Understanding Load History Effects for Improved Defect Assessments in Reactor Pressure Vessel Steels

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School of Mechanical, Aerospace and Civil Engineering
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Nomenclature

\( a \)       Crack length
\( a_0 \) Initial crack depth
\( a_n \) Notch depth
\( B \) Specimen thickness
\( E \) Young’s Modulus
\( E' \) Plane strain/stress effective young’s modulus
\( G_f \) Fracture energy
\( G \) Shear Modulus
\( G_c \) Critical energy at fracture
\( I_N \) Integration constant
\( J \) The J integral crack driving force
\( J_c \) J at failure
\( J^{\text{mat}} \) Elastic plastic constraint corrected fracture toughness
$J_e$ Elastic component of $J$

$J_{mat}$ Elastic plastic fracture toughness

$J_p$ Plastic component of $J$

$J_{res}$ Residual crack driving force

$K_0$ Normalising SIF

$K_1$ Initial loading SIF

$K_2$ Warm pre-stress pre-load SIF

$K_{mat}^C$ Constraint corrected fracture toughness

$K_f$ Fracture toughness with warm pre-stress benefit

$K_{LC}$ Linear elastic fracture toughness

$K_{Scmed}$ Median SIF at failure

$K_{mat}$ Fracture toughness

$K_{min}$ Minimum SIF for cleavage failure

$K_{P_j}^P$ Primary SIF calculated from $J$

$K_r$ Proximity to fracture ratio

$K_{S_j}^S$ Secondary SIF calculated from $J$

$K_u$ Difference between load and unload SIF

$L_r$ Proximity to plastic collapse ratio

$L_{rmax}$ FAD cut-off point

$m$ Constraint based toughness exponent

$m_1$ Wallin Weibull shape parameter

$m_b$ Beremin shape function

$n$ Strain hardening exponent

$n_i$ Unit vector

$P$ Load

$Q$ The $Q$ parameter measure of crack tip constraint

$r$ distance from crack tip

$r_c$ Critical distance from crack tip

$r_p$ Plastic zone size

$s_u$ Scaling stress

$s_w$ Weibull stress

$t$ Thickness
<table>
<thead>
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<td>T</td>
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<td>$T_i$</td>
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<td>$\sigma_{ij}$</td>
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</table>
$\sigma_0$ Normalising reference stress

$\bar{\sigma}$ Flow stress

$\sigma_d$ Carbide or carbide/matrix interface strength

$\bar{\sigma}_d$ Weibull fitting parameter to describe probability of stresses greater than $\sigma_d$

$\sigma_f$ Stress at fracture

$\sigma_{UTS}$ Ultimate tensile strength

$\sigma_y$ Yield stress

$\nu$ Poisson's ratio

**Abbreviations List**

<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
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<tbody>
<tr>
<td>A533b</td>
<td>A533b Class 1 Ferritic RPV Steel</td>
</tr>
<tr>
<td>ABWR</td>
<td>Advanced Boiling Water Reactor</td>
</tr>
<tr>
<td>ALARP</td>
<td>As Low as Reasonably Practicable</td>
</tr>
<tr>
<td>BCC</td>
<td>Body-Centred-Cubic</td>
</tr>
<tr>
<td>BCT</td>
<td>Body-Centred-Tetragonal</td>
</tr>
<tr>
<td>BWR</td>
<td>Boiling Water Reactor</td>
</tr>
<tr>
<td>CEGB</td>
<td>Central Electricity Generating Board</td>
</tr>
<tr>
<td>CGOD</td>
<td>Clip Gauge Opening Displacement</td>
</tr>
<tr>
<td>CMOD</td>
<td>Crack Mouth Opening Displacement</td>
</tr>
<tr>
<td>CT</td>
<td>Compact Tension</td>
</tr>
<tr>
<td>DBTT</td>
<td>Ductile to Brittle Transition Temperature</td>
</tr>
<tr>
<td>EDF</td>
<td>Électricité de France</td>
</tr>
<tr>
<td>EPFM</td>
<td>Elastic Plastic Fracture Mechanics</td>
</tr>
<tr>
<td>EPSRC</td>
<td>Engineering and Physical Sciences Research Council</td>
</tr>
<tr>
<td>FAD</td>
<td>Failure Assessment Diagram</td>
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<tr>
<td>FCC</td>
<td>Face-Centred-Cubic</td>
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<tr>
<td>FE(A)</td>
<td>Finite Element (Analysis)</td>
</tr>
<tr>
<td>FTT</td>
<td>Fracture Toughness Testing</td>
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<tr>
<td>Abbreviation</td>
<td>Description</td>
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<tr>
<td>GDA</td>
<td>Generic Design Assessment</td>
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<tr>
<td>IoF</td>
<td>Incredibility of Failure</td>
</tr>
<tr>
<td>HAZ</td>
<td>Heat Affected Zone</td>
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<tr>
<td>HRR</td>
<td>Hutchinson, Rice and Rosengren Stress Field</td>
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<tr>
<td>HSLA</td>
<td>High Strength Low Alloy</td>
</tr>
<tr>
<td>IAR</td>
<td>Irradiated, Annealed, Reirradiated</td>
</tr>
<tr>
<td>IARA</td>
<td>Irradiated, Annealed, Reirradiated, Annealed</td>
</tr>
<tr>
<td>JEDI</td>
<td>J-Equivalent Domain Integral</td>
</tr>
<tr>
<td>JSW</td>
<td>Japan Steel Works</td>
</tr>
<tr>
<td>LA</td>
<td>Local Approach</td>
</tr>
<tr>
<td>LCF</td>
<td>Load Cool Fracture</td>
</tr>
<tr>
<td>LEFM</td>
<td>Linear Elastic Fracture Mechanics</td>
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<tr>
<td>LOCA</td>
<td>Loss of Coolant Accident</td>
</tr>
<tr>
<td>LUCF</td>
<td>Load Unload Cool Fracture</td>
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<tr>
<td>NCCEF</td>
<td>National Composites Certification and Evaluation Facility</td>
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<tr>
<td>ND</td>
<td>Neutron Diffraction</td>
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<tr>
<td>NPP</td>
<td>Nuclear Power Plant</td>
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<td>ONR</td>
<td>Office for Nuclear Regulation</td>
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<td>PS</td>
<td>Plane Stress</td>
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<td>PWHT</td>
<td>Post-Weld Heat Treatment</td>
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<td>PWR</td>
<td>Pressurised Water Reactor</td>
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<tr>
<td>$\varepsilon$</td>
<td>Plane Strain</td>
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<tr>
<td>RKR</td>
<td>Ritchie, Knott, Rice Local Approach Methodology</td>
</tr>
<tr>
<td>RPV</td>
<td>Reactor Pressure Vessel</td>
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<td>SD</td>
<td>Standard Deviation</td>
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<tr>
<td>SENB</td>
<td>Single Edge Notched Bend</td>
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<tr>
<td>SIF</td>
<td>Stress Intensity Factor</td>
</tr>
<tr>
<td>SSY</td>
<td>Small Scale Yielding</td>
</tr>
<tr>
<td>TEM</td>
<td>Transmission Electron Microscopy</td>
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</table>
Abstract

The reactor pressure vessel (RPV) is a major component in all current nuclear reactor designs. It consists of several forged components joined through a welding process. The process of welding introduces residual stresses to the structure. Residual stresses are those which are present with no external loading on the structure. Plastic strains may also have been introduced to the heat affected zone (HAZ) of the weld. Both residual stresses and plastic strains are considered as load history effects.

Cleavage fracture is a brittle mode of failure and would be catastrophic were it to occur in an RPV weld. The structural material used in the forgings that make up the RPV is ferritic steel, which, at the usual operational temperatures of nuclear plant, would not be susceptible to cleavage fracture. However, long-term neutron irradiation embrittles the RPV. This embrittlement, when combined with lower temperatures during routine plant shutdown events, renders cleavage fracture a concern.

Most structural integrity assessments consider a single parameter fracture toughness value, obtained by testing deeply cracked specimens. It is known that geometrical factors such as reduced crack depths, or loading conditions such as tension instead of bending, increase the effective fracture toughness of the material, through a reduction in hydrostatic stresses, known as a loss of constraint.

What is not clearly understood is the combined effects of the load history on constraint. The work presented in this thesis provides a methodology, using a 2-parameter fracture toughness approach, to predict failure through cleavage, at various levels of constraint, with both residual stresses and plastic strains. The methodology employs a failure curve in the J-Q space, derived from as received conditions, which is shown to well predict the toughness of specimens under the various initial conditions described above.

This thesis presents an experimental programme and series of numerical analyses that enabled the derivation of the J-Q failure curve, and validated its ability to predict failure for various constraint levels and load histories.
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I, Jack Beswick, declare that no portion of the work referred to in this thesis has been submitted in support of an application for another degree or qualification of this or any other university or other institute of learning.

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Acknowledgements

First, I would like to thank my supervisors: Professors Andrey Jivkov and Andrew Sherry, Dr Peter James and John Sharples. Their superlative knowledge of structural integrity and the nuclear industry overall, and their willingness to disseminate it at the drop of a hat have enabled the completion of the work presented in this thesis; and their unwavering support and encouragement have provided me with the confidence to be able to complete the document you are reading. Thank you also to the EPSRC and Wood for sponsoring the project.

Further thanks are given to Dr David Stanley and Caroline Lalley in the EngD centre, and Bev Knight at the University of Manchester for administrating the course so well and providing sound advice on many of the issues that arose during the project.

Sincere thanks to the team at Wood for their experimental support, especially Paul Hutchinson, Peter Birkett, Matthew Jones, Mike Hilton, Mark Callaghan and Colin Austin; and all the others at Walton and Newton House for welcoming me.

Thanks to the team in the NCCEF for their support with the machining and pre-staining, especially Tom Lucas, Dr Eddie McCarthy, Professor Costas Soutis, Ryan Delve and Chris Hyde.

I would like to finally thank all of my friends, family and colleagues for their constant friendship, advice and encouragement Jill, Phil, Sam, Joey, Alice, Jacob, Adam and Todor for their patience, kindness and constant belief in me throughout the doctorate.
1. Introduction

The reactor pressure vessel (RPV) is an integral part of all current nuclear power plant (NPP) designs. The RPV houses the core, inlet and exhaust nozzles for the coolant, and its structural integrity is essential for the safe operation of the NPP. All forged components are joined together through a welding process. The process of welding introduces significant residual stresses within the structure, which are of major concern when assessing the integrity of the RPV.

Residual stresses are those that are still present when the initial loading has been removed; these stresses are self-equilibrating (they cause the body to exert zero total force) and may be caused through welding, or installation or manufacturing defects. They have significant effects on the effective fracture toughness of the materials within the structure, which is a parameter used to determine critical defect or crack sizes; or loads at which the structure containing these defects is likely to fail.

The cleavage fracture toughness of a material is its ability to resist fracture through transgranular cleavage. This mechanism of failure would be catastrophic to an RPV and therefore the NPP as a whole, potentially allowing coolant to be released and causing a core meltdown. The cleavage fracture toughness, in the presence of residual stresses, is of significant interest throughout the life cycle of the RPV, and especially for life extension projects, as end of life RPVs are more susceptible to cleavage fracture: The effective fracture toughness of RPV steels is not only affected by weld induced residual stresses, but damage induced through neutron irradiation, which leads to radiation hardening and embrittlement. A reduction in temperature also increases the likelihood of cleavage fracture occurring through reducing the mobility of defects in the crystal lattice of the steel.

All of the above factors (residual stresses, radiation, temperature) have significant effects on the effective fracture toughness of a material because they increase or decrease the level of plasticity in the material; and welding processes may also introduce significant plasticity. For brittle, linear-elastic materials (such as ceramics or glass at room temperature), there are validated analytical fracture mechanics approaches used for defect assessments in structures, which are both relatively simple to carry out and validated through testing. This linear elastic fracture mechanics (LEFM)
approach is also applicable for metals where the plastic zone around a crack is suitably small, which is known as small scale yielding (SSY) conditions.

Where the effect of plasticity is not negligible more complex approaches are required to predict failure and assess structural integrity. These approaches include elastic-plastic fracture mechanics (EPFM), which expand on LEFM to account for plasticity. An alternative is the local approach (LA) methodology which includes statistical analysis of fracture initiators in the microstructure around the crack tip and can be used to predict the probability of failure under prescribed loads.
1.1. Project Summary

Ferritic RPV steel welds that are not stress-relieved can contain substantial residual stresses and may have a plastic strain history, both of which are encompassed under ‘load history effects’. These load history effects have to be considered in structural integrity assessments based on real or postulated crack-like defects. This is of practical importance; with, for example, respect to safe-end nozzle transition welds (usually dissimilar metal welds connecting the RPV nozzles to primary circuit pipework) and control rod drive mechanisms welds, which cannot be post-weld heat treated (PWHT) in-situ.

Ferritic steels exhibit cleavage fracture at sufficiently low temperatures that it is not a concern during normal plant operation, but cleavage probability increases with plant lifetime due to neutron irradiation-induced material damage. The combined effects of both residual stress and plastic strain on effective cleavage fracture toughness are still not sufficiently well understood and their combined effects cannot be accurately characterised by a single fracture toughness parameter without undue conservatism being built into the assessment. The hypothesis examined in this thesis is that a 2-parameter approach, e.g. using J-Q two-parameter fracture mechanics, may provide a more accurate and less conservative assessment, where J is the crack driving force and Q is a parameter used to quantify crack tip constraint; a measure of crack tip plasticity growth under loading, predominantly affected by geometry and loading type (i.e. tension or bending), which is also affected by the stress triaxiality at the crack tip. Using this two-parameter framework an engineer can perform finite element analysis (FEA) of the cracked geometry under a given loading condition that may include a residual stress profile to calculate a loading curve in the J-Q space. The intersection between the loading curve and a material specific, J-Q, 2-parameter, fracture toughness curve may then be used in a defect assessment, with the 2-parameter curve replacing the usual 1-parameter fracture toughness value.

The outstanding issue investigated in this thesis was to assess whether a unique material failure curve in the J-Q space exists for a specific RPV steel irrespective of the initial stress-strain conditions, geometry and loading. This required that as many as possible failure points in the J-Q space were known and these were reached starting
from different initial stress/plastic strain conditions. The underlying challenge was the limited knowledge of the effects of prior plasticity on cleavage fracture probability in general and of initial plastic deformations when present in particular. Progress towards developing new understanding of prior plasticity and residual stress on cleavage probability was made by considering four different scenarios involving testing and numerical analyses of specimens: (1) without initial plastic strains and without residual stresses; (2) with initial homogeneous plastic strains and without residual stresses; (3) without initial plastic strains and with residual stresses; (4) with initial plastic strains and with residual stresses, as shown in Table 1:

<table>
<thead>
<tr>
<th></th>
<th>Without plastic strain</th>
<th>With plastic strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>Without residual stress</td>
<td>(1)</td>
<td>(2)</td>
</tr>
<tr>
<td>With residual stress</td>
<td>(3)</td>
<td>(4)</td>
</tr>
</tbody>
</table>

Table 1: Experimental and numerical test scenarios

The aim of this research was to investigate the existence of a unique material failure curve by undertaking a combination of numerical analysis and targeted experimental work with respect to the above four scenarios. These were realised by fracture toughness testing (FTT) of specimens extracted from RPV steel plates of as received material and plates subjected to an appropriately selected uniform level of plastic strain. Selected sets of specimens were tested directly, while others were subjected to mechanically induced tensile residual stresses by compression ahead of the crack prior to testing. Specimens with different constraint conditions were studied to sample different failure points along various J-Q loading paths. Numerical analyses through validated finite element analysis (FEA) of the tests were performed and the ability of two parameter fracture mechanics to predict the experimental failure points across the tested scenarios were investigated, with recommendations given as to how a material specific failure curve could be used by an engineer assessing the defect tolerance of a welded structure with a known residual stress field and or plastic strain history.
2. Literature Review

2.1. Ferritic RPV Steels

Typical RPV designs incorporate high strength low alloy (HSLA) steel forgings, arranged similarly to those shown in Figure 2-1, joined through welding processes. The welded structure is internally clad with a corrosion resistant stainless steel. The work in this thesis is concerned with the outer layer of HSLA steel, which is more susceptible to cleavage fracture, as it is harder and less ductile than the inner stainless-steel cladding. One such steel is A533b Class 1 manufactured by Japan Steel Works (JSW) (hereafter referred to as A533b), and is the material under investigation in this project. This material has been used in the fabrication of Japanese boiling water reactor (BWR) and pressurised water reactor (PWR) RPVs since the 1960s and is still used in the current advanced boiling water reactor (ABWR) design (1), which has recently passed the UK nuclear regulator’s (ONR) Generic Design Assessment (GDA), and is planned to be built on the Welsh Island of Anglesey when the full site license is obtained. The material has high purity due to its iron content being extracted solely from recycled steel and the inclusion of a double degassing process to remove hydrogen and oxygen during ladling. Removal of phosphorus and sulphur is also carried out through refinement processes; schematics of which are shown in Figure 2-2. The chemical composition of the steel following these processes is shown in Table 2.
2.1.1. Microstructure of A533b Class 1 Steel

The material used in this project was a plate of A533b, produced using the same methods described to make the material for the RPV forgings, with the plate rolled as opposed to being forged in dies. A533b and similar types of low alloy steel used in RPV forgings are typically referred to as ferritic steels, in that their matrices are predominantly comprised of the $\alpha$-ferrite phase of an iron-iron carbide (Fe-Fe$_3$C) constitution. $\alpha$-ferrite has a body-centred-cubic (BCC) crystal structure. This type of crystal structure contains 48 potential slip systems, along which plastic deformation usually occurs, but none of these are more closely-packed planes than others, so there is no preferential crystallographic slip system, which strengthens the material. Face-
centred-cubic (FCC) crystal structures, found in austenitic stainless steels, have 12 close packed planes that enable slip, increasing ductility.

The plate used in this project was annealed (heated into the austenite phase to ~900°C) for 4.25 hours; quenched (cooled rapidly in water); tempered at ~650°C for 6.7 hours; air-cooled and cross-rolled (2). Microscopy of A533b plate was carried out (3) and revealed areas of both ferrite and martensite rich regions, with varying levels of carbon in the martensite areas, as shown in Figure 2-3. The introduction of the metastable martensite phase occurs through the annealing and subsequent quenching of the steel. The presence of the martensite provides the steel with strength and hardness, whilst being the most brittle of the Fe-Fe₃C phases. The tempering process reduces the hardness introduced by the martensite as it enables carbides to precipitate, releasing interstitial carbon atoms trapped in solution. The martensite’s microstructure changes through tempering, from the pure body-centred-tetragonal (BCT) structure, which leads to needle type grains, to very small Fe₃C (cementite) particles within the α-ferrite matrix (tempered martensite). This change in crystal structure drastically improves ductility and toughness and is a desirable within RPV steels. The carbon rich (brittle) martensite banded regions are parallel to the rolling direction of the plate, which effects anisotropic material behaviour.

Figure 2-3: Micrographs showing ferrite (left) and martensite (right) rich bands, and close-ups of microstructures (3)
2.1.2. Cleavage Fracture Mechanism

Cleavage fracture is a brittle failure mode, although it can be preceded by restricted plastic flow. It is defined as the “rapid propagation of a crack along a particular crystallographic plane”, where the preferred separation plane \{100\} is that which has fewest bonds and most spacing between planes (4).

Cleavage fracture in ferritic steels usually begins with the breaking of a brittle second phase particle, such as a carbide, in the presence of a macroscopic defect, acting as a stress concentrator. The breaking of the carbide nucleates a microcrack, which then propagates through a single grain of ferrite. The ferrite surrounding the carbide then undergoes plastic deformation (caused by mobilised dislocations), which may arrest the crack growth, or if the fracture toughness of the ferrite is too low, will propagate into neighbouring grains, as shown in Figure 2-4. If the microcrack propagation is not arrested through the plasticisation of ferrite grains, or dislocation pile ups at grain boundaries, it will cause macroscopic brittle fracture (propagating up to the speed of sound), with some initial plasticity (5).

Figure 2-4: Transgranular cleavage associated with cracked grain boundary carbides (6)
2.1.3. Ductile to Brittle Transition

The brittleness associated with cleavage fracture increases as the ability of the microstructure to arrest the crack growth decreases. With ample slip systems in the ferrite matrix in place there is opportunity for dislocations within the crystal lattice to accumulate, which absorb the energy that would otherwise be used to create new fracture surfaces. More ductility means the material is more likely to fail through plastic collapse. At higher temperatures dislocations are more mobile and this is more likely to occur, therefore at operating temperatures of NPP (~280°C), cleavage fracture in RPV steels is not a primary concern. At these temperatures, the ferritic steel’s fracture toughness is sufficiently high that it is assumed that any failure at a defect would be through ductile tearing.

Where cleavage fracture becomes a concern is at shutdown temperatures. UK nuclear plant licensing requires that a complete shutdown is demonstrated during the Periodic Safety Review. As temperature decreases there reaches a point where there is a sharp decrease in the fracture toughness of ferritic steels, known as the ductile to brittle transition temperature (DBTT). At this temperature, the probability of failure through cleavage fracture increases. As temperature decreases the probability of cleavage increases to a point where brittle fracture is more likely than plastic collapse, if the structure were to fail at a defect. In practice this is designed against, there are however considerations that are not yet sufficiently well understood to discount completely the possibility of cleavage fracture.

Neutron irradiation has been widely understood to provide an increase in the yield strength of RPV steels. Increase in yield strength leads to a decrease in ductility, which is necessary to arrest cleavage crack growth. Margolin et al. (7) empirically proved that irradiation also resulted in a decrease in the parameter $\bar{\sigma}_{\text{d}}$, which is a fitting parameter used to describe a Weibull distribution of the probability of local microstructural stresses greater than the strength of either carbides or carbide/matrix interfaces in a unit volume ($\sigma_0$). This reduction implies that an increase in the radiation dose the RPV steel is subjected to causes a shift, and a change in shape in the fracture toughness transition curve, as shown in Figure 2-5.
A reduction in the load at which cleavage fracture initiates has also been shown to exist in the presence of residual stresses. James et al. (8) showed that similar SENB specimens at the same temperatures had reduced effective fracture toughness values when residual stresses are present, as shown in Figure 2-6. From a microstructural perspective weld induced residual stresses are created by non-uniform cooling in and around the welds that can lead to shrinkage mismatch in the components, which, alongside phase changes in both the parent and weld material, lead to internal stresses remaining after welding.
The combination of radiation embrittlement and residual stresses is an area of high concern. As discussed earlier, at operating temperatures there is less concern because cleavage fracture considered unlikely. However, the combination of the increase in the temperature at which fracture toughness sharply reduces, caused by radiation embrittlement, with the reduction of effective fracture toughness caused by residual stresses, suggests there is potential for cleavage fracture to occur during cool down cycles of the reactors. Current reactor designs opt for very high safety factors to prevent this occurring, although this leads to undue conservatism in reactor design and structural integrity assessments.

2.1.4. Annealing and Post Weld Heat Treating

It has been shown that steels, which have suffered irradiation embrittlement, can benefit from heat-treating processes. A 2006 paper (9) demonstrated a decrease of 96°C in the DBBT of irradiated A533b steel following post-weld heat treatment (PWHT), as shown in Figure 2-4. The figure shows the Charpy energy transition curves, where the tests measure impact energy as opposed to load at fracture in cracked specimens. The work discovered through atom probe tomography that irradiation caused the formation
of clusters of silicon, manganese, copper phosphorous and nickel in ferritic steels. These precipitates are known to cause the embrittlement and hardening which is associated with the reduction in toughness and ductility in RPV steels.

What was demonstrated in this paper (9) was that the annealing process did not dissolve the clusters back into the matrix, but the recuperation of material properties was made available through the coarsening of the clusters. This meant that the precipitates grew larger and were less capable of impeding dislocations, increasing ductility. It is worth noting that radiation has various effects in microstructure, so the energy absorption reduction caused by precipitation hardening is as likely to have an effect on cleavage fracture as the shift in breakage from carbides to carbide-matrix interface as mentioned in section 2.1.2. This study focused on solely on precipitates, and this hardening mechanism was proven to be somewhat reversed through the heat treating.

![Figure 2-7: Unirradiated; irradiated, annealed, reirradiated (IAR); irradiated, annealed, reirradiated, annealed (IARA) ductile to brittle transition curves (9)](image)

Weld induced residual stresses, described in detail in section 0, are also reduced through heat-treating. It is not however always possible to conduct PWHT on certain
RPV welds, such as safe-end transitions or control-rod drive mechanism welds. It is these non-stress-relieved welds that are of significant concern to the industry, and the areas where the research presented in this thesis is applicable.

2.2. Fracture Mechanics

Fracture Mechanics is the science of predicting the effects of stress concentrators such as cracks or crack like defects on the failure of a structure. The aims of these analyses include understanding at which load the crack will propagate through the material, leading to a catastrophic failure; or if the material will undergo plastic collapse, where large scale yielding around the crack or defect will cause the structure to be rendered unfit for purpose.

2.2.1. Griffith’s Energy Criterion

Fracture mechanics has a single analytical solution, in that it is derived from first physical principles. Griffith (10) first suggested that for a crack of length 2a in an infinite body, for a defect to propagate, the strain energy in the system (Us) must be greater than the energy required to create new fracture surfaces (UE) thus the entire system’s energy is reduced through the propagation of a crack. The equation for the elastic strain energy released by the growth of a crack is:

\[ U_s = \frac{\sigma^2}{E} \pi a^2 t \]  \[1\]

where \( \sigma \) is the applied stress, \( E \) is the Young’s modulus, \( a \) is the crack length and \( t \) is the material’s thickness. The surface energy that is absorbed by the creation of new surfaces can be expressed as:
\[ U_E = 4at\gamma_s \]  \hspace{1cm} [2]

where \( \gamma_s \) is the surface tension of the material.

