A study of the generation and creep relaxation of triaxial residual stresses in stainless steel

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Abstract

This paper presents results from a numerical and experimental research programme motivated by the need to predict creep damage generated by multi-axial states of stress in austenitic stainless steels. It has been hypothesized that highly triaxial residual stress fields may be sufficient to promote creep damage in thermally aged components, even in the absence of in-service loads. Two prerequisites to test this hypothesis are the provision of test specimens containing a highly triaxial residual stress field and an accurate knowledge of how this residual stress field relaxes due to creep. Creep damage predictions may then be made for these specimens and compared to damage observed in experiments. This paper provides solutions to both of these prerequisites. Cylindrical and spherical test specimens made from type 316H stainless steel are heated to 850 °C and then quenched in water. Finite element predictions of the residual stress state, validated by extensive neutron diffraction measurements, are presented which confirm the high level of triaxiality present in the specimens. The specimens are then thermally aged at 550 °C and numerical predictions of the residual stress relaxation are given, again validated by extensive neutron diffraction measurements. The results confirm the validity of the creep relaxation models employed. In addition, the results show the influence of specimen size and permit comparisons to be made between three different types of neutron diffractometers.
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1. Introduction

The principal deterioration mechanisms in high temperature plant are creep damage, microstructural degradation, high temperature fatigue, creep-fatigue, embrittlement, carburisation, thermal shock, erosion, and high temperature corrosion of various types (Furtado and Le May, 2002). Creep is one of the most serious high temperature damage mechanisms and creep damage parameters are usually calculated using either a time
fraction approach or a ductility exhaustion approach. The presence of multi-axial stress states, often introduced in plant by welding, further complicates calculations of creep damage. In the case of the time fraction approach, there are a number of models that predict the effect of state of stress on creep rupture strength, e.g. Huddleston (1985) developed a model based on data determined from stainless steels. The R5 assessment procedure (R5, 2001), used widely in the United Kingdom power industry and beyond, uses a ductility exhaustion approach to calculate creep damage and includes a model for use under triaxial states of stress. Spindler (2004) developed this model by considering cavity nucleation and growth, and multi-axial creep data obtained from type 304 and 316 stainless steels. The ductility exhaustion approach relies on accurate predictions of the creep strains which arise during thermal relaxation of pre-existing residual stress fields and the local multi-axial creep ductility. The model assumes that both the hydrostatic stress and the deviatoric stresses contribute to damage accumulation.

The purpose of the present study was to provide a platform for future validation of these creep damage models. To provide such a platform, test specimens are needed which contain a highly triaxial residual stress field. The test specimens must also be amenable to experimental measurements of residual stress in order that finite element predictions of the residual stress state may be validated. The test specimens can then be aged at elevated temperatures in order to examine how the deviatoric and hydrostatic stresses relax. Predictions of the relaxed stress state may then be validated against further measurements. Once confidence has been gained regarding the magnitude and distribution of the original residual stress field and its subsequent relaxation by thermal ageing, the specimens may then be used for further creep damage studies.

Recent work (Hossain et al., 2006, 2004) has confirmed the potential of water quenching spherical and cylindrical specimens as a mechanism for creating test specimens which contain a residual stress field of sufficiently high magnitude and triaxiality to generate creep damage if the specimens are raised to an elevated temperature. It was shown that differing levels of triaxiality could be achieved over a relatively large volume of the test specimen by adjusting the cooling conditions and changing the specimen dimensions. The geometry of the specimens allows the residual stress field to be measured by neutron diffraction, as well as reducing the complexity of the supporting finite element study. This earlier research, however, did not consider the thermal ageing process and its influence on residual stress state, either experimentally or numerically and, furthermore, only presented results from a limited number of measurements on as-quenched specimens. This paper, in contrast, presents a comprehensive set of measurement results obtained on a significantly higher number of as-quenched specimens (and conducted at three independent measurement facilities) and then proceeds to present numerical predictions and experimental measurements of the relaxed residual stress field existing in the specimens when they have been subjected to short-term and long term thermal ageing at 550 °C.

In the course of the paper, experimental results are presented from three independent neutron sources, located across Europe. Thus, in addition to providing validation of the numerical simulations, a relatively unique opportunity exists for comparing residual stress measurements made using three different neutron diffractometers, including both continuous and pulsed neutron sources. The paper starts by describing the material used and the specimens created and then proceeds to detail the numerical simulations of the quenching and thermal ageing processes. A description of the measurements performed is then given, followed by the results obtained. In the final sections of the paper, results are compared and conclusions drawn over whether the original objectives have been met.

