by the cooling rates, affect the ductility of the material at a set temperature and deformation rate. Understanding the mechanisms that govern defect production is crucial to refine the manufacturing process and improve product quality.

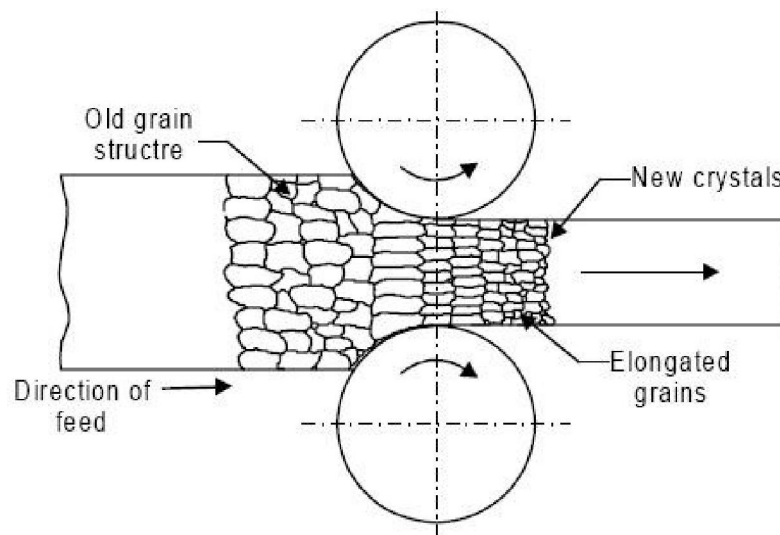


Figure 1.8: Recrystallisation process during hot rolling (Jumat, 2011).

1.4 AREA OF STUDY

Being able to understand, describe and ultimately avoid practises that would lead to defects is a priority of the hot rolling industry. The present volatile steel market requires steel manufacturers to become competitive by increasing production efficiency and thus, minimise cost. This can be realised if rolling processes are optimised so that the amount of rolled product experiencing edge cracking is kept to a minimum. Even a small reduction in the damaged product output, achieved by restructuring the processing route, could lead to an amplified change in the competitiveness of the steel manufacturer. Additionally, limiting the amount of scrap products from rolling routes could have a great environmental impact as CO₂ emissions would be reduced.

The way to accomplish the above is by understanding the parameters that cause or affect in any way damage under hot rolling conditions. Having this understanding, could help the development

of a material model able to predict this type of damage evolution. Such models can be implemented in Finite Element (FE) analysis of a rolling process in order to minimise damage by adjusting the processing parameters (i.e. rolling speed, cooling times etc.) accordingly (Thomson and Burman, 1980).

Unfortunately, there are many challenges associated with modelling hot deformation processes. For one, there are plenty of parameters affecting damage mechanisms under the complex thermal and straining histories the material experiences (Farrugia, 2006). Being able to assess which variable would be most important is often tricky and sometimes using a simplistic approach cannot describe the underlying phenomena sufficiently. On the other hand, involving multiple variables could give rise to an unnecessarily complicated model, difficult to calibrate. A balance must be achieved where the constitutive model should be simple enough for calibration but also able to capture the primary thermo-mechanical parameters that lead to damage.

This research concentrates on understanding the damage mechanisms by which ductility break-up such as edge cracking (see Figure 1.9), occurs in free cutting steel (FCS) under hot rolling conditions. Free cutting steels are inherently low ductility steels and are prone to developing defects during hot rolling. This is due to alloying elements that are often added in FCS to promote machinability and improve surface quality, nonetheless compromising material ductility. Examples include lead (Pb), sulphur (S), phosphorus (P), manganese (Mn) and silicon (Si). FCSs are very important for engineering applications due to the chip formation mechanism that increases their machinability (Farrugia and Cheong, 2009; Yaguchi, 1986). This in turn reduces tool wear, extends tool life, improves productivity and lowers energy consumption (Kishi and Eda, 1976). It is therefore imperative that a solid understanding of the causes of ductility defects during hot rolling of FCSs is achieved.

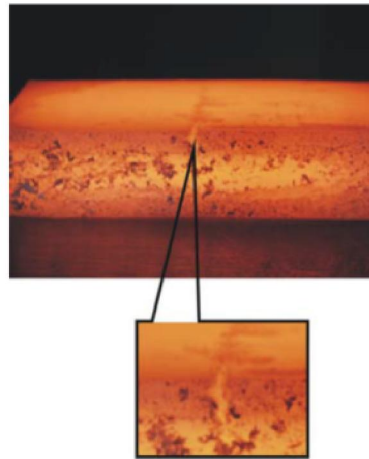


Figure 1.9: Edge cracking in a steel billet (Kainz et al., 2008).

Many parameters have been identified as damage agents in hot rolling processes for FCS. These can be grouped into macro-, meso- and microscale factors. Microscale parameters include grain size distribution, second phase non-metallic inclusion distribution and orientation, and chemical composition. All of the above contribute to the mechanical behaviour of the microstructure which is further influenced by the thermo-mechanical routes during rolling. For example the thermal profiles across the material, the stress-strain paths and the triaxiality (defined as the ratio of the hydrostatic stress to the equivalent stress) experienced during deformation are very important factors that influence and promote damage (Farrugia, 2006; Farrugia et al., 2007; Sellars, 1990).

During this research, triaxiality, a factor relating to stress-state, was investigated as the primary parameter influencing the onset of microcracks at certain locations of tested specimens. In the real case of hot rolling, the change of stress-state during the deformation has been recognised as one of the defining factors that alters damage mechanisms in the work piece (Farrugia, 2013; Foster et al., 2009b; Thomson and Burman, 1980). In Figure 1.10 an example of material deforming under different stress-states, depending on the position of the material in the work piece is shown. In this case the material at the centre of the work piece, due to the forces applied during rolling, experiences compression. Because of material continuity, the material at the edge of the work piece experiences a tensile strain. The fact that the edge material is more frequently damaged has to be linked with the different stress-state at that location. This also implies that edge cracking is associated with tensile stress conditions, thus, motivating the dominance of tensile testing throughout this thesis.

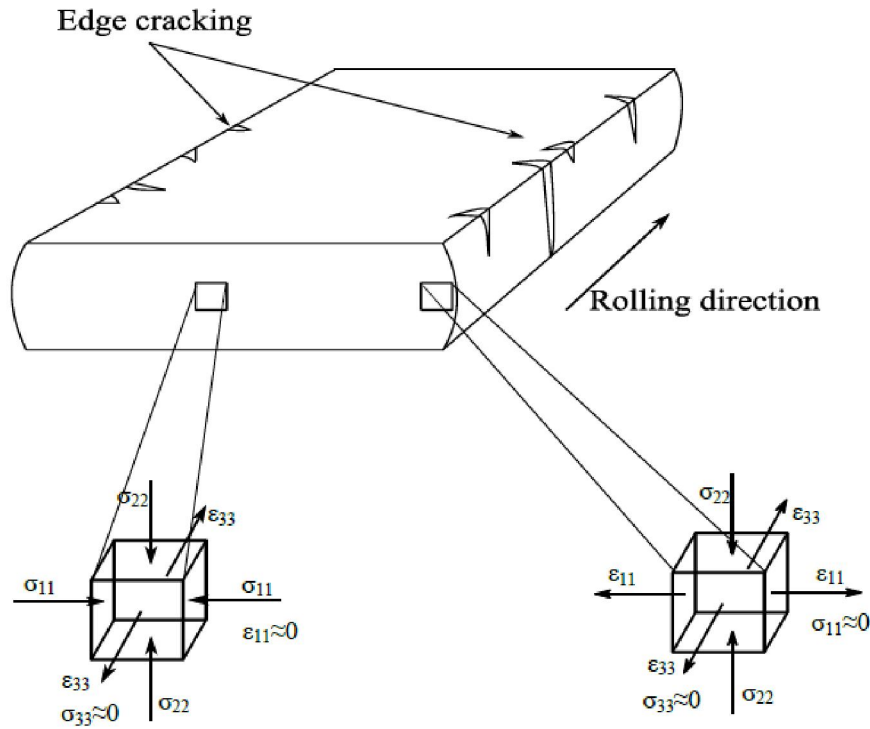


Figure 1.10: Stress-state example of hot rolling.

1.5 RESEARCH OBJECTIVES

The main aim of this research is to characterise the effects of stress-state on damage evolution under hot forming conditions and subsequently, to produce an accurate multi-axial, damage constitutive model. The material considered in the present work is leaded free cutting steel (LFCS). The material has been provided by Corus and tested in an as-cast condition at high temperatures and moderate to high strain rates. The objectives of this present work are:

- 1) To characterise the viscoplastic behaviour of the material via uniaxial tensile tests for the temperature range of 900-1200 °C and the strain rate range of 0.1-10 s⁻¹. The testing conditions have been chosen to represent the thermo-mechanical conditions experienced by the material in the first few passes of hot rolling.
- 2) To simulate the longitudinal thermal gradients, present in high temperature tensile testing performed using a Gleeble 3800 thermal-mechanical testing system. Being able to accurately simulate the conditions of the physical testing gives an advantage and provides confidence on the numerical results outlined in this work.

- 3) To experimentally investigate, through tensile testing, the effect of stress-state on the initiation of damage under high temperatures and moderate strain rates. This was achieved using double notched samples of varying notch radii.
- 4) To numerically simulate the effect of stress-state induced from the sample geometry on damage initiation and compare the experimentally observed damage profiles with calculated stress-state results, thus, giving insight into which stress-state parameters control damage.
- 5) To develop a multiaxial damage equation describing the observed effect of stress-state on damage initiation, calibrate the model using multiaxial data and experimental damage profiles and finally predict damage initiation patterns for a wide range of conditions encountered in the hot rolling route.

1.6 THESIS STRUCTURE

An extensive review of the literature on damage mechanisms, experimental observations of damage occurring under hot forming conditions and multiaxial damage models is carried out in Chapter 2. Emphasis is on tensile notch bar testing and the effect of stress-state on damage.

In Chapter 3 the material and testing equipment used are presented. Additionally, the experimental programme which is divided in two sections; thermal testing and tensile testing (uniaxial and double notch), is outlined.

A method for simulating the high temperature testing conditions accurately in ABAQUS is presented in Chapter 4. In this chapter, predictions of the longitudinal thermal temperature profiles, present in high temperature Gleeble testing, are presented and compared to experimental measurements.

Chapter 5 includes the experimental results from the uniaxial tensile testing, double notch tensile testing and microstructural analysis of unbroken notches of the failed double notch samples in an effort to understand damage initiation as a function of stress-state (geometry of the notch). A discussion of the presented results is also given in this chapter.

FE simulation results based on the double notch testing presented in Chapter 5, including longitudinal thermal profiles, are presented in Chapter 6. Comparisons of the simulated results, which do not include damage, to the experimental results from Chapter 5 are carried out in an

effort to understand the influence of principal stress, hydrostatic stress and equivalent stress on damage initiation.

In Chapter 7 the constitutive damage equations that describe the effect of stress-state on damage, as observed experimentally and numerically in Chapters 5 and 6 are presented. Predictions of this calibrated unified multiaxial damage model are also given in this chapter.

CHAPTER 2

EXPERIMENTAL AND MODELLING TECHNIQUES FOR DUCTILE DAMAGE EVOLUTION UNDER MULTIAXIAL STRESS CONDITIONS

SUMMARY

Although the interest of this research is damage during hot rolling, other major damage types encountered in metallic systems are also outlined so that a comprehensive understanding of damage as a whole is achieved.

In the following sections a review on notched bar testing is presented as a way of understanding the constraints stress-state imposes on material behaviour and ultimately damage. The importance of stress-state when considering damage progression was recognised experimentally for steel very early on (Kirkcaldy, 1860) and was further corroborated by Bridgman almost a century later (Bridgman, 1952).

Finally, the fundamental damage modelling principles are explained and important damage models that apply to both creep and ductile damage under multiaxial conditions are discussed. The emphasis is on models that account for the effects of stress-state in their damage evolution description.

2.1 DAMAGE DEFINITIONS-MECHANISMS

2.1.1 DAMAGE DEFINITION

In modern engineering and manufacturing processes, materials have to endure adverse mechanical, thermal and environmental conditions which often cause detrimental microstructural changes that decrease their ability to withstand loading; effectively a reduction in strength. To describe this loss in the carrying capacity of the material, the term damage is used. Each

manufacturing process, (cold forming, hot forming, superplastic forming etc.) or engineering application (structural loading, irradiation, corrosion etc.) favours a different type of damage. This essentially reveals that the application conditions, along with the material microstructure are the defining factors that dictate which damage mechanism will lead to material failure.

The primary damage types occurring in metallic systems are ductile damage, creep damage, superplastic damage and damage under conditions of hot forming. Other types such as fatigue damage, irradiation damage and spalling are also important but have not been detailed in this section as the damage mechanisms are dissimilar to high temperature deformation. In all damage types, although the micromechanisms by which damage is conceived may differ, the general process remains the same. This is the nucleation, growth and finally coalescence of micro-defects to form microcracks which eventually fuse together to create macroscopic cracks, thus, leading to material failure (see Figure 2.1).

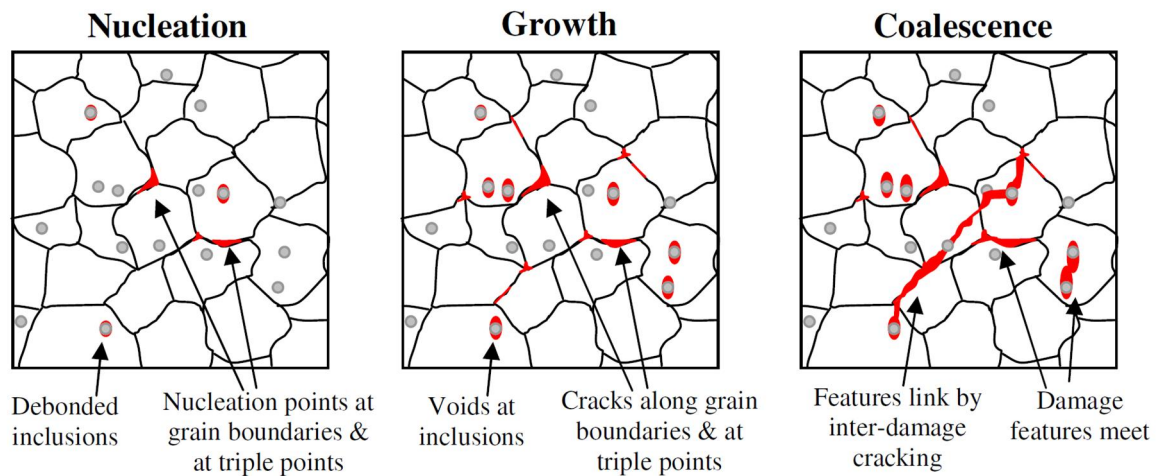


Figure 2.1: Schematic of nucleation and growth of voids and microcracks at second phase particles and grain boundaries (Foster, 2007).

2.1.2 DUCTILE DAMAGE

Ductile damage, also known as fibrous fracture, dimpled rupture or microvoid coalescence, is often defined in three stages; void nucleation, growth and finally coalescence that leads to macro or visible cracks and ultimately material failure. This type of damage is encountered in metallic systems at low temperatures, usually less than one third of the melting temperature ($T < 0.3 \times T_m$) and has attracted much attention over the years as it has been encountered in various cold working processes. Cold forming is important for mass production of commercial goods since the ability to form parts at room/low temperatures makes it an extremely economic and environmentally

friendly option. In cold metal working deformation is usually facilitated by multiplication of dislocations.

The mechanism by which ductile damage takes place was identified as void nucleation at inclusions and/or second phase particles. This was first hypothesised by Tipper (1948) who stated that ductile damage is caused by decohesion of the second phase particles from the matrix. Experimental evidence for this hypothesis was given later by Puttick (1959) who sectioned specimens of copper and iron, deformed under tension, and examined them using optical microscopy. Since then there have been many experimental observations of non-metallic inclusions acting as points of void nucleation, either by matrix/particle interface decohesion (see Figure 2.2a) or by inclusion cracking (see Figure 2.2b) (Beremin, 1981; Cox and Low, 1974; Goods and Brown, 1979; Pardoen et al., 1998). Interface decohesion can be influenced by a number of factors such as particle size, distribution, volume fraction, stress-state and strength of the matrix material.

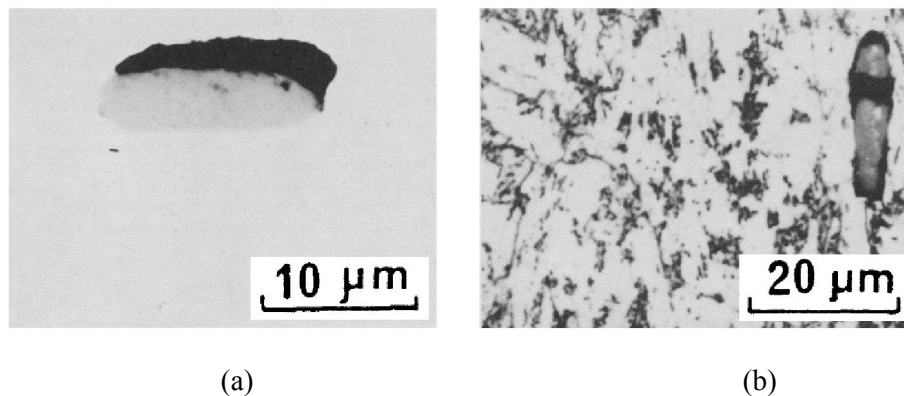


Figure 2.2: Damage around inclusion; (a) inclusion/matrix decohesion and (b) inclusion cracking. Modified from Beremin (1981).

2.1.3 CREEP DAMAGE

Creep can be defined as the time-dependent deformation of materials subjected under a static load at elevated temperatures, often ranging between a third to half of the material's melting point ($0.3 \times T_m < T < 0.5 \times T_m$). This type of damage plays an important role in industrial design for the power generation industry (fossil fuel plants, nuclear plants, petrochemical industry etc.) and for assessing the lifetime of structural parts. There are three stages in creep deformation; primary, secondary and tertiary. The primary stage is defined as the instantaneous strain observed at the initial load application, followed by a period of strain rate decrease. At the end of this stage, when a steady state and creep rate have been reached, the secondary or stationary stage begins. A

component under creep would typically spend the longest time deforming under this stationary stage at a slow but constant deformation rate. The strain rate would accelerate once the component has entered the tertiary stage of creep during which creep failure will also take place (Sposito et al., 2010).

In creep damage, failure takes place by intergranular linkage of cavities and microcracks that nucleate, grow and accumulate during the deformation (Nix, 1988). In creep type deformation, cavities are thought to nucleate on grain boundaries, as seen in Figure 2.3. There are a few mechanisms that have been proposed to explain this process. A popular explanation is that cavities nucleate because of grain boundary sliding. According to this view, cavities would be most likely to form at a grain boundary triple junction. This microdamage mechanism is more prevalent at low temperatures, high stresses and for coarse grained materials (Waddington and Lofthouse, 1967). A study by Argon et al. (1980) showed that this is indeed an important mechanism for materials with small particles, sparsely distributed at grain boundaries. Nonetheless, grain boundary cavities have also been observed in pure materials and on grain boundaries transverse to the loading direction, suggesting that this is not the only acting mechanism.

Dyson (1983) proposed the interaction of dislocations with grain boundaries as another cavity nucleating mechanism, often termed as ‘dislocation pile-ups’. During this mechanism, a large normal stress develops when a slip band is stopped by a grain boundary or a hard second phase particle. When the magnitude of this stress concentration and the rate of internal loading are large enough then a cavity can nucleate (Yoo and Trinkaus, 1986).

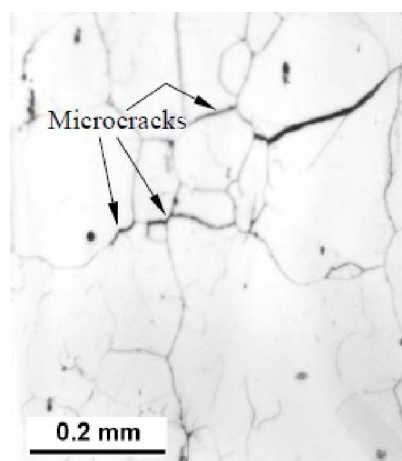


Figure 2.3: Intergranular creep microdamage, σ_{net} = 404 MPa, modified from (Bettinson, 2001).

After cavities have nucleated in a material deforming under creep conditions, growth usually via vacancy diffusion takes place. At elevated temperatures, during diffusive cavity growth atoms move from the cavity surface onto the surrounding grain boundaries (Hull and Rimmer, 1959). In addition to this unconstrained grain boundary diffusion growth mechanism, a surface diffusion growth mechanism was also proposed by Chuang and Rice (1973). This results in faster cavity growth, by shortening the effective diffusion path length. This occurs when matter, diffusing from the cavity surface, is accommodated by local separation of the surrounding grains. Grain boundary sliding can also contribute as a growth factor, although it is more typical of superplastic damage.

2.1.4 SUPERPLASTICITY DAMAGE

Under high temperatures and controlled strain rates some fine grained (less than 10 μm), materials can experience extremely high tensile elongation, ranging from 200 to several thousand percent (Langdon, 1991). This type of deformation is called superplasticity and is used in sheet metal forming to produce complex, light and strong engineering parts, often for aerospace applications. Typical superplastic deformation conditions are testing temperatures of at least $0.5 \times T_m$ and constrained strain rates of the range $10^{-2} - 10^{-4} \text{ s}^{-1}$.

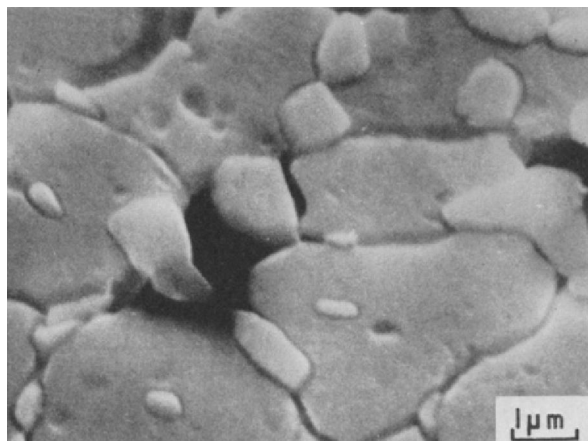


Figure 2.4: Superplastic damage at triple junctions of a nickel silver alloy (Livesey and Ridley, 1978).

The mechanisms by which material deforms under conditions of superplasticity are grain rotation and grain boundary sliding (Langdon, 1991). Cavity nucleation (see Figure 2.4) in this type of deformation can be attributed to a few mechanisms; intergranular slip interaction with hard second phase particles or grain boundaries, grain boundary sliding, fragmented particles or

vacancy condensation on grain boundaries. These cavities can grow by stress-directed vacancy diffusion, superplastic diffusion or plastic deformation. Stress-directed vacancy diffusion increases void volume by vacancies directed into the void along one common grain boundary. This mechanism is most important for small cavities. Superplastic diffusion works in a similar way but the vacancies are directed along multiple grain boundaries. This is only activated when the material deforms at a strain rate of 10^{-4}s^{-1} (Raj and Ashby, 1975). The last mechanism, plastic deformation, is the most dominant growth mechanism for superplastic deformation and is due to translation of grains (Chokshit and Langdon, 1987). Knowledge of the void volume fraction in the deformed material can help determine the dominant void growth mechanism as it has a unique relationship with strain. This relationship is, linear for stress-directed vacancy diffusion, for superplastic diffusional growth the cube root of void volume fraction is proportional to strain and for plastic deformation there is an exponential relationship between the two (Raj and Ashby, 1975).

2.1.5 DAMAGE AT HIGH TEMPERATURES

Damage in hot metal forming conditions has a lot of similarities with the aforementioned damage types. Since hot metal forming is performed under high temperatures, it is likely that there will be damage and deformation mechanism similarities with creep and superplastic deformation. Although hot metal forming temperatures are typically at the high end of the range, often above $0.6 \times T_m$, mechanisms such as grain boundary diffusion and sliding are encountered, resulting in void nucleation at triple grain junctions. Such ‘creep-like’ mechanisms are more likely for fine grained materials at high temperatures and low strain rates (Liu et al., 2006). Although, this set of conditions is admittedly not frequently encountered in hot metal forming. Indeed, most materials have a coarse microstructure and the strain rates tend to be much higher than the rates used in creep or superplastic forming. The short deformation times make it unlikely for grain boundary diffusion and sliding to evolve, instead dislocation-based deformation is generally more prominent. The deformation behaviour is, therefore, similar to that seen in cold metal working. It follows that the damage mechanisms would also present similarities between the two metal forming methods.

Often in hot metal working, damage is manifested as voids around second phase particles inside grains (Foster et al., 2007). This ‘ductile-type’ damage is frequently seen alongside damage at grain boundaries. Grain boundary damage forms as cracks along the grain boundaries and is termed microwedge cracking. It is possible that grain boundary diffusion mechanisms are

responsible for weakening the grain boundary bonds resulting in microcracks. These two damage mechanisms act simultaneously in high strain rate hot forming but material and forming conditions such as grain size, temperature and strain rate dictate which mechanism prevails (Lin et al., 2005). In Figure 2.5 the two microdamage mechanisms are shown in a sectioned area for free cutting steel that failed in tension under hot forming conditions of high temperature (900 °C) and a high strain rate (0.1 s^{-1}). The micrograph is taken from a uniaxial sample for which the stress-strain results are shown in Chapter 5.

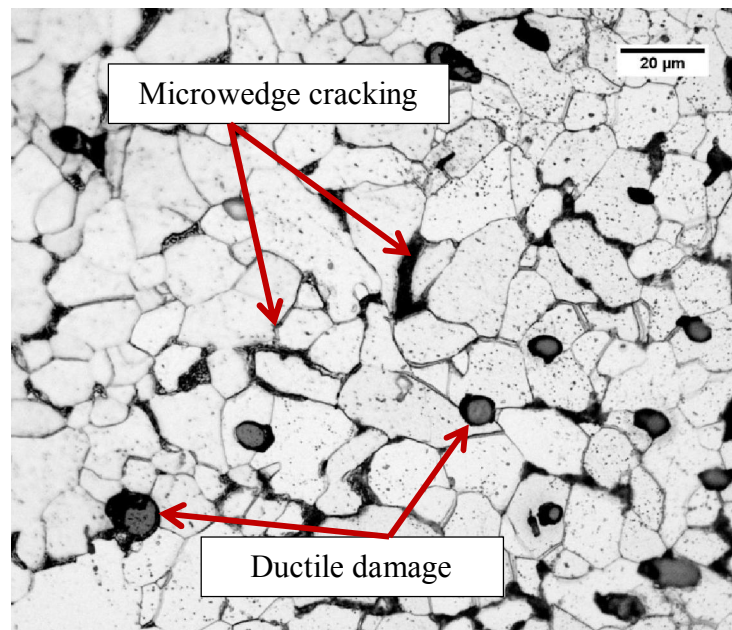


Figure 2.5: Microwedge cracking and ductile damage observed from a uniaxial tension test under $T=900 \text{ °C}$ and $\dot{\epsilon}=0.1 \text{ s}^{-1}$ on free cutting steel.

2.1.6 SUMMARY

In this section, main damage types such as creep damage, ductile damage, superplastic damage and hot forming damage were reviewed to gain insight into the micromechanics of each one. Generalising those mechanisms one could say that there are two main categories when it comes to damage; grain boundary damage and damage around second phases. These two mechanisms can be facilitated by diffusion processes, multiplication of dislocations and their interactions, grain sliding and rotation or debonding of second phase particles from the matrix. Each type of damage proceeds through the preferred mechanism according to the deformation conditions, namely deformation rate, elongation limits and temperature. Specific damage mechanisms are also known to be favoured according to material microstructure and composition (Liu et al., 2006). Although,

one damage mechanism is often prevalent for each damage type, it is not unusual for two damage mechanisms to be present simultaneously and perhaps competing with each other. This is commonly observed for damage in hot forming/rolling (Liu et al., 2006, 2005; Puncreobutr et al., 2012). Unlike the combining nature of hot forming damage, creep damage is mostly intergranular as is superplastic damage and ductile damage is manifested as voids forming around second phases. The following schematic (see Figure 2.6) summarises the main damage mechanisms seen in (a) creep, (b) superplastic damage, (c) ductile damage and (d) hot forming damage.

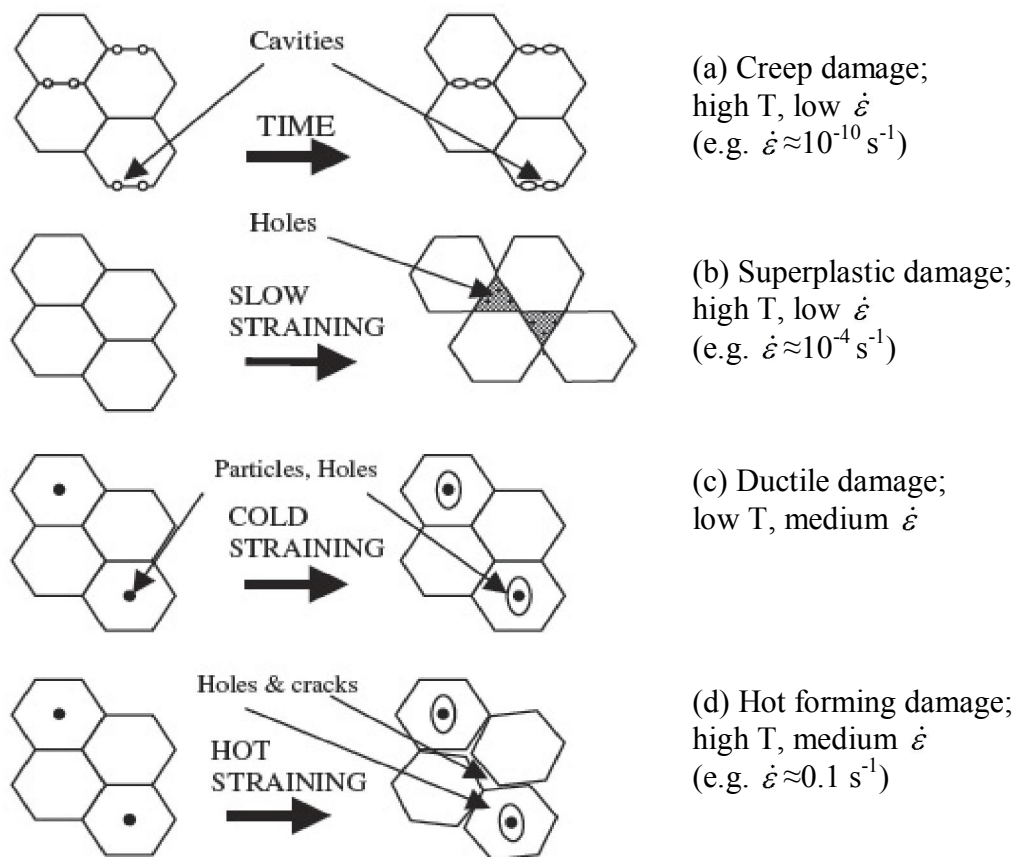


Figure 2.6: Schematics of the dominant mechanisms in (a) creep damage, (b) superplastic damage, (c) ductile damage and (d) hot forming damage. In creep damage is intergranular where grain boundary diffusion controls the process. This takes place at high stress and temperature creep. For case (b) it is grain boundary sliding and rotation that leads to voids at triple junctions of grain boundaries. In ductile damage voids grow under plasticity conditions around second phase particles. Finally for damage observed under hot forming conditions a combination of grain boundary (Lin et al., 2005).

2.2 EXPERIMENTAL OBSERVATION OF DAMAGE

2.2.1 HOT WORKABILITY TESTS

Ductility can be defined in a few different ways in hot working. It can be expressed as a percentage reduction of the cross sectional area of a sample in tension or the number of revolutions to failure in a torsion experiment. Compression and bend tests can also be used to assess the hot workability of materials. For example, ductility is measured as a minimum reduction in height in compression tests and in bend tests measuring the average crack depth can provide useful information about the ductility of the material.

Each test can provide useful information about how the intrinsic material parameters respond to the applied working process, essentially the stress and strain state. This notion can be summarised by the following equation (Cockcroft and Latham, 1968):

$$\text{Hot workability} = f_1 \text{ (material)} \times f_2 \text{ (working process)} \quad 2.1$$

where, f_1 is a function that represents the intrinsic ductility of the material and f_2 is a function of all the external parameters imposed by the chosen working process or test. Choosing a suitable test to measure hot workability is, therefore, dictated by the factors to be investigated.

2.2.2 DUCTILITY MEASUREMENT FROM TENSILE TESTS

Fast strain rate, hot tensile tests have been established as a good measure of assessing hot ductility for steels (Mayland et al., 1976). This is due to the fast testing output and the simplicity of the ductility measure in a tension test. Furthermore, in tension tests a wide range of samples can be used to evaluate different stress-states. Uniaxial samples provide the basis for the most fundamental type of tests with the simplest stress-state where the maximum principal, equivalent and hydrostatic stresses are equal ($\sigma_1 = \sigma_e = \sigma_H$). These samples feature a long reduced gauge at their centre where the deformation, strain rate and temperature are assumed to be uniform (see Figure 2.7a). In order to induce more complicated stress-states, notched samples are used (Figure 2.7b). Bridgman (1952) was the first to provide evidence via tension tests that stress-state and strain to failure are inversely proportional. Since this work multiple studies on the effect of stress-

state on failure were carried out using notched samples in a Bridgman style analysis (Becker et al, 1988; Hancock & Brown, 1983; Isobe, Yashirodai, & Murata, 2014; D. Kwon & Asaro, 1990).

Uniaxial Notched

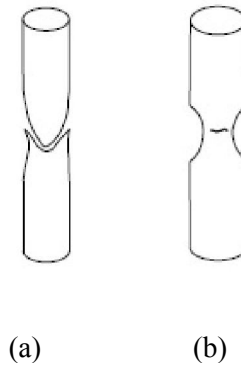


Figure 2.7: Typical samples used in tension tests. Modified from Benzerga and Leblond (2010).

Although tension tests can be fast and simple they still have some limitations, in particular necking. During necking, the deformation rate of the material accelerates uncontrollably and does not represent the originally applied strain rate. This difficulty can be addressed if a non-constant deformation speed is applied (Sellars and Tegart, 1972). Furthermore, temperature variations in the heated sample, depending on the method of heating, can lead to erroneous data due to the temperature dependent flow properties of the material. These are serious considerations in high temperature testing, especially, for uniaxial samples (Kardoulaki et al., 2014).

Nonetheless, uniaxial tension tests have been used extensively to determine and map the deformation and damage mechanisms present in hot working conditions. Early studies on ductile fracture in steel have used uniaxial, ASTM type samples to examine mechanisms such as void initiation and growth (Park and Thompson, 1988). Later, Liu et al. (2006) used axisymmetric uniaxial tension tests to conduct an experimental program that took into account the effects of strain rate, material, temperature, recrystallization and grain size on damage evolving under hot rolling conditions. Performing similar, fundamental investigations with uniaxial samples is often the first step in a rigorous examination of material damage (Bao and Wierzbicki, 2004; Cortese et al., 2013; Li et al., 2011; Voyiadjis et al., 2012). Although the stress-state experienced by uniaxial samples is unlikely to be encountered in a real life situation, it is important to be able to express a fracture criterion in a simple form before moving on to the complicated multiaxial case scenario (Hayhurst, 1972).

2.2.3 DAMAGE IN NOTCHED AXISYMMETRIC TENSILE SAMPLES

Early experimental evidence on the importance of stress-state was given by Bridgeman (1952) who proved that reducing the tensile stress-state increases ductility. Following him, McClintock (1968) and Rice and Tracey (1969) sought to model this relationship between void growth and stress triaxiality in their respective models of a long cylindrical and spherical void enlarging in a remotely uniform plastic deformation field. In both studies it was found that more triaxial states of stress speed up void growth. The inverse relationship established in those works between the strain to failure and triaxiality has also been validated by various experimental studies like those of Hancock and Mackenzie (1976) or Mackenzie et al. (1977) for structural steels at room temperature. More specifically, the latter study highlighted the importance of understanding how damage initiates for different stress-states. Following that idea, Hancock and Brown (1983) carried out an investigation on structural steels, using the same techniques as Hancock and Mackenzie (1976), and confirmed that damage in notched bar samples occurs in the centre of the notch plane, where triaxiality is maximum.

In a contrasting study published by Alves and Jones (1999), it was found that damage in notched bars of mild steel initiated in regions of low triaxiality and it was hypothesised that strain may control void nucleation rather than stress. As done in previous studies (Hancock and Brown, 1983; Hancock and Mackenzie, 1976; Mackenzie et al., 1977), the Bridgman analysis was used and its results were compared to experimental results and FE calculated results. The study highlighted that Bridgman analysis should be used with caution since it does not allow for material and geometry dependencies. Many more studies aiming to map the influence of hydrostatic stress on material ductility have been conducted for various metals (Bao, 2005; Goto et al., 1999; Helbert et al., 1995; Jun, 1991; Mirone, 2007), with some examining the effects over a wide range of strain rates (Clausen et al., 2004; Mirza et al., 1996), but none on the combination of high temperature and high strain rate.

Circumferentially notched samples are also extensively used in creep tension experiments to examine how multiaxial stress-states can influence material ductility and damage initiation (Gao and Zhang, 2009; Goyal et al., 2013; Hayhurst et al., 1984). Significant efforts have been made to establish an appropriate notch design so that a uniform stress-state with minimum stress gradients is achieved in the notch region (Hayhurst et al., 1977). In this study, a variety of notched sample designs were studied by altering design parameters like b/a and R/a (see Figure 2.8). From

these, a circular Bridgman type design ($b/a = 1.67$, $R/a = 0.67$) and a variation of that design ($b/a = 1.9$, $R/a = 0.5$) were deemed most suitable for the study of multiaxial rupture. In later studies by the same author, (Hayhurst et al., 1984) both Bridgman type notched samples (Figure 2.8a) and British Standard (B.S.) notched samples (Figure 2.8 b) were used to assess the creep rupture characteristics experimentally and numerically. In this work, the Bridgman type notched samples provided the basis of understanding creep fracture in a complex, yet uniform stress field whereas the B.S. notched samples were used to examine a more realistic and complicated, non-uniform stress-state. From these tests it became apparent that crack propagation is highly dependent on the material, which in turn defines the stress rupture criterion, the geometry of the specimen and the stress fields that it induces, and the evolution of those two factors with time.

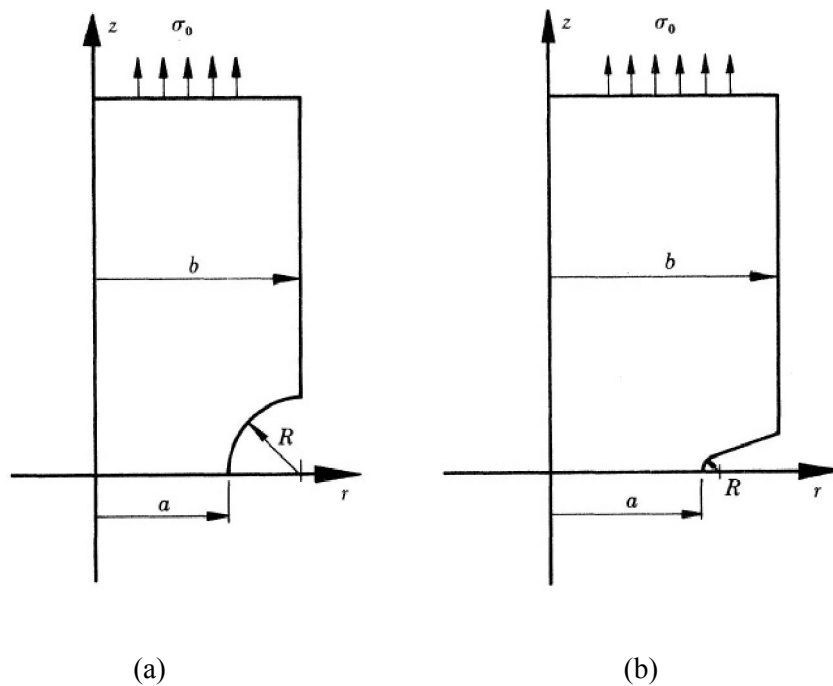


Figure 2.8: Tensile notched bar geometries; (a) Bridgman type notched sample ($b/a = 1.67$, $R/a = 0.67$) and (b) British Standard (B.S.) notched bar ($b/a = 1.41$, $R/a = 0.055$) (Hayhurst et al., 1984).

Another important study was conducted by Dyson and Loveday (1981) and showed that two distinct failure modes can be observed in notched samples depending on the level of stress experienced. It was found that at low stress levels, cracks started from the outside of the notch and progressed towards the centre, known as ‘outside-in’ damage. At high stress levels, damage initiated at the centre of the notch, namely ‘inside-out’ damage. These tests were performed on

Nimonic 80A and the notched samples were designed according to the standard Bridgman type geometry ($b/a = 1.9$, $R/a = 0.5$). Apart from experimental studies, many researchers have numerically investigated the effects of stress-state distributions using notched bars in creep (Al-Faddagh, Fenner, & Webster, 1982; Hayhurst & Henderson, 1977; Kirkcaldy, 1860; Othman, Dyson, Hayhurst, & Lin, 1994).

2.2.4 EXPERIMENTAL STUDIES OF STRESS-STATE EFFECTS ON DAMAGE

Mirza et al. (1996) investigated the effect of triaxiality on ductile fracture for three materials over a wide strain rate range (10^{-3} to 10^4 s $^{-1}$). The materials investigated were mild steel, Remko iron and an aluminium alloy. The study used notched tensile bars of different notch radii for two sizes of bar specimens, one with a 15 mm gauge diameter and one with 9 mm gauge diameter. For the 15 mm gauge diameter specimen, five different notch radii were used whilst for the smaller sized sample four were used. The study showed that for all materials the ductility was strongly dependent on triaxiality, as expected. The degree of dependency changed depending on the material, with ferrous materials exhibiting a strong dependency and the aluminium alloy exhibiting a weaker dependency. The results for mild steel showed that the strain rate and stress triaxiality combined can control the mode of fracture. As the strain rate was increased, a transition to a brittle fracture mode was observed for samples with low levels of triaxiality.

A numerical study of high strain rate tensile tests on smooth and notched bars of structural steel Weldox 460 E was performed by Borvik et al. (2003). In this study the main findings were that triaxiality does not stay constant with deformation, the triaxiality is almost insensitive to strain rates or temperature and that the material damage is driven by a combination of accumulated plastic strain and stress triaxiality. The study suggested that it is a competition between the high triaxiality values in the centre of the specimens and the high plastic strains at the notch root that determine the failure sites. It was, therefore, concluded that failure could take place at regions where stress triaxiality is not the highest. This finding was in contrast to other studies that predicted failure always taking place in the centre of the specimen, where triaxiality is maximum, for both smooth and notched bars.

A study of similar findings was conducted by Mirone (2007) who tested smooth and notched samples made of aluminium, copper and steel, to evaluate the role of triaxiality on material damage. Again the eventual material failure was attributed to a competition between high triaxiality values occurring in the centre of a sample and high plastic strains in the notch root. It was found that for notched samples, larger voids were observed at the centre of the sample,

promoted by high triaxiality whilst smaller ones were found near the notch root. For the sharpest notch, where equivalent strain is highly non-uniform, the strain induced effect was found to dominate and failure was initiated on the notch root rather than the high triaxiality centre.

Bao and Wierzbicki (2004), in an effort to assess a number of ductile fracture criteria performed a study where different tests were employed to give a wide range of different stress and strain states for an aluminium alloy (2024-T351). Upsetting tests and tensile tests of smooth and notched bars of two different geometries were performed at room temperature. In this study it was assumed that fracture in the tensile tests always occurred at the centre of the notch. In a later study by the same author (Bao, 2005) investigating stress-state effects on damage on the same material, a more detailed study of the stress-state developed due to the geometry of the sample was performed. The experiments included tensile tests of flat specimens, smooth and notched bars and flat-grooved plates. The study focused on four parameters and their effect on damage evolution; equivalent strain, stress triaxiality and stress and strain ratios. The triaxiality values were calculated numerically (FE) and were not based on the Bridgman solution. The numerical results predicted that fracture will initiate at the centre of the notch on the bar specimens, consistent with classical observations. This study argues that the equivalent strain and stress triaxiality were the two most important parameters that affect crack formation.

Recently Li et al. (2011) conducted an extensive experimental study to investigate the effects of stress-state on ductile fracture. The experiments were performed at room temperature for an aluminium alloy, Al-6061(T6). The investigation included several kinds of tensile tests (smooth bar, different radii notched bars, shear induced tensile tests) and upsetting tests on various sample geometries, including flanged and tapered upsetting tests. This study aimed to characterise the modes of ductile fracture that occur for each one of the examined stress-states and then use a range of ductile fracture criteria to assess their validity for the examined stress-state. The study concluded that in the majority of examined cases, the ductile fracture mode was mixed shear-dimple fracture. This was the case for both smooth bar samples and notched bar samples. Although, notched bar samples with smaller radii were shown to exhibit a ‘cup-cone’ dominated fracture mode meaning that the maximum principal stress controlled this type of fracture.

2.2.5 DAMAGE IN DOUBLE NOTCHED AXISYMMETRIC TENSILE SAMPLES

Ample sets of uniaxial and notched experimental data have been collected over the years to aid understanding and interpret the effects of stress-state on damage. Most studies have focused on finding a link between stress-state and material failure from the observed flaws, post failure.

Hence, initiation and early damage development have not been crucial parts of most studies. Understanding damage initiation however, is an important milestone that has to be achieved in order to be able to fully describe how stress-state affects the evolution of damage.

Double notched samples have helped researchers move towards that goal. The idea behind using a double notched sample lies in the fact that the unbroken notch holds information about the initial stages of material damage. The material in the unbroken notch would be arrested in a stage between damage initiation and failure. This essentially has the same effect as performing an interrupted tensile test without the introduced experimental errors of inaccurate timing, which can be very difficult to control in fast strain rate tests. In Figure 2.9 a schematic of a typical double notched sample design is shown.

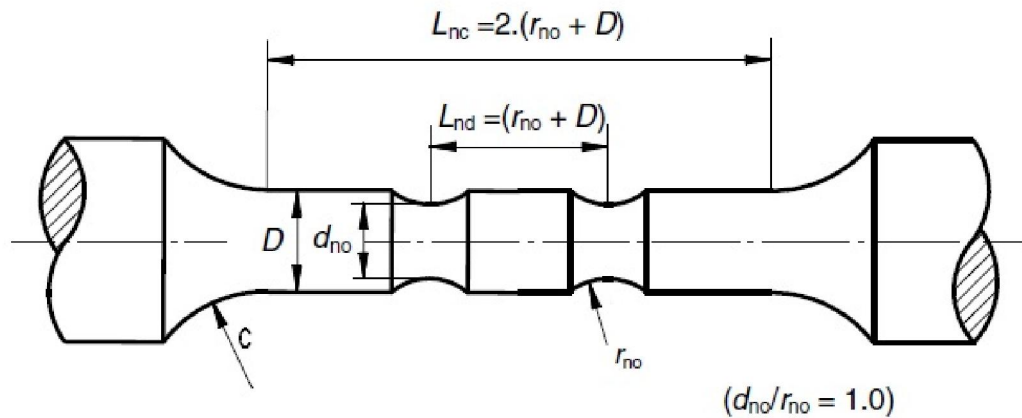


Figure 2.9: Double, Bridgman type notched sample template (Webster et al., 2004).

An important aspect that has to be evaluated when using this type of specimen is stress interference between the two notches. A Code of Practice on notched bar testing of multiaxial creep rupture was first published in 1991 by the Notch Bar Creep Testing Working Party of the High Temperature Mechanical Testing Committee (Webster et al., 1991). This aimed to standardise the testing procedures for notch bar testing and simplify the analysis and interpretation of the results. A few years later, a new report that validated and recommended revisions on the procedures that were suggested by the first report was published (Al-Abed et al., 1999). In those studies, testing with double notched samples was suggested as a method of facilitating metallographic examinations of the unfailed notch to determine how early stage damage evolves. The second validation report (Al-Abed et al., 1999) identified the possibility of stress interaction between the two notches and examined this via FE analysis. The results of this work confirmed

that the initial design recommendations for the samples are sufficient, as the stress field interactions were found to be insignificant. Finally, a more recent report on interpreting the data from notched bar creep testing was published in 2004 (Webster et al., 2004). These studies created a structured basis for double notch bar testing and established this method as a way to identify damage initiation patterns as a function of stress-state.

Specifically in creep damage assessment, double notched samples have been used extensively (Al-Abed et al., 1999; Othman, Lin et al., 1993; Webster et al., 2004). Othman et al. (1993), following the recommendations from the first Code of Practice for creep testing, compared the response of a single notched sample to a double notched sample using FE analysis and found that fracture lifetime results were within a 5% error bracket. This reinforced the merit of using double notched samples as a way of performing both a micro and macro analysis on data from examination of the damage mechanisms and the failure response respectively.

Kwon et al. (2004) used double notched samples to assess the multiaxial rupture behaviour and stress-state sensitivity of Durehete 1055 and 2.25Cr1Mo steel. As expected, the ductility was strongly influenced by the state of multiaxiality in the sample. Additionally, the two materials showed different creep crack propagation paths. In 2.25Cr1Mo steel the cracks developed near the notch root and were concentrated there whilst damage in the Durehete 1055 samples was initiated in the central region of the notch and was more spread out over that general area. Based on that evidence, it was proposed that the multiaxial stress failure mechanism is material dependent.

The dependency of creep damage on multiaxiality in a low-carbon, Nitrogen-added 316 stainless steel was also examined in a study by Takahashi et al. (2008). These results showed that multiaxiality did not affect the reduction of life in this particular steel, although, the distribution of damage examined from the unbroken notch changed as a function of the stress-state imposed by the sample geometry.