The rate of change of energy of the system with respect to the change in crack length can thus be expressed as:

\[
\frac{dU}{da} = \frac{dU_s}{da} - \frac{dU_E}{da} \hspace{1cm} [3]
\]

where \( U \) is the energy of the entire system. When the system is at energetic equilibrium, just before a crack propagates, \( \frac{dU}{da} = 0 \), from which the equation for the critical stress at fracture (\( \sigma_f \)) was derived:

\[
\sigma_f = \sqrt{\frac{2E\gamma_s}{\pi a}} \hspace{1cm} [4]
\]

### 2.2.2. Irwin’s Modification

The theory Griffith derived was experimentally validated using glass specimens (10). The theory however produced large discrepancies when compared with experimental results with ductile materials such as metals. It was suggested by Irwin (11) that the reasoning behind this was due to the onset of plasticity around the crack tip. He proposed that yielding occurred at the crack tip, which required far more energy than that required for the creation of new fracture surfaces. Therefore, a plastic energy term (\( \gamma_p \)) was added to the fracture stress equation (12):
\[ \sigma_f = \sqrt{\frac{2E(y_s + y_p)}{\pi a}} \]  

[5]

Incorporating the above into Griffith’s energy criterion the fracture energy \((G_f)\) term could then be characterised, known as the strain energy release rate (where rate refers to change with respect to crack area, as opposed to time), or the energy absorbed by the creation of new fracture surfaces + the creation of the plastic zone at the crack tip. At the critical fracture stress the strain energy release rate reaches a critical value to enable the onset of fracture \((G_c)\), which can also be described as the fracture toughness of the material (12):

\[ G_f = \frac{\pi \sigma^2 a}{E} \]  

[6]

\[ G_c = \frac{\pi \sigma_c^2 a}{E} \]

### 2.2.3. Stress Intensity Factor

An alternative method to the energy method to describe crack tip stresses involves the use of a constant known as the stress intensity factor \((K)\), which is also described as the crack driving force. This parameter was developed as a scaling parameter to define stresses at any point \(P\) ahead of the crack, as illustrated by the Westergaard (13) functions in Figure 2-8. The functions are derived for a linear elastic material only, and do not consider crack tip plasticity. Subscript \(I\) denotes Mode I crack opening, where the stress acts normal to the crack surfaces as opposed to shear methods, Modes II and III, which are considered to be less significant in most structural applications. The Westergaard functions are applicable for all cracks in all geometries. The scaling parameter \((K)\), however, is affected by the body and crack geometry, boundary conditions and loading method (4).
Mathematical derivations for $K_I$ are readily available in handbooks for basic two-dimensional geometries and loading scenarios. The most basic of these derivations is that for a crack of length $2a$ in an infinite plate, loaded under uniform tensile stress ($σ$), normal to the crack:

$$K_I = σ\sqrt{πa}$$  \[8\]  

As the geometry and loading conditions become more complex dimensionless calibration functions ($Y$) are incorporated into equation 8 so the generic equation for $K_I$ becomes:

$$K_I = Yσ\sqrt{πa}$$  \[9\]  

The value of $Y$ has been calculated numerically for various standard crack and fracture toughness test specimen geometries, which allow stress intensity factors for common defects to be calculated.
2.2.4. Plane Stress and Plane Strain

In real world applications, defects exist in 3-dimensional structures. This invokes new considerations for the calculation of $K_i$. At a free surface, or in very thin structures, a structure can be said to be loaded in plane stress conditions: the stress is constrained to act biaxially in that it is assumed to be either zero or constant in one of the principal directions. Where structures are very thick and one principal dimension is much larger than the others, the strain in the direction of the largest dimension is approximately zero. Material is constrained from moving by the surrounding material, unlike at a free surface. This is known as the plane strain condition.

As illustrated by Figure 2-9 the stress acts at the centre of the specimen in the $x$, $y$ and $z$ directions. Towards the free surface the stress in the $z$ direction tends towards zero, as there is no material to constrain movement. At this point the material is under plane stress conditions. In reality cracked structures are very rarely under plane stress conditions, except for very thin geometries such as foils, or at the free surface of three-dimensional geometries. The two-dimensional theoretical $K$ calculations such as those used to derive equation 7 are also considered to be under plane stress considerations. Most real crack tip regions are however considered to be under plane strain conditions,
in that there is assumed to be enough material to prevent strain in the largest dimension.

The following adjustments are made for calculations involving the material’s Young’s modulus. These definitions are derived using Hooke’s law for a three-dimensional element, and setting the stress (for plane stress), or strain (for plane strain) in one of the principal directions equal to zero. $E'$ refers to the effective Young’s Modulus under the required condition, and $\nu$ is the material’s poisons ratio. PS denotes plane stress and $\varepsilon$ plane strain:

$$E'(PS) = E$$  \hspace{1cm} [10]$$E'(P\varepsilon) = \frac{E}{1 - \nu^2}$$

### 2.2.5. Fracture Toughness

Equation 6 was defined as the fracture toughness based on the energy criteria postulated by Griffith and Irwin. A similar value is obtained using the stress intensity factor method: the value of $K_i$ reaches a critical point at which a crack will propagate through the material. As the stress intensity factor method is only applicable to material behaving elastically, the method of failure will be fast fracture, where the crack propagates up to the speed of sound through the material, and no yielding of the crack tip is assumed, which follows Griffith’s criteria in that all of the strain energy is released into the creation of new surfaces. The linear elastic fracture toughness, or critical stress intensity factor, known as $K_{ic}$, is described as follows:

$$K_{ic} = Y\sigma_f\sqrt{\pi a}$$  \hspace{1cm} [11]

Most simple engineering calculations assume plane strain conditions, so $K_{ic}$ is often also described as the material’s plane strain fracture toughness. It is therefore
possible to draw comparison between the two ‘fracture toughness’ methods that have been previously described. For a linear elastic material:

\[ G_c = \frac{K_{IC}^2}{E'} \]  

[12]

A schematic of a standard (and that used for the experimental work presented in this thesis) fracture toughness test (FTT) specimen is the single edge notch bend (SENB) shown schematically in Figure 2-10. These specimens are loaded to failure, and the value of \( K_I \) at this occurs is defined as the fracture toughness of the material, (which is also referred to as \( K_{mat} \)) from which the specimen is made, if only a single test is performed. Often more than one test is performed, and the analysis of multiple tests is described in section 2.4.2.

Figure 2-10: SENB specimen schematic

The following equation is used to calculate the linear elastic stress intensity factor from applied load \( F \) for SENB specimens:
\[ K_I = \frac{FS}{BW^{1.5}} \cdot f \left( \frac{a}{W} \right) \]

\[ f \left( \frac{a}{W} \right) = \frac{3 \left( \frac{a}{W} \right)^{0.5} \left[ 1.99 \left( \frac{a}{W} \right) \left( 1 - \frac{a}{W} \right) \left( 2.15 - \frac{3.93a}{W} + \frac{2.7a^2}{W^2} \right) \right]}{2 \left( 1 + \frac{2a}{W} \right) \left( 1 - \frac{a}{W} \right)^{1.5}} \]  \[13\]

where \( S = 4W \) and \( a/W < 0.6 \) (as shown in Figure 2-10). This equation is reported to give <0.2% error for linear elastic calculations (14). The SENB is one type of specimen used to calculate the fracture toughness of materials. BS 7448 Part 1 (15), the British standard for fracture toughness testing, and ASTM 1820 (15) the contemporaneous American standard, state that the size of the specimen must conform to:

\[ W - a, B, a > 2.5 \left( \frac{K_{ic}}{\sigma_{ys}} \right)^2 \]  \[14\]

where \( \sigma_{ys} \) is the 0.2% proof stress of the material (\( K_{ic} \) is estimated prior to testing, or a sacrificial test is carried out with the material in question) to give approximate dimensions. For the material under consideration in this thesis, based on the test results presented in sections 4.1 and 5.2, where the 0.2% proof stress was measured at 733MPa and the fracture toughness, \( K_{ic} \), at 59MPa\(\sqrt{m}\), this gives a minimum pertinent dimension of 15mm. Whilst equation 14 may formulate a starting point for the design of the specimens, the standard which recommends equation 14 considers deep cracks where \( W-a \) and \( a \) are similar values. Non-standard tests, with very shallow cracks, similar to those presented in this thesis could not be achieved if \( a \) were to be considered in equation 14, so in these cases it is sufficient to only consider \( W-a \) and \( B \) in equation 14.

A maximum, limiting factor for \( K_{ic} \) (where \( K_{ic} \) is the fracture toughness calculated from the J integral from fracture toughness tests as described in section 2.3.3) is described in ASTM 1921 (16), an American standard which extends (15) to include details of the calculation of various pertinent parameters to describe the scatter in fracture toughness test results (see section 2.4.2):
\[
K_{lc\ (limit)} = \left( \frac{E(W - a)\sigma_{ys}}{30(1 - \nu^2)} \right)^{1/2}
\]

which again may be used to enable decisions to be made with regards to the design of the specimen. The British and American standards (15), (17) offer similar diagrams to assist an engineer in designing their specimens, through describing the ratios of pertinent dimensions, an example of which is shown in Figure 2-11. The recommended \(a/W\) (see Figure 2-10) ratio is between 0.45 and 0.7, as generally fracture toughness estimates for use in engineering assessments consider conservative values extracted from deeply cracked specimens, where it is assumed the effects of plasticity are negligible, known as small scale yielding (SSY), and is described in more detail in section 2.3.1.

Figure 2-11: SENB dimension ratios (15)

SENB specimens are one of two generic designs most commonly used for fracture toughness test programmes, the other being compact tension (CT) specimens. These specimens are not considered in this thesis; SENB specimens are used, as they enable varying levels of crack tip constraint to be built in, through simply varying crack depth, which is more difficult using the CT design.
2.3. Elastic Plastic Fracture Mechanics

Excluding Irwin’s (11) modification for plasticity the calculations described so far are only technically applicable for linear elastic materials. In reality, there is limited applicability in structural integrity assessments for LEFM. This is because metals, when loaded to destruction, will usually fail through plastic collapse, where yielding, necking, and eventually ductile rupture occur. As explained in section 2.1.2, cleavage fracture, although apparently a brittle mode of failure, is preceded by some crack tip plasticity, which must be considered in engineering assessments.

2.3.1. Small Scale Yielding

The Westegaard functions 1/√r singularity implies infinite stresses at the crack tip (r=0). This is not possible and is accounted for through the yielding of the material at the crack tip, initially postulated by Irwin (11). Where the plastic zone ahead of the crack tip is sufficiently small however the LEFM principles apply at distances greater than zero and it is possible to assess structural integrity using these methodologies. This is known as small scale yielding (SSY). Using equations 7 and equating the principal stresses to the yield stress (σy) of the material it is possible to define the size of the plastic zone (rp) for plane stress and plane strain conditions (12):

\[
\begin{align*}
    r_{p(PS)} &= \frac{K^2}{2\pi\sigma_y^2} \\
    r_{p(PS)} &= \frac{K^2}{6\pi\sigma_y^2}
\end{align*}
\]

When these plastic zones are small compared to the dimensions of the cracked geometry and the crack itself, SSY conditions apply, and linear elastic calculations can be used to define stress fields ahead of a crack. If these areas are large however alternative
analysis methods are used to describe the stress field ahead of a crack, which consider the onset of plasticity in the material surrounding the crack.

### 2.3.2. The J Integral

The J integral, first postulated by Rice in 1968 (18) is a path-independent measure of strain energy release rate for elastic plastic materials whose stress strain behaviour can be characterised by the Ramberg-Osgood hardening equation:

\[
\varepsilon = \frac{\sigma}{E} + \alpha \left(\frac{\sigma}{\sigma_0}\right)^n
\]  

where \( E \) is the Young’s modulus, \( \sigma_0 \) is a reference stress (usually taken as the 0.2% proof stress), \( \alpha \) is a constant, and \( n \) is known as the strain hardening exponent. This equation is often used for fitting curves to stress strain data of metals.

The J integral is defined by equation 18 with a schematic also given in Figure 2-12:

\[
J = \int \left( w dy - T_i \frac{\partial u_i}{\partial x} ds \right)
\]  

![Figure 2-12: Contour Path at Crack Tip for J Integral Calculation](image-url)
where \( w \) is the strain energy density, \( T_i \) are components of the traction vector \( (T_i = \sigma_{ij} n_j) \), where \( n_j \) is a unit vector acting perpendicular to the contour \( u_i \) are components of the displacement vector and \( ds \) is the length increment along contour \( \Gamma \) (12). The critical value \( J_c \) at which failure occurs is again known as the fracture toughness of the material and is \( J \) equal to \( G \) for linear elastic materials.

It is possible to characterise the stress and strain fields ahead of a defect using the J integral through calculation of the HRR fields (after Hutchinson (19) and Rice & Rosengren (20)), which are the elastic-plastic analogue to the Westegaard functions described by equation 7 (4):

\[
\sigma_{ij} = \sigma_0 \left( \frac{J}{\alpha l_n \sigma_0 \varepsilon_0 r} \right)^{\frac{1}{n+1}} \tilde{\sigma}_{ij}(\theta, n)
\]

\[
\varepsilon_{ij} = \alpha \varepsilon_y \left( \frac{J}{\alpha l_n \sigma_0 \varepsilon_0 r} \right)^{n/(n+1)} \tilde{\varepsilon}_{ij}(\theta, n)
\]

where \( l_n \) is an integration constant; and \( \tilde{\sigma}_{ij} \) and \( \tilde{\varepsilon}_{ij} \) are non-dimensional functions of \( \theta \) and \( n \). The values of these dimensionless have been tabulated by Shih (21).

The calculation of the J integral is usually carried out numerically. This can be achieved using FEA software such as the commercial package Abaqus (22). The software requires the definition of concentric contour regions within the analysis mesh, and for three-dimensional calculations a cylindrical region is defined around the crack line as can be seen in Figure 2-13 (23).
Technically the J-Integral is only for use with material models which are defined by equation 17 and therefore display non-linear elasticity. Furthermore, it assumes proportional loading in that all elements of the stress tensor increase monotonically through loading. In reality, these conditions rarely hold true in elastic-plastic analyses. J-integral analysis is however widely considered as the best approximation of the crack-driving force and is usually considered valid in structural integrity assessments.

2.3.3. Measuring J Experimentally

During fracture toughness testing of specimens, such as the SENB shown in section 2.2.5, the load and crack mouth opening displacement (CMOD) are measured. Following testing crack growth is measured. This involves regarding the crack front under a microscope and measuring how far the crack has extended at 9 equidistant points across the thickness of the specimen (with a 1%B offset). This is easily discernible through a change in appearance of the fracture surface when observed under an optical microscope. It is also necessary to measure any ductile tearing prior to fracture. Again, this is discernible by a change in appearance of the fracture surface, although in the case of cleavage fracture this is usually negligible, or zero in some cases. The load, CMOD, and crack growth measurements are used to calculate J at failure \((J_c)\) using the following methodology:
The elastic component of the crack driving force $J_e$ is calculated from the stress intensity factor for the specimen type ($K_i$, such as equation 13 for an SENB specimen):

$$J_e = \frac{K_i^2}{E} \quad [20]$$

and the plastic component $J_p$:

$$J_p = \frac{\eta_p U_p}{B(W - a_0)} \quad [21]$$

where $U_p$ is the plastic component of work done during the test, calculated through integration of the load displacement curve, and removing the elastic component (see Figure 2-14), $a_0$ is the initial measured crack length (averaged across the crack front), and $\eta_p$ is a plasticity correction factor. As described in section 2.2.5, the recommended crack depth ratios for standard fracture toughness tests is $0.45 < a/W < 0.7$ as recommended in the fracture toughness testing standards ASTM 1820 (15) and BS7448 part 1 (17). These standards simply give the value of $\eta_p$ as 2 for these high constraint geometries. BSI EN ISO 15653 (24), which details further assessing weld fracture toughness properties, and ASTM 1921 (16), which formalises definition of $T_0$ and the master curve (see section 2.4.2) defines $\eta_p$ as, for shallow notches, where $0.1 < a/W < 0.45$:

$$\eta_p = 3.667 - 2.199 \left( \frac{a_0}{W} \right) + 0.437 \left( \frac{a_0}{W} \right)^2 \quad [22]$$

Equation 22 is based on a series of plane strain finite element analyses in which $J$ was numerically calculated, for $0.05 < a/W < 0.70$ for various material models (25). The very low constraint geometries tested for the work presented in this thesis, with a crack depth of $a/W=0.05$ are not represented in the standards but are included in the referenced work. This geometry is part of the novelty of the work presented in this
thesis and comparison is made section 5.2 between the experimental J values obtained using the above $\eta_p$ factor and those calculated using finite element analyses.

The plastic and elastic components are summed to give a final J at failure ($J_c$) such that:

$$J_c = J_e + J_p$$ \[23\]

and then finally where necessary $J_c$ may be converted to critical stress intensity factors ($K_{Jc}$) using equation 20.

![Figure 2-14: Load vs CMOD plot showing plastic area under curve](image-url)
2.4. UK Nuclear Industry Structural Integrity Assessments

Failure of certain components within nuclear plant must be proven through robust and thorough structural integrity assessments that failure is so unlikely that it is considered incredible. These are components, such as the RPV and primary circuit, that if they were to fail would pose a serious threat to people or the environment, in that their failure may lead to catastrophic consequence such as a loss of coolant accident (LOCA). It is these components which are subjected to defect tolerance assessments.

The ONR does not offer a prescriptive standard for an engineer to follow when proving the safety of such component. The only criteria for engineers when proving the safety of nuclear plant is any risk to safety is as low as reasonably practicable (ALARP). In the case of defect tolerance assessments of components with real or postulated defects, this requires that ‘incredibility of failure’ is proven. This means proving that failure of such a component is so unlikely it would be considered incredible. There are no prescriptive standards in the UK to enable an engineer to prove this, so instead a set of guidelines was commissioned in 1976 by the Central Electricity Generating Board (CEGB). This document is known as R6 (26), and is currently owned by EDF energy, the only civil nuclear plant operator in the UK. It describes in detail how defect tolerance assessments may be carried out, with the items pertinent to this thesis described below.

2.4.1. The Failure Assessment Diagram

The basis of R6, (26) is the failure assessment diagram (FAD). It is used for prediction of whether, when and how a defect is likely to cause failure of a structure; under either normal operating loads, or a potential loading spike (a drastic increase in pressure within the RPV causing an unusually high hoop or axial stress) caused by an abnormal pressure transient. R6 details methods for the incorporation of both primary and secondary stresses into the FAD assessment, where they are known to be in vicinity of a defect. Primary stresses are those which contribute to the plastic collapse of a structure, and secondary stresses do not, examples of which are weld induced residual stresses and thermal stresses.
The Option 1 FAD is a generic failure curve, against which any defect size, in any material, of any geometry, can be compared. The failure curve has been approximated from various data from experiments and numerical simulations. The two axes of the curve are labelled $K_r$ (which is the ratio of the stress intensity factor at the defect and the materials fracture toughness), and $L_r$, which is the ratio between the load on the structure and the structure’s collapse or limit load. The generic failure curve has the equation:

$$K_r = f(L_r) = \left[ 1 + 0.5L_r^2 \right]^{-\frac{1}{2}} \left( 0.3 + 0.7e^{p(-0.6L_r^6)} \right)$$

$$L_r < L_{rmax}$$

$$K_r = 0$$

$$L_r \geq L_{rmax}$$

The defect under consideration can then be plotted as an assessment point using its calculated $K_r$ and $L_r$ values, and the position of this point can be compared against the failure curve, which enables consideration as to how the defect may grow, or what effects change in loading or material may have on the likelihood or mechanism of failure. The FAD can be used for real cracks found in service, or postulated cracks during the design phase. The results from the FAD can be used to determine critical crack sizes within the structure, or critical loads on the structure. These can also be combined with fatigue and creep calculations, to predict the amount of time the structure will remain safe for, under various standard, or optimum, loading conditions.

An annotated example of an FAD is shown in Figure 2-15. The black dot is a structural assessment point, which represents the defect under examination. If the load increases the assessment point moves in a straight line drawn from the origin through the original point. The diagram shows how other parameters affect the position of the assessment point.
The FAD is an invaluable tool in predicting not only under what loading the defect is likely to cause failure, or critical crack sizes; but also, how the structure will fail. As with all fracture mechanics considerations the FAD considers whether the failure will be brittle, plastic, or an elastic plastic mechanism. Towards the left or top of the curve the failure mechanism is likely to be brittle, and as the curve begins to slope plasticity begins to be significant and at high values of $L_r$ it could is predicted failure would be through a ductile mechanism. The vertical red line is a cut-off point at $L_{r_{\text{max}}}$:

$$L_{r_{\text{max}}} = \frac{\bar{\sigma}}{\sigma_y}$$

$$\bar{\sigma} = \frac{\sigma_{ys} + \sigma_{UTS}}{2}$$

[25]

$\bar{\sigma}$ is referred to as the flow stress $\sigma_y$ and $\sigma_{UTS}$ and are the material’s the yield and ultimate tensile strengths, respectively. There are options to modify the FAD, including a
material specific Option 2 curve, which alters the failure curve based on a material’s specific uniaxial tensile properties. In this case:

\[
f_2(L_r) = \left[ \frac{E \varepsilon_{ref}}{L_r \sigma_y} + \frac{L_r^3 \sigma_y}{2E \varepsilon_{ref}} \right]^{-1/2}
\]

where \( \varepsilon_{ref} \) is a reference strain at a reference true stress=\( L_r \sigma_y \), obtained from the material’s stress strain curve (26).

There is also the option to modify the curve using a normalising constraint parameter \( \beta \) and material properties \( \alpha \) and \( m \) which describe toughness increase through a reduction in \( \beta L_r \). Further description of these variables is provided in section 2.5

\[
K_r = f(L_r)[1 + \alpha(-\beta L_r)^m]
\]

2.4.2. The Master Curve

Cleavage fracture toughness test results show significant scatter. When ascertaining a fracture toughness value for a specific material from a programme of several tests it may be necessary to describe both average and bounding values in terms of probability of failure, and the extent of the scatter. Also, as described in section 2.1.3, the fracture toughness of steels is temperature dependant, and there is a small transition range in temperature from ductile to brittle behaviour. Therefore, it is useful for engineers to be able to describe the probability of failure in a statistical manner, and ascertain the temperature at which the ductile to brittle transition occurs.

The scatter in cleavage fracture has been shown to follow a Weibull distribution which was initially described as (28):
\[ P_f = 1 - \exp \left[ - \left( \frac{K_i}{K_0} \right)^{m_1} \right] \]  \tag{28}

where \( P_f \) is the probability of failure, \( K_i \) is the applied stress intensity factor and \( K_0 \) is a normalisation stress intensity factor where the failure of probability is 63%. \( m_1 \) is the shape factor of the Weibull distribution which describes the magnitude of scatter, or the slope of the Weibull distribution.

Wallin (29), demonstrated based on 2 statistical models for cleavage fracture that the value of \( m_1 \) is 4, although comparison with experimental data showed \( m_1 \) varied between 2 and 10, and was skewed such that when describing tests the value was usually above 4. A limiting minimal value (\( K_{\text{min}} \)) of \( K_i \), at which cleavage fracture is impossible (in that cracks cannot propagate further than one grain), was therefore included in the distribution, such that:

\[ P_f = 1 - \exp \left[ - \left( \frac{K_i - K_{\text{min}}}{K_0 - K_{\text{min}}} \right)^{m_1} \right] \]  \tag{29}

The range of \( K_{\text{min}} \) is 10-20MPa\( \sqrt{\text{m}} \), concluded Wallin (29). Figure 2-16 shows convergence on the proven shape factor \( m_1=4 \) on as the number of tests increased using \( K_{\text{min}}=20\text{MPa}\sqrt{\text{m}} \), as shown in Figure 2-16.
The Wallin distribution described above (29) was shown to describe the scatter in cleavage fracture toughness data and can be used to predict the probability of failure for a steel under an applied stress intensity factor. This distribution is advocated ASTM 1921 (16) which formalises the calculation of the normalising stress intensity factor, $K_0$, used in equation 29. For a number of tests, $N$, at a single temperature:

$$K_0 = \left[ \frac{1}{N} \sum_{i=1}^{N} \left( \frac{K_{JC(i)} - K_{min}}{\Delta K} \right)^4 \right]^{1/4} + K_{min} \quad \text{(MPa√m)} \tag{30}$$

This value corresponds to a cumulative failure probability of 63%. The median failure probability value ($K_{Jcmed}$) is calculated by:

$$K_{Jcmed} = K_{min} + (K_0 - K_{min})[\ln(2)]^{1/4} \quad \text{(MPa√m)} \tag{31}$$

Using this value, it is possible to draw a design curve based on a programme of fracture toughness tests, which allows an engineer to estimate the fracture toughness of the material in question at a given temperature. This curve is described by:
\[ K_{J\text{med}} = 30 + 70 \exp[0.019(T - T_0)] \text{ (MPa\(\sqrt{m}\))} \]  \[32\]

where \( T \) is the test temperature and \( T_0 \) is the reference temperature used as the DBTT (see section 2.1.3), at which \( K_{J\text{med}} = 100\text{MPa}\sqrt{m} \), as shown in Figure 2-17, which also includes the 5% and 95% tolerance bounds of failure probabilities.

\[ 
\]

Figure 2-17: A533b Master Curve (16)

ASTM 1921 (16) further expands on these calculations allowing for data to be size corrected to the standard specimen thickness of \( B=25\text{mm} \), and for different failure probability percentile curves to be included. Furthermore, the document includes details as to how to incorporate several test temperatures into the analysis. At least 6 tests at each temperature should be included for the master curve approach to be valid. R6 advocates the use of the 5\textsuperscript{th} percentile failure probability as the fracture toughness value.
$K_{mat}$ used in $K_r$ calculations which is calculated using equation 33. For other fracture toughness percentile values 5% in equation 33 may be replaced with the required percentile failure probability.

$$K_{mat5\%} = 20 + \left[ \ln \left( \frac{1}{1-5\%} \right) \right]^{1/4} \left( 11 + 77 \left[ 0.019(T - T_0) \right] \right) \text{ (MPa√m)} \quad [33]$$

### 2.4.3. Warm pre-stressing

R6 section III.10 details methodologies for calculating a fracture toughness benefit which is achieved through a warm pre-stressing process. This involves loading a defect above the DBBT to a stress intensity factor higher than the cleavage fracture toughness of the material at the temperature of interest. This is physically explained through the combination of the increase in yield strength through temperature reduction alongside the creation of a compressive residual stress field at the crack tip. There are three main pathways to this warm pre-stress benefit shown schematically in Figure 2-18. The left figure shows the pathways to warm pre-stress benefits, load-cool-fracture (LCF), at the top offers the most benefit, and load-unload-cool-fracture (LUCF), is the least beneficial. This is demonstrated in the right figure which shows how the increased fracture toughness following warm pre-stressing ($K_t$), reduces as partial unloading occurs following the initial warm pre-stress, where $\kappa$ is a measure of the amount of unloading occurring, so $\kappa=1$ is an LUCF cycle, and $\kappa=0$ is the LCF pathway, where no unloading occurs. The $\gamma$ transitions refer to where the unloading transition takes place.
R6 recommends using equation 34 to calculate the benefit to fracture toughness, which interpolates between the LCF and LUCF benefits.

$$K_f = K_2 + \sqrt{K_{mat}\Delta K_u} + 0.15 K_{mat} \text{ (MPa\sqrt{m})}$$  \[34\]

where $\Delta K_u = K_1 - K_2$, which for the LCF cycle=0, as $K_1=K_2$. $K_{mat}$ is the original, uncorrected material cleavage fracture toughness at the temperature of interest. (30) also demonstrated a reduction in tensile residual stresses from the WPS process, but R6 explicitly refutes the use of any claim of the combination of a reduction in residual stress, and an increase in toughness, caused by the warm pre-stressing process.
2.4.4. Local Approaches to Fracture

Several methods exist to model cleavage fracture at a microscopic level and how these microscopic effects contribute to the probability of global fracture. These approaches can include statistical methods which are used to calculate the probability of the necessary conditions for cleavage fracture occurring under a prescribed principal stress.