2. Material and test specimens

A number of specimen designs were explored with the intention that well-defined residual stress fields could be introduced into the specimen of sufficient magnitude and triaxiality to enable the study of creep relaxation at elevated temperature. For this reason, spherical and cylindrical test specimens were machined from type 316H and 316L stainless steel of dimensions shown in Table 1. The choice of material was motivated by a practical industrial application. Specifically, reheat cracking has been observed in austenitic stainless steels and nickel based alloys. The cracking is associated with creep straining in weldments either during post-weld-heat-treatment or during high temperature service (Bouchard et al., 2004). A total of 261 incidents of reheat cracking in the steam pipe-work systems of UK Advanced Gas Cooled Reactors have been reported by Coleman et al. (1998). The cracks were observed to develop in strain hardened zones adjacent to non-stress
relieved welds in AISI Type 316H stainless steel components after 10,000–50,000 h operation in the temperature range 490–530 °C.

The dimensions of the specimens, and in particular the length-to-diameter ratio of the cylinders, were chosen so that the triaxiality factor, defined as $\sigma_h/\sigma_e$, where $\sigma_h$ is the hydrostatic stress and $\sigma_e$ is the von Mises equivalent stress, attained a maximum value. Full details of the original design study may be found in Hossain et al. (2006). The choice of both spherical and cylindrical specimens was motivated by consideration of the creep damage model which would subsequently be used, detailed in Section 3. Spherical specimens would, in principle, contain an equi-triaxial stress field at the centre of the specimen, and a correspondingly infinite triaxiality factor. The creep damage model, however, contained the triaxiality factor as a parameter and could potentially become numerically unstable. It was thus considered desirable to also consider cylindrical specimens which did not suffer from this potential problem.

It can be seen from Table 1 that 12 specimens were fabricated in total, eight of spherical geometry and four of cylindrical geometry. However, only seven specimens (S1, S3, S5, S6, S7, S9 and S11) were available for residual stress measurements, the remaining specimens (S2, S4, S8, S10 and S12) being used to obtain stress-free lattice spacings. All but two of the test specimens were made from type 316H stainless steel. Two spheres were fabricated from type 316L stainless steel in order to permit the sensitivity of predictions and measurements to small changes in material properties to be determined. The 316H material was supplied as a single block of well-characterised ex-service steam header material, and had the mechanical properties shown in Fig. 1 (R66). The 316L material was less well characterised and estimates of material properties from the literature were used for the numerical predictions involving specimen S1.

A residual stress field was generated in each of the specimens by heating them to 850 °C and then rapid spray water quenching them using an experimental rig consisting of a cylindrical furnace, a cooling chamber, and a support and quick release mechanism. The furnace was isolated from the cooling chamber using a removable heat resistant plate. The cylindrical furnace consisted of three zones, each with independent temperature controllers. The cooling chamber consisted of two concentric hollow cylinders, sealed at each end, so that water could be pumped between the two cylinders. The inner cylinder wall contained a series of holes so that uniform water spraying of a sample when placed in the chamber could be achieved. Water from the chamber flowed into a bucket below the experimental rig. Rapid water spray quenching was deemed more appropriate than simply dropping the specimens into water so as to minimise the effects of film boiling and avoid the associated complexity of the finite element analysis. The procedure is fully documented in Hossain et al.
(2004). Use of type 316 austenitic stainless steel for the test specimens ensured no phase transformations were present during the quenching process, greatly assisting the numerical modelling by ensuring that no residual stresses were introduced by volumetric changes caused by phase transformations. It is emphasised that all specimens were subjected to the same initial quenching procedure.

After the quenching procedure, two cylindrical specimens (S9 and S11) and four spherical specimens (S7, S8, S10 and S12) were placed in a furnace at 550 °C in order to promote creep relaxation of the original, quench-induced, residual stress field. The six specimens were then subjected to three different soak times. Two samples (S11 and S12) were subjected to a short term thermal age of 1.25 h, two samples (S9 and S10) were soaked to a medium-term age of 1800 h, and two specimens (S7 and S8) were long-term aged to 3200 h. Use of these different soak time periods enabled the numerical models detailed in Section 3 to be validated at more than one temporal point and permitted the time evolution and relaxation of the residual stress state to be monitored.