Recently, similar studies have been conducted to examine how creep damage characteristics vary as a function of multiaxiality so that a better understanding is achieved for a particular material leading to more sophisticated lifetime predictions. Four notched samples were used by Isobe et al. (2014) to study creep damage initiation patterns and mechanisms. From comparisons of micrographs from the notched samples to those of uniaxial samples, it was found that the damage mechanisms change from a creep dominated deformation type failure of the uniaxial samples to a more ductile type fracture for the notched specimens. The study also reported that creep cracks initiated at areas where the triaxiality factor was highest, as calculated from the numerical

analysis. Double notched samples have also been used to perform semi-quantitative analysis on the effect of element additions on damage mechanisms (Khadkikar et al., 1989).

2.2.6 OTHER SPECIMEN DESIGNS

As explained in previous sections, much research has been concentrated on examining the effects of stress-state on damage initiation and mechanisms. Circumferentially notched bars with varying notched profiles are indeed a very popular choice of sample for this purpose. Double notched bar samples have also been used readily as they have the added advantage of facilitating post-failure metallographic examinations of damage initiation from the unfailed notch. However, there is still a wide range of multiaxial stress-states that can be achieved by various other sample designs.

Hayhurst et al. (Hayhurst and Henderson, 1977; Hayhurst, 1972; Hayhurst and Felce, 1986) have extensively studied the effect of multiple states of stress on creep rupture via uniaxial, notched, biaxial and triaxial testing. For the purpose of biaxial testing, cruciform biaxial plate specimens were used in tension conditions (Hayhurst, 1972). Pilling and Ridley (1986) also performed biaxial tests to study the role of hydrostatic pressure on cavity nucleation and growth in superplastic alloys. Although biaxial tests provide a simple way to study multiaxiality, they are restricted to sheet materials and therefore have limited applicability. For triaxial tension testing, Hayhurst and Felce (1986) aimed to design a sample that would create a uniform stress-state of high triaxiality, similar to that encountered near a crack tip. The designed sample, shown in Figure 2.10, has a complicated geometry and it was found to perform inferior to a Bridgman type notched sample since a constant, uniform triaxial stress-state could not be achieved.

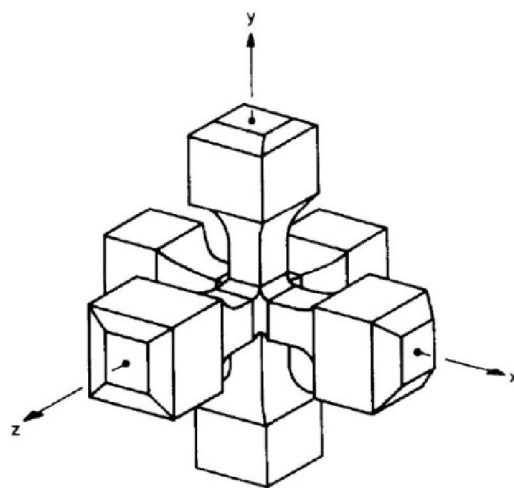


Figure 2.10: Triaxial tension creep specimen (Hayhurst & Felce, 1986).

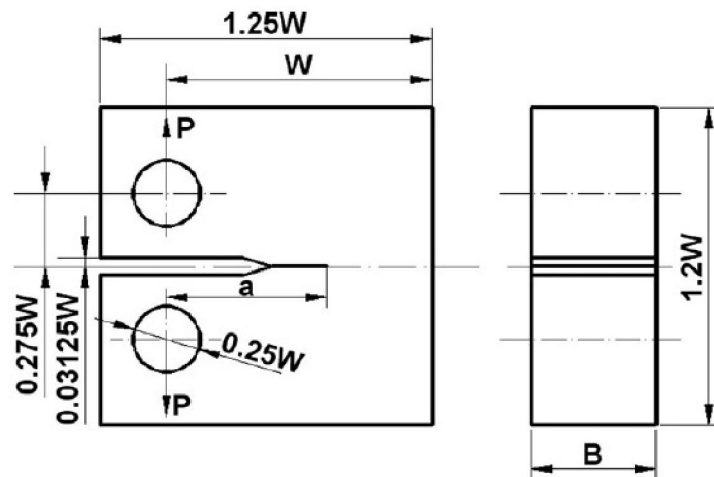


Figure 2.11: Geometry of a CT specimen from Wen et al. (2013) .

Various other specimens, designed for use in stress-state testing specifically for hot rolling applications such as, double collar, flying saucer, hexagon compression, octagon compression, upsetting and revised plane strain compression tests have been used and assessed for suitability by Corus RD & T, UK in a study by Farrugia (2006). From these the double collar, flying saucer and the plane strain compression tests were deemed most suitable for hot rolling applications. These specimens, although useful for assessing damage under several stress-states in one sample alone, create complications in metallographic studies as the interaction between the stress-states can make the distinction between damage mechanisms unclear. Additionally, if resistance heating is used, the non-uniform temperature fields that occur across the sample volume have to be considered carefully and accounted for.

In an effort to study the transition of stress-state from compressive to tensile, simulating hot rolling deformation conditions, Foster et al. (2009b) created a novel test called conical splay (see Figure 2.12). To avoid the non-uniform temperature fields this sample is heated first in a furnace and then transferred to the press for deformation. This test is useful as multiple triaxialities can be achieved in one sample but it is prone to cracks forming around the circumference of the outer ring. This would alter the stress triaxiality of the sample in a way that cannot be accounted for in the FE simulations, making the correlation of the microstructural examinations to the FE calculated stress-state difficult.

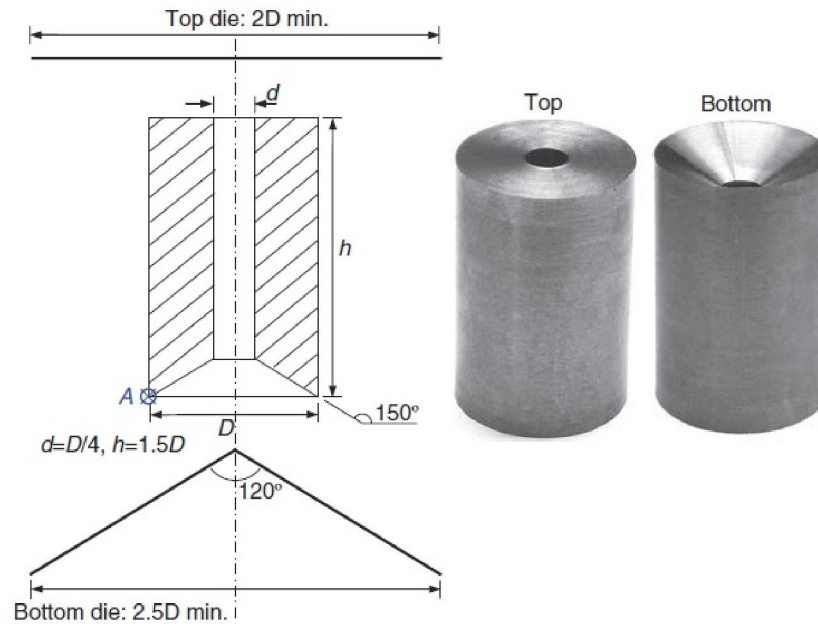


Figure 2.12: Schematic of the conical splay test and testing apparatus and the machined specimen where $D=40$ mm (Foster et al., 2009b).

2.3 DAMAGE MODELLING UNDER MULTIAXIAL LOADING CONDITIONS

2.3.1 BASIC CONCEPTS

Here the basic concepts of Continuum Damage Mechanics (CDM), based on the early works of Kachanov (1958) and Rabotnov (1969) who were the first to propose a damage variable for creep, are summarised. CDM material models can describe and predict microcrack initiation, growth and material failure without a detailed account of the microscopic deformation mechanisms that are involved. Damage is defined as a variable, often a tensor, and state variables are used to describe the effect of damage on the material behaviour in terms of dislocation density, loss of stiffness, microcrack density, etc. In order to predict material failure, a damage activation criterion and a damage evolution model are necessary.

Damage has been defined by Maire (2005) as the appearance of new surfaces in a material. However, damage can also include the enlargement of pre-existing voids such as microvoids created during the casting process or from porosity. When considering an element from a material section, damage can be described as the ratio of a damaged area of the element to a virgin,

original area prior to deformation where no voids exist. If A is the virgin material area and A_D the damaged area, then damage, ω_n , is defined as:

$$\omega_n = \frac{A_D}{A} \quad 2.2$$

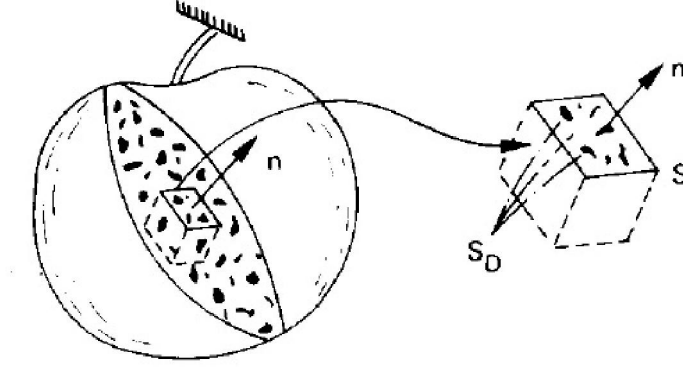


Figure 2.13: Definition of the damage element (Chaboche, 1986).

From this definition, for a virgin material with no voids ($A_D = 0$), $\omega_n = 0$; for a material at failure, where the material has fractured so the full area corresponds to damaged area ($A_D = A$), ω_n would be equal to 1. The damage variable can take values within the range $0 \leq \omega_n \leq 1$. For anisotropic materials where damage can progress in a different way depending on orientation, ω_n is a tensor. In the case of isotropic materials, damage is represented as a scalar variable, ω . For a material with voids the new effective area, A_E , of the material can be expressed as:

$$A_E = A - A_D = A - A \times \omega_n = A(1 - \omega_n) \quad 2.3$$

If this effective area, A_E , replaces the area, A , in the original stress definition:

$$\sigma = \frac{F}{A} \quad 2.4$$

Then it can be expressed as an effective stress, σ_E :

$$\sigma_E = \frac{F}{A_E} = \frac{F}{A(1-\omega_n)} = \frac{\sigma}{1-\omega_n} \quad 2.5$$

The same principle can be applied to Young's modulus, E . An effective Young's modulus, E_E , which would represent the elasticity of the damaged material, can be expressed as:

$$E_E = (1-\omega_n)E \quad 2.6$$

According to this, the elastic strain, ε^e , of a damaged material can be written as:

$$\varepsilon^e = \frac{\sigma_E}{E} = \frac{\sigma}{(1-\omega_n)E} = \frac{\sigma}{E_E} \quad 2.7$$

Damage can be measured in a number of ways depending on the material and the deformation route. Examples of this are measurements as a void volume fraction, a function of the change in Young's modulus or a definition according to the creep deformation rate. Determining the correct measure depends on the damage mechanisms that are important for the particular application.

2.3.2 FUNDAMENTAL CREEP DAMAGE EQUATIONS

As discussed earlier, Kachanov (1958) was the first to introduce the idea of a damage variable that lead to the development of the CDM framework that has since been used by many researchers. Kachanov's damage parameter, ω , is a scalar value that ranges from 0 and 1 representing virgin state and failure in the material respectively. Kachanov (1958) developed a simple empirical model for use in applications of creep deformation, in which, the evolution of damage was described only as a function of applied stress and the level of damage:

$$\dot{\omega} = \left[\frac{\sigma}{a_0(1-\omega)} \right]^\gamma \quad 2.8$$

where a_0 and γ are temperature dependent material constants. Although the simplicity of the model can be advantageous as no knowledge of the controlling deformation mechanisms is needed, this can also be a major limitation for the same reason.

To address this limitation, damage models corresponding to individual damage mechanisms encountered in creep have been developed. For example, damage due to multiplication of mobile dislocations was described in a model developed by Dyson and Mclean (1990) and Lin et al. (1993). This typically takes the form:

$$\dot{\omega}_1 = \eta(1 - \omega_1)^2 \dot{\epsilon}_c \quad 2.9$$

where ω_1 is damage due to mobile dislocations and η is a material parameter. In this case, the evolution of damage is proportional to the creep strain rate, $\dot{\epsilon}_c$. This model has been developed for nickel superalloys under the assumption that velocity of dislocations around γ particles can be treated as a diffusive drift.

Damage due to creep constrained cavity nucleation and growth has been described by Dyson and Mclean (1990) and Lin et al. (1993) as follows:

$$\dot{\omega}_2 = D \dot{\epsilon}_c \quad 2.10$$

where D is a material constant equal to $\epsilon_f/3$ according to Lin et al. (1993) where ϵ_f is the strain to failure in uniaxial tension. Here $\dot{\omega}_2$ describes grain boundary creep constrained cavitation and is linear to the creep strain rate and sensitive to the fracture strain which indicates the propensity of the material to develop creep constrained grain-boundary damage.

Although the aforementioned models are simple and easy in their calibration, the assumption that creep damage is controlled by only one mechanism is frequently not the case. More sophisticated models address this limitation by introducing multi-variable damage parameters that describe one or more material mechanisms according to their dominance over creep deformation for the investigated material. These mechanisms were summarised previously in section 2.1.6. Othman et al. (1993) developed a model that represents damage with two state variables that take into account active mechanisms in creep deformation such as grain boundary cavity nucleation and growth, and multiplication of mobile dislocations. Similar models (Dyson and Gibbons, 1987; Kowalewski et al., 1994) were formulated with varied descriptions of the chosen state variables that describe damage depending on the controlling mechanisms. This approach in modelling allows more flexibility in controlling the mechanisms that contribute most to material failure, but as always there is the trade-off of complexity that has to be considered since the model has to be validated against experimental results.

2.3.3 VOID GROWTH MODELS FOR DUCTILE DAMAGE

For ductile type fracture the process of void growth and coalescence was first modelled by McClintock (1968) and Rice and Tracey (1969). These micromechanical models treated the growth of an isolated defect represented by a cylinder in the McClintock model (Figure 2.14a) and a sphere in the Rice and Tracey model (Figure 2.14b), in a remote, perfectly plastic field.

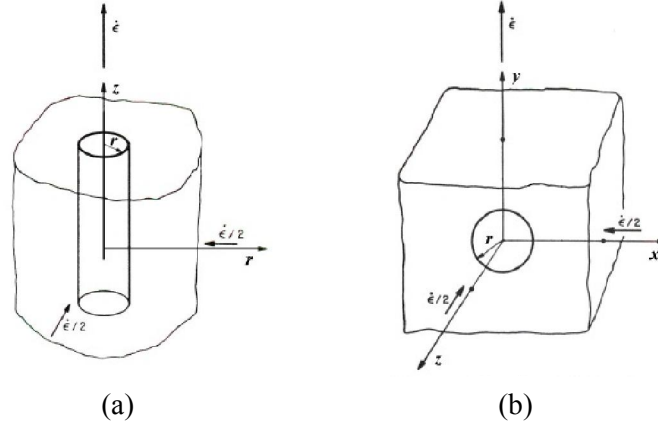


Figure 2.14: Void growth models of (a) McClintock and (b) Rice and Tracey. Modified from (Rice and Tracey, 1969).

In McClintock's work (1968) the failure criterion was defined as the point when the void radius equals half the mean spacing between voids or, in other words, when the walls of two adjacent voids come in contact. From this work it was established that the growth rate of the cylindrical void, $\dot{r}/r$, increases exponentially with positive transverse stress. Building upon this work, Rice and Tracey (Rice and Tracey, 1969) treated the same case in the more realistic scenario of a spherically shaped void. As in McClintock's work, the dependence of the void growth on the stress-state was evident. Under uniaxial tensile loading conditions the void growth according to the Rice and Tracey model is expressed as follows:

$$\frac{\dot{r}}{r\dot{\epsilon}} = 0.558 \sinh\left(\frac{3\sigma_H}{2\sigma_e}\right) + 0.008 \cosh\left(\frac{3\sigma_H}{2\sigma_e}\right) \quad 2.11$$

The second term of this equation could be omitted as it is typically a lot smaller in comparison to the first term. Therefore, void growth could be approximated by the following expression:

$$\frac{\dot{r}}{r\dot{\epsilon}} = 0.558 \sinh\left(\frac{3\sigma_H}{2\sigma_e}\right) \quad 2.12$$

Assuming constant triaxiality and independence of the initial, r_i , and final, r_f , void radii to triaxiality this relationship can be integrated to give the failure strain under multiaxial conditions, ϵ_f^* , as:

$$\epsilon_f^* = 1.792 \ln\left(\frac{r_f}{r_i}\right) / \sinh\left(\frac{3\sigma_H}{2\sigma_e}\right) \quad 2.13$$

Normalising with the uniaxial failure strain, ϵ_f , for which $\frac{\sigma_H}{\sigma_e} = 0.33$, gives:

$$\frac{\epsilon_f^*}{\epsilon_f} = 0.521 / \sinh\left(\frac{3\sigma_H}{2\sigma_e}\right) \quad 2.14$$

For higher triaxialities ($\frac{\sigma_H}{\sigma_e} > 0.8$) the growth rate expression is approximated by:

$$\frac{\dot{r}}{r\dot{\epsilon}} = 0.283 \exp\left(\frac{3\sigma_H}{2\sigma_e}\right) \quad 2.15$$

And the equivalent ratio of multiaxial to uniaxial failure strain is given by:

$$\frac{\epsilon_f^*}{\epsilon_f} = 1.65 / \exp\left(-\frac{3\sigma_H}{2\sigma_e}\right) \quad 2.16$$

Although the two studies described provided fundamental insight into the field of ductile damage modelling they both represent oversimplified cases of ductile void growth. One of the assumptions, for example, was the presence of a lone void in a remote plastic field so that void interaction is not taken into account. This limitation was addressed by Needleman (1972) who modelled the growth of an array of circular voids under plane strain conditions. In his work, the circular voids were arranged in a periodical fashion (far from a real-life representation) but this was an important start for the study of cluster effects on coalescence that were previously neglected.

Another limitation of the previous McClintock (1968) and Rice and Tracey (1969) models was that softening of the matrix was not taken into account. In an effort to advance void growth models, Gurson (1977) calculated an approximate yield surface for porous ductile materials which incorporated the effects of softening that were previously ignored. In this model, the dilatation of spherical voids in an incompressible matrix is simulated under axisymmetric stress-state conditions, taking into account pre-existing damage. This yield function was formulated in terms of the equivalent, σ_e , hydrostatic, σ_H , and flow, $\bar{\sigma}$, stresses of the matrix material as well as the void volume fraction, f , as follows:

$$\varphi(\sigma_e, \bar{\sigma}, f, \sigma_H) = \left(\frac{\sigma_e}{\bar{\sigma}}\right)^2 + 2q_1 f \cosh\left(\frac{3q_2 \sigma_H}{2\bar{\sigma}}\right) - 1 - (q_1 f)^2 = 0 \quad 2.17$$

For the original Gurson model, parameters q_1 and q_2 are equal to 1. If $f = 0$, representing a material with no voids, the model reduces to the conventional von Mises model. In this case, material deformation is homogeneous and loss in load carrying capacity is only achieved when void volume fraction takes its maximum value. Comparisons of the original Gurson model predictions to experimental data did show considerable discrepancies making it apparent that further improvements to the model were necessary (Besson, 2009).

Numerical studies on the void interaction effects during localised deformation were performed by Tvergaard (1981, 1982). These studies were an alteration to the original Gurson model, accounting for the unrealistically large void volume fraction prediction at the onset of material damage. In this modification the parameters q_1 and q_2 take the values 1.5 and 1 respectively (Koplik and Needleman, 1977). Although this improved the accuracy of the model, the reduction in the void volume fraction prediction was still not sufficient. To provide more flexibility and accuracy to the model, an additional void-coalescence criterion was incorporated. This was done by Tvergaard and Needleman (1984) who introduced an additional function, $f^*(f)$, thus shifting from homogenous to localised deformation. This function can be interpreted as an effective porosity. When a critical porosity value is reached, the coalescence process takes place thus promoting damage.

$$f^*(f) = \begin{cases} f & \text{if } f \leq f_c \\ f_c + \frac{f_u - f_c}{f_f - f_c} (f - f_c) & \text{if } f > f_c \end{cases} \quad 2.18$$

where, f_f is the void volume fraction at fracture or fracture porosity and $f_u^* = 1/q_1$. Fracture occurs when $f^*(f) = 1/q_1$. The addition of this function as well as the parameters q_1 and q_2 , collectively termed the Gurson-Tvergaard-Needleman (GNT) model, have somewhat shifted the micromechanical focus of this model to a more phenomenological one. Since the works of Tvergaard and Needleman, many other researchers have improved the Gurson model to account for various aspects such as; void shape effect (Gologanu et al., 1993), plastic anisotropy (Benzerga and Besson, 2001), void/particle interaction (Siruguet and Leblond, 2004), void size effect (Wen, 2005) etc.

2.3.4 STRESS-STATE MODELLING OF DAMAGE EVOLUTION

Since the pioneering models developed by McClintock and Rice and Tracey that identified the importance of stress-state and its effect on void growth (see section 2.3.3), many researchers have concentrated their efforts on understanding this relationship and developing models and fracture criteria that describe it sufficiently. These models can be based on cavity/void growth (Rousselier, 1987), nucleation or empirical relationships derived from fitting to experimental data.

Many multiaxial models have been developed for creep deformation and a convenient form of expressing the multiaxial ductility has been established as the ratio of multiaxial to uniaxial strain to failure, $\varepsilon_f^*/\varepsilon_f$; often termed multiaxial ductility factor (MDF) (M W Spindler, 2004). Ductile fracture criteria that assume fracture occurs when the accumulated plastic strain reaches a critical, material dependent value, C , have also been proposed. These take the following general form:

$$\int_0^{\varepsilon_f} f(\text{stress} - \text{state}) d\varepsilon = C \quad 2.19$$

where, f , is a function that includes components or invariants of the stress tensor, ε , is the equivalent strain, ε_f is the equivalent strain to failure and C is a material constant. In the following sections the reviewed multiaxial models have been described according to their dependence on triaxiality, σ_H/σ_e , maximum principal stress, σ_1 or equivalent stress, σ_e .

2.3.4.1 Stress triaxiality dependency

Cocks and Ashby (1980) developed a model of an array of spherical voids on a grain boundary, growing due to grain boundary diffusion mechanisms in a material deformed under power law creep. In this widely used model, the voids are assumed to grow without losing their spherical shape, such that high triaxiality is assumed. Damage due to continuum cavity growth, ω , is defined as the squared ratio of the void radius to the spacing between the voids, r^2/l^2 . From this model the ratio of the multiaxial to uniaxial strain to failure (or else the multiaxial ductility) is expressed as:

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \sinh \left[\frac{2}{3} \left(\frac{n-0.5}{n+0.5} \right) \right] / \sinh \left[2 \left(\frac{n-0.5}{n+0.5} \right) \frac{\sigma_H}{\sigma_e} \right] \quad 2.20$$

where, n , is the steady state creep exponent, a material constant ranging from 2-9. In the original model, it was observed that the predictions are not as accurate for low triaxialities, such as $\sigma_H/\sigma_e < 0.8$, and for creep exponents lower than 3. This was done intentionally so that high triaxiality predictions have a bigger bias towards uniaxial cases. A simple modification to this model was done by Wen and Tu (2014) to reduce this bias and produce accurate predictions across the full creep exponent range.

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \exp \left[\frac{2}{3} \left(\frac{n-0.5}{n+0.5} \right) \right] / \exp \left[2 \left(\frac{n-0.5}{n+0.5} \right) \frac{\sigma_H}{\sigma_{eq}} \right] \quad 2.21$$

Another creep void growth model was developed by Budiansky et al. (1982). In this study, void growth in a non-linear viscous material was considered. This was investigated for a spherical void, assuming power law creep, under various triaxialities and creep exponents, n . The multiaxial ductility for this model was described as:

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \left[(n+\zeta) / \left(3n \frac{\sigma_H}{\sigma_e} + \zeta \right) \right]^n \quad 2.22$$

where $\zeta = (n+1)(n+0.4319)$. This work also suggested that voids would retain their shape whilst growing at sufficiently high triaxialities ($1.5 < \sigma_H/\sigma_e < 3$), otherwise the voids would grow to be oblate shaped.

A popular fracture model that includes the effect of stress-state was developed by Johnson and Cook (1985). In this model that considered high strain rates and high temperatures, the critical equivalent strain to failure, ε_f , was defined as a function of triaxiality as such:

$$\varepsilon_f = \left[a_1 + a_2 \exp \left(a_3 \frac{\sigma_H}{\sigma_e} \right) \right] \left(1 + a_4 \ln \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right) \left(1 + a_5 \frac{T - T_0}{T_m - T_0} \right) \quad 2.23$$

where, a_1 , a_2 , a_3 , a_4 , a_5 , $\dot{\varepsilon}_0$ and T_m are material constants. This model has been mainly determined from high triaxiality tension tests so its validity is not confirmed for small or negative triaxialities. The expression used in this fracture model was first suggested by Hancock and Mackenzie (1976) and the notion that the strain to failure is a function of triaxiality can also be traced back to the void growth models of McClintock (1968) and Rice and Tracey (1969). The Johnson-Cook model has the advantages of simple formulation, easy calibration and wide availability of material constants but it does not provided a way of changing the triaxiality value in the loading process.

Dhar et al. (1996) extended the original model developed by Lemaitre (1985) to separately describe the nucleation of voids (first term of equation 2.24) and their growth (second term of equation 2.24) for AISI-1090 spheroidised steel. In this proposed model the void growth is expressed as a function of triaxiality but the term of void nucleation is not. According to this model the damage rate is:

$$\dot{\omega} = c \dot{\varepsilon}_p + (\alpha_1 + \alpha_2 \omega) \frac{K^2}{2E} (\varepsilon_p)^{2m} f \left(\frac{\sigma_H}{\sigma_e} \right) \dot{\varepsilon}_p \quad 2.24$$

where, c , α_1 , α_2 are material constants, K and m are hardening related parameters and function

$f \left(\frac{\sigma_H}{\sigma_e} \right)$ is defined as:

$$f \left(\frac{\sigma_H}{\sigma_e} \right) = \frac{2}{3} (1 + \nu) + 3 (1 - 2\nu) \left(\frac{\sigma_H}{\sigma_e} \right)^2 \quad 2.25$$

This model expresses damage as a function of both plastic strain and stress-state (i.e. triaxiality) which gives the model a phenomenological character. Nonetheless, the results of this model have

been verified from experimental results of uniaxial tests for only one temperature (1090 °C) (LeRoy et al., 1981).

Another, material independent, damage formulation based on the early work of Lemaitre (1985), that models the reduction of material ductility for increasing triaxiality was proposed by Bonora (1997) in the following form:

$$\omega = \omega_0 + (\omega_{cr} - \omega_0) \left\{ 1 - \left[1 - \frac{\ln(\varepsilon^* / \varepsilon_{th}^*)}{\ln(\varepsilon_f / \varepsilon_{th})} \cdot f\left(\frac{\sigma_H}{\sigma_e}\right) \right]^a \right\} \quad 2.26$$

where, ω_0 , the initial value for damage, ω_{cr} , the critical damage value for which damage takes place, ε_{th} , the uniaxial threshold strain for damage and ε_{th}^* , the multiaxial threshold strain are material parameters. Variable a is the damage exponent that defines the shape of the damage evolution curve and is related to the mechanisms of void nucleation and growth. Here $f\left(\frac{\sigma_H}{\sigma_{eq}}\right)$ is

the same term proposed by Lemaitre (1985), as seen in equation 2.25 above. All these material parameters have a phenomenological character and can be determined from uniaxial tension tests. This model was validated for several different materials and stress-states.

2.3.4.2 *Maximum principal stress dependency*

Other models have incorporated the effect of stress-state on damage via a ductile damage criterion based on the maximum principal stress, σ_1 . Cockcroft and Latham (1968) developed such a model for bulk forming operations based on the observation that ductile fracture occurs at sites where the tensile stress is maximum. This fracture criterion is only applicable for a range of small and negative stress triaxialities. The equation they proposed is as follows:

$$\int_0^{\varepsilon_f} \sigma_1 d\varepsilon = C \quad 2.27$$

Later Oh et al. (1979) modified the Cockcroft and Latham model to include the equivalent stress:

$$\int_0^{\varepsilon_f} \frac{\sigma_1}{\sigma_e} d\varepsilon = C \quad 2.28$$

Also building on the Cockcroft and Latham model (1968), Brozzo et al. (1972) included a dependence on the hydrostatic stress in the following model:

$$\int_0^{\varepsilon_f} \frac{2\sigma_1}{3(\sigma_1 - \sigma_H)} d\varepsilon = C \quad 2.29$$

2.3.4.3 Hydrostatic stress dependency

Manjoine (1975) proposed a form for the multiaxial ductility factor that was empirically derived from notched bar testing performed on type 304 steel at 593 °C:

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \frac{\sigma_e}{3\sigma_H} \quad 2.30$$

Later work by the same author (Manjoine, 1982) using notched bar testing on a range of steels gave the following empirical equation:

$$\frac{\varepsilon_f^*}{\varepsilon_f} = 2^{(1-3\sigma_H/\sigma_e)} \quad 2.31$$

Another phenomenological fracture criterion was developed by Norris et al. (1978) and was based mainly on plastic strain but was also weighted by hydrostatic stress according to this expression:

$$\int_0^{\varepsilon_f} \frac{1}{1 - c\sigma_H} d\varepsilon = C \quad 2.32$$

where C is a material constant. This fracture criterion was validated for uniaxial, notched-bar and pre-cracked compact tension tests at room temperature.

2.3.4.4 Mixed stress-state criteria models

The models by Rice and Tracey (1969), Manjoine (1975), Budiansky et al. (1982) and Cocks and Ashby (1980) employ only one mechanism to describe the growth of voids which makes them convenient to apply. However, in some cases it is essential to be able to describe the effect of stress-state where a transition between different mechanisms is observed.

Such a model was developed by Hales (1994) who considered plastic deformation, diffusion controlled cavity growth and constrained cavity growth. The proposed ductility factor for diffusion controlled cavity growth was given as:

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \frac{\sigma_e}{\sigma_1} \quad 2.33$$

And similarly for constrained cavity growth as:

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \frac{2\sigma_e}{3(\sigma_1 - \sigma_H)} \quad 2.34$$

Spindler (2004) also created a model for multiaxial creep ductility which considered the evolution of both cavity nucleation and growth as a function of hydrostatic, maximum principal and equivalent stress. This was expressed in the empirical equation below:

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \exp \left[p \left(1 - \frac{\sigma_1}{\sigma_e} \right) + q \left(\frac{1}{2} - \frac{3\sigma_H}{2\sigma_e} \right) \right] \quad 2.35$$

where p and q are constants that need to be determined using numerical regression techniques. The p parameter represents cavity nucleation and the effects of maximum principal stress on cavity growth. When $p = 0$ and $q = 1$ this model turns into the Rice and Tracey model (1969), describing only void growth. If $p = 0$ and $q = 2 \ln(2)$ then the model turns into the Manjoine (1982) model. Spindler (2004) derived empirical values for p and q for type 304 and 316 stainless steels using biaxial creep rupture tests. For the case of the 316 steel, for which creep strain at rupture is not a function of strain rate, $p = 0.15$ and $q = 1.25$. For the 304 steel the value of p increases to 2.38 and q decreases to 1.04 to account for the ductility dependency on strain rate. Although this model showed good agreement with experimental data, as the values for

p and q were fitted in an empirical way, it also demonstrates a physical basis. The model has been validated with biaxial test results reaching triaxiality values of about 0.7.

Another mixed criteria model for the prediction of creep damage based on a combination of maximum principal stress, σ_1 , and equivalent stress, σ_e , was proposed by Hayhurst (1983):

$$\dot{\omega} = \frac{[a(\sigma_1/\sigma_0) + (1-a)(\sigma_e/\sigma_0)]^v}{(1-\omega)^\eta} \quad 2.36$$

where, a , is a weighing parameter varying from 0 to 1 and was used to calibrate the damage dependency of the particular material to σ_1 or σ_e . For example, damage evolution in aluminium alloys is strongly linked to σ_e , therefore, in this case parameter a would be zero. For copper, where the damage evolution is mainly controlled by σ_1 , $a = 0.7$. This parameter makes the model formulation flexible for the prediction of damage in a range of materials.

More recently Lin et al. (2005) developed another stress-state creep damage equation based on a combination dependency for σ_1 and σ_e . This work was originally based on a model proposed by Kowalewski et al. (1994) for aluminium alloys and takes the following form:

$$\dot{\omega} = D \left(\frac{(\sigma_1 + |\sigma_1|)/2}{\sigma_e} \right)^\gamma \dot{\epsilon}_p \quad 2.37$$

where, D is a material constant and $\gamma = \beta\sigma_1$.

Here β is a material constant that is used to describe the stress-state dependency on the damage evolution and to model the lifetime and tertiary creep deformation behaviour of the material. The term $(\sigma_1 + |\sigma_1|)/2$ is used to prevent a negative damage rate in the case of a compressive σ_1 . In the case of a positive σ_1 , equation 2.37 becomes similar to the original Kowalewski model (1994):

$$\dot{\omega} = D \left(\frac{\sigma_1}{\sigma_e} \right)^{\beta\sigma_1} \dot{\epsilon}_p \quad 2.38$$

If $\beta < 0$, then the damage evolution for the particular material exceeds the effective stress control and a low value for σ_1 would reduce the creep lifetime. If $\beta > 0$ then the material damage evolution is controlled by σ_e . If $\beta = 0$ then the creep lifetime and tertiary creep deformation of the material are only controlled by σ_e . This model has been validated via tension-torsion creep tests for aluminium and copper.

Building upon this study, Lin et al. (2013) developed a plane stress CDM model to predict forming limit diagrams for aluminium alloys under hot stamping conditions. In this model, the effects of maximum principal, hydrostatic and effective stress on the overall material damage are included, meaning this model does not account for just one specific damage mechanism. The equation describing damage under multiaxial conditions in hot stamping was formulated as:

$$\dot{\omega} = \frac{\Delta}{(\alpha_1 + \alpha_2 + \alpha_3)^\varphi} \left\langle \frac{\alpha_1 \sigma_1 + 3\alpha_2 \sigma_H + \alpha_3 \sigma_e}{\sigma_e} \right\rangle^\varphi \cdot \frac{\eta_1 \sigma_e}{(1 - \omega)^{\eta_2}} (\dot{\epsilon}_p)^{\eta_3} \quad 2.39$$

In this equation, α_1 , α_2 and α_3 are weighting parameters that define the dependency of the particular material to the stress-state parameters, namely the maximum principal, hydrostatic and equivalent stress. Their values are defined according to experience and it has been determined that α_1 and α_3 range from 0 to 1, whilst α_2 ranges from -1 to 1 due to the fact that the hydrostatic stress can be negative. If either of these coefficients equals zero then the damage evolution of the material is not affected by the associated stress under hot stamping conditions. Exponent φ controls the combination effect from the three stresses on damage evolution and, thus, defines the shape of the forming limit diagram. The parameter, Δ , is a correction factor linked to the testing data obtained. Finally, η_1 , η_2 and η_3 are temperature dependent material constants. In this case, the model was verified using uniaxial tension data. This model provides flexibility as the damage evolution of a material can be described by a combination of stress-state parameters. This advantage is offset by the multiple material parameters that have to be determined from experimental data.

It is, therefore, clear that stress-state modelling has attracted much attention over the years, especially for creep conditions. Although it is more popular in creep applications, it was the McClintock and Rice and Tracey models, developed for ductile void nucleation, growth and coalescence, which highlighted the importance of stress-state and its effect on void growth. Transferability of these fundamental models between applications is not uncommon. For example,

the work of Hellan (1975) has proven that the Rice and Tracey model can be also used for creep applications such as cavity growth by power law creep. Additionally, research conducted by Lin et al. (2002) has proven that the Cocks and Ashby model can be used in superplastic conditions. The multiaxial ductility factor proposed by Cocks and Ashby has also been applied in hot forming damage modelling by Foster et al. (2009a). In the following section, the inclusion of stress-state factors in damage modelling applied to hot forming conditions will be further reviewed.

2.3.4.5 Stress-state modelling in high temperature forming

In damage modelling for the application of hot forming conditions, where stress-state has a significant impact on damage evolution, the triaxiality factor is often included as a way of incorporating stress-state effects. For example, the work of Nicolaou and Semiatin (2003) included this term in a model applied to damage occurring during secondary hot working or superplastic forming. This model was originally developed by Pilling and Ridley (1986) for aluminium, superplastic alloys based on experimental data. The empirical relationship they presented was formed as below:

$$\frac{\dot{\omega}^*}{\dot{\omega}} = \left(1 + 2 \frac{P}{\sigma_e} \right) \quad 2.40$$

where, P , is the hydrostatic pressure and corresponds to compression if negative and tension if positive. Nicolaou and Semiatin (2003) in an effort to quantify the effect of stress-state on cavitation, rewrote this expression to include the term, $\frac{\sigma_H}{\sigma_e}$ as follows:

$$\frac{\dot{\omega}^*}{\dot{\omega}} = \left(\frac{1}{3} + 2 \frac{\sigma_H}{\sigma_e} \right) \quad 2.41$$

This work was performed for a titanium alloy, Ti-6Al-4V, and tensile tests performed for a range of temperatures and strain rates were used to measure cavitation, corresponding to a specific stress-state. These measurements were then compared against predictions from the Rice and Tracey (1969) and Pilling and Ridley (1986) models to find that the first model performs well for high triaxialities and the latter for low triaxialities, such as compressive conditions.

A constitutive model, specifically developed for hot rolling of free cutting steel, was proposed by Liu et al. (2005). This model is able to describe dislocation build up and material hardening,

recovery and recrystallization, and also accounts for grain growth. Damage was divided in two terms; grain boundary, ω_{gb} , and plasticity induced damage, ω_{pi} , since these two types of damage often coexist in free cutting steels under hot rolling conditions.

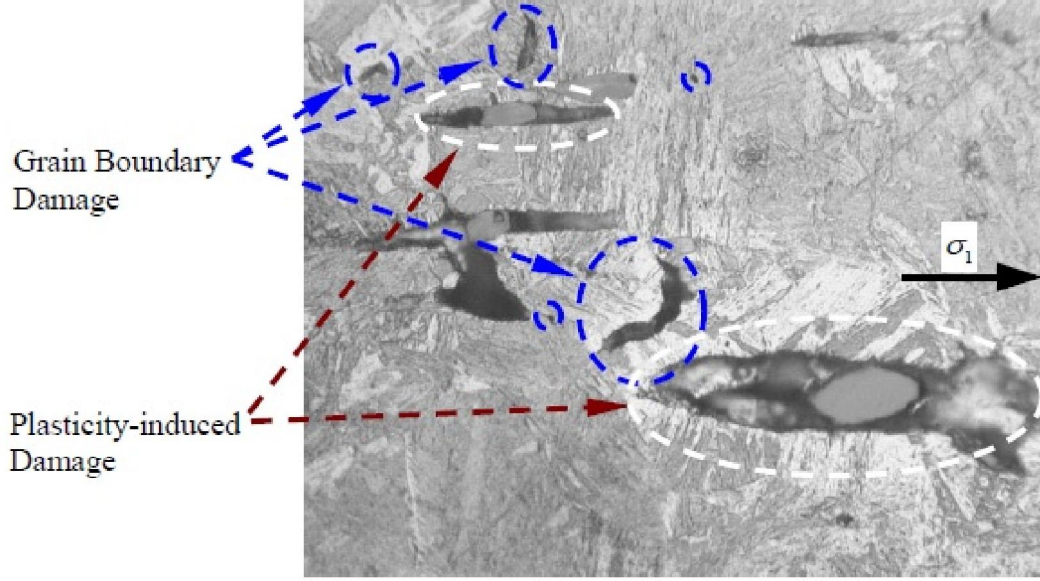


Figure 2.15: Co-existence of grain boundary and plasticity induced damage in free cutting steel (as-rolled) tested in tension at 1273 K (Liu, 2004).

This model describes damage in a phenomenological way for the nucleation and growth of damage as follows:

$$\dot{\omega}_{gb} = \alpha_4 \eta \left(\frac{\sigma_1}{\sigma_{eq}} \right)^{n_2} \left\{ \alpha_7 (1 - \omega_{gb}) \dot{\rho} + \left[\frac{1}{(1 - \omega_{gb})^{n_3}} - (1 - \omega_{gb}) \right] \dot{\epsilon}_p \right\} \quad 2.42$$

$$\dot{\omega}_{pi} = \alpha_9 \left(\frac{\sigma_1}{\sigma_e} \right)^{n_4} \left[(1 - \omega_{pi}) \dot{\rho} + \alpha_8 \frac{\omega_{pi} d}{(1 - \omega_{pi})^{n_5}} (\dot{\epsilon}_p)^{n_6} \right] \quad 2.43$$

$$\dot{\omega}_T = \dot{\omega}_{gb} + \dot{\omega}_{pi} \quad 2.44$$

where α_4 , α_7 , α_9 , α_8 , n_2 , n_3 and n_4 are material constants to be defined from experimental results. The terms η and d are used to describe the effects of grain size and strain rate on the evolution of grain boundary related damage. In this model η is defined as below:

$$\eta = \exp \left[-\alpha_5 \left(1 - \frac{d}{d_c} \right)^2 \right] \quad 2.45$$

where d is the average grain size, $d_c = \alpha_6 (\dot{\epsilon}_p)^{-n_1}$ and α_5 , α_6 and n_1 are material constants. In equation 2.42, the first term represents grain boundary damage nucleation and the second term grain boundary damage growth. Two mechanisms were proposed to model grain boundary damage nucleation; (a) grain boundary sliding and rotation, accommodated by diffusion and (b) vacancy condensation on grain boundaries (see Figure 2.16). The grain boundary growth term is strain controlled in this model.

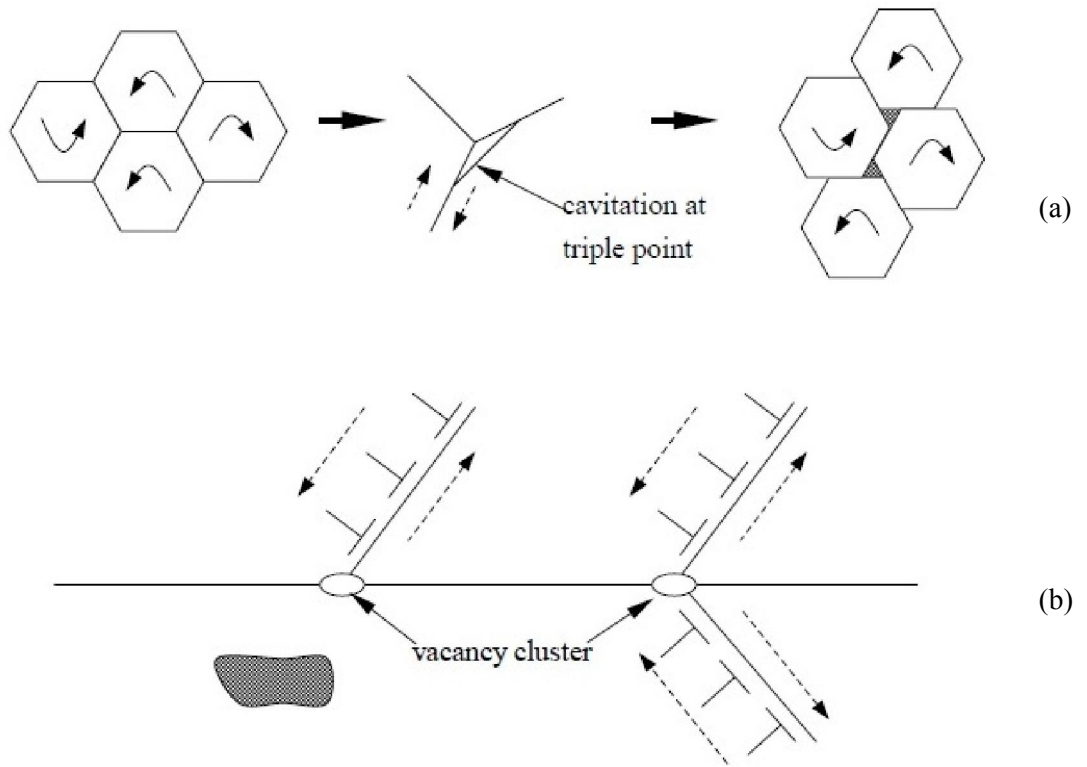


Figure 2.16: Possible mechanisms for grain boundary damage nucleation: (a) grain boundary sliding and grain rotation and (b) vacancy condensation (Liu, 2004).

Equation 2.43, which describes the plasticity induced damage evolution, is also separated in two terms. The first term represents the plasticity induced damage nucleation, hence the decohesion of the inclusion/matrix interface, and the second term represents the strain-controlled growth.

Coalescence is not included as a separate mechanism in this model but as accelerated growth of voids instead. By normalising the damage rate of this model with the damage rate of the uniaxial

tension case, the influence of stress-state on damage becomes clear and can be represented by the following, simple term:

$$\frac{\dot{\omega}^*}{\dot{\omega}} = \left(\frac{3\sigma_H}{\sigma_e} \right)^n \quad 2.46$$

where, n , is a material constant, taking the value of 2 for that particular application.

Later work by Foster et al. (2009a) aimed to further improve the proposed equation set by Liu et al. (2005). To do so, a simple stress-state dependent equation set was formulated. In this work recrystallisation was omitted from the described mechanisms as it was believed to complicate the underlying damage mechanisms (Lin et al., 2007). The stress-state dependency was added by means of the multiaxial creep ductility term developed by Cocks and Ashby (1982) and was examined with a novel mechanical test (conical splay test) specifically developed to assess the effect of a range of stress triaxialities on damage. In this model the damage initiation term is a function of stress-state but a coalescence term was also developed and coupled with the damage initiation term to create a higher stress-state dependency. The work has similarities to the work of Cocks and Ashby (1982) also in the definition of the inclusion size and distribution term, ξ . In Foster's work, $\xi = (d_{av}/l_{av})^2$, where d_{av} and l_{av} are the average inclusion diameter and average distance between inclusions. This term is very similar to the damage due to cavity growth proposed by Cocks and Ashby (1982), $\omega = r_h^2/l^2$, where r_h and l are the void radius and spacing between voids respectively. In Foster's work the voids grow at inclusion sites, parallel to the straining direction, and maintain the same cross-section as the original inclusion particle (see Figure 2.17).

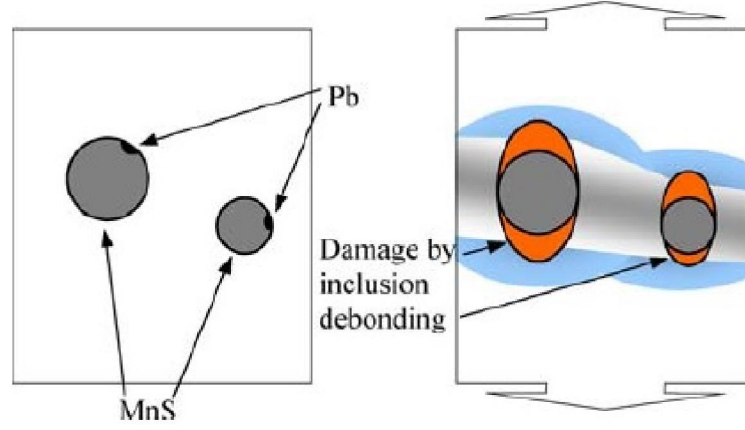


Figure 2.17: Inclusion related damage as proposed by Foster et al. (2009a).

According to the above assumptions, inclusion related damage was formulated as follows:

$$\dot{\omega}_i = z_1 \xi \dot{\epsilon}_p^\gamma \sinh \left(2 \frac{n_d - 0.5}{n_d + 0.5} \frac{\sigma_H}{\sigma_e} \right) / \sinh \left(\frac{2}{3} \frac{n_d - 0.5}{n_d + 0.5} \right) \quad 2.47$$

where, z_1 , is a material constant related to the initial damage area, γ , is an exponent used to bring time dependency into the equation and n_d is also a material constant. The damage coalescence rate is described as:

$$\dot{\omega}_C = z_2 \left\langle \frac{\sigma_1}{1 - \omega_C} \right\rangle_+ \xi \sinh \left(z_3 \sqrt{\rho} \omega_i \right) \dot{\epsilon}_p \quad 2.48$$

where, z_3 , is a material constant and z_2 is a temperature dependent material constant. Although an effort was made to describe the effect of stress-state on damage based on a phenomenological approach, the chosen testing method with the conical splay does not maintain a constant stress-state during the test due to the cracks that form around the outer edge of the test piece. It is therefore not a viable way to characterise the effects of stress-state on damage and a more controlled method, that is also easier to examine, should be utilised. Such a method could include notched tension tests with varying notch acuity ratios to induce different stress-states, as was done in this study.

This model was further developed by Kaye (2012) who proposed an equation set to describe the evolution of damage due to the growth of voids around inclusions, $\dot{\omega}_{pl}$, the growth of voids around grain boundaries, $\dot{\omega}_{gb}$, and their combined effect on material softening through coalescence, $\dot{\omega}_c$:

$$\dot{\omega}_{pl} = z_1 \Psi \frac{\dot{\epsilon}_p^\gamma}{\dot{\epsilon}_0^{1-\gamma}} \left(\frac{1}{3} + 2 \frac{\sigma_H}{\sigma_e} \right)^2 \quad 2.49$$

$$\dot{\omega}_{gb} = z_2 \frac{\dot{\epsilon}_p^{\gamma_2}}{\dot{\epsilon}_0^{1-\gamma_2}} \left(\frac{\sigma_1}{\sigma_e} \right) \quad 2.50$$

$$\dot{\omega}_c = z_3 \left\langle \frac{\sigma_1}{1-D_c} \right\rangle_+ \left\{ \exp \left(\frac{z_4 \frac{\dot{\epsilon}_p^{\gamma_3}}{\dot{\epsilon}_0^{1-\gamma_3}} \sqrt{\rho} \omega_{pl}}{L} \right) + \eta \exp(\omega_{gb}^{\gamma_4}) \right\} \quad 2.51$$

where, z_1 , z_2 , γ , γ_2 and γ_3 are material coefficients and Ψ is a phenomenological parameter to account for inclusion distribution according to:

$$\Psi = \psi_0 \left(1 + \frac{\ln \dot{\epsilon}}{\ln |\dot{\epsilon}_0|} \right) \left(1 - \frac{T - T_0}{T_0} \right) \quad 2.52$$

where, $\psi_0 = \left(\frac{D}{N} \right)^s$, D is the size of inclusions and N is the nearest neighbour for a specific inclusion distribution.

