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

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⁴Dept. of Mechanical, Materials & Manufacturing Engineering, University of Nottingham, Nottingham NG7 2RD, UK

ABSTRACT
The high temperature low cycle fatigue behaviour of P91 weld metal (WM) and weld joints (cross-weld) is presented. Strain-controlled tests have been carried out at 400 °C and 500 °C. The cyclic behaviour of the weld material (WM) and cross-weld (CW) specimens are compared with previously published base material (BM) tests. The weld material is shown to give a significantly harder and stiffer stress-strain response than both the base material and the cross-weld material. The cross-weld tests exhibited a cyclic stress-strain response which was similar to that of the base material. All specimen types exhibited cyclic softening but the degree of softening exhibited by the cross-weld specimens was lower than that of the base material and all-weld tests. Finite element models of the base metal, weld metal and cross-weld test specimens are developed and employed for identification of the cyclic viscoplasticity material parameters. Heat affected zone (HAZ) cracking was observed for the cross-weld tests.

1. INTRODUCTION
P91 martensitic steels are widely employed in fossil-fuel, steam turbine power plants due to their excellent mechanical properties at high temperatures; these include high resistance to creep and oxidation, low coefficient of thermal expansion, as well as good weldability. However failures commonly occur at the welded regions of P91 power plant piping connections. Traditionally power plants have operated in a continuous, uninterrupted manner,
where start-ups and shut-downs are relatively infrequent. A considerable amount of work has been carried out in understanding power plant weldments under sustained loads at elevated temperature i.e. the creep behaviour of welded joints (see [1-8]). However, modern power plants are required to operate in a flexible mode of operation in order to compliment the unpredictable nature of renewable energies and to respond quickly to peak demands. This has led to more frequent start-stop plant operation, subjecting plant systems to cyclic (fatigue) loading. To this end, a significant amount of research has been reported on modified 9Cr-1Mo steels, in terms of high temperature low cycle fatigue behaviour (HTLCF) (e.g. [9 - 13]); however, relatively little is known about power plant weldments in terms of their HTLCF and creep-fatigue (CF) behaviour. HTLCF tests carried out on specimens fabricated from modified 9Cr-1Mo welded joints have been investigated [14 - 18]. Yang et al., [14] reported that welded joint specimens exhibited similar fatigue lives to base material. Mannan et al. [15] reported that with the introduction of a tensile hold, the fatigue life of P91 steel decreased rapidly with increasing hold time. A tensile hold of 1 h reduced life to approximately 25% of the continuous cycling life (i.e. no hold time). The reduction in life under creep-fatigue conditions is primarily attributed to the reduction in strength of P91 steel due to microstructural degradation associated with the coarsening of precipitates and dislocation substructures. Takahashi [16] carried out CF tests on P91 welded specimens. Comparisons made with the base material under identical testing conditions revealed that the welded specimens exhibited reduced life across all strain ranges and temperatures tested. It was also reported that cycles to failure decreased with increase in hold period. These observations are supported by results presented by [17], where the effect of hold period was found to reduce life compared with pure HTLCF type tests. In addition; a hold period in compression was found to be more detrimental to fatigue life than hold periods in tension. The lowering of life under compression hold was attributed to the increased rate of crack propagation due to oxidation effects. Shankar et al [17, 18] reported that the stress response of both the base metal and the weld joint are similar. It was also reported that the base material and the weld joint exhibited similar fatigue lives at 550 °C; however, at 600 °C lower lives were observed for the weld joints. Sandhya et al. [19] evaluated the HTLCF behaviour of modified 9Cr-1Mo weld joints, tested in air and dynamic sodium environment. The sodium environment resulted in increased fatigue life. This was attributed to the inhibited formation of surface oxidation, which retarded the onset of crack initiation. The present paper is concerned with experimental characterisation of the weld regions of P91 steel and identification of the temperature dependent material parameters for a cyclic viscoplasticity material model which incorporates non-linear kinematic hardening (NLKH), isotropic softening and strain-rate (creep) effects.
2. EXPERIMENTAL