As has already been indicated, only seven of the specimens in Table 1 were available for residual stress measurements. This was because the experimental technique used in this paper to measure internal residual stresses, neutron diffraction (details of which are provided in Section 4), required stress-free lattice spacings in order to determine the internal strains. The stress-free lattice spacings were obtained by making measurements on samples extracted from specimens S2, S4, S8, S10 and S12. Using samples which had undergone the same thermal and mechanical load history as those measured ensured that the stress-free measurements were made on material with the same local microstructure. In order to ensure that measurements were made in a stress-free state, ‘comb’ samples (fully described in Hossain et al. (2004) and illustrated in Fig. 2) were extracted from the specimens following fabrication and thermal ageing (if relevant). The ‘comb’ samples were manufactured using electro-discharge machining (EDM) and consisted of seven fingers, positioned across the diameter of the spherical specimens.

3. Finite element models

3.1. Quenching model

Only the salient features of the numerical quenching model are presented here. A more detailed description of the model may be found in Hossain et al. (2006, 2004). The finite element analysis was carried out using the ABAQUS 6.3 FE code (Hibbit, 1080) and consisted of an uncoupled heat transfer analysis with a subsequent
thermal non-linear stress analysis using a kinematic hardening model. Due to symmetry, a quarter axi-symmetric model was used, as shown in Fig. 3. The boundary conditions included convective heat transfer on the outer surfaces with a heat transfer coefficient of 7000 W/(m² K) (Sen et al., 2000), and an adiabatic condition on both the axial and radial axes. The heat transfer coefficient was chosen to ensure that predictions of both surface and sub-surface specimen temperatures during the quench process agreed with the values recorded by thermocouples. It was assumed that the surface heat transfer coefficient was independent of surface temperature (Gür et al., 1996). Both materials, 316L and 316H, were assumed elastic with strain hardening plasticity and with a yield stress that decreased with temperature. The mechanical and physical properties for type 316H stainless steel were obtained from (R66) and are shown in Fig. 1. The mechanical and physical properties of type 316L stainless steel were of a similar form to Fig. 1, but with lower yield stresses at all temperatures, e.g. 187 MPa 0.2% offset yield stress compared with 286 MPa for the 316H. Full details of 316L material properties were given in Hossain et al. (2004). The specimen was initially assumed to be at a uniform temperature of 850 °C, where it was assumed in a stress-free state. The specimen was then quenched in water.
at room temperature until the entire cylinder or sphere reached the equilibrium quenchant temperature of 20°C. During the heat transfer analysis the temperature distributions were stored in the ABAQUS results file. This temperature–time history was then used as an input loading condition in the thermal stress analysis step. The transient stresses were large enough to cause significant plastic flow, so residual stresses remained after the specimens reached the coolant temperature. The effect of phase transformation on residual stress and distortion was considered unimportant for the stainless steel. The results of the finite element analyses for the quenching process, specifically results for samples S1, S3, S5 and S6 (as defined in Table 1), are presented and discussed in Section 5, where they are compared to experimental measurements.

3.2. Creep model

In order to predict the stress relaxation occurring in the specimens caused by creep at elevated temperatures, a number of further steps were added to the finite element model described in Section 3.1. These further steps consisted of firstly raising the temperature of the quenched specimens to the operating temperature of the furnace, 550°C, and secondly simulating thermal exposure at 550°C for the appropriate number of hours for each of the relevant specimens (samples S7, S9 and S11 in Table 1). The temperature was raised within the finite element analysis by instantaneously raising the temperature field within the whole model. The temperature-dependent material properties shown in Fig. 1 were used in the analysis. Creep deformation was modelled using an empirical law for primary and secondary creep of AISI Type 316H steel (Holt, 1996), implemented in the high temperature assessment procedure (R5, 2001) and based on the empirical model found in RCC-MR (1985), i.e.

\[
\bar{e}_p = \frac{C_1}{100} \bar{\sigma}^{n_1} t^{n_2},
\]

\[
\bar{e}_s = \bar{e}_{ip} + C \bar{\sigma}^{n_3} (t - t_{ip}),
\]

where \(\bar{e}_p\) and \(\bar{e}_s\) are the equivalent creep strains in the primary and secondary ranges respectively, \(t\) is the time in hours, \(\bar{\sigma}\) is the von Mises equivalent stress (MPa), \(t_{ip}\) is the time in hours of transition from primary to secondary behaviour (where the primary and secondary strain rates are equal) and \(\bar{e}_{ip}\) is the equivalent creep strain at that time, i.e. at the end of the primary creep stage. The constants \(C_1, C_2, C, n_1\) and \(n\) are temperature-dependent coefficients and are \(2.9618 \times 10^{-12}, 0.042131, 5.2900 \times 10^{-26}, 4.18\) and \(8.20\) respectively at 550°C, \(R60\), and are specific to the material considered in this study.