The Beremin (31) group suggested in 1983 that probability of cleavage fracture occurring \( (P_f) \) could be modelled by a Weibull distribution, using the following equation:

\[
P_f = 1 - \exp \left[ - \left( \frac{\sigma_w}{\sigma_u} \right)^{m_B} \right]
\]  \[35\]

where \( m_B \) (subscript \( B \) used to differentiate between this and \( m \) parameter described in section 2.4.2) describes the shape of the distribution, taken be 22 for the Beremin study (31), and \( \sigma_u \) is a material property used as a scaling factor. The research introduced a parameter, known as the “Weibull Stress” \( (\sigma_w) \), which is calculated through equation 36:

\[
\sigma_w = m_B \sqrt{\sum_j \left( \frac{\sigma_j}{\sigma_1} \right)^{m_B} \frac{V_j}{V_0}} \quad \text{(MPa)} \]  \[36\]

The idea behind this distribution is that the volume around the crack tip \( (V_0) \), (which must be a region small enough, \( \sim10 \) grains, that its stress gradient is small enough that it may be considered as quasi-homogenous), is divided into \( j \) unit cells with volumes \( (V_i) \), which experience the maximum stress \( (\sigma_j) \). The approach allows for the size effect of ferrite grains or to be accounted for in fracture toughness predictions. The above expressions are derived from the probability of a microcrack, sufficiently long enough to propagate into the neighbouring grain, existing in a unit volume ahead of the macroscopic crack tip. This enables the probability of cleavage fracture to be predicted under a prescribed global stress.
2.5. Crack Tip Constraint

As was discussed when explaining plane stress and plane strain conditions, defects in stressed structures alter the distribution of the stresses ahead of the crack tip, in that the stress triaxiality is altered. R6 (26), describes constraint as follows: “(A) material’s resistance to fracture is increased when specimens with shallow cracks, or specimens in tension, are tested. These conditions lead to lower hydrostatic stresses (reduced triaxiality) at the crack tip, referred to as lower constraint (26).” It has been widely observed that this loss of constraint, which is affected by geometry and loading, as opposed to being a material property, has significant effects on the fracture resistance of certain components, especially where large-scale-yielding is observed. Due to this, two-parameter fracture toughness approaches have been developed, which quantify constraint for structural integrity approaches.

2.5.1. T Stress

Initial consideration of two parameter fracture mechanics focussed on linear elastic cases. Considering the Westergaard functions (equations 7) in a cylindrical coordinate system it was possible to add terms to the equations (32), such that the stress functions could be rewritten as below, where δ_{ij} is the Kroneckor delta tensor, and T is known as the T stress:

\[ \sigma_{ij} = \frac{K_I}{\sqrt{2\pi r}} f(\theta) + T \delta_{1i} \delta_{1j} \]  

The T stress parameter represents stress acting parallel to the crack flank and is used to quantify constraint. Its value (due its linear elastic limitations) is proportional to load and affected by crack and specimen geometry. Thus, a parameter was introduced to relate all of the above, known as \( \beta_T \) (33). The \( \beta_T \) parameter is used to modify \( K_I \) based
on the level of constraint and was tabulated by Sherry et al (34) for various standard FTT specimen geometries.

\[
\beta_T = \frac{T\sqrt{\pi a}}{K_I} = \frac{T}{L_T\sigma_y}
\]

For SENB test specimens the \( \beta_T \) value provided in R6 section 5.4 (26) is given as a function of the crack depth to wall thickness ratio, \( a/W \):

\[
\beta_T = -0.9893 + 4.874(a/W) - 9.6956(a/W)^2 + 11.434(a/W)^3
- 5.9061(a/W)^4
\]

2.5.2. The Q Parameter

O’Dowd and Shih first postulated the Q parameter (35). They suggested that the HRR fields as described in equation 19 were insufficient to describe the stress field ahead of a crack tip where constraint is low. The Q parameter was added, as shown below, which incorporates elastic-plastic effects and due to this is no longer limited to act parallel to the crack flank:

\[
\sigma_{ij}(r\sigma_0/J, \theta) = \sigma_{ij}^{SSY}(r\sigma_0/J, \theta) + Q\sigma_0\delta_{ij}
\]

This parameter is accepted in the R6 guidelines (26) as also valid for structural analysis of low constraint components, alongside the T-stress analysis. As Q is analogous to the T-stress it follows that there is an analogous \( \beta_Q \). Again, there are R6 advocated compendium solutions for the value of \( \beta_Q \) (36), the applicability of which are discussed in section 7.1 of this thesis.
The calculation of Q is carried out numerically through creating a boundary layer model loaded to a prescribed value of J or K. A boundary layer model produces an idealised SSY field, where theoretically there is no loss of constraint, and Q=0, because the plastic zone ahead of the crack is very small compared to the other pertinent dimensions; the boundary layer model simulates a semi-infinite crack in an infinite body. This small scale yielding stress field is then compared with the stress field at a normalised distance \( r_0/J = 2 \), for the finite-geometry model under consideration, where \( r \) is the distance from the crack tip, and \( \sigma_0 \) is a normalising reference stress, such as the 0.2% proof stress of the material. The difference in the normalised stress is the Q parameter, at that increment of load (quantified by the crack driving force, J or K). The value of Q may then be plotted against J, to produce a loading curve as shown schematically in Figure 2-19.

\[
\beta_Q = \frac{Q}{L_r}
\]

Figure 2-19: J-Q loading curve methodology schematic (3)
Following calculation of the J-Q loading curves it is possible to populate them with failure data ($J_c$ values), such that several failure points are drawn in the J-Q space. The purpose of the research presented in this thesis was to assess the viability of a material specific failure curve drawn through these failure points for predicting failure for the load histories displayed in Table 1. Such a curve is shown schematically in Figure 2-20.

To employ such a curve in practice requires an engineer to conduct FEA of the cracked structure, simulating as accurately as possible the load history in the structure. The engineer could then calculate a J-Q loading curve for the defect, and the J value where this intersects the fracture toughness locus would be the new, constraint based $J_c$ value, or the predicted load at failure.

![Figure 2-20: J-Q failure curve schematic](image)

R6 predominantly uses $K_i$ values to quantify toughness and crack driving force, so the recommendations for the derivation of a failure curve such as that shown in Figure 2-20 is as follows, where $K_c^{\text{mat}}$ is the constraint corrected fracture toughness:

$$K_c^{\text{mat}} = f(L_r)[1 + \alpha(-\beta L_r)^m]$$  \hspace{1cm} [42]
2.6. Residual Stress

One aspect of the work presented in this thesis is focussed on the effects of residual stress on the postulated 2-parameter J-Q fracture toughness curve. As explained in section 1, residual stresses are those that exist when all exterior loading has been removed and are self-equilibrating; the net force they exert $= 0$. They are introduced through contraction of material that has been heated through welding, which cannot return to its original position, inducing significant stress on the surrounding material (37). Other mechanisms for the introduction of residual stress of interest to RPV structural integrity assessments include microstructural phase changes or plastic deformation introduced through installation or manufacturing errors.

2.6.1. Residual Stresses in R6

The FAD, as presented in section 2.4.1, can be used to assess structures where residual stresses are present (26). There are three levels of residual stress profiling, the simplest of which suggests they should be modelled as equal to the yield stress of the material (Level 1), in each of the principal directions. This leads to a simple vertical shift in the loading line on the failure assessment diagram. This approach is both overly simple and conservative.

For more accurate assessments weld induced residual stress must be separated into its various components. Level 2 residual stress profiling in R6 entails utilisation of compendia where the residual stress profiles of standard weld geometries have been compiled. These stress profiles detail the longitudinal and transverse stress in the through thickness and transverse directions in relation to the weld. The procedures involve using various equations to characterise the size of the yielded zone, and material constants at the weld (defined through parameters such as weld travel speed, process efficiency and material thermal properties). These parameters are then used to characterise the stress profile in the directions described above to be included in the $K_r$ calculation for the assessment point included in the FAD (26), an example of a standardised residual stress distribution is shown in Figure 2-21.
Both Level 1 and Level 2 assessments may prove excessively conservative for a suitable safety factor to be achieved in the structural assessment (26). When this is the case more detailed methods of residual stress measurement are necessary, which are achieved through finite element analysis assessment or experimental measurement such as through neutron diffraction, described in section 2.6.2.
2.6.2. Residual Stress Measurement Through Neutron Diffraction

Neutron diffraction uses radiation in a similar way to X-rays to measure stress in a material but is preferable as the have wavelengths used are comparable to crystal lattice spacing in metals (38). This means that residual stresses can be measured at depths much greater than surface layer measurements, which can penetrate to 5μm and are available through X-rays (39).

![Figure 2-22: Neutron diffraction arrays (40)](image)

A schematic of the two methods of neutron diffraction stress measurement techniques is shown in Figure 2-22. The vector in which the strain is measured is shown as Q. The major issues with the measurement of residual stress using neutron diffraction is that the equipment only exists in certain facilities, so is therefore expensive and time consuming for experimentation. Furthermore, as with all residual stress measurements there are likely to be some inaccuracies. The measurement techniques discussed above equate the strains measured to stress using Hooke’s law, which is only applicable in the elastic regime. Therefore, inaccuracies will exist with any near yield or post yield stresses. The nature of thermal residual stress involves inelastic strains caused through phase transformations and thermal contraction, so the use of Hooke’s law may not be sufficient for the stress/strain relationship. With neutron diffraction R6 concedes that
issues with initial lattice spacing parameters and intergranular stresses may also lead to experimental inaccuracies (26).

2.6.3. Mechanical Residual Stress Introduction

Several works have demonstrated how mechanical compression could be used to induce residual stresses into fracture toughness specimens. An in-plane compression method suggested by James et al. (41) involved the longitudinal compression of a deeply, circular notched SENB specimen, showed significant tensile residual stresses allowing for detailed investigation of the effective fracture toughness of specimens with residual stress fields.

The out-of plane compression, or double-side punching method first suggested by Mahmoudi et al. (42), which was utilised by Hurlston (3), involved the compression of a circular area ahead of the crack on the free surface of an SENB specimen. In Hurlston’s (3) work FE simulations were carried out to assess the residual stress fields introduced through this method. These involved the displacement of a circular node set which was positioned at a prescribed x and y distance from the crack (with the crack front parallel to the z axis), as shown schematically in Figure 2-23. An extensive parametric study involving modification of y, x and r was carried out to simulate residual stress distributions that would offer the maximum tensile residual stresses in the ligament ahead of a pre-machined notch. The parametric study concluded that the dimensions to offer the most significant tensile residual stress were x=2mm, y=10mm, r=5mm, for notch depths \( a_n/W \) =0.1 and 0.3. Figure 2-23 also shows how the numerically calculated residual stress fields from this project were validated using the neutron diffraction method described in section 2.6.2.
2.6.4. J in the Presence of Residual Stresses

The calculation of J using equation 18, using the FE method, where initial inelastic or thermal strains are in the crack vicinity prior to loading or crack extension, usually results in path dependence issues. One aspect of these inaccuracies is the difficulty involved in defining the stress, strain and strain energy density fields accurately close to the crack tip (43), necessary for the J calculation. Another aspect of the problem with prior inelastic strains being incorporated into the model is that the path independence of the J integral may rely on the stresses at the crack increasing in proportion with the loading, where “the different components of the stress tensor vary in constant proportion to one another and change monotonically with increase of the external load” (44). This can be prevented through the introduction of residual stresses and render the accuracy of the standard J integral questionable. Works by Lei et al (45), Lei (44) and Beardsmore (43) offered modifications of equation 18 to enable the computation of the J Integral where the crack tip loading is non proportional and where strains are present before cracking.
The earlier work by Lei et al (45) offers modifications to the J calculation, allowing users to compensate for residual stresses (strains). This adjustment splits the strain tensor $\varepsilon_{ij}$ into an initial (residual) strain $\varepsilon_{ij}^0$ and an applied (mechanical) strain $\varepsilon_{ij}^m$. $\varepsilon_{ij}^0$ remains as constant through subsequent deformation and $\varepsilon_{ij}^m$ increases with stress as defined by the material’s constitutive law. This allows the modification of the J integral to be defined to include initial strain, and remain path independent, where $w$ is the mechanical strain energy density (see below), and $A$ is the area surrounded by the contour, $\Gamma$ (45):

$$W = \int_0^{\varepsilon_{ij}^m} \sigma_{ij} d\varepsilon_{ij}^m \quad [43]$$

$$J = \int_C \left( W \delta_{1j} - \sigma_{ij} \frac{\partial u_j}{\partial x_1} \right) n_i \, ds + \int_A \left( \sigma_{ij} \frac{\partial \varepsilon_{ij}^0}{\partial x_1} \right) \, dA$$

Equation 43 does not however, account for non-proportional loading effects. The non-proportional loading may lead path dependence using equation 43, rendering them less accurate for calculations of J. The later works by Beardsmore (43) and Lei (44) offer similar solutions for this issue. Their modifications remove the proportional assumptions through separation of the strain energy density ($W$) term into the initial plastic strain energy ($W_{p0}$) density and the mechanical ($W_m$). Further separation of the strain into its constituent parts is also considered so the strains are separated into $\varepsilon_{ij}^e$ (applied elastic), $\varepsilon_{ij}^p$ (applied plastic), $\varepsilon_{ij}^{p0}$ (initial plastic); and $\varepsilon_{ij}^l$ (initial thermal, where applicable). This leads to:

$$\varepsilon_{ij}^* = \varepsilon_{ij}^{p0} + \varepsilon_{ij}^l$$

$$\varepsilon_{ij}^m = \varepsilon_{ij}^p + \varepsilon_{ij}^e - \varepsilon_{ij}^{p0} \quad [44]$$

Furthermore the contour region is again expanded over a longer path ($C=\Gamma+C_a+C+C_0$) to allow for integration over an area as can be seen in Figure 2-24, where $q$ is a vector field which $=\eta$ (a unit vector parallel with the crack tip) on $\Gamma$ and $=0$ on $C$ (43), $m_i$ is a unit vector parallel to $C_0$. 
This equation was expanded by Beardsmore (43), using Green’s theorem, for use in a post-processing code known as JEDI (J-Equivalent Domain Integral), to be used with the Abaqus FE software (22):

\[
J = \int_C W^m \delta_{ik} - \sigma_{ij} \left( \frac{\partial u_j}{\partial x_k} \right) m_k q_k \, ds
\]

Both Lei and Beardsmore proved that the above equations are suitable to obtain path independent J integral values for non-proportional loading conditions, (43), (44). An example of the J results for an SENB specimen (see Figure 2-10) with crack inserted into a residual stress field, calculated using JEDI (43) (equation 46), and Abaqus (22) (equation 43), is shown below. As can be seen there is convergence much closer to the crack tip on a J value, demonstrating path independence at a shorter distance from the crack tip. Convergence begins several contours ahead of the crack tip because of the unrealistically high stresses at the crack tip due to the model being based on a sharp notch (no blunting), causing singularity issues with the calculations.
2.6.5. J-Q Behaviour in the Presence of Residual Stresses

Hurlston (3) previously considered the effects of residual stresses on J-Q loading curves for A533b RPV steel, and suggested a 2-parameter fracture toughness curve, including residual stress, which is displayed in Figure 2-26. To expand on these results the work presented in this thesis extends into much lower constraint geometries, and therefore lower values of \( Q \), and also considers the effects of global plasticity to simulate that which may be seen around a weld (see section 2.7). The work presented in (3) formed a starting point for the design of the research presented in this thesis.

Hurlston’s (3) work discussed the effects of inducing residual stresses achieved through a local out of plane compression, as described in section 2.6.3. The intention of the work presented in this thesis was to consider if a similar failure curve to that presented in Figure 2-26 could be derived using results from a similar test programme with a much larger variation in initial constraint levels. The failure curve derived in Figure 2-26 was derived not using the R6 (26) recommended constraint toughness correction procedure as described in section 2.5.2, but a solution based on the RKR (after Ritchie, Knott and Rice (46)) local approach. The simple RKR procedure predicts failure when stresses at a critical distance ahead of the crack \( r_c \), generally considered arbitrary,
provide the fracture process zone is enclosed) reach a critical value ($\sigma_f$-considered a material property).

To obtain the curves shown in Figure 2-26 the RKR procedure was followed and the critical stress was chosen as the mean failure stress of all test cases ($P_f=0.5$, $P_f=0.05$, 0.95 are the bounding cases), with $r_c$ taken as 0.2mm, shown to fully enclose the fracture process zone (3). Beardsmore’s (43) methodology was then followed which showed that correcting the material fracture toughness to constraint based toughness could be achieved by using equation 47, which equates the RKR stress at fracture to the HRR opening mode stress as defined in equation 19, and incorporating into equation 40:

$$J_{mat}^c = J_{mat} (1 - Q \sigma_0 / \sigma_f)^{n+1}$$  \[47\]

![Figure 2-26: Fracture toughness curve in the J-Q space for two Cracks Depths (a/w=0.22, 0.42) with and without residual stresses (RS). P_f denotes percentile of amount of experimentally tested failures (3)](image)

As can be seen in Figure 2-26 the RKR/Beardsmore approach for constraint based toughness corrections predicts well failure in Hurlston’s case, and comparison with this
failure curve at that derived from this project is made is section 7.1 of this thesis. There are interesting points for discussion raised by the J-Q behaviour presented in Figure 2-26. There can be seen a reduction of Q for the high constraint geometry following the introduction of residual stress, whereas for the shallower crack, Q is increased following the introduction of a similar stress field. It is apparent that there are some combinations of geometrical and loading conditions whereby the residual stress no longer increases Q, and constraint, but reduces it. Hurlston (3) postulated that the mechanically introduced residual stress field, as described in 2.6.3, served to increase stress triaxiality (and therefore constraint) in the a/W=0.22 specimens, but actually reduce it in the higher constraint geometries. A detailed analysis of this hypothesis is included in the results of this thesis (see section 7.3) and show it to be correct. This effect therefore means that constraint is reduced in the deeper cracked geometry and therefore could be considered to have a positive structural effect.

This observation however requires significant consideration. The J values used in this plot are extracted from FE simulations using the methodology described in section 2.6.4, and specifically equation 43. They show an increase in average values of J, for both geometries when the residual stress field is introduced. This however includes a residual J value, in that the calculations include a crack driving force from the secondary stresses acting alone before any external primary load is exerted on the structure. Therefore, with the reduction in Q and increase in J at failure it could be said that the structure may be able to withstand higher loads where residual stresses are present. This however is incorrect, as shown in Figure 2-27.
What can be seen from the actual experimental data is that the ability of the specimens to withstand load is significantly reduced where a residual stress field is presented. This comparison between load and J values calculated numerically using finite element analyses, including residual stresses, are discussed in detail in section 5.4. But what can be seen from Figure 2-26 is that the indication is that there is a failure curve in the J-Q space, for medium and high constraint geometries, but its applicability to structural integrity assessments is not intuitive, in that careful consideration needs to be given as to how the actual resistance load is interpreted through the J values used in the calculations.

2.7. Plasticity in Welds

It is known that the welding process introduces plastic strains in welds. It is however difficult to quantify the amount of plasticity a welding process may introduce to the heat affected zone (HAZ). The process of welding causes irregular expansion of the material over the heat affected zone, which also shrinks irregularly during cooling. This alongside microstructural changes in the material makes it difficult to extract specimens of material, which could be subjected to tensile tests, as the distribution of the plastic strains would also be irregular. Therefore, this section analyses how welds have been
examined to estimate the amount of plasticity in the HAZ, and how testing has been carried out to better understand how these levels of plasticity have affected the material behaviour.

2.7.1. Finite Element Modelling

R6 (26) advocates in its Level 3 residual stress assessment the use of the FE method to model welds and their effects on parameters that would influence the structural integrity of NPP components. One such project (47) has demonstrated how the equivalent plastic strain in the heat-affected zone (HAZ) around an austenitic repair weld, on a circumferential girth weld in a boiler spine pipe, could be estimated. The simulation considered a 12-pass repair weld through the addition of elements to simulate the weld beads, as shown in Figure 2-28.

The simulation was carried out using the FE software Abaqus (22), with appropriate material models and annealing algorithms applied during the passes (47). The elements were applied in such a fashion that they existed in the original mesh, without any material properties. As each pass was added the thermal and material models were applied to the appropriate elements such that the surrounding elements were subjected to the relevant stresses. The conclusions from this report found that within the parent metal structure at separate points the equivalent plastic strain accumulated was 3.8%, 5.2% and 3.6% (at various locations within the weld). This simulation considered austenitic parent material only but the methodology is applicable to ferritic steels, and whilst there are variations in yield stress and hardening behaviour...
in austenitic steels the simulations give an estimate as to the amount of plastic strains an RPV weld may be subjected to.

### 2.7.2. Hardness Mapping

Brayshaw et al. (48) described the changes in hardness to the materials used in a dissimilar metal transition weld. Through microindentation of a sample section of the joint it was possible to create a ‘hardness map’ of the weld, which is shown in Figure 2-29. The hardness map shown uses the unit HV1, which denotes Vicker’s hardness (49). This method involves indenting a pyramidal diamond into the material to a specific load and measuring the size of the indent area, such that a value in the units of HV1 is obtained. Mapping software was then used to create the plot shown in Figure 2-29.

![Hardness Map](image)

**Figure 2-29: Hardness map of dissimilar metal weld containing ferritic (SA508-4N), weld (Alloy 82) and austenitic (316LN) material (48)**

It was observed that in the heat affected zone of the ferritic steel there are significant increases in hardness, caused by a reduction in grain size (also observed on the austenitic side of the weld). The increase in hardness was shown to increase from an average of 241HV1 in the parent ferritic material, to an optimum of 343HV1 in the HAZ.

Work has been carried out on a wide range of test data to offer an empirical relationship between an increase in hardness to an increase in yield stress, for both austenitic and ferritic steels (50) (51). Hart (52) expanded these relationships by
considering that the welding process would affect the change in hardness, and therefore the HAZ of a weld would not have the same relationship between hardness and yield stress so he suggested a new HAZ based relationship. The evidence collected in his study is shown on the left of Figure 2-30. Applying his relationship to the stress-strain curve for SA508, it can be seen that there is a good estimate of the yield stress of the parent material. This demonstrates confidence in the use of the relationship to estimate the yield stress of the HAZ material using his relationship, which is applied to stress stain curve (provided by the author of (48)) for the parent, ferritic material in the hardness map shown in Figure 2-29. Converting the change in yield stress to plastic strain, as shown in Figure 2-30, a value of 4.25% plastic strain is suggested to have been shown in ferritic HAZ of the dissimilar metal weld.

![Figure 2-30: Empirical yield stress to hardness conversions and stress strain curve for SA508 (52)](image)

### 2.7.3. Effects of Plastic Strain on Tensile Properties of Metals

Standard plasticity theory suggests that after a metal reaches its elastic limit, dislocations within the crystal lattice are mobilised and the material is no longer governed by Hooke’s law, where stress is proportional to strain, and the ratio with the ratio being the Young’s modulus of the material. Beyond this limit the material is irreversibly transformed and displays new tensile behaviour. The standard prediction for
This new material behaviour is that the material will begin at the prescribed level of plastic strain for the as received material, follow the same modulus line, and continue elastically to a new yield stress, where it will follow the same hardening curve as the original material, as shown schematically in Figure 2-31.

![Stress-Strain Diagram](image)

Figure 2-31: Standard plasticity theory effect on tensile behaviour of metals

It has been observed in the literature that this approximation is not necessarily accurate to describe material which has seen significant amounts of plasticity, specifically in that a reduction in Young’s modulus could be effected in material which has seen significant plastic strains. Benito et al. (53) observed through standard tensile testing of pure iron a reduction in the Young’s modulus from 210 to 196 GPa at 6% plastic strain, and then a slight increase in modulus at higher levels of plastic strain. They attribute this reduction to the bowing of dislocations which increases dislocation density and corroborate this experimentally as shown in Figure 2-32, where the dislocation density values were calculated from the results of transmission electron microscopy (TEM). At a level of 8% strain the dislocations become pinned reducing the length of the dislocation lines and therefore dislocation density, subsequently
demonstrating a slight increase in modulus which is no longer affected by increasing plastic strain.

![Dislocation density effect on Young's modulus of raw iron](image)

**Figure 2-32:** Dislocation density effect on Young’s modulus of raw iron (53)

Yang et al. (54) demonstrated empirically a more significant reduction in modulus in SPCE carbon steel sheet metal, through both macroscopic tensile testing and nano-indentation of the material with a spherical indenter. The nano-indentation process allowed modulus measurements to be taken across a grain of the material, and they showed a significant reduction in modulus at the grain boundaries as shown in Figure 2-33.
This reduction in modulus at the grain boundaries was attributed to the increase in movable dislocations caused by the pre-straining process accompanying dislocation pile up, which was used to explain the significant reduction in modulus observed with increasing plasticity in the tensile tests, shown in Figure 2-34. Both papers offer strong arguments for a reduction in modulus caused by an increase in dislocations and dislocation movement caused through pre-straining.
A recent paper (55) further elaborated on this work by comparing results of various methods of modulus measurement including measurement of resonant frequencies and ultrasonic pulses in the material, alongside mechanical tensile testing. This work corroborates the previous findings in that modulus is reduced through pre-straining, and that this plateaus at a certain amount of plasticity. This work reports this amount of plastic strain as 2%, significantly lower than the ~5% reported in (53). This work also examined non-linearity in the nominally elastic region of stress-strain curves for various steels in both as received and pre-strained conditions, which is apparent in the unloading portion of the stress-strain data output from the pre-straining process, and then in the subsequent reloading.

They suggest that calculating a value of modulus, from a chord between points in the non-linear region, during unloading, offers a consistently accurate prediction of the modulus when re-loading. This approach however raises questions because a modulus value, for use in modelling and simulation, should be a representation of the elastic behaviour in the material, including this non-linearity is actually demonstrating plasticity at 0MPa yield stress. This is discussed in the paper (55) however and shows good prediction in terms of modelling of situations where pre-strained material shows spring back, in this case for steels used in the automotive industry.
The work presented in (53), (54) and (55) shows empirical evidence that introducing plastic strains to metals causes a reduction in at least the effective Young’s modulus, with physical explanations as to why this occurs. It is clear that this is an important effect, and this becomes apparent when trying to model the behaviour of materials which have seen significant plasticity. This is pertinent for the work presented in this thesis and is discussed in detail when presenting the material properties used in the modelling of the pre-strained specimens fracture toughness tested and presented in section 4.1.

2.7.4. Effects of Plastic Strain on Cleavage Fracture Toughness of Steels

There is limited information on the effects of global plastic strains on cleavage fracture toughness. Much of the literature available is based on upper-shelf fracture toughness however and widely reports a reduction in toughness with increasing pre-strain. One paper however (56) summarises empirical data and describes a model based around the including a critical strain at fracture criterion in the HRR fields described in section 2.3.2, and then incorporating standard plasticity theory assumptions as shown in Figure 2-31, to estimate a reduction in toughness.