2.1 FABRICATION OF TEST SPECIMENS

Base material (BM) test specimens were fabricated from an ex-service P91 header which had undergone 35,168 hrs of service. Prior to installation, the header was manufactured by a hot rolling process, normalised at 1050 °C for 0.5 h and tempered at a temperature of 765 °C for 1 h. The chemical composition of the base material is given in Table 1. The BM specimens were fabricated parallel to the rolling direction of the header. The BM tests were carried out on an Instron 8862 TMF test rig at the University of Nottingham; see [20], for more details on the experimental configuration. The chemical composition of the P91 electrode used here for the weld is also given in Table 1. A section of the header was used to generate a weld, from which P91 welded specimens could be fabricated. This was achieved by cutting the header in two and machining a chamfered edge and a square edge. The header was then re-aligned and a (P91) backing plate was affixed to allow deposition of weld material (WM) between the gap created by the two edges. The weld itself was generated by shielded metal arc multi-pass welding, which was consistent with the method employed to carry out a field repair of such a header. Upon completion of the welding process, a post weld heat treatment process was carried out. This was performed using heating blankets which were applied to the internal and external surfaces of the header, maintaining the weld region at a constant temperature of 760 °C for a period of 80 minutes, after which the weld was allowed to cool at a rate of 50 °C/hour. The header (inclusive of the weld junction) was then cut into strips, one of which is shown in Figure 1. Figure 1 shows a cross-sectional view of the weld (revealed using Marbles reagent). It can be seen that one of the fusion boundaries between the weld and the base material is perpendicular to the longitudinal direction of the strip. This ensures that the fusion boundary is perpendicular to the longitudinal axis of the test specimens. From these strips cylindrical threaded-end specimens with a gage length of 14 mm were machined, as illustrated in Figure 2.

Figure 3 (a and b) depicts the positions from which the specimens were fabricated relative to the base and weld material. Figure 3a shows the specimen positioned in the middle of the weld material, creating an all-weld test specimen. The use of the chamfered edge creates an extended gap to ensure that the entirety of the specimen gage length is comprised of weld material. Figure 3b illustrates the position of the cross-weld specimen, whereby the heat affected zone (HAZ) is positioned in the middle of the gage length, creating a cross-weld `(CW) specimen. The total width of the HAZ was found to vary between approximately 2.5 and 3 mm, depending on specimen, which is consistent with measurements given by [16].
2.2 TESTING PROCEDURE

All testing of the welded type specimens was carried out in the HTLCF test rig (Instron 8800) at NUI Galway. Heating of the specimen was achieved via a furnace heating system controlled to within ± 3 °C. Fully reversed, total strain-controlled tests were carried out at temperatures of 400 °C and 500 °C for three different strain ranges of 0.6%, 0.8% and 1% for a strain ratio of $R = -1$, at a constant strain-rate of 0.033% s$^{-1}$. Strain was measured using a high temperature extensometer within a closed-loop feedback control system. Creep-fatigue tests were carried out using a trapezoidal waveform, at 400 °C and 500 °C for tensile hold times of 120 s.

The fatigue life, $N_f$, for each specimen was taken as the cycle number corresponding to a 30% drop in load range relative to the load range of the 150th cycle, which is consistent with previous testing of similar materials [22]. Experience with testing of this material has shown that this is a reasonable measure of number of cycles to accentuation of damage leading to fracture within a negligible number of additional cycles and allowed comparison with as-new P91 base material tests [21].

3. RESULTS

Figures 4 and 5 show the measured stress-strain loops for the first cycle for each of the WM and CW test specimens at 400 °C. Figure 6 shows the cyclic softening evolution with number of cycles for three different strain-ranges at 400 °C for the WM. Figure 7 shows the measured response of the CW specimen at 500 °C at $N = 1$, and Figure 8 shows the cyclic softening behaviour of the CW tests at 500 °C, for different strain ranges. Some slight initial hardening is observed over the 10 cycles followed by continuous softening. The 0.8% and 1% tests display similar softening behaviour over the first 100 cycles.