Although this paper restricts the use of the creep model to making predictions of the relaxation of the residual stress field due to thermal exposure, it is informative to note that the model was used in conjunction with a separate creep damage model, based on the ductility exhaustion approach described in R5 (2001), which permitted a creep damage parameter to be determined at any point in the specimen at the end of every time increment of the finite element analysis. The creep damage parameter, \(D_c\), was defined as

\[
D_c = \int_0^t \frac{\dot{\bar{e}}_c}{\bar{e}_c} \text{d}t
\]

where \(\dot{\bar{e}}_c\) is the instantaneous von Mises creep strain rate at time \(t\), and \(\bar{e}_c\) is the corresponding multi-axial creep ductility, expressed as the von Mises strain at failure, which is a function of the strain rate and stress state. Within the creep damage model, an empirical approach was adopted for describing the effects of stress state on ductility, (Spindler, 2004). This included two material constants, \(p\) and \(q\), which described the decreasing ductility with increasing stress triaxiality according to

\[
\frac{\bar{e}_f}{\bar{e}_{uni}} = \exp \left[ p \left(1 - \frac{\sigma_1}{\sigma_e}\right) \right] \exp \left[ q \left(\frac{1}{2} - \frac{3\sigma_h}{2\sigma_e}\right) \right],
\]

where \(\bar{e}_{uni}\) is the uniaxial ductility, \(\sigma_e\) is the von Mises stress, \(\sigma_1\) is the maximum principal stress and \(\sigma_h\) is the hydrostatic stress. The first term of Eq. (4) represents cavity nucleation. The second term represents cavity growth by creep deformation. Crack initiation is expected when \(D_c \geq 1\). The constants \(p\) and \(q\) are empirically derived from multi-axial tests on a particular material and two sets of values have been proposed by Spindler...
for AISI Type 316 stainless steel (Spindler, 2004). Thus, although predictions of creep relaxation are made using a model based on the von Mises equivalent stress, Eqs. (1) and (2), it can be seen that specimens containing a high hydrostatic component of residual stress are required for studies of creep damage, as predicted using Eqs. (3) and (4). Furthermore, the form of Eq. (4) influenced the decision to fabricate cylindrical test specimens in addition to spherical test specimens, as discussed in Section 2. In spherical test specimens, the equivalent stress was predicted to be zero at the centre of the specimen and this, in turn, could potentially lead to numerical problems implementing Eq. (4).

In the finite element analysis of the creep relaxation, the creep strain for both the primary and secondary stages was calculated for each time increment. An initial time increment of 0.01 h was used in the analysis with subsequent time increments varying in magnitude. The variable time increment used was chosen by allowing a maximum tolerance (difference between the creep strain increment calculated using the conditions at the beginning and end of the increment) of $5 \times 10^{-5}$ (Holt, 1996). From the creep strain increments calculated independently for both the primary and secondary creep, the larger increment was subsequently selected. This meant that the changeover from primary to secondary creep occurred when the primary and secondary rates became equal, the primary law applying when the predicted primary creep rate was higher than the secondary creep rate and vice-versa. The strains, displacements and stresses at the end of the analysis increment were then calculated from the constitutive relations.

4. Residual stress measurements

The internal residual stress fields were measured in the samples listed in Table 1 using the neutron diffraction technique. The table also shows the corresponding ‘comb’ samples which were used to determine the appropriate stress-free lattice parameter, $d_0$. For brevity, only the pertinent features of this technique are presented here and further details may be found in Fitzpatrick and Lodini (2003). It has long been known that neutrons have a wave-like character, and an associated wavelength, $\lambda$, which for thermal neutrons is of the same order of magnitude as inter-atomic spacings in solids. As such, whenever a beam of neutrons is incident on a solid, diffraction phenomena occur which in crystalline solids may be used to infer information concerning the spacing between layers of atoms, and consequently the strain. It may easily be shown that if $2\theta$ is the angle between the incident beam and the diffracted beam then with a polycrystalline sample constructive interference (and a subsequent peak in intensity) occurs when Bragg’s law is satisfied

$$2d_{hkl} \sin \theta = \lambda. \quad (5)$$

In this equation, $d_{hkl}$ is the interplanar distance between planes of Miller indices $(hkl)$.