The standard plasticity assumptions include an increase in yield stress, a reduction in ductility (in that the strain at the ultimate tensile strength is lower) and a parallel modulus. Using these assumptions, the model recalculates the critical strain and therefore toughness. Figure 2-35 shows the empirical data and results from the suggested model for reduction in toughness with increasing pre-strain. Unfortunately, no direct comparison is made with the test data and the model results, due to insufficient information, although the work claims the output from the model offers the same trends as the data (56).
Lewis and Truman (57) considered the possibility of using local approaches to predict the effects of pre-straining on fracture toughness. They investigated using a Beremin (31) type model, which incorporated a linear correlation between defect nucleation and strain (58) in the prediction of failure of high constraint CT specimens with varying levels of pre-strain. The work concluded that it was possible to qualitatively predict a reduction in mean failure load of pre-strained material, although further calibration was required to formulate the exact reduction with increasing pre-strain. The best results of the local approach investigation are shown in Figure 2-36, where AR denotes as received. The results are promising and are compared with the tests presented in section 7.2 of this thesis.
2.7.5. Effects of Plastic Strain on J-Q Loading Curves

Figure 2-37 displays a complete set of results for a set of preliminary scoping calculations carried out as preparation for the work presented in this thesis. Fracture toughness testing simulations were carried out in a similar manner to those described in section 4.2. The results presented in Figure 2-37 are J-Q loading curves for 16 different scenarios: with and without a mechanically introduced residual stress field similar to that described in section 2.6.3, (denoted N/NR) respectively), at 0, 2, 5, and 10% equivalent plastic strain field introduced through a hardening field option in the commercial finite element software Abaqus, which simply shifts the stress strain curve (as shown in Figure 2-31) used to describe the tensile data used in the model (denoted H0, H2, H5, H10 respectively). For this analysis 2 crack depths were used: a/W =0.1 and 0.5 (denoted 01, 05 respectively).
Figure 2-37 displays two clear trends. The first is that, in general, the addition of a plastic strain field serves to increase Q, as can be seen moving from the blue to yellow curves. Secondly it can be see that in the high constraint geometry (a/W=0.5) the introduction of a residual stress field causes a decrease in Q. The converse is apparent for the low constraint conditions, in that Q increases with the addition of a residual stress field.

The set of results were published in (59), and warrant some discussion. The method of calculating Q for the plots shown in Figure 2-37 involves changing the material properties of the boundary layer model to correspond with the finite geometry specimens, such that an equivalent plastic strain field is applied to the SSY field, when calculating Q for the pre-strained material. This is discussed further in 7.2 and is an important output of work presented in this thesis, as it is shown after the material has seen significant plastic strains, it is important to consider the reference SSY field as that from a new material, which displays the tensile behaviour of the plastically strained material. For the high constraint geometry (a/W=0.5), increasing the level of plastic strain shows little change in Q through loading, demonstrating that SSY conditions are prevailing. Where residual stress is concerned, there is, as was seen in the Hurlston’s (3)
work shown in Figure 2-26, a switch between a decrease in Q at high constraint, to an increase at low. This again demonstrates a converse influence on stress triaxiality between the 2 scenarios, which is discussed at length in section 7.3 of this thesis.

2.8. Outcomes of Literature Review

The purpose of this literature review was to justify the methodologies employed to meet the objectives of this thesis, which were to investigate the viability of a material specific, constraint based, 2-parameter cleavage fracture toughness curve in the J-Q space, that would predict failure under various load history effects such as residual stresses and plastic strains, that would be comparable to those a flaw in an RPV weld may be subjected to.

The relevant microstructural details were included, including the structure of ferritic steels and specifics about the material in question, namely JSW A533b Class 1. There are details of the micromechanisms of cleavage fracture, alongside why this may be a risk in ferritic steels which have been subjected to long term neutron irradiation, when they are at lower temperatures than the design operating temperature of the RPV. This justifies the low temperature (-140°C) at which the fracture toughness test programme was carried out (detailed in section 3) which has been shown to be a temperature which can guarantee cleavage fracture in the material in question (3).

Following the microstructural details there is an explanation of the J-Q methodology for quantifying constraint, with details of the supporting fracture mechanics principles that have enabled the failure curve in question to be hypothesised, including the methods of calculating the J integral such that it remains path independent in the presence of residual stresses. Furthermore, there are details as to the current status of a J-Q failure curve considering residual stresses, which is used a reference for the calculations presented in this thesis. There are details as to how introducing residual stress fields through mechanical methods such as double side-punching are valid simulations of welds in RPVs.

Finally, there are details as to the amount of plastic strains welding can introduce into ferritic steels in RPV structures, with one option described involving converting the
change in hardness in the material to a change in yield stress which equates to 4.25% plastic strain in a dissimilar metal weld. There is also description of observations that significant amounts of plasticity may affect the material differently than standard plasticity theory suggests, most noticeably through a reduction in Young’s modulus through increasing dislocation movement. This has been shown to lead to non-linearity in the elastic region of stress-strain curves of pre-strained materials. These observations justified the necessity of the tensile test programme detailed in section 3.4 to compare the as received material to that with significant plastic strains.

2.9. Gaps in the Knowledge

Several key gaps not reported in the literature are evident, with regards to the overall aim of this thesis as described in section 1.1:

- The material specific J-Q failure curve methodology has not been tested in its ability to predict failure of defects in structures with complex load histories.
- There is limited understanding of the effects of significant plastic strains on the cleavage fracture toughness of ferritic RPV steels, only that there is a general trend to a reduction with increasing pre-strain.
- There is no information of how the postulated reduction in cleavage fracture toughness in pre-strained RPV steels is affected through reduced constraint, and no comparison as to how this effect compares with as received material.
- Non-linearity has been observed in the usually considered ‘elastic’ region of the stress-strain curve of pre-strained material, but no consideration of this behaviour has been applied in the calculation of the fracture toughness of steels.
- There is limited information as to the effects of constraint on defects in a residual stress field, and no information at such low constraint that there is a significant increase in fracture toughness.

It is these gaps in the knowledge that are addressed in this thesis, and demonstrate the novelty of the work presented in the following sections.
3. Experimental Methods

To assess the viability of a material specific fracture toughness curve in the J-Q space, it was necessary to collect failure data from fracture toughness tests under the various different initial conditions shown in Table 1. The project objectives were to consider the interaction between the residual stress and plastic strain fields a defect in an RPV weld may be subjected to. Furthermore, as the J-Q failure curve is postulated to provide a constraint based fracture toughness for the specific material, it was necessary to consider varying levels of constraint. This section provides details for the methods used to explore these variables physically.

3.1. Experimental Overview

It was decided that SENB specimens would be used for any fracture toughness testing, and that residual stress introduction would be carried out using the side punching method as described in section 2.6.3. This was for several reasons:

First, using SENB specimens (as opposed to other standard specimen types such as compact tension) allows for varying levels of constraint to be introduced through simply changing the crack depth (a). It has been observed in the literature (3) (60) (59) that varying the crack depth to specimen thickness ratio (a/W) in this type of specimen has significant effects on the J-Q loading curves, and the value of J at which failure occurs (J_c).

Secondly, the side punching method used to introduce residual stress fields to this type of specimen has been shown to accurately imitate residual stress fields (42), (3) defects may be subjected to near RPV welds. The modelling used to simulate these stress fields has been validated using neutron diffraction (3), with details of the methods used for both experiments and FE modelling well documented such that reproduction could be carried out.
Thirdly, as described in section 2.6.3, the output from Hurlston’s project (3) included a J-Q failure curve considering residual stress at different levels of constraint. By reproducing this work, it would be possible to add comparison between the J-Q loading curves calculated for inclusion in this thesis.

It was therefore decided that similar specimens to those used in (3) would be tested to allow the best comparison with this previous work. For control purposes nominal a/W ratios of 0.4 and 0.2 would be tested. As (3) considered similar a/W ratios (actually 0.42 and 0.22) with residual stress it was decided these scenarios did not need to be re-tested; carrying out the control tests would provide suitable validation for the comparison of this project’s results with those included in (3), and justify the use of any additional failure data to add to the failure curve shown in Figure 2-26.

In addition to these crack depths it would be necessary to consider a very low constraint option. This would enable the results to form a failure curve, as opposed to just two distinct sets of results, which would be obtained with only 2 crack depths. It has been observed (60) that with a/W ratios at below 0.1 there are significant increases in the apparent fracture toughness of a similar material. It was therefore decided to choose a crack depth of a/W=0.05 for the very low constraint option. The addition of such low constraint geometries also adds novelty to the work, as these have not been investigated with residual stress fields.

Residual stress, as discussed, was to be introduced through out of plane compression using the double side-punching method, as shown schematically in Figure 2-23. This method would allow residual stress to be introduced to the very low constraint geometries (crack depth a/W=0.05), and would again enable comparison with the results reported in (3). An FE simulation was carried out to investigate how the punches would be positioned, and what level of indent they would apply to the specimens, such that they would subject the cracks to the same residual stress field seen in the previous project. This is detailed in section 3.3.3.

The level of pre-strain the specimens would be subjected to was decided upon based on various considerations. It was necessary to simulate the same amount of plasticity a ferritic weld may experience in an RPV structure. As plastic strain in welds is not easy to measure directly, the literature was examined to estimate what may be a suitable amount of plastic strain the specimens should be subjected to (see section
2.7.2). It was also necessary to see a discernible difference in the J-Q loading curves between the virgin material and those subjected to pre-straining. Preliminary J-Q calculations, displayed in Figure 2-37, showed a large distinction between loading curves with 5% equivalent plastic strain and the virgin material, more so than 2%, and with a less significant increase between the 5% and 10% loading lines. As described in section 2.7.2, 5% plastic strain is a reasonable estimate of the amount a ferritic weld might see (rounded for simplicity and conservatism). Therefore, this amount was decided upon for the experiments.

The above decision-making process enabled a set of test scenarios to be decided upon. These scenarios are shown in Table 3, where an ‘x’ denotes those which were tested.

<table>
<thead>
<tr>
<th></th>
<th>With Residual Stress</th>
<th>Without Residual Stress</th>
</tr>
</thead>
<tbody>
<tr>
<td>a/w=0.05</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0% Equivalent Plastic Strain</td>
<td>x</td>
<td>x</td>
</tr>
<tr>
<td>5% Equivalent Plastic Strain</td>
<td>x</td>
<td>x</td>
</tr>
<tr>
<td>a/w=0.2</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0% Equivalent Plastic Strain</td>
<td></td>
<td>x</td>
</tr>
<tr>
<td>5% Equivalent Plastic Strain</td>
<td></td>
<td>x</td>
</tr>
<tr>
<td>a/w=0.4</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0% Equivalent Plastic Strain</td>
<td></td>
<td>x</td>
</tr>
<tr>
<td>5% Equivalent Plastic Strain</td>
<td></td>
<td>x</td>
</tr>
</tbody>
</table>

Table 3: Experimental scenarios tested
3.2. Specimen Design

The material chosen for this project was A533b, ferritic reactor pressure vessel steel. The microstructure of this material is described in section 2.1.4. This material was chosen due to its availability at the sponsoring company (Wood), and because it had been used in Hurlston’s (3) project, so comparisons could be drawn with the previous results. The availability of material determined machining requirements and the number of tests per scenario (see Table 3) that could be carried out. The design of the specimens and methods used to introduce plastic strains was also decided based on the amount of material available.

3.2.1. SENB Fracture Toughness Specimen Design

SENB specimens were used for the fracture toughness testing phase of this project, the reasons for this are described in section 3.1. This type of specimen has a standardised design in BS7448 (17), and is shown schematically in Figure 2-10. The dimensions of each specimen were to be 240 x 50 x 25mm. The standard (17) specifies an a/W ratio of between 0.45 and 0.55 such that conservative, high constraint fracture toughness can be obtained. As the purpose of this project was to test lower constraint scenarios, it cannot be said that any designs or test procedures were entirely compliant with the standards. All specimens were designed, however to be as far as possible in conformance. This is also applicable to any test procedures carried out (such as fracture toughness testing).

One such specification is that a fatigue pre-crack must be included in the final a/W ratio of at least 2.5%W or 1.3mm, whichever is greater. The fatigue pre-crack is grown from a notch that is machined into the specimen in such a way so as not to introduce heat damage. Fatigue pre-cracking is the process of cyclically bending the specimen in such a way that $K_i$ remains low enough that no residual stresses are introduced into the specimen. This allows a crack, that would be similar in nature to those seen in structures under cyclic loading, to grow under fatigue from a pre-machined notch.
A fatigue pre-crack length of 2.5mm was chosen. This was based on the very low constraint option (a/W=0.05=2.5mm). It was decided that for this option a defect that would be entirely fatigue pre-crack (i.e. no notch), would provide the least difficulties when machining. This meant that a 5mm notch could be created in oversized specimens (W=55mm), and then the edge with the notch could be skimmed up to the notch tip, leaving only the fatigue pre-crack in the specimen. The a/W=0.2 and 0.4 specimens would therefore also have a fatigue pre-crack of 2.5mm; therefore, their notch depths would be 7.5 and 17.5mm respectively. A machining drawing of the a/W=0.4 specimen is shown in Figure 3-1. Also shown is the final drawing of the a/W=0.05 specimen, following the skimming of the notch.

![Figure 3-1: a/W=0.4 (L), 0.05 (R) specimens machining drawings (all dimensions in mm)](image)

As can be seen there is a difference in the design. When carrying out the fracture toughness testing, it is necessary to attach a clip (strain) gauge to knife-edges, to allow measurement of crack mouth opening displacement (CMOD). In deeper notched specimens, it is possible to machine the knife-edges directly into the notch. For the very shallow crack, it was necessary to attach the knife-edges with bolts; therefore, tapped holes were included in this design.
3.2.2. Flat Dog-bone Tensile Specimens for Plastic Strain Introduction

It was apparent from the size of the specimens that significant loading capabilities would be required to introduce 5% plastic strain into the SENB specimens. For ease of analysis and quality of results it was decided that a uniform plastic strain throughout the specimen would be ideal. Because of this plasticity would be introduced through uniaxial tension. This would allow significant plasticity to be introduced without also introducing a residual stress field, unlike bending or rolling the material. An 8806 250MN Instron testing machine was available at the National Composites Certification and Evaluation Facility (NCCEF) at The University of Manchester, which had sufficient capacity to introduce the strain required.

The specimens used for this process were designed to be similar in shape to flat tensile test specimens, such that the gauge length would be of sufficient size, that from it, 2 SENB specimens could be machined. As there were to be 2 different sizes of specimen required (W=50 and W=55mm), two slightly different gauge areas were required. Sufficient extra material was also included in the gauge lengths to allow for machining waste and reduction in area caused through pulling the dog-bone specimens to 5% plastic strain. The grip sections were designed such that there would be sufficient area to ensure grip within the rig, and that radii could be included so the strain would not be concentrated at the ends of the specimen. This design was completed in conjunction with expert opinions from The Welding Institute, who have significant experience in this type of procedure. Validation of the design is given in section 3.3.1. A drawing of the specimens is shown in Figure 3-2, the wider specimens (right) had slight alterations made to the relevant dimensions to accommodate the extra 5mm in the gauge width required for the waste area where the a/W=0.05 SENBs would have their notches removed.
3.2.3. Cut Design from Raw Plate

The raw plate provided by Wood was limited in size, so steps were taken to maximise the number of specimens that could be machined from the material available. It was decided that using a water jet cutter would minimise waste whilst allowing the plate to be cut into smaller pieces from which the specimens could be machined, maximising material usage and minimising cost. Such a water jet cutter was available at The School of Materials, The University of Manchester. The use of the water-jet cutter also prevented alteration of material properties that would have occurred through using any heat based cutting tools. The cuts from the water-jets were designed considering the SENB and dog-bone designs, such that material could be used efficiently. This meant using the areas between the dog-bone gauge lengths for the SENB specimens that were not to be subjected to plastic strains (as received material). The profiles of the dog-bones were arranged on the plate to give the cut design shown in Figure 3-3. This design used the material in such a way the production of 80 test specimens was possible, such that 10 tests per scenario (see Table 3) could be completed.
3.2.4. Machining Details

Following the water jet cutting, there remained 10 x 70mm thick pieces with the dog-bone profile and 10 x 70mm thick blanks from the areas between them. These were shipped to CNC Precision (Redditch) for further machining. The water jet cutter was unsuitable for any further machining because of the precision required and depth of the cuts to split the dog-bones to the appropriate thickness (as per the drawings in Figure 3-2). Initially returned from the machinists there were 10 x a/W=0.4 SENB specimens, 10 x a/W=0.2 SENB specimens, 20 x a/W=0.05 SENB specimens (all as received material); and 20 x dog-bone specimens (half with each nominal width: 50mm or 55mm). These virgin SENB specimens were subjected to fatigue pre-cracking, and the dog-bone specimens were subjected to the pre-straining process. Following pre-straining, the dog-bones were returned to CNC Precision where a similar second batch of pre-strained SENB specimens were machined from the gauge lengths. After this batch had been returned, fatigue pre-cracking was again carried out and all (40) a/W=0.05 SENB specimens were returned to CNC Precision to have 5mm skimmed from the notched face. Following this the designated specimens had residual stress introduced (20 x
a/W=0.05 specimens: half made from virgin, half from pre-strained material), and after all prior processes were completed, fracture toughness testing on all specimens was carried out. Further details of all pertinent experimental processes are given in section 3.3.

![Figure 3-4: Photographs showing (L-R): water jet cutting, cut arrangement, finished dog-bone specimens](image)

### 3.3. Experimental Details and Validation

This section details the experiments and the analysis used to ensure the desired output had been achieved in terms of the residual stress fields and plastic strains.

#### 3.3.1. Pre-straining of Dog-Bone Specimens

To introduce 5% plastic strain to the dog bone specimens it was calculated that a stress of ~640MPa would be required, based on the ambient temperature tensile test data available for the material (61). This equated to load requirements of 1.13MN for the specimens with the largest gauge area. The 8806 2.5MN Instron™ testing machine located in the NCCEF at The University of Manchester offered sufficient load capacity for these requirements. As described in section 3.2.2 the dog-bone specimens were designed such that they could fit within the grips of this machine and they would be suitable for this kind of testing. The machine manual was consulted to ascertain the appropriate grip pressure to ensure no slippage occurred during the pre-straining process.
4 strain gauges were attached to each specimen, equally spaced along the gauge length, starting from 50mm from where the radii in the grips ended (see Figure 3-2), such that no strain gauges were crushed in the grips, and were able to have lead wires (to measure the voltage signal) attached when the specimens were installed. 2 gauges were installed on both the front or the back faces, and each lead wire was used for the same position gauge in each test so it was clear where on the specimens each of the readings were taken from. Calculations were carried out to confirm that ~55,000με (the software that converted strain gauges’ output voltage output readings in με) would be required to achieve 5% plastic strain, in that following unloading there would be a 5% increase in the gauge lengths. 19 of the 20 dog-bone specimens were tested in an identical fashion: the specimens were pulled to 55,000με, (average from 4 gauges, if no gauge failure, otherwise an average from the remaining gauges), and then immediately unloaded. A strain rate of 0.5mm/minute was applied. Figure 3-5 shows the average true stress and true strain reading for all 19 tests. As can be seen all tests are within ±0.2% of the 5% plastic strain required following the unloading.

![Figure 3-5: Average true stress vs true strain readings from pre-straining process](image)

Evident in Figure 3-5 is slight non-linearity during the unloading portion of the stress strain curve. A533b is a material which has been shown to exhibit the Bauschinger
effect (8) (62) after a compressive load, and it could be possible that this, and the non-linearity observed on reloading, described in detail in section 4.1, could be attributed to the Bauschinger effect. The Bauschinger effect is a reduction in yield stress upon compression, such that for cyclic behaviour a material must be modelled using cyclic stress/strain properties and a potentially a kinematic hardening model. However this would not be the case as both works (8) (62) stipulated that for the material to exhibit the Bauschinger effect it must undergo a compressive load, as opposed to simply an unload from compression. The non-linearity observed in Figure 3-5 is therefore attributed to increased dislocation density as observed in the literature, described in section 2.7.3.

3.3.2. Fatigue Pre-Cracking

As explained in section 3.2.1 each of the test specimens included a 2.5mm fatigue pre-crack. These cracks were grown ahead of a notch, the detailed geometry of which is shown in Figure 3-1. Each notch had similar geometry; the only differences were the length of the notch or the inclusion (or not) of integral knife-edges for the clip gauge to be attached to. The process of fatigue pre-cracking was carried out at Wood’s Walton House site on Ansler ± 100kN vibraphore machines.

Each pre-crack was completed in 4 steps (0.625mm each). The specimens had lines drawn on at each step, and the load was stepped down incrementally from a $K_i$ value of 25 to 20MpaVm, such that no residual stress would be introduced, and there would also be minimal plasticity introduced at the crack tip, to simulate as close as possible in-service cracks. The crack growth was observed on video cameras and after the final step had been completed a final measurement of the crack length was taken, on both sides. Table 4 shows average fatigue pre-cracking data from this process.
<table>
<thead>
<tr>
<th>Specimen Batch</th>
<th>Mean (and standard deviation-SD) of no. of cycles to complete 2.5mm pre-crack</th>
<th>Mean (and SD) visible crack depth (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>a/W=0.4</td>
<td>327,329 (120,837)</td>
<td>19.85 (0.19)</td>
</tr>
<tr>
<td>a/W=0.4 pre-strained</td>
<td>222,130 (38,300)</td>
<td>20.06 (0.06)</td>
</tr>
<tr>
<td>a/W=0.2</td>
<td>298,958 (55,204)</td>
<td>9.93 (0.08)</td>
</tr>
<tr>
<td>a/W=0.2 pre-strained</td>
<td>196,220 (11,855)</td>
<td>10.06 (0.07)</td>
</tr>
<tr>
<td>a/W=0.05</td>
<td>198,989 (103,728)</td>
<td>7.54 (0.05)</td>
</tr>
<tr>
<td>a/W=0.05 pre-strained</td>
<td>129,195 (38,300)</td>
<td>7.50 (0.04)</td>
</tr>
</tbody>
</table>

Table 4: Fatigue pre-crack data

All of the average crack depth values fell within 1% of the required a/W ratios (the a/W=0.05 batches had a 7.5 mm crack depth because this measurement was prior to 5mm being skimmed from the notched edge). What can also be seen is a significant reduction in the number of cycles to achieve the same desired crack depth in the pre-strained specimens, demonstrating a reduction in toughness was introduced through the pre-straining.

3.3.3. Residual Stress Introduction

The final experimental pre-processing of the specimens prior to the final fracture-toughness testing involved the introduction of a residual stress field to half of the a/W=0.05 specimens (10 x pre-strained, 10 x virgin material). As described in section 3.1, the idea behind this process was to introduce a stress field ahead of the final fatigue pre-crack that would be as close as possible to those described in (3). There were several differences however between the original process (see section 2.6.3) and the process
used for this project. The most significant was the fact that the residual stress fields were applied after the fatigue pre-cracking process. In the previous project (3) the stress fields were applied before the pre-crack was inserted. Preliminary FE simulations were conducted to as closely as possible follow Hurlston’s methodology to assess what effects this would have on the residual stress fields after fatigue pre-crack introduction. These are shown in Figure 3-6.

The details of the FE simulation methods similar to those employed to obtain the results shown in Figure 3-6, are included in section 4.2 of this project, although the results in the figure used the same material properties as (3). Simulations were completed for the actual specimen geometries, with the punch the same relative distance from the final fatigue pre-crack tip (as opposed to the notch tip, as shown in Figure 2-23), for determination as how to introduce similar residual stress fields as close as possible those in the previous project.

Hurlston (3) conducted neutron diffraction measurements of the residual stress fields seen in his specimen, both before (shown in Figure 2-23), and after fatigue pre-

Figure 3-6: Residual stress fields following fatigue pre-crack insertion from FE simulations (FE) and Neutron Diffraction measurements (ND) (3)
crack introduction, which are displayed in Figure 3-6, which were then used to validate the numerical simulations such that the stress field ahead of the crack in the actual specimens could be said to simulate those in the FE simulations. Therefore, the results from these preliminary FE simulations were compared with those from Hurlston’s to see if it would be reasonable to expect similar residual stress fields if the side-punching method were used in such a way that the punch would have the same relative position and indent-depth from the crack tip, and would be employed after the introduction of a pre-crack. The results from Hurlston’s FE simulations and those from this work, alongside the neutron diffraction results are shown in Figure 3-6.

It was decided that these results validated the process of using the side punch the same relative distance from the fatigue pre-crack, and applying the same total indent of 0.5mm (2I, see Figure 2-23), to achieve similar residual stress fields. Also, because neutron diffraction had been carried out at two different crack depths, with similar results, using a third crack depth and similar methodology would likely yield similar neutron diffraction results, so this analysis would not be used for this project. It was also decided that using the same level of indent would be suitable for the pre-stretched material, because the residual stress fields in the FE simulations were in good agreement.

The decision to proceed without further experimental validation of the residual stress fields was finally justified because the output of this work did not require identical residual stress field’s as those from Hurlston’s project. The postulated J-Q failure curve methodology is designed to be material specific, and independent of initial stress/plastic strain conditions. Therefore slight differences in stress field (which can be seen in Figure 3-6) would not affect the derivation of the curve, and would still allow results from Hurlston’s work to be compared with said curve. The neutron diffraction was used to validate Hurlston’s (3) FE methodology, which was recreated faithfully for this work, with some improvements to accuracy. This, alongside the validation of the FEA against load/CMOD and J integral experimental data presented in section 5, sufficiently validated the residual stress field predictions obtained from the FEA.

The original tooling from Hurlston’s project was still available at Wood’s Walton House site. This tooling consisted of a jig, in which to sit the specimen, and four punches made from maraging steel, which is hard enough to perform the indenting into A533b
steel without deformation. Because the punch positions were now different, new holes had to be drilled into the jig, so the punches would be the correct distance from the crack tip, otherwise the process was carried out as far as possible to exactly recreate that described in (3).

The side punching was carried out using the 250MN Schenk testing machine at Wood’s Walton House site. Once the specimen had been positioned in the jig correctly, a small amount of load was applied to maintain its position. Following this the load was ramped up to 240kN for the virgin material and 273kN for the pre-strained material, to achieve a nominal total indent of between 0.5mm (measured using micrometer before and after side-punching). It would have been desirable to use displacement control for this process, but the displacements were too small for this to be possible with the equipment available. Because of this some trial and error was necessary to determine the load required. This meant some of the specimens saw indents beyond the range stated, with the total indents shown in Table 5. The range of indent depths necessitated a sensitivity study as to what effects on the residual stress field at the crack tip deeper indentations may have, which is presented in section 5.3.

<table>
<thead>
<tr>
<th>Test #</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
<th>6</th>
<th>7</th>
<th>8</th>
<th>9</th>
<th>10</th>
</tr>
</thead>
<tbody>
<tr>
<td>As Received</td>
<td>0.69</td>
<td>0.55</td>
<td>0.49</td>
<td>0.48</td>
<td>0.51</td>
<td>0.47</td>
<td>0.49</td>
<td>0.46</td>
<td>0.49</td>
<td>0.46</td>
</tr>
<tr>
<td>Pre-strained</td>
<td>0.42</td>
<td>0.60</td>
<td>0.56</td>
<td>0.43</td>
<td>0.49</td>
<td>0.48</td>
<td>0.50</td>
<td>0.48</td>
<td>0.46</td>
<td>0.49</td>
</tr>
</tbody>
</table>

As received mean=0.509mm (SD=0.07mm)  
Pre-strained mean=0.491mm (SD=0.05mm)

Table 5: Measured indents in mm after side punching
3.3.4. Fracture Toughness Testing

All fracture toughness testing was carried out at Wood’s Newton House site, on a Schenk 250kN testing machine. The tests were carried out following all subsequent pre-processes, detailed in sections 3.3.1-3.3.3. All tests were carried out to, as far as possible comply with the relevant standards (16) (17), despite the non-standard geometry. The tests were carried out at -140°C, to ensure as far as possible a cleavage mode of failure, that could potentially simulate optimum radiation embrittlement that may be seen in RPV steels during service combined with the low temperatures associated with plant shutdown. Furthermore, this temperature was the same as the previous tests with which the results from this project will be compared (3).