A comparison of the WM and CW first cycle responses at a strain range of 0.8% at 400 °C with the corresponding BM results from [21, 23] is shown in fig 9a. It is clear that the WM response is significantly harder than that of the BM and CW tests. The CW tests are similar to the BM tests in terms of stress range, although the CW tests exhibit slightly larger inelastic strain range. Figures 9b, 9c and 9d show corresponding comparisons for 0.6% and 500 °C, in all cases showing the same trend in terms of comparative WM behaviour.

Figures 10 and 11 show the stress versus plastic strain amplitude data extracted at half-life ($N_f/2$, where $N_f$ is defined in terms of 30% load drop) from the test results for the WM and CW test specimens, for 400 °C and 500 °C, in comparison with the BM data from [21, 23]. This highlights the cyclically harder response of the WM relative to
the CW and BM specimens, particularly at 400 °C, at which the CW gives a response very similar to the BM; this, in turn, indicates that the BM dominates the response of these test specimens. At 500 °C the WM exhibits the hardest response; however, in this case, the CW tests show a significantly harder response than the BM tests, suggesting a more significant influence of the WM on the CW response at the higher temperature.

A power law relationship is commonly used to represent this relationship for cyclic analyses of components and materials, as follows:

$$\frac{\Delta \sigma}{2} = K' \left(\frac{\Delta \varepsilon}{2}\right)^{n'}$$  \hspace{1cm} (1)

where $K'$ and $n'$ are the cyclic strength coefficient and cyclic strain hardening exponent, respectively. The temperature-dependent $K'$ and $n'$ values were extracted for the half-life, $N_f/2$, (where $N_f$ corresponds to 30% load drop) and are presented in Table 2.

Figure 12 shows the measured stress relaxation behaviour for the three different types of specimen from the first quarter cycle of a creep-fatigue tensile hold test at a temperature of 500 °C. The peak tensile strain was held constant at 0.5 % for a period of 120 s.

3.1 IDENTIFICATION OF HEAT AFFECTED ZONE

The HAZ has been found to be comprised of narrow sub-regions of coarse and fine-grained microstructures. Numerous researchers have reported [4-7] that these sub-regions have different creep strengths. Samples for optical microscopy were extracted from tested specimens which were etched using Vilella's reagent (5 ml HCl + 100 ml H$_2$O + 1 g Picric acid). Figure 13 shows cracking in the HAZ of the CW test specimens; cracking was also observed in the BM of the CW specimens. Figure 14 shows a Vickers hardness profile measured across the sectioned CW test specimen. This also corroborates the higher hardness of the WM region, as demonstrated in the measured cyclic responses. Figure 15 (a to d) shows micrographs of the different material zones annotated in Figure 14. Figure 15a and b show the grain structure of the WM and BM, respectively. The figures show that the grain sizes of both materials are similar. Figure 15c shows the fusion boundary between the WM and the fine-grain HAZ (FG-HAZ) and Figure 15d shows the coarse-grain HAZ (CG-HAZ), intercritical HAZ (IC-HAZ) and the BM regions on the CW specimen of Figure 14. The identification of these regions allowed estimation of the HAZ width and location. The HAZ cracks are seen to be consistent with Type IV cracking, i.e. located in the IC-HAZ region.
3.2 CONSTITUTIVE BEHAVIOUR AND FINITE ELEMENT MODELLING.

The present paper is particularly interested in understanding the constitutive behaviour of the WM and HAZ under cyclic loading conditions. In order to predict the material response in this region a series of finite element (FE) models were utilised incorporating the two-layer viscoplasticity material model (e.g. see [24-26]) to simulate the cyclic stress-strain behaviour of the BM, WM and HAZ at various temperatures and strain ranges. This material model can be represented by an elastic-plastic network in parallel with an elastic-viscous network, as depicted in the rheological diagram of Figure 16.