In order to make strain measurements in crystalline materials a supply, or flux, of neutrons is thus required. This flux may be provided in two ways. Firstly, neutrons which are generated as part of a controlled nuclear fission reaction – i.e. in a nuclear reactor – may be collimated and controlled to produce a steady supply of thermal neutrons. Sources which produce neutrons in this manner are termed steady-state (or continuous), with the exception of the source at Dubna which is a pulsed reactor. An alternative method of neutron production is to accelerate protons and then fire these high energy protons onto a target of high atomic mass. This causes the target nuclei to be elevated to a highly excited state, from which they immediately decay by emitting neutrons and other particles. The neutrons may then be collimated and focussed onto the sample and used to conduct strain scans. Sources which produce neutrons in this manner are termed spallation sources. The question often arises as to the usefulness of always classifying reactors as continuous sources and spallation sources as pulsed sources. Although this is the most common way of classifying source types, it is not absolutely necessary e.g. the sources at Dubna and SINQ do not adhere to this classification.

In both types of neutron source, an incident neutron beam is directed at the specimen, and the diffracted neutrons are counted with detectors. For steady state sources it is most common to use a monochromatic incident beam, i.e. set $\lambda = \text{constant}$, and measure the scattered intensity as a function of scattering angle, $\theta$. For pulsed sources it is usual to use a polychromatic incident beam and a fixed scattering angle, i.e. set $\theta = \text{constant}$, and measure the scattered intensity as a function of wavelength, $\lambda$. In situ this is achieved by measuring the time-of-flight (TOF), $t$, of neutrons between the moderator or the chopper and the detector, which is directly related to the wavelength through the de Broglie relationship,
\[ t = \frac{L}{v} = \frac{\lambda mL}{h}, \]  
(6)

where \( h \) is Planck’s constant, \( m \) is the mass of a neutron, \( v \) is neutron velocity and \( L \) is the distance between the moderator and the detector.

To measure absolute values of residual strain, a stress-free lattice spacing, \( d_{0}^{hkl} \), must also be measured. This permits the strain component in a direction defined by the geometry of the incident and diffracted beam, \( e_{i} \), to be determined as

\[
e_{i} = \frac{d_{i}^{hkl} - d_{0}^{hkl}}{d_{0}^{hkl}} = \frac{\Delta \lambda}{\lambda} - \cot \theta \Delta \theta,
\]

(7)

where use has been made of Eq. (3). For constant wavelength strain scanners, \( \Delta \lambda = 0 \) and \( e_{i} = -\cot \theta \Delta \theta \), and for pulsed beam instruments, \( \Delta \theta = 0 \) and \( e_{i} = \Delta \lambda \theta = \Delta t/t \). Once the residual strain components have been measured, the residual stresses may be determined by application of Hooke’s law.

Three different types of neutron diffractometers (and four instruments) were used to determine the residual stresses in the samples detailed in this paper. The nature and location of the diffractometers used are given in Table 2. Samples S3, S6 and S7 were measured on the ENGIN-X instrument at the ISIS facility located at the Rutherford Appleton Laboratory, UK. Sample S1 was measured on the ENGIN instrument, the precursor to ENGIN-X. ENGIN-X is a purpose-built neutron scattering instrument for engineering residual stress measurements. Its main features include a sample-positioning device, focussing collimator slits, two position sensitive detectors at 90° from the incident neutron beam and a masking slit for the incident beam. The incident and diffracted beams define the sampling or gauge volume within the specimen. The diffraction data was analysed using the Rietveld refinement technique, which has been shown to provide elastic strains analogous to the bulk equivalent strains (Daymond et al., 1999). A 2 × 2 × 2 mm³ gauge volume was used for the 30 mm diameter samples and a 4 × 4 × 4 mm³ gauge volume was used for the single 60 mm diameter sample (S6), with strain scans made in all three principal orientations – radial, axial and hoop. The stress-free \( d_{0}^{hkl} \) measurements were made on the ‘comb’ samples detailed in Section 2 and (Hossain et al., 2004). Specimen S2 was used to provide stress-free \( d_{0}^{hkl} \) values for specimen S1, S4 was used to provide stress-free \( d_{0}^{hkl} \) values for S3, S4 was also used to provide stress-free \( d_{0}^{hkl} \) values for S6, and S8 was used to provide stress-free \( d_{0}^{hkl} \) values for S7. In each case, an average value of \( d_{0}^{hkl} \), irrespective of orientation, was used to calculate the residual stresses in the specimens.