In this model an attempt was made to develop damage equations that carry more of a physical meaning, although this has led to a very complicated damage equation set. For example, in the damage equation set alone there are 10 material parameters that need to be calibrated. In the model a stress-state consideration was also included but since it was not calibrated using multiaxial stress-state data there is no evidence to whether this factor describes the stress-state influence on damage. Finally, the inclusion spacing parameter Ψ is based on calculations on inclusion clusters for the nearest neighbour. This means that a large amount of statistical data on the material's inclusion clusters need to be available which creates another difficulty for the

calibration of this model. In this case, the inclusion distributions were obtained using X-ray micro-tomography (Kaye et al., 2013).

2.4 CONCLUSIONS

Being able to account for multiaxiality is very important when assessing damage and creating a damage model as almost no structure is loaded under uniaxial conditions and hot forming is conducted under complex stress-state conditions. Therefore, being able to understand how damage initiates under various stress-states is crucial for the development of a constitutive model to be used under hot forming conditions.

Multiple specimen designs were reviewed in an effort to choose the most suitable for the study of damage initiation under various stress-states. It has been concluded that double notched samples are the most suitable for the aims of this project. This is for two reasons, first altering the design by varying the notch acuity ratio of the sample can produce multiple stress-states and second, by utilising the second or unbroken notch the damage initiation sites can be examined optically.

The advances in damage modelling under multiaxial loading conditions and damage modelling of hot forming conditions were also reviewed. CDM models developed for hot forming damage frequently include stress-state terms in their damage formulations but calibration of these models is often done with uniaxial data. Multiaxial damage models have to include a consideration for the stress-state influence on damage and be validated through multiaxial stress-state data.

CHAPTER 3

EXPERIMENTAL PROGRAMME

SUMMARY

In this chapter the material and testing procedures used for characterisation through thermal, tension and optical tests are presented. The samples and experimental set up used for the thermal, uniaxial and double notch bar tension tests are detailed. Additionally, the temperature and mechanical load testing conditions are presented and the optical examination methods are given. The data processing methods for generating the stress versus strain curves from the measured tension data, presented in Chapter 5, are also outlined.

3.1 MATERIALS

The material used in this study was a low carbon, leaded, free cutting steel (LFCS). All material was provided by Tata Steel UK in as-cast form. The as-cast material was provided in slabs cut from a near continuous cast billet of cross section 180x180 mm, 320 mm long (see Figure 3.1). The chemical composition of the billet material is shown in Table 3.1 below. All samples were machined in the same orientation, along the axis of rolling, to maintain the same micromechanical properties.

Table 3.1: Chemical composition of leaded free cutting steel (billet).

Element	C	Mn	S	Pb
Weight (%)	0.06	0.99	0.28	0.33

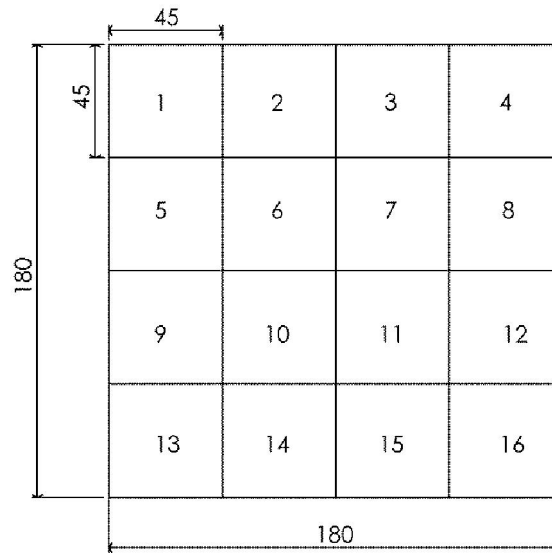


Figure 3.1: Billet cross-section schematic, as provided by Tata UK.

Free cutting steel (FCS) is extensively used in engineering applications due to its excellent machinability. This property arises from microstructural characteristics of the metal such as non-metallic second phase particles and lead additions (LFCS) which act as an inner lubricant due to their low melting point (Kishi and Eda, 1976; Shaw et al., 1961). The augmented chip formation of FCS during manufacturing helps extend tool life, improve productivity and lower energy consumption since minimal tool wear can be achieved at high machining speeds (Yaguchi, 1986).

However, there are drawbacks in using alloying elements to promote machinability. Edge cracking is often encountered in hot rolling of FCS, which leads to costly scrap (Hwang et al., 2005). Research on the thermal, mechanical and microstructural factors contributing to the nucleation, growth and coalescence of micro-defects in hot rolled FCS is therefore necessary to optimise the surface quality and consistency of the finished hot rolled products.

There is extensive evidence in the literature to show that the non-metallic manganese sulphide (MnS) inclusions found in FCS can act as stress raisers and nucleation sites at high temperature when a specific triaxial strain threshold has been reached, thereby inducing premature damage and eventually failure. This was first observed experimentally as edge cracking in hot rolled FCS by Baker and Charles (1972). It has since been confirmed experimentally (Kim et al., 2011; Maloney and Garrison, 1989; Sabirov and Kolednik, 2005) and studied using FE modelling (Foster et al., 2007, 2009b; Luo, 2001). The viscoplastic response of FCS has also been found to

change in response to parameters such as inclusion size and spacing that vary spatially across bloom sections (Foster et al., 2006; Kaye et al., 2013).

In this project, the focus of the study was on how the stress-state, induced from sample geometry, can affect the viscoplastic behaviour of the material. In order then to eliminate microstructural factors such as changes in second-phase particle size and distribution, all samples used in the tension experiments were machined from the same location in the billet. This location is labelled 13 in Figure 3.1.

3.2 TEST EQUIPMENT

The Gleeble 3800 thermal-mechanical testing system, shown in Figure 3.2, was used for both thermal (see section 3.3) and high temperature tension (section 3.4) testing performed in this work. This machine was chosen as it is capable of fast and accurate heating of samples to very high temperatures. To achieve this, the Gleeble uses resistance heating where low voltage, alternating current is passed through the sample to heat it up. The temperature control is achieved as follows. A controlling thermocouple, TC 1, is welded on the surface at mid-length of the tested sample, thus, measuring the temperature in real time. This information is instantly compared, at every time step, to the temperatures prescribed in the thermal cycle which is input by the user. The electrical current applied is then controlled appropriately to achieve an accurate match between the prescribed and real time temperature. Throughout the heating stage, and the mechanical loading, the water cooled Gleeble jaws are in constant contact with the grips that hold the sample (see Figure 3.3). This way only the gauge of the sample, where the deformation is concentrated, is heated to the prescribed temperature. This cooling method is very effective but leads to axial symmetric, for uniaxial samples, temperature profiles. These will be examined in further detail in Chapter 4.

Apart from fast and accurate temperature control the Gleeble can also provide the advantage of conducting tests in vacuum. This minimises the oxidation on the sample surface and heat loss via convection. Avoiding oxidation of the sample is particularly important in high temperature testing as surface scaling can interfere with the controlling thermocouple weld, causing it to become detached. When temperatures are as high as 1200 °C, K-type thermocouples are used to achieve accurate temperature readings. The Gleeble allows for a maximum of four thermocouples to be used in one experiment, which is useful for observing thermal profiles that are often present in Gleeble high temperature testing.

A C-gauge transducer was used to measure the diameter reduction, Δd , of the sample during testing. The C-gauge was mounted at mid-length of the sample for uniaxial samples and at one of the two notches for the double notched samples. From this measurement average axial stress and strain were then calculated as follows:

$$\sigma = \frac{4F}{\pi(d + \Delta d)^2} \quad 3.1$$

$$\varepsilon = 2 \ln \left(\frac{d}{d + \Delta d} \right) \quad 3.2$$

where, d is the initial diameter, Δd is the diametrical change as measured from the C-gauge and F is the force as measured by the Gleeble load cell.

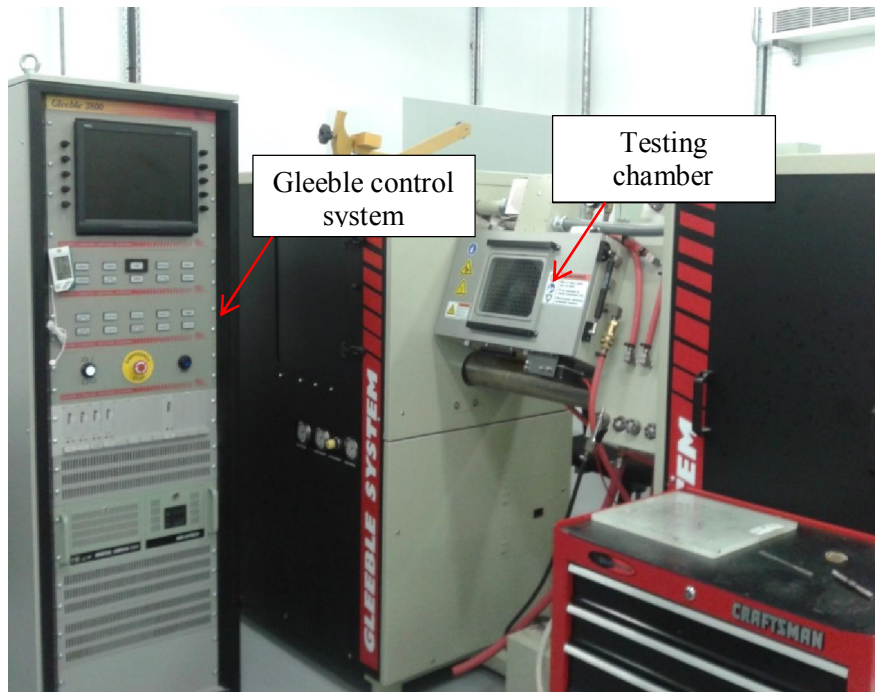


Figure 3.2: The Gleeble 3800 thermal-mechanical testing system.

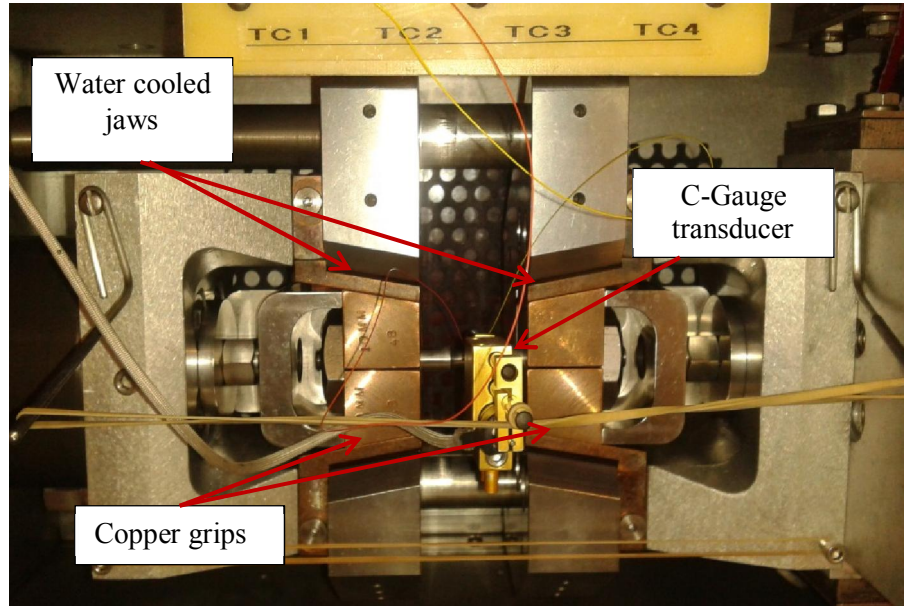


Figure 3.3: Sample set up inside the Gleeble testing chamber.

3.3 THERMAL TESTS

All test specimens used in this investigation are axisymmetric cylinders and follow the standards for use in Gleeble tension tests. For this reason, all samples feature an outer nominal diameter of 10 mm and are 130 mm in length with threads at both ends. A line drawing of the specimen geometry used in the thermal testing is shown in Figure 3.4. The gauge length is 6 mm and so is the radius of the gauge; $a=3$ mm. The shoulder radius is 10 mm, designed to achieve a higher triaxiality ratio of 0.47 as opposed to 0.33 which is the typical ratio for plain dog-bone specimens.

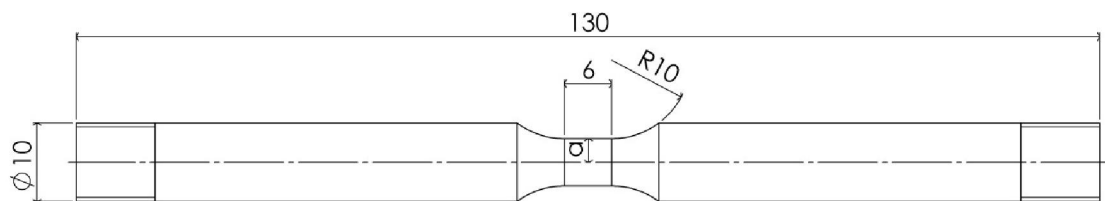


Figure 3.4: Dog-bone sample design for use in Gleeble thermal testing where the reduced gauge radius, $a=3$ mm.

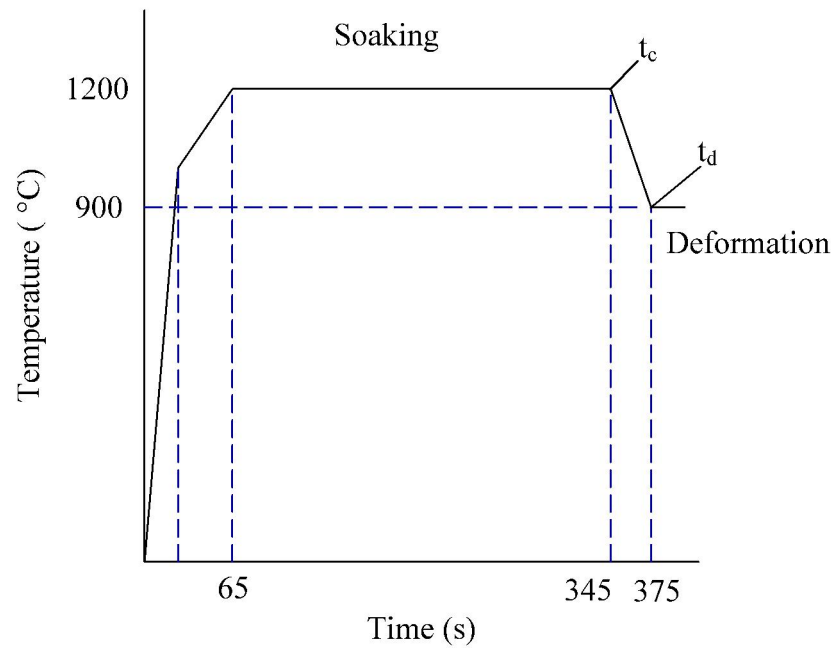


Figure 3.5: Schematic of a typical prescribed thermal cycle for high temperature tension tests. Annotated are the end of the soaking, t_c and the onset of deformation, t_d .

Thermal tests were performed to determine the temperature profiles induced along the length of this specimen geometry. For these tests the thermal profile shown in Figure 3.5 was used and the specimens were not mechanically loaded. Four thermocouples were welded on the sample surface at specified positions shown in Figure 3.6b. TC 1, the controlling thermocouple, is positioned at the mid-span of the sample whilst TM 2, TM 3 and TM 4 are welded 3, 6 and 12 mm away from the centre position respectively on one side of the specimen, since the heating is axisymmetric. The complete set up of the specimen, thermocouples and grips is shown in Figure 3.6.

The same general method for identifying temperature gradients was used in the uniaxial and double notched samples. In the case of the uniaxial sample, the controlling thermocouple, TC 1, was welded at the centre of the gauge area. This was not the case for the double notched samples as in order to achieve an axisymmetric thermal profile, TC 1 could not be applied in any one notch. Due to this, the prescribed temperature and the notch deformation temperature were slightly different. To find out what this temperature difference was, thermal tests were carried out on the double notch samples according to the method outlined here. Knowledge of this temperature difference was then used to overestimate the thermal cycle so that the notch temperature corresponded to the deformation temperature.

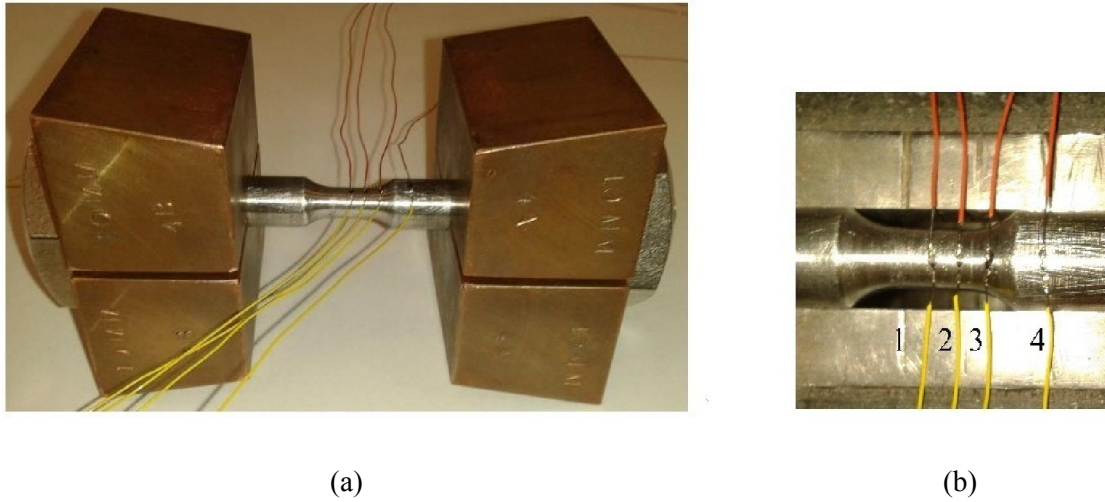


Figure 3.6: Thermal tests set up with (a) all four thermocouples attached and (b) the positioning of the thermocouples where TC 1 is at the centre (0mm), TM 2 is 3, TM 3 is 6 and TM 4 is 12 mm from the centre.

3.4 UNIAXIAL AND NOTCHED BAR TESTING

For the purpose of tensile testing, specimens designed according to British standards, shown in Figure 3.7a, and two variations of the general template for double notched samples, shown in Figure 3.7b, were used to achieve a different stress-state. The two double notch sample geometries were labelled DN_R5 and DN_R1.25 according to their notch profile radius, shown in Figure 3.7c. Both uniaxial and double notched samples are designed to have the same radius and hence notch depth, for the reduced gauge.

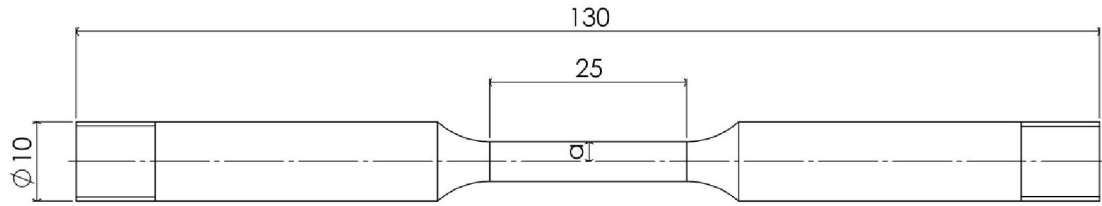
A blunt notch generates a small notch acuity ratio ($a/2R$) of 0.25 whilst a sharper notch generates a higher one of 1. This translates to triaxiality values of 0.55 and 1.02 for the blunt and the sharp notch, respectively, as calculated by the Bridgman triaxiality formula (Bridgman, 1952) for the centre of the necking zone and the initial value of R :

$$\frac{\sigma_H}{\sigma_e} = \frac{1}{3} + \ln\left(1 + \frac{a}{2R}\right) \quad 3.3$$

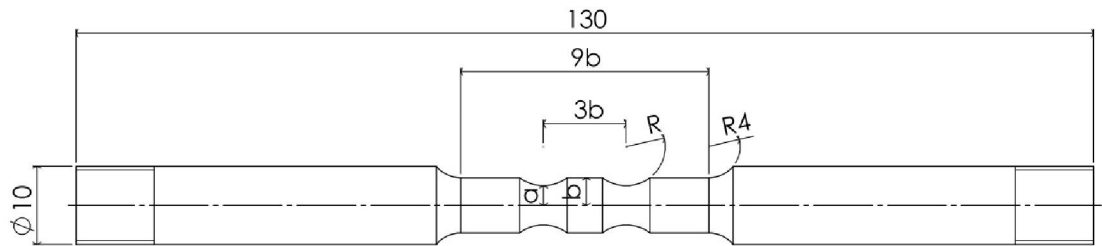
where, σ_H , is the hydrostatic stress and σ_e the equivalent stress.

The triaxiality values chosen here (see Figure 3.8) represent the upper bound of the triaxiality range encountered in long product hot rolling which varies from 0 to 1 for tensile triaxial states

(Farrugia, 2006). In Figure 3.8 a notch acuity ratio of 0 represents the case of the uniaxial sample which generates a triaxiality of 0.33.



(a) Uniaxial (ASTM) sample design



(b) Double notched (DN) nominal design



(c) Double notched (DN) sample dimensions for samples DN_R5 (left) and DN_R1.25 (right)

Figure 3.7: Sample designs. The notch spacing is kept constant along with the notch depth ratio, b/a . Notch diameter, $2a = 5$ mm, shank diameter, $2b = 7.05$ mm.

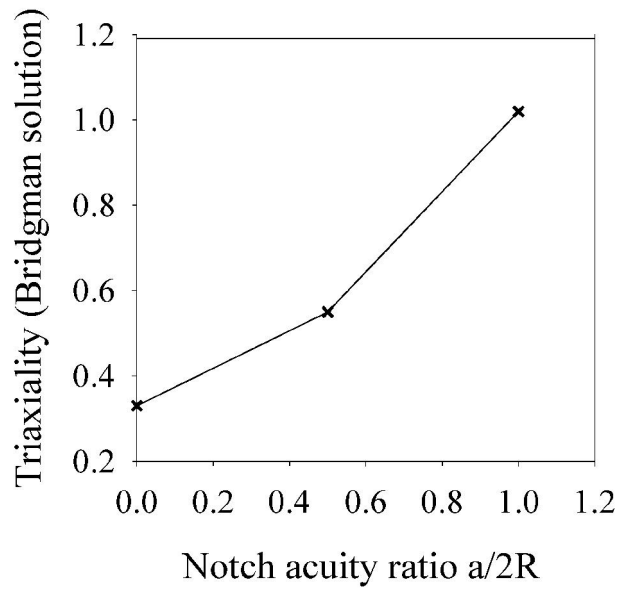


Figure 3.8: Triaxiality values, as calculated by the Bridgman equation (equation 3.3), for each sample used in the tension tests.

The general double notched template design (see Figure 3.7b) has been kept the same for both samples DN_R1.25 and DN_R5 ensuring that the notch spacing and notch depth ratio (b/a) remain constant. In particular, by keeping the notch depth ratio at approximately $\sqrt{2}$ and the notch spacing at $3b$, as defined in Figure 3.7b, the possibility of stress field interaction from the two notches is eliminated. This has been proven by Al-Abed et al. (1999) and Webster et al. (2004) who calculated the stress interference between the two notches on this nominal double notch template design for a maximum temperature of 650 °C.

In order to ensure that the use of this template design for double notched bars would not generate stress field interference in the higher temperatures applied in this work (900-1200 °C), FE simulations of the two double notched designs were performed for the highest temperature of the applied range, namely 1200 °C. For the purpose of these simulations, one full notch and a half notch were simulated for each sample as this would reveal any possible stress interference at the shank of the sample, between the two notches.

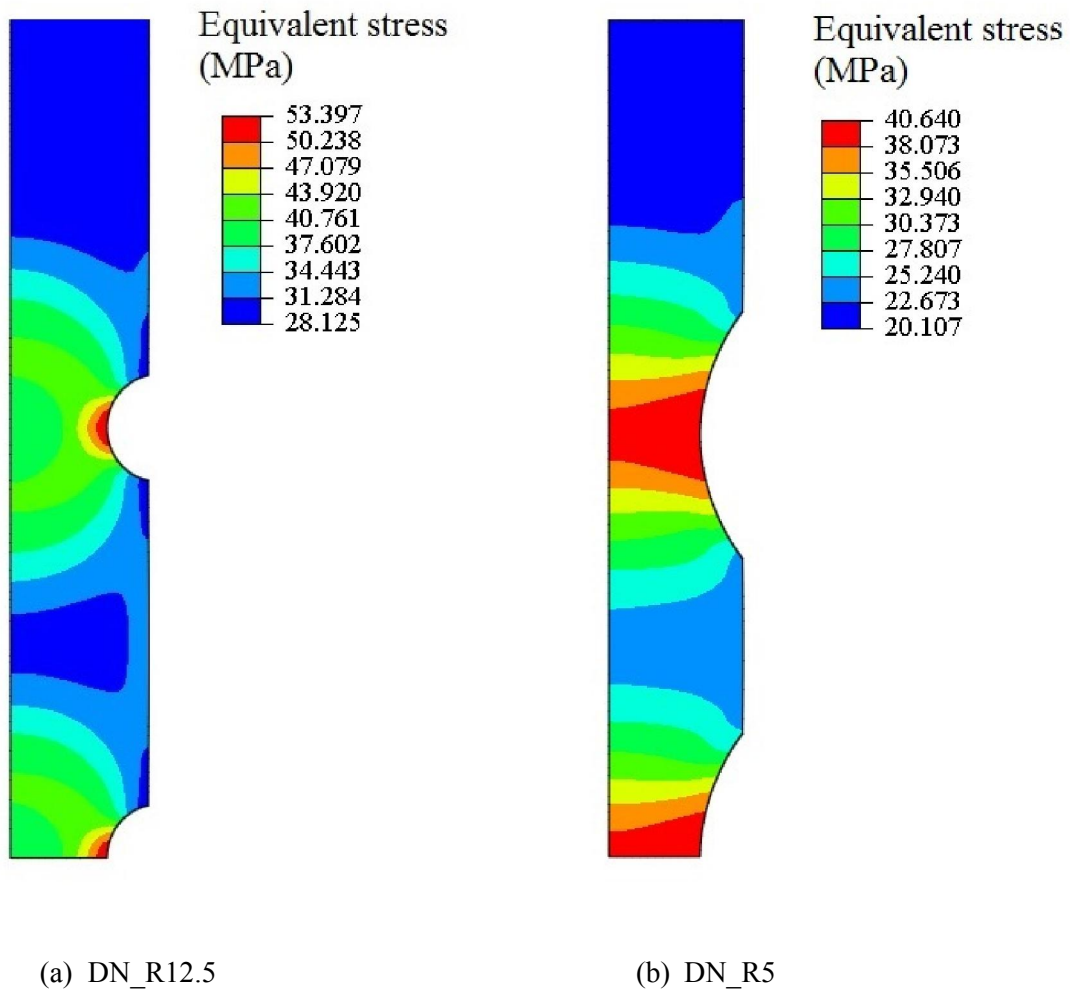


Figure 3.9: FE contours for equivalent stress at a temperature of 1200 °C and a strain rate of 0.1s⁻¹ for (a) sample DN_R12.5 and (b) DN_R5. The results here represent a strain of 0.1.

In these simulations no damage was included as only the stress-state was of interest. The material model and simulation methods detailed in Chapter 6 were used to generate the results of Figure 3.9. Figure 3.9 shows the FE results for the equivalent stress from simulations carried out on both double notched designs at 1200 °C at a strain rate of 0.1 s⁻¹. The results of Figure 3.9 are plotted for an axial strain of 0.1 and show that no significant stress interference is present between the two notches. This verifies that both samples are suitable for studying the viscoplastic response of the material at high temperatures since it was proved that measurements from the C-gauge are accurate from either of the two notches in a sample, up to just before failure occurs. This is useful as in double notch testing the C-gauge was positioned at the right hand side notch in all tests for consistency; there was no way of knowing which notch would fail.

3.5 THERMAL AND MECHANICAL CONDITIONS

A schematic of a typical prescribed thermal cycle used in the high temperature tensile tests is shown in Figure 3.5 (see section 3.3). This is an example chosen to simulate the thermal conditions a workpiece undergoes in the first few passes of the roughing stage during primary rolling.

Initially the sample is heated up to 1000 °C at a rate of 40°C/s and then at a much slower rate (5 °C/s), to avoid overshoot, the temperature is risen to 1200 °C. The specimen is then held for 4 minutes at 1200 °C so that a uniform austenitic microstructure is achieved (soaking phase). Subsequently, the test piece is cooled down to the prescribed target test temperature, 900 °C at a rate of 10 °C/s. The sample is held at this temperature for the onset and the entire duration of the mechanical loading. Natural cooling follows failure of the sample.

As mentioned above, the thermal parameters were selected so that deformation of material entering the initial roughing mills is physically simulated. A workpiece is usually heated to about 1250 °C before entering the first roughing mills. During the pass between mills this temperature reduces, primarily due to conduction from contact with the rolls and convection by air cooling; radiation cooling also takes place. It is therefore the edges of the workpiece that loose the most heat with temperatures there dropping by up to 300 °C. The temperature of the bulk of the material, on exit from the roughing mills is approximately 900 °C. Hence the temperature range of 1300-900 °C is of interest for the study of edge cracking in hot rolling conditions. Due to testing difficulties such as adherence of thermocouples for the duration of the test, (usually 6 minutes) the temperature range was adjusted to 1200-900 °C instead.

To ensure that the sample was deformed at a constant strain rate, a continuously changing velocity profile was used. The calculation of this profile was based on the effective gauge of the sample; the heated portion of the specimen. Thermal tests are essential to define this length prior to calculating the velocity profiles and, therefore, the required strain rate for each test.

For the mechanical loading, a strain rate range of 0.1-10 s⁻¹ was selected as this includes typical deformation rates encountered in hot rolling deformation conditions. Although this strain rate range was appropriate for the uniaxial tension tests performed, it was reduced to 0.01-1 s⁻¹ for the double notched tension tests as the highest strain rate of 10 s⁻¹ was not suitable. In such fast tests, if the C-gauge is not placed on the failed notch there is not enough time for the unbroken notch to deform sufficiently so that the measurement is higher than the error margin. Additionally, in the

faster tests, the unbroken notch does not have enough time to develop damage, which means that damage initiation and failure are too close together to identify separately.

3.6 METALLOGRAPHIC PREPARATION AND TECHNIQUES

Double notched samples, after failure, were subjected to optical examination so that the effect of the stress-state on microstructure could be observed. For this reason, the non-failed notch was examined, where the material was captured in a pre-failure state, after being deformed under various thermal and mechanical conditions. This was to provide information on the damage initiation sites and micromechanical mechanisms taking place at a close to failure condition of the material.

The specimens were sectioned longitudinally along their extended gauge section to approximately 0.1 mm off the gauge centreline, as shown in Figure 3.10. Subsequently they were mounted in Bakelite, ground, diamond paste polished, up to 1 μm , and finally etched in a solution of 2% Nital to reveal their grain boundaries. The optical examination was performed in two stages as the samples were first examined as-polished to get a general idea of the areas damage is concentrated at in relation to the testing conditions and the stress-state and afterwards etched and then examined again to acquire a more detailed view of damage and the interactions with second phase particles and grain boundaries.



Figure 3.10: Axial sectioning of double notched samples for optical examination of the unbroken notch.

CHAPTER 4

PREDICTION OF THERMAL FIELDS AND ERROR ANALYSIS IN RESISTANCE HEATING

SUMMARY

This Chapter focuses on the simulation of Gleeble tensile tests of leaded FCS at high temperatures. Firstly, the results from the thermal Gleeble tests are presented. Then, a coupled electro-thermal model, implemented in ABAQUS 6.11 was used to facilitate the prediction of transient temperature distributions along the specimen. The calculated thermal data were validated against experimental results. The temperature data generated from this model serve as input for subsequent mechanical simulations of high temperature tensile tests, also implemented in ABAQUS 6.11. The mechanical simulations used a set of unified temperature-dependent constitutive equations to describe the viscoplastic behaviour of the leaded FCS. A more accurate representation of the phenomena taking place during a high temperature Gleeble tensile test is therefore modelled by means of two simple FE simulations. The effect of the longitudinal thermal gradients present in Gleeble high temperature tests was assessed by comparing the results of a commonly performed isothermal FE tensile analysis to one where the FE calculated axial thermal profile was considered.

4.1 THERMAL GRADIENTS IN GLEEBLE TESTING

Axisymmetric high temperature tensile or compressive tests are widely used to physically simulate the complex phenomena taking place during hot rolling of metals (Bao, 2005; Li et al., 2011) and characterise ductile damage (Needleman and Tvergaard, 1984). The Gleeble thermal-mechanical testing system has now become a standard machine for this type of testing as it can combine accurate high rate heating and mechanical loading in a vacuum environment (Gomez and Perez, 2002; Pietrzyk and Jedrzejewski, 2001).

Axial, but also, radial thermal gradients are a common phenomenon, well documented by many researchers during Gleeble high temperature tensile tests (Kim et al., 1996; Changli Zhang et al.,

2011). Depending on specimen geometry, this could cause heterogeneous deformation during testing due to the temperature dependent material flow properties. With this in mind, accurate numerical modelling of Gleeble tension tests at elevated temperatures requires knowledge of the induced surface temperature distribution.

Although thermal gradients can be somewhat minimised by using suitable sample geometries (Norris and Wilson, 1999) or a specific jacket around Gleeble tensile specimens (Misra and Nagaraju, 2004), it is hard to completely eradicate them. For this reason, finite element models able to predict the temperature versus time response of tensile specimens under Gleeble thermomechanical loading have been previously developed (Brown et al., 1997; S. Norris and Wilson, 1999). Despite the important contribution of these studies, they only examine the electro-thermal aspect of the problem. The complex electro-thermomechanical nature of Gleeble tensile tests is often not taken into account, with studies usually assuming a uniform temperature distribution when performing FEM studies (Liu and Lin, 2003; Wang et al., 2006).

A more comprehensive study to evaluate the effect of Gleeble related errors such as initial specimen temperature profiles but also interface friction and heat generation was performed by Bennett et al. (2010) on axisymmetric compression tests. Zhang et al. (2010) also developed a coupled electro-thermomechanical model of the Gleeble tensile tests for inverse identification of the constitutive parameters. This model predicted the formation of hot tears at the end of solidification and of cracks at lower temperatures in the solid state of ultra-high-strength steel. The study pointed out the importance of accounting for thermal gradients, inherent in Gleeble testing, for accurate rheological material characterisation. Liu et al. (2005) also considered the problem of Gleeble thermal gradients on FCS and incorporated them in numerical studies on the effect of necking and the resulting stress-state. The limitation of this study lies in the fact that experimental thermal gradient data are needed to perform the numerical analysis of the hot tensile test. This requires a validation process involving multiple thermal tests to record the axial thermal gradients and is therefore inefficient. What is more, the thermal gradients have to be input manually using partitions in an FE model which can affect the mesh quality and in turn the simulated results. Nonetheless, limited research has been conducted to fully investigate the effect of the reported thermal gradients on the stress-state of tensile samples.

4.2 THERMAL TESTS RESULTS

4.2.1 STANDARD DOG-BONE SAMPLE

Strong axial thermal gradients were observed, as expected, for the standard dog-bone sample (see section 3.3) from the thermal tests. Figure 4.1 shows the experimentally recorded longitudinal temperature profiles at two instances in the thermal cycle; $t_c = 230$ sec and $t_d = 330$ sec representing the end of the soaking period and the beginning of the deformation period respectively, as defined in Figure 3.5 (see section 3.3). Symbols represent the recorded data whilst the lines represent the best fit. The data in Figure 4.1 were measured along half of the sample due to its axisymmetric geometry.

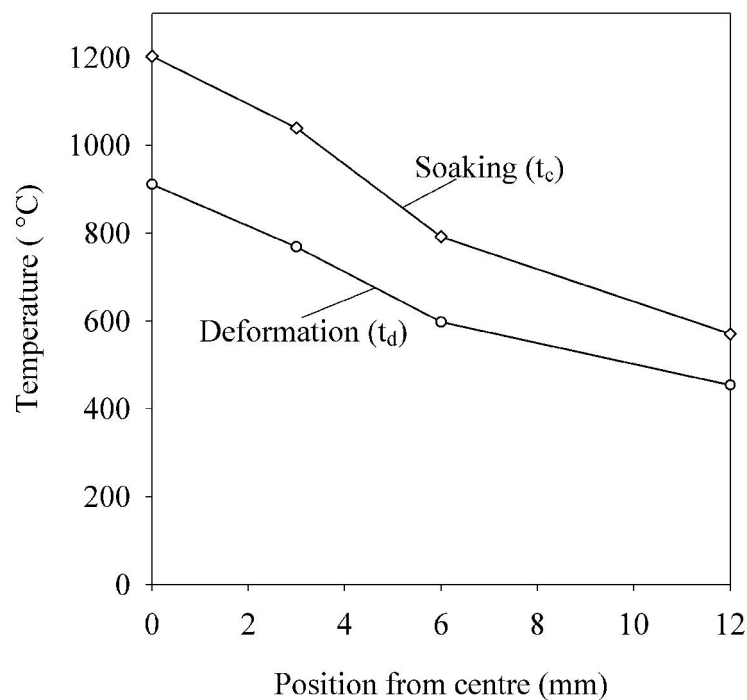


Figure 4.1: Axial thermal profiles for times t_c and t_d .

The transient thermal response obtained from all four thermocouples is also presented in Figure 4.2. It is noted that the temperature distribution observed along the gauge (0-3 mm) is non-uniform whilst outside of the gauge area, temperature drops slowly. The observed thermal profile results deviate from the theoretical temperature profile which assumes uniform heating of the gauge region with the rest of the sample at considerably lower temperatures. In the

aforementioned case all deformation takes place in the gauge region under uniform thermal conditions.

Hence, tensile testing using a Gleeble machine can result in deformation of different regions of the material in and near the gauge by different amounts. This is due to the different flow properties the material has for the differently heated regions of the gauge, together with the respective microstructure gradient and features (e.g. inclusions) that may be present. The strain rate applied to the tensile sample would also be non-uniform during the test. The findings of this experiment are conclusive to the fact that potentially inaccurate and misleading mechanical response data can be reported if modelling is performed using isothermal conditions along the gauge length, in high temperature Gleeble test simulations.

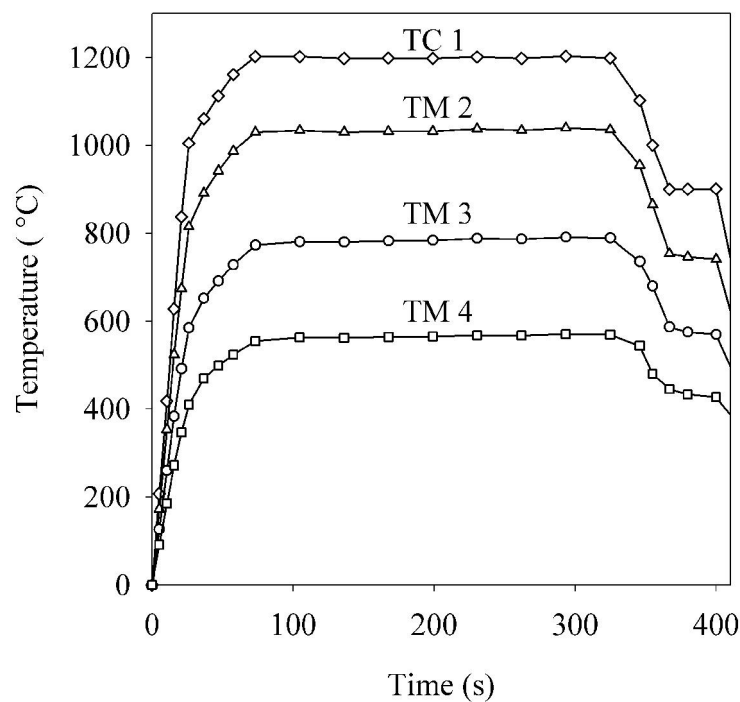


Figure 4.2: Transient response from TC 1, TM 2, TM 3 and TM 4.

4.2.2 DOUBLE NOTCHED SAMPLES

Following the same method outlined in Chapter 3, section 3.3, thermal tests were performed on double notched samples so that the effective gauge length is identified. Additionally, since the deformation of the two notches is of interest, the temperature drop between the centre of each

sample, where the controlling thermocouple TC 1 is welded (to achieve axisymmetric heating), and the centre of each notch had to be identified.

Initially, copper grips were used to perform a thermal test on double notched sample DN_R5 to identify if the grips can provide the required heating in both notches. The thermal test was performed for the lowest temperature of the studied range, 900 °C. Four thermocouples were welded on the sample in the following locations; TC 1 on the sample centre, TM 2 and TM 3 on the centres of each notch and TM 4 on the outer edge of the right hand side notch. Sample DN_R5 was used for the grip selection test as it features a longer gauge. The transient thermal response from the four applied thermocouples is shown in Figure 4.3. It is obvious that the copper grips are not suitable for this application as there is a temperature difference of 100 °C between the centre of the sample and the centre of each notch. Additionally, the temperature difference reaches more than 300 °C between the sample centre and the outer edge of one of the notches. It is also worth noting that the measured temperature responses from TM 2 and TM 3, which correspond to the centres of the two notches, should be identical here and the slight difference in the reading is attributed to experimental errors in the thermocouple positioning. From these results it becomes clear that a different set of grips should be used for the double notch tests, thus, steel grips were used instead. Steel grips allow for less heat transfer between them and the water cooled plates, extending the heated area.

The temperature history of the sample when the test was set up with steel grips is shown in Figure 4.4. In this case, thermocouple TM 2 measures the temperature at the centre of the right hand side notch and TM 3 measures the temperature at the outer edge of the same notch. An additional thermocouple TM 4 was also positioned on the gauge area but fell during the test. From the results of Figure 4.4 a small temperature difference of 30 °C is observed between the readings of TC 1 and TM 2. Therefore, to achieve the target test temperature at the notch centres for this sample, the thermal cycle was overestimated by this difference of 30 °C. A very similar thermal profile was identified for double notched sample DN_R1.25 with also a 30 °C temperature difference between the sample centre and the notch centre.

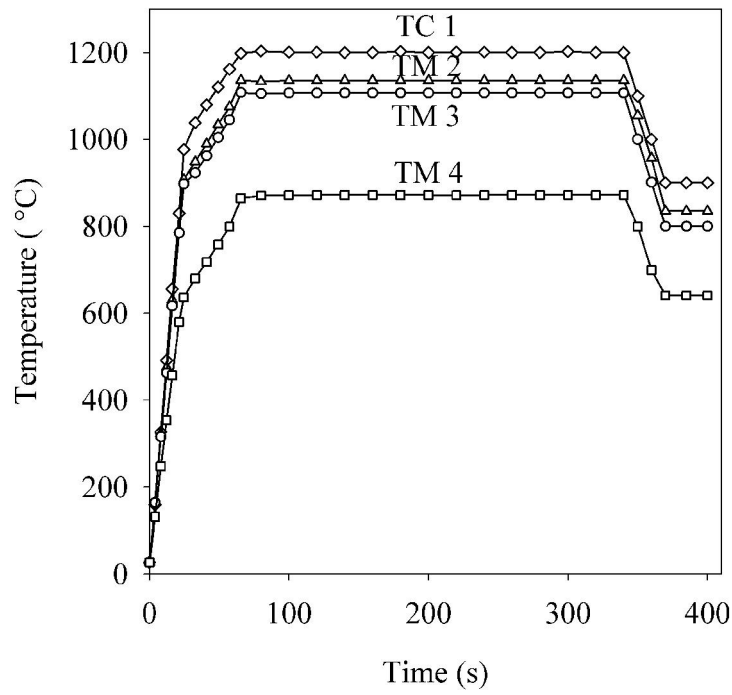


Figure 4.3: Transient response for TC 1, TM 2, TM 3 and TM 4 for double notched sample DN_R5 with copper grips.

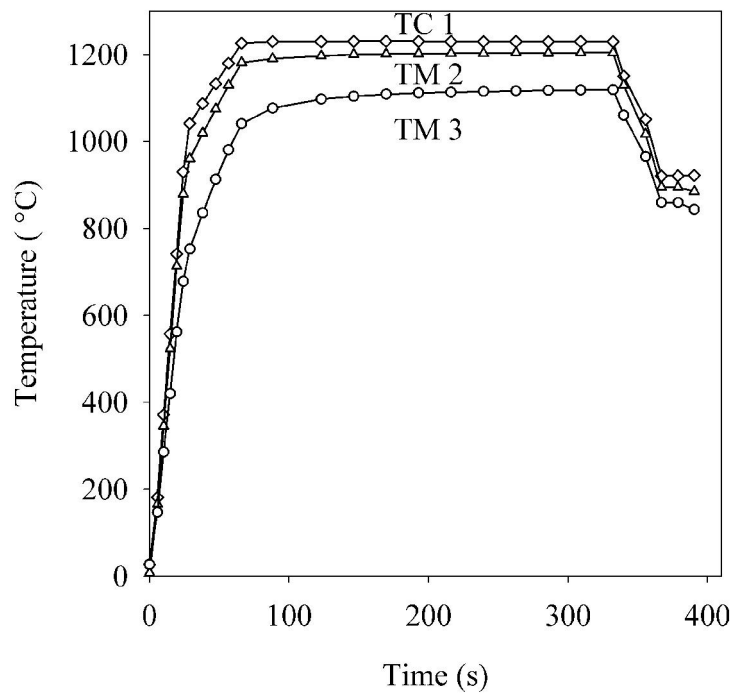


Figure 4.4: Transient response for TC 1, TM 2 and TM 3 for double notched sample DN_R5 with steel grips.

4.3 ELECTRO-THERMAL FE MODEL

The electro-thermal FE analysis presented in this section aims to simulate the high temperature axisymmetric Gleeble tests described in Chapter 3 (section 3.3) and calculate the experimentally observed longitudinal thermal profiles.

The FE model, shown in Figure 4.5, comprises of the assembly of the tensile sample and the copper grips. Both the sample and the grips were modelled as axisymmetric deformable bodies. This meant that cylindrical grips were assumed instead of their actual rectangular geometry to reduce computational cost. The bottom of the test piece is the mid-axis where the axis of symmetry has been applied. Additional displacement and symmetry boundary conditions were applied to represent the kinematic restrictions of the test set up. Four nodes were created at the exact positions of thermocouples TC 1, TM 2, TM 3 and TM 4 (see Figure 3.6b), thus representing them. Node 1 (N1 in Figure 4.5a) represents the controlling thermocouple TC 1 whilst the rest are simply used for recording temperatures at the specific points.

In this analysis the sample is heated by electrical current, applied on surface A (shown in Figure 4.5a), the contact surface of the grips with the water-cooled copper plates. The current amplitude was controlled using a feedback loop system, in a similar fashion to the Gleeble machine. This loop compares the instant sample temperature, recorded from node N1, to the prescribed sample temperature, defined in a subroutine, at each time step and corrects the current amplitude according to a function of the error between the two compared temperatures. This process was implemented using user subroutine UAMP and is explained in detail in section 4.4. Zero electrical potential was applied across the middle of the sample at surface D. The initial temperature for the whole assembly was set to a room temperature of 23 °C.

Heat loss via conduction, from the copper grips to the water-cooled copper plates occurs during the test. This copper-copper contact results in high heat transfer along that interface. In order to reduce computational power and avoid adding an extra component to the assembly, this heat loss was modelled as convection taking place on the contacting grip surface A. A high convection heat transfer coefficient of $1000 \text{ Wm}^{-2}\text{K}^{-1}$ was used and the sink temperature was defined to be 23 °C. This coefficient was determined empirically for the best fit of the thermal simulation results to the experimental temperature data obtained for selected positions along the gauge section of the specimen. This achieved satisfactory cooling of the copper grips for the applied range of test temperatures. The heat loss from the sample to the grip at surface B was modelled using thermal

contact conduction. The thermal interface coefficient for surface B is unknown and dependent on the pressure applied and the roughness of the two contacting surfaces.