The contributions of elastic-plastic and elastic-viscous networks are apportioned by a user-specified parameter, \( f \), which is defined by the following relationship:

\[
f = \frac{K_v}{K_v + K_p}
\]

where \( K_p \) and \( K_v \) are the elastic moduli of the elastic-plastic and elastic-viscous networks, respectively, and \( K_v + K_p \) is the total instantaneous elastic modulus which is taken to equal Young’s modulus. Figures 17a to 17c show the FE models developed for the BM, WM and CW specimens, respectively. The constitutive parameters for the BM and WM specimens were identified first, from the corresponding single-material tests, and validated via comparison of the FE and measured responses. These identified BM and WM parameters were then utilised within a multi-material CW simulation (Figure 17c) to identify the constitutive behaviour of the HAZ material (treated as a single material zone). The measured width of HAZ was found to vary between about 2.5 mm and 3 mm. The FE results presented employ a value of 2.7 mm, but a sensitivity analysis to this width established negligible effect on the predicted stress-strain response of the CW specimen. The overall height of the model is 12.5 mm which corresponds to the distance between the extensometer legs.

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

3.2.1 Step 1 - Estimation of NLKH parameters

The cyclic yield stress \( k \) is identified from the linear portion of the true stress versus plastic strain loop at the half life, averaging across the different strain ranges at each temperature, following the approach of Lemaitre and
Chaboche [27]. The NLKH coefficients $C$, $\gamma$ are identified using the following relationship between the measured cyclic stress and plastic strain amplitude [27]:

$$\frac{\Delta\sigma}{2} - k = \frac{C}{\gamma} \tanh \left( \frac{\Delta\epsilon_p}{2} \right)$$

(3)

The ratio $C/\gamma$ is determined as the asymptotic value of $\Delta\sigma/2-k$ versus the inelastic strain amplitude, which is taken to be the value of the backstress at the half-life. For example, Figure 18 shows the identification of $C$ and $\gamma$ for the WM material at 400 °C and 500 °C. The identified NLKH constants for the WM are presented in Table 3. Figure 19 shows the comparison of predicted NLKH stress-strain responses and the measured data at the half-life, at 500 °C, for two strain ranges. The NLKH constants for the BM at 400 °C and 500 °C are identified in the same manner and are presented in Table 4.

Once the WM and BM NLKH parameters have been identified and validated using the single material models, a CW FE model was employed to simulate the stress-strain response of the CW tests. In the first instance the material constants $C$ and $\gamma$ of the HAZ are assumed to equal those of the BM. The constant $k$ was taken to be a value $\sim 4 \%$ lower than the corresponding value of $k$ for the BM. This was based on the observation that the results of Figure 14 show that hardness within the HAZ is $4 \%$ lower than that of the BM. The Young's modulus of the HAZ was determined using a three-spring model with the three in-series springs representing the three material zones, as follows:

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

(4)

where $k_T$, $k_{BM}$, $k_{HAZ}$, $k_{WM}$ are the material stiffnesses and $x_T$, $x_{BM}$, $x_{HAZ}$, $x_{WM}$ are the lengths of each material zone. With the estimated constants $E$ and $k$, the global strain-controlled response of the CW FE model was compared with the half-life test data to iteratively identify $C$ and $\gamma$. Fig 20 depicts a comparison between the resulting NLKH response for the CW FE model and the corresponding CW test for two strain ranges at a temperature of 500 °C.

### 3.2.2 Step 2 - Identification of isotropic hardening parameters

The cyclic softening curves presented in Figure 6 illustrate substantial softening behaviour occurring during testing of the WM specimens. The constant $Q$ is the asymptotic value of the isotropic variable, $R$, at the stabilized cycle. Due the fact that the continuous cyclic softening occurred for all tests, the constant $Q$ was taken to be the stress
difference between the first cycle and that taken at the half-life. In fact, as discussed in [20, 28] and elsewhere, the softening response can be divided into three stages, an initial non-linear stage, a linear steady-state stage and a damage (failure) stage, also non-linear. The model employed here for evolution of isotropic softening, which is the Chaboche model, e.g. see [27], only deals with the primary non-linear stage and is given as follows:

\[ R = Q(1 - e^{-bp}) \]  