Samples S9 and S11 were measured on the POLDI instrument (Pulse Over Lap time-of-flight Diffractometer) (Stuhr et al., 2005) at the SINQ facility located at the Paul Scherrer Institut, Switzerland. SINQ is a quasi-continuous spallation source but POLDI allows the angular information present in the diffracted neutron beam, usually ignored by time-of-flight diffractometers, to be used additionally to the time-of-flight of the neutrons for the determination of residual strains (Stuhr, 2005). A 2 × 1.5 × 2 mm³ gauge volume was used in the measurements and strain scans were again made in the radial, axial and hoop directions. The analysis of the measured data was carried out using an in-house program which determined, via a direct comparison of lattice spacing between the stressed components and their corresponding stress-free combs for a selected reflection plane, the residual stress distributions in the specimens. The stress-free \( d_{0}^{hkl} \) measurements were made on

<table>
<thead>
<tr>
<th>Specimen/‘comb’</th>
<th>Instrument and measurement location</th>
<th>Type of neutron source</th>
<th>Measurement principle</th>
</tr>
</thead>
<tbody>
<tr>
<td>S1(S2)</td>
<td>ENGIN at ISIS, UK</td>
<td>Spallation</td>
<td>Time-of-flight</td>
</tr>
<tr>
<td>S3(S4)</td>
<td>ENGIN-X at ISIS, UK</td>
<td>Spallation</td>
<td>Time-of-flight</td>
</tr>
<tr>
<td>S5(S4)</td>
<td>REST at Studsvik, Sweden</td>
<td>Continuous</td>
<td>Monochromatic diffraction</td>
</tr>
<tr>
<td>S6(S4)</td>
<td>ENGIN-X at ISIS, UK</td>
<td>Spallation</td>
<td>Time-of-flight</td>
</tr>
<tr>
<td>S7(S8)</td>
<td>ENGIN-X at ISIS, UK</td>
<td>Spallation</td>
<td>Time-of-flight</td>
</tr>
<tr>
<td>S9(S10)</td>
<td>POLDI at SINQ, Switzerland</td>
<td>Quasi-continuous spallation</td>
<td>Time-of-flight</td>
</tr>
<tr>
<td>S11(S12)</td>
<td>POLDI at SINQ, Switzerland</td>
<td>Quasi-continuous spallation</td>
<td>Time-of-flight</td>
</tr>
</tbody>
</table>
‘comb’ samples extracted from S10 (for specimen S9) and S12 (for specimen S11). Again, an average value of \(d_{0}^{hkl}\) irrespective of orientation was used.

Finally, sample S5 was measured on the REST instrument (REsidual Stress and Texture measurement) at the Studsvik Neutron Research Laboratory (NFL), Sweden. The (311) reflection of 316H stainless steel, along with a \(2 \times 2 \times 2 \text{ mm}^3\) gauge volume, was used for the measurements and strain scans were again made in the radial, axial and hoop directions. The stress-free interplanar spacing \(d_{0}^{hkl}\) was again obtained by making measurements on the appropriate ‘comb’ sample, S4. Measurements on the REST instrument were made with a nominal neutron wavelength of 1.7 Å, and measurements on all instruments assumed the elastic constants \(E = 195.6 \text{ GPa}\) and \(\nu = 0.294\), where \(E\) is Young’s modulus and \(\nu\) is Poisson’s ratio.

It may be noted that the gauge volume employed in the measurements was related to the size of the specimens, e.g. a \(2 \times 2 \times 2 \text{ mm}^3\) gauge volume was used for the 30 mm diameter cylinder and a \(4 \times 4 \times 4 \text{ mm}^3\) gauge volume for the 60 mm diameter cylinder. The finite element analysis described in Section 3 and corresponding results given in Section 5 showed that the residual stress distributions did not vary abruptly along the measurement path and varied negligibly normal to the measurement path in the vicinity of the measurement points. Therefore, the strains averaged over the gauge volumes selected in the experiments were considered representative of the actual strains. The measurement uncertainties are shown in the results section by error bars in the figures.

5. Results and discussion

5.1. As-quenched predictions and measurements

Fig. 4 shows the finite element predicted and neutron diffraction measured residual stresses across the mid-section of a 30 mm diameter, as-quenched 316H stainless steel sphere (S3). The finite element simulations predicted higher tensile residual stresses in the centre of the sphere than measured using the ND method. Nevertheless, the measurements revealed the presence of an essentially triaxial residual stress field in a central core of about 10 mm diameter. The difference between the predictions and measurements was attributed to the material properties used in the finite element analysis. A kinematic hardening model was used when the use of a mixed hardening model would more realistically model the true behaviour. This was not possible,
however, due to a lack of materials data. Fig. 5 presents the same results, but now with the residual stresses normalised with respect to the yield stress, and distance normalised with respect to the radius of the sphere. This permits a direct comparison with equivalent results obtained for the 30 mm diameter, as-quenched 316L stainless steel sphere (S1) (Hossain et al., 2004). While there is general agreement between the two sets of measurements, it can be seen that higher triaxial residual stresses were obtained in the 316L sample. This was consistent with the finite element predictions given in Hossain et al. (2004).