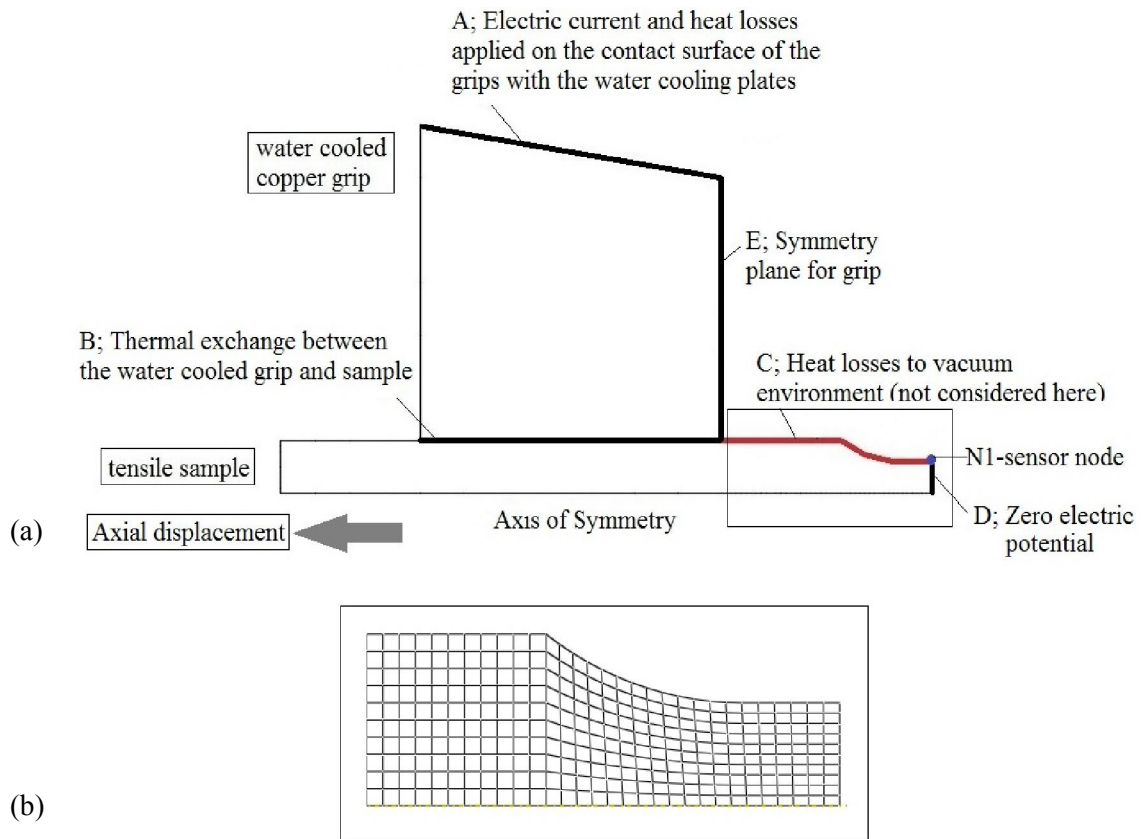


Figure 4.5: (a) Finite Element model of the tensile sample and copper grip and (b) Mesh of the area of interest of the tensile sample.

In this case the contact between the grips and the sample is assumed to be smooth and under high pressure, thus, resulting in a high thermal interface coefficient. Curve fitting of the predicted thermal data to experimental data was implemented to estimate an accurate interface thermal coefficient and the best fit was obtained with a coefficient of $4000 \text{ Wm}^{-2}\text{K}^{-1}$. This value is in agreement with values used for the same purpose in the literature (Norris and Wilson, 1999). The tests were conducted in vacuum therefore there are no heat losses to the surroundings by convection. Although surface heat loss due to radiative cooling along the free span of the sample (surface C) does occur, this was not considered as the radial thermal profile is not of interest in this simulation.

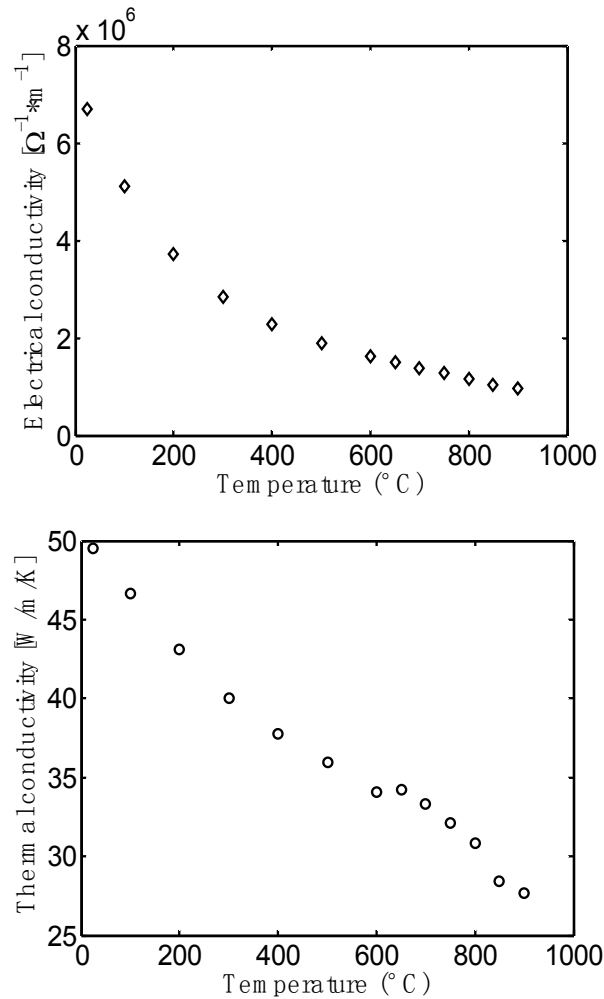


Figure 4.6: Simulated electrical and thermal material properties for the leaded free cutting steel that was considered.

A coupled thermal-electrical analysis, implemented in ABAQUS/Standard was used to simulate the Joule heating that takes place during Gleeble tests. In this type of analysis the coupled thermal-electrical equations used, are solved simultaneously at every node for both temperature and electrical potential. More details about the equations used in this analysis can be found in the ABAQUS Theory Manual (Systemes, 2011). For this Standard ABAQUS analysis, 4-node linear axisymmetric coupled thermal-electrical quadrilateral (DCAX4E) elements were used to reduce computational cost. These thermal-electric elements allow the Joule heating effect to be simulated as they couple electrical and thermal problems due to heat generation from the applied current and the heat transfers across the model. The area of interest of the sample, which extends from the gauge centre to about 12 mm from the centre (position TM 4) was meshed with 430 elements (see Figure 4.5b) to capture in detail the thermal gradients that appear in that region. In more detail,

the centre section of the gauge was meshed with 10 elements across. The rest of the model implements a coarser mesh (total number of elements is 676) since no information is required from other parts rather than the gauge area. Basic sensitivity studies were carried out and the generated mesh was based on experience of similar FE simulation problems (Liu et al., 2005).

Figure 4.6 shows the thermal and electrical temperature dependent properties of the FCS material obtained from Tata Steel internal simulation with JMatPro. For simplicity, the specific heat and density of the material were kept constant at 8257 J/kg/K and 7786 kg/m³, respectively, as the use of these material properties as a function of time does not have a significant effect on the temperature field results (difference in thermal fields within 4%).

4.4 TEMPERATURE CONTROL ALGORITHM

User subroutine UAMP was used to provide the transient temperature feedback control that regulates the electrical loading of the sample, thus, achieving the prescribed thermal cycle (see Figure 4.7). The control aspect of the system was implemented using a sensor able to provide temperature feedback at the beginning of every step. The controlling thermocouple, TC 1, was modelled as a node, N1, shown in Figure 4.5a. A small, empirical value for the electric current amplitude was assigned at the start of the simulation. The prescribed temperature for the controlling node is calculated at the start of every step of the electro-thermal simulation. The subroutine calculates this temperature according to a specific time interval (A, B, C, D, E as shown in Figure 4.7). The FE calculated temperature, as recorded from N1, is then compared to the prescribed temperature as described by equation 4.1. If this difference is sufficiently low (below or equal to 10⁻²) then the current amplitude remains constant, otherwise the current is corrected according to equation 4.2, which is a function of the obtained temperature difference:

$$\Delta T_i = T_i^p - T_i^s \quad 4.1$$

$$\Delta I = \left(1 + \left(\frac{\Delta T_i}{T_i^p} \right)^\alpha \right) \pm \beta \quad 4.2$$

where, T_i^p and T_i^s are the prescribed and measured, sensor temperature at every time step, α , is a coefficient that remains constant according to the corresponding time interval (A, B, C, D or E) and β is a correction coefficient equal to 0.001. The correction term is then multiplied by the current used in the previous step:

$$I_i = I_{i-1} \times \Delta I \quad 4.3$$

This procedure is repeated at every time step. The outline of this process is shown in the flow chart of Figure 4.8. The values coefficient α takes for each time interval are given in Table 4.1 below.

Table 4.1: Coefficient α used in the UAMP subroutine to achieve the replication of the prescribed heating cycle.

Interval	α
A	0.15
B	0.25
C	0.25
D	0.27
E	0.27

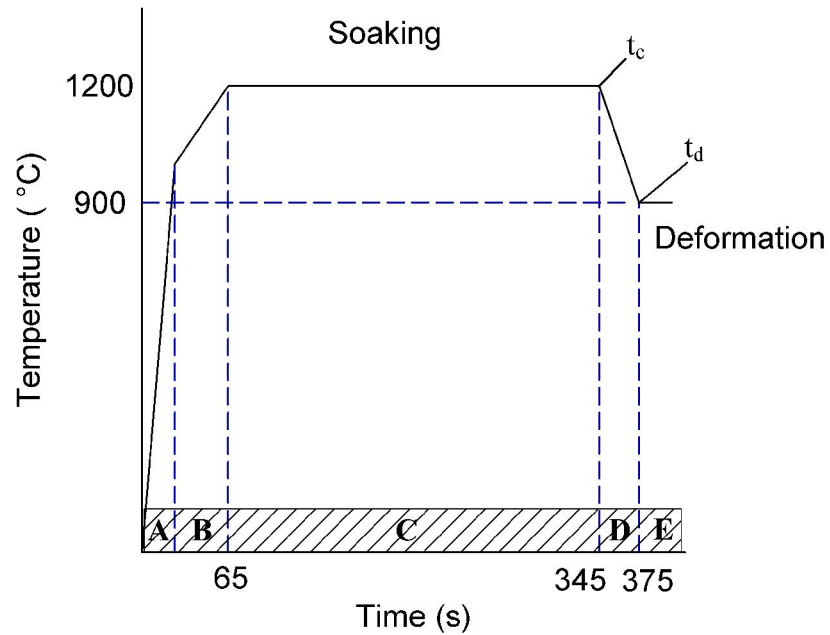


Figure 4.7: Temperature control with prescribed thermal cycle. Intervals A, B, C, D and E, separated by the dashed lines, denote the different heating and cooling stages of the thermal cycle.

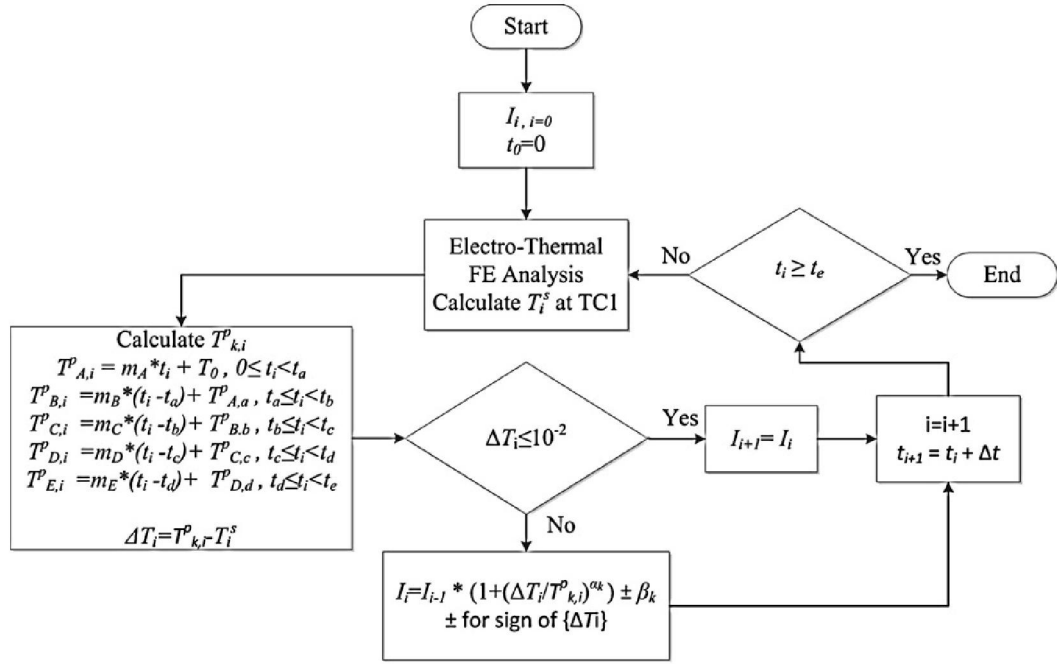


Figure 4.8: Flow chart describing the user subroutine UAMP.

4.5 ELECTRO-THERMAL PREDICTION RESULTS

The electro-thermal FE simulation was able to reproduce axial thermal gradients along the sample gauge, as observed experimentally, in high temperature Gleeble tensile tests. Figure 4.9 illustrates the FE calculated longitudinal temperature profiles by means of temperature contours for times t_c and t_d (see Figure 4.7). Additionally, the FE predicted (lines) temperature history of the sample as recorded from the four nodes is compared to the experimentally recorded (symbols) temperatures, at the same positions, in Figure 4.10. The results of this figure show that the subroutine serves its purpose as a reliable feedback control system with predictions from node N1 reproducing the recordings of TC 1 exactly. The FE predicted (lines) and experimentally observed (symbols) temperature gradients along the specimen surface are presented in Figure 4.11 for specific times of the heating cycle, namely t_c and t_d .

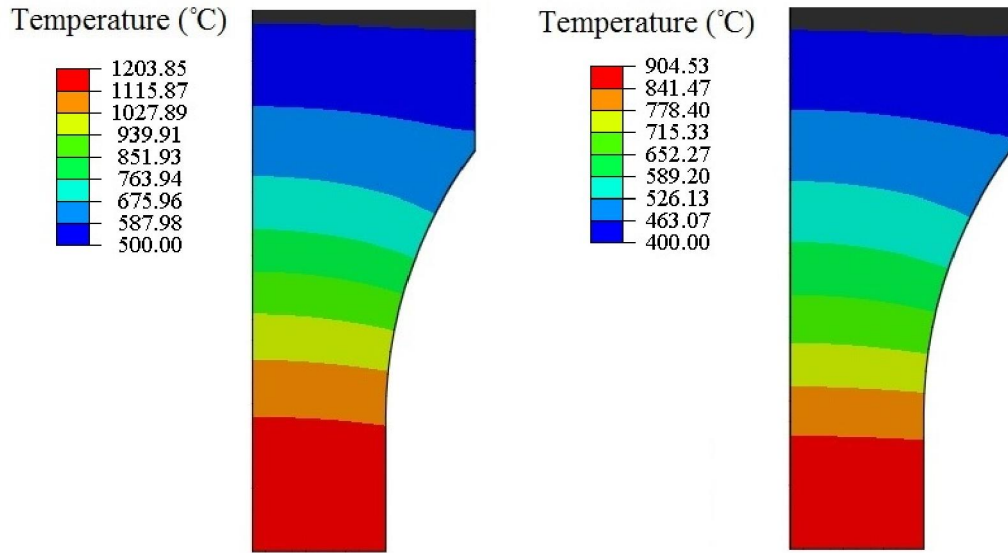


Figure 4.9: Predicted temperature fields (in °C) along the gauge at times (a) t_c , end of soaking and (b) t_d , beginning of deformation, as indicated in (Figure 4.7).

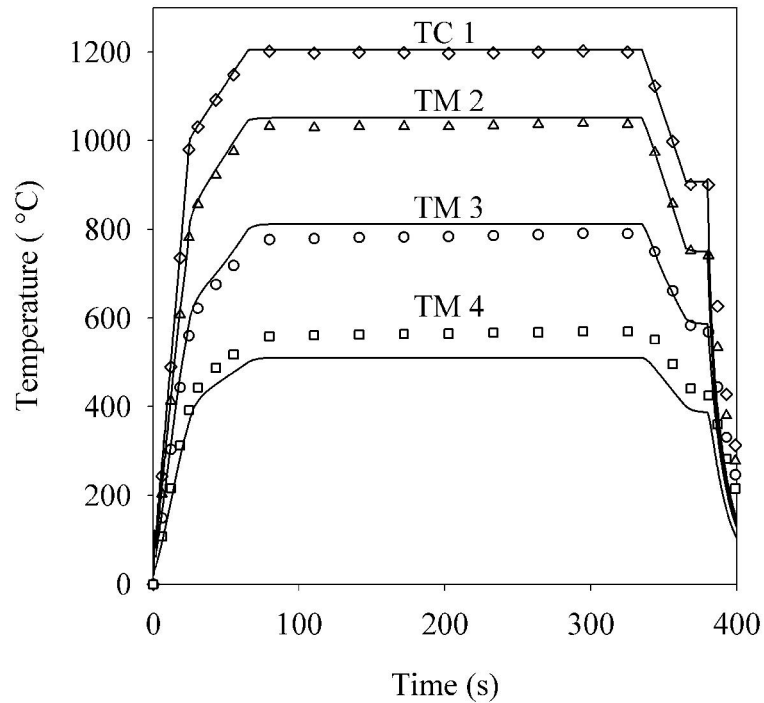


Figure 4.10: Temperature history along the sample as recorded from the four thermocouples. Symbols represent the experimental data and solid lines the FE calculated thermal data.

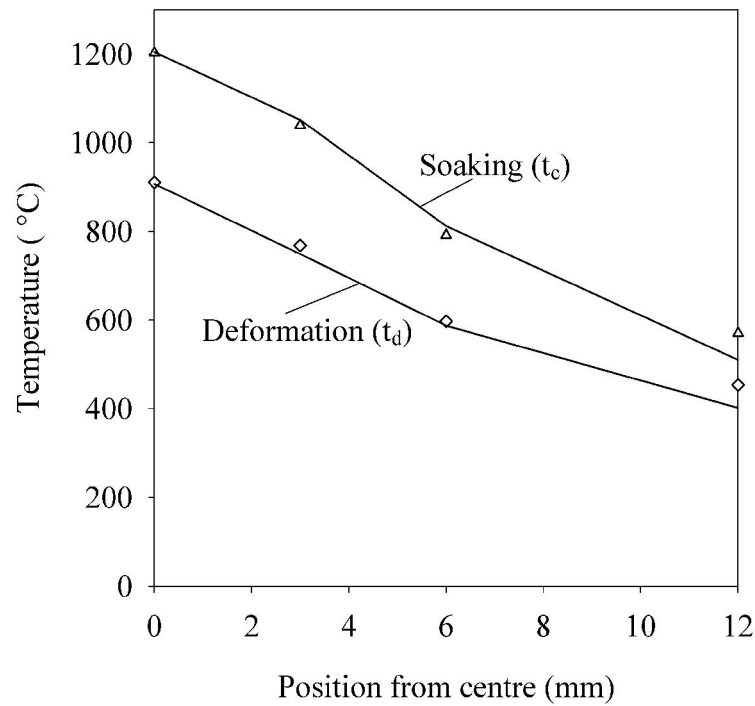


Figure 4.11: Comparison of experimental (symbols) and FE computed (lines) longitudinal temperature profiles at different stages of the heating cycle, namely at times t_c , end of soaking and t_d , beginning of deformation, as indicated in Figure 4.7.

Good agreement is found between the experimental and FE calculated results for thermocouples TC 1, TM 2 and TM 3. Greater inconsistencies are observed for thermocouple TM 4. This is expected seeing as thermocouple TM 4 is positioned on the testpiece shaft, nearest to the water cooled grips. Sample cooling, which becomes possible from contact with the grips, is difficult to be simulated accurately due to experimental factors that cannot be measured easily. Such an example is the force applied when the grips are being fastened. This factor, which will influence the heating profile cannot be measured in this set up and therefore cannot be accurately included in the model. For this reason, the greater inaccuracy of the reading from thermocouple TM 4 is justified. In this model, TM 4 served as a reference point to aid with the fit of the modelled data to the experimental data. This was successful for the three thermocouples, positioned within the gauge section of the sample, which are the main focus of this model as the majority of the deformation takes place in that region.

Inconsistencies can also be attributed to inaccurate attachment of the thermocouple as precise welding of its two wires on a curved surface can be challenging. Great manual dexterity and excellent spot welding skills are required for accurate positioning. Taking into consideration the

maximum accuracy with which the thermocouples can be welded into position in this experimental arrangement, the predicted results are in good agreement with the experimental data.

This validates the algorithm implemented in the user subroutine UAMP as a useful tool for reliable calculation of the axial thermal gradients present in Gleeble high temperature tensile tests. Using data calculated from electro-thermal simulations where a prescribed thermal profile can be successfully reproduced at the controlling node will reduce the number of experimental investigations where a complicated set up with multiple thermocouples is required.

4.6 MECHANICAL FE MODEL

An explicit mechanical FE analysis was used to simulate the high temperature Gleeble axisymmetric tensile tests. The same FE model with the same mesh as for the electro-thermal simulation was used. This served the purpose that all predicted temperature data corresponding to each element from the electro-thermal model could be extracted and input as a predefined thermal field in the mechanical model. The temperature data that was input in the mechanical simulation corresponded to time $t = t_d$, the start of the deformation interval for which temperature conditions remain constant. The input thermal data remain constant throughout the mechanical simulation. In this model, a mechanical rather than a coupled thermo-mechanical simulation was implemented as the temperature fields, although inhomogeneous, are in steady state. As the heating of the sample in the Gleeble is controlled via the thermocouples, any additional heat generated by plastic work beyond what is already generated by the electric current would be felt by the thermocouples and the current amplitude would be modulated accordingly to maintain the target temperature. Therefore, any conversion of plastic work into heat was not included in the model.

After the successful input of the FE calculated axial temperature profiles, the sample was pulled in uniaxial tension, recreating the high temperature Gleeble tension test. To simulate tension in the specimen, a displacement controlled load was applied at surface E (see Figure 4.5a). To ensure that the testpiece is deformed at a near-constant strain rate, a continuously changing velocity profile, based on assuming a constant gauge length was used. A near-constant strain rate of 0.1 s^{-1} was used to achieve a 2 mm deformation. This mechanical tension simulation was also conducted under the same conditions for a simple isothermal case where a temperature of $900 \text{ }^{\circ}\text{C}$ was applied at the extended gauge area of the sample and was kept constant throughout the simulation. Four-node thermally coupled axisymmetric quadrilateral elements of type CAX4RT were used.

A material model, specific to leaded free cutting steel, was used to accurately simulate the deformation behaviour. This material model was implemented via the user defined subroutine VUMAT, for use in ABAQUS/Explicit. Heating being converted into plastic work has not been considered in this model as the temperature profile is kept constant throughout the deformation time. The set of constitutive equations used is described in more detail in Chapter 6. Here the equations are given together with the material coefficients that were calculated from previous work on a similar material from Foster et al. (2007). In this study, damage was omitted since the strain state of the testpiece and not the reduction of the flow stress at the final stages of deformation was of interest. In the equations presented below, where a dot notation is used it describes a rate whilst the $\langle \dots \rangle_+$ notation is used to indicate that negative values are set to zero.

$$\dot{\varepsilon}_p = \dot{\varepsilon}_0 \left\langle \frac{\sigma - (k + R)}{K} \right\rangle_+^n \quad 4.4$$

$$\dot{\bar{\rho}} = k_1 (1 - \sqrt{\bar{\rho}}) \dot{\varepsilon}_p - c_1 \bar{\rho}^{c_2} \quad 4.5$$

$$R = B \sqrt{\bar{\rho}} \quad 4.6$$

$$\sigma = E (\varepsilon_T - \varepsilon_p) \quad 4.7$$

The temperature dependent material constants are determined using the equations below:

$$k = k_0 \exp \left(\frac{Q_p}{\kappa T} \right) \quad 4.8$$

$$c_1 = c_{1(0)} \exp \left(-\frac{Q_R}{\kappa T} \right) \quad 4.9$$

$$E = \frac{E_0}{\cosh^2 (K_1 (T - T_0))} \quad 4.10$$

$$B = \frac{B_0}{\cosh^2 (K_1 (T - T_0))} \quad 4.11$$

Table 4.2: Material constants for viscoplastic damage constitutive equations determined for a leaded free-cutting steel as calculated by Foster et al. (2009b).

Constant	E_0 (MPa)	K (MPa)	k_0 (MPa)	Q_p (J/mol)	n (-)
Value	8.7×10^3	11.0	0.2	4.2×10^4	1.7
Q_R (J/mol)	$c_{1(0)}$ (s^{-1})	k_1 (-)	K_1 ($^{\circ}C^{-1}$)	c_2 (-)	B_0 (MPa)
1.73×10^4	1.0	2.8	3.5×10^{-3}	0.55	176

Equation 4.4 describes the plastic strain rate which is dependent on the initial yield, k and isotropic hardening, R . The initial yield point, k , is a temperature dependent material constant described in equation 4.8. Isotropic hardening is described by equation 4.6 where B is a material constant dependent on temperature, as described in equation 4.11. Equation 4.5 describes the normalised dislocation density evolution. The first term of the equation describes the formation of dislocation and dynamic recovery and the second part describes static recovery. A detailed derivation and explanation of this equation can be found in Lin et al. (2003). The coefficient c_1 in the normalised dislocation density equation is temperature dependent and takes the form given in equation 4.9. Equation 4.10 is used to describe the elastic modulus for the temperature range $T > 0.6T_m$ as it has been proven successful in work carried out by Liu et al (2004). In this case T_0 is not a determined material constant but a reference temperature, 900 $^{\circ}C$ in this case, at which E_0 and B_0 are defined. Finally an expression for the flow stress is described by equation 4.7.

The material constants of this model were first approximated using semi-empirical methods. Following this an in-house built EP optimisation algorithm, developed by Cao and Lin (2008) was implemented to determine a more accurate fit. Final determination of the material constants was achieved by comparing the computed fittings to experimental data like high temperature Gleeble tensile data. The equations were calibrated from experimental data for strain rates of the range $0.1-10 \text{ s}^{-1}$ and temperatures ranging from 900-1200 $^{\circ}C$. Stress-state behaviour of the material correlating to strain rates and temperatures outside the specified range can also be predicted using these equations.

4.7 FE RESULTS

The tensile simulations were carried out for a range of temperatures and strain rates to investigate the effect of FE calculated axial thermal fields on tensile strain state. The range of 900-1200 °C was chosen for the test temperatures and strain rates of 0.1, 1 and 10 s⁻¹ were chosen to physically simulate the conditions met in typical cogging/roughing continuous bar mills where initial speed is low during transition from as-cast to a wrought structure.

In Figure 4.12 the calculated axial plastic strain field from the tensile test simulations for both the FE calculated, non-isothermal, (Figure 4.12a) and isothermal (Figure 4.12b) temperature fields are shown for different stages of a total elongation of 2 mm. The results of this simulation were produced using a strain rate of 0.1 s⁻¹ and a target temperature of 1200 °C. The results indicate that the differences between maximum axial strains grow significantly as the elongation gets closer to the strain to failure (60% in case (iii) of 2 mm). For this particular case, a higher maximum axial strain (0.80 for an overall deformation of 2 mm) is reached for the case where an FE calculated temperature profile is used. The corresponding simulated maximum axial strain for the isothermal case was 0.32, approximately 2.4 times lower than the finding of the temperature gradient field simulation. The axial strain contours for the two cases (Figures 4.12a(iii) and 4.12b(iii)) also differ in uniformity. Although the predicted axial strains are considerably lower for the isothermal field case, the strain values are also more evenly distributed in the gauge area of the tensile specimen. In contrast, the highest values for axial strains of the FE calculated thermal field simulation are concentrated in the central area of the gauge, and a steeper transition is observed as the strains take lower values towards the end of the gauge area. This represents the real life case more closely. For the same conditions, when a smaller elongation of 1 mm is considered (case ii), the maximum strain for the FE calculated thermal case is 0.29 whilst the strain obtained for the isothermal case is 0.15. Here the strain values differ less, but still by 93.3%.

The ratio of the maximum axial plastic strain for the non-isothermal case over that of the isothermal case for a 2 mm elongation is shown in Figure 4.13 for three strain rates and the temperature range of 900-1200 °C. From the figure it is clear that for each strain rate there is a different optimal target temperature for which the isothermal assumption can be applied sensibly. For target temperatures above 1000 °C, the difference between the isothermal and non-isothermal cases diminishes with increasing strain rate. For example, at 1100 °C, the difference is 92.5 %, 51.4 % and 14.2 % at strain rates of 0.1 s⁻¹, 1 s⁻¹ and 10 s⁻¹, respectively. At 1000 °C, the absolute difference is minimal and an optimal strain rate exists at which the two simulations give the same maximum strain. Contrary to what is observed for temperatures above 1000 °C, at 900 °C the

difference in strains increases monotonically with increasing strain rate and the isothermal case maximum axial strain is always larger than the equivalent non-isothermal strain. For the temperature range of 1000-1200 °C the maximum axial strain is higher for the cases where a longitudinal FE calculated thermal gradient field has been utilised. However at a target temperature of 900 °C, the isothermal tensile simulation produces a higher maximum strain, although the absolute percentage difference is only 14.5 %. For smaller strain rates that target temperature is also lower but in all cases a target temperature of 1200 °C produces the highest discrepancy between the two cases.

Non-uniform temperature profiles in the gauge area cause non-uniform deformation since the ductility of the material is temperature-dependent, i.e. lower temperature regions deform less. In order to accommodate the applied elongation, the hotter region in the centre of the gauge must deform more than it would if the temperature were uniform. Strain localisation in the centre of the gauge is more pronounced when the overall temperature is high and the applied strain rate is low; ductility is higher under those conditions. This gives rise to the trend shown in Figure 4.13 for temperatures greater than 1000 °C.

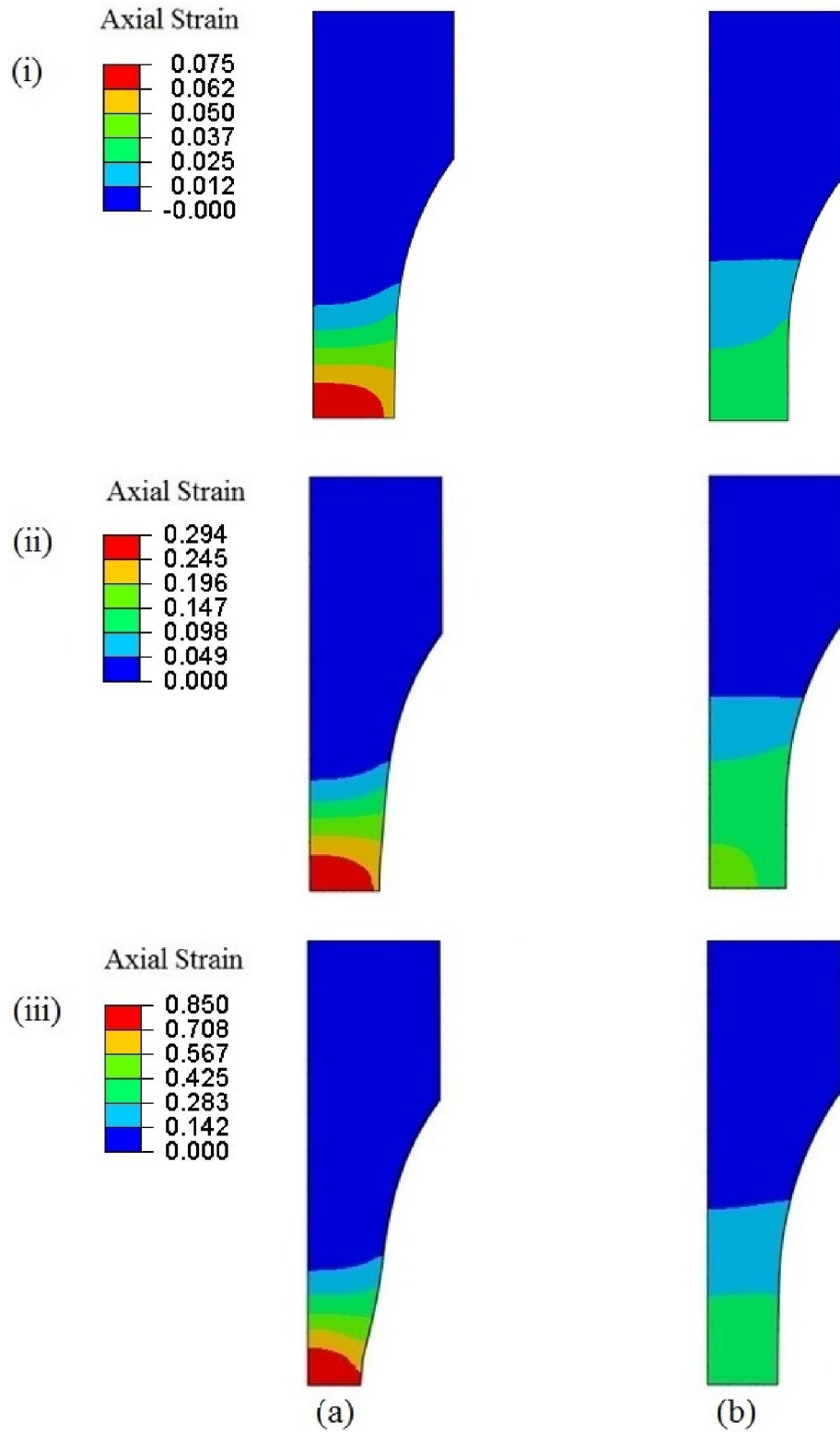


Figure 4.12: Comparison of computed axial plastic strain for specimens with (a) FE-calculated temperature fields and (b) isothermal temperature fields at a total elongation of (i) 0.2 mm, (ii), 1 mm and (iii) 2 mm. The target temperature in this simulation is 1200 °C and the strain rate is 0.1 s⁻¹.

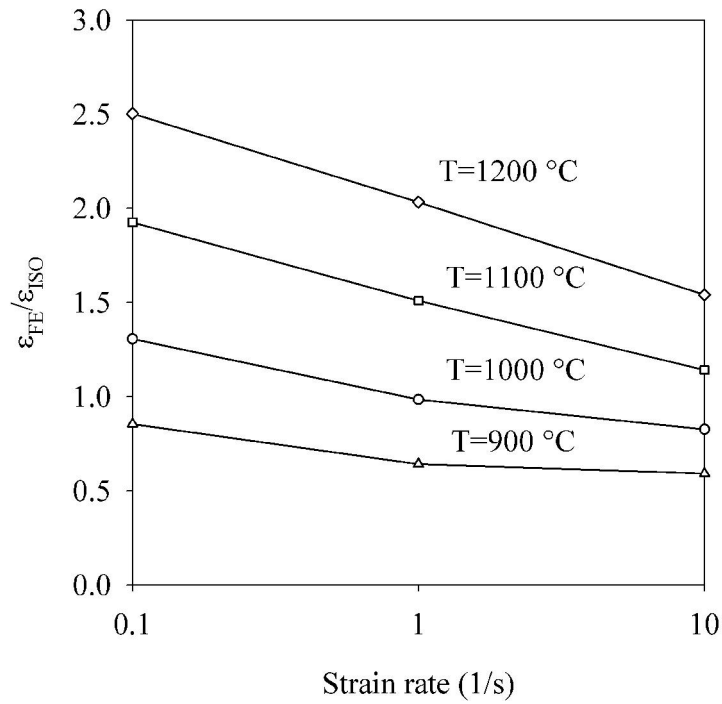


Figure 4.13: Maximum axial strain of the non-isothermal case over that of the isothermal case versus strain rate for four target temperatures and three strain rates ($\dot{\epsilon}=0.1, 1, 10 \text{ s}^{-1}$).

4.8 THERMAL PREDICTIONS FOR DOUBLE NOTCH SAMPLES

A very similar approach to the one described in the earlier sections was used to calculate longitudinal thermal fields for the double notched samples used in this work, namely DN_R5 and DN_R1.25. In this case, to simplify the simulation work that will later on include calculations of damage, the grips were omitted. Instead, convection cooling was applied directly on the sample surface. A best fit to experimental thermal data approach was used to determine the value of the convection heat transfer coefficient accurately. A good agreement between the FE calculated and experimental thermal profiles was achieved using a convection coefficient of $1500 \text{ Wm}^{-2}\text{K}^{-1}$. Figure 4.14 and 4.15 represent the calculated thermal contours for samples DN_R5 and DN_R1.25 respectively for two times in the thermal cycle, t_c and t_d .

Including the thermal profiles from the high temperature Gleeble testing in the double notched sample simulations is an important milestone that will improve the accuracy of the stress-state results (Chapter 6) and damage estimations (Chapter 7).

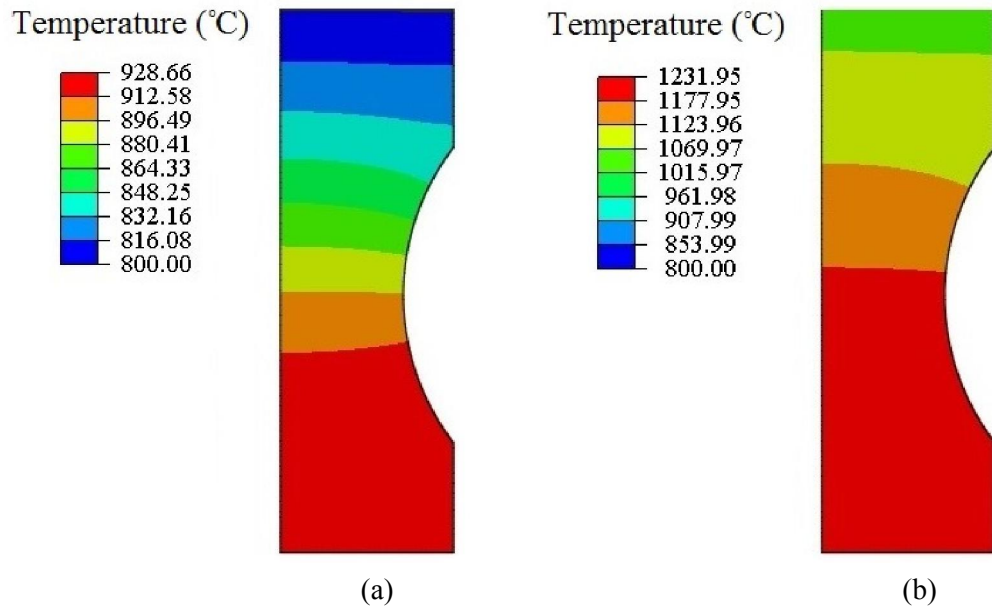


Figure 4.14: Thermal contours of double notch sample DN_R5 for two times in the thermal cycle; (a) t_d and (b) t_c .

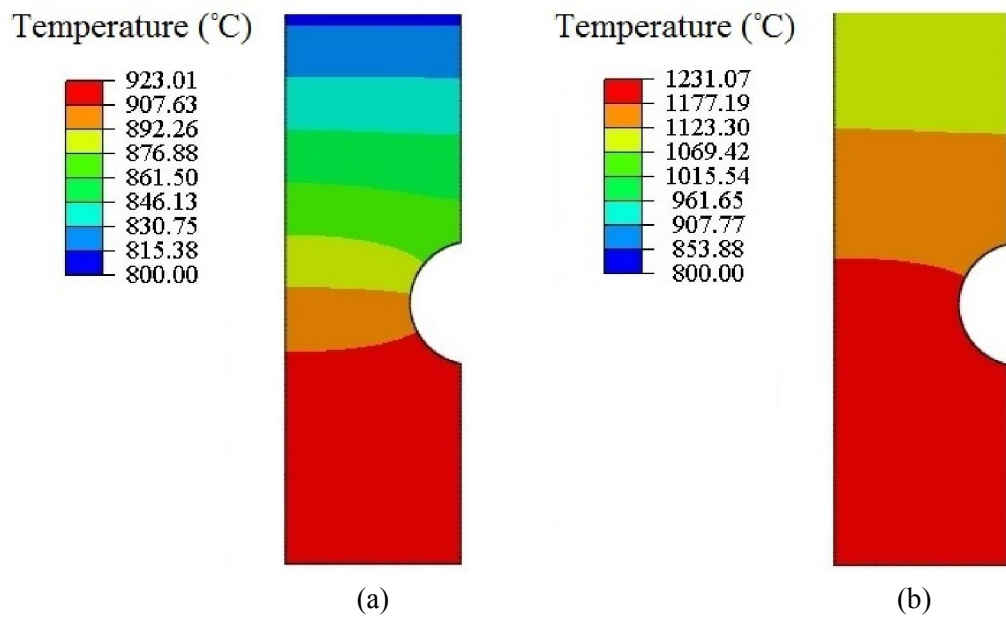


Figure 4.15: Thermal contours of double notch sample DN_R1.25 for two times in the thermal cycle; (a) t_d and (b) t_c .

4.9 CONCLUSIONS

Gleeble high temperature thermal tests have revealed the presence of strong axial thermal gradients. This has been tested in standard dog-bone samples and double notched samples. From the double notch thermal tests, the temperature difference from the sample centre to the notch centre was identified to be 30 °C. Knowledge of this temperature difference allowed to correct for it and, therefore, apply the required target temperature on the notched areas of the sample.

An efficient feedback algorithm that controls the current input of an electro-thermal numerical analysis for predicting thermal fields observed in high temperature Gleeble testing was presented and validated. The algorithm was implemented in the commercial FE software ABAQUS through the user defined subroutine, UAMP. Electro-thermal numerical analyses utilising this algorithm were carried out for a range of elevated temperatures to predict the thermal fields observed in Gleeble high temperature testing. The FE calculated thermal fields were compared with the captured experimental measurements from Gleeble tensile tests and were found to be in good agreement. A series of high temperature tensile test simulations were performed in ABAQUS to evaluate if the use of isothermal profiles as opposed to FE calculated thermal profiles, has a significant effect on the predicted strain state. All tensile simulations were carried out using a temperature-dependent viscoplastic constitutive model, specific to leaded free cutting steel. For most of the tested conditions the maximum axial strain was higher when an FE calculated thermal profile was used and larger differences were identified for low strain rates, high temperatures and large deformations. Additionally, it was found that for a fixed temperature above 1000 °C the difference between the FE calculated versus the isothermal maximum axial strain increases with decreasing strain rate. For cases of mid-range target temperatures and medium to high strain rates (i.e. 1 s⁻¹ at 1000 °C, or 10 s⁻¹ at 1100 °C) the isothermal assumption produced similar results to those of the FE calculated temperature field simulations.

Nonetheless, the isothermal assumption was found to be generally inaccurate. It should therefore be used with caution and mainly for simulating tests at mid-range target temperatures and moderate to high strain rates. In most cases the plastic deformation is under predicted by the isothermal assumption. The proposed method can accurately calculate the thermal profiles of high temperature Gleeble uniaxial tension tests, induced by Joule heating. These FE calculated thermal profiles can then easily be used in mechanical simulations to produce a numerically accurate prediction of the stress and strain state.

CHAPTER 5

EXPERIMENTAL RESULTS OF TENSILE TESTING ON UNIAXIAL AND DOUBLE NOTCHED SAMPLES

SUMMARY

In this chapter, the results of high temperature Gleeble tension testing under conditions of hot rolling are presented. Both uniaxial and multiaxial tension results, from the double notched samples, are presented. Uniaxial tension results are presented to understand the viscoplastic material response under purely uniaxial conditions. The double notch tension tests were carried out to investigate the effect of stress-state on damage initiation for leaded free cutting steel (LFCS). Both sets of tests were carried out under conditions analogous to those of hot rolling; high temperatures (900-1200 °C) and moderate strain rates ($\dot{\epsilon}=0.01-1 \text{ s}^{-1}$). A comparison between the uniaxial and multiaxial tension test results is also presented here. All stress versus strain graphs were generated using the methods outlined in Chapter 3. To understand the damage initiation patterns, influenced from the stress-state, micrographs from the optical microstructural examination of selected double notched samples are presented. These were taken from the sectioned unfailed notch, after testing to failure the other notch of each sample, to examine failure initiation sites since the material has been captured in a state very close to failure. The examined micrographs reveal distinctive damage paths relating to stress-state.

5.1 TENSION TEST RESULTS

5.1.1 UNIAXIAL TESTS

High temperature tensile experiments were carried out on the uniaxial (ASTM) and double notched samples (DN_R5 and DN_R1.25). Firstly, tension tests were performed on the uniaxial samples to identify the viscoplastic behaviour of the material and gather sufficient experimental data to later calibrate the constitutive model (equations without damage). A list of the testing conditions used for the ASTM samples is shown in Table 5.1. The experimental results from tests

performed at $T=900\text{ }^{\circ}\text{C}$ for three strain rates, namely $\dot{\epsilon} = 0.1, 1, 10\text{ s}^{-1}$ are shown in Figure 5.1 and the results from tests carried out at $\dot{\epsilon} = 1\text{ s}^{-1}$ for three different temperatures, $T=900, 1100, 1200\text{ }^{\circ}\text{C}$ are shown in Figure 5.2. The stress versus strain graphs were produced from the experimental data according to the methods outlined in Chapter 3. A typical viscoplastic response is identified from Figures 5.1 and 5.2, with ductility increasing for higher temperatures and lower strain rates and flow stresses increasing with increasing strain rate. Additionally it is worth noting that for low temperatures (i.e. $T=900\text{ }^{\circ}\text{C}$) and high strain rates (i.e. $\dot{\epsilon}=1\text{ s}^{-1}$) material failure takes place just after the ultimate tensile strength is reached whilst for high temperatures (i.e. $T=1200\text{ }^{\circ}\text{C}$) and lower strain rates (i.e. $\dot{\epsilon} = 0.01\text{ s}^{-1}$) the onset of damage is somewhat delayed.

Previous investigations (Foster, 2007; Kaye, 2012; Liu, 2004) of free cutting steel from different production routes also confirm the observed low ductility and a similar viscoplastic response. Additionally, Foster (2007) who studied a similar as-cast leaded free cutting steel grade from a bloom product, postulated that the reason for postponed damage onset at high temperatures and low strain rates could be due to softening mechanisms such as recrystallisation and recovery. It is worth noting that the samples used in these previous investigations were not purely uniaxial and therefore a direct comparison with that data cannot be made.

Foster and later Kaye (Foster, 2007; Kaye, 2012) used an axisymmetric tension specimen, G4A, often used in the industry and similar to the dog-bone sample used in the thermal tests of Chapter 4. This sample geometry features a 6 mm gauge to concentrate deformation and heating but the close proximity of the notched radius to the gauge area brings uncertainty to whether these results can be classified as uniaxial. The use of a purely uniaxial sample, according to the ASTM British tensile standards (Loveday et al., 2004) here ensures that the viscoplastic material response has been recorded for pure uniaxial tension and that the constitutive model can be later calibrated with accuracy for uniaxial conditions.

Table 5.1: Uniaxial tension test program.

Sample name	Strain rate (s^{-1})	Temperature ($^{\circ}\text{C}$)
ASTM	0.01	900
	0.1	900
	1	900
	0.1	1100
	1	1100
	1	1200
	10	1200

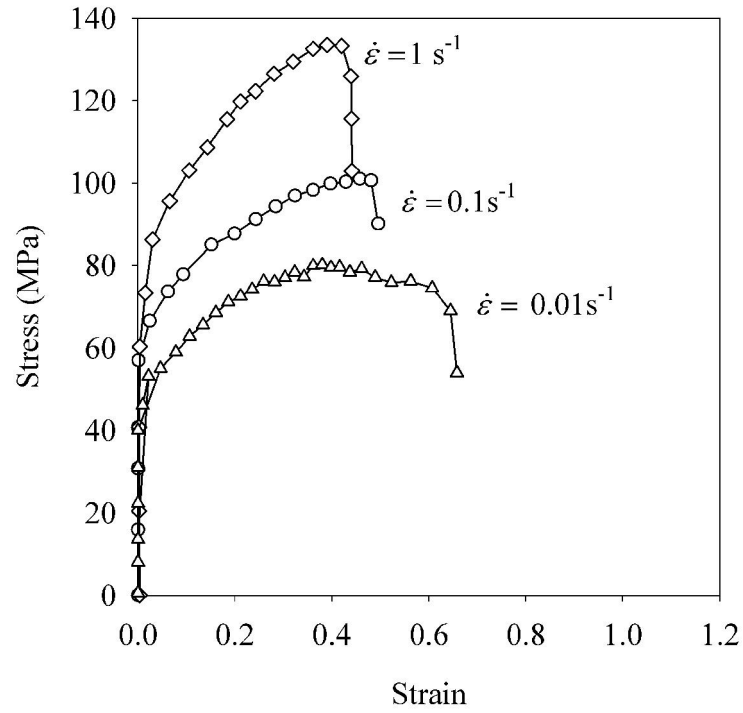


Figure 5.1: Average axial stress versus axial strain for loaded free cutting steel, from uniaxial tension tests of ASTM samples at $T=900\text{ }^{\circ}\text{C}$.

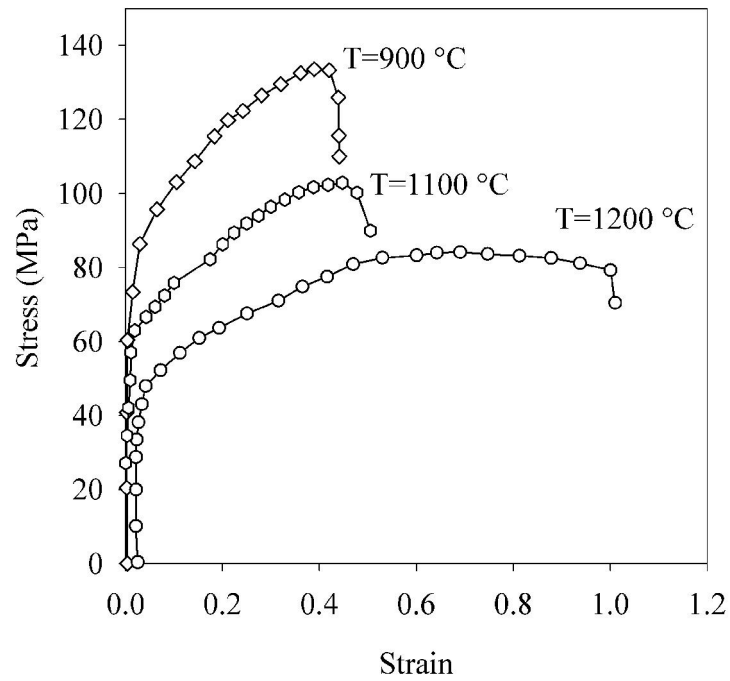


Figure 5.2: Average axial stress versus axial strain for loaded free cutting steel, from uniaxial tension tests of ASTM samples at $\dot{\epsilon} = 1\text{ s}^{-1}$.