The test method involved mounting the specimen on rollers as shown in Figure 3-7. The specimens were cooled using liquid nitrogen in an environmental chamber, with the temperature monitored by a thermocouple, which was installed in contact with the specimen. To ensure that the entire specimen had cooled to the appropriate temperature, the cooling continued after the temperature reached -140°C for 1min/mm thickness of the specimen. Once the specimen was sufficiently cooled the test was carried out. Load was applied as shown in Figure 3-7, until the specimen failed. In all cases presented as results the specimen broke cleanly in two (2 of the very low constraint tests did not fail at all and the results are omitted from the analysis presented later in this thesis).
Following cleavage of the test piece, measurements of the crack were taken at 9 equidistant points through the thickness of the specimen (starting from a position of 0.01B into the thickness of the specimen), as required by ASTM 1921 (16). This analysis is used to assess the extent of ductile tearing during the testing if applicable, and the fatigue pre-crack dimensions, to ensure the results are valid.

### 3.4. Tensile Test Programme

Preliminary scoping calculations for this work, published in (59), were based on a consideration of standard plasticity theory, whereby a lateral of 5% shift in the stress strain behaviour was modelled to simulate the pre-straining process, using historical tensile data (61) for A533b. This approximation did not represent, when modelled using finite element analysis as described in section 4.2, the behaviour the material showed in the fracture toughness tests, specifically the load v CMOD behaviour against which the models were validated. As described in section 2.7.3, it has been reported in the literature that pre-straining material has subtle effects in the supposed elastic behaviour of material that has been pre-strained, to similar levels as that used test programme presented in this thesis. It was therefore necessary to obtain accurate tensile data for
the pre-strained material to be used in the finite element analysis, which required a tensile test programme to be carried out.

The tensile test programme was carried out according to BS 10002 (63). The specimens were of standard circular section design, with a gauge diameter of 15mm and a machined gauge length of 43mm. The specimens were machined from the material in the broken half-specimens following the fracture toughness test programme. They were all taken from the $a/W=0.4$ geometries, to ensure crack tip plasticity would not have extended into the region from where the specimens were to be extracted. Specimens were extracted from both pre-strained and as received SENB halves. The SENBs from which the specimens were selected were those which fell in the central region of the failure load scatter bands for their respective test batch, to minimise the possibility of any potentially strengthening or weakening material anomalies. A machining drawing of the specimen design is shown in Figure 3-8.

![Figure 3-8: Tensile test specimen drawing, all dimensions in mm](image)

The tests were carried out approximately a year after the initial pre-straining had been carried out. It was therefore necessary to determine if any strain aging effects had occurred in the pre-strained material, which would therefore determine if the material behaviour measured directly from the pre-strained SENB halves would be representative of the material at the time of fracture toughness testing, which was carried out between 4 and 9 months of the initial pre-straining process. To assess the tensile behaviour of the pre-strained material, it was therefore decided to compare the behaviour of specimens that were pre-strained immediately before testing, with those machined from the pre-
strained material. It was necessary to carry out testing at both ambient temperature and \(-140^\circ C\), because both the side-punching (carried out at ambient temperature) and fracture toughness testing (carried out at \(-140^\circ C\)) processes were modelled. This decided the tensile test matrix shown in Table 6. The number of tests was decided based on funding and time constraints and it was decided 3 tests per scenario, denoted with an x in Table 6, would be tested, totalling 18.

<table>
<thead>
<tr>
<th>Tensile test temperature</th>
<th>As received</th>
<th>Loaded to 5% plastic strain at ambient temperature prior to test</th>
<th>Pre-Strained</th>
</tr>
</thead>
<tbody>
<tr>
<td>20°C</td>
<td>x</td>
<td>x</td>
<td>x</td>
</tr>
<tr>
<td>-140°C</td>
<td>x</td>
<td>x</td>
<td>x</td>
</tr>
</tbody>
</table>

Table 6: Tensile test matrix, 3 x each test carried out

All of the tests were carried out at Wood’s Walton House Site. They were conducted on a Zwick tensile testing machine with a 50kN load cell. The tests were displacement controlled at a strain rate of 0.025%/sec using TestXpertII (64) software inbuilt to the test system. Prior to all testing pertinent measurements were carried out including gauge diameter measured to a resolution of 10\(\mu\)m. When the material was to be pre-strained during the tensile test programme the machine was set to achieve the same level of strain as with the dog-bone specimens (5.5%, see section 3.3.1), and then unloaded, immediately prior to the full tensile test being carried out. These specimens had their gauge diameters re-measured to ensure a valid stress could be calculated from the subsequent, full tensile test (in some cases this step was omitted so estimates were made based on the cases where it was carried out). The strain was measured through an extensometer up to 1.5%, and then through crosshead travel beyond this strain. The results of this test programme are presented in section 4.1 of this thesis.
3.5. Summary

- A test programme was designed and carried out to assess the effects of significant plasticity and tensile residual stress fields on the effective fracture toughness of a ferritic RPV steel.

- To present a wide range of constraint, SENB specimens with 3 crack depths were designed: high constraint \((a/W=0.4)\), medium constraint \((a/W=0.2)\) and very low constraint \((a/W=0.05)\).

- A residual stress field was mechanically introduced through the double side-punching process, which has been shown to leave a stress field comparable to that a defect near a weld may be subjected to.

- Significant uniaxial plasticity was introduced prior to specimen machining through pre-straining the blank material.

- To better understand the material behaviour of the pre-strained material, tensile tests were carried out on specimens machined from broken halves of the tested SENBs.
4. Numerical Modelling

To calculate J-Q loading curves (see section 2.5.2) for the experiments described in section 3, it was necessary to calculate the stress fields ahead of the crack for all of the test scenarios, requiring validated FE simulations to be carried out. To model the experiments accurately it was necessary to model both the introduction of residual stresses and the fracture toughness testing. Due to the subtleties reported in the literature regarding the effects of pre-straining on the material behaviour of steels in the elastic regime, the tensile tests that were carried out directly on the pre-strained material meant that pre-straining process need not be modelled, its effects could be included in the material properties of the pre-strained simulations. The results from these tensile tests (described in section 3.4) are included in section 4.1, and the implications of the values used and approximations made is discussed in section 7 and 8 of this thesis. The results from the tensile test programme are included in this section as they describe the material properties included in the simulations.

4.1. Material Properties - Tensile Test Results

The elastic plastic material behaviour was modelled on the results of the tensile test programme described in section 3.4. It was necessary to create validated elastic-plastic material data that would enable the accurate modelling of both the side punching process carried out a 20°C and the fracture toughness testing which was carried out at -140°C. Furthermore, it was necessary to model both the pre-strained material and the as received, requiring a total of 4 material models (shown in Table 7) to be approximated from the data, utilising a consistent methodology. This required detailed consideration of two factors: the Lüder’s strain apparent in tensile tests of as received ferritic steels, and the subtle effects of pre-straining on material properties of metals reported in the literature as described in section 2.7.3.
The Lüder’s phenomenon materialises as bands of localised plasticity caused by the pinning of dislocations to Cottrell atmospheres (present in interstitial carbon atoms in ferritic steels). This results in the flattening of the stress strain curve, which continues until the dislocations are released from the atmospheres after which standard work hardening and uniform plasticity occur. This is displayed in Figure 4-1, which displays the results of one of the tensile tests carried out for this project.

<table>
<thead>
<tr>
<th>Temperature</th>
<th>As received</th>
<th>Pre-strained to 5% plastic strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>-140°C</td>
<td>A</td>
<td>B</td>
</tr>
<tr>
<td>20°C</td>
<td>C</td>
<td>D</td>
</tr>
</tbody>
</table>

Table 7: Material models

Figure 4-1: Lüder’s strain in A553b tensile test
This Lüder’s effect is not apparent in areas of stress concentration, the increased stress prevents the dislocation pinning. In most real-world situations, therefore (apart from uniaxial tensile testing), and in terms of modelling and simulation, Lüder’s strain is can be considered as artificial. Therefore, when considering finite element modelling of steels which display Lüder’s behaviour, it is necessary to consider how it should be treated in terms of the tensile properties of the material. Hurlston (3) and James (41), both suggested methods for approximating the stresses seen in the material at the low strains where Lüder’s behaviour is seen, and there is also guidance in RSE-M, the French fracture standard (65). In all cases Ramberg-Osgood curve fits (equation 17) are applied to the stress strain behaviour beyond the Lüder’s.

James (41), explored the various methods and demonstrated good modelling results for A533b by implementing a curve fit “to fit the initial yielding region and to re-join or form a close tangent to higher strains.” He also considers a ‘shift’ of the stress strain curve beyond the Lüder’s region, to join up with where yielding occurs, but discounts this as viable because it would be difficult to accurately assess the ‘length’ of the Lüder’s strain region, as there is some curvature between the elastic limit and the beginning of the Lüder’s strain. This however was considered for the material models used in the simulations presented in this thesis as it was decided that engineering judgement could be applied as to how far to shift the stress-strain curve, and this shift may be necessary when considering both the pre-strained (which would not demonstrate Lüder’s, as, the dislocation pinning is removed in the initial pre-straining process) and as received materials with the same methodology. An example of James’ methodology is shown in Figure 4-2.
As the methodology displayed in Figure 4-2 demonstrated good modelling results this formed the basis for modelling of the material properties for use in the simulations presented in this thesis. As can be seen the curve fit ‘overcuts’ the actual test data, and the limit of proportionality falls below actual yield point. A similar methodology was applied to the tests reported here. It was however decided after consideration of the physical processes involved that shifting the stress/strain curve to the ‘beginning’ of the Lüder’s strain would be appropriate. Therefore, for each test scenario (see Table 7), the methodology shown in Figure 4-3 would be applied to convert the tensile test data to data that could be employed as true stress/plastic strain data for use in the finite element analyses.
4.1.1. Young’s Modulus

Initially the Young’s modulus for use in the material models was decided. Several of the test results displayed non-linearity in the ‘elastic’ region of the stress strain curve. This was especially apparent in the pre-strained material, as was expected based upon the literature as discussed in section 2.7.3. Therefore, it was necessary to consider from where in this region the modulus would be taken. As the modulus is a measure of the elastic response of the material, and steels do not display non-linear elasticity, the modulus values for each test was taken from the linear region of the data, which consistently occurred between $0 < \sigma < \sigma_{y}/3$, although the length of the straight portion was larger in the as received material. Because there were 6 tests where the pre-straining was carried out in the tensile specimens, it was possible to obtain 9 Young’s modulus values for the as received material at ambient temperature. The values of Young’s modulus obtained for each of the tests is shown in Table 8, with the
corresponding values of modulus before and after the pre-straining process plotted against each other.

<table>
<thead>
<tr>
<th>Test #</th>
<th>Scenario A (As Received, -140°C)</th>
<th>Scenario B (Pre-strained, -140°C)</th>
<th>Scenario C (As Received, 20°C)</th>
<th>Scenario D (Pre-strained 20°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 (SENB)</td>
<td>215.3</td>
<td>230.8</td>
<td>204.0</td>
<td>197.5</td>
</tr>
<tr>
<td>2 (SENB)</td>
<td>244.9</td>
<td>250.7</td>
<td>202.2</td>
<td>202.4</td>
</tr>
<tr>
<td>3 (SENB)</td>
<td>227.6</td>
<td>227.7</td>
<td>207.7</td>
<td>205.5</td>
</tr>
<tr>
<td>4 (Tensile)</td>
<td></td>
<td>220.0</td>
<td>225.7</td>
<td></td>
</tr>
<tr>
<td>5 (Tensile)</td>
<td></td>
<td>203.3</td>
<td>186.8</td>
<td></td>
</tr>
<tr>
<td>6 (Tensile)</td>
<td></td>
<td>236.8</td>
<td>204.7</td>
<td></td>
</tr>
<tr>
<td>7 (Tensile)</td>
<td></td>
<td></td>
<td>238.7</td>
<td></td>
</tr>
<tr>
<td>8 (Tensile)</td>
<td></td>
<td></td>
<td>213.4</td>
<td>174.9</td>
</tr>
<tr>
<td>9 (Tensile)</td>
<td></td>
<td></td>
<td>188.2</td>
<td>172.8</td>
</tr>
<tr>
<td>Mean (and SD)</td>
<td>229.27 (14.87)</td>
<td>228.22 (15.97)</td>
<td>207.93 (16.53)</td>
<td>190.28 (15.58)</td>
</tr>
</tbody>
</table>

Table 8: Young's modulus values in GPa obtained from tensile test programme (SENB/Tensile indicates in which specimen pre-straining was carried out, where applicable), those used in modelling highlighted in green

It can be seen there is significant scatter in the modulus values. There is also discernible reduction in average modulus through pre-straining at room temperature, although this was not the case in the low temperature tests. Previous work (3) on this material used a value of 220 GPa for modelling and achieved good results for similar
simulations. Therefore, the modulus value used for scenario A was the closest value of 215.3 GPa, which was the lowest value from the tests.

It was found after a parametric study that only a significantly reduced modulus would capture the load/CMOD behaviour displayed in the pre-strained fracture toughness tests, so the lowest value low temperature value of 203.3GPa from the tests was used in the modelling. This implied a reduction of 5.3%, which was less than reported in the literature, as described in section 2.7.3. The work reported there, however, considered a different material at different temperatures, so no direct application of those observations could be made to these tests. Therefore, using a value for the pre-strained modulus extracted directly from the tests, yet lower than the value for the as received, was justified.

For the moduli at room temperature, a similar methodology was applied. As this data would be used for defining the material behaviour during the side punching, it was again decided to use, for the as received material, as close a value from the tests as possible as Hurlston (3), which was 207.7GPa. Consistency was necessary, so the closest value from the tests showing the same reduction (5.3%) was used for the pre-strained material (197.5GPa).

### 4.1.2. Ramberg Osgood Curve Fits

Following determination of the Young’s modulus, Ramberg-Osgood curve fits were applied according to equation 17, where $\epsilon_0$=offset strain referred to in Figure 4-3. $\sigma_{0.2\%}$ (the stress at 0.2% strain) and n values were chosen to offer the best curve fits to all of the true stress/plastic strain data for the given test scenario. The curve fits with the corresponding values for $\sigma_{0.2\%}$ and n are shown in Figure 4-4 - Figure 4-7, with scenarios A-D (see Table 7) displayed top to bottom. For scenario C (as received, 20°C), where many ‘extra’ measurements were taken during the pre-straining process, only the full tests were considered for this aspect of the material modelling, as they displayed the full hardening curve. It should be noted that above necking the comparison stops for the derivation of the curve fits, as the standard true from engineering stress/strain conversion is not applicable.
Figure 4-4: Ramberg-Osgood Curve fit to hardening region (scenario A), as received material, -140°C

Figure 4-5: Ramberg-Osgood Curve fit to hardening region (scenario B), pre-strained material, -140°C
Figure 4-6: Ramberg-Osgood Curve fit to hardening region (scenario C), as received material, -20°C

Figure 4-7: Ramberg-Osgood Curve fit to hardening region (scenario D), pre-strained material, 20°C
As can be seen in Figure 4-5 and Figure 4-7 the pre-strained material does not display Lüder’s strain in that there is no discernible flattening of the curve in the low strain region, as can be seen with the as received material, therefore only the as received material would have the stress-strain curve ‘shifted’ to the left to omit this behaviour. The amount of Lüder’s strain seen in the as received material was measured at 0.9% at 20°C and 1.9% at 140°C, so this value was subtracted from all strains at a given stress.

### 4.1.3. Elastic Limit

For each of the material models the elastic limit of the material was approximated where deviation from linearity was seen in the stress-strain behaviour. It was necessary to assign this value to complete the material models such that a curve fit could join between the Ramberg-Osgood fits to the main hardening curve and the stress where plasticity would begin in the modelling. To observe the elastic limit, the modulus value for each of the test scenarios displayed in Table 8 was used to calculate plastic strain and observation of where significant deviation from 0 was carried out. For the as received material this behaviour was fairly standard, in that the stress-strain behaviour was fairly linear up to an easily observed yield. The pre-strained material however displayed significant non-linearity in the very low strain regions, and it proved to be necessary to capture this in the material models. To maintain consistency a value of stress was
chosen to the nearest 10MPa increment that would have mean plastic strain of less than $\pm 0.0055\%$, with engineering judgement also used in defining a suitable elastic limit with the pre-strained material which showed significant variation and curvature in the low strain region of the tensile curves. Close ups of the low-strain regions of the true stress/plastic strain behaviour are shown in Figure 4-8 - Figure 4-11.

![Figure 4-8: Elastic limit, scenario A (as received material, -140°C)](image)

![Figure 4-9: Elastic limit, scenario B (pre-strained material, -140°C)](image)

![Figure 4-10: Elastic limit, scenario C (as received material, 20°C)](image)

![Figure 4-11: Elastic limit, scenario D (as received material, 20°C)](image)
As can be seen in Figure 4-8 - Figure 4-11, there is significant non-linearity in the pre-strained material, at both temperatures, from very low stresses. It can be seen most prominently in the results at 20°C, where there is plastic strain at almost zero stress in some of the tests, although an elastic limit of 220MPa was chosen because this is the stress where there is discernible deviation in plastic strain from 0. There are clearly dislocations mobilised by the pre-straining which enable plastic deformation at very low stresses. These results and their importance in terms of modelling the material behaviour are discussed further in section 5, 7 and 8 of this thesis.

### 4.1.4. Final Material Models

Following determination of modulus, Ramberg-Osgood curve fits (shifted to omit Lüder’s strain in the as received specimen) and elastic limit of the material it was necessary to finalise the material models again using a consistent methodology, to approximate the transition between the elastic limit stress and the hardening curve. It was decided based on a parametric study which also considered tracing exactly the yield behaviour in the low strain region of the pre-strained material that the most accurate material behaviour could be achieved by curve fitting between the elastic limit and a point where the curve could transition smoothly into the Ramberg-Osgood fit to the hardening curve.

The most suitable fits to allow smooth tangential transition required the 2 curves to join at approximately 2%, and used equation 17, and a similar methodology to that described in section 4.1.2. This resulted in the curve fits shown in Figure 4-12 - Figure
4-15, converted back to true/stress strain to allow comparison between the material models and all of the test data simultaneously.

Figure 4-12: Material model (scenario A), as received material, -140°C

Figure 4-13: Material model (scenario B), pre-strained material, -140°C
Figure 4-14: Material model (scenario C), as received material, 20°C

Figure 4-15: Material model (scenario D), pre-strained material, 20°C
These material models were used as input data for the finite-element simulations and showed good results in terms of accurately modelling the load/CMOD behaviour during the fracture toughness testing, as shown in section 4.2.3. The importance of this
analysis is discussed in section 5.1, where it is shown to have significant (and potentially disregarded) bearing on how structural integrity assessments of pre-strained material are carried out.

It is conceded that there are also several uncertainties within the models. The selection of modulus value was shown to have significant effect on the load/cmmod behaviour of the simulations, and the values employed (see Table 8) were used such that they would best fit the experimental data, as were the low-strain Ramberg-Osgood fits to the data, following a parametric study considering the different values observed in the tensile tests. It was observed for example that including higher plastic strains below the 0.2% proof stress, and a higher modulus, may yield similar load/cmmod results. The methodology presented in sections 4.1.1 to 4.1.3 was utilised to ensure a repeatable approach was applied for each of the models, that could be validated against the experimental data. It is possible that with improved understanding of the mechanisms causing the behaviour of the pre-strained material, more accurate material models could be created that would accurately and repeatably predict the low-strain tensile behaviour of this type of material.

4.2. SENB Fracture Toughness Test Modelling

The commercial finite element analysis software package Abaqus (22) was used to carry out all numerical simulations of the testing programme. This section describes how the fracture toughness testing (described in section 3.3.4) and side-punching process (described in sections 2.6.3 and 3.3.3) were simulated such that crack tip stress fields could be calculated to enable calculations of J-Q loading curves for all of the test scenarios.

4.2.1. Geometry and Mesh

Meshes were created for each of the SENB geometries, whose dimensions were 240 x 50 x 25mm, with nominal crack depth to thickness ratios of a/W=0.4, 0.2 and 0.05. Using
the part-extrude menu a 120mm (specimen length/2, x axis, from origin) x 50mm (W, y axis, from origin) x 12.5mm (B/2, depth, z axis, from origin) (see Figure 4-18 for axis orientations; Figure 2-10 for B and W definitions) rectangular block was constructed. This part was defined as a deformable solid and used to represent a quarter model of an SENB specimen, so that symmetry could be exploited through two planes, along the centre of the length of the specimen and through the ligament and crack front (xy and yz planes), as shown in Figure 4-18. The symmetry in the ligament was used to create the crack, so part of the ligament, that was to be the crack and notch, was left unconstrained. The cracks were modelled as sharp and straight horizontal through the thickness of the specimen. To improve accuracy, the average fatigue pre-crack depths, measured after the fracture toughness testing, were used as the crack depths in the simulations. These values are presented in Figure 4-17 and Table 11, although the crack depths will continue to be referred to as their nominal values in the remainder of this thesis.

![Graphical representation of fatigue pre-crack measurements](image)

Figure 4-17: Graphical representation of fatigue pre-crack measurements
The notches for the $a/W=0.4$ and 0.2 specimens were included in the model, however the knife edge detail was omitted for simplicity as this would have negligible effect on the crack tip stress fields and $J$ calculations. For the $a/W=0.05$ specimens, the knife edges were included in the simulation geometry. These were modelled as simple rectangular forms, connected to the specimen at the same horizontal difference from the centre of the specimen as the bolts, and extending to the same distance from the bottom of the specimen as the average of the 3 actual measured knife edge geometries. This was to enable accurate clip-gauge opening displacement measurements to be taken for comparison with the output from the testing system.

<table>
<thead>
<tr>
<th></th>
<th>$a/W=0.4$</th>
<th>$a/W=0.02$</th>
<th>$a/W=0.05$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nominal crack depth</td>
<td>20</td>
<td>10</td>
<td>2.5</td>
</tr>
<tr>
<td>Average Measured Crack Depth</td>
<td>20.61</td>
<td>10.63</td>
<td>3.22</td>
</tr>
</tbody>
</table>

Table 11: Comparison between nominal design crack depth and average measured values through thickness of the specimen, which were used as the crack depths in simulations.
The loading and rest rollers were simulated as analytical rigid (non-deformable) semi-circular surfaces. To model the side punch, a third part was created, a circular analytical rigid surface with radius 10mm. This was given a trapezoidal section to represent accurately the side punch. The specimen model was partitioned to allow for the correct meshing techniques to be used to evaluate suitable contour integrals at the
crack tip, based on the suggestions in the Abaqus fracture mechanics handbook (66). At the crack tip a semi-circular partition was created with a radius of 2.5mm, and further partitions were created to aid with meshing. The crack tip area was seeded such that there were 45 concentric rings, the smallest with a radius of 2.5μm (bias ratio= 1000), to allow for the contour integrals to be evaluated very close to the crack tip. This area was applied mesh controls which were ‘hex-dominated, swept mesh, medial axis’, the rest of the specimen was seeded at 1mm, with mesh controls ‘hex, swept mesh, medial axis’. The element type used was linear 3D stress, reduced integration with 8 nodes (C3D8R). Linear elements were used to reduce analysis time and were considered sufficient with the mesh refinement used. Reduced integration elements with only one integration point again reduce processing time, and to prevent the shear-locking effect whereby overlap of integration points can introduce singularity errors to the calculation.

The crack tip area was defined as a ‘contour integral’ in the engineering features module, with the crack front running through the thickness at the focus of concentric rings. Symmetry was applied to the crack definition and the midside node parameter was set at 0.25, towards the crack tip. The positioning of the midside node parameter in this way is suggested in the Abaqus Fracture Mechanics Handbook (66) and allows the elements at the crack tip to be collapsed such that the nodes at the crack tip become a singularity.

Further partitions were applied in the area where the punches were to be indented. The partitioning strategy was used to create further concentric rings with in the area where the punches would contact the specimen would be refined and allow for better stress distribution to the central ligament, where residual stress measurements were to be taken. Figure 4-18 illustrates the mesh for the a/W=0.4, with the a/W=0.2 specimen being identical apart from the overall crack depth and specimen thickness.

The assembly was created with the lowermost central axis of one (loading) roller aligned with the top left edge of the quarter specimen model, and the corresponding axis with a position 100mm along the bottom of the specimen, for the rest pin, to create a span (S, see Figure 2-10) of 200mm. Surfaces were assigned to the top, bottom and front of the specimen in the features module and the outer faces of the rollers and side punch, and reference points were assigned to the centre on the outer face of the roller, and the centre of the face where the roller was to contact the specimen. Contact pairs
were defined between all surfaces that would be in contact with each other during the simulation.

4.2.2. Boundary Conditions and Loading

Prior to any loading initial fields were applied to the specimen model to enable the side punching to be simulated at 20°C and the fracture toughness testing to be simulated at -140°C. Using the predefined fields option in Abaqus it was possible to assign different temperatures to the specimen during different time steps, to allow accurate simulations of these procedures, therefore the first step had an ambient temperature field applied. A later step was added to cool the model from 20°C to -140°C.

To simulate the process of side-punching a ‘static, general’ step (all steps for this analysis were considered as ‘static, general’, non-linear geometry assumptions were also removed from all steps) was applied after the initial step. This step was used to simulate the side punch entering the front of the specimen. During this step a displacement boundary condition in the y direction of -0.34mm, was applied to the reference point on the punch part, this resulted in a final indent after unloading of 0.5mm, as shown in Figure 5-16.

It was necessary to constrain the movement of the specimen during the side punching simulation, and as during the testing the top (in the orientation shown in Figure 4-18) of the specimen was pushed hard up against the jig which enabled the side punches to be correctly positioned, the top surface of the specimen was constrained in the y direction.

The subsequent step was as above but the punch displacement was removed, and then in step 3 the boundary conditions constraining the specimen in the jig were removed. The following step involved the cooling of the specimen to a cryogenic temperature, through removing the initial temperature field and applying a new one at -140°C. A further ‘static/general’ step was created after the cooling step. During this step, a displacement boundary condition in the y direction was applied to the loading roller’s reference point to exert the primary load on the specimen. This was set at a value to go approximately 20% above the maximum load measured during each of the test
scenarios. The displacement was set to a prescribed number of increments such that an appropriate amount of measurements could be taken (increments of \( t=0.01 \)).