Fig. 6 shows the predictions and measurements across the mid-section of a 30 mm diameter, 30 mm long, as-quenched type 316H stainless steel cylinder (S5). Similar comments may be made to those associated with Fig. 4. There is generally good agreement between predictions and measurements, with the predictions higher than measurements near the centre of the cylinder. An essentially triaxial residual stress field is present in a central core of about 10 mm diameter. The magnitude of the residual stress field is similar to that measured in the sphere S3. Fig. 7 shows the predictions and measurements across the mid-section of a 60 mm diameter, 60 mm long, as-quenched type 316H stainless steel cylinder (S6). Overall a good correlation is seen. It may be noted that the measurement of strain in the axial direction at the centre of the specimen required a total flight path length within the sample of ~85 mm. This resulted in excessive noise in the data with the limited duration of counting time available, resulting in unreliable axial strain data being obtained at the centre of the sample (thus measured residual stresses close to the centre of the cylinder are not shown in Fig. 7). Nevertheless, as the measurements agree closely with predictions near the surface of the cylinder, it is reasonable to assume the presence of an essentially highly triaxial residual stress field in a central region of the quenched cylinder, of diameter about 20 mm. Fig. 8 presents the same results using a normalised distance coordinate to permit direct comparison of different sized specimens. There is, in general, very good agreement between the two sets of measurements. The maximum difference is towards the surface of the cylinders where the radial stresses differ by about 70 MPa.

5.2. Thermally aged predictions and measurements

Fig. 9 shows the predictions and measurements across the mid-section of a 30 mm diameter, 30 mm long, type 316H stainless steel cylinder following quenching and thermal ageing at 550 °C for 1.25 h (S11). There is,
generally, good agreement between predictions and measurements. Fig. 10 shows the predictions and measurements for an identical cylinder, this time thermally aged for 1800 h (S9). Again, there is generally good agreement between predictions and measurements, except perhaps near the surface of the cylinder where the measured hoop stresses were more compressive than predictions. Fig. 11 shows the predictions and measurements for another identical cylinder, this time thermally aged for 1800 h (S9).
ments across the mid-section of a 30 mm diameter type 316H sphere following quenching and subsequent thermal ageing at 550 °C for 3200 h (S7). Again, there is excellent agreement, except near the surface of the sphere where the magnitudes of the measured stresses were higher than finite element predictions.

Fig. 8. A comparison between measured residual stress distributions in an as-quenched, 30 mm diameter, 30 mm long cylinder (S5) and an as-quenched, 60 mm diameter, 60 mm long cylinder (S6).

Fig. 9. Measured and predicted residual stress distribution across mid-section of a 30 mm diameter, 30 mm long, cylinder (S11) following quenching and subsequent short term thermal ageing at 550 °C for 1.25 h.
Fig. 12 shows the stress relaxation caused by thermal ageing in the long term thermally aged sphere (S7) by comparing measurements with the equivalent as-quenched specimen (S3). It may be clearly seen that the residual stresses towards the centre of the sphere relax by approximately 60%, whereas there is a relatively small reduction in residual stress towards the surface of the sphere.

Fig. 10. Measured and predicted residual stress distribution across mid-section of a 30 mm diameter, 30 mm long, cylinder (S9) following quenching and subsequent long term thermal ageing at 550 °C for 1800 h.

Fig. 11. Measured and predicted residual stress distribution across mid-section of a 30 mm diameter sphere (S7) following quenching and subsequent thermal ageing at 550 °C for 3200 h.

Fig. 12 shows the stress relaxation caused by thermal ageing in the long term thermally aged sphere (S7) by comparing measurements with the equivalent as-quenched specimen (S3). It may be clearly seen that the residual stresses towards the centre of the sphere relax by approximately 60%, whereas there is a relatively small reduction in residual stress towards the surface of the sphere. Fig. 13 shows a similar comparison for the ther-
mally aged cylinders (S5, S9 and S11), although only the hoop component of stress is shown for clarity. Although there are fewer measurement points, a similar trend may be observed. The residual stresses relax by a greater amount in the centre of the cylinders compared to near the surface, relaxing to about 75% of their original value.