5.1.2 DOUBLE NOTCH SAMPLE RESULTS

After establishing the viscoplastic response of the material for uniaxial tension, high temperature tensile tests on double notched samples DN_R5 and DN_R1.25 were performed. The aim of these tests was to understand how stress-state differences, influenced by the geometry of the samples, affect ductility. Table 5.2 and Table 5.3 show the testing conditions used for each double notched sample along with the repeats for some of the tests. In Figure 5.3 the tension test results for sample DN_R5 deformed at two strain rates, $\dot{\epsilon} = 0.1$ and 1 s^{-1} and two temperatures $T = 900, 1200$ °C are shown. The viscoplastic response from the tests is as expected, with higher flow stress as strain rate increases and reduced ductility for lower temperatures but with differences in ductility between testing conditions being very small. A significant reduction in the ductility of the material is identified, with the maximum strain to failure, ϵ_f , at 0.5 for a test performed at $T = 1200$ °C and $\dot{\epsilon} = 0.1 \text{ s}^{-1}$ and the lowest strain to failure at 0.38 for a test performed at $T = 900$ °C and $\dot{\epsilon} = 0.1 \text{ s}^{-1}$. This shows that the multiaxial stress-state, induced from the notch of the sample, has significantly altered the capacity of the material to deform plastically.

Table 5.2: Double notch tests matrix for sample DN_R1.25.

Sample name	Strain rate (s^{-1})	Temperature (°C)	Repeat
DN_R1.25	0.1	900	x
	0.1	1100	
	0.1	1200	x
	1	900	x
	1	1200	

Table 5.3: Double notch test matrix for sample DN_R5.

Sample name	Strain rate (s^{-1})	Temperature (°C)	Repeat
DN_R5	0.1	900	x
	0.1	1100	
	0.1	1200	x
	1	900	x
	1	1200	

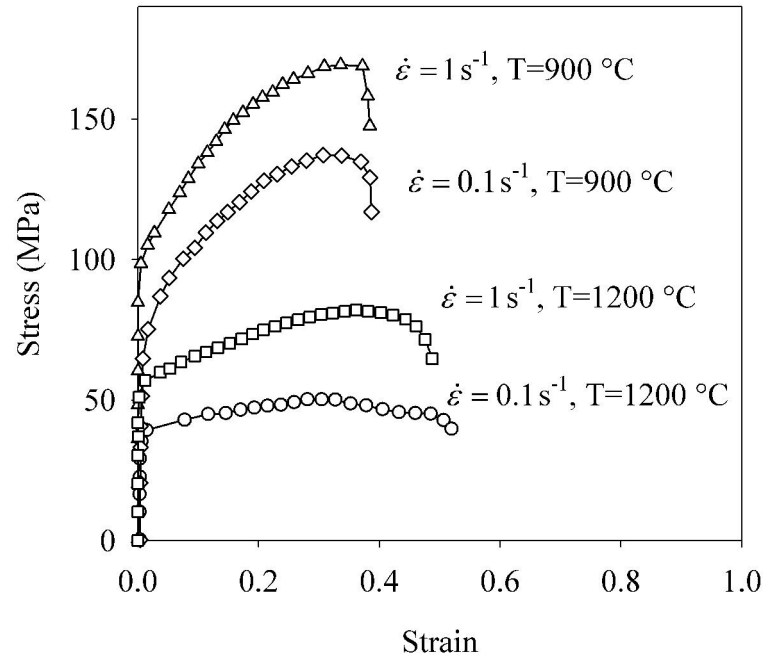


Figure 5.3: Axial stress versus axial strain for sample DN_R5 at four conditions.

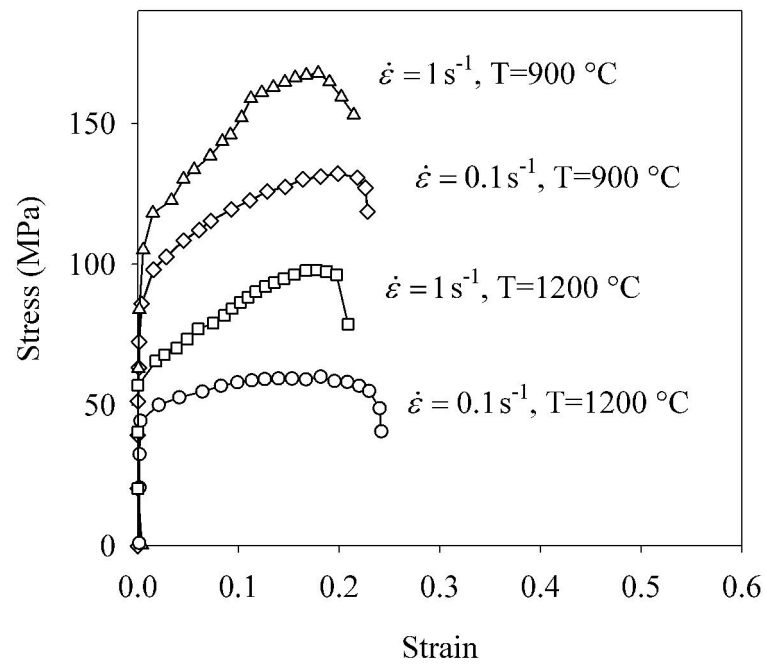


Figure 5.4: Average axial stress versus axial strain for sample DN_R1.25 for four conditions.

In Figure 5.4 the test results for the same four conditions as those presented in Figure 5.3 were plotted for sample DN_R1.25. These results show an even greater reduction in ductility with strain to failure, for all conditions examined, below 0.3. These results also showed that whilst the flow stresses are still higher for higher strain rates, strain to failure values do not change significantly for lower temperatures. For example, a sample tested at a strain rate of 0.1 at $T=900$ °C has a strain to failure of 0.22, whilst at $T=1200$ °C the strain to failure increases marginally to 0.24. This validated that the ductility of the material is restricted by stress-state.

5.1.3 COMPARISON OF UNIAXIAL AND DOUBLE NOTCH SAMPLE RESPONSE

In this section a comparison between the uniaxial and multiaxial tensile test data, presented in sections 5.1.1 and 5.1.2, respectively, is given. Figures 5.5 and 5.6 show the average axial stress plotted against the effective plastic strain, obtained from tensile experiments for the uniaxial sample (black symbols), sample DN_R5 (blue symbols) and sample DN_R1.25 (green symbols). In Figure 5.5 the data for the three different samples for two conditions with the same strain rate, $\dot{\epsilon} = 0.1 \text{ s}^{-1}$ at temperatures $T=900$ and 1200 °C are compared. In Figure 5.6 the results for $\dot{\epsilon} = 1 \text{ s}^{-1}$ at the same temperatures are shown for all three samples.

Data from all four conditions shown in Figures 5.5 and 5.6 support the same trend; the higher the triaxiality (i.e. the sharper the notch), the lower the strain to failure. This trend has been well documented by many other researchers (Alves and Jones, 1999; Hancock and Mackenzie, 1976; Isobe et al., 2014; Mirza et al., 1996). The reduction in strain to failure as triaxiality increases is more evident at high temperatures. This suggests that the stress-state influence dominates in these tests over the testing conditions (i.e. high temperature in this case). The geometry induced stress-state influences not only strain to failure but also flow stress. Flow stresses measured from the uniaxial sample tests are considerably lower than those recorded for both double notched samples. When comparing the flow stresses from the double notched samples alone, a higher triaxial state (sharp notch DN_R1.25) is associated with higher flow stresses. This is most evident for high temperatures, $T=1200$ °C, and both examined strain rates in Figures 5.5 and 5.6. The results highlight the impact of the stress-state, created by the notch geometry, on the viscoplastic response of the material.

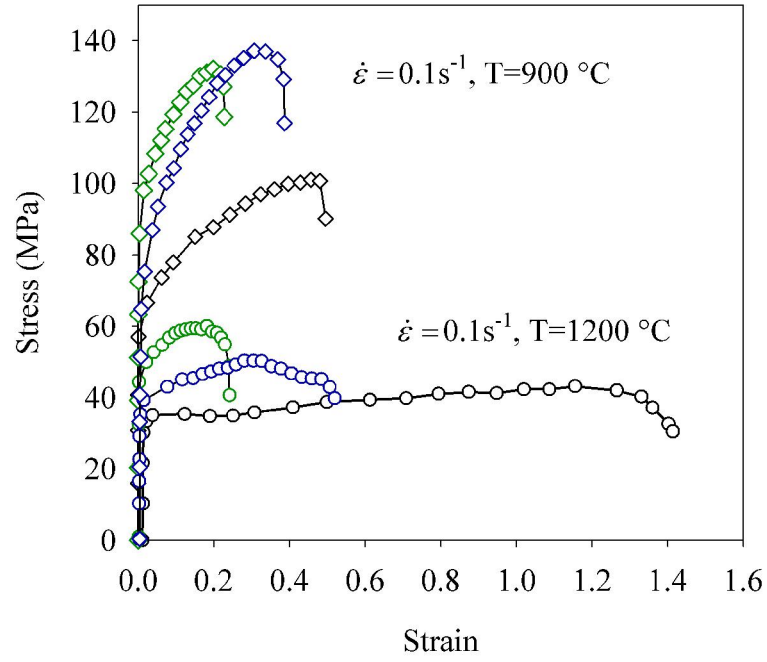


Figure 5.5: Average axial stress versus axial strain for samples ASTM (black symbols), DN_R1.25 (green symbols) and DN_R5 (blue symbols) for conditions; (i) $\dot{\epsilon} = 0.1\text{ s}^{-1}$, $T=900\text{ }^{\circ}\text{C}$ (diamond) and (ii) $\dot{\epsilon} = 0.1\text{ s}^{-1}$, $T=1200\text{ }^{\circ}\text{C}$ (circle).

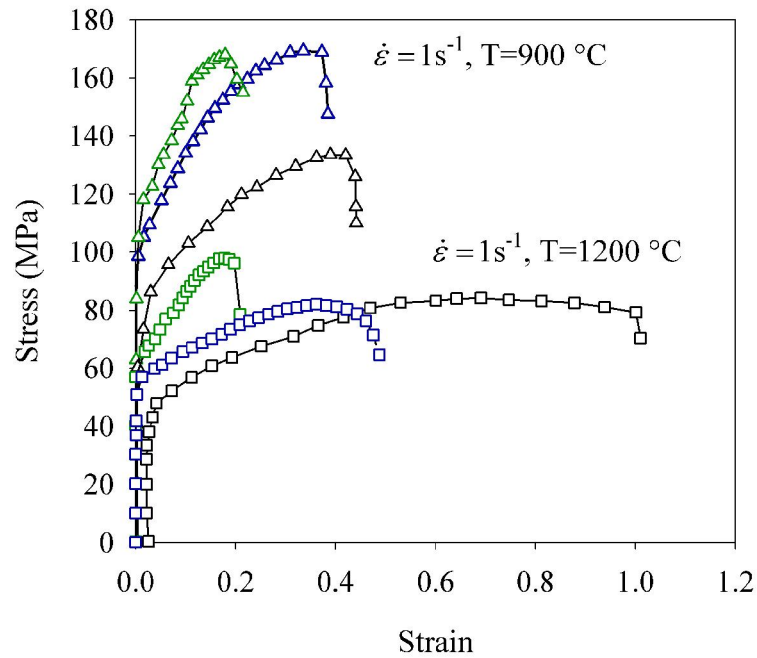


Figure 5.6: Average axial stress versus axial strain for samples ASTM (black symbols), DN_R1.25 (green symbols) and DN_R5 (blue symbols) for conditions; (i) $\dot{\epsilon} = 1 \text{ s}^{-1}$, $T = 900 \text{ °C}$ (triangle) and (ii) $\dot{\epsilon} = 1 \text{ s}^{-1}$, $T = 1200 \text{ °C}$ (square).

5.2 MICROSTRUCTURAL ANALYSIS

5.2.1 MESOSCALE DAMAGE OBSERVATIONS

In order to assess the influence of stress-state on damage initiation, low magnification optical microscope images were taken from the double notched samples to reveal the damage propagation, prior to failure. Optical microscope images revealing damage distribution from the unbroken notch of the blunt double notched sample DN_R5, at selected conditions, are presented in Figure 5.7. Figure 5.7(i) and 5.7(ii) show damage initiation sites of samples tested at the same strain rate, $\dot{\epsilon} = 0.1 \text{ s}^{-1}$, but different temperatures; (i) at $T = 900 \text{ °C}$ and (ii) at $T = 1200 \text{ °C}$. At $T = 900 \text{ °C}$, voids were observed both at the notch root and centre of the sample. However, at $T = 1200 \text{ °C}$, damage was concentrated at the centre of the specimen and it is believed that voids nucleate at the centre and then progress to the notch root. This feature is described as ‘inside-out’ failure, a term first used by Dyson and Loveday (1981). When comparing the damage features shown in Figures 5.7(ii) and 5.7(iii), where the specimens were tested at the same temperature, $T = 1200 \text{ °C}$ but at

different strain rates; (ii) at $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and (iii) at $\dot{\epsilon}=1 \text{ s}^{-1}$, a change in damage features is observed again. At the higher strain rate, more widespread damage was observed, which is similar to the damage feature observed for the low temperature and low strain rate case shown in Figure 5.7(i). This may be because less stress relaxation occurs for the specimen tested at either low temperature or high strain rate and that hardening of the matrix due to the high triaxiality cannot compensate for growth and coalescence of voids.

As a general trend it is observed that the blunt notch sample, with lower triaxiality, shows patterns of evenly spread damage overall. For case (i) in Figure 5.7 the damage is almost equally divided between the notch root and the centre. In case (ii) of Figure 5.7, damage is accumulated more centrally but spreads over a substantial portion of the sample, exhibiting distinct ‘inside-out’ behaviour, and finally in case (iii) extensive damage is shown in the notch root but also near the centre.

In Figure 5.8 micrographs from the sharp double notched sample, DN_R1.25 are shown for the same conditions as in Figure 5.7. In case (i) of Figure 5.8, where the sample was tested at $T=900 \text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1 \text{ s}^{-1}$, damage commences at the notch root, at a direction normal to the notch surface. For this case no damage is observed at the centre of the notch. The micrographs in case (ii), where the sample has been deformed at the same strain rate as in the case (i) but at a higher temperature, $T=1200 \text{ }^{\circ}\text{C}$, unfortunately do not show damage very clearly as the material has been captured in an early stage of damage. Nonetheless, it can be observed that the material at the notch centre features more voids. The sample in Figure 5.8 (iii), deformed at a higher strain rate of $\dot{\epsilon}=1 \text{ s}^{-1}$ and a higher temperature ($T=1200 \text{ }^{\circ}\text{C}$), is unfortunately captured in a relatively early damage stage, where no distinctive damage pattern could be observed, but it is still clear that more voids have nucleated and grown near the notch root, also normal to the notch surface as in case (i) for this sample geometry. As explained for the blunt notch above, it is hypothesised that the higher temperature conditions would reduce the bias to the notch root and limit damage around that area, as seen here.

The damage initiation sites were examined more closely after etching was performed and higher magnification images are shown for each case. Some distinct characteristics were found for the sharp notch sample (DN_R1.25) when tested at $T=900 \text{ }^{\circ}\text{C}$ at a strain rate of 0.1 s^{-1} (see Figure 5.8(ii)) and for the blunt notch when tested at $T=1200 \text{ }^{\circ}\text{C}$ at the same strain rate (see Figure 5.7(ii)). In the first case damage occurs from ‘outside-in’ and in the latter from ‘inside-out’.

From the results of Figures 5.7 and 5.8 it can be postulated that damage initiation in a sharp notch is more concentrated in specific areas (notch root in this case) whilst damage initiation sites are more evenly spread on a blunter notch. A possible reason for this is that the geometrical constraint imposed by the sharp notch can lead to the material in the notch root and centre of the sample to strain by different amounts and under different strain rates. The discrepancy of the strain between the notch centre and root for sharp notch geometries has also been studied by Alves and Jones. (1999). In their work it was proposed that the combination of strains and triaxiality is a more reliable factor for predicting damage initiation regions instead of only locations in the notch plane where the value of stress triaxiality is high. Similar results are encountered here where the locations of damage initiation in the examined samples do not always correspond to positions of high triaxiality according to the Bridgman equation, which assumes the highest values to be in the centre, especially for the sharp notch geometries. This is discussed in detail in Chapter 6, in addition to the apparent temperature and strain rate dependences.

To summarise, according to the damage profiles of the micrographs presented in this section, the following observations can be made:

- As triaxiality increases and plastic straining increases at the notch root, damage tends to initiate from the notch root;
- As deformation temperature increases, damage tends to initiate from the notch centre;
- As deformation rate increases, damage tends to initiate from the notch root;

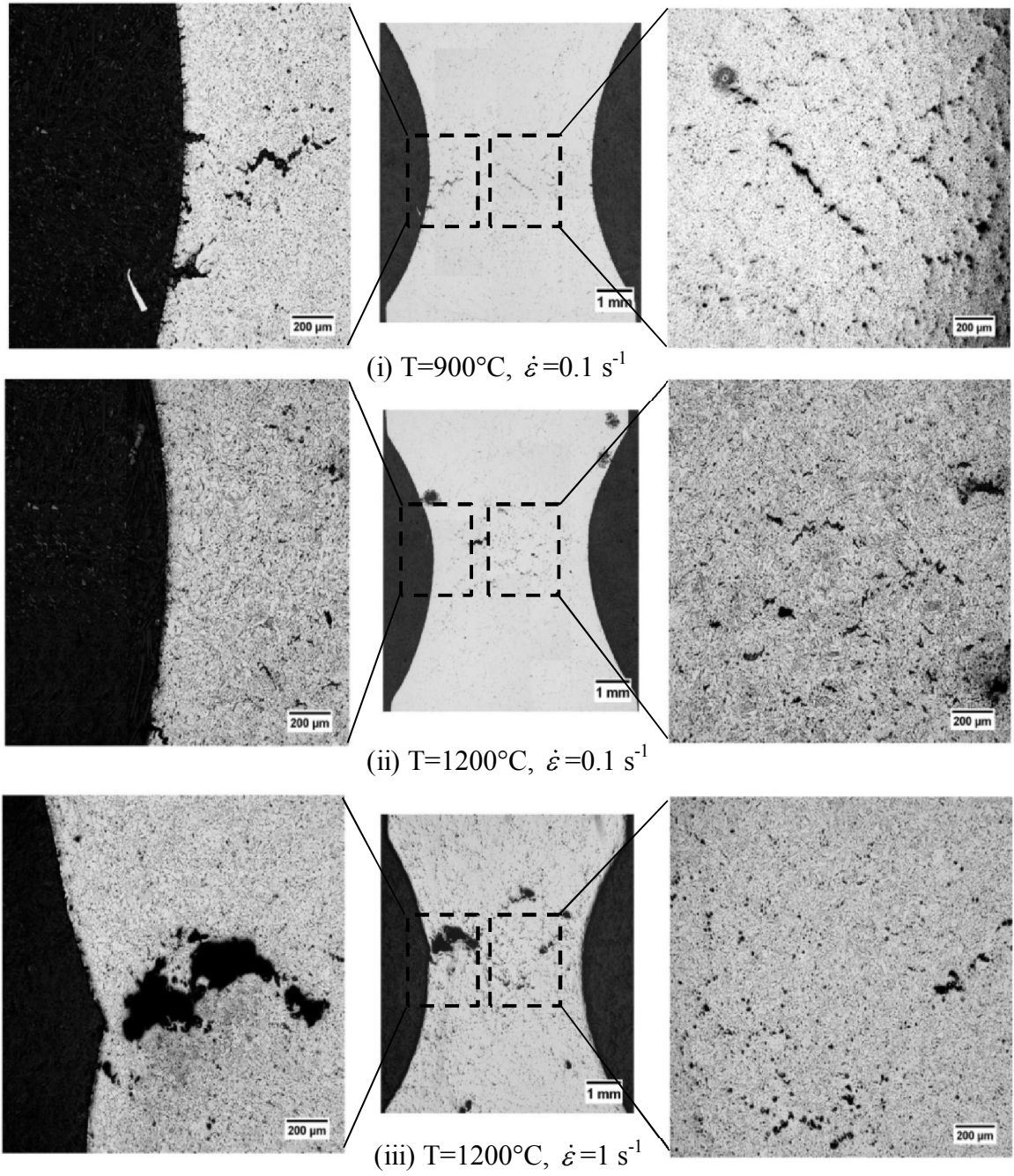


Figure 5.7: Optical microscope images of samples DN_R5 for conditions (i) $T=900^{\circ}\text{C}$, $\dot{\epsilon}=0.1 \text{ s}^{-1}$, (ii) $T=1200^{\circ}\text{C}$, $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and (iii) $T=1200^{\circ}\text{C}$, $\dot{\epsilon}=1 \text{ s}^{-1}$.

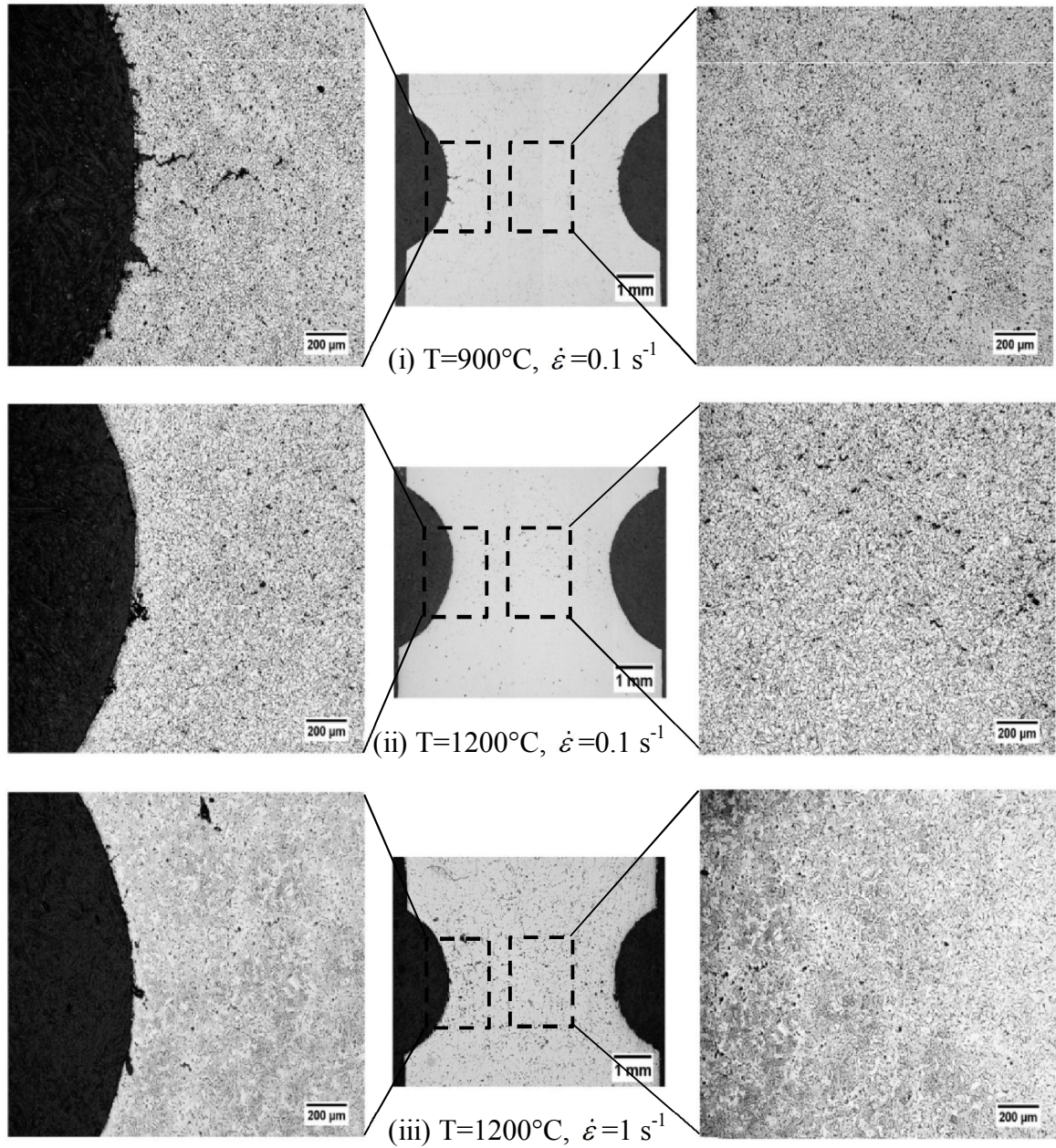


Figure 5.8: Optical microscope images of samples DN_R1.25 for conditions (i) $T=900^{\circ}\text{C}$, $\dot{\epsilon}=0.1 \text{ s}^{-1}$, (ii) $T=1200^{\circ}\text{C}$, $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and (iii) $T=1200^{\circ}\text{C}$, $\dot{\epsilon}=1 \text{ s}^{-1}$.

5.2.2 MICROSCALE DAMAGE OBSERVATIONS

To further investigate the stress-state effect, on microdamage mechanisms that occur just before fracture, a more detailed microscopic examination was performed for samples DN_R5 and DN_R1.25. Figure 5.9 shows micrographs from DN_R5 samples examined at a magnification of x50. The micrograph in Figure 5.9a is that from a sample tested at $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and $T=900^{\circ}\text{C}$ whilst

Figure 5.9b is that from a sample tested at the same strain rate but at $T=1200\text{ }^{\circ}\text{C}$. Again here the unfailed notch was examined and the areas picked for detailed examination correspond to the damage initiation areas identified in Figure 5.7 for these conditions. Two damage mechanisms were identified to coexist in the samples examined, as expected based on the literature (Lin, Liu et al., 2005; Liu et al., 2006). Both inclusion based damage and grain boundary damage were present but they became more or less prominent depending on the deformation temperature at which the sample was tested.

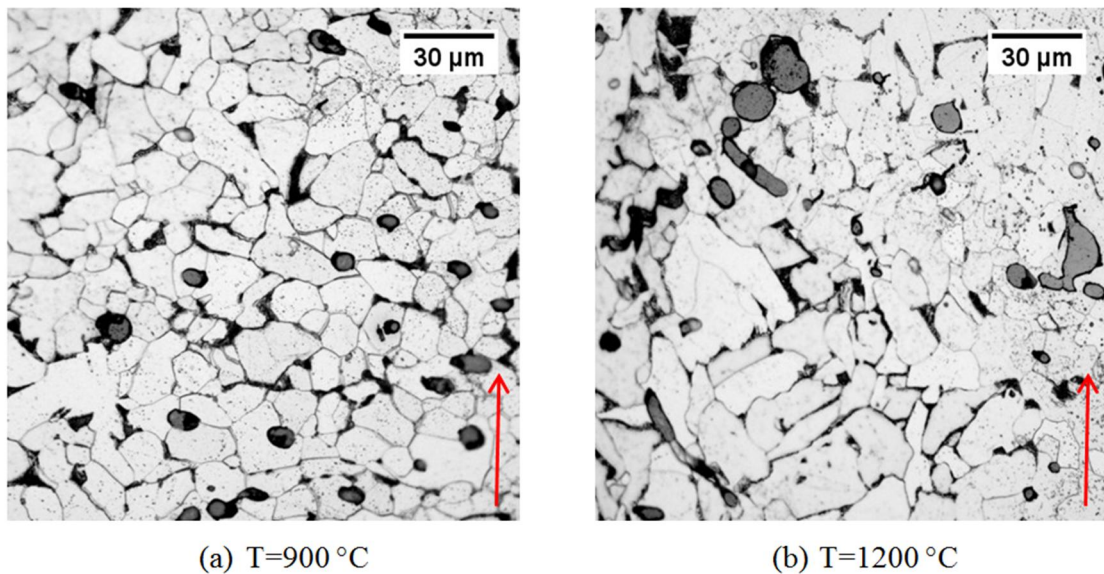


Figure 5.9: Micrographs taken from the damage initiation site for each sample as identified in Fig. 5 at magnification 50x for sample DN_R5 for deformation temperatures (a) $900\text{ }^{\circ}\text{C}$ and (b) $1200\text{ }^{\circ}\text{C}$ at a strain rate of 0.1 s^{-1} . The arrows at the bottom right of the micrographs indicate the loading direction.

Substantial damage around second phase particles is observed for sample DN_R5 tested at $T=900\text{ }^{\circ}\text{C}$ in Figure 5.9a. Grain boundary damage is also present in small amounts. The reverse trend is seen in the case of $T=1200\text{ }^{\circ}\text{C}$ in Figure 5.9b. Here inclusion related damage is still present but less dominant. There is evidence that second phase particles have debonded from the matrix, but the majority of them have no sizable voids nucleating around them. This is especially evident since there is a cluster of sizable inclusions, closely packed showing little void growth around them and no void coalescence. The fact that there is no substantial inclusion related damage in this case has to be partially affected by the fact that the particular sample region features a small distribution of second phase particles. Instead, intergranular damage is extensive in this micrograph.

Figure 5.10 also shows high magnification micrographs from double notch samples DN_R1.25 tested at $\dot{\epsilon}=0.1 \text{ s}^{-1}$ at (a) $T=900 \text{ }^{\circ}\text{C}$ and (b) $T=1200 \text{ }^{\circ}\text{C}$. Very similar microdamage response is observed for this sample with a higher proportion of inclusion related damage at the low temperature and higher proportion of grain boundary damage at the high temperature. Still both mechanisms are present in the two examined cases. Evidently the average inclusion size in Figure 5.10b is significantly smaller than that in Figure 5.10a and this would influence the dominant damage mechanism. In such cases where the second phase particles are small in size and not closely spaced together, the material can redistribute the load within the matrix better, making void growth less likely (Foster et al., 2006).

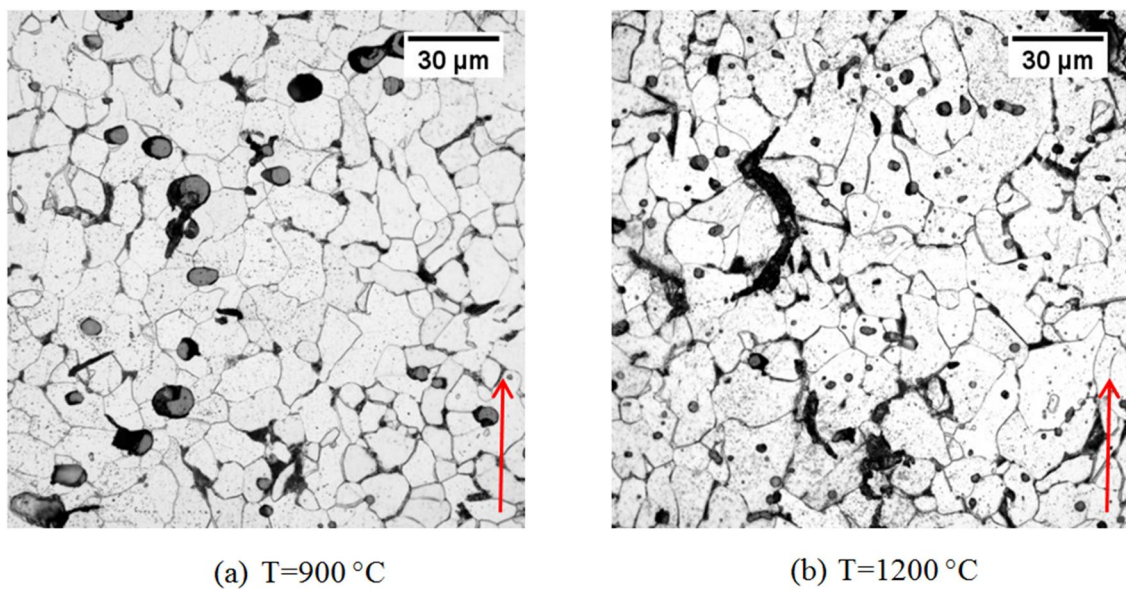


Figure 5.10: Micrographs taken from the damage initiation site for each sample as identified in Fig. 5 at magnification 50x for sample DN_R1.25 for deformation temperatures (a) 900 $^{\circ}\text{C}$ and (b) 1200 $^{\circ}\text{C}$ at a strain rate of 0.1 s^{-1} . The arrows at the bottom right of the micrographs indicate the loading direction.

It must be noted that, the second phase particle distribution in this leaded free cutting steel is not uniform and cannot be controlled since clusters of inclusions, often encountered in this material can change the mechanism of particle decohesion and promote damage in those areas. Such a cluster of inclusions is shown in Figure 5.11 from a micrograph of the DN_R1.25 sample, tested at $T=900 \text{ }^{\circ}\text{C}$ at a rate of 0.1 s^{-1} . From this micrograph it can be seen again that two damage mechanisms are present at the same time; ductile damage at second phase particles and limited grain boundary damage. In this image it is clear how closely spaced inclusions can create a high

stress network where damage is locally amplified. Outside of this inclusion cluster area whilst damage around inclusions is still obvious, it is at a much lower level.

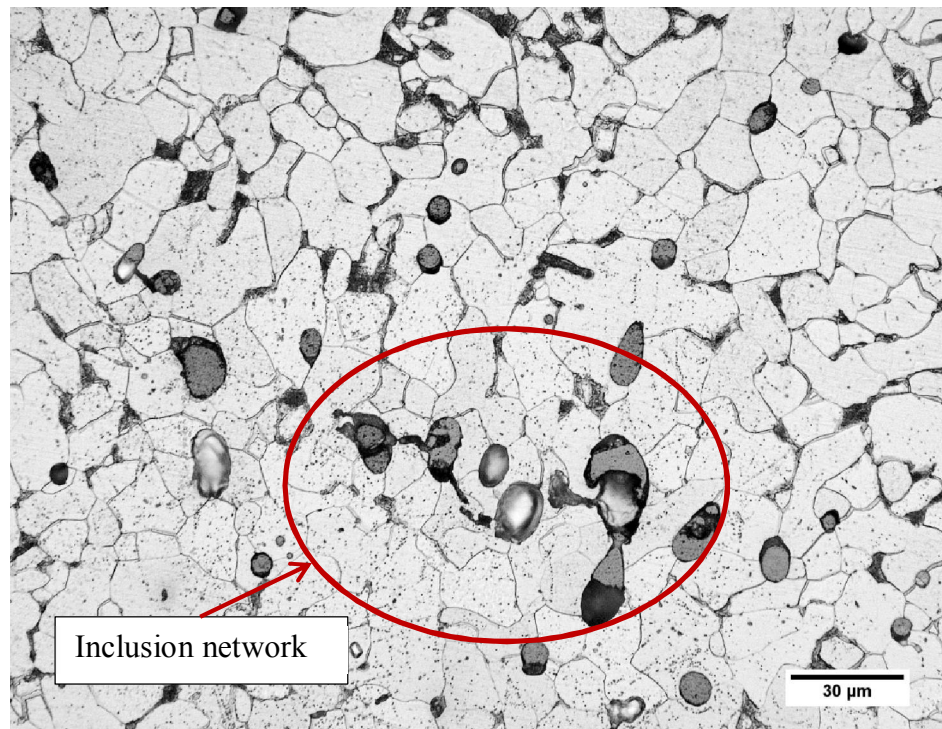


Figure 5.11: High magnification micrograph of sample DN_R1.25 (unbroken notch) at $T=900\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$.

It should be noted that the emphasis of the microstructural analysis presented here was on identifying the damage initiation patterns relative to the notch root and notch centre so that it could be correlated to the simulated stress-state of each sample (see Chapter 6). Future work could quantify the different types of damage and extend this model to link damage micromechanisms to the stress-state. This would greatly enhance the ability of the model to make predictions but also increase the complexity.

5.3 CONCLUSIONS

In this chapter the results from high temperature Gleeble uniaxial and double notch tension tests were presented. The uniaxial tensile test results confirmed the expected viscoplastic response of the material and were gathered to later calibrate the viscoplastic damage model for pure tension conditions. Additionally, the viscoplastic response of the double notched samples was investigated and the influence of stress-state on material ductility was examined.

The effect of stress-state on damage initiation was investigated under conditions analogous to those of hot rolling via metallographic examinations of failed double notched samples where the notch that did not fail was examined to reveal the pre-failure response. From these tests two distinct cases of damage initiation were observed. At low temperatures ($T=900\text{ }^{\circ}\text{C}$) and low strain rates ($\dot{\epsilon}=0.1\text{ s}^{-1}$) damage in the sharp notch (sample DN_R1.25) developed in the notch root first; damage occurs from the ‘outside-in’. The opposite is seen for the blunt notch (sample DN_R5) deformed at the same strain rate but a higher temperature ($T=1200\text{ }^{\circ}\text{C}$). Damage in this case is from the ‘inside-out’. Therefore, stress-state has been identified as an important parameter affecting material ductility because of the geometrical constrain that arises from sharp notch geometries.

The high magnification micrographs examined reveal the presence of two damage mechanisms; grain boundary damage and inclusion based damage. For a low deformation temperature i.e. $900\text{ }^{\circ}\text{C}$, extensive amount of damage around second phase particles has been observed. On the other hand, for higher deformation temperatures ($1200\text{ }^{\circ}\text{C}$), grain boundary damage has been discovered to be more significant. In all cases particle/matrix decohesion has been recognised to be one of the main mechanisms by which voids nucleate.

CHAPTER 6

NUMERICAL STUDIES OF STRESS-STATE VARIATIONS FOR NOTCHED BAR SAMPLES

SUMMARY

In this chapter, the experimental results of the failure initiation sites are compared with the computed values of the stress fields around the notch profiles. The computed values were obtained from FE analysis using a tailored constitutive model, calibrated from uniaxial tension tests data (presented in Chapter 5). Additionally, temperature profiles occurring during the high temperature mechanical testing have been accommodated for and included in the FE calculations of the stress-state fields. This analysis aims to identify the link between damage initiation and stress-state. For this reason, the profiles of three stresses often associated with multiaxial damage were chosen to be examined across the notch plane. Those are; maximum principal stress, σ_1 , hydrostatic stress, σ_H , and equivalent stress, σ_e .

6.1 NUMERICAL PROCEDURES

6.1.1 MATERIAL MODEL CALIBRATION

A material model, originally developed by Lin et al.(2004) for a C-Mn steel, was used in this study to describe the viscoplastic behaviour of leaded FCS. In the original model, recrystallisation was incorporated and its interactive effects were represented in the dislocation density and grain size equations. A leaner formulation of the same model, tailored for a leaded as-cast FCS, was presented by Foster et al. (2009a) and was used in this study. In this model, recrystallisation was omitted from the constitutive equations as it was not deemed a defining mechanism of the particular material, as evident from the shape of the flow stress (see Chapter 5). The model also excluded the description of damage since only the stress-state was of interest at this stage. The equations are summarised below:

$$\dot{\varepsilon}_p = \dot{\varepsilon}_0 \left\langle \frac{\sigma - (k + R)}{K} \right\rangle_+^n \quad 6.1$$

$$\dot{\bar{\rho}} = k_1 (1 - \sqrt{\bar{\rho}}) \dot{\varepsilon}_p - c_1 \bar{\rho}^{c_2} \quad 6.2$$

$$\sigma = E (\varepsilon_T - \varepsilon_p) \quad 6.3$$

$$R = B \sqrt{\bar{\rho}} \quad 6.4$$

The temperature dependant material constants are:

$$k = k_0 \exp \left(\frac{Q_p}{\kappa T} \right) \quad 6.5$$

$$n = n_0 \exp \left(\frac{-Q_n}{\kappa T} \right) \quad 6.6$$

$$K = K_0 \exp \left(\frac{Q_K}{\kappa T} \right) \quad 6.7$$

$$k_1 = A_0 \exp \left(\frac{-Q_A}{\kappa T} \right) \quad 6.8$$

$$c_1 = c_{1(0)} \exp \left(\frac{-Q_R}{\kappa T} \right) \quad 6.9$$

$$c_2 = c_{2(0)} \exp \left(\frac{-Q_{c2}}{\kappa T} \right) \quad 6.10$$

$$E = \frac{E_0}{\cosh^2 (K_1 (T - T_0))} \quad 6.11$$

$$B = \frac{B_0}{\cosh^2 (K_1 (T - T_0))} \quad 6.12$$

Equation 6.1 describes the plastic strain rate as a function of the threshold stress, k , isotropic hardening, R , flow stress, σ , drag stress, K and the viscoplastic exponent, n . The threshold stress,

k , below which no plastic deformation (including creep) occurs, is a temperature dependent parameter defined according to equation 6.5. The isotropic hardening of the material, R , is modelled proportional to the square root of the dislocation density as given in equation 6.4. In equation 6.4, B , is a temperature dependent stress (equation 6.12). Drag stress, K , and the viscoplastic exponent, n , are both related to temperature via an Arrhenius relationship (equations 6.7 and 6.6 respectively). The normalised dislocation density equation (equation 6.2) models the accumulation of dislocations due to plastic deformation together with the dynamic recovery of the dislocation density (first term). The second term of equation 6.2 represents the static recovery, i.e. the reduction of dislocations (annealing) under hot rolling conditions. Both parameters c_1 and c_2 in the normalised dislocation density equation are temperature dependent according to equations 6.9 and 6.10 respectively. The elastic modulus, E , is modelled with a similar expression to B (see equation 6.11). In equations 6.11 and 6.12, T_0 , is a reference temperature (900 °C) at which E_0 and B_0 are defined. Finally the flow stress is given in equation 6.3.

The values for the temperature dependent material constants are given in Table 6.1. These constants were determined from the uniaxial experimental data, presented in Chapter 5, for the temperature range 900-1200 °C and strain rates varying from 0.1-10 s⁻¹. The model calibration was performed manually in MATLAB and the results are shown in Figure 6.1, for the applied temperature range, and Figure 6.2, for the strain rate range. In both Figure 6.1 and 6.2 solid lines represent calculated values and symbols represent experimental data. In this model calibration, temperature dependency was added to more constants as compared to the model of Chapter 4. This was possible due to the fact that damage was not included, therefore making it easier to calibrate even with a higher number of temperature dependent coefficients.

Table 6.1: Material constants used in the viscoplastic constitutive equations for a leaded FCS.

Constant	T_0 (°C)	K_0 (MPa)	Q_K (J/mol)	k_0 (MPa)	Q_p (J/mol)	K_1 (1/K)	B_0 (MPa)	Q_{c2} (J/mol)
Value	900	2.7	2.5e4	3.174	14738.23	0.004	160	5263.8
A_0 (-)	Q_A (J/mol)	$c_{1(0)}$ (1/s)	Q_R (J/mol)	$c_{2(0)}$ (-)	Q_{c2} (J/mol)	E_0 (MPa)	n_0 (-)	Q_n (J/mol)
265.07	39503.3	5.52	5543.15	0.8	5.263e3	19.9e3	6.4	9e3

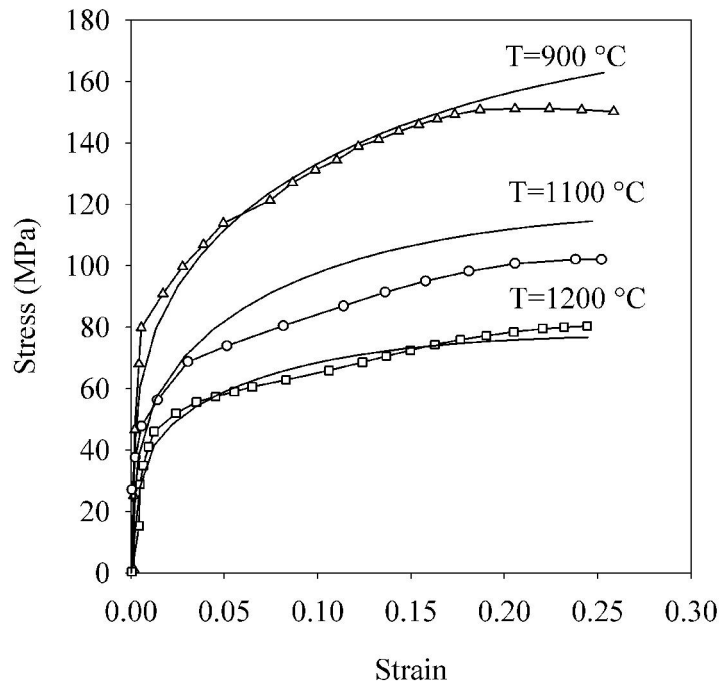


Figure 6.1: MATLAB fitting (lines) versus experimental values (symbols) for $\dot{\epsilon}=1\text{ s}^{-1}$ at three temperatures.

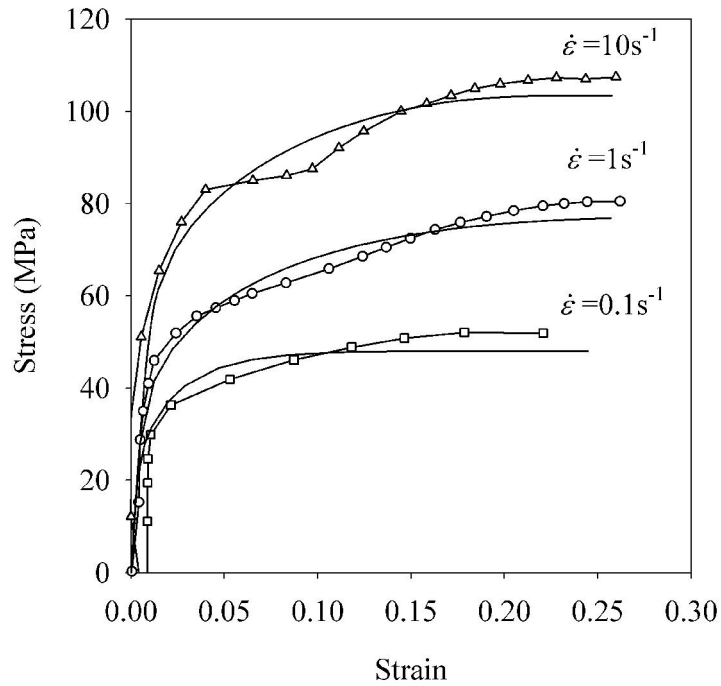


Figure 6.2: MATLAB fitting (lines) versus experimental values (symbols) for $T=1200\text{ }^{\circ}\text{C}$ at three strain rates.

6.1.2 FE MODEL

FE analysis was used to gain insight into the distribution of the stress and strain across the notch plane of each double notched sample used in this investigation without consideration of any ductile damage that would change the stress/strain distribution. The aim here is to understand how the triaxiality, maximum principal stress, hydrostatic and equivalent stress influence void nucleation and growth and how much they ultimately affect the failure initiation. For this purpose, an explicit mechanical FE model of the notched samples was created to simulate the high temperature Gleeble tensile tests performed in the experimental investigation. In order to accurately represent the experimental conditions, the thermal profiles induced by the Gleeble machine, calculated in Chapter 4, were used in the mechanical simulations. A detailed description of the electro-thermal simulations performed to calculate these thermal profiles was outlined in Chapter 4.

The two axisymmetric models used in the mechanical, dynamic, explicit simulations are shown in Figure 6.3; the same models were used in the electro-thermal simulations so that mesh consistency was achieved. Symmetry conditions were applied on both the transverse and longitudinal planes of the samples, as indicated in Figure 6.3. The displacement was applied as a continuously changing velocity profile calculated assuming a constant gauge length so that the model is deformed at a near-constant strain rate. The position where the displacement was applied is also shown in Figure 6.3.

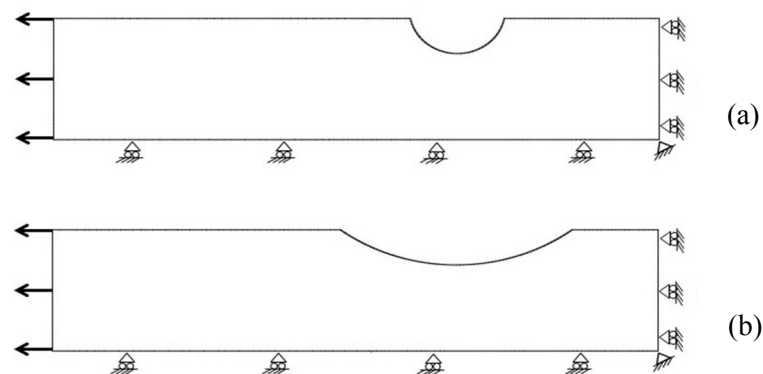


Figure 6.3: Finite element models used for the analysis of samples (a) DN_R1.25 and (b) DN_R5.

Basic mesh sensitivity studies were performed to ensure the correct mesh size was used. Six mesh sizes were investigated, with the largest mesh size being 1 (62 elements for sample DN_R1.25

and 64 elements for sample DN_R5) and the smallest being 0.1 (3181 elements for DN_R1.25 and 3498 elements for DN_R5). These six simulations were performed for both notched sample geometries, DN_R5 and DN_R1.25, at a high temperature, $T=1200\text{ }^{\circ}\text{C}$, at $\dot{\epsilon}=0.1\text{ s}^{-1}$. The differences in the produced stress versus strain curves from the six different models were insignificant but it was still important to evaluate how particular stresses change as a function of the mesh size for individual nodes. For this reason the equivalent stress, σ_e , and maximum principal stress, σ_1 , were chosen. Figure 6.4 shows the effect of mesh size on sample DN_R5 for the two selected stresses and Figure 6.5 for sample DN_R1.25. In both figures the data were extracted from a node on the notch root of the samples.

From Figure 6.4 it is clear that the changes in the examined values are not significant, although, for both cases a plateau in the values is reached as mesh size density increases. The same behaviour is seen in Figure 6.5 for the sharp notch DN_R1.25. However, the changes in the stress values of Figure 6.5 are higher as compared to the equivalent ones for the blunt notch.