### 4.2.3. Model Verification

To initially validate the model a mesh sensitivity analysis was carried out. This was to assess that the contours at the crack tip were small enough to prevent discrepancy in the Q calculations. The mesh sensitivity analysis was also used to assess the effects of omitting the NLGeom (large strain assumptions) in the calculations. Therefore, the normalised stress vs normalised distance plots that would form the basis for the Q calculations (see Figure 2-19) were calculated with contours decreasing in size from \(~5\mu m\) to \(~1.25\mu m\). The mesh sensitivity was carried out on a 2D \( a/W=0.4 \) geometry, with the same mesh profile as would be used in the 3D analyses. The model was loaded to the highest \( J_c \) value measured during testing for that particular geometry. The results are shown in Figure 4-19.

![Figure 4-19: Mesh and small strain assumption sensitivity analysis results](image-url)
As can be seen in Figure 4-19 the chosen mesh size with the smallest crack tip of 2.22µm was sufficiently small to converge on a single curve for use in the Q calculations, with no discernible difference with a more refined mesh. The large strain model using the same mesh size also converges on the same curve at a value of 2, demonstrating validity in using small strain assumptions and a normalised distance of 2 in the Q calculations, as is common practice.

Prior to completing full simulations using plastic properties and inducing residual stress, 3D linear elastic simulations were carried out to validate the simulations of the fracture toughness testing. These linear elastic tests were based on equating the $K_i$ values calculated from the simulations to the handbook solution given in R6 (26) and BS7448 (17) (equation 13). The only material property used was the Young’s modulus at $-140°C=215.3$GPa.

The reaction force at the roller at each increment of displacement was calculated. These values were multiplied by 4 (as a quarter model was used), and inserted as F into equation 13. A value of $K_i$ was calculated from the simulations, through conversion from the strain energy release rate, $J$, using equation 20. These values were compared with the handbook solutions for all of the crack geometries, at both the mid-plane and as a weighted average through the thickness of the specimen. These results are shown in Figure 4-20 - Figure 4-22.

![Figure 4-20: Comparison of linear elastic a/W=0.4 $K_i$ values calculated from simulations at mid-plane, weighted average through specimen thickness and handbook solution (17), (26)](image-url)
As can be seen in Figure 4-20 – Figure 4-22 there is good agreement for the $K_i$ values for the all crack geometries for the linear elastic validation, although there is an upper-bound discrepancy of ~5% in the $a/W=0.4$ geometries when considering the maximum mid-plane value. This is significantly higher error than the 0.2% claimed in R6 (26) for this type of analysis. The handbook solution however is from a programme of curve-fitting analyses to 2D plane strain finite element simulations. What is likely to be
causing the discrepancy shown in Figure 4-20 is that true plane-strain conditions are not apparent in this simulation in the mid-plane.

Therefore, to further validate the mesh and corroborate this hypothesis, a 2D-plane strain assessment was carried out using a very similar mesh profile; and exactly the same loading techniques, contact controls, material properties, contour integral regions and boundary conditions. This result is presented in Figure 4-23. As can be seen there is almost zero discrepancy in this analysis which justified the modelling techniques. It also provided validation for consideration an average value of J through the thickness of the specimens in the elastic-plastic analysis. As can be seen there here and in section 5.2, there is an apparent 3-dimensionsal effect on $K_I$, whereby an average value is closer to the plane-strain assumptions used to derive the handbook solutions, than that at the mid-plane of the specimen which physically should be closer to plane strain as there is most material to prevent strain in the largest dimension, and furthermore there is a symmetry boundary condition to prevent strain in the z direction as shown in Figure 4-18.

![Figure 4-23: Comparison of linear elastic a/W=0.4 $K_I$ values calculated by from a 2D plane strain simulation and handbook solution (17), (26)](image-url)
4.3. Summary

- A high accuracy 3D FE simulation was carried out for each of the test scenarios.
- Where residual stress was included through side punching, this process was also simulated.
- The material properties used as input data were derived from a set of tensile test results carried out as part of the work presented in this thesis.
- Both a reduced modulus and elastic limit were incorporated into the pre-strained material models, based on non-linearity displayed in the low-strain regions of the stress strain curves.
5. Fracture Toughness Test and Numerical Simulation Results

The analysis presented in section 4.2.3 is the closest-to-an-analytical solution available for linear elastic modelling of structures with defects, which has justified the modelling approach, but did not consider elastic-plastic materials, with which this project is concerned. It was therefore necessary to validate the material properties used, especially considering the approximations used in the material models (see section 4.1) to remove Lüder’s strain and incorporate the lower elastic limit and Young’s modulus observed in the pre-strained material. Therefore, comparison with the experimental results were made.

The validation was 2-fold; comparing the FE results with both the load/CMOD (CGOD for a/W=0.05 simulations including the non-integral knife edges) measurements extracted from the testing equipment, and the J-integral calculations as per ASTM 1921 (16) (this standard was chosen as the $\eta$ factor used in equation 23 enables the standardised calculation of $J$ for the shallower cracks tested in this project).

5.1. Load/CMOD (CGOD) Results without Residual Stress

The first set of results, presented in Figure 5-1 - Figure 5-4, are for the as received and pre-strained material at the 3 nominal crack depths of a/W=0.4, 0.2 and 0.05. in the case of the a/W=0.05 specimen geometries, the x-axis is labelled clip-gauge opening displacement (CGOD), because the results from the displacement were compared with the displacement of the retrofitted knife-edge that was included in the simulations. This replaces crack-mouth opening displacement (CMOD), which is equal to the CGOD in the specimen geometries which could include an integral knife edge (a/W=0.4, 0.2). There is comparison with the simulation results with the load calculated at the reference point of the analytical, rigid, load roller; and the CM(G)OD at central the mid-plane of the specimen, at the knife edge position. For the elastic plastic analysis, the knife edge was treated as a very soft (E=0.1GPa), elastic material, such that it would have negligible effect on the crack tip stress fields in the actual specimen, whilst still capturing the displacements and rotations of the knife edge position.
Figure 5-1: Load v CMOD plots for a/W=0.4 test scenarios

Figure 5-2: Load v CMOD plots for a/W=0.2 test scenarios
There are several points for discussion raised from observation of Figure 5-1 - Figure 5-4. It is evident that the pre-straining process has served to reduce the average failure load in all geometries. This is attributed to an increase in the number of microcracks caused by the pre-straining process. There is empirical evidence as reported in a seminal work.
by Gurland (67), shown in Figure 5-5, that ~1% of particles which act as microcrack initiators in a similar material are broken at 5% strain. This increase in microcracks clearly would clearly reduce the fracture toughness of the material based on the fracture process as described in 2.1.2. Furthermore, the increase in microcracks would corroborate the reduced elastic limit shown in the pre-strained material (see section 4.1), as they would act as new dislocation sources.

![Figure 5-5: (Left) Percentage of microcrack initiators broken as function of strain, with (right) 1500x magnification photograph of microstructure of 1.05% C steel showing microcracks after tensile deformation (67)](image)

The increase in mobile dislocations causing the reduced elastic limit also explains the most important observation, which is that of the experimental load/CM(G)OD traces. This necessitated the tensile test programme described sections 3.4 and 4.1, to obtain better understanding of the observed behaviour. As can be seen in Figure 5-1 - Figure 5-3, the pre-strained material exhibits greater CMOD than the as received material at the same load. When considering standard plasticity theory this would not be the case, the material would yield later and follow the same elastic behaviour as shown in Figure 5-6, where the prediction is from applying a 5% plastic strain field to the as received simulation, which, as described in section 2.7.3, is a simple lateral shift in the stress strain behaviour, with an identical modulus before and after pre-straining.
As can be seen in Figure 5-1 - Figure 5-3 the finite element calculations show good agreement with the test results; the load/CM(G)OD traces lie within the scatter band for all of the geometries considered. This validates the material models described in section 4.1, which included a reduced modulus and elastic limit for the pre-strained material. This is a very important observation for structural integrity assessments, because it has significant implications on the calculations of J in finite element analyses, in that the crack tip field displacements $\frac{\partial u}{\partial x}$ are altered, or if measuring J experimentally has effects on the area under the load CMOD curve, (see section 2.3.3). This is demonstrated through the J calculations from both the standard plasticity material model and that used in the simulations, as shown Figure 5-7.

Figure 5-6: Standard plasticity theory load/CGOD prediction for a/W=0.2 test scenarios
At a maximum load of 183kN in the a/W=0.05 scenario the J value calculated using the pre-strained material model is 143% the J value calculated using the standard plasticity model in the same simulation. This is a very significant difference and has obvious implications in terms of how plastic strains are considered in defect tolerance assessments. As is clearly evident, when considering standard plasticity theory there is significantly higher load at a given value of J than considering the more realistic material model. Should a defect appear in the HAZ of a weld, where plastic strain may be ascertained by finite element analyses or hardness measurements such as are described in sections 2.7.1 and 2.7.2, the predicted crack driving force at a prescribed load (based on only a lateral shift in the stress-strain curve using standard plasticity theory, which would likely be seen as a logical modelling approach) may be significantly lower than the actual plastically damaged material. This would lead to non-conservative predictions of failure and is of high importance to structural integrity assessments.
5.2. J Integral Calculations without Residual Stress

To further validate the models, and as a necessary step in the extraction of J-Q failure curves, the crack driving force, quantified by J, was calculated from the simulations, using equation 18, and compared with the experimental results calculated using the methodology described in section 2.3.3. The results are shown in Figure 5-8 - Figure 5-13. The figures include the results from a weighted average J calculation through the thickness of the specimen, and the value at the central, mid plane. Both sets of results are included because generally the mid-plane value would be used in integrity assessments as this is assumed to be plane strain conditions.

The J values included in Figure 5-8 - Figure 5-13 is that taken at contour 33, where r=≈2mm. This was found to be, in all scenarios, a distance where a path independent J value could be extracted and is demonstrated by further series added to the figures, shown in purple (note a second horizontal axis at the top of the figures).

![Figure 5-8: Load v J comparison with test calculations and FE r v J, a/W=0.4 as received](image)
Figure 5-9: Load vs. $J$ comparison with test calculations and FE vs. $J$, $a/W=0.4$ pre-strained

Figure 5-10: Load vs. $J$ comparison with test calculations and FE vs. $J$, $a/W=0.2$ as received
Figure 5-11: Load v J comparison with test calculations and FE v J, a/W=0.2 pre-strained

Figure 5-12: Load v J comparison with test calculations and FE v J, a/W=0.05 as received
The figures show in general very good agreement with the test results for the J calculations. The largest percentage deviation in terms of load prediction of J is shown in Figure 5-13, where at the highest test load of 174kN, J is under-predicted by around 15%. As shown in the load/CGOD trace for this test scenario, at higher loads the FE load also over-predicts the CGOD values, demonstrating that the pre-strained material model may, at higher level of plastic strain show more yielding than is physically apparent. It was however necessary to apply consistent methodology between the as received pre-strained material models, and the excellent overall agreement against both the load/CMOD traces and J integral results validated the modelling procedures used.

What is also clearly apparent from observation of Figure 5-8 - Figure 5-13 is that there was better agreement between the average through-thickness J’s and the experimental results than those at the mid-plane, which was ~10% higher in the a/W=0.4 geometries and ~9% in the a/W=0.2 geometries, with almost no difference in the a/W=0.05 results. This enabled the decision to be made as to use the average value of J in all J-Q analyses. Although the stress field used to calculate Q in this analysis was calculated at the mid plane of the specimen, it was decided that as J is a global
parameter representative of the overall load on the structure; using average values as close as possible to the experimental results would produce most consistency when making industrial recommendations.

5.3. Residual Stress Results

Prior to completing load/CMOD calculations for the specimens which included residual stress, it was necessary to calculate the residual stress fields that the crack tip was exposed to. This was achieved through numerically simulating the side punching methodology described in section 3.3.3. As described in section 3.3.3, it was desirable to have residual stress fields that would be comparable with those that were used in a previous project, so that comparison could be made between these results and those described in (3).

The previous project validated (3) the residual stress fields against neutron diffraction measurements as described in 2.6.2, and the preliminary scoping calculations displayed in Figure 3-6 offered good agreement with the FE calculations from the previous project. However, as the final work used new material models, it was necessary to re-calculate the residual stress fields, and validate the modelling against the experimental data from the side punching. The experimental data available were load/displacement traces from the side-punching rig and an LVDT attached to the side-punching tooling so this was used to initially compare the modelling with real-world data. These comparisons are shown in Figure 5-14 and Figure 5-15.
As can be seen in Figure 5-14, there is maximum discrepancy of 12.9% between the highest load required to achieve the final indent of 0.5mm between the as received FE model and the experimental value. This is most likely explained through the orientation of the material. The tensile test programme, from which the material models used in this analysis were extracted, was carried out using specimens machined in the
rolling direction of the plate. The side-punching load was, however, applied perpendicular to the rolling direction. It is likely that this difference in orientation caused the discrepancy between the FE and experimental results.

As shown in Figure 2-3, the material displayed martensitic bands parallel to the rolling direction of the plate. As can be seen in Figure 5-15, the pre-strained material model appears to capture better the material behaviour than the as received, suggesting the as received material in this orientation has a higher hardening exponent. This would suggest the martensitic bands have a heterogeneous effect on the material behaviour; the flow stress is lower in the transverse direction to the martensitic bands.

Furthermore, the early yielding in the pre-strained material model does not accurately capture the behaviour in the transverse direction, showing that the effects of pre-straining on the dislocation mobility as discussed in section 4.1.3 is also heterogeneous and more prominent in the direction of the pre-straining. Finally, there can be seen variation in the maximum load in the tests which resulted in the variation in final indent caused. The maximum effect of this was in the pre-strained material, because of the higher loads required for compression to the indent. This was modelled as shown in Figure 5-15 as ‘FE Max’, where a maximum punch displacement was applied of 0.78mm (compared to 0.62mm to achieve the final indent of 0.5mm after relaxation) such that comparisons could be made between the residual stress fields to ascertain if the results from these tests could be used in the final analysis. These comparisons are shown in Figure 5-16.
The residual stress fields from all the analyses shown in Figure 5-16 are clearly comparable. It can be seen that there is minimal difference in residual stress field caused by differing between the maximum (0.65mm) and nominal indents (0.50mm) in the pre-strained simulations, validating the use of the test results which resulted in larger indents in the final analysis with the results from the model considering only the nominal indent. The stress fields all appear to follow similar behaviour to Hurlston’s, which enables comparison in the J-Q analysis. It was not imperative that the residual stress fields were identical for any J-Q comparison, because the stress fields are part of the calculations, it was merely desirable that they were comparable. For further validation, as in section 5.1, the simulation results were compared with the load/CGOD data output from the experiments, these are shown in Figure 5-17, and the good agreement shown further validated the modelling approaches and material models employed.
Figure 5-17: Load v CGOD plots for a/W=0.05 with residual stress test scenarios

Figure 5-18: Failure loads for all a/W=0.05 test scenarios, including specimens with residual stress and pre-straining

Figure 5-18 displays the failure loads for all of the a/W=0.05 test scenarios. As can be seen there is a significant reduction in mean caused by the residual stress in both the pre-strained material and as received, although it is much higher in the pre-strained material, (33% as opposed to 17%). The general reduction in mean is most simply
explained by there being a stress already acting on the crack, so overall one would expect there to be less resistance to fracture in the same geometry. Again, the postulated increase in microcracks caused by the pre-straining are likely playing a significant role, in that they would increase the likelihood of failure. The addition of the residual stress field to the increased amount of microcracks is increasing the probability of cleavage.

5.4. J Calculations including Residual Stress

For the residual stress scenarios, J was calculated experimentally using the exact same methodology as described in section 5.2. This methodology however has no consideration of the crack driving force from the residual stress field and therefore only provided details of the primary load at failure of the tested specimen. Therefore, finite element calculations were carried out using equation 43. These enabled calculation of path independent J values with the inclusion of the initial stress/strain fields. Comparison of the finite element predictions and the experimental values are shown in Figure 5-19 and Figure 5-20.

![Graph showing Load v J comparison with test calculations and FE r v J, a/W=0.05 residual stress](image)
As can clearly be seen in Figure 5-19 and Figure 5-20 there is relatively good convergence on a path independent value of $J$ at the chosen distance from the crack-tip of ~2mm, as was demonstrated in the simulations which did not involve residual stress. This is not achievable calculating $J$ in the usual way and demonstrates the validity of the use of equation 43 to obtain a path independent value of $J$ at this level of constraint. There is also a significant residual $J$ value in both cases. What is however also apparent from observation of the figures is that the output from the calculations does not accurately match the test results, as one may expect because the experimental results only consider the primary $J$ values.

The JEDI methodology described in section 2.6.4 was considered to account for the effects on non-proportional loading on the $J$ calculations, for the as received scenario. A comparison between the outputs $J$ calculated by both equation 43 ($J$ Original), and equation 46 (JEDI Non-proportional), with the test results summed with the residual $J$ value, calculated from the FE simulations, of 31kJ/m$^2$ are shown in Figure 5-21. Again, there is no clear prediction of the failure load vs experimental $J$ behaviour, although the prediction from equation 18 is closer to the results with the addition of the
$J_{res}$ values. Based on this, and because path independence had been demonstrated using equation 43, this methodology was used in all subsequent analyses.

![Figure 5-21: Load v J comparison with JEDI methodology](image)

The results presented here show some correlation between the experimental $J_c$ values and those predicted by the FE simulations, although the $J_{res}$ caused by the residual stress field, when incorporated into the calculations, seems insufficient in predicting $J$ through loading, when added to the experimental test results. This is however to be expected. The calculations of the $J$ values with residual stress are not comparable with the test results, as the experimental calculations do not consider the secondary stresses at all. Therefore, as an aside, it was also considered how the experimental failure $J_c$ values would compare to the, primary only, $J$ vs load behaviour for the same geometry. These comparisons are shown in Figure 5-22 and Figure 5-23.
As can be seen there is very good agreement between the evolution of $J$ through loading following the introduction of a residual stress field, despite this not being considered in the $J$ calculations. This is because in these scenarios the $a/W=0.05$ load CMOD traces are similar which can be seen when comparing Figure 5-3 and Figure 5-17. This therefore implies that residual stress does not have a significant effect on the primary $J$ v load behaviour.
5.5. Summary

- The FE simulations were validated against load/CMOD and J integral experimental data in all cases and offered good agreement.
- The pre-strained material displayed load-CMOD behaviour that could not be captured using standard plasticity theory derived material properties, but the material models used performed well.
- The mean through-thickness J values offered better agreement with the experiments than the mid-plane values, and are therefore used in all subsequent analyses.
- The J values with residual stress calculated using equation 43 showed path independence at r=2mm and are used in all subsequent analysis.
- The pre-strained material has a lower fracture toughness than the as received.

5.6. Key Results and Observations

- The reduced toughness is attributed to an increase in microcracks introduced through the pre-straining process.
- The introduction of microcracks would also explain the reduced elastic limit and modulus in the pre-strained material as they would act as new dislocation sources.
- Standard plasticity theory over-predicts the load vs J behaviour of the materials and is therefore non-conservative which is of key significance in integrity assessments.
6. J-Q Analysis Tools

To enable the calculation of the Q parameter it was necessary to create an idealised SSY field, achieved through boundary layer analysis as described in section 6.1. Also, because this project required an in-depth analysis of the methodology used to calculate the Q parameter, with several variables considered, an Abaqus-Python script was written to allow calculation of J-Q loading curves automatically. The code is described in section 6.2.

6.1. Boundary Layer Analysis

The boundary layer model is used to calculate the SSY field, which has no loss of constraint through loading such that Q=0. The stress field from this high constraint field would then be compared to that of the finite geometry specimens to quantify the reduction in constraint calculated using equation 40.

6.1.1. Geometry and Mesh

The geometry of the boundary layer model was chosen to be suitably large such that a semi-infinite crack in an infinite body which ensures SSY conditions could be justifiably approximated. In accordance with (35) and (3) a semi-circular 2-D mesh was created with a radius of 1m. The mesh contained 136 concentric rings and was biased, as with the finite geometry simulations, such that the smallest elements had a radius of 2.5µm (bias ratio=15000), and 20 spokes were created, as shown in Figure 6-1. Plane strain elements were used (CPE8R). Again, a crack was defined, with the crack front the focus of the mesh and the crack extension direction along the ligament.
6.1.2. Material Properties

The material properties used for the boundary layer model were identical to those used for the 3D analysis, although only the cryogenic values were used as it was this temperature that the stress fields and J integral through loading would be calculated. As discussed in section 7.2, it was necessary to consider using both the pre-strained and as received material models in the boundary layer analysis. The material behaviour had altered significantly through the introduction of the significant plastic strain field, so it was necessary to compare the results of using the original, as received SSY field, with one calculated using the pre-strained properties, when considering J-Q loading curves from the pre-strained test scenarios.

6.1.3. Boundary Conditions and Loading

Symmetry along the ligament was exploited in the model as with the fracture toughness test simulations. The loading was applied by a displacement: Using equations 32 it is possible to apply an exact crack driving force to the crack through a displacement field, where G=the shear modulus of the material (35):
\[ u_x = \frac{K_t}{2G} \sqrt{ \left( \frac{r}{2\pi} \right) \cos \left( \frac{\theta}{2} \right) (3 - 4\nu - \cos \theta) } \]

\[ u_y = \frac{K_t}{2G} \sqrt{ \left( \frac{r}{2\pi} \right) \sin \left( \frac{\theta}{2} \right) (3 - 4\nu - \cos \theta) } \]

The displacements calculated by equation 48 were applied to the outermost nodes of the model and ramped at in 100 equal increments up to a value of \( K_t = 200 \text{MPa}\sqrt{\text{m}} \). This was achieved using the equation command in Abaqus and applying a displacement boundary condition. This displacement field was also applied when plastic properties were applied, which allowed for calculation of an incremental J value to be calculated for comparison with the 3D analysis.

**6.1.4. Model Verification**

As can be seen in Figure 6-2 the application of the displacement boundary condition to the linear elastic case has provided the required stress intensity factor of 200MPa\(\sqrt{\text{m}} \). Also illustrated in the figure are the J values for the same simulation, with the elastic-plastic material properties applied, as can be seen there is negligible difference in the \( K_t \) values, except for very close to the crack tip (in a region very small compared to the 1m radius of the model) to guarantee SSY conditions.
Due to the numerous variables considered in the calculation of the Q parameter for the research presented in this thesis, such as the use of the average or mid-plane J values, and the use of the elastic limit or 0.2% proof stress as the reference stress ($\sigma_0$) in equation 40, it became apparent that using spreadsheets to calculate Q would be an inefficient use of time. Therefore, a script (Entitled ‘JAQ’- J And Q) was written using the programming language ‘Abaqus-Python’ that could fully automate the extraction of the J-Q plots and display them on the ‘Abaqus Viewer’ (22) interface. What follows is a breakdown of the pertinent points within the code, with visualisation of the outputs. The examples given are from various 2D and 3D simulations, and the overall process the code follows is shown below:

1) Extract the J-Integral and opening mode stress data for each loading increment from the FE software
2) Plot the normalised stress ($\sigma/\sigma_0$) versus the normalised distance ($r\sigma_0/J$)
3) Interpolate to find the normalised stress at a normalised distance of 2
4) Subtracting the normalised SSY field stress from that of the normalised stress at the normalised distance of 2, which is the Q value
5) Plotting the Q value against J for each increment of loading.

6.2.1. Code Breakdown

The JAQ code initially opens the .odb file, which is the Abaqus (22) output from the simulations. The user must enter the name of the specific file into the code. The code then displays the model’s mesh in the viewport and draws a path along the nodes ahead of the crack tip. Again, the user must enter the nodes for this path. After the path has been drawn the code calculates the length of the path in terms of numbers of nodes. Figure 6-3 shows the path as drawn by the JAQ code.

Following the drawing of the path and calculating its length the JAQ code extracts the opening mode stresses and J-Integral values for use in the J-Q loading curve calculations. In the case of the weighted average J values, these were approximated from vertical straight paths along the ligament of the crack, which offered the closest values to those calculated from spreadsheets as shown in section 5. The distance
through the thickness, from the mid plane of the specimen to where these values were taken, differed between test scenarios. In all cases there was negligible difference between the J values used and those obtained from the weighted average. As can be seen in Figure 5-20, the largest discrepancy between the mid plane and average J values was the $a/W=0.05$, pre-strained with residual stress scenario. Therefore, to validate this methodology, the weighted average J values compared to those used in the code for this scenario are shown in Figure 6-4.

![Figure 6-4: Maximum discrepancy between weighted average J values compared to those used in JAQ code](image)

To calculate the pertinent values for the J-Q loading curves ($\sigma/\sigma_0$ and $r_\sigma/J$), the variables were converted to lists in Abaqus-Python, transposed and multiplied as necessary. These were also reconverted into variables, which could be optionally read by the Abaqus (22) viewer and plotted in the graphical user interface (GUI) if necessary, as shown in Figure 6-5.
Following the production of the normalised stress vs. normalised distance plots the JAQ code then calculates the Q stress at each increment of loading. This is achieved through linear interpolation to find the normalised stress values at a normalised distance of 2. To calculate Q the boundary layer model was subjected to the JAQ code and the constant value of normalised stress at the normalised distance of 2 was obtained. This value is then subtracted from the normalised stress values at each of the load increments and assigned to another variable, whose second column is the appropriate J value. Finally, the J-Q variable is converted to Abaqus-readable form, which can then be plotted in the Abaqus Viewer GUI. An example is shown in Figure 6-6. The data can then be copied directly from the Abaqus interface for use in other data processing software. An example of the JAQ code for the a/W=0.4, as received scenario is included in the Appendix.
6.3. Summary

- A boundary layer model was created and verified to demonstrate SSY conditions for use in the Q calculations
- Abaqus-Python software was developed to automate the calculation of J-Q loading curves
7. J-Q Failure Curve Results

This section presents the results from the experimental and numerical analyses described in section 5, in the J-Q space. The results include J-Q loading curves, calculated from the numerical simulations using the JAQ code as described in section 6.2. They are populated with failure data from the fracture toughness tests, with the J-Integral values used in the Q calculation taken as the average through the thickness of the specimen. \( K_{\text{mat}} \) fracture toughness values are extracted from the high constraint master curve approach, described in section 2.4.2, and there is consideration of failure curves incorporating both the 5\textsuperscript{th} percentile (lower bound) and mean high constraint toughness values.

7.1. J-Q Failure Curve for As Received Material

The calculations were first carried out with the as received material only. It was possible to use these results to validate the J-Q loading curves, and justify decisions as to whether to use the 0.2\% proof stress or elastic limit as the normalising reference stress \( (\sigma_0) \) in the Q calculations (equation 40). The value of the reference stress used is typically the material’s 0.2\% proof stress as advocated by O’Dowd and Shih (35) and R6 (26). For the research presented in this thesis, and most importantly for the pre-strained material, more detailed consideration was required. Therefore, 2 options were considered, using either the elastic limit used in the material model or the mean measured 0.2\% proof stress measured during the tensile test programme, as the reference stress, \( \sigma_0 \). Only the cryogenic tensile data were considered because these material properties were those in effect in the actual fracture toughness tests. These are shown in Table 12, where the elastic limit was that calculated for the material models using the methodology described in section 4.1.3.
Comparison between the as received $a/W=0.4$ and $0.2$ J-Q loading curves were initially made to assess whether the change in $\sigma_0$ would have significant impact. In all cases the boundary layer analysis as described in section 6.1 was normalised by the reference stress under consideration, such that a different SSY field was obtained for each value of reference stress. The J-Q loading curves for both options are shown Figure 7-1.