Fig. 12. A comparison between measured residual stress distributions in an as-quenched, 30 mm diameter sphere (S3) and a quenched and aged 30 mm diameter sphere (S7).

Fig. 13. A comparison between the measured hoop residual stress distribution in an as-quenched, 30 mm diameter, 30 mm long cylinder (S5), a quenched and short term thermally aged 30 mm diameter, 30 mm long cylinder (S11) and a quenched and long term thermally aged 30 mm diameter, 30 mm long cylinder (S9).
as-quenched value in the short term thermally aged sample and to about 50% in the long term thermally aged cylinder. The predictions and measurements for specimens S5, S9 and S11 are plotted as von Mises equivalent stresses in Fig. 14. There exists an acceptable comparison between the predictions and measurements. The pre-

Fig. 14. Measurements and predictions of the von Mises equivalent residual stress distributions in 30 mm diameter, 30 mm long cylinders (S5, S11, S9) following different thermal ageing treatments.

Fig. 15. Numerical predictions of the time evolution of the residual stress state at the centre of a 30 mm diameter, 30 mm long cylinder following quenching and subsequent thermal ageing at 550 °C. Measurements at three temporal points (S5, S11, S9) are also shown.
dictions tend to overestimate the equivalent stresses, although both predictions and measurements highlight the low equivalent stresses (and hence high triaxiality) in the central region of the cylinders for all of the thermal ageing soak times. Fig. 15 shows numerical predictions and measurements of the time evolution of the stress state at the centre of cylinders S5, S9 and S11. The time evolution of both the von Mises equivalent stress and the hydrostatic component of residual stress are both shown. It may be noted that although the creep relaxation model employed, Eqs. (1) and (2), contains only the von Mises equivalent stress, the hydrostatic component of stress also decreases with time. This is due to the redistribution of residual stresses, not creep relaxation per se. The form of the two plots, and in particular the low and decreasing value of \( \sigma_e \) and the relatively high ratio \( \sigma_h/\sigma_e \), suggests that creep damage, as predicted by Eqs. (3) and (4), should continue to accumulate as the specimens are thermally aged. The measurements made at the three different time coordinates agree acceptably with predictions.

6. Closing remarks and conclusions

Numerical studies using finite element analysis have demonstrated that highly tensile, triaxial, residual stresses can be obtained in the centre of quenched cylinders and spheres. These tensile residual stresses are balanced by compressive residual stresses near and at the surface. Neutron diffraction measurements have confirmed these predictions. A creep relaxation model was then used to predict how these residual stresses relax when the specimens were subjected to thermal soaks of different duration. Neutron diffraction measurements were again performed and validated the predictions. The residual stresses were shown to remain highly triaxial even after stress relaxation, thus ensuring the specimens proposed were entirely suitable for further creep damage studies.

Generally, agreement between predictions and measurements was good. In general the finite element analysis tended to overestimate the magnitude of residual stress at the centre of the specimens, and slightly underestimate the magnitude of residual stress near the surface. Two reasons for this discrepancy are proposed, both concerning the material model assumed in the numerical models. The first reason is due to the use of a kinematic hardening material model. It is thought use of a more realistic mixed hardening model would improve numerical predictions. The second reason is that during quenching the surface of the specimens experienced rapid cooling for a time period of less than about 10 seconds. This suggests that the material underwent relatively high strain rates. However, the finite element analysis used conventional quasi-static yield data.

The results also showed that predictions and measurements were relatively insensitive to the choice of specimen material (by considering the results for type 316L and 316H specimens) and the size of the specimens. One of the original assumptions made in the research was that the effect of phase transformations on the final residual stress state for austenitic stainless steels was negligible, and this has been vindicated. It is assumed, however, that if the specimens were fabricated from a material which underwent a phase change during quenching, such as a ferritic steel, then additional residual stresses would have been generated from volumetric changes associated with the phase change.

It was also reassuring to note that measurement results obtained on three different neutron diffractometers (and four independent instruments) were entirely consistent. The only practical difference observed from a residual stress measurement perspective was in the time required to make the measurements, which was in direct proportion to the neutron flux available. In conclusion, the work in this paper set out to fulfill two prerequisites necessary to test the hypothesis that highly triaxial residual stress fields may be sufficient to promote creep damage in thermally aged components, even in the absence of in-service loads. These were the provision of test specimens containing a highly triaxial residual stress field and an accurate knowledge of how this residual stress field relaxes due to creep. Both of these prerequisites have been satisfied with the precaution that lower than predicted residual stresses were measured.

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