The results of the mesh sensitivity study suggest that using a mesh size of at least 0.1 for both notched samples is necessary for this analysis, as accurate stress values need to be extracted from nodes along the notch plane of each sample. The meshes for each simulated sample are shown in Figure 6.6 below. The elements used were four-node axisymmetric reduced integration quadrilateral (CAX4R) and simulations were performed in ABAQUS 6.11.

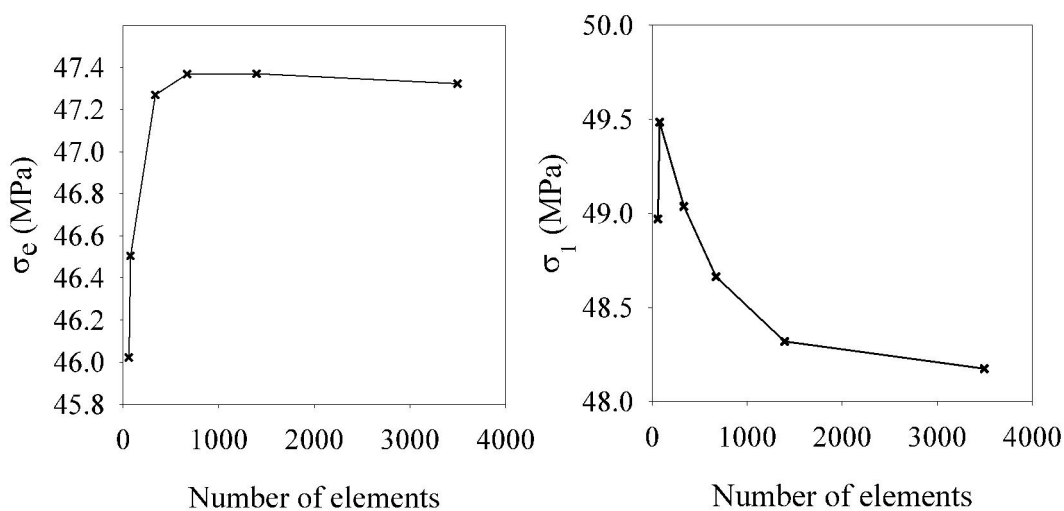


Figure 6.4: Convergence study for an element on the notch root of sample DN_R5 simulated at $\dot{\epsilon}=0.1\text{ s}^{-1}$ and $T=1200\text{ }^{\circ}\text{C}$ for maximum principal, σ_1 , and equivalent stress, σ_e , at $t=0.15$ sec.

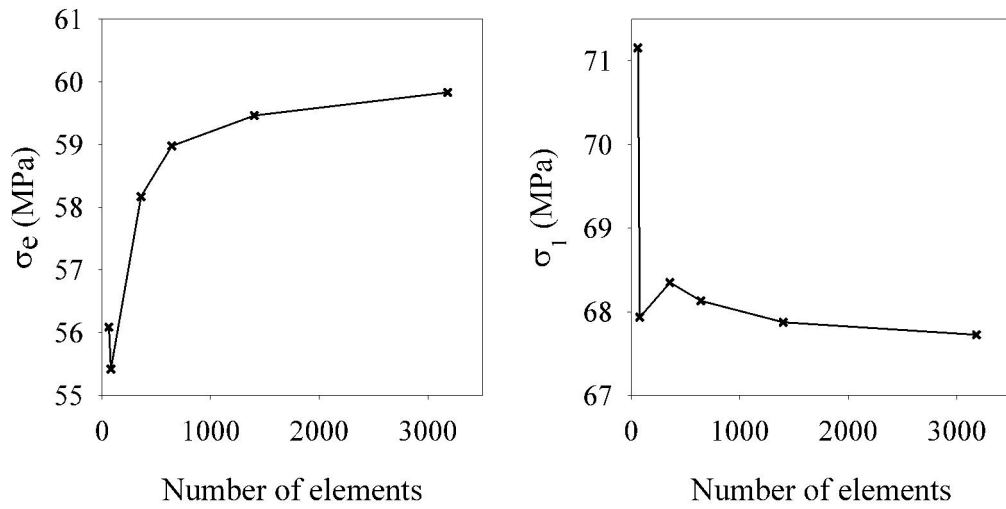


Figure 6.5: Convergence study for an element on the notch root of sample DN_R1.25 simulated at $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and $T=1200 \text{ }^{\circ}\text{C}$ for maximum principal and equivalent stress at $t=0.6 \text{ sec}$.

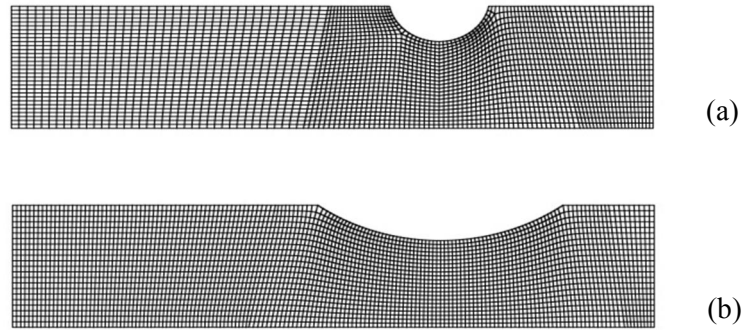


Figure 6.6: Meshes used for the double notched simulations of samples DN_R5 and DN_R1.25.

6.1.3 EXPERIMENTAL VALIDATION

A set of simulations corresponding to the experiments were carried out for both double notched samples under two conditions; $\dot{\epsilon}=0.1 \text{ s}^{-1}$, $T=900 \text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1 \text{ s}^{-1}$, $T=1200 \text{ }^{\circ}\text{C}$. The constitutive equations described in section 6.1.2 were implemented via a user-defined subroutine VUMAT in ABAQUS/Explicit.

In Figure 6.7 experimental data from the double notched samples DN_R1.25 and DN_R5 are compared with the FE predictions. The comparison was performed for two strain rates, $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and $\dot{\epsilon}=1 \text{ s}^{-1}$, and two temperatures, $900 \text{ }^{\circ}\text{C}$ and $1200 \text{ }^{\circ}\text{C}$. A satisfactory match between the

computed and experimental data confirmed that the FE model accurately describes the viscoplastic behaviour of the leaded FCS and was, therefore, calibrated correctly. The FE simulations were terminated at the steady state condition, and were used to analyse stress redistribution in the material under different deformation conditions, before failure occurred.

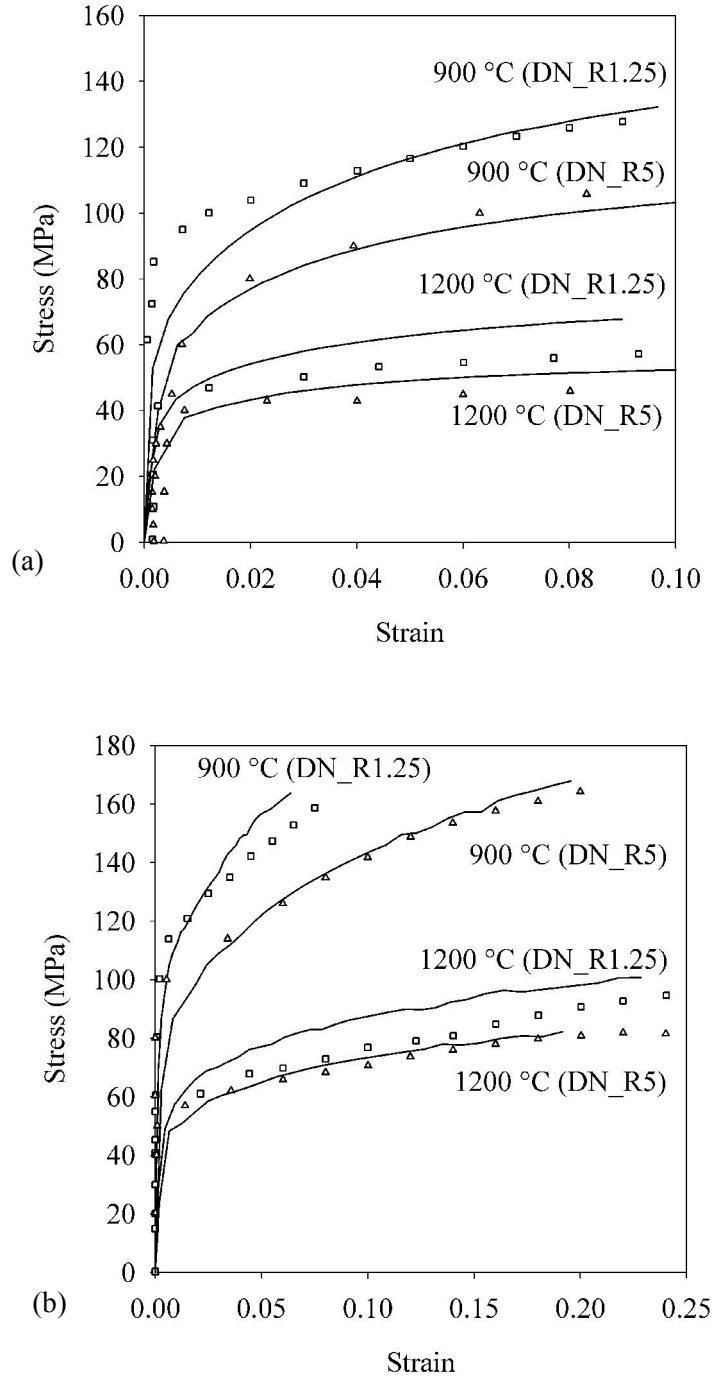


Figure 6.7: Comparison of computed (solid curves) and experimental (symbols) stress-strain relationships for (a) $\dot{\epsilon} = 0.1 \text{ s}^{-1}$ and (b) $\dot{\epsilon} = 1 \text{ s}^{-1}$ at temperatures 900 °C and 1200 °C.

6.2 COMPUTATIONAL RESULTS AND ANALYSIS

6.2.1 EFFECT OF NOTCH GEOMETRY ON THE STRESS-STATE VARIATION

In Figure 6.8 the axial strain histories for samples (a) DN_R1.25 and (b) DN_R5 are shown for two locations on the notch plane; A, the notch centre and C, the notch root, identified on the FE axial strain contours. These values were calculated for $T=900\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$. The computational results confirm the hypothesised trend from the experimental results of Chapter 5, section 5.2.1. In Figure 6.8a for the sharp notch (DN_R1.25) the strain levels from location A and location C differ significantly; the material in the notch root is strained considerably more, almost by a factor of 3. This explains the non-uniform damage profiles identified experimentally in the sharp notch (see Figure 5.8, Chapter 5). Conversely, for the blunt notch (DN_R5) the strains from locations A and C are similar, with the maximum strain at the notch root just 1.2 times higher than at the centre, suggesting that damage would initiate more evenly across the notch plane. This is also confirmed experimentally in Chapter 5, Figure 5.7.

For the same deformation conditions ($T=900\text{ }^{\circ}\text{C}$, $\dot{\epsilon}=0.1\text{ s}^{-1}$), the maximum principal, σ_1 , hydrostatic, σ_H , and equivalent, σ_e , stresses are plotted against equivalent strain in Figure 6.9 for the same locations A and C, as defined in Figure 6.8. A significant difference in equivalent stress between A and C is seen for the sharp notch, when compared with the blunt notch. This non-uniform stress distribution in the sharp notch may account for the particularly high strain and damage levels at the notch root where stress is highest. Conversely, the more uniform stress distribution of the blunt notch accounts for a more even damage profile, as seen experimentally and is further supported by the quasi uniformity of the strain level.

From the experimental results presented in Chapter 5 and from the literature (Hancock and Brown, 1983; Hancock and Mackenzie, 1976; Mackenzie et al., 1977) it is clear that the higher the value of triaxiality (sharper notch geometries), the lower the ductility. It is, therefore, important to know how triaxiality is distributed across the notch planes and how that correlates with the damage initiation sites.

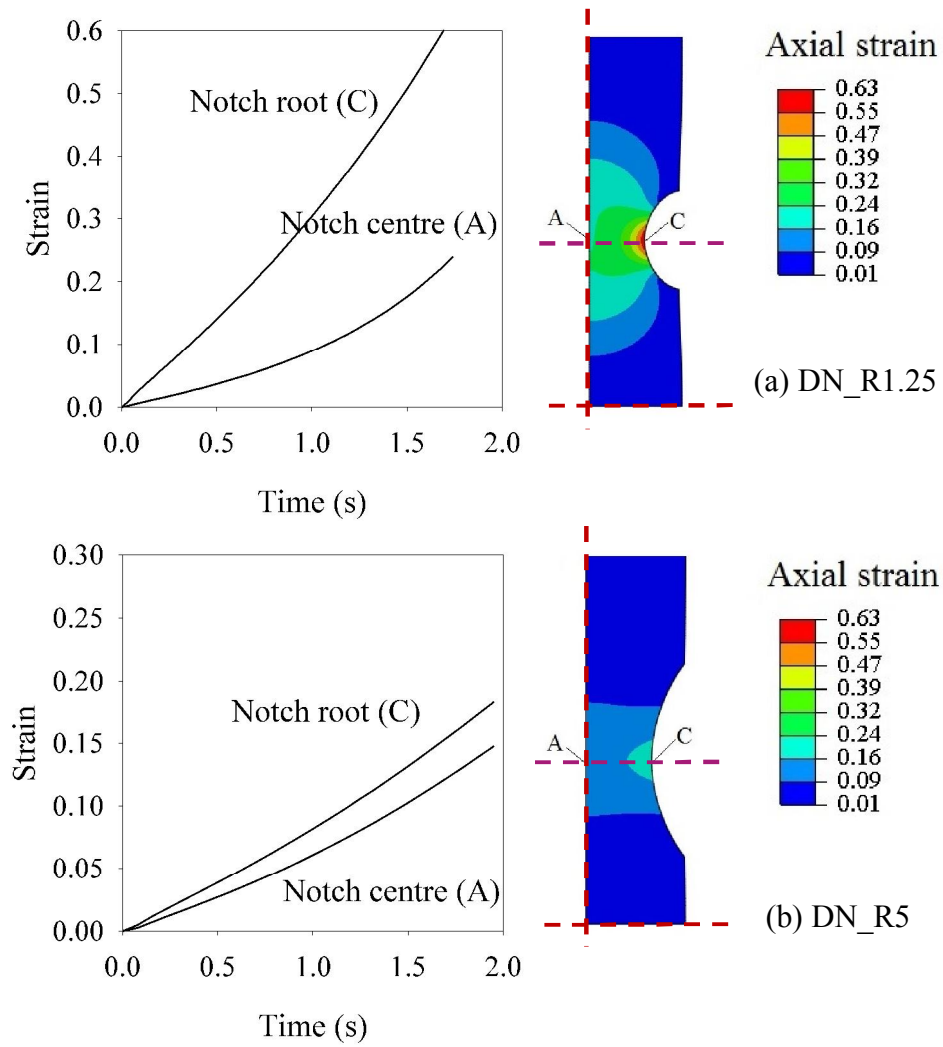


Figure 6.8: Strain histories, in the loading direction, for a point in the notch centre (location A) and a point in the notch root (location C) for (a) sample DN_R1.25 and (b) DN_R5 calculated at $\dot{\epsilon} = 0.1 \text{ s}^{-1}$ and $T=900 \text{ }^{\circ}\text{C}$. Contour plots for the axial strain in the same direction are also shown here together with the annotated locations from where the data were extracted from. The dashed red lines indicate the symmetry axes.

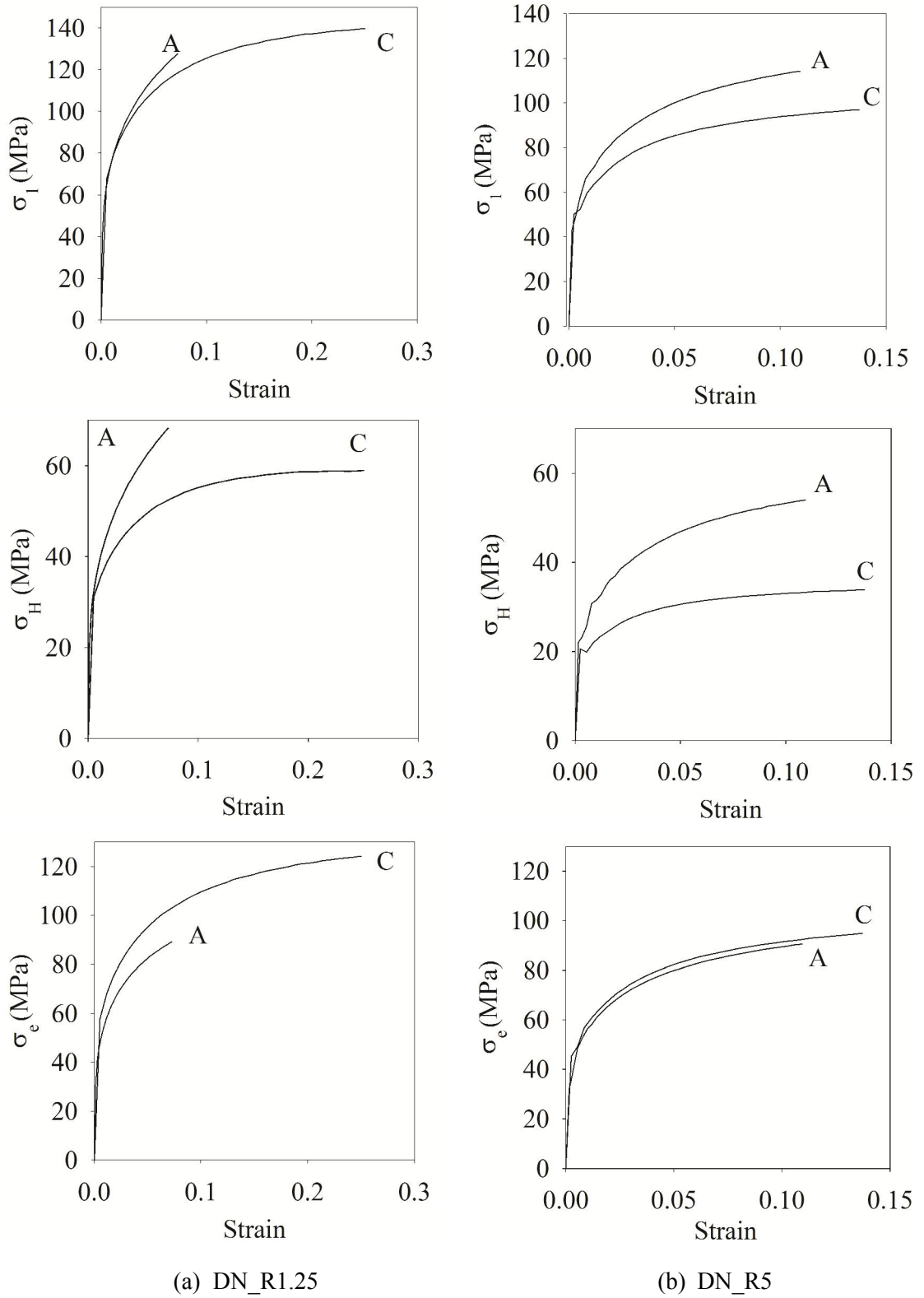


Figure 6.9: Maximum principal, hydrostatic and equivalent stresses versus strain curves for samples DN_R1.25 and DN_R5 at $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and $T=900 \text{ }^{\circ}\text{C}$ calculated for a point on the outer centre of each sample (C) and at a point at the inner centre of the sample (A) as defined in Figure 6.8.

In Figure 6.10, FE calculated triaxiality values are plotted against the normalised distance from the inner notch centre for samples DN_R1.25 and DN_R5 for two strain rates; $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and $\dot{\epsilon}=10 \text{ s}^{-1}$. Here the model was used to extrapolate to testing conditions that have not been examined experimentally (i.e. $\dot{\epsilon}=10 \text{ s}^{-1}$) so that the two extreme cases of the strain rate range for hot rolling applications can be considered. The triaxiality changes as a function of distance across the notch plane. The blunt notch, DN_R5, has a lower average triaxiality compared to the sharper notch DN_R1.25, as expected, and its maximum triaxiality corresponds to the centre of the notch whilst for the sharp notch the peak value is located between the notch centre and the notch root. For both sample geometries the lowest triaxiality is calculated at the notch root. A sharp decrease in triaxiality is calculated for the sharp notch (DN_R1.25), whilst for the blunt notch the decrease in triaxiality from the notch centre to the notch root is considerably more gradual. Whilst this trend is true for both examined strain rates, for each individual sample the drop in triaxiality decreases with higher strain rate; although, this effect is small when considering the two orders of magnitude change in the strain rate. For example, in the case of the blunt sample, DN_R5, the percentage decrease from the notch centre to the notch root for $\dot{\epsilon}=0.1 \text{ s}^{-1}$ is 38.4% and the equivalent value for $\dot{\epsilon}=10 \text{ s}^{-1}$ is 30.2%. In the case of the sharp notch the percentage decrease is very similar for the two examined strain rates, although the decrease is still smaller for the higher strain rate (28.4% compared to 29.2% decrease for $\dot{\epsilon}=0.1 \text{ s}^{-1}$).

Therefore, the positions of high triaxiality across the notch plane for both samples do not correlate well with the positions of damage initiation observed experimentally (see section 5.2.1). This is especially true for the sharp notch (see Figure 6.10) where damage is evidently concentrated in the notch root, the location with the lowest triaxiality as calculated for $\dot{\epsilon}=0.1 \text{ s}^{-1}$ at $T=900 \text{ }^{\circ}\text{C}$ in the same figure. Consequently, triaxiality is not the only parameter affecting damage initiation and, perhaps, a simplistic relationship that assumes high triaxiality to be the sole cause of damage initiation is misleading. Additionally, many studies investigating stress-state effects and damage initiation use Bridgman's analysis (Hancock and Brown, 1983; Mackenzie et al., 1977) so triaxiality is always found to be highest in the centre of the specimens for axisymmetric specimens, in contrast with the results presented here. In some studies (Becker et al., 1988; Needleman and Tvergaard, 1984), although the highest triaxiality was calculated at the centre of the investigated samples, damage was observed to initiate at other sites as well as the centre of the notch. For this reason, the cause of damage initiation should instead be thought of in terms of plastic strain inhomogeneity, hardening of the matrix and stress triaxiality. Microstructural

discontinuities such as second phase particles and porosity would also influence the damage evolution, but these aspects are not examined in this work.

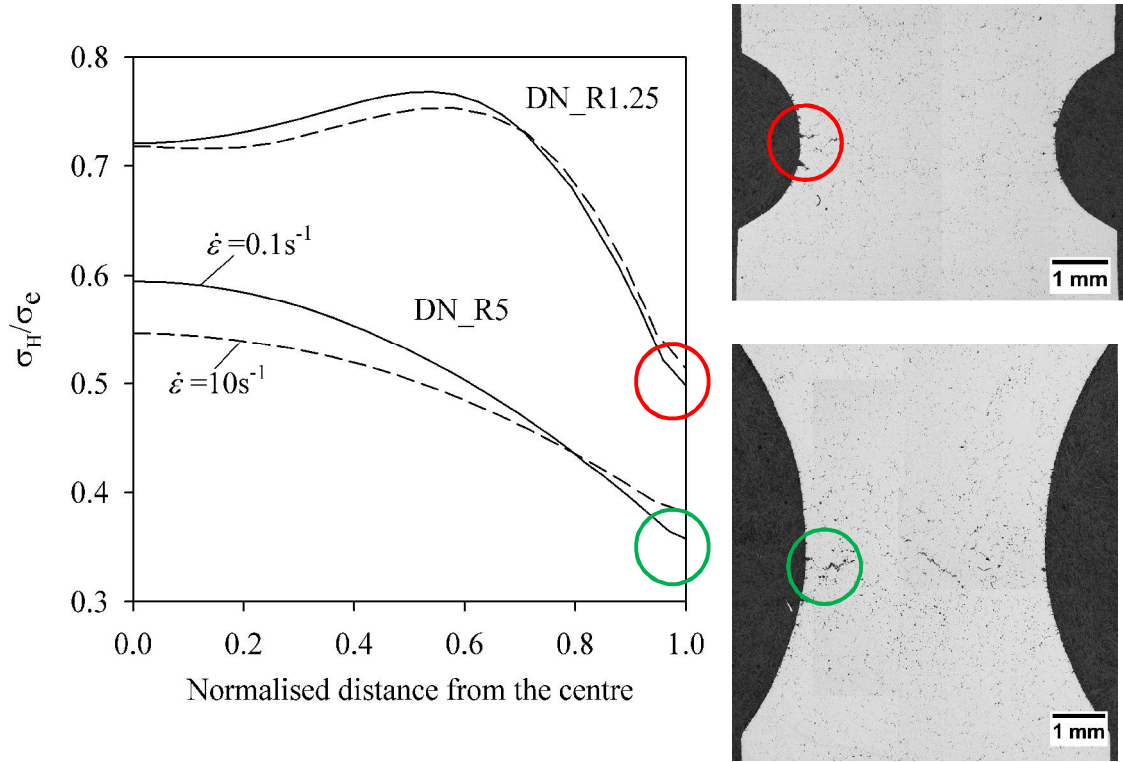


Figure 6.10: Triaxiality distribution across the notch plane for samples DN_R5 and DN_R1.25 for $T=900\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1$ (solid line) and 10 s^{-1} (dashed line). The computed data correspond to an average axial strain of 0.1. Micrographs of the damage profiles of samples DN_R5 and DN_R1.25 are also shown for conditions $T=900\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$ for comparison with the FE calculated triaxiality profiles.

Figure 6.11 shows σ_H/σ_e against equivalent strain for three points on the minimum cross section of the notch; A (centre), B (middle) and C (outer). This was done for both double notched samples straining at a rate of 0.1 s^{-1} and at temperatures of $900\text{ }^{\circ}\text{C}$ (solid curves) and $1200\text{ }^{\circ}\text{C}$ (dotted curves). The results indicate that temperature variation does not affect the triaxiality significantly but, as expected, position does. From the triaxiality distributions it can be noticed that the Bridgman solution (1.02 for the sample DN_R1.25) does not apply well to sharp notch geometries, whereas it gives a closer estimation for the blunt notch DN_R5 (0.55). It is also evident, that although for the blunt notch the selected locations A, B and C follow similar strain paths, this is not the case for the sharp notch sample. Figure 6.11 confirms the trend observed in Figure 6.8 and Figure 6.9, that the sharp notch imposes a geometrical constraint that in turn leads to inhomogeneous straining of the material from different points on the notch plane. In addition,

the effect of temperature seems to have a greater effect on the triaxiality of the centre position, A, of the blunt notch than the same position in the sharp notch. This results in more damage at the centre of the blunt notch when deforming at a higher temperature, as shown in Figure 5.5(ii) in Chapter 5.

To understand how the damage features presented in Chapter 5 have initiated, stress distributions of samples DN_R5 and DN_R1.25 are plotted against the normalised distance across the minimum cross section of each notch in Figure 6.12 for a strain rate of 0.1 s^{-1} . The distributions were calculated from FE analysis for temperatures 900 °C and 1200 °C, and the values of the extracted maximum principal, σ_1 , hydrostatic, σ_H , and equivalent, σ_e , stresses were normalised with respect to the net stress, σ_A (the average stress across the notch root). The values correspond to an average axial strain value of 0.1, when steady state has been reached (before non-uniformity takes place during tensile testing for all tensile geometries considered). In Figure 6.12a, σ_1/σ_A shifts its maximum from the centre to the notch root for increasing notch acuity ratio. For both sample geometries, as the temperature decreases σ_1/σ_A increases at the notch root and decreases at the centre of the notch. The increase at the notch root is more significant for the sharp notch geometry. Therefore, it is expected that for the same testing conditions the sharp notch would experience more damage initiating at the notch root. This is indeed the case as seen in Figures 5.7(ii) and 5.8(ii), in Chapter 5. The calculated values of σ_1/σ_A are shifted towards the centre at higher temperatures for both geometries. This bias towards damage at the centre for high temperatures is confirmed experimentally by comparing Figure 5.7(i) to 5.7(ii), for the blunt notch, and Figure 5.8(i) with 5.8(ii), for the sharp notch. Looking back at the results for DN_R5 in Figure 5.7 (in particular, case (ii)) where damage distinctly commences from the ‘inside-out’, it can be seen that the maximum principal stress is highest in that region. In fact, the normalised stresses experienced at the notch centre are 23% higher than those at the notch root.

A similar trend was observed for the σ_H/σ_A distributions, shown in Figure 6.12b. Although in this case, the values for the blunt and sharp notch are similar in the central region, it is shown to increase for the sharp notch and decrease for the blunt notch towards the notch root. The hydrostatic stresses across the notch planes for both samples are lower than the net stress, σ_A , for all cases presented in Figure 6.12b.

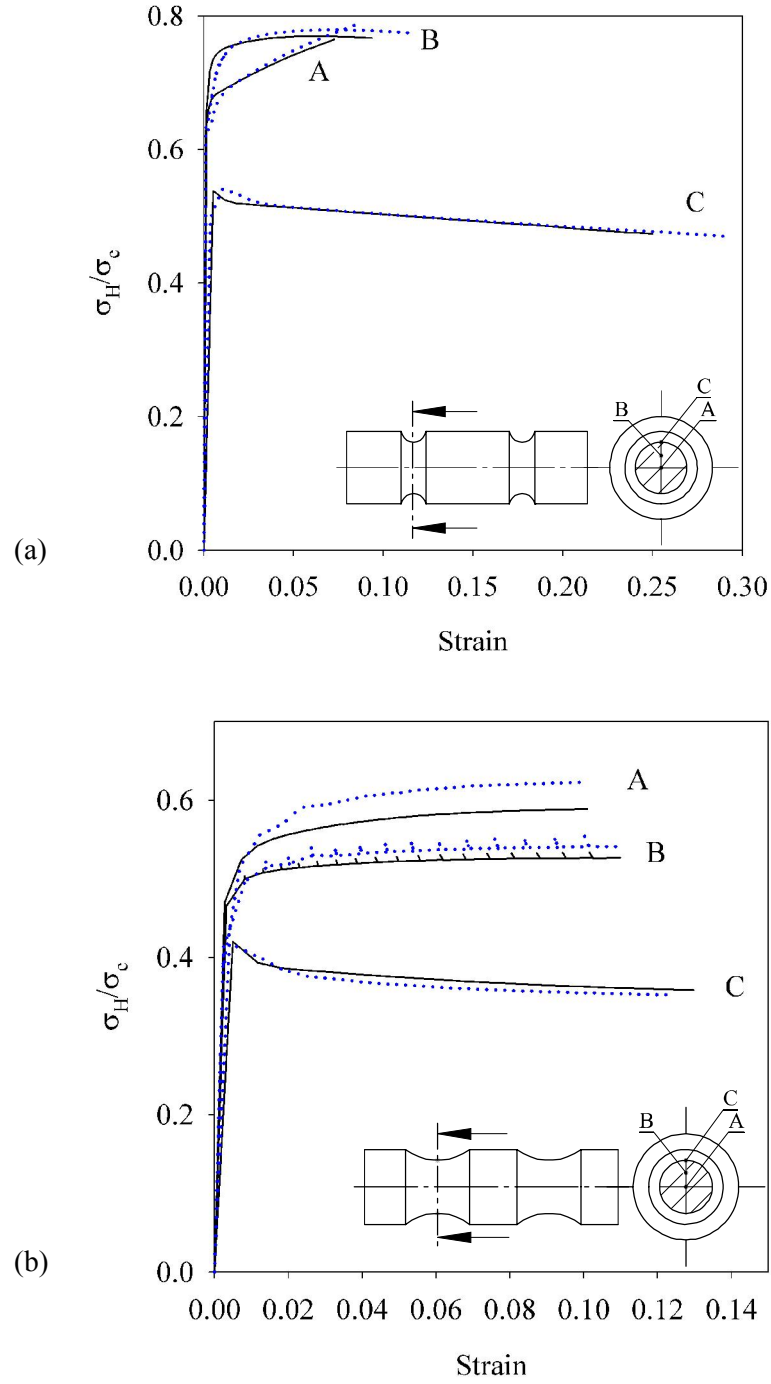


Figure 6.11: Triaxiality evolution with equivalent strain for three points on the notch plane for (a) DN_R1.25 and (b) DN_R5 at T=900 °C (solid curves) and T=1200 °C (dotted curves) for $\dot{\epsilon}=0.1 \text{ s}^{-1}$.

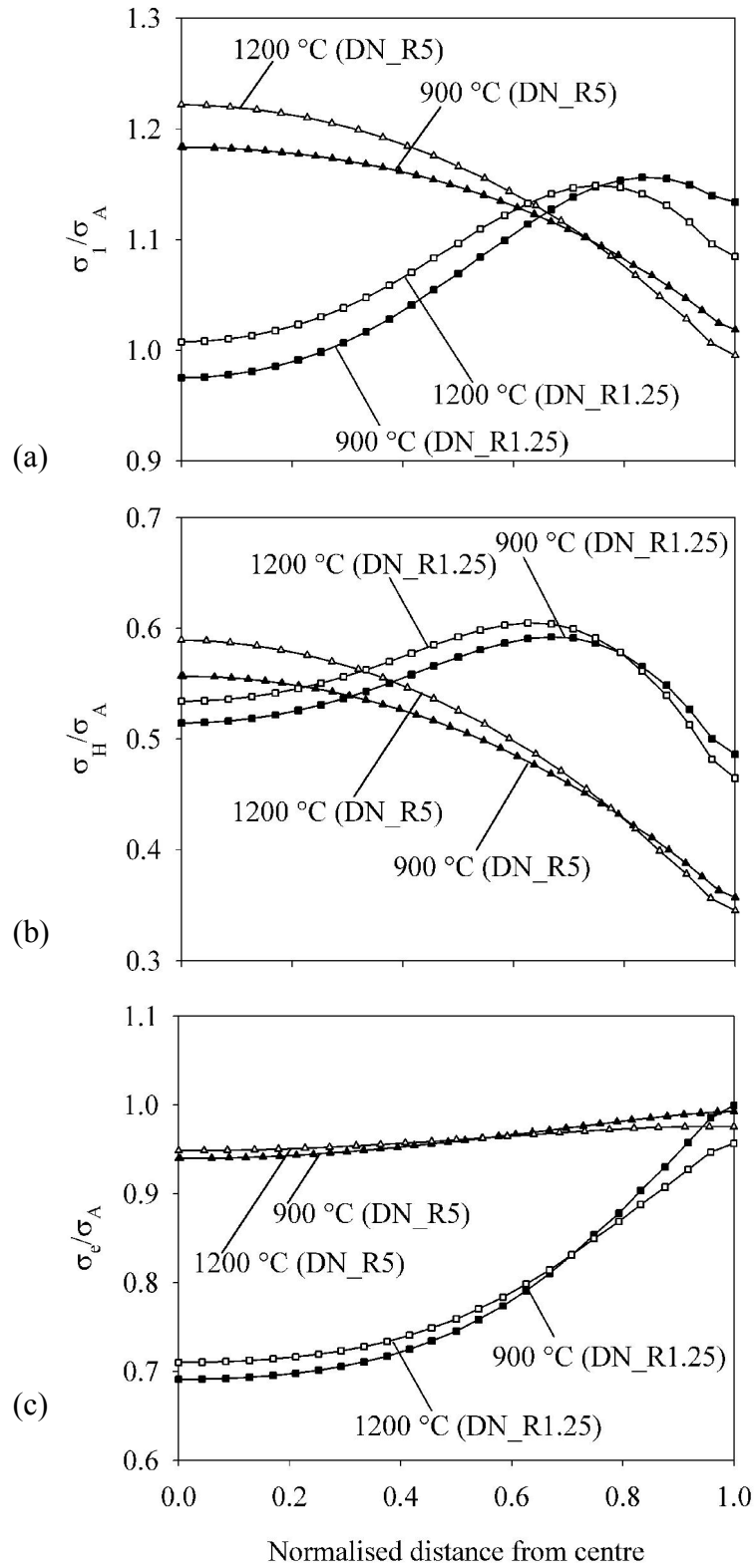


Figure 6.12: Evolution of normalised, with net stress, (a) maximum principal stress, (b) hydrostatic stress and (c) equivalent stress across the notch plane for samples DN_R1.25 and DN_R5. Data were extracted from ABAQUS tensile simulations calculated for $\dot{\epsilon} = 0.1 \text{ s}^{-1}$ for the case where steady state is reached and true strain equals 0.1.

The distributions for the equivalent stress, σ_e/σ_A , are very different from those presented in Figure 6.12a and Figure 6.12b. The equivalent stress values for the blunt notch remain fairly constant across the notch plane, suggesting that nucleated voids would grow more or less equally across the notch plane, whilst for the DN_R1.25 sample a sharp increase is observed when approaching the notch root. For all cases presented here the equivalent stress is lower than the net stress but those two values almost equalise towards the notch root for both sample designs; this is more distinct for the lower temperature case ($T=900\text{ }^\circ\text{C}$).

From the results of Figure 6.12 it can be summarised that:

- High triaxial stress-states tend to concentrate high stresses at the notch root, as seen experimentally in Chapter 5 (see Figure 5.8(i) compared to Figure 5.7(i));
- As temperature increases, stresses become higher in the centre for both sample geometries; this is also in line with the experimental profiles of Figures 5.7(i) and 5.7(ii) for blunt notch, DN_R5, and Figures 5.8(i) and 5.8(ii) for sharp notch, DN_R1.25;

This correlation of σ_e and σ_1 with regions of observed damage means it can be postulated they are the key parameters in damage initiation for the particular FCS. In particular, the stress profiles as a function of geometry are in very good agreement with the experimental observations. On the contrary, the effect of the hydrostatic stress on the damage features observed in Figures 5.7 and 5.8 does not seem to be straightforward. The stress distributions for σ_H and σ_1 follow very similar paths but, as the hydrostatic stress is always lower than the net stress it cannot be the key parameter that influences damage initiation. It is possible that the hydrostatic stress contributes to void growth rather than void initiation.

6.2.2 TEMPERATURE AND STRAIN RATE EFFECTS ON STRESS-STATE VARIATIONS

The effects of strain rate and temperature were also examined to provide a clearer picture. To this end, FE calculations were performed on a range of strain rates, $\dot{\epsilon}=0.01\text{-}10\text{ s}^{-1}$, for sample DN_R5, for two temperatures, $T=900\text{ }^\circ\text{C}$ and $T=1200\text{ }^\circ\text{C}$, as in the previous sections.

In Figure 6.13a the distribution of the ratio of maximum principal stress to equivalent stress, σ_1/σ_e , is plotted versus the normalised distance from the inner centre of the notch plane for a

range of strain rates ($\dot{\epsilon}=0.01\text{-}10\text{ s}^{-1}$). The values of σ_1/σ_e exhibit the sharpest decrease, from the notch centre to the root, for the lowest strain rate, namely $\dot{\epsilon}=0.01\text{ s}^{-1}$. As the strain rate increases the decrease in σ_1/σ_e becomes less sharp. To compare, the percentage decrease in the values of σ_1/σ_e for $\dot{\epsilon}=0.01\text{ s}^{-1}$ and $\dot{\epsilon}=10\text{ s}^{-1}$ is 21% and 13.7%, respectively. From this it would be expected that as the deformation rate increases, damage would increase its bias towards the notch root and would also become more widespread across the notch plane.

In Figure 6.13b triaxiality distribution, σ_H/σ_e , is plotted against normalised distance from the centre, for the same set of conditions as in Figure 6.13a and the same range of strain rates. In this case, although the trend of decreasing steepness in the distribution of triaxiality as the strain rate increases is still the same, the reduction from the lowest strain rate when compared to the highest is more significant than in Figure 6.13a. Here, the percentage reduction in triaxiality is 43.3% for the lowest strain rate ($\dot{\epsilon}=0.01\text{ s}^{-1}$), as compared to a 30.2% reduction for the highest strain rate ($\dot{\epsilon}=10\text{ s}^{-1}$).

In Figure 6.14, the percentage reductions of σ_1/σ_e and σ_H/σ_e from the notch centre to the notch root are plotted as a function of the range of strain rates examined for two temperatures, $T=900\text{ }^{\circ}\text{C}$ and $T=1200\text{ }^{\circ}\text{C}$. What becomes obvious from the data is that there is a gradual decrease in the percentage reduction as the strain rate increases for both temperatures, although for both stress ratios this behaviour is more pronounced for the higher temperature, $T=1200\text{ }^{\circ}\text{C}$. Another important trend observed from the data of Figure 6.14 is that, as the strain rate increases the percentage reductions for the two temperatures almost coincide. It is therefore evident that deformation at a low temperature or a high strain rate can have a similar effect in terms of damage initiation sites. In both cases the stresses shift from the centre of the notch towards the notch root, under elastic-plastic conditions. This relates to the stress redistribution that has limited time to act as identified both for conditions of low temperature and high strain rate, and can cause the damage initiation sites to appear near the notch root rather than the centre. Conversely for low strain rates and high temperatures there is more time for the temperature activated plasticity/creep to occur and cause damage to initiate at the centre of the notch.

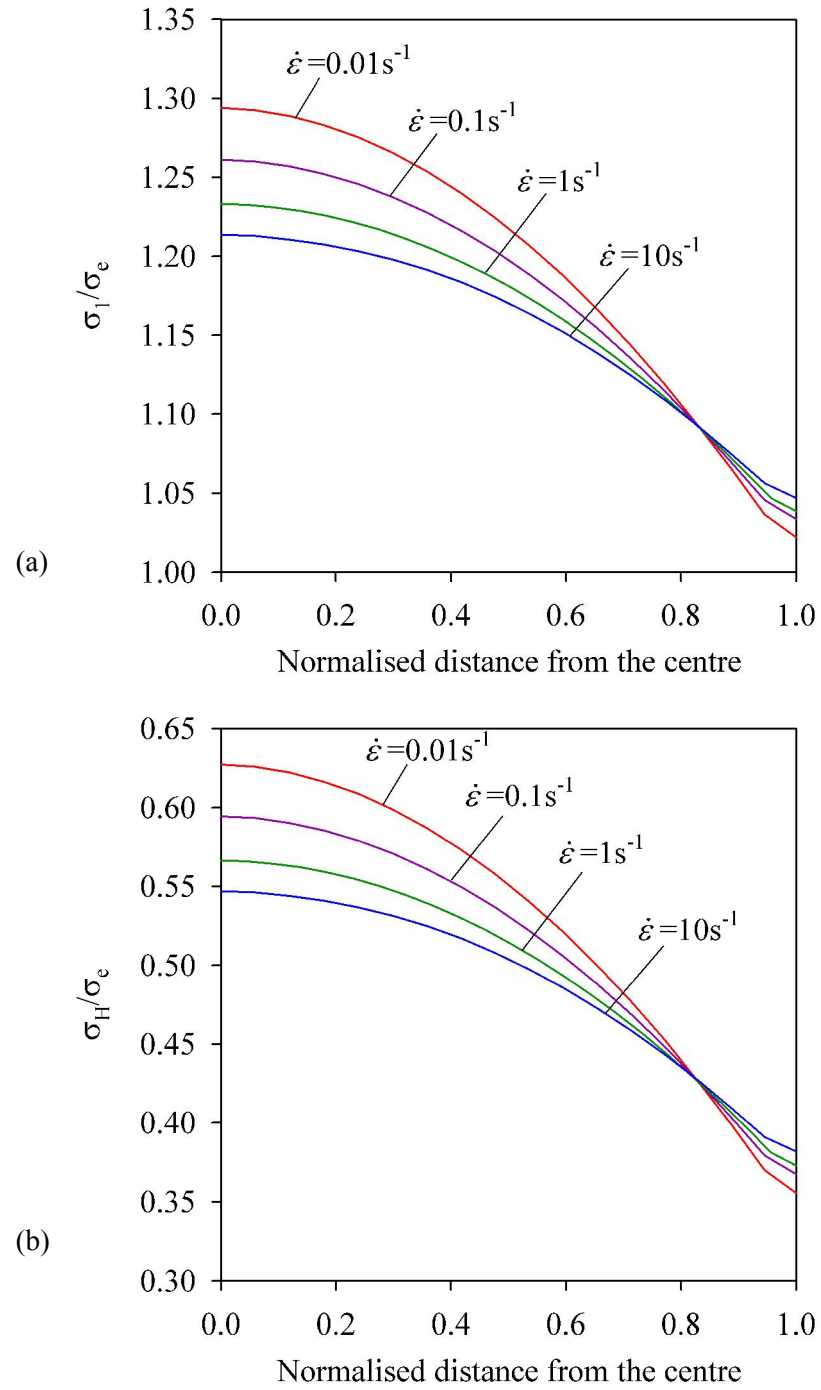


Figure 6.13: Distributions of (a) σ_I/σ_e and (b) σ_H/σ_e across the central notch plane of sample DN_R5. Data were extracted from ABAQUS tensile simulations calculated for a range of strain rates ($\dot{\epsilon} = 0.01\text{--}10\text{ s}^{-1}$) at 900 °C for the case where steady state is reached and true strain equals 0.1.

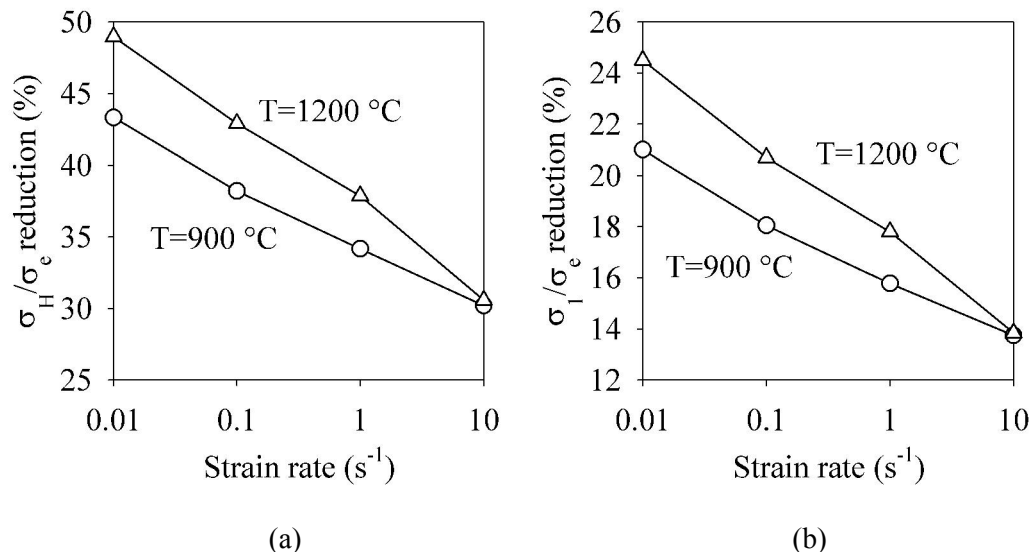


Figure 6.14: Percentage reduction of (a) σ_H/σ_e and (b) σ_1/σ_e between the notch centre and the notch root for temperatures $T=900\text{ }^{\circ}\text{C}$ and $T=1200\text{ }^{\circ}\text{C}$ versus strain rate, ranging from $\dot{\epsilon}=0.01\text{--}10\text{ s}^{-1}$ for blunt notch DN_R5.

6.3 CONCLUSIONS

In this chapter the effect of stress-state on damage initiation was investigated for a FCS under hot rolling conditions. Blunt (DN_R5) and sharp (DN_R1.25) double notched bar samples were chosen for this purpose. All FE calculations were performed using a viscoplastic material model describing the behaviour of leaded FCS under hot deformation conditions. The model was calibrated for the leaded FCS used here from experimental uniaxial tension results and the predictions were found to be satisfactory.

The findings of this Chapter are summarised below:

- Sharp geometries lead to inhomogeneous straining of the material in the notch and this in turn leads to more localised damage in areas where the material strains for longer and at higher rates. This is seen for sample DN_R1.25 where the maximum strain calculated for an element at the notch root deforms 3 times more than an element at the notch centre. The strain difference between elements from the notch root and centre for sample DN_R5 is considerably smaller; the notch root element is only 1.2 times more than the element at the centre. Experimentally this is seen in Figure 5.7(i) where the blunt notch exhibits

widespread damage both at the notch centre and root whilst in Figure 5.8(i) the sharp notch micrographs reveal damage heavily localised at the notch root.

- The distributions of σ_e/σ_A show a strong skew towards the notch root for the sharp notch, which translates to more localised damage sites. On the contrary, the distribution of σ_e/σ_A for the blunt notch is almost even across the notch plane; this explains the more extensive damage distribution that was observed experimentally for sample DN_R5 (see Figure 5.7) as opposed to sample DN_R1.25 (see Figure 5.8). Furthermore, according to the stress profiles, mainly for the maximum principal stress, low temperatures and high strain rates favour damage at the notch root.
- As strain rate increases or temperature decreases, the values of σ_1/σ_e and σ_H/σ_e drop at the centre of the notch and increase at the notch root. This redistribution is more significant for high strain rates and low temperatures, meaning that under those conditions damage would be more likely to initiate from the notch root. This was proven experimentally for the case of sample DN_R5 in Figures 5.7(ii) and 5.7(iii).

CHAPTER 7

MODELLING OF DAMAGE EVOLUTION UNDER MULTIAXIAL STRESS-STATES

SUMMARY

In this chapter the development of a continuum damage mechanics (CDM) model for the prediction of multiaxial damage under hot forming conditions is outlined. Initially, a uniaxial damage model is presented and the calibration results are shown. Subsequently, based on the numerical analysis, performed in Chapter 6, of the stress-state dependency of maximum principal stress, hydrostatic stress and triaxiality, a multiaxial damage model was formulated and presented. The calibration of the multiaxial damage model is outlined and the calibration results are presented. Finally, the numerically (FE) produced multiaxial damage profiles are presented and compared with the experimental results from the double notched samples, shown in Chapter 5. Further damage predictions based on the newly formulated multiaxial damage model are presented for a range of temperatures and strain rates. These were extended to one additional double notched sample geometry so that a wider range of stress triaxialities was studied.