As can be seen in Figure 7-1 there is significant variation in the J-Q behaviour when varying the reference stress, which becomes more prominent as $Q$ decreases, and plasticity increases. This type of analysis has not been reported in the literature and is
one of the novel aspects of this work. To enable a decision to be made as to which values would be used in the final analysis, comparison was made with the J-Q loading curves reported in Hurlston’s (3) work, which considered very similar geometries.

Hurlston (3) considered a/W=0.42 and 0.22 SENB specimens with the same B, W and S values (see Figure 2-10 for definitions). As the work presented in this thesis considered crack depths only 1mm shorter than these (nominally, in actuality the models used cracks that were only ~0.3mm shorter than Hurlston’s because the average fatigue pre-crack lengths were used in the simulations) it was decided that there was enough similarity to warrant comparison between J-Q results. What was discovered was that normalising the stress field by the elastic limit offered best agreement with Hurlston’s work. The J-Q loading curves were then also compared with $\beta_T L_r$ solutions for the test results, calculated using the methodology described in section 2.5.1. The $L_r$ values were equated to $J_c$ values when included in the figure to observe if the elastic $\beta_T$ constraint parameter could be used to validate the elastic plastic behaviour displayed by the J-Q loading curves. To ensure consistency with all the calculations the elastic limit was used in both the Q and $L_r$ calculations. These results are shown in Figure 7-2.

![Figure 7-2: J-Q loading curve comparison with Hurlston’s (3) and T-stress solutions (26)](image-url)
It is evident from observation of Figure 7-2 that there is very good agreement with the results and those from Hurlston’s (3) work, which validates the calculations of the J-Q loading curves, despite minor differences in material models and geometry. There is also relatively good agreement with the $\beta_T$ results, despite the fact they are derived from elastic analyses. In the a/W=0.4 geometry the results incorporating $\beta_T$ tend to 0, as opposed to the J-Q loading curve, where Q decreases slightly through loading. This indicates that the high constraint geometry is predicted by the T-stress to be pure small-scale yielding. The introduction of plasticity shows a slight diversion from this and that some plasticity develops reducing constraint slightly. In the lower constraint, a/W=0.2 scenario, there is an over-prediction of the constraint benefit without plasticity, which is not seen in the a/W=0.05 geometry, as shown in Figure 7-3.

![Figure 7-3: J-Q loading curve comparison with T-stress solutions (26), all as received geometries](image-url)
An explanation for the greater discrepancy between the $a/W=0.2$ scenario and the $a/W=0.05$ geometry is given in that the R6 (26) $\beta_T$ factor for SENB specimens is calculated using a polynomial fit to a series of FE analyses, some of which are published in (34), a compendium of solutions which itself is based on several references.

An example of the set of results from which the fit is calculated is displayed in Figure 7-4 (34). For the $a/W=0.2$ geometry the value of $\beta$ calculated from the polynomial fit (equation 39) is -0.292, which gives the T-stress values shown in Figure 7-3. Observation of Figure 7-4 shows the $\beta_T$ value could be as high as -0.2. This gives much better agreement between the Q and T stress values, as shown in Figure 7-5. This analysis further validates the Q calculations, and also demonstrates that using the elastic $\beta_T$ constraint factor predicts well constraint loss at high levels of plasticity, provided elastic plastic J calculations are used to describe the crack driving force vs $L_0$, behaviour. It is also demonstrated that, when necessary, exploring beyond standard handbook solutions is for constraint based fracture toughness predictions is necessary, as recommended in R6 (26).
It was also possible to further validate some of the calculations incorporating the $\beta_Q$ parameter, which again is derived from a series of 2D plane strain finite element simulations, considering Ramberg-Osgood curve fits to tensile data (36). Figure 7-6 is a comparison with the compendium solutions presented in terms of $\beta_Q$ vs $L_r$. The results presented in the compendium are presented as plots for various $a/W$ and strain hardening exponent values ($n$). The as received material displayed, in the hardening region, a strain hardening exponent of $n=6.6$ as shown in Figure 4-4, therefore the compendium solutions for $n=10$ and $n=5$ are shown for comparison. The material models used in the simulations presented in this thesis were not however standard Ramberg-Osgood fits, as described in section 4.1. The compendium solutions also used a different formula for the limit load of an SENB than that suggested in R6 (26) so again for consistency the load from the simulations was used to appropriately calculate $L_r$, and the elastic limit of 600MPa, as opposed to the 0.2% proof stress was used in all calculations. The compendium paper (36) from which the results in Figure 7-6 also considered the elastic limit as the reference stress.
What is observed in Figure 7-6 is relatively good agreement between the R6 advocated compendium solutions (36) for the \( Q \) calculations, in that the curves follow similar trends for all geometries. There are potentially several reasons for the discrepancies; most significantly the material models employed, but also mesh refinement, plane strain assumptions and loading simulation types. What is demonstrated, however, is that the tabulated, elastically calculated \( \beta_T \) factor calculations including in the main R6 document (26), as shown in Figure 7-5, seem to describe constraint loss better than the elastic plastic solutions.

In the elastic analysis \( \beta_T \) is constant. As can be seen in Figure 7-6, the \( \beta_Q \) factors calculated using the methodology presented in this thesis, also vary less through loading, than those presented in (36), which would account for the better agreement in Figure 7-5 than Figure 7-6. That said the general similarities in the trends of the curves implies the \( Q \) calculations are validated, and alongside the T-stress validation and the agreement with the previous work (3), confidence is demonstrated in the calculations and justifies the subsequent failure curve analysis.
To present the results from the tests in such a way that would be useful in structural integrity assessments it was necessary to apply statistical methodologies to assess confidence bounds for fracture toughness values. The master curve approach as described in section 2.4.2, is technically only suitable for high constraint toughness predictions, therefore the calculations considered the a/W=0.4 geometry. The experimentally calculated $J_c$ results were converted to $K_{jc}$ results using equation 20, and calculations were carried out according to ASTM 1921 (16) and described in section 2.4.2 to calculate the Wallin Weibull distribution, with a $K_{min}$ value of 20MPa$\sqrt{m}$, is shown in Figure 7-7, where it is compared with standard Weibull distribution median rankings.

![Figure 7-7: Wallin distribution vs median Weibull rankings, a/W=0.4](image)

It is evident from observation of Figure 7-7 that the Wallin distribution calculated using equation 29 describes relatively well the scatter in results and enables calculation of the 5th, 50th and 95th percentile fracture toughness values. The 5th percentile value was compared with results from a previous programme of tests (41) which supplied a temperature dependent prediction of the 5th percentile toughness value, which at -140°C equated to 57.99MPa$\sqrt{m}$. The value predicted from the Wallin distribution displayed in Figure 7-7 was 59.30MPa$\sqrt{m}$, the negligible difference demonstrating the validity of the test programme and the calculations used. Using the master curve approach, it was also possible to extract a value of $T_0$, calculated as -136.16°C. Again, this
agrees well with a previously published value of -135°C (68), further validating the test programme.

Although the master-curve approach for the calculation of $T_0$ is strictly limited to high constraint specimens, it was decided that any J-Q failure curve may be described in terms of the 5th, 50th and 95th percentile confidence bounds. Therefore, the same comparison between median rankings and the Wallin distribution was made for the lower constraint geometries, and are displayed in Figure 7-8 and Figure 7-9.

Figure 7-8: Wallin distribution vs median Weibull rankings, $a/W=0.2$

Figure 7-9: Wallin distribution vs median Weibull rankings, $a/W=0.05$
Note that the same scale is used for \( a/W = 0.4 \) and 0.2, whereas the range of \( K_{ic} \) is significantly higher for the \( a/W = 0.05 \) test results. This implies that, in these higher constraint geometries, the standard single-parameter fracture toughness approach stands up well, it is in the very low constraint geometry where constraint plays a much more significant role in any toughness consideration; the fracture toughness values are altered significantly. Furthermore, examination of Figure 7-7 - Figure 7-9 demonstrates that the Wallin distribution describes suitably well the data for the \( a/W = 0.2 \) and \( a/W = 0.4 \) geometry. Considering the \( a/W = 0.05 \) results it can be seen that 2 of the test results lie significantly below the others in terms of \( K_{ic} \), which skews the distribution. There was however enough agreement that it was decided that it would be suitable to use the 5th, 50th and 95th percentile toughness values in each test scenario, although it would be more accurate to use the mean test results for the main comparisons.

The recommendation in R6 (26) for constraint based toughness prediction is to correct the high constraint toughness values \( K_{mat} \) to a constraint based toughness value \( K'_{mat} \) using equation 49:

\[
K'_{mat} = K_{mat}[1 + \alpha(\beta L_p)^m]
\]  

[49]

This approach considers linear elastic stress intensity factors, to correct high constraint \( K_{ic} \) toughness values to those at lower constraint, with \( K_{ic} \) the linear elastic stress intensity factor calculated from the failure load, as opposed to the elastic-plastic \( K_{ic} \) value. However, for the academic purposes of the work presented in this thesis it was decided that the \( K_{ic} \) test results would be used to assess the viability of such a curve, and, as justified above, the mean test values would be considered in the initial failure curve analysis. Therefore equation 49 would be converted to equation 50 to incorporate the elastic plastic test results, and the Q stress calculated from the simulations, as described in section 6:

\[
K'_{ic,mean} = K_{mat,mean}[1 + \alpha(-Q)^m]
\]  

[50]
where $K_{\text{mat, mean}}$ would be taken as the mean toughness value from the $a/W=0.4$ test results, in this instance, $Q$ would be treated as 0, which would assume high constraint, SSY conditions for this geometry. Although the J-Q loading curve, as shown in Figure 7-5, does tend slightly away from zero, it was necessary to have a high constraint mean toughness value for use in equation 50, and the good agreement with other project’s (68), (41), high constraint $T_0$ and $K_{\text{mat}}$ predictions for this material validated this assumption. The J-Q mean toughness curve, for as received material, is shown in Figure 7-10. Also included in the figure are the curves calculated for the 5th and 95th percentile $K_{\text{mat}}$ values, calculated using the master curve approach at high constraint, and the same $\alpha$ and $m$ values.

![Figure 7-10: J-Q mean failure curve, as received material, $\alpha=0.62$, $m=1.54$](image)

R6 advocates the calculation of $\alpha$ and $m$ in several ways, the first option being “testing specimens having different geometries in the range of $\beta L$ of interest” (26). Therefore, the decision to apply curve fits, defined by equation 50, Could easily be justified. It was decided that for simplicity, a single value of $\alpha$ and $m$ would be used when considering the lower and upper bound 5th and 95th percentile toughness curves. As can be seen from Figure 7-10 these curves capture all of the test results, although the
upper bound appears to over-predict the higher $J_C$ values as constraint is reduced. What was observed from the results was that different fits could be applied to the upper and lower bound toughness values, although this approach would prove of limited usefulness in industrial structural integrity assessments, especially the upper bound toughness prediction, as this value would not be used for incredibility of failure scenarios.

Observation of Figure 7-10 shows the lower bound curve, using the same $\alpha$ and $m$ values as those used to predict the mean toughness benefit through constraint loss predicts well the toughness with decreasing constraint. What was observed was that using fits to the lower bound 5th percentile toughness values, extracted from the Wallin distributions as shown in Figure 7-8 and Figure 7-9, over-predict the mean toughness benefit through decreasing constraint, as can be seen in Figure 7-11.

This analysis is less accurate as it can be seen that despite fitting the curves to the lower bound value of toughness, there is more significant increase in toughness through constraint than is apparent at higher loads, therefore if an engineer were to try and calculate the specific toughness of the material at a prescribed load, the results
would be non-conservative. This justifies further the use of the mean values for defining the failure curve.

Comparison was made at this stage with Hurlston’s (3) results, which used Beardsmore’s (43) methodology to use the RKR failure stress criteria to derive a J-Q failure curve, as described in section 2.6.5. That curve showed that in higher constraint geometries the RKR approach could be used to predict mean failure, with the critical stress taken as the mean stress from the test failures, calculated at a distance $r_c=0.2\text{mm}$ using FEA. The critical stress at failure was calculated as $3.10\sigma_y$ from Hurlston’s analysis, which is defined as a material property, therefore, as the same material and very similar material models used for the calculations presented in this thesis, the failure curve derived using RKR should align well with that presented in Figure 7-10. Comparison is shown in Figure 7-12. Hurlston’s failure points are also included to demonstrate where they lie compared to the failure curves derived in this thesis.

![Figure 7-12: J-Q failure curves, $\alpha=0.91$, $m=1.39$, compared with RKR derived curve with Hurlston's failure points included (3)](image-url)
As is evident in Figure 7-12 the failure curves derived using the R6 equations and the tests presented in this thesis do not align with the RKR prediction derived using Hurlston’s (3) RKR methodology, which in principal should be similar as the same material is was used. What can also be seen is that the curves derived using the R6 equations contain most of the J-Q failure points from Hurlston’s work, although the lower bound curve may be excessively conservative, due to the generally higher J values at failure observed in Hurlston’s results which is attributed to the reduced scatter due to fewer tests being carried out. The general positioning of all the failure points however does provide confidence in the material-specific failure curve.

To better assess if the RKR curve could be used in a similar way to Hurlston (3) at very low constraint, consideration was given to the \( f_s \) parameter. The value used in Hurlston’s (3) work was 3.10 \( \sigma_y \), and the range between all the scenarios considered was merely \( \pm 0.04 \sigma_y \). Therefore, considering a single \( \sigma_f \) value as a material property for use in the RKR approach is justified as the range is small. The same calculations were made for the as received results presented in this thesis and are shown in Table 13. The \( \sigma_y \) value used in the calculations was the as received elastic limit, 600MPa.

<table>
<thead>
<tr>
<th>Test Scenario</th>
<th>a/W=0.4</th>
<th>a/W=0.2</th>
<th>a/W=0.05</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mean ( J_c ) Value (kJ/m(^2))</td>
<td>40.45</td>
<td>42.59</td>
<td>74.24</td>
</tr>
<tr>
<td>( \sigma_f/\sigma_y )</td>
<td>3.25</td>
<td>3.23</td>
<td>2.95</td>
</tr>
</tbody>
</table>

Table 13: RKR critical stress values, as received condition

The results in Table 13 indicate that, as with Hurlston’s (3) results there is very good agreement in the higher constraint conditions. This would lend weight to using an RKR defined failure curve in this region in the J-Q space. However, in the very low constraint geometry difference in critical stress is an order of magnitude larger, demonstrating that the simple RKR methodology may require further calibration for these geometries. A further explanation for the discrepancy between the failure curves
shown in Figure 7-12 is that the RKR derived curve, based on Beardsmore’s methodology (43) is dependent on using a standard Ramberg-Osgood fit to tensile data, which was not suitable for the modelling presented in this thesis (see section 4.1).

### 7.2. J-Q Failure Curve for Pre-strained Material

To consider a J-Q failure curve for the pre-strained material the same methodology was applied as in section 7.1. The most significant decision involved with this was using the elastic limit, as opposed to the 0.2% proof stress, as $\sigma_0$ in the Q calculations, to enable comparison with the as received material, the implications of this are discussed further in section 8. Comparison with the various handbook solutions, as were performed with the as received material, were omitted at this stage, due to the non-conventional material models employed: there would be poor agreement with the handbook solutions as these were obtained from standard Ramberg-Osgood material models.

As this type of analysis provides a novel aspect of this work, the SSY field, obtained from the boundary layer model, was a significant consideration. It could be assumed by an engineer making an assessment that the pre-strained material properties need only be considered in the actual finite geometry modelling, and that the SSY field would be suitably obtained using as received material properties. It was necessary to ascertain whether this would be a suitable assumption, or whether the SSY fields needed recalculating as an entirely new material. The results of this comparison are shown in Figure 7-13. For this initial analysis, when deciding the details involved with the Q calculations, the high constraint geometry was considered, as with the discussion for the as received material, presented in Figure 7-1. Also, as with all of the J-Q calculations, consistency was maintained so for this comparison, the elastic limit was used as the reference stress, which for the pre-strained material = 400MPa.
As can be seen in the figure the use of the as received stress field as the reference SSY field against which Q is calculated causes the J-Q loading curve for the pre-strained material into a region significantly higher than zero. Whilst even using the pre-strained material properties for the boundary layer model also causes a value of Q slightly above 0 at low loads, having such a highly positive value of Q would not be desirable. The current convention is that any constraint-based benefit in toughness should be considered where Q < 0, and for use in this analysis this is necessary such that equation 50 stands, as using the R6 curve fitting methodology, $K_{\text{mat}} = K_{\text{mat}}$ when Q > 0. This is again an important consideration in structural integrity assessments as it demonstrates that when considering a failure curve for material which has seen significant plastic strains, it is necessary to reconsider the SSY field with the same material properties.

To discern a failure curve, it was again necessary to consider the master-curve high constraint toughness predictions. This type of analysis has not been reported in the literature previously, so it was necessary to compare toughness and $T_0$ values between the pre-strained and as received materials. The high constraint, Wallin distribution for the pre-strained, a/W=0.4 scenario is presented in Figure 7-14, with comparison of pertinent toughness values demonstrated in Table 14.
As can be seen there is a significant reduction in toughness due to the pre-straining process, alongside an increase in $T_0$, the physical reasons for which are discussed in section 5.1. The purpose of Table 14 is to allow quantification of actual toughness values that would be used in industrial structural integrity assessments. The values show a reduction in 5% toughness ~25%, which is of extreme significance in integrity assessments. As can be seen this results in an increase in $T_0$ of 31°C, which again would be of significance in proving incredibility of failure at shutdown temperatures combined with the potential increase in $T_0$ caused by radiation embrittlement in RPV components.

Table 14: Toughness values for $a/W=0.4$ geometry
It was possible to make comparisons with the results from the model presented in section 2.7.4. The yield strain (at the 0.2% proof stress) in the as received material was ~0.4%, and the pre-strain was 5.5%. As can be seen in Figure 2-35, this would equate to a reduction in toughness through pre-straining to ~15% of the original. This value considers the toughness values in terms of $J_c$. The results in Table 14 when converted to $J_c$ values show the pre-strained toughness is 57% that of the as received material, demonstrating the mechanistic modelling presented in section 2.7.4. is insufficient to describe the actual reduction in toughness observed in the tests presented in this thesis. The model considers standard plasticity theory and a critical strain criterion, and equates the increase in yield stress and reduction in ductility to a reduction in toughness. There is no mention of microcracks, or reduced modulus or elastic limit as has been observed in the material used for the tests presented in this thesis. This demonstrates that using simple mechanistic modelling based on standard assumptions is inaccurate in predicting the toughness reduction following pre-straining of ferritic steels.

When comparing the results of the local approach method also described in section 2.7.4 there can be seen better correlation with the test data. The results displayed in Figure 2-36 show that the model predicts that the mean failure load with 5% pre-strain is 84% of the as received condition. When considering load, as opposed to $J_c$ as above, the difference observed in the tests reported in this thesis was a 31% reduction in mean failure load following pre-straining. This demonstrates better agreement than the mechanistic model, although it is clear there is still a large discrepancy and that modelling approaches in the literature do not accurately predict toughness reduction with pre-straining.

To investigate the applicability of the J-Q failure curve presented in Figure 7-10 the same methodology was applied to the pre-strained results as described in section 7.2, such that the only difference between the two failure curves is that the original, high constraint toughness was that extracted from the pre-strained results. This failure curve is presented in Figure 7-15. The results presented in the figure are calculated using the elastic limit of 400MPa as the normalising reference stress, which again follows the same methodology as with the as received material.
It is evident from Figure 7-15 that using the same methodology to calculate the J-Q failure curve, for both the as received and pre-strained material, it is possible to predict the fracture toughness with decreasing constraint. As can be seen the mean toughness values are within range of those predicted from the $\alpha$ and $m$ values used to derive the as received mean failure curve. There is a maximum discrepancy of mean toughness of 10%, in the very low constraint geometry, which means the toughness prediction in that region is slightly non-conservative. This is however a very good result, demonstrating that through simply changing the high-constraint toughness value it is possible to derive a J-Q failure curve from the as received material, to calculate constraint based fracture toughness values for material which has seen significant plastic strains.
To predict failure with residual stress, the J values calculated using equation 43 from the finite element simulations were incorporated into the Q calculations exactly as they were for the other scenarios, with the J values used the weighted averages displayed in Figure 5-19 and Figure 5-20 to remain consistent with the other calculations. The material in this case had been subjected to a significant stress intensity factor caused by the double side punching process, at room temperature, which is clearly on the upper shelf of the fracture toughness curve (T₀ had been calculated at -136°C, see section 7.1). The pre-loading K₂ values were calculated from the J values as calculated from the FE simulations. It was shown that there was no reduction in J after releasing the boundary conditions following the side-punching and during cooling in the FE simulations, meaning that the side-punching process, as described in sections 3.3.3, exerted an LCF warm pre-stressing process, as described in section 2.4.3. The pertinent values to calculate the toughness benefit used in equation 34, for both the as received and pre-strained material are shown in Table 15.

<table>
<thead>
<tr>
<th></th>
<th>K₂ (MPa√m)</th>
<th>Kᵣ (MPa√m)</th>
<th>Toughness benefit = Kᵣ - Kₘ₅% (MPa√m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As Received</td>
<td>81.9</td>
<td>90.8</td>
<td>31.5</td>
</tr>
<tr>
<td>Pre-strained</td>
<td>65.1</td>
<td>71.7</td>
<td>27.7</td>
</tr>
</tbody>
</table>

Table 15: Warm pre-stress parameters

The toughness benefit was calculated by subtracting the 5% Kₘ₅% value from the fracture toughness following the warm pre-stress benefits, Kᵣ, as the 5% Kₘ₅% value is commonly used as the fracture toughness of a material, and this would be the methodology used in industrial applications such as R6 (26). It was assumed that this increase in toughness would also apply to the mean toughness values, for use in the failure curve calculations. This is considered conservative, and therefore applicable,
based on the results displayed in Table 14. As can be observed when considering the high constraint toughness values, there is a difference between the 5% toughness value of the as received and pre-strained values of ~15MPa√m, increasing to ~28MPa√m for the mean values. If this is considered as a reference as to how any benefit in toughness affect the material, it is justified that using an additive approach, considering only the benefit to the 5% K$_{mat}$ toughness, is clearly the most conservative approach. Furthermore, the warm pre-stress benefit would be very similar in all cases for the same test scenario, so would not provide any greater benefit to tests which failed at higher loads than those at lower loads, so the approach is justified. Therefore, the toughness benefit was included as described above, as an addition to the mean failure curve, and therefore failure curves, for the both the pre-strained, and as received material. Again, the original $\alpha$ and m values, used to extract the initial, as received curve, were applied. These results are shown in Figure 7-16 and Figure 7-17.

![Graph showing J-Q mean failure curve, as received material with residual stress, $\alpha=0.62$, m=1.54](image-url)
As can be seen, for both materials, the mean toughness is again predicted very well using the curve fitting parameters derived from the initial as received J-Q failure curve, again demonstrating the applicability of the J-Q failure methodology. For all test scenarios, it has been possible to predict the mean toughness values for test programmes of 8 tests or more to within 10%. This facilitates an answer the initial hypothesis of this thesis, which was: Does the J-Q failure methodology apply to material which has seen both residual stress and significant plastic strains? As has been demonstrated, the J-Q failure curve is applicable for the material subjected to the various load history effects under consideration, with the only major variable being the high constraint toughness value used in the calculations.

7.4. Stress Triaxiality Observations

Hurlston’s (3) work demonstrated that there was a similar failure curve which included residual stress, with J values calculated using the same methodology as described here. There is no discussion of a benefit in toughness from the warm pre-stressing process,
nor is there description of the $J_{res}$ values introduced through the simulation of the side punching. It was therefore necessary to recreate as well as possible the simulations that might have been carried out to further corroborate the failure curve methodology postulated in this thesis. As was the case in section 7.1, because the material was the same and the very similar geometries used, it was decided to apply the side-punching simulation steps to the a/W=0.2 and 0.4 geometries, with the side punch positioned the same distance ahead of the nominal crack tip, as with the a/W=0.05 scenarios. The J-Q loading curves, with both the average J values, and the mid-plane values (used by Hurlston) are shown in Figure 7-18.

As can be seen from observation of the loading curves there is the best agreement between Hurlston’s (3) results and the J-Q loading curves calculated using the mid-plane J values, which one would expect because those results also used the mid-plane value. Where the average J value is used there is a greater discrepancy, most apparent in the higher constraint geometry (a/W=0.2). What can be seen throughout,
and was introduced in section 2.6.5, is that there is, provided one uses the same J calculation method, a crossover where Q for the a/W=0.2 scenario is less negative than for a/W=0.4. This is not seen in the as received cases, which follow the convention for constraint in that Q decreases with crack length. What is apparent is that the inclusion of a residual stress field changes the J-Q behaviour, when considering crack depth. This is illustrated in Figure 7-19, which includes all of the J-Q loading curves, with and without residual stress. In Figure 7-19, all of the J-Q curves are calculated using a consistent methodology, in that the J values, calculated using equation 43, are the weighted average through the thickness of the specimen, and the elastic limit is used for the reference stress, σ₀, to calculate Q using equation 31.

Figure 7-19: J-Q loading curves, as received material, all geometries

What can be observed in Figure 7-19 is that in the medium and very low constraint geometries, Q increases with the introduction of a residual stress field, whereas in the high constraint geometry, the value of Q is lower at the same load. Constraint is defined in R6 as a reduction in hydrostatic stresses at the crack tip (26),
therefore to better understand the reasons for this, a single element analysis was carried out. This element chosen was a distance \( r_c = 0.2 \text{mm} \) from the crack tip (to correspond with Hurlston’s (3) RKR analysis as described in section 2.6.5), at the mid-plane to assess how stress triaxiality, defined by the ratio of hydrostatic to Von Mises equivalent stress, was affected by the introduction of a residual stress field. For this investigation, the mid-plane J-values were used because these were used by Hurlston (3), and the crossover described above is more prominent using the mid-plane J values.

![Figure 7-20: Stress triaxiality at distance \( r_c = 0.2 \text{mm} \) from crack tip, as received material, all geometries with and without residual stress fields](image)

It is clearly evident from observation of Figure 7-20 that the crack tip stress triaxiality is significantly affected by the introduction of a residual stress field in all scenarios. Initially examining the as received scenarios; each of the geometries, at early loading, begin with a similar value of stress triaxiality. There is a ‘bedding in’ period where there is a small drop in triaxiality, and then in each case it increases through loading. Subsequently, as load increases, there is less increase in hydrostatic stress with a shortening in crack depth, meaning that constraint is lost, as would be expected.
For the cases with residual stress, the $a/W=0.2$ and $0.4$ geometries behave in a very similar fashion, although again the higher constraint geometry shows a greater increase in triaxiality than with the $a/W=0.2$ scenario. This implies that there are lower hydrostatic stresses in the scenario with residual stress than without in the $a/W=0.4$ scenario, and the converse effect in the $a/W=0.2$ scenario. This explains why the introduction of the residual stress field causes a crossover in the J-Q behaviour for the $a/W=0.4$ and $0.2$ geometries.