(5)

\( R \) is calculated from the test data as the change in maximum (or minimum) stress in each cycle relative to that of the first cycle. The equivalent plastic strain \( p \) is equal to two times the plastic strain range for each cycle and its value accumulates as cycles increase. The constant \( b \) is determined by fitting Equation 5 to the test data, as shown in Figure 21. The identified isotropic material constants for the WM and BM are given in Tables 3 and 4, respectively. This figure shows the correlation for the BM and WM for a strain range of 0.8 % at 500 °C. Good agreement is observed between the model and the measured data from the initial cycle to the half-life, after which the model and the test behaviours begin to diverge. The implementation of a more complex softening model, with a linear secondary softening stage, e.g. [20], will be investigated in future work. This would allow for more accurate prediction of stress evolution.

Essentially, the softening is related to the cyclic deformation effects on evolution of dislocation density (reducing) and annihilation of dislocations at low-angle (sub-grain) boundaries, eventually leading to the disappearance of the low angle boundaries, effectively coarsening the sub-grain microstructure [29]. Future work will address this through the development of physically-based constitutive modelling.

The HAZ isotropic parameters were again (as for the NLKH parameters) obtained by iterative comparison of the CW FE model response and the measured response, with the WM and BM values taken as input. The identified value of \( Q \) is significantly lower than for the BM and WM, indicating that the HAZ does not cyclically soften as much as WM or BM. fig 22 shows the comparison between the resulting CW model response (combined isotropic-NLKH) and the test data at \( N = 1 \). Figure 23 shows a comparison between the predicted cyclic softening behaviour of the combined isotropic-NLKH CW model with the measured cyclic softening behaviour of the CW test at 500 °C at \( \Delta \varepsilon = 0.8 \% \).

3.2.3 Step 3 - Identification of creep constants
The Norton creep constants $A$ and $n$ for the WM and BM were identified from the stress relaxation experimental data in the form of $\log(\dot{\varepsilon})$ versus $\log\sigma$ (see [24] for more details), using the Norton equation for steady-state creep, as follows:

$$\dot{\varepsilon} = A\sigma^n$$  \hspace{1cm} (6)

The identified constants were compared with the theoretical Norton creep relaxation response to ensure accuracy. The WM and BM constants were then utilised within the CW FE elastic-creep model under stress-relaxation conditions to facilitate identification of the HAZ creep constants. The resulting CW FE predicted stress relaxation behaviour is compared with the test data in fig 24 and the identified creep constants are given in Tables 3 to 5.

Figure 25 shows the validation of the resulting identified cyclic viscoplasticity parameters for WM (Table 3), using the two-layer material model, for the first cycle and at the half-life at 500 °C. Figures 26 and 27 show the sample results for validation of the identified HAZ cyclic viscoplasticity parameters (Table 5), for $N = 1$ and at half-life against the measured data. The quality of comparison was equivalent at 400 °C and for other strain-ranges.

A single material model utilising the identified HAZ modelling constants was employed to predict the stress-strain behaviour of the HAZ material. Figure 28 shows the measured stress-strain of the WM, BM and CW at $N = 1$ compared against the predicted response of the single material HAZ model at 500 °C for $\Delta\varepsilon = 0.8\%$. The figure shows that the predicted response of the HAZ model is 31 % softer than the measured WM material response and 14 % softer than the BM. A sensitivity study on the effect of HAZ length in the FE CW specimen model established negligible effect on predicted stress-strain response of the CW specimen, varying the HAZ length between 2.5 mm and 3 mm, covering the approximate measured range of values.