7.1 UNIAXIAL DAMAGE MODEL DETERMINATION

7.1.1 VISCOPLASTIC UNIAXIAL DAMAGE MODEL EQUATIONS

The unified viscoplastic equations, originally presented in Chapter 6, were used here but were extended to include damage. These take the following form:

$$\dot{\epsilon}_p = \dot{\epsilon}_0 \left\langle \frac{\frac{\sigma}{1-\omega} - (k+R)}{K} \right\rangle_+^n \quad 7.1$$

$$\dot{\bar{\rho}} = k_1 \left(1 - \sqrt{\bar{\rho}}\right) \dot{\varepsilon}_p - c_1 \bar{\rho}^{c_2} \quad 7.2$$

$$R = B \sqrt{\bar{\rho}} \quad 7.3$$

$$\sigma = E(1 - \omega) (\varepsilon_T - \varepsilon_p) \quad 7.4$$

$$\dot{\omega} = \frac{z_1 \sigma_e}{(1 - \omega)^{z_2}} \left(\dot{\varepsilon}_p \right)^{z_3} \quad 7.5$$

The temperature dependent material constants are determined using the equations below:

$$k = k_0 \exp\left(\frac{Q_k}{\kappa T}\right) \quad 7.6$$

$$K = K_0 \exp\left(\frac{Q_K}{\kappa T}\right) \quad 7.7$$

$$c_1 = c_1 \exp\left(\frac{-Q_{c1}}{\kappa T}\right) \quad 7.8$$

$$k_1 = A_0 \exp\left(\frac{-Q_A}{\kappa T}\right) \quad 7.9$$

$$E = \frac{E_0}{\cosh^2(K_1(T - T_0))} \quad 7.10$$

$$B = \frac{B_0}{\cosh^2(K_1(T - T_0))} \quad 7.11$$

The damage related parameters z_1, z_2, z_3 are also temperature dependent:

$$z_1 = z_{01} \exp\left(\frac{-Q_{z_1}}{\kappa T}\right) \quad 7.12$$

$$z_2 = z_{02} \exp\left(\frac{Q_{z_2}}{\kappa T}\right) \quad 7.13$$

$$z_3 = z_{03} \exp\left(\frac{Q_{z_3}}{\kappa T}\right) \quad 7.14$$

Equations 7.1 and 7.4 were modified from those in Chapter 6 in order to account for damage, hence the reduction in the cross sectional area of the original undamaged material ($\omega = 0$). This modification was done using the basic principles of continuum damage mechanics (Chaboche, 1986), as explained in Chapter 2, where the effective stress of a damaged material can be described as, $\sigma_E = \sigma/(1 - \omega)$. Equation 7.5 which describes the damage rate of a material under uniaxial stress conditions was originally proposed in a recent work of Lin et al. (2013). Damage, ω , takes values in the range $0 \leq \omega \leq 1$ with $\omega = 0$ representing the undamaged, original material and $\omega = 1$ meaning the material has failed. An upper limit on the local damage is set to 0.7. This is in line with the work of Lin et al. (2013) who found that an increase in damage from 0.7 to 1 (material failure), has little effect on the strain increment in stress-strain curves. Restricting the highest value for damage is common practice (Lin et al., 2013; Mohamed et al., 2012) in damage simulations as it significantly reduces the required computational time without sacrificing accuracy.

7.1.2 UNIAXIAL MODEL CALIBRATION

The viscoplastic model was previously calibrated in Chapter 6 without the consideration of damage evolution. Therefore, recalibration was necessary for the set of equations presented here as the addition of the damage equation has an effect on the viscoplastic response that cannot be neglected. In the process of recalibration of the damage model, the temperature dependency of each coefficient was re-evaluated.

Although in the model presented in Chapter 6 there were 8 temperature dependent material constants, namely, $k, n, K, k_1, c_1, c_2, E$ and B , in this calibration the temperature dependency has been removed from constants c_2 and n in an effort to simplify the calibration process. The calibration was performed in MATLAB manually for the temperature range of 900-1200 °C and the strain rate range of 0.01-1 s⁻¹, using uniaxial tension data. Figure 7.1 shows a comparison between the uniaxial experimental data used to calibrate the model and the MATLAB calculated

model predictions. The calibration results show a good fit for this model. In Table 7.1 the temperature independent values of the material constants are given.

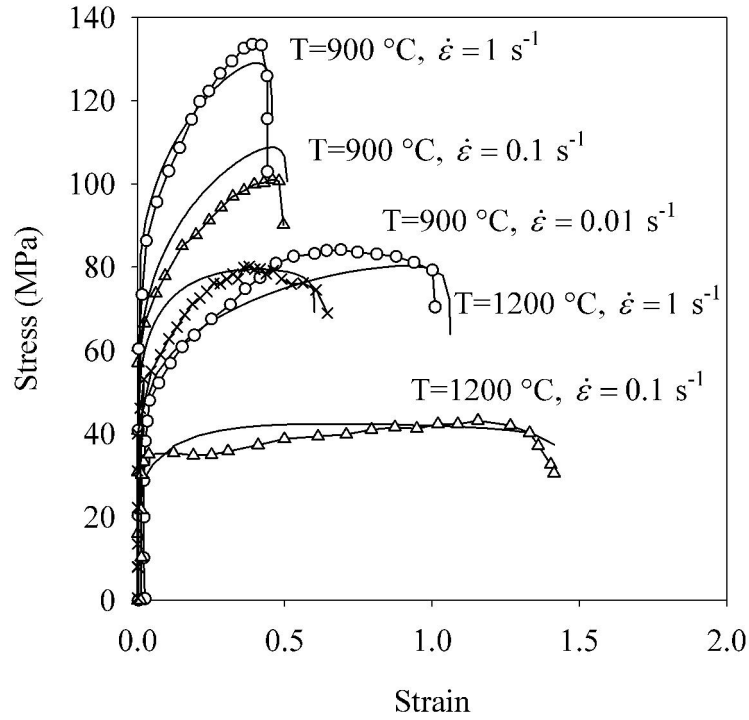


Figure 7.1: Damage model calibration with uniaxial experimental data (symbols) versus model data calculated in MATLAB (lines).

Table 7.1: Temperature independent parameters as calibrated from uniaxial experimental data.

Parameter	Value
n (-)	2.5000
K_0 (MPa)	21.9600
Q_K (J/mol)	4563.8000
k_0 (MPa)	0.0147
Q_k (J/mol)	77065.7900
A_0 (-)	1.1500
Q_A (J/mol)	9522.0200
c_2 (-)	0.6700
c_1 (1/s)	2032.0400
Q_{c_1} (J/mol)	110260.2600
B_0 (MPa)	169.9900

K_1 (1/K)	0.0028
E_0 (MPa)	19900.0000
z_{01} (-)	0.000125
Q_{z_1} (J/mol)	-12563.2800
z_{02} (-)	23.2300
Q_{z_2} (J/mol)	5639.8800
z_3 (-)	0.6013
Q_{z_3} (J/mol)	4662.6500

7.2 MULTIAXIAL DAMAGE MODEL DETERMINATION

7.2.1 MULTIAXIAL DAMAGE EQUATION

As stated earlier, the damage formulation of equation 7.5 applies only to uniaxial damage conditions. A stress-state parameter has to be incorporated in this equation so that it can sufficiently describe the effect of stress-state on damage. In Chapter 6 an extensive investigation was undertaken to understand which parameter affects damage the most in multiaxial conditions. The results of Chapter 6 showed that out of the three parameters investigated (σ_1 , σ_H and σ_e), the maximum principal stress, σ_1 , and equivalent stress, σ_e , gave the best correlation with regions of observed damage. For this reason they have been chosen to formulate the stress-state parameter of the damage equation as follows:

$$\dot{\omega} = \left(\frac{\sigma_1}{\sigma_e} \right)^{\beta \sigma_1} \frac{z_1 \sigma_e}{(1 - \omega)^{z_2}} \left(\dot{\epsilon}_e^p \right)^{z_3} \quad 7.15$$

where, β is a temperature dependent coefficient defined as:

$$\beta = \beta_0 \exp \left(\frac{Q_\beta}{\kappa T} \right) \quad 7.16$$

This stress-state factor is based on a model proposed by Kowaleski et al. (1994) and has also been used in the works of Lin et al. (2005). This simple formulation of the damage equation allows the representation of the stress-state effect on damage. The value of the temperature dependent

coefficient β dictates the magnitude of the dependency between stress-state and strain to failure. Coefficient β takes positive values here and has a reverse linear relationship with the strain to failure. This trend is shown in Figure 7.2 where the strain to failure reduces significantly as the value for coefficient β increases. When $\beta = 0$, equation 7.15 reverts back to equation 7.5, thus, representing the uniaxial damage response.

Apart from predicting strain to failure with greater accuracy for multiaxial damage conditions, the stress-state factor added in equation 7.5 is able to predict the location of damage initiation. When omitting the stress-state coefficient (i.e. when damage is uniaxial) damage initiation always occurred at the notch root. From the experimental investigation of Chapter 5 it is clear that this is not the case. Instead, the damage initiation sites are closely linked to the areas of high maximum principal stress, σ_1 , suggesting that material damage is maximum principal stress controlled. Due to the high dependency of σ_1 in the new damage rate equation (equation 7.15), prediction of damage initiation sites becomes possible. Figure 7.3 shows the FE calculated damage contours corresponding to $\beta = 0$ and $\beta = 0.12$, compared to the experimentally obtained damage initiation sites of sample DN_R5 tested at $T=1200\text{ }^\circ\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$.

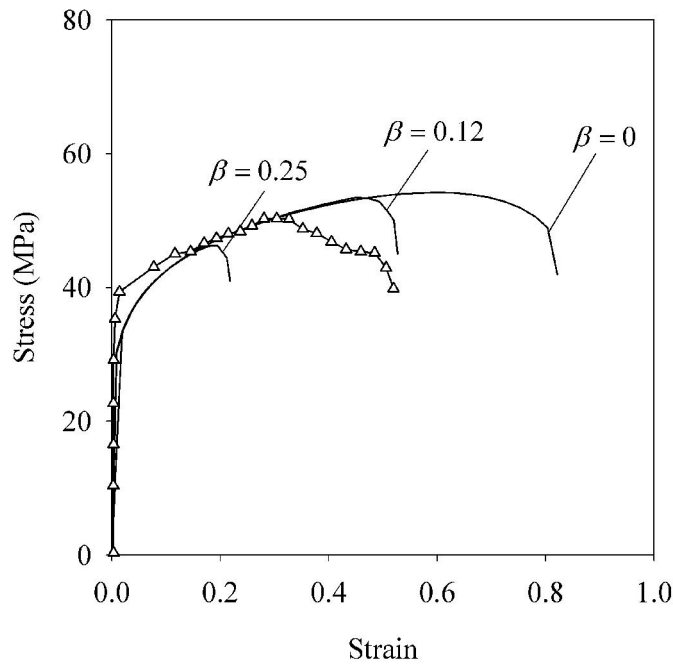


Figure 7.2: Variation of the stress strain curves as a function of stress-state exponent β for sample DN_R5 as calculated from FE analysis at $T=1200\text{ }^\circ\text{C}$ and $\dot{\epsilon} = 0.1\text{ s}^{-1}$.

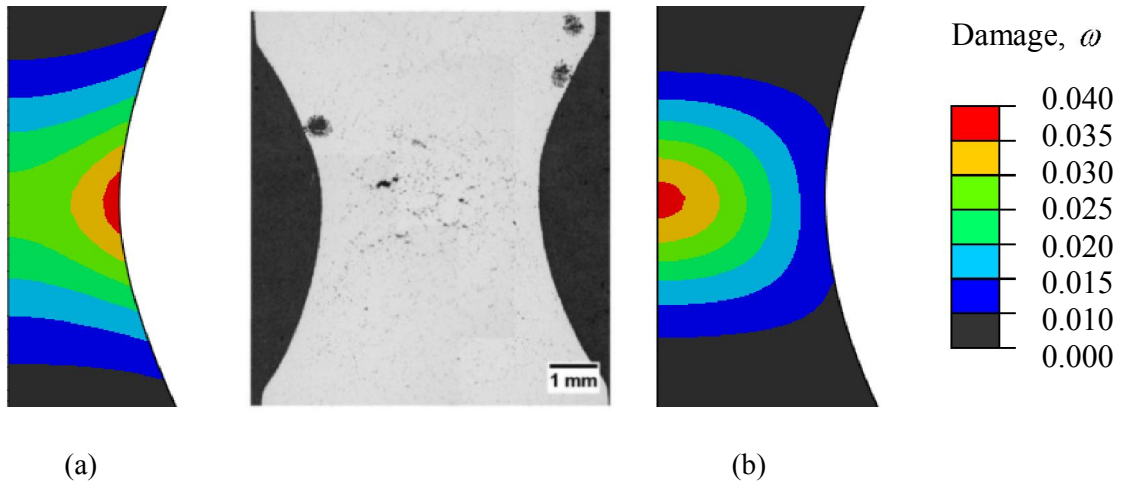


Figure 7.3: FE contours of damage for (a) $\beta = 0$ and (b) $\beta = 0.12$ versus the experimentally obtained damage profile of sample DN_R5 for $T=1200\text{ °C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$.

7.2.2 MULTIAXIAL MODEL CALIBRATION FROM FLOW STRESS DATA

The new multi-axial constitutive damage model was implemented into a user defined VUMAT subroutine that was used in conjunction with the double notched FE models, presented in Chapter 6. The calibration of the multi-axial coefficient β was performed manually in ABAQUS 6.11 and it was found that a value of 0.01 works well for $T=900\text{ °C}$ whilst for $T=1200\text{ °C}$, this value corresponds to 0.12. The results of Figure 7.4 and Figure 7.5 show the FE calculated model predictions (lines) versus the experimental results (symbols) obtained for samples DN_R1.25 and DN_R5, respectively, for three sets of conditions. The model predictions are fairly accurate over this examined range, reproducing the general viscoplastic response and improving on the strain to failure estimations (see Figure 7.3). Although, for sample DN_R5 greater deviations are found in the viscoplastic response for the low temperature examined and strain rates of 0.1 and 1 s^{-1} . Despite these variations, the model has a good overall accuracy especially when taking into account the complexity of the sample design and therefore the resulting stress and strain measurements in this study.

Apart from validating the modelled viscoplastic response with multi-axial stress strain data, it is also important that the model predicts the correct damage profile across the notch plane for each sample. This has to be validated through comparisons of the FE damage contours with the experimentally observed damage profiles, presented for selected conditions in Chapter 5.

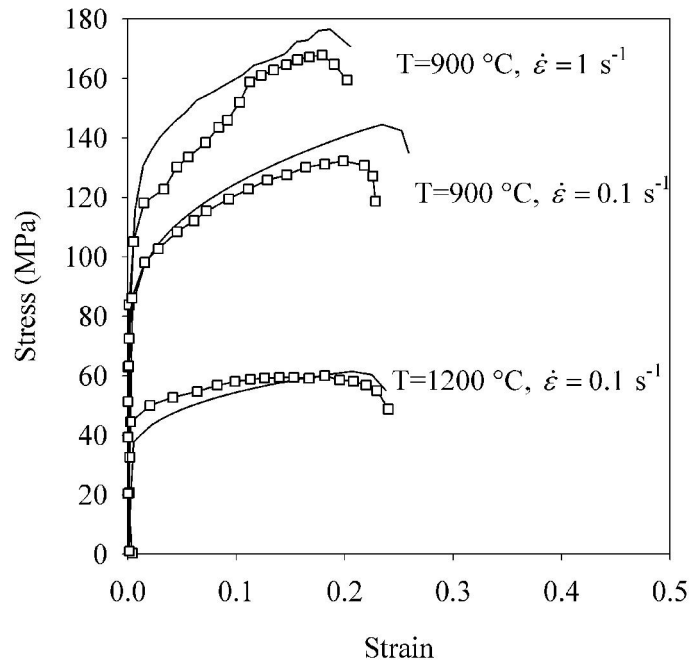


Figure 7.4: FE model calibration for sample DN_R1.25. Lines represent FE calculated values and symbols triaxial tension experimental data.

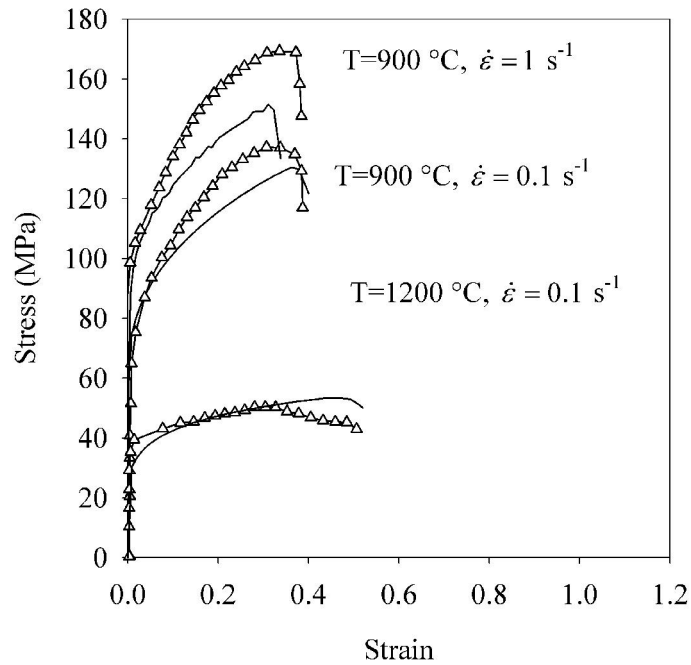


Figure 7.5: FE model calibration for sample DN_R5. Lines represent FE calculated values and symbols triaxial tension experimental data.

7.2.3 MULTIAXIAL MODEL CALIBRATION FROM DAMAGE PROFILES

In this section the model damage predictions are validated against the experimentally observed damage patterns identified in Chapter 5. Figures 7.6-7.9 show the FE contours for the maximum principal stress, σ_1 and damage, ω for two temperatures $T=900\text{ }^{\circ}\text{C}$ and $T=1200\text{ }^{\circ}\text{C}$ at the same strain rate $\dot{\epsilon}=0.1\text{ s}^{-1}$ for samples DN_R5 and DN_R1.25. In particular, Figures 7.6 and 7.7 show the FE contours of sample DN_R5 at temperatures $T=900\text{ }^{\circ}\text{C}$ and $T=1200\text{ }^{\circ}\text{C}$, respectively, at three times; when (a) $t = 0.2 \times t_f$, (b) $t = 0.7 \times t_f$ and finally (c) $t = 0.99 \times t_f$ or simply at failure. Here t_f corresponds to the time of failure in the simulations, when $\omega = 0.7$. In both figures, the FE damage contours of time (c), when the material is really close to failure, are compared with the experimental damage sites for the same conditions. Figures 7.8 and 7.9 compare the same conditions for the sharp notch DN_R1.25.

In Figure 7.6 the distribution of maximum principal stress is shown to be fairly widespread within the notch area. This influences the calculated damage contours which show a similar trend. At time $t = 0.2 \times t_f$ (Figure 7.6a) the damage contours show almost an even distribution across the notch plane, although the damage values are very low with the maximum value at 0.002. At time $t = 0.7 \times t_f$ (Figure 7.6b) there is a significant increase in the damage values, damage progression, and the initiation sites remain widespread across the notch plane. The central area of the notch has slightly higher damage values but this difference is not significant. At time $t = 0.99 \times t_f$ (Figure 7.6c) the damage values have jumped to 0.7, at which point failure occurs. In this stage, the damage initiation sites are concentrated on the central area of the notch and extend significantly across the notch plane. This calculated damage contour is very similar to the experimentally observed, one also shown in Figure 7.6c. In the micrograph, damage initiation sites are widespread over the notch plane with damage being almost equally divided between the centre and the notch root. The calculated damage contour and micrograph are in good agreement since the widespread damage across the notch plane is predicted.

In Figure 7.7 where sample DN_R5 is examined at a higher temperature ($T=1200\text{ }^{\circ}\text{C}$) and the same strain rate, similar distribution in the maximum principal stresses is observed but the damage initiation sites are predicted to be in the centre of the notch only and there is no sign of widespread damage. Specifically at time $t = 0.7 \times t_f$ this becomes obvious. At that time damage values increase enough to show the distinct concentration at the notch centre. Finally, at time $t = 0.99 \times t_f$ damage values reach 0.7 and these are in the centre of the notch, as expected. This

agrees with the experimental observations, shown in the micrograph of Figure 7.7c, where ‘outside-in’ damage is observed.

In Figure 7.8 the maximum principal stress and damage contours are shown for sample DN_R1.25 for $T=900\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$ at three times with respect to the time of fracture; (a) $t = 0.2 \times t_f$, (b) $t = 0.7 \times t_f$ and (c) $t = 0.99 \times t_f$. For case (c) the damage contour, which predicts damage very close to failure condition is compared with the experimentally observed damage initiation sites for this condition. Looking at the maximum principal stress distributions it becomes apparent that the highest values are seen on the whole area of the notch plane and are not concentrated at a specific point such as the notch root. This is distinctively different to the damage contours, compared at the same time. It is evident from the damage contours for all times examined that the notch plane site where damage initiates, does not shift throughout the simulation. This site is the notch root and when compared to the experimentally observed damage initiation sites shown in the micrograph of Figure 7.8c it is clear that the prediction is accurate.

In Figure 7.9 the FE contours for maximum principal stress and damage are shown for sample DN_R1.25 for the same times as in Figure 7.8 with relation to the time of fracture t_f , at a higher temperature $T=1200\text{ }^{\circ}\text{C}$. Here the highest values of maximum principal stress are concentrated near the notch root. When examining the damage contours, for case (a) when damage values are very low, damage initiates very close to the notch root. As the damage values increase and damage progresses, the highest values shift slightly from the notch root towards the centre and this is where damage finally takes place according to the prediction shown in Figure 7.9c. This corresponds well with experimentally observed damage as seen in the enlarged version of the micrograph of Figure 7.9c.

To conclude, the FE damage contours shown in Figures 7.6-7.9 for the two sample geometries DN_R5 and DN_R1.25 demonstrate fairly accurate predictions of the damage initiation sites seen experimentally. More specifically, the model is able to predict the ‘inside-out’ damage characteristics seen in the blunt notch DN_R5 for $T=1200\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$ and the ‘outside-in’ damage observed for DN_R1.25 at the same strain rate at $T=900\text{ }^{\circ}\text{C}$. Additionally, it was established that the maximum principal stress contours do not solely dictate the damage contour shape. Instead the values of the full multiaxial coefficient added to equation 7.5 have to be examined to understand how this influences the predicted damage profiles. Finally, to fully justify the addition of this multiaxial term, the damage profiles have to be compared against the damage predictions of the uniaxial damage equation (equation 7.5).

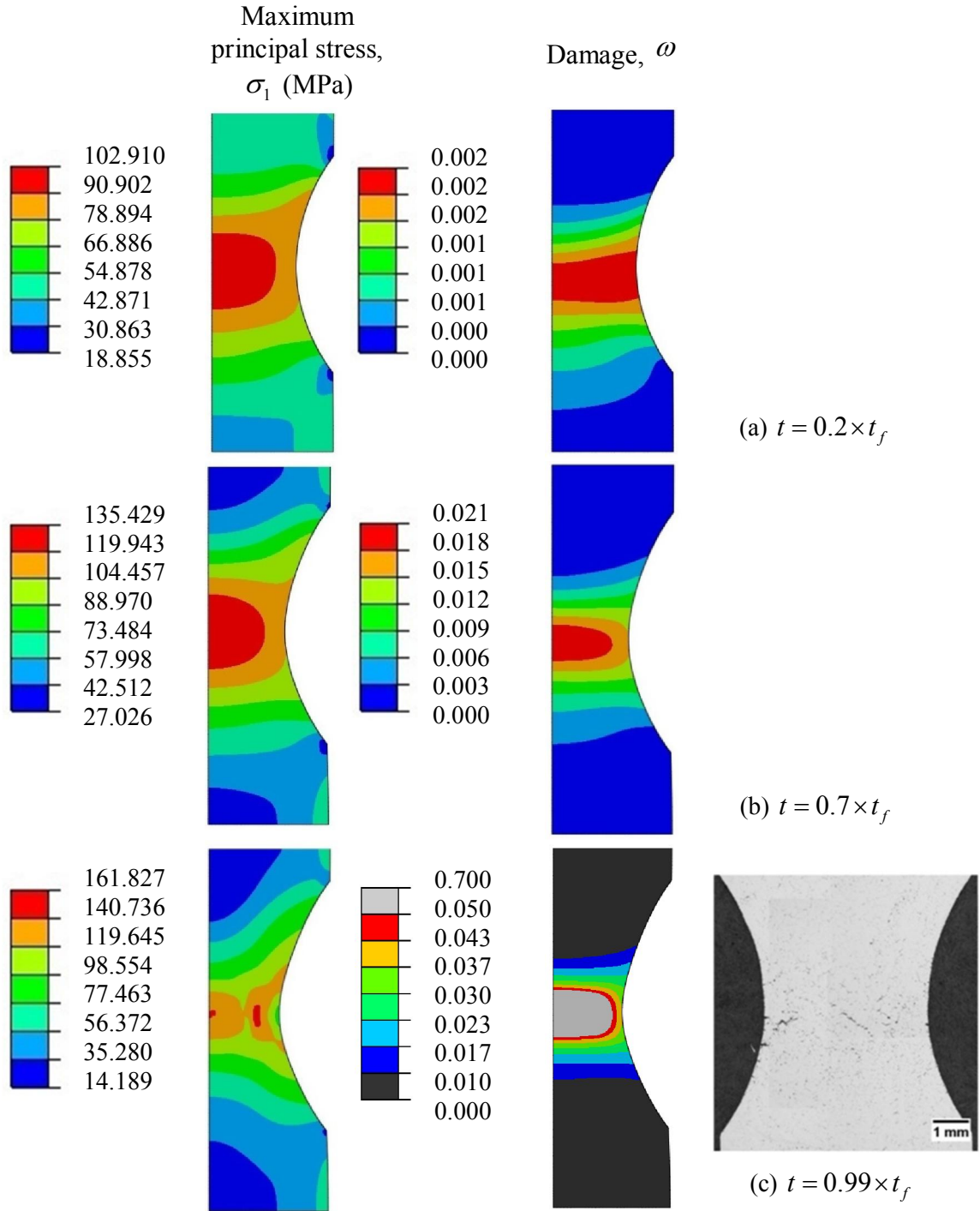


Figure 7.6: Maximum principal stress and damage contours for sample DN_R5 calculated for $T=900\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$ ($\beta=0.01$) taken at times (a) $0.2t_f$, (b) $0.7t_f$ and (c) $0.99t_f$. Damage profile at time (c) is compared with the experimentally observed damage profile.

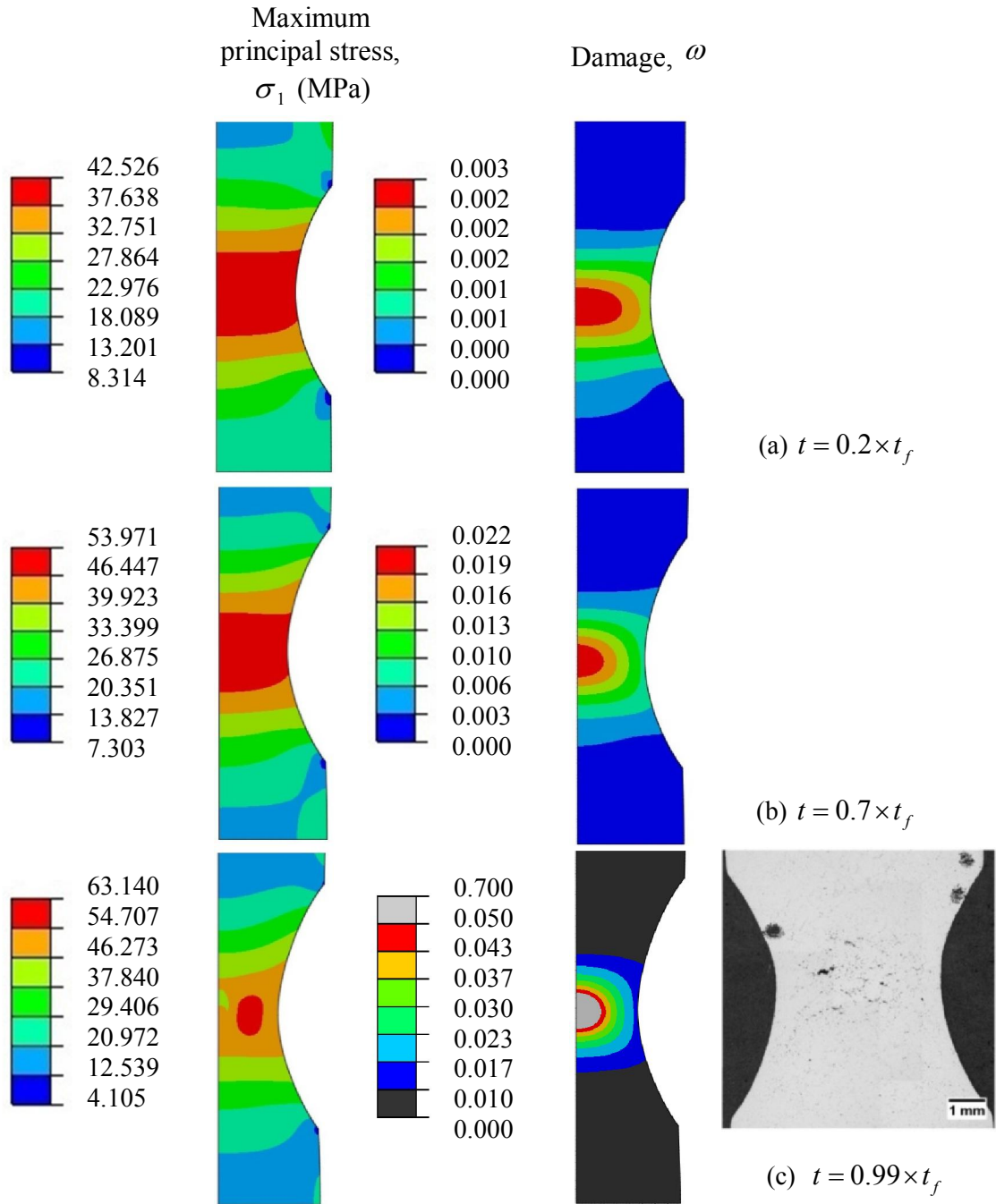


Figure 7.7: Maximum principal stress and damage contours for sample DN_R5 calculated for $T=1200$ °C and $\dot{\epsilon}=0.1$ s $^{-1}$ ($\beta=0.12$) taken at times (a) $0.2t_f$, (b) $0.7t_f$ and (c) $0.99t_f$. Damage profile at time (c) is compared with the experimentally observed damage profile.

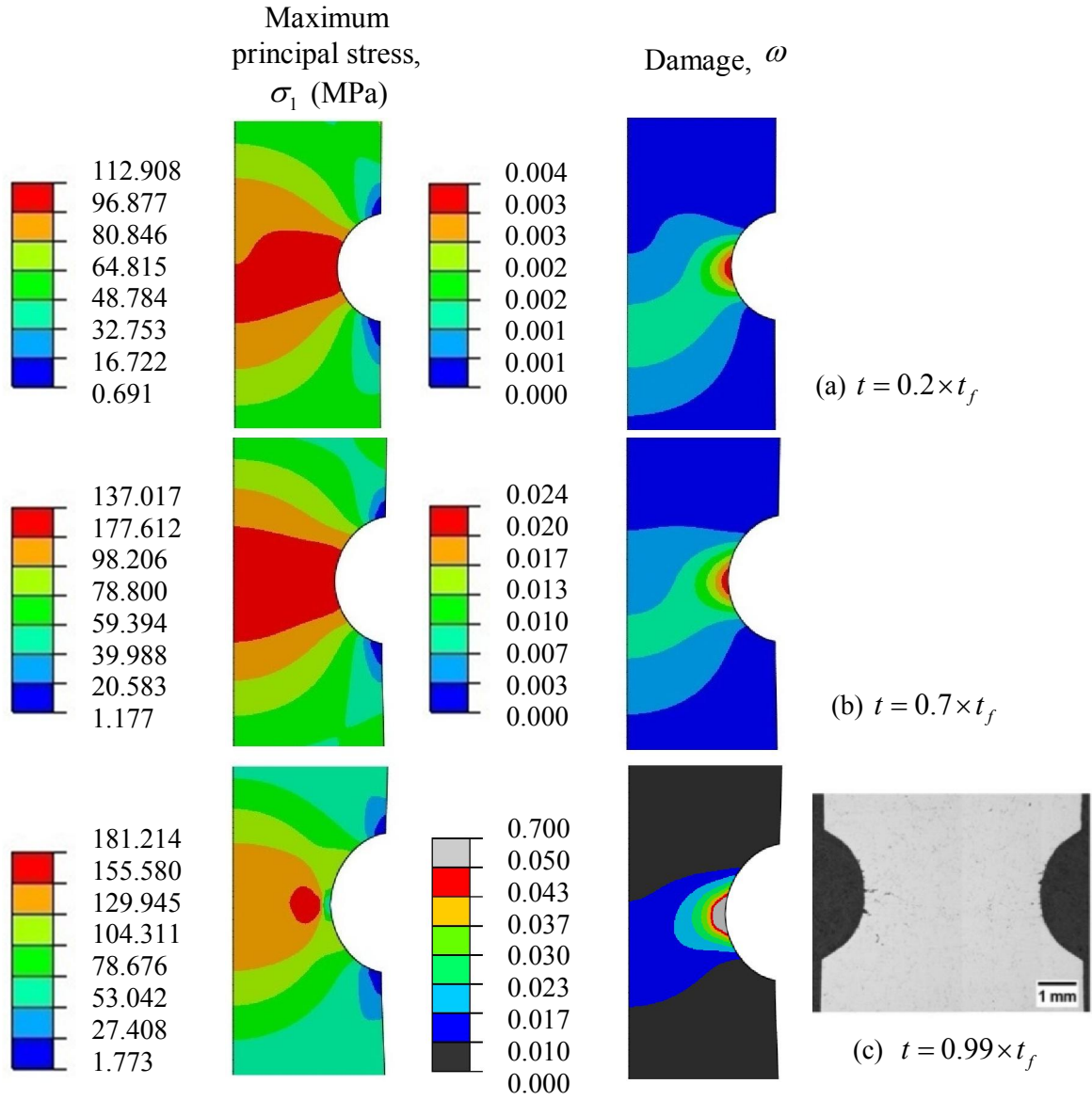


Figure 7.8: Maximum principal stress and damage contours for sample DN_R1.25 calculated for $T=900^\circ\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$ ($\beta=0.01$) taken at times (a) $0.2t_f$, (b) $0.7t_f$ and (c) $0.99t_f$. Damage profile at time (c) is compared with the experimentally observed damage profile.

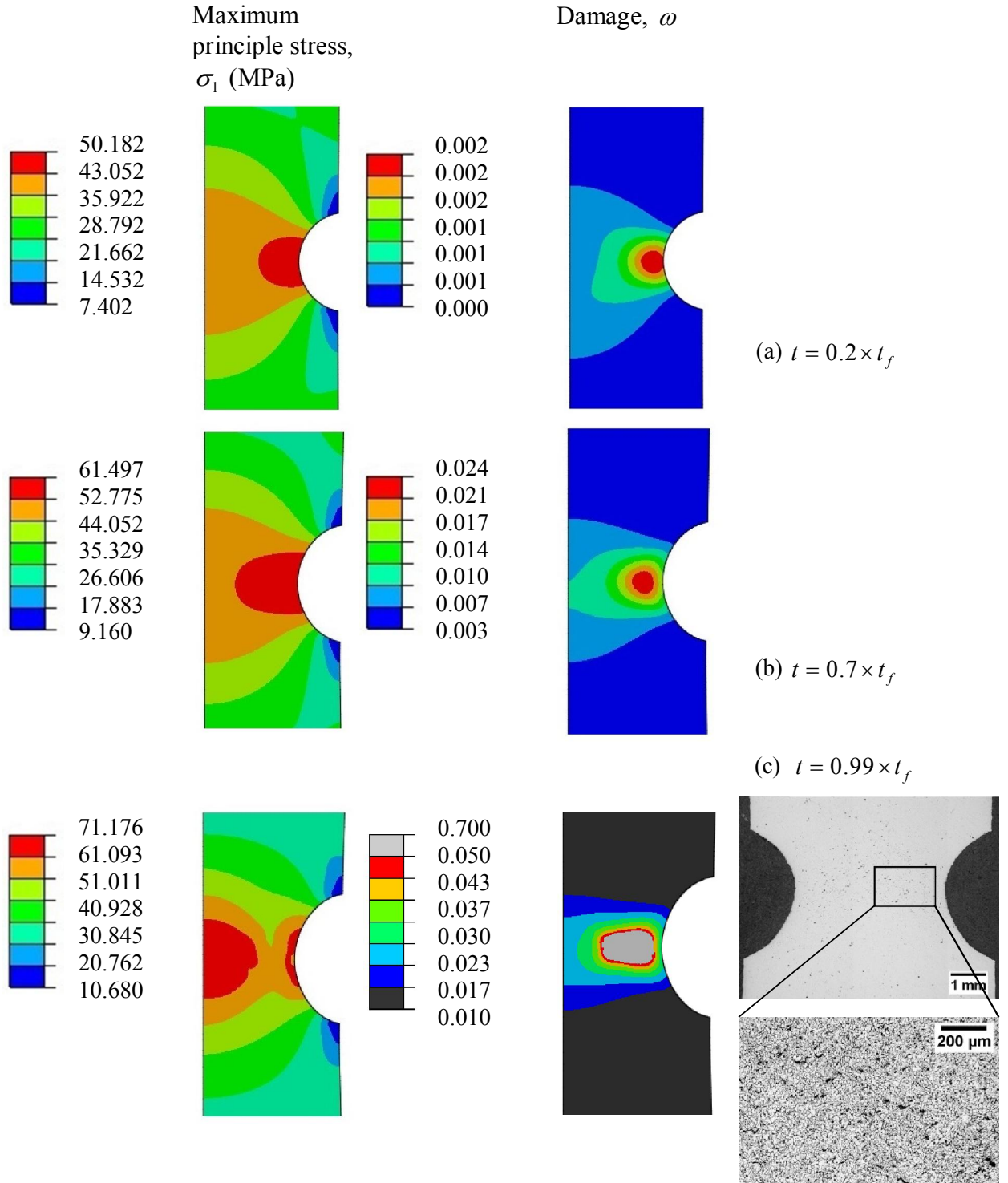


Figure 7.9: Maximum principal stress and damage contours for sample DN_R1.25 calculated for $T=1200\text{ }^{\circ}\text{C}$ and $\dot{\epsilon}=0.1\text{ s}^{-1}$ ($\beta=0.12$) taken at times (a) $0.2t_f$, (b) $0.7t_f$ and (c) $0.99t_f$. Damage profile at time (c) is compared with the experimentally observed damage profile.

Figure 7.10 shows the damage profiles across the normalised distance from the centre of the notch plane for two cases; solid lines represent the damage profiles when the multiaxial damage equation (equation 7.15) was used and the dashed lines represent the damage predictions with the uniaxial damage equation (equation 7.5). The damage profiles are shown for two temperatures at $\dot{\epsilon}=0.1 \text{ s}^{-1}$, the blue lines correspond to $T=1200 \text{ }^{\circ}\text{C}$ and the pink lines to $T=900 \text{ }^{\circ}\text{C}$. Figure 7.10a shows the damage profiles for sample DN_R5 and Figure 7.10b for sample DN_R1.25. In all cases the damage values correspond to time $t = 0.7 \times t_f$.

For sample DN_R5 (Figure 7.10a) it is seen that in the case of the uniaxial profiles, the dashed lines, the damage values are fairly constant across the notch plane and increase slightly at the notch root for both temperatures examined. More specifically, for $T=1200 \text{ }^{\circ}\text{C}$ the damage values increase more steeply at the notch root meaning that in this case, damage initiation sites are expected to be in the notch root mainly. In the case of $T=900 \text{ }^{\circ}\text{C}$ where damage values are more even across the notch plane, it is expected that damage sites would be widespread across the notch plane. These predictions contradict the experimentally observed damage profiles for sample DN_R5 at $T=1200 \text{ }^{\circ}\text{C}$. According to the experimental observations, at this temperature, damage occurs from ‘inside-out’. For $T=900 \text{ }^{\circ}\text{C}$ the uniaxial prediction is acceptable. The damage profiles predicted from the multiaxial damage equation (solid lines) are significantly different to the uniaxial profiles. For $T=1200 \text{ }^{\circ}\text{C}$ the multiaxial profile is much closer to the experimentally observed damage as the highest damage values are in the notch centre and they reduce towards the notch root, meaning that damage occurs from ‘inside-out’. For $T=900 \text{ }^{\circ}\text{C}$ the highest damage values are also at the centre of the notch and reduce towards the notch root but this reduction is very small. It is therefore expected that the damage initiation sites would be widespread across the notch plane with possibly more damage at the notch centre. This prediction is similar to the experimentally observed damage profile for this temperature.

For sample DN_R1.25 (Figure 7.10b) it is obvious that the uniaxial damage profiles are very similar for both temperatures examined. Damage values from the uniaxial damage simulation results are small in the central region of the sample and increase towards the notch root, with the maximum value of damage seen at the notch root. For the multiaxial damage profiles different distributions are seen for the two different temperatures. The multiaxial profile for $T=900 \text{ }^{\circ}\text{C}$ is very similar to the distribution of the uniaxial profiles. For $T=1200 \text{ }^{\circ}\text{C}$ a distinctive difference in the distribution of damage values across the notch plane is observed. In this case, the damage values are low at the central area of the notch and increase close to the notch root but then reduce again at the notch root.

Therefore, damage under this condition is high near the notch root but not at the notch root. When compared with the experimentally observed damage for these conditions, it is obvious that at $T=900\text{ }^{\circ}\text{C}$ (Figure 7.8c) the damage is indeed highest at the notch root with minimal damage in the centre of the notch. Therefore, both the uniaxial and multiaxial profiles correctly predict this behaviour. For $T=1200\text{ }^{\circ}\text{C}$ (see Figure 5.8(ii)) this is different as no damage is observed at the notch root. Instead, damage sites are seen to initiate nearer the notch centre, although this cannot be seen clearly. For this temperature, the damage profile predicted by the multiaxial damage simulation results is more representative of the experimental case.

Figure 7.11 shows the distributions of the multiaxial coefficient, $\left(\frac{\sigma_1}{\sigma_e}\right)^{\beta*\sigma_1}$, added in the uniaxial equation 7.5 as a function of the normalised distance from the centre for the same conditions as examined Figure 7.10. This aims to explain the shapes of the multiaxial damage profiles seen in Figure 7.10. There is a certain correlation between the distribution of the values of the multiaxial coefficient and the resulting multiaxial damage predictions. It is clear that for $T=1200\text{ }^{\circ}\text{C}$ the multiaxial coefficient values are more affected due to the choice of the multiaxial exponent, β , and, therefore, the multiaxial damage predictions are distinctively different to the uniaxial predictions for that temperature. For $T=900\text{ }^{\circ}\text{C}$ where the chosen value for $\beta=0.01$, the values for the multiaxial coefficient are notable smaller and therefore do not affect the uniaxial damage profiles significantly.

From the above comparison of the uniaxial to the multiaxial damage predictions across the notch planes of samples DN_R5 and DN_R1.25, it becomes apparent that the uniaxial equation gives an acceptable prediction only for low temperature i.e. at $T=900\text{ }^{\circ}\text{C}$. In the uniaxial case, damage is always predicted to start at the notch root. This is not accurate since different notch profiles and testing conditions influence the damage initiation sites, as seen experimentally. As a general comment from these results it is observed that the multiaxial damage profiles are able to predict damage better for higher temperatures ($T=1200\text{ }^{\circ}\text{C}$) and for the blunt notch geometry (sample DN_R5).

The proposed multiaxial damage model was validated through both multiaxial stress-strain data and experimentally observed damage profiles, unlike a plethora of stress-state models that are calibrated from uniaxial data (Foster et al., 2009a; Kaye, 2012; Liu et al., 2005). In the next section, damage predictions for one more double notched sample geometry are presented for a range of temperatures and strain rates.

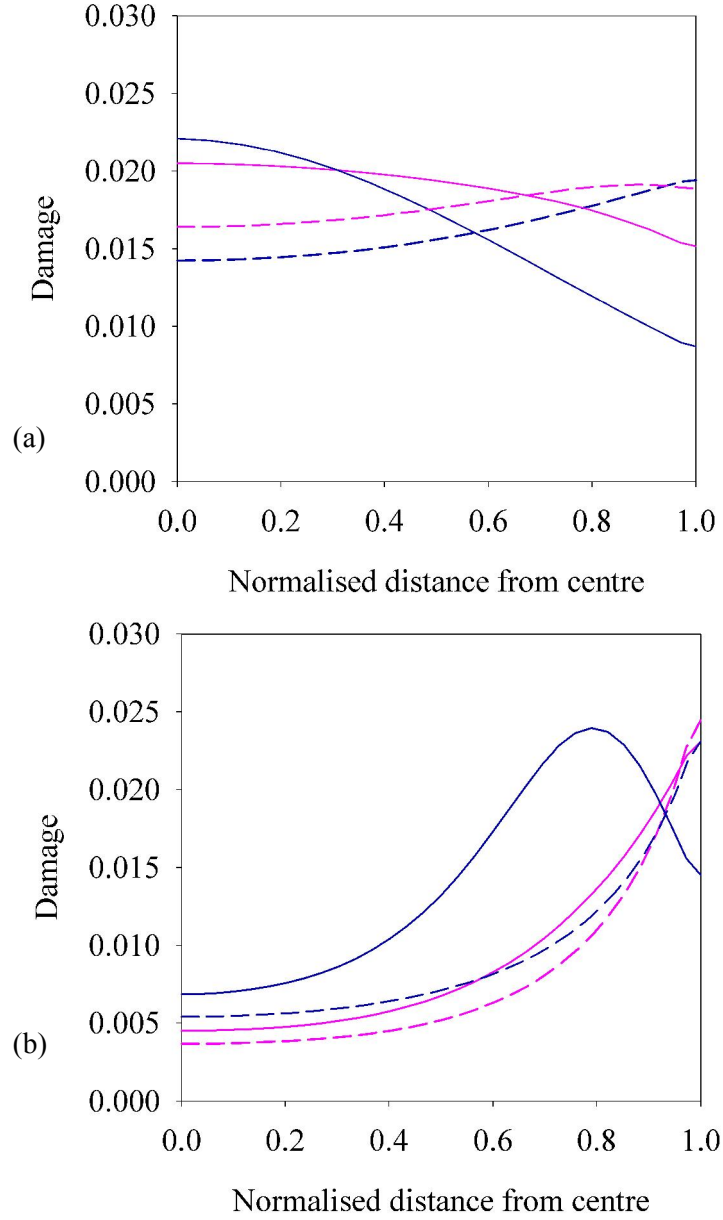


Figure 7.10: Damage profiles with calibrated β values (solid lines) and with $\beta = 0$ (dashed lines) across the notch plane for (a) sample DN_R5 and (b) sample DN_R1.25 at $t = 0.7 \times t_f$. Blue corresponds to simulations at $T=1200$ °C and pink at $T=900$ °C, both at $\dot{\epsilon}=0.1$ s $^{-1}$.

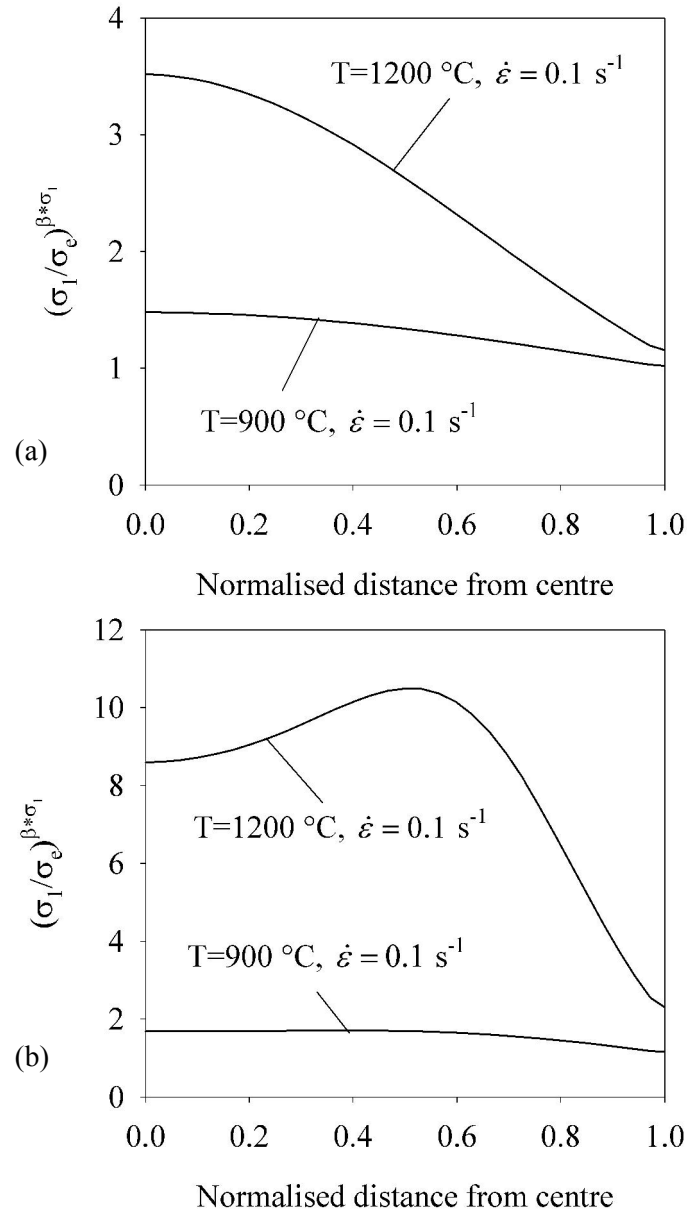


Figure 7.11: Profiles of the multiaxial coefficient in equation 7.15 across the notch plane for sample (a) DN_R5 and (b) DN_R1.25 at $t = 0.7 \times t_f$ for the two conditions examined in Figure 7.10; $T=900\text{ }^{\circ}\text{C}$, $\dot{\epsilon} = 0.1\text{ s}^{-1}$ and $T=1200\text{ }^{\circ}\text{C}$, $\dot{\epsilon} = 0.1\text{ s}^{-1}$.