For the $a/W=0.05$ scenarios, the introduction of the residual stress field has a similar, but much more pronounced effect. As has been described throughout this thesis, the residual stress field ahead of the crack was comparable, after fatigue pre-crack introduction, to those introduced in Hurlston’s work, according to FE simulations of the test processes. What is apparent however from this analysis is that the similar stress field has a much more pronounced effect on crack tip triaxiality in very low constraint scenarios. Immediately after the side punching, there are very high hydrostatic stresses, which contribute to a higher $J_{\text{res}}$ value (demonstrated by the different starting positions of the curves with residual stress), and much higher stress triaxiality than all of the other scenarios. Following the introduction of the primary load, however, the principal, crack opening stress contributes much more significantly and the stress triaxiality reduces. It does not however go as low as with the as received scenario and therefore still displays higher constraint, as can be seen in Figure 7-20 and Figure 7-19 by the fact that the Q values are higher in the loading curves with residual stress.

It was also prudent to assess the effects of decreasing crack depth on stress triaxiality, indeed in some works (i.e. (34)) Q is calculated not using the opening mode stress as has been the case for this thesis, but hydrostatic stresses, which represents a measure of triaxiality at a normalised distance from the crack tip through loading. Visualisation of stress triaxiality was therefore obtained in the form of contour plots, at the closest loading increment to the mean failure load for the geometry in question, in the as received condition. The results are shown in Figure 7-21.
Observation of the contour plots shows that triaxiality is significantly affected by crack geometry. As crack size decreases, the size of the contours also reduces, demonstrating a steeper reduction in triaxiality ahead of the crack. It is also evident that the higher constraint geometries have similar triaxiality fields at their mean failure loads, which again are very similar (within 2.14kJ/m², see Table 13). At very low constraint the stress triaxiality at mean failure is significantly lower, despite the average Jₑ values being much higher. This corroborates what has been previously observed in that constraint
does not change significantly in these geometries, and that the use of a failure curve for this material in this region of J-Q space may not be necessary; 1 parameter toughness would suffice. This also demonstrates that using stress triaxiality alone as measure of fracture toughness would not be suitable, as there is no similarity in the fields. As with Q, it would be necessary to have a correction process to equate triaxiality to toughness in the low constraint geometries.

This analysis has given a more detailed, constraint based, insight into the crack tip stress fields for different initial loading conditions, than just considering the crack opening mode, stress based, Q parameter which has formed the main body of this thesis. Whilst the J-Q failure curve methodology has been shown to enable toughness prediction for the various load history scenarios tested, both in this work and Hurlston’s (3), this analysis has enabled stress based reasoning for the actual positioning of the curves in J-Q space, and has demonstrated that: At high constraint (a/W=0.4), the hydrostatic stresses do not increase with loading as much as they would with no residual stress field, but still increase. As crack depth reduces (a/W=0.2), there is a point where the increase in triaxiality is almost equal through load, regardless of the initial triaxiality state. At very low constraint (a/W=0.05), the stress field begins at much higher triaxiality than the lower constraint geometries, but the introduction of primary load contributes much more significantly than the hydrostatic stresses and constraint drops rapidly. With no initial stress field the hydrostatic stresses and therefore triaxiality are lowest, making this the lowest constraint scenario under consideration.

7.5. Summary

- J-Q loading curves were calculated for the as received material and compared with handbook solutions, where good agreement was obtained.
- A J-Q failure curve was derived using the R6 (26) methodology, whereby the α and m parameters were calculated to best fit the mean toughness data for the as received material.
• A failure curve was created derived using the same $\alpha$ and $m$ parameters for the pre-strained material and demonstrated good prediction of the mean toughness values, when incorporating the high constraint pre-strained toughness value.

• The side-punching process had introduced a significant residual J value into the specimens, creating a WPS benefit in toughness.

• When the WPS benefit was added to the original fracture toughness in both the as received and pre-strained materials the original failure curve predicted well failure in the low constraint specimens with residual stress.

• A single element analysis close to the crack tip was carried out to demonstrate why there is a switch from reduction to increase in $Q$ with increasing crack depth, showing that a significant residual stress field actually decreases stress triaxiality (and therefore constraint) in high constraint conditions.

7.6. Key Results and Observations

• The elastic $\beta_1$ constraint parameter provided a good description of elastic-plastic constraint loss in the as received material, if $L_r$ is equated to the appropriate elastic-plastic crack driving force quantified by J.

• Simple mechanistic modelling considering standard plasticity theory drastically over predicts the reduction in cleavage fracture toughness in pre-strained material.

• The best failure curve predictions were obtained with the mean toughness values.

• The use of the mean toughness values is conservative in predicting the lower bound values.
8. Discussion

To discuss the work presented in this thesis it was necessary to deconstruct the methodology applied to enable the calculation of the J-Q failure curves presented in section 7, in order to ascertain if using simplified industrial methodology could be used, more quickly and cheaply, to predict the failure behaviour observed. As was shown in section 7.1, the $\alpha$ and $m$ parameters used to define the as received (and all subsequent) failure curves, were calculated through examination of the test results from 3 different levels of constraint for the as received material. This required an extensive test programme, which was costly in terms of both materials and time. R6 (26) does advocate several other options for determination of the $\alpha$ and $m$ parameters, including using local approach methodologies or consulting a series of look up tables (69), which provide $\alpha$ and $m$ values based on $E/\sigma_y$, the strain hardening exponent $n$, and the Beremin (31) shape parameter, $m_B$, (described in section 2.4.4).

The look up tables are for use with the R6 methodology and would therefore consider the 0.2% proof stress, as opposed to the elastic limit as the reference stress, $\sigma_0$. Therefore, the first simplification involved with the deconstruction of the methodology was to reassess $\alpha$ and $m$ using the 0.2% proof stress, in both the as received and pre-strained materials. As was discussed in section 7.1, applying the curve fits to the mean values, as opposed to 5th percentile toughness values, was conservative, so this approach was continued for this analysis. The results are shown in Figure 8-1 and Figure 8-2. For clarity, the scales used are the same as those used when presenting the results considering the elastic limit as $\sigma_0$, for easy comparison with Figure 7-16 and Figure 7-17.
Figure 8-1: J-Q failure curve, as received material, $\sigma_0=0.2\%$ proof stress, $\alpha=0.82$, $m=1.47$

Figure 8-2: J-Q failure curve, pre-strained material, $\sigma_0=0.2\%$ proof stress, $\alpha=2.5$, $m=2.0$
As can be seen it was necessary to use different $\alpha$ and $m$ parameters to achieve good fits with the mean toughness data, between the as received and pre-strained materials (without residual stress). Their values are noticeably different which implies that using the 0.2% proof stress as the reference stress in the $Q$ calculations predicts the toughness benefit through constraint loss as that of an entirely new material. Using the elastic limit enables an engineer to consider the material differently: it is possibly to predict toughness benefit through that achieved when testing the as received material. The physical explanation for this is that the increase in toughness through constraint loss is caused by the creation of plastic zones at the crack tip. Therefore, it is apparent that consideration of the very onset of plasticity, as opposed to that after 0.2% plastic strain, allows more accurate prediction of the toughness benefit as the plasticity increases. Provided this methodology is applied consistently between materials, is can be seen that the toughness increase, in $J$-$Q$ space, also remains consistent.

When considering the residual stress prediction, it can again be seen that in the analysis of the pre-strained material there is less accuracy than when the elastic limit is used as the reference stress. Whereas with the as received material the mean toughness is again well predicted by the results without residual stress, the pre-strained residual stress toughness predictions are excessively conservative. As explained in section 7.3, the introduction of the residual stress field significantly increases the stress triaxiality at the crack tip, most predominantly in low constraint geometries. As primary loading is added to the residual stress field, this triaxiality decreases, decreasing constraint. This would clearly be influenced by the development of the plastic zone through primary loading. Using the 0.2% proof stress as the normalising reference stress, at which plasticity develops, causes, in all cases, a prediction of less plasticity and therefore less constraint loss. What is observed in the pre-strained with residual stress scenario is that the reduced plasticity under-predicts the constraint benefit in toughness. Where this early plasticity is less prominent, as with the as received material (demonstrated well in Figure 4-8 - Figure 4-11), the scenarios without residual stress predict well those with; because the elastic limit and 0.2% proof stress are similar. In the pre-strained case, however it is again necessary to consider the low-strain plasticity to accurately predict its effects on fracture toughness.

The overall aim of the thesis was to assess whether a $J$-$Q$ failure curve could be categorically determined, and if so what parameters would be needed as input. It has
been shown that the toughness benefit through constraint loss in material which has seen significant plastic strains and or a tensile residual stress remote from the crack could be well predicted by using the J-Q methodology, provided the engineer has detailed knowledge of the low-strain tensile behaviour of the material, specifically the stress at the onset of plasticity. Elastic-plastic finite element analyses are generally expensive in terms of an engineer’s time, and may also be computationally expensive. Therefore, it was pertinent to remove as far as possible the need for finite element calculations.

The FAD approach to structural integrity assessment is described in section 2.4.1. To incorporate the constraint benefit into the FAD, there are two recommendations in R6. An engineer can either include the constraint corrected toughness $K_{C_{\text{mat}}}$ in the calculation of $K_r$, which is calculated using equation 51:

$$K_{C_{\text{mat}}} = K_{\text{mat}} [1 + \alpha (-\beta L_r)^m], \quad \beta < 0$$  \[51\]

Or the failure curve $f(L_r)$ of the FAD can be corrected using the $\alpha$, $\beta$ and $m$ parameters through equation 24 as described in section 2.4.1. Either of these methods is advised within the R6 framework therefore, for clarity and to allow all scenarios for each material to be plotted against a single failure curve, equation 51 was used to correct the material fracture toughness.

The aim of this discussion was to assess if a simplified methodology such as those described in R6 could be used to predict failure as well as the J-Q failure curves, therefore, the as received scenarios, with the constraint corrected toughness calculated using equation 51, were used in the calculation of $K_r$. In both cases the elastic limit was used in the calculation of both $L_r$ and $K_r^p$. To calculate $K_r$ for the residual stress case, the following methodology was applied: $K^s$ was calculated as $K_r^p$, in that the elastic plastic $J$ calculated from the side punching process was converted to $K_i$ using equation 20. As recommended in R6 (26), a multiplicative plasticity factor $V$ was included in the calculations such that:
\[ K_r = \frac{K_l^p + V K_f^5}{K_{mat}} \]  \tag{52}

In the case with residual stress, \( K_{mat} \) was taken (uncorrected) as that with the WPS benefit, as described in 7.3, and in all cases where corrected is indicated, in Figure 8-3 and Figure 8-4, \( K_{mat} = K_{c,mat} \). In the cases without residual stress the 5% toughness from the master curve analysis were used as the \( K_{mat} \) value, as shown in Table 14. In all cases the value used was that tabulated in R6 (26), derived from the elastic T-stress, as was used in the calculations displayed in Figure 7-3. This was included because it would be simpler than using the changing values of \( \beta Q \), shown in Figure 7-6, and the tabulated T-stress \( \beta T \) factors actually offered better agreement overall than the elastic plastic values in the R6 advocated compendium (34). Despite the a/W=0.2 case offering better agreement with a higher value of \( \beta \), the original tabulated value was used in this case, so it could be said the R6 methodology had been followed accurately. \( \alpha \) and \( m \) were again calculated to best fit the data such that the lower bound failure points were as close as possible to the failure curve, as the lower bound toughness values were used in the \( K_r \) calculations.

For the as received case, as A533b is a standardised RPV steel, a generic Option 1 failure curve was used to assess if this simple methodology would be sufficient in assessing toughness benefit with constraint loss. This is shown in Figure 8-3. Because the material’s tensile behaviour was altered significantly through the pre-straining process, the second FAD, as shown in Figure 8-4, offers comparison between 2 material specific Option 2 FAD curves. This was to enable further discussion of the impact of using the either the elastic limit or 0.2% proof stress in calculations regarding the fracture toughness of the pre-strained material.
Figure 8-3: As received FAD, $\alpha=0.75$, $m=0.65$

Figure 8-4: Pre-strained option 2 FAD, $\alpha=0.75$, $m=0.65$
There are several points for discussion raised from observation of Figure 8-3 and Figure 8-4. First it can be seen the $\alpha$ and $m$ values differ significantly from those used in the J-Q elastic plastic methodology. This is simply explained because the FAD analysis considers linear elastic behaviour. The purpose of the FAD is that it removes the necessity for an engineer to carry out complex J integral analysis, and the plasticity effects are accounted for in the shape of the failure curve. The methodology presented in section 7 required detailed FE simulations to be carried out that are costly in terms of computation and an engineer’s time. What is shown by the differences in $\alpha$ and $m$ is that were an engineer to use an elastic-plastic J-Q failure curve for linear elastic analysis, the appropriate conversions would have to be made.

It can be seen that using the same $\alpha$ and $m$ values offer good agreement between the as received and pre-strained materials, and with the FAD, through altering the material toughness in the $K_r$ calculations. There is however greater discrepancy than with the J-Q calculations and their prediction of the mean toughness values. As can be seen in Figure 8-3 the FAD is non-conservative for all of the as received cases without residual stress. The most non-conservative case is the high constraint $a/W=0.4$ scenario, where one test failed at significantly lower load than the others. It was observed that reducing the toughness by $\approx 10\text{MPaVm}$ would offer better agreement with the FAD in all of the cases without residual stress. However, as the $K_{\text{mat}}$ value used had been validated against a previous test programme (2) and the master curve methodology had been applied correctly, it was appropriate to present the results using the original $K_{\text{mat}}$ in the $K_r$ calculations.

Furthermore, in the case of the as received material, using a reduced toughness would render the FAD excessively conservative when considering scenarios with residual stress. There was a significant $K_S^J$ introduced through the side punching process, according to the FE simulations where $J$ was calculated through equation 18, resulting in a significant $K_{\text{res}}$ value. This, combined with the scatter in the results, leads to the FAD appearing conservative in terms of its prediction of failure in this case. Furthermore, both $a/W=0.05$ scenarios from the as received material only had 8 valid tests completed, 2 less than the other scenarios. It is predicted with more tests there would better agreement with the lower bound datum from these scenarios and the FAD curve.
The scatter in cleavage fracture results is well documented and clearly observable in the test results presented in this thesis. This is corroborated through the better predictions when applying the mean toughness value, as with the J-Q methodology, which demonstrates that; because of the significant scatter, prediction of average, as compared to bounding, toughness values is more accurate with the number of tests carried out. It was shown, however, in section 7, that using the mean failure curve to extrapolate the lower bound toughness values was conservative using the J-Q methodology, so it would be recommended to use the mean values in such analysis as described in section 7.

The hypothesis that increasing the number of tests would offer better agreement between the FAD Option 1 curve and the failure points, could be further validated through observation of the pre-strained FAD, which displays 10 completed tests for each scenario, and clearly offers better agreement with the lower bound data. It can also be seen that the spread in the high constraint (a/W=0.4) test scenarios is less skewed, and there are no noticeable outliers. On the whole, the pre-strained scenarios failed at much lower loads. As a result of this there is less scatter in the tests: it is known with that cleavage fracture the larger the values of $K_{tc}$, the greater the scatter, as is discernible through observation of the results with decreasing constraint. This, therefore, implies that the change in the material which effected a reduced toughness in the pre-strained material, also reduced the scatter in the results. This enabled better high-constraint toughness prediction using the master curve approach and therefore better agreement with the FAD.

A further observation is that in Figure 8-4 the use of the Option 2 curve also improves the failure prediction of the FAD, most noticeably in the very low constraint (a/W=0.05) scenario. It can be seen that the Option 1 FAD would be excessively conservative (in that the failure points would be further away from the failure curve). It is also evident that using either the 0.2% proof stress or elastic limit in the $f(L)$ calculations results in different failure curves, especially at higher values of $L_r$, with, as has been the case throughout this thesis, better agreement using the elastic limit. The $L_r$ and $K^p_1$ calculations also incorporated the elastic limit as $\sigma_y$, so better agreement would be expected. Obviously using the 0.2% proof stress drastically reduces $L_r$, and from observation of Figure 8-4, this would mean that the use of the Option 2 failure curve would not provide any significant change in failure prediction using this methodology.
This observation once again necessitates consideration beyond standard plasticity theory when assessing material which has seen significant plastic strains. It has been evident throughout this thesis that the onset of plasticity, at much lower stresses than the 0.2% proof stress, has a significant effect on the failure behaviour of the material, and that if the material is considered to simply have an increase in yield stress, with a parallel Young’s modulus, the failure assessment may be non-conservative and therefore potentially dangerous, demonstrating that this low-strain analysis of tensile behaviour must be carried out and considered in structural integrity assessments of pre-strained material.
9. Summary, Conclusions, and Future Work

A brief overview of the work presented in this thesis is as follows, with further recommendations included:

- A significant experimental programme was carried out to obtain failure data of SENB specimens fabricated from a ferritic RPV steel, with defects which had been subjected to a tensile residual stress field and/or significant uniaxial plastic strain.
- The tests were carried out on high, medium and very low constraint geometries.
- Detailed 3D finite element simulations of the tests were carried out to calculate the crack tip stress fields.
- It was not possible to model the load vs J behaviour of pre-strained material using standard plasticity theory.
- The use of standard plasticity theory to model load vs J behaviour was non-conservative.
- A tensile test programme was also carried out to ascertain the effects of pre-straining on the low strain uniaxial tensile behaviour of the material, for use as input in the simulations.
- Significant uniaxial pre-straining caused a reduced modulus and lower elastic limit in the ferritic material.
- The high constraint $K_{mat}$ fracture toughness of A533b material with 5% uniaxial pre-strain was 75% that of the as received material.
- Mechanistic modelling considering standard plasticity theory did not predict the reduction in toughness observed through the introduction of significant plastic strains, local approaches offer improved but still inaccurate predictions.
- It would be pertinent to ascertain if mechanistic modelling or local approach calculations could be updated to include the reduced modulus and lower elastic limit apparent in the pre-strained material, alongside the postulated microcrack population, to ascertain if the reduction in toughness observed in the tests presented in this thesis could be predicted.
• The reduction in toughness of the pre-strained material is most likely caused through breaking of cleavage initiators in the material during the pre-straining process.
• The reduction in modulus and elastic limit in the material which had seen 5% plastic strain would corroborate the increase in microcracks as they would act as new dislocation sources.
• To validate the hypothesis that broken microcracks effect the reduction in modulus, elastic limit and fracture toughness in the pre-strained material it would be desirable to carry out detailed fractography of the pre-strained specimens.

• A boundary layer model was created to calculate the SSY field for the as received and pre-strained materials.
• It was necessary that the SSY field for use in J-Q calculations for the pre-strained material was calculated using the same material properties as the finite geometry stress field.

• An Abaqus-Python script was developed to automate the calculation of J-Q loading curves.
• J-Q loading curves were calculated for all of the test scenarios, and populated with failure data from the tests.
• A failure curve was derived from the as received results and was able to predict constraint based toughness that predicted toughness benefit through constraint loss in pre-strained material and defects in a residual stress field.
• To predict constraint based toughness benefit using the as received failure curve it was necessary to consider the onset of plasticity, as opposed to yielding beyond the 0.2% proof stress.
• To use the J-Q failure curve, it was also necessary to know the high constraint toughness under the initial condition in question.
• The failure curves derived in this thesis were obtained through fits to test data. To achieve prediction for the pre-strained material it was necessary to have both the as received failure curve and the high constraint toughness of the pre-strained material. The next stage in any work on this subject would be to
ascertain if the $\alpha$ and $m$ parameters could be derived for the as received and pre-strained material without testing. This would most likely be achieved through local approach calculations.

- A single element analysis was carried out to assess the effects of geometry and a residual stress field on crack-tip stress triaxiality.
- At high constraint stress triaxiality does not necessarily increase with the addition of a tensile residual stress field.
- The stress triaxiality fields at low constraint are significantly lower at mean failure load than in higher constraint scenarios.

- The industrial FAD methodology was also used to assess the test results and showed good lower bound toughness prediction.
- Using the R6 (26) FAD methodology also offers good failure prediction provided the material’s elastic limit is used in the calculations.

- It would be appropriate to further compare the methodologies described above with toughness data from materials at different levels of pre-strain, and material from weld HAZ’s that had been subjected to significant plastic strains and weld induced residual stresses, simulating realistic plant conditions.
10. References


8. James, P, Hutchinson, P and Madew, C. *Provisional Results for an Experimental Investigation into the Effect of Combined Primary and Secondary Stresses When Considering the Approaches of R6 and the recently Developed g() Factor.* 2011. ASME Conference on Pressure Vessels and Piping.


27. Jivkov, A P. Introduction to Structural Integrity, Lecture 5: The FAD. 2012. [Lecture Notes].


61. **AMEC, Clean Energy Ltd.** Tensile Test Data Results, 58KV-140/65KVRT.


64. **Zwick/Roell**. testXpertII. [Software].


Appendix: JAQ Code

There follows an annotated Abaqus-Python script as described in section 6.2, with annotations shown in # red

#opens odb

from abaqus import *
from abaqusConstants import *
session.Viewport(name='Viewport: 1', origin=(0.0, 0.0), width=166.076553344727, height=277.283325195313)
session.viewports['Viewport: 1'].makeCurrent()
session.viewports['Viewport: 1'].maximize()
from viewerModules import *
from driverUtils import executeOnCaeStartup
executeOnCaeStartup()
import os
os.chdir("C:\Work\Current ODBs\Final\GM")
odb = session.openOdb(name='GM04.odb')
session.viewports['Viewport: 1'].setValues(displayedObject=odb)

#draws path

session.Path(name='Path-1', type=NODE_LIST, expression=[('SENB-1', (87, '3213:3185:-1', 82, '2894:2890:-1', 81, '2993:3000:1', ))])

# defines number of nodes in path

pth = session.paths['Path-1']
pathlength=[]
pathlengthsum=[]
for i in range(len(pth.expression[0][1])):
    if type(pth.expression[0][1][i])==str:
        a=pth.expression[0][1][i].split(";")
        pathlength.append(a)
    else:
        a=[pth.expression[0][1][i]]
        pathlength.append(a)
for i in range(len(pathlength)):
    if len(pathlength[i])==1:
        x=1
        pathlengthsum.append(x)
    else:
        y=(int(pathlength[i][1])-int(pathlength[i][0]))/int(pathlength[i][2])
        pathlengthsum.append(y)
pathlength=int(sum(pathlengthsum))

# creates lists

stresses=[]
Js=[]
normdist=[]
normstress=[]
plotpoints=[]
data=[]
rawstress=[]
sig0=732
incs=len(odb.steps['Load'].frames)

# extracts direct stress (S22) at all contours greater than 2 um radius

session.viewports['Viewport: 1'].odbDisplay.setPrimaryVariable(variableLabel='S', outputPosition=INTEGRATION_POINT, refinement=(COMPONENT, 'S11'))
for i in range(incs):
    session.viewports['Viewport: 1'].odbDisplay.setFrame(step=0, frame=i)
    session.XYDataFromPath(name='S11_Frame_'+str(i), path=pth, includeIntersections=False, pathStyle=PATH_POINTS, numIntervals=10, shape=UNDEFORMED, labelType=TRUE_DISTANCE_Y)
    xystr = session.xyDataObjects['S11_Frame_'+str(i)]
    stresses.append(xystr)
for j in range(len(stresses[i])-pathlength):
    while stresses[i][j][0]<2e-3:
        stresses[i].pop(j)

# extracts abaqus J Integral values

for i in range(pathlength):
    session.XYDataFromHistory(name='J_Contour_'+str(i+1), odb=odb, outputVariableName='J-integral: J at J_CRACK-1__PICKEDSET160-10__Contour_33 in ELSET ALL ELEMENTS', steps=('Load',), )
    xyJ = session.xyDataObjects['J_Contour_'+str(i+1)]
    Js.append(xyJ)

# calculate normalised stress vs normalised distance

normstress=[[stresses[i+1][j][1]/sig0 for i in range(incs-1)] for j in range(pathlength-1)]
normdist=[[stresses[i+1][j][0]*sig0)/(Js[j+1][i][1]) for i in range(incs-1)] for j in range(pathlength-1)]
plotdata=[[(normdist[i][j],normstress[i][j]) for i in range (pathlength-1)] for j in range (incs-1)]

# plot normalised stress vs normalised distance for all increments

xQuantity = visualization.QuantityType(type=PATH_Y)
yQuantity = visualization.QuantityType(type=STRESS)
curves=[]
cs=[]
for i in range(incs-1):
    session.XYData(name='Nstr_Ndist_'+str(i+1), data=plotdata[i]), sourceDescription='Entered from keyboard', axis1QuantityType=xQuantity, axis2QuantityType=yQuantity,
    xy=session.xyDataObjects['Nstr_Ndist_'+str(i+1)]
    c=session.Curve(xyData=xy)
    curves.append(c)
import bisect

indices = []
Qs = []
newnormdist = [[plotdata[i][j][0] for j in range(len(plotdata[i]))] for i in range(len(plotdata))]
newnormstress = [[plotdata[i][j][1] for j in range(len(plotdata[i]))] for i in range(len(plotdata))]

for i in range(len(newnormdist)):
    indices.append(bisect.bisect(newnormdist[i], 2))
    for i in range(len(indices)):
        if indices[i] > 0 and indices[i] <= pathlength - 2:
            x0 = newnormdist[i][indices[i]-1]
            x1 = newnormdist[i][indices[i]]
            y0 = newnormstress[i][indices[i]-1]
            y1 = newnormstress[i][indices[i]]
            y = ((y1 - y0) / (x1 - x0) * (2 - x0)) + y0
            Q = y - 3.005
            Qs.append(Q)
        else:
            Q = 0
            Qs.append(Q)

# plot J-Q Curve

xQuantity = visualization.QuantityType(type=PATH_X)
yQuantity = visualization.QuantityType(type=STRESS)
JQplotdata = [(Qs[i], Js[indices[i]][i][1]) for i in range(len(Qs))]
session.XYData(name='J_V_Q', data=(JQplotdata), sourceDescription='Entered from keyboard', axis1QuantityType=xQuantity, axis2QuantityType=yQuantity,)

chartName = xyp.charts.keys()[0]
xy = session.xxyDataObjects['J_V_Q']
c = session.Curve(xyData=xy)
xyp = session.XYPlot('J_V_Q')
chartName = xyp.charts.keys()[0]
chart = xyp.charts['Chart-2']
chart.setValues(curvesToPlot=(c), )
session.viewports['Viewport: 1'].setValues(displayedObject=xyp)
session.charts['Chart-2'].axes1[0].axisData.setValues(maxValue=-1, maxAutoCompute=False)
session.charts['Chart-2'].axes1[0].axisData.setValues(minValue=0.2, minAutoCompute=False)
session.charts['Chart-2'].fitCurves(fitAxes1=True, fitAxes2=False)
session.charts['Chart-2'].axes2[0].axisData.setValues(maxValue=0, maxAutoCompute=False)
session.charts['Chart-2'].axes2[0].axisData.setValues(minValue=100, minAutoCompute=False)
session.charts['Chart-2'].fitCurves(fitAxes1=False, fitAxes2=True)