4. CONCLUSIONS

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

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Fig. 28. Comparison between predicted HAZ stress-strain response and measured BM, WM and CW responses at N = 1

Table 1: Chemical composition (wt%) of the P91 base material prior to service
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<th>Element</th>
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Table 2. Cyclic material constants for BM, WM, and HAZ for 400 °C and 500 °C.
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Table 3. Cyclic viscoplasticity material parameters for WM.
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<th>$C$ (MPa)</th>
<th>$Q$ (MPa)</th>
<th>$b$</th>
<th>$A$ (MPa s$^{-1}$)</th>
<th>$n$</th>
<th>$f$</th>
</tr>
</thead>
<tbody>
<tr>
<td>400</td>
<td>235</td>
<td>202040</td>
<td>925.2</td>
<td>288600</td>
<td>-111</td>
<td>0.88</td>
<td>$2.3 \times 10^{-6}$</td>
<td>21</td>
<td>0.05</td>
</tr>
<tr>
<td>500</td>
<td>185</td>
<td>195375</td>
<td>889.5</td>
<td>205690</td>
<td>-122</td>
<td>1.1</td>
<td>$1.77 \times 10^{-5}$</td>
<td>20.2</td>
<td>0.05</td>
</tr>
</tbody>
</table>

Table 4: Cyclic viscoplasticity material parameters for BM.
<table>
<thead>
<tr>
<th>$T$ (°C)</th>
<th>$k$ (MPa)</th>
<th>$E$ (MPa)</th>
<th>$\gamma$</th>
<th>$C$ (MPa)</th>
<th>$Q$ (MPa)</th>
<th>$b$</th>
<th>$A$ (MPa s$^{-1}$)</th>
<th>$n$</th>
<th>$f$</th>
</tr>
</thead>
<tbody>
<tr>
<td>400</td>
<td>205</td>
<td>188000</td>
<td>1112.58</td>
<td>220290</td>
<td>-72.4</td>
<td>0.283</td>
<td>$5.19 \times 10^{-58}$</td>
<td>19.8</td>
<td>0.07</td>
</tr>
<tr>
<td>500</td>
<td>175</td>
<td>171220</td>
<td>763</td>
<td>136000</td>
<td>-96</td>
<td>0.56</td>
<td>$4.28 \times 10^{-40}$</td>
<td>13.6</td>
<td>0.1</td>
</tr>
</tbody>
</table>

Table 5. Cyclic material parameters for HAZ material.
<table>
<thead>
<tr>
<th>$T$ ($^\circ$C)</th>
<th>$k$ (MPa)</th>
<th>$E$ (MPa)</th>
<th>$\gamma$ (MPa)</th>
<th>$Q$ (MPa)</th>
<th>$b$</th>
<th>$A$ (MPa s$^{-1}$)</th>
<th>$n$</th>
<th>$f$</th>
</tr>
</thead>
<tbody>
<tr>
<td>400</td>
<td>188</td>
<td>173191</td>
<td>729</td>
<td>125000</td>
<td>-40</td>
<td>0.3</td>
<td>2.5 × 10$^{-22}$</td>
<td>26.1</td>
</tr>
<tr>
<td>500</td>
<td>169</td>
<td>161957</td>
<td>649</td>
<td>121076</td>
<td>-47</td>
<td>0.31</td>
<td>1.97 × 10$^{-21}$</td>
<td>25.56</td>
</tr>
</tbody>
</table>
fig5.tiff
500 °C

T.P. Farragher  PVT-13-1088 27

fig7.tif
$500 \, ^\circ C$

$\Delta \varepsilon_T = 0.8 \%$

- WM
- BM [21, 23]
- CW

Strain (%)

Stress (MPa)
The graph shows the stress (MPa) over time (secs) at 500 °C. The different lines represent different conditions: WM, BM [21, 23], and CW.
fig17.tiff
fig21.tiff
Figure 23. Tensile stress vs. normalized life at 500 °C. The solid line represents the model prediction, while the dotted line represents the experimental data.
500 °C
\[ \Delta \epsilon_f = 0.8\% \]

Stress (MPa)

Strain (%)

-600

-400

-200

0

200

400

600

N=1, Model
N=1, Expt.
N=Half-life, Model
N=Half-life, Expt
500 °C
$\Delta \varepsilon_T = 0.8 \%$

Stress (MPa)

Strain (%)

HAZ (Predicted)
WM (Measured)
BM (Measured), [21, 23]
CW (Measured)

fig28.tiff