7.3 MULTIAXIAL MODEL PREDICTIONS

7.3.1 NEW GEOMETRY SIMULATION AND THERMAL RESULTS

In order to study a broader range of triaxialities and the effect of the resulting stress-state on damage characteristics across sample notches, one more sample geometry was simulated. The notch geometry of this sample was selected to produce a multiaxial stress-state with high triaxiality. For this purpose, the double notched sample geometry template (see Figure 3.7b) was utilised and only the notch radius was changed. The new sample geometry is shown in Figure 7.12. Based on this new notch profile, where the notch radius is $R=0.5$ mm, the resulting triaxiality calculated according to the Bridgman equation (see equation 3.3) (Bridgman, 1952) is 1.58. Therefore, when taking into account the simulation results from sample DN_R0.5, the triaxiality range investigated here is extended to 0.55-1.58. Double notched sample geometry was used in this simulation in order to keep the thermal gradients across the notch consistent for all simulated samples. Using a single notch sample geometry in this simulation, although would be equivalent to FE results of a double notched sample, it would also have a symmetric thermal profile unlike that of a double notched sample.

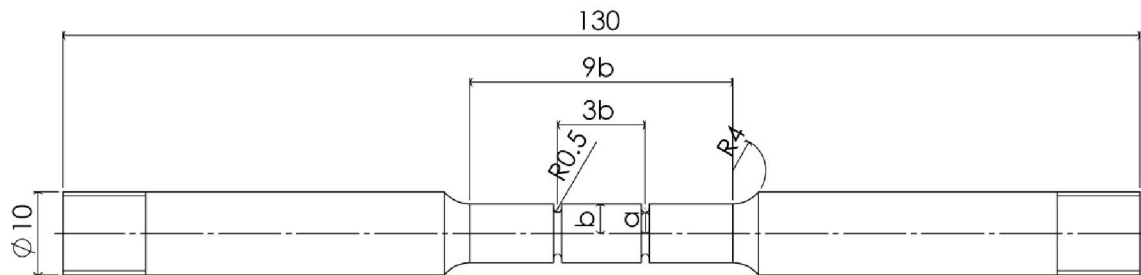


Figure 7.12: Sample geometry of DN_R0.5. The notch spacing is kept constant along with the notch depth ratio, b/a . Notch diameter, $2a=5$ mm, shank diameter, $2b=7.05$ mm.

Sample DN_R0.5 was simulated in ABAQUS 6.11 according to the same methodology used for samples DN_R5 and DN_R1.25, presented in Chapter 6. The mesh used across the notch of this sample is shown in Figure 7.13. For the whole geometry, 5580 nodes and 5377 CAX4R elements were used. The same mesh was also used in the electro-thermal simulations that were carried out to obtain the thermal fields for the sample. For the electro-thermal simulations DCAX4E elements were used instead of CAX4R elements.

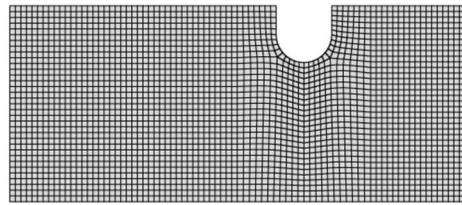


Figure 7.13: Mesh used for the DN_R0.5 simulation.

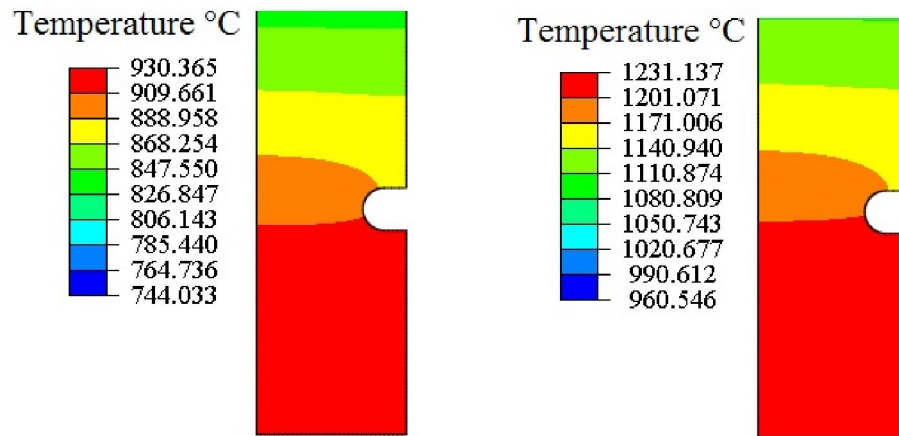


Figure 7.14: FE calculated temperature fields for sample DN_R0.5 at two temperatures (a) T=900 °C and (b) T=1200 °C.

The electro-thermal simulations were carried out according to the method outlined in Chapter 4. Figure 7.14 shows the results of the FE calculated thermal fields for two times in the thermal cycle (see Figure 4.7), namely (a) $t = t_d$ when T=900 °C and (b) $t = t_c$ when T=1200 °C. These FE calculated thermal fields were then used as input for the tensile simulations carried out for this sample geometry. The mechanical, tension simulations were performed for four different temperatures, thus, capturing the damage characteristics across the selected temperature range for hot forming conditions, 900 -1200 °C. In all simulations performed the multiaxial damage model formulation, utilising equation 7.15 was used. In the next section, a comparison between the results of damage profiles across the centre of the notch plane of the three simulated sample geometries is given to understand the effect of multiaxial stress-state on damage.

7.3.2 TRIAXIALITY EFFECT

In this section, the damage profiles across the notch plane are shown for three different sample geometries, namely DN_R5, DN_R1.25 and DN_R0.5 with corresponding triaxialities, according

to the Bridgman solution, of 0.55, 1.02 and 1.58, respectively. This is done to identify the effect that stress-state can have on damage. The results shown in Figure 7.15 were extracted from the centre of the notch plane for two temperatures and one strain rate; (a) $T=900\text{ }^{\circ}\text{C}$, $\dot{\epsilon}=0.1\text{ s}^{-1}$ and (b) $T=1200\text{ }^{\circ}\text{C}$, $\dot{\epsilon}=0.1\text{ s}^{-1}$. For comparison reasons, the results were extracted at a time before damage reached the maximum value of 0.7, which indicates failure, $t = 0.7 \times t_f$.

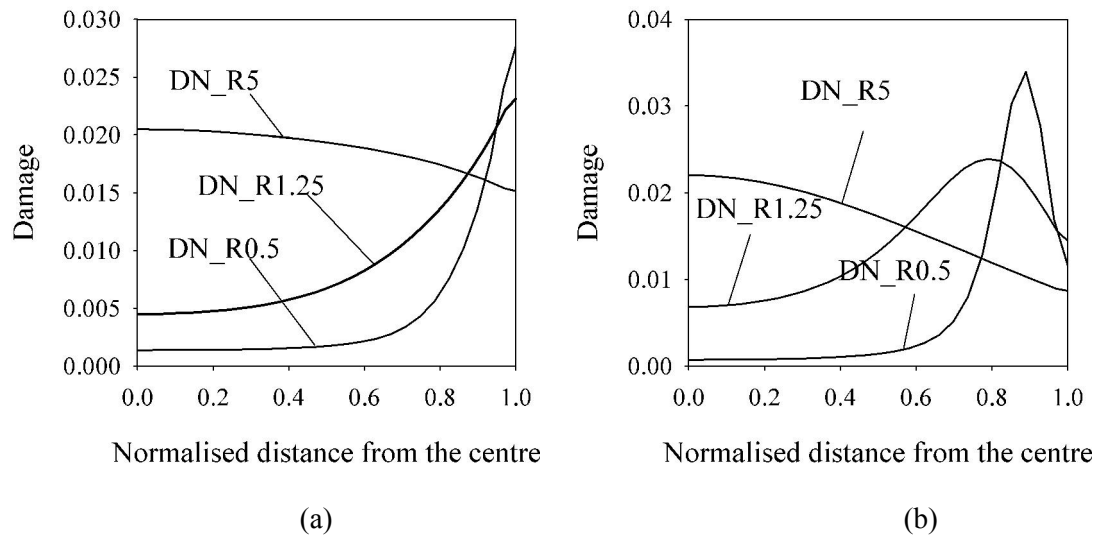


Figure 7.15: Damage profiles across the notch plane for samples DN_R0.5, DN_R1.25 and DN_R5. The damage values were simulated for two conditions; (a) $T=900\text{ }^{\circ}\text{C}$, $\dot{\epsilon}=0.1\text{ s}^{-1}$ and (b) $T=1200\text{ }^{\circ}\text{C}$, $\dot{\epsilon}=0.1\text{ s}^{-1}$ and were extracted at time $t = 0.7 \times t_f$.

For $T=900\text{ }^{\circ}\text{C}$ it is observed that for the high triaxiality samples DN_R0.5 and DN_R1.25, the highest damage values are calculated at the notch root. Instead, damage is highest in the centre for sample DN_R5. In general, the damage values for DN_R5 do not change significantly across the notch plane but for samples DN_R1.25 and DN_R0.5 the damage values at the notch root are significantly higher than at all other locations, the minimum being experienced at the notch centre. This sharp rise in the damage values is more distinctive for the sharpest notch, DN_R0.5.

At a higher temperature, $T=1200\text{ }^{\circ}\text{C}$, there is a shift of the maximum damage values from the notch root to a location close to the notch root for samples DN_R1.25 and DN_R0.5. For the sharpest sample DN_R0.5 the maximum damage location is calculated to be at $r/a = 0.9$ and for DN_R1.25 at $r/a = 0.8$. For sample DN_R5 the highest damage value remains at the notch centre, although, the damage values drop sharply when compared to the damage values of $T=900\text{ }^{\circ}\text{C}$. It is also worth noting that although the damage values for DN_R0.5 are very close to zero at

the central area of the notch, for sample DN_R1.25 the damage values at the corresponding area have increased slightly when compared with the equivalent values for $T=900\text{ }^{\circ}\text{C}$.

As a general remark, the sharp notch DN_R0.5 seems to not be affected by temperature changes as much as the other two samples and also to restrict fracture at the notch root. Additionally, simulation results for sample DN_R0.5 indicate that very limited damage takes place in the notch when excluding the damage at the notch root. This suggests that damage develops very fast and therefore the material has very limited ductility. As triaxiality reduces, damage across the notch plane becomes more widespread with sample DN_R5 at $T=900\text{ }^{\circ}\text{C}$ being an excellent example where damage is almost uniform across the notch plane.

In the next section the effect of temperature and strain rate on the damage profiles of the three simulated samples will be examined more closely to identify at which temperature and strain rate damage sites shift from the notch centre to the notch root or vice versa.

7.3.3 EFFECT OF TEMPERATURE AND STRAIN RATE

In this section the effect of temperature and strain rate on damage values across the centre of the notch plane is investigated for samples DN_R0.5, DN_R1.25 and DN_R5. For each sample investigated here, 12 simulations were performed to cover all possible condition combinations for the temperature range $900\text{--}1200\text{ }^{\circ}\text{C}$ (moving at intervals of $100\text{ }^{\circ}\text{C}$) and the strain rate range of $0.01\text{--}1\text{ s}^{-1}$. Performing these simulations, resulted in a map of all the possible damage patterns for the temperature and strain rate range of hot forming conditions. In Figure 7.16-7.18 the FE damage results for samples DN_R0.5, DN_R1.25 and DN_R5, across the centre of each notch plane, are shown for temperatures $900, 1000, 1100$ and $1200\text{ }^{\circ}\text{C}$ at $\dot{\epsilon}=0.01$ (blue), 0.1 (green) and 1 s^{-1} (red). All data here were extracted at time $t = 0.7 \times t_f$ of the simulation. For ease of comparison the FE calculated damage values and FE calculated triaxiality values for the notch centre and root have been tabulated for all three samples and temperatures for two strain rates ($\dot{\epsilon}=0.1, 1\text{ s}^{-1}$) (see appendix).

Looking at the results for sample DN_R0.5 shown in Figure 7.16 for $T=900\text{ }^{\circ}\text{C}$ it is observed that there are no significant differences between the damage profiles for the three different strain rates. In all cases, damage is low at the notch centre and increase sharply after location $r/a=0.93$, reaching a maximum at the notch root ($r/a=1$). When increasing the temperature slightly to $T=1000\text{ }^{\circ}\text{C}$, very similar damage characteristics are observed; damage values are minimal in the

notch centre and remain minimal until $r/a=0.86$, at which point damage starts to increase sharply. At this temperature, the damage profiles for $\dot{\epsilon}=0.01 \text{ s}^{-1}$ and 0.1 s^{-1} are almost identical whilst for $\dot{\epsilon}=1 \text{ s}^{-1}$, although the differences are minor the peak damage value shifts slightly from the notch root. Increasing the simulated temperature by $100 \text{ }^{\circ}\text{C}$, the shift in peak damage values becomes more obvious. At $T=1100 \text{ }^{\circ}\text{C}$ peak damage values are predicted to develop around the range of $r/a=0.89-0.93$. This shift in maximum damage values is stronger for the higher strain rates, $\dot{\epsilon}=0.1$ and 1 s^{-1} for which peak damage occurs at location $r/a=0.89$ whilst for the lowest strain rate the equivalent location is at $r/a=0.93$. As in previous cases, the damage values at the notch centre are minimal, although in this case, the damage values are lower at the centre for higher strain rates and as the strain rate lowers, the damage values at the centre of the notch increase slightly. This contributes to the effect that as strain rate increases the rise in damage values at the notch root becomes sharper. For example, for $\dot{\epsilon}=0.01 \text{ s}^{-1}$ damage is more widespread in comparison with the other two examined strain rates, although, this difference is not significant as the difference between the maximum and average damage values for $\dot{\epsilon}=0.01 \text{ s}^{-1}$, 0.020, is only marginally lower than the equivalent values for $\dot{\epsilon}=0.1 \text{ s}^{-1}$, 0.022, and $\dot{\epsilon}=1 \text{ s}^{-1}$, 0.023. Finally, at $T=1200 \text{ }^{\circ}\text{C}$ the peak damage values for all strain rates investigated have shifted away from the notch root. At this temperature, the peak damage values are simulated to be around $r/a=0.89-0.9$. Additionally, at $T=1200 \text{ }^{\circ}\text{C}$ it becomes more obvious that at lower strain rates the increase of damage values from notch centre to notch root is significantly smoother; the difference in maximum to average damage value for strain rates 0.01 s^{-1} , 0.1 s^{-1} and 1 s^{-1} is 0.023, 0.025 and 0.032, respectively. Therefore, for high temperatures and low strain rates more damage is expected to develop around the notch root area than at high strain rates where damage is expected to develop only at the notch root.

In Figure 7.17 the damage results for sample DN_R1.25 are shown. Following a similar damage pattern to that of sample DN_R0.5, at $T=900 \text{ }^{\circ}\text{C}$ the peak damage values are calculated to be at the notch root. Additionally, damage values at the notch centre, are slightly lower for the highest strain rate ($\dot{\epsilon}=1 \text{ s}^{-1}$) indicating that under these conditions (low temperatures and high strain rates) at that location, damage would be minimal. As temperature increases to $T=1000 \text{ }^{\circ}\text{C}$, the increase in damage values becomes less steep for strain rates 0.01 s^{-1} and 0.1 s^{-1} whilst for $\dot{\epsilon}=1 \text{ s}^{-1}$ the rise in damage values is steeper than in the other two strain rates but still less pronounced compared to the $T=900 \text{ }^{\circ}\text{C}$ response. Additionally, at $\dot{\epsilon}=1 \text{ s}^{-1}$ the peak damage shifts slightly beyond the notch root ($r/a=0.97$). At $T=1100 \text{ }^{\circ}\text{C}$ the peak damage values have shifted location significantly, for

all three strain rates, with damage values for $\dot{\epsilon}=1 \text{ s}^{-1}$ to have shifted the most, calculated at location $r/a=0.84$. At the same time, damage values at the centre of the notch have increased for $\dot{\epsilon}=0.01 \text{ s}^{-1}$, thus, making the damage at this rate more widespread across the notch plane; the difference between the peak and average damage value is at 0.009, indicating a shift from the distinctive ‘outside-in’ damage pattern (at $T=900 \text{ }^{\circ}\text{C}$) to a more evenly spread damage profile. The difference between the peak and average damage value for $\dot{\epsilon}=1 \text{ s}^{-1}$ is 0.012. At an even higher temperature, $T=1200 \text{ }^{\circ}\text{C}$, this effect is exaggerated even more with the damage profile for $\dot{\epsilon}=0.01 \text{ s}^{-1}$ to be even more smooth across the notch plane. At $T=1200 \text{ }^{\circ}\text{C}$ the difference between the peak and average damage value is 0.006 for $\dot{\epsilon}=0.01 \text{ s}^{-1}$, 0.012 for $\dot{\epsilon}=0.1 \text{ s}^{-1}$ and 0.017 for $\dot{\epsilon}=1 \text{ s}^{-1}$.

Finally, in Figure 7.18 the damage results for the selected temperature and strain rate range are shown for sample DN_R5. For this sample, where the generated stress triaxiality is lower, damage is highest at the notch centre, unlike the behaviour seen for samples DN_R0.5 and DN_R1.25. Damage simulated at $T=900 \text{ }^{\circ}\text{C}$ is rather widespread across the notch plane for all examined strain rates. For this temperature the difference between the peak and average damage is very small; 0.001 for $\dot{\epsilon}=1 \text{ s}^{-1}$, meaning that widespread damage across the notch would be visible at testing conditions of $T=900 \text{ }^{\circ}\text{C}$ at any strain rate. Increasing the simulation temperature to $1000 \text{ }^{\circ}\text{C}$ a more significant deviation was observed, with damage being significantly higher at the notch root for $\dot{\epsilon}=0.01 \text{ s}^{-1}$. Damage across the notch plane follows a very similar course for $\dot{\epsilon}=0.1$ and 1 s^{-1} . At $T=1100 \text{ }^{\circ}\text{C}$ damage showed a slightly higher reduction from the notch centre to the notch root when compared to the equivalent reduction for $T=1000 \text{ }^{\circ}\text{C}$. The highest strain rate, $\dot{\epsilon}=1 \text{ s}^{-1}$, shows the maximum reduction; difference in peak to average damage is 0.005, whilst the other two strain rates follow similar paths with this difference at 0.003. At $T=1200 \text{ }^{\circ}\text{C}$, a sharp decrease of damage values from centre to notch root is observed for all strain rates. This reduction is more significant at $\dot{\epsilon}=1 \text{ s}^{-1}$ where the difference between peak and average damage is 0.011 whilst the equivalent value for $\dot{\epsilon}=0.01$ and 0.1 s^{-1} is 0.008.

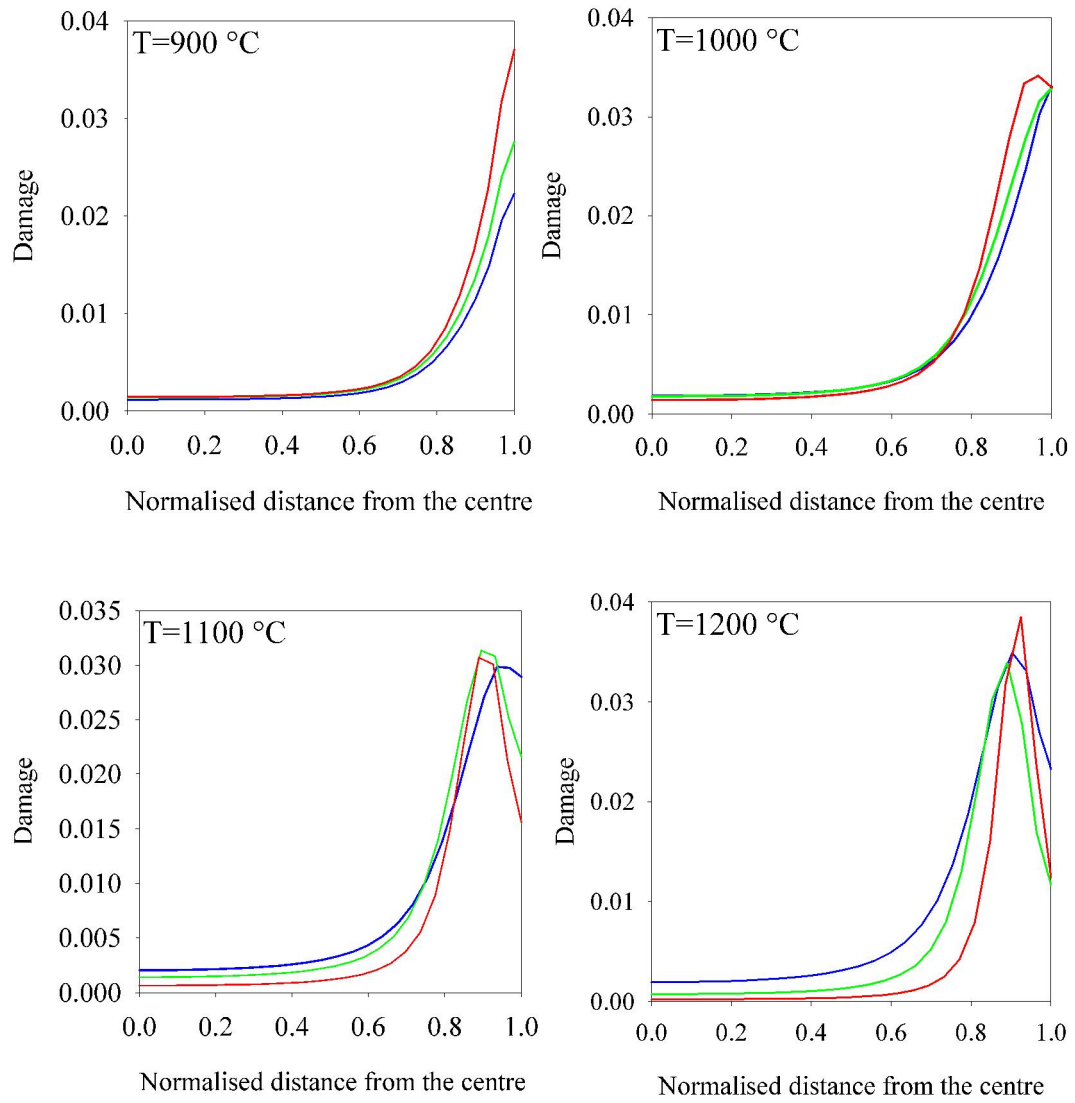


Figure 7.16: Damage profiles across the notch plane of sample DN_R0.5 extracted for four temperatures; 900, 1000, 1100 and 1200 °C and three strain rates; $\dot{\epsilon}=0.01$ (blue), 0.1 (green) and 1 s⁻¹ (red). The data were extracted at approximately $t = 0.7 \times t_f$.

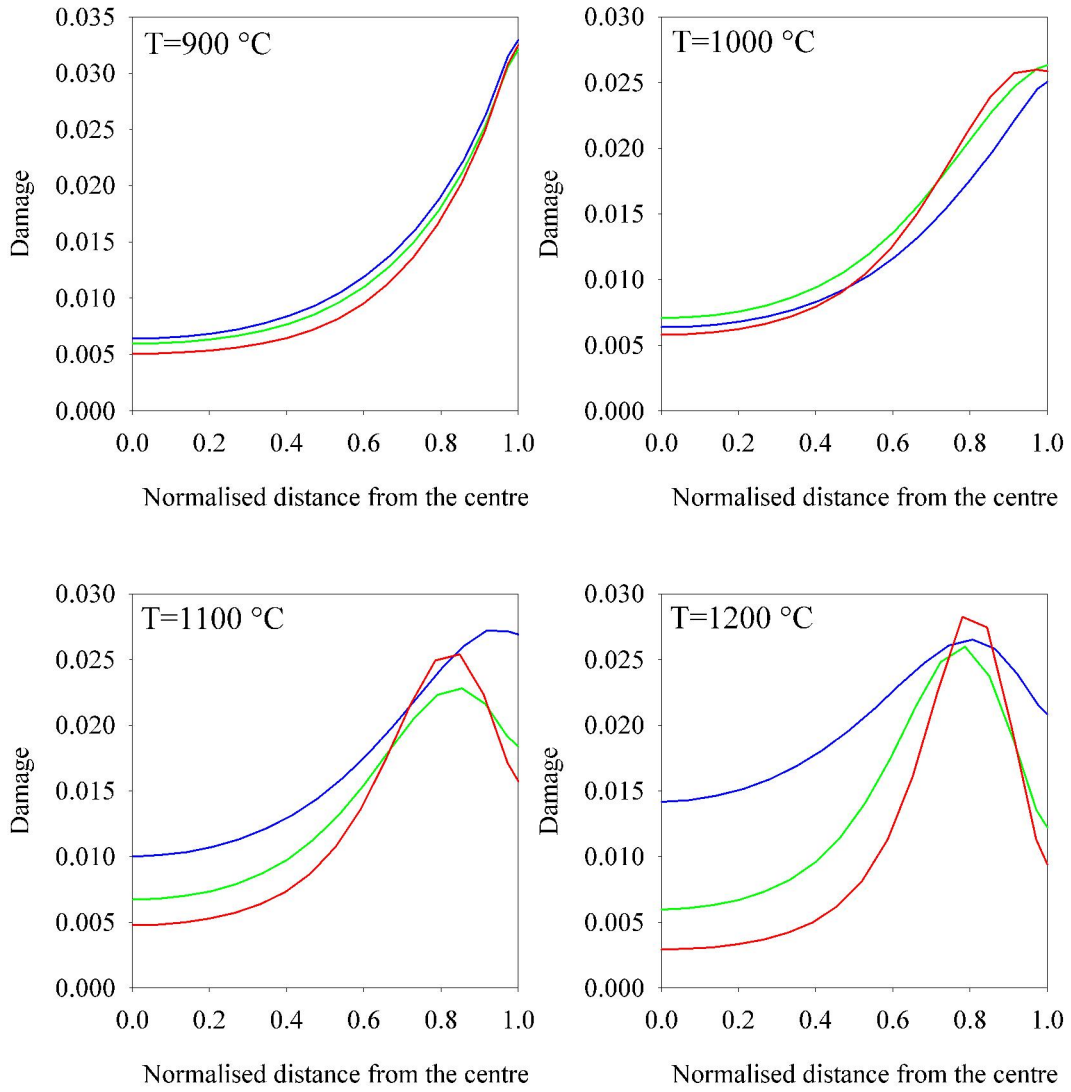


Figure 7.17: Damage profiles across the notch plane of sample DN_R1.25 extracted for four temperatures; 900, 1000, 1100 and 1200 °C and three strain rates; $\dot{\epsilon} = 0.01$ (blue), 0.1 (green) and 1 s⁻¹ (red). The data were extracted at approximately $t = 0.7 \times t_f$.

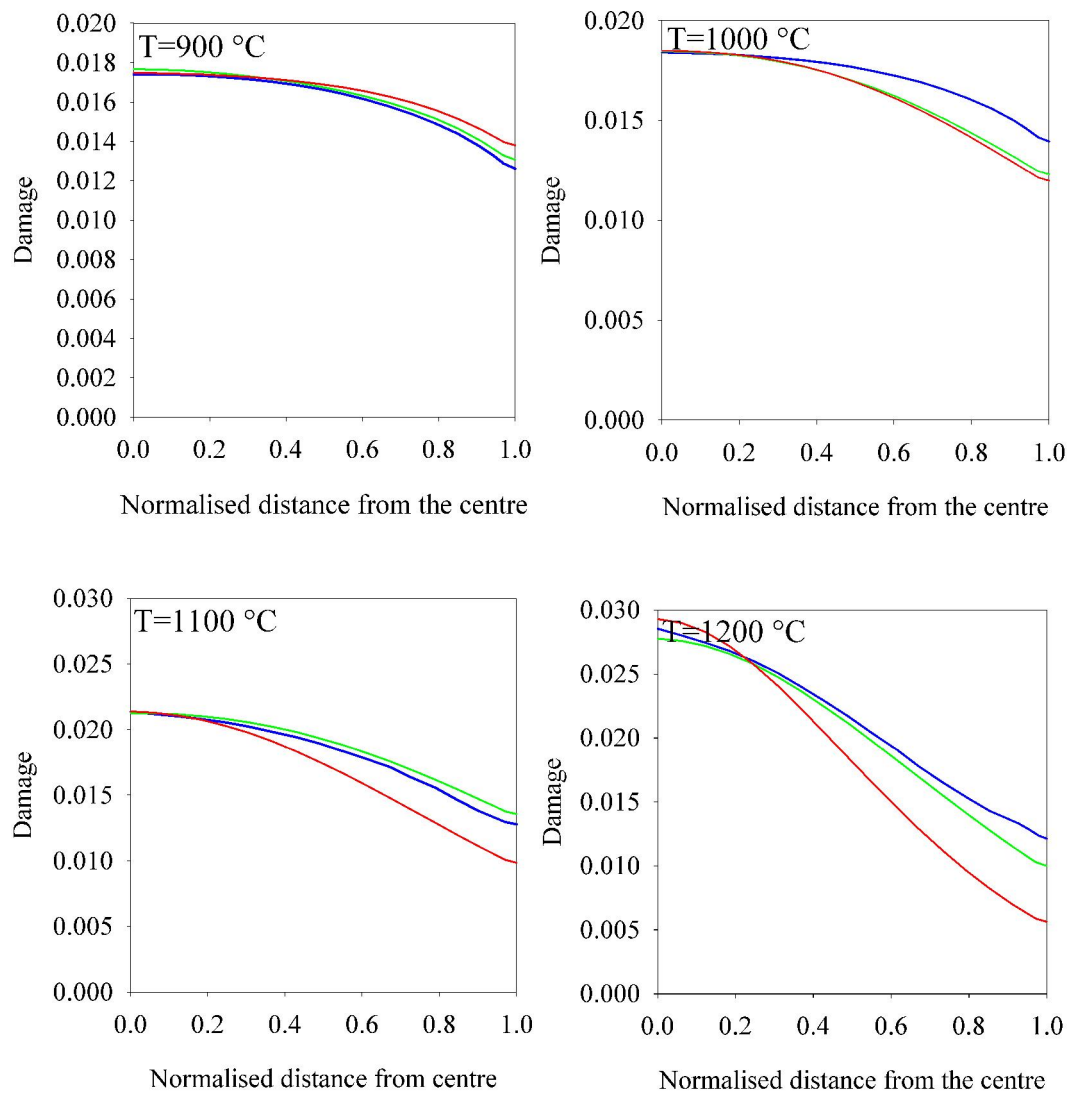


Figure 7.18: Damage profiles across the notch plane of sample DN_R5 extracted for four temperatures; 900, 1000, 1100 and 1200 °C and three strain rates; $\dot{\epsilon}=0.01$ (blue), 0.1 (green) and 1 s⁻¹ (red). The data were extracted at approximately $t = 0.7 \times t_f$.

From the results of Figures 7.16-7.18, the following general conclusions can be drawn:

- At very high triaxialities, sample DN_R0.5, damage always occurs at or very close to the notch root regardless of the test conditions.
- At high triaxialities, sample DN_R1.25, damage is prone to occur at the notch root at low temperatures, 900-1000 °C but the damage bias to the notch root becomes less significant as temperatures increase to 1100 and 1200 °C. Additionally, as temperatures increase beyond 1000 °C, a clear trend develops as low strain rates, $\dot{\epsilon}=0.001$ s⁻¹ reduce the damage bias to the notch root even more. At the highest temperature, T=1200 °C this

strain rate trend becomes most obvious, as at this temperature the lowest strain rate produces significantly even damage values across the notch plane.

- At lower triaxialities, sample DN_R5, a clear shift to the damage sites is observed, with high damage at the notch centre for all test conditions examined here. For low temperatures, 900-1000 °C, damage reduces smoothly from the notch centre to the notch root. As temperature increases to 1100 °C the decrease in damage from the centre to the notch root becomes more obvious and this trend is even more exaggerated for T=1200 °C. At 1200 °C, the strain rates have a greater effect on damage initiation sites with higher strain rates accommodating damage at the notch root.
- For all examined geometries, at high temperatures there is a more distinct difference between the damage profiles observed for different strain rates. In all cases the lowest reduction in damage values is simulated for the lowest strain rate, meaning that at low rates damage sites would be less prone to solely occur at the notch root for sample DN_R0.5, near the notch root for sample DN_R1.25 and at the notch centre for sample DN_R5.

7.4 CONCLUSIONS

The results from the experimental and numerical analysis of the double notched samples detailed in Chapters 5 and 6, respectively, were taken into account to develop a damage model able to predict damage initiation sites developed under multiaxial conditions.

The developed multiaxial model was implemented in ABAQUS via a VUMAT user defined subroutine and the multiaxial damage equation was calibrated using multiaxial stress-strain data from Gleeble high temperature testing. The results were validated through both multiaxial stress-strain data and their experimental damage observations. The results from the multiaxial damage profiles were compared against the uniaxial equivalent values and the multiaxial predictions were proven to be more accurate. Therefore, the developed multiaxial material model was proven able to make fairly accurate predictions of damage initiation sites under conditions of hot forming for complex stress-states.

Following thorough validation, damage predictions were carried out for the full range of hot forming test conditions and an additional sample geometry, thus, expanding the investigated triaxiality range to 0.55-1.58. From these predictions it follows that, at high triaxialities (i.e. sharp notches) damage tends to start at the notch root and testing conditions do not exert significant influence. For lower triaxialities and therefore blunter notch geometries test conditions begin to

significantly change the location of damage initiation sites. It is established that at low temperatures (900-1000 °C) the notch geometry seems to influence the damage locations the most, whilst at high temperatures strain rate has a bigger impact on damage location.

CHAPTER 8

CONCLUSIONS AND FUTURE WORK

SUMMARY

In this work the effect of stress-state on damage under hot rolling conditions has been investigated. To this end, double notched samples were used to identify early stage damage features relating to stress-state and the testing conditions. The experimental work has resulted in identifying two distinct damage behaviours; for a low triaxiality sample (DN_R5) ‘inside-out’ damage was identified for high temperatures and low strain rates and for higher triaxiality samples (DN_R1.25) ‘outside-in’ damage was identified for low temperatures and low strain rates. FE stress-state analysis of these tests has resulted in correlating stress-state parameters with the observed damage. In this work the maximum principal stress and the effective stress gave the best correlation and were therefore used to develop a multiaxial damage model. The model which has a strong stress-state dependency was calibrated with multiaxial stress-strain data, unlike previous models for a similar material which put emphasis on the stress-state dependency to damage (Foster et al., 2009a; Kaye, 2012; Liu and Lin, 2003).

In this chapter the conclusions from this work are summarised and some suggestions for further work are given. The conclusions are split in four sections regarding the literature review findings, the experimental work, the multiaxial constitutive modelling and the predictions made by FE analysis of high temperature tensile tests.

8.1 CONCLUSIONS

8.1.1 LITERATURE REVIEW

The main damage mechanisms that occur during creep, superplasticity and ductility were reviewed in Chapter 2. Different types of damage are controlled by different mechanisms according to temperature, strain rates, microstructure of the material and stress-state. A common way of quantifying stress-state is triaxiality; a factor used frequently to model the effects of stress-state. Many studies have used notched bar testing to explore the effects of different stress-states

on damage initiation. Capturing damage initiation through notched bar tests is often challenging as interrupt tests are necessary to prevent failure. To overcome this, double notched bar testing (primarily seen in creep applications) was used. Double notch bar testing facilitates the identification of early damage development patterns, owing to the symmetry of the sample, as one of the two notches does not break but is instead captured in a pre-failure state. A notable omission from the literature is the stress-state dependency of damage initiation characteristics for the high temperature and high strain rate conditions typical of hot rolling. The work presented in this thesis aims to provide insight into this area.

8.1.2 EXPERIMENTAL WORK

Three sets of experimental programmes were undertaken. Firstly, the thermal gradients in a standard dog-bone sample for Gleeble tensile testing were observed. This was done to identify if these gradients could alter the deformation evolution of the sample as material flow properties are temperature dependant. The identified longitudinal thermal gradients were severe enough so that this could be a possibility. For this reason a user defined subroutine was developed and incorporated into FEM studies of the tensile tests used in validating constitutive models. It was found for that particular sample, specifically at high temperatures, that there is a big variation in the experienced strain between the isothermal and calculated thermal cases. Subsequently, the subroutine has been used in all further FEM studies as a way to accurately represent the performed experiments.

The second set of experimental work using the Gleeble was uniaxial tension testing. This was done as a fundamental study on the viscoplastic response of the material and is based upon previous studies that used near-uniaxial, dog-bone samples (Foster, 2007; Kaye, 2012). These dog-bone samples can induce stress-state influences from the notch radius and, therefore, cannot be categorised as uniaxial. In this study uniaxial samples based on British standards were used to ensure pure uniaxial tension conditions. The uniaxial test results were then used to fit the constitutive model (equations without damage).

Finally, double notch tensile tests were implemented so that an investigation of the role of stress-state was possible, by altering the notch radius. The stress-state had the expected result on ductility; the higher the stress triaxiality, the lower the ductility. Microstructural analysis using an optical microscope was performed on select samples so that the damage initiation sites were identified. To do so the unfailed notch of the double notched samples were sectioned and examined. From these tests two distinct damage evolution profiles were identified. For low

temperatures (900 °C) and low strain rates (0.1 s^{-1}) damage initiated from the ‘outside-in’, especially in the sharp notch DN_R1.25. ‘Inside-out’ damage paths were identified in the blunt notch case (DN_R5) for 0.1 s^{-1} at 1200 °C. The locality of this damage could, thus, be correlated to the local stress-state, as calculated through FEM using the experimentally determined temperature gradients (section 8.1.3)

8.1.3 STRESS-STATE STUDY OF NOTCHED SAMPLES

Following the experimental work that identified specific damage features developing under certain testing conditions and stress-states (induced by the sample geometry) an FE study was performed on the previously tested double notched samples to identify the stress-state parameters that are most strongly linked to the observed damage patterns. For the FE simulations, a viscoplastic material model was used to describe the behaviour of FCS under hot forming conditions, without considering damage. The parameters investigated in this study were the maximum principal stress, σ_1 , the equivalent stress, σ_e , and the hydrostatic stress, σ_H . It was found that the maximum principal and equivalent stress were the most predictive of the observed damage profiles. In particular, the distributions of those values across the minimum area of the notch plane for sample DN_R1.25 showed a strong skew to the notch root whilst for sample DN_R5 the values were rather evenly distributed across the plane. This agreed with the observed experimental results for sample DN_R5, where damage was mostly widespread, and for sample DN_R1.25 that revealed strong features of ‘outside-in’ damage, especially for low temperatures and high strain rates.

8.1.4 MULTIAXIAL DAMAGE MODEL DEVELOPMENT AND PREDICTIONS

Taking into consideration the results from the numerical study of the stress-state in notched samples a new multiaxial damage equation was developed. The multiaxial equation incorporated the stress-state parameters that were identified to control damage, σ_1 and σ_e . The multiaxial damage equation was then added in the now unified multiaxial viscoplastic damage model and was calibrated using stress-strain data from double notched samples rather than uniaxial data as it is common in similar studies. The multiaxial material model was then validated through both stress-strain data and observed damage profiles. It was shown to predict multiaxial damage sites accurately under the conditions of hot forming.

Finally, in order to expand the triaxiality range investigated, a new notched sample geometry, DN_R0.5 (i.e. notch radius = 0.5 mm), was simulated. This expanded the triaxiality range investigated from 0.55-1.02 to 0.55-1.58. The multiaxial damage model was used for damage predictions of the new sample geometry and the previously simulated samples (DN_R1.25 and DN_R5) for a wide range of testing conditions: temperatures from 900-1200 °C and strain rates of 0.001-1 s⁻¹.

These simulations indicated that for sharp notches (high triaxialities) damage is prone to occur at the notch root, ‘outside-in’ damage, regardless of the testing conditions, as seen for sample DN_R0.5. As triaxiality reduces, samples DN_R1.25 and DN_R5, the testing conditions have a more significant effect on damage initiation positions. It was found that for lower temperatures, geometry has an important role so the strain rate at those temperatures does not influence the damage profiles significantly. Therefore, damage for sample DN_R1.25 occurs mainly in the notch root and for sample DN_R5 in the centre. At higher temperatures, T=1100-1200 °C, the applied strain rate starts to have an effect on the damage profiles. At this temperature range, the lower strain rates cause damage to have a more even distribution across the notch plane whilst for higher strain rates damage is skewed to the location where it is more prone to happen for each sample; notch root for samples DN_R0.5 and DN_R1.25, and notch centre for sample DN_R5. It is, therefore, expected that at high temperatures and low strain rates damage will be more evenly spread compared to the case of high temperature and high strain rates where ‘inside-out’ damage is predicted for sample DN_R5 and ‘outside-in’ damage for DN_R1.25.

8.2 SUGGESTIONS FOR FUTURE WORK

The work presented in this thesis goes somewhat to understanding the damage behaviour of free cutting steel under hot rolling conditions by incorporating realistic thermal gradients into FE analysis and by developing a multiaxial model for damage that is validated experimentally. Nonetheless, there are clear ways in which future work could build upon this understanding and further enhance the analysis of damage evolution during manufacturing. For example:

- The electro-thermal FE analysis could be expanded to include radial thermal gradients which have not been included in the analysis performed in this work. Additionally the models can be expanded to include coupled thermal-mechanical analyses to better capture the effects the temperature can have on slower strain rates.

- Expand the stress-states investigated in this study, by varying the geometry of the notches ($a/2R$), to both higher and lower tensile states. This study can also expand to more testing conditions for temperature and strain rate. A combination of tension and compression tests could also be introduced so that a comprehensive understanding of damage evolution under different stress-states is achieved. Additionally, tests at varying strain rates could be conducted to represent the strain rate evolution during hot rolling. The link of the triaxiality, maximum principal, hydrostatic and equivalent stresses with the damage sites via FEM can also be obtained in the same way it was studied here. This can lead to a constitutive model that has a broader physical basis.
- This work could be extended to plain free cutting steel (FCS) so that the influence of the lead (Pb) second phase particles found in leaded free cutting steel (LFCS) is removed and the stress-state effects can be isolated for the MnS inclusions.
- Performing interrupted tests using double notched samples would mean that extra information is gathered with only one test as both notches would be strained at a point before failure and therefore they could both be microscopically examined to better understand damage initiation. Recrystallisation could also be investigated with interrupted tests of double notched samples.
- In both of the aforementioned experimental programmes a bigger statistical analysis has to be introduced. For example, at least 5 tests of each condition should be repeated for every sample geometry. This will give more confidence in the reproducibility of the results. Performing the experimental programmes with a large number of repeats is always helpful for statistical analysis but this has to be balanced with the costs of materials, the cost and time of manufacturing the samples and finally the length of the experiments. In this instance, performing multiple repeats for every condition tested would have been unrealistic due to the limited timeframe and the cost of the material.
- The effect of second phase particles, their distribution and size on damage evolution can also be examined via synchrotron X-ray tomography. Therefore, damage evolution could be quantified accurately. *In-situ* tomography would be particularly interesting, as with *ex-situ* tomography interrupt tests are necessary and due to the fast strain rates investigated for hot deformation conditions it is hard to capture early stage damage (Kaye et al., 2013; Puncreobutr et al., 2012). Although multiple *in-situ* studies to examine the effects of the second phase particles are common, these have concentrated on room or low temperatures and uniaxial samples (A.B. Phillion et al., 2006; Babout et al., 2004; Puncreobutr et al., 2012). It would be very beneficial to investigate the inclusion effects, for a range of high temperatures reaching up to 1200 °C which is now possible due to improved furnace

designs that can reach up to 1500 °C (Buffiere et al., 2010). It would also be useful to perform these tests for a range of notch sample geometries, to achieve different triaxialities, so that the evolution of voids around inclusion is related to the stress-state. Using this information, micro FE models incorporating the inclusion morphology and spacing can be developed to study their effect on double notch tensile testing. An appropriate inclusion size parameter can then be defined and added in the viscoplastic model thus enabling the calibration of the beta value with relation to the second phase particles.

- Other than studying the volume effect of inclusions via destructive tensile testing and *in-situ* X-ray tests, it would also be interesting to examine the effects of stress-state on the surface level. This could be achieved by SEM tests that could also be performed *in-situ* and give insight to the interactions of stress-state with inclusions as inclusion debonding and fracture could be observed in more detail. It would be of interest to prepare plane strain samples with a known/evaluated number and distribution of inclusions in addition to a notch, machined to induce a stress-state dependency. It would then be possible to exclude the effect of inclusions on damage as that would be known for all tested cases.
- The inclusion effect could also be studied numerically in a parametric FE study to systematically control the distribution of the inclusions via designed clusters of particles that vary in size and orientation to examine such aspects with respect to porosity and stress-state. A variety of aspects of the inclusion clusters can be examined in these studies such as number, size and shape of particles, microporosity of the matrix, loading direction and triaxiality. Similar studies have been performed by Horstermeyer et al. (2003) to motivate the development of a phenomenological model for void nucleation of an aluminium alloy.
- Finally, texture studies using Electron-Back-Scattering-Diffraction (EBSD) would be useful to determine how existing and newly developed grains change their orientation during a hot rolling test. Such studies have been performed on a Mg-based AZ61 alloy where the macro and microtexture analysis was performed on specimens that have been hot rolled in two or three passes showing superplastic deformation characteristics with large reductions per pass (del Valle et al., 2003). Additionally, Pinna et al. (Pinna et al., 2015) have used a combination of Crystal Plasticity Finite Element (CPFE) and EBSD analysis performed using microgrids for an aluminium alloy to assess the reliability of CPFE simulations for the deformation of metals in high temperature conditions. A similar approach could be used for the free cutting steel used in this study.

Future studies extending the work presenting here should aim to provide insight to the underlying parameters governing damage in free cutting steel. A more mechanistic understanding of such processes can go a long way to creating a highly predictive set of tools that could help direct experimental programmes and improve industry standards.

List of publications

- Kardoulaki, E., Lin, J., Balint, D., and Farrugia, D. (2014a). Investigation of the effects of thermal gradients present in Gleeble high-temperature tensile tests on the strain state for free cutting steel. *J. Strain Anal. Eng. Des.*, vol.49, 521–532.
- Kardoulaki, E., Lin, J., Balint, D., and Farrugia, D. (2014b). Triaxiality effect on material damage evolution in hot rolling. In *Key Engineering Materials* (pp. 1041–1048).
- Kardoulaki, E., Lin, J., Balint, D., and Farrugia, D. (2015a). A study on the effect of stress state on damage evolution in hot deformation of free cutting steels using double notched bars. *Philos. Mag. A.*, under review
- Kardoulaki, E., Lin, J., Balint, D., and Farrugia, D. (2015b). Modelling the multiaxial damage evolution of free cutting steel under hot forming conditions. *Int. J. Mech. Sci.*, to be submitted

Appendix

Table 1: FE calculated triaxiality and damage values for notch centre and root of sample DN_R5, DN_R1.25 and DN_R0.25 for temperatures 900, 1000, 1100 and 1200 °C at a strain rate of 0.1s^{-1} . Values are extracted for $t = 0.7 \times t_f$

	Temperature (°C)	900		1000		1100		1200	
		Centre	Notch root	Centre	Notch root	Centre	Notch root	Centre	Notch root
DN_R5	Bridgman Triaxiality	0.55		0.55		0.55		0.55	
	FE triaxiality	0.65	0.34	0.59	0.35	0.56	0.35	0.54	0.35
	Damage	0.01	0.01	0.02	0.01	0.00	0.01	0.02	0.00
DN_R1.25	Bridgman Triaxiality	1.02		1.02		1.02		1.02	
	FE triaxiality	0.84	0.45	0.79	0.45	0.76	0.46	0.71	0.49
	Damage	0.00	0.03	0.00	0.02	0.00	0.01	0.00	0.00
DN_R0.25	Bridgman Triaxiality	1.58		1.58		1.58		1.58	
	FE triaxiality	0.78	0.56	0.77	0.55	0.75	0.57	0.74	0.59
	Damage	0.00	0.02	0.00	0.03	0.00	0.02	0.00	0.01

Table 2: FE calculated triaxiality and damage values for notch centre and root of sample DN_R5, DN_R1.25 and DN_R0.25 for temperatures 900, 1000, 1100 and 1200 °C at a strain rate of 1s^{-1} . Values are extracted for $t = 0.7 \times t_f$

	Temperature (°C)	900		1000		1100		1200	
		Centre	Notch root	Centre	Notch root	Centre	Notch root	Centre	Notch root
DN_R5	Bridgman Triaxiality	0.55		0.55		0.55		0.55	
	FE triaxiality	0.61	0.34	0.57	0.35	0.56	0.36	0.56	0.36
	Damage	0.01	0.01	0.01	0.01	0.02	0.00	0.02	0.00
DN_R1.25	Bridgman Triaxiality	1.02		1.02		1.02		1.02	
	FE triaxiality	0.79	0.46	0.73	0.47	0.71	0.48	0.69	0.50
	Damage	0.00	0.03	0.00	0.02	0.00	0.01	0.00	0.00
DN_R0.25	Bridgman Triaxiality	1.58		1.58		1.58		1.58	
	FE triaxiality	0.75	0.57	0.74	0.57	0.73	0.59	0.74	0.61
	Damage	0.00	0.03	0.00	0.03	0.00	0.01	0.00	0.01

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