Assessment of creep rupture properties for dissimilar steels welded joints between T92 and HR3C

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ABSTRACT Dissimilar steels welded joints, between ferritic steel and austenitic stainless steel, are always encountered in high-temperature components in power plants. As two new grade ferritic steel and austenitic stainless steel, T92 (9Cr0.5Mo2WVNb) and HR3C (TP310HCbN), exhibit superior heat strength at elevated temperatures and are increasingly applied in ultra-supercritical (USC) plants around the world, a complete assessment of the properties for T92/HR3C dissimilar steels welded joints is urgently required. In this paper, metallographic microstructures across the joint were inspected by optical microscope. Particularly, the creep rupture test was conducted on joints under different load stresses at 625 \( ^\circ \)C to analyse creep strength and predict their service lives, while their fractograph were observed under scanning electron microscope. Additionally, finite element method was employed to investigate residual stress distribution of joints. Results showed that the joints were qualified under USC conditions, and T92 base material was commonly the weakest part of them.

Keywords creep rupture; dissimilar steels welded joints; finite element method; HR3C; T92.

INTRODUCTION

With the worsening of global energy crisis and environmental pollution, higher energy utilization and lower \( \text{CO}_2 \) emission are presently the two prior criteria for designing fossil power plants,1 which will still serve as the dominant energy form for at least 20 years.2–4 As for fossil power boilers, it is well known that the enhancement of steam parameters can facilitate improvement in thermal efficiency, which can result in reduction in not only the fossil costs but also the \( \text{CO}_2, \text{SO}_2 \) and \( \text{NO}_x \) emission.5 Concretely, compared with supercritical (SC) boilers (24 MPa, 565 \( ^\circ \)C), modern ultra-supercritical (USC) boilers operating around 30 MPa and 600 \( ^\circ \)C can lead to both an increase of 8–10% in thermal efficiency (from 37% to 45–47%) and a reduction of 20–25% in \( \text{CO}_2 \) emission.1,6–10 Development of USC boilers with increasingly higher steam parameters is an added incentive for boiler material research. Consequently, heat-resistant steels, mainly referring to ferritic steels and austenitic stainless steels, have been pursued simultaneously since the emergence of USC boilers for applications in boiler components, including superheater, reheater, header, turbine, steam piping and so on. Subsequently, for the sake of reliable services in USC boilers, a wealth of research has been carried out on these heat-resistant steels as F12 (X20CrMoV12.1), T91 (9Cr1MoVNb), TP347H (18Cr10NiNb), etc. to investigate characteristics of their base materials and performance deterioration after long-term services.11–29 Among them, the T92 (9Cr0.5Mo2WVNb), approximately similar to T91 but with a little modification in chemical compositions for preferable high temperature properties than T91, and the HR3C (TP310HCbN) are the two typical representatives of new grade ferritic steels and austenitic stainless steels for their superior heat strength above 620 \( ^\circ \)C, and will certainly have a broad application prospect in forthcoming USC boilers. However, applications of T92 and HR3C are currently a bit constrained due to the lack of sufficient literature and experience on...
creep rupture performances of them and their welded joints, let alone the dissimilar steels welded joints between them. Therefore, a thorough assessment of the comprehensive properties, particularly the creep properties of T92/HR3C dissimilar steels welded joints, seems pretty urgent.

In this paper, besides various conventional mechanical tests including tensile test, bending test and hardness survey, optical microscope (OM) was also applied to inspect the metallographic microstructures across the dissimilar steels welded joint between T92 and HR3C. Moreover, creep rupture test was particularly employed under different load stresses at 625 °C to investigate the creep features of the joints, whose fractograph was then observed by using scanning electron microscope (SEM) as well. Furthermore, the residual stress distribution of the welded joint after welding was calculated by using finite element method (FEM), which was a tentative approach to evaluate residual stress of the dissimilar steels welded joint between these two novel materials through the computational simulation method. Finally, based on the analysis results, not only the creep rupture performances and the degradation curves of T92/HR3C dissimilar steels welded joints were reported, but also the mechanism of voids initiating creep rupture was concretely discussed, which may have critical significance in both service-life prediction and future heat-resistant steels preparation for boiler components.

**EXPERIMENTAL**

Tested materials were nominal T92 and HR3C heat-resistant steels with scales of 48O D × 8.4 mm thick and 48.26O D × 10.16 mm thick, respectively. Chemical compositions as well as heat treatment conditions of their base materials are listed in Table 1, which are in accordance with the requirements of ASME SA-213 T92 and TP310HcBN specifications. Etched in agent of picric acid (2, 4, 6-trinitrophenol) 1.25 g, HCl 20 ml, ethanol 10 ml and H2O 10 ml for 40 s, the metallographic microstructure of T92 sample is presented in Fig. 1a, which displays a typical tempered lath martensitic microstructure. Similarly, metallographic microstructure of HR3C sample was also obtained after being etched in the agent of CuSO4 4 g, HCl 20 ml and ethanol 20 ml for 20 s. As is shown in Fig. 1b, HR3C presents a fine-grained

### Table 1 Chemical compositions and heat treatment conditions of T92 and HR3C samples (wt%)

<table>
<thead>
<tr>
<th>Elements</th>
<th>C</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>Nb</th>
<th>Ni</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>N</th>
<th>Al</th>
<th>W</th>
<th>B</th>
</tr>
</thead>
<tbody>
<tr>
<td>T92 Sample</td>
<td>0.11</td>
<td>8.76</td>
<td>0.36</td>
<td>0.21</td>
<td>0.059</td>
<td>0.25</td>
<td>0.46</td>
<td>0.016</td>
<td>0.002</td>
<td>0.39</td>
<td>0.044</td>
<td>0.01</td>
<td>1.63</td>
<td>0.0033</td>
</tr>
<tr>
<td>ASME SA-213</td>
<td>0.07–0.13</td>
<td>8.50–9.50</td>
<td>0.30–0.60</td>
<td>0.15–0.25</td>
<td>0.04–0.09</td>
<td>≤0.40</td>
<td>0.30–0.60</td>
<td>≤0.020</td>
<td>≤0.010</td>
<td>≤0.50</td>
<td>0.03–0.07</td>
<td>≤0.04</td>
<td>1.50–2.00</td>
<td>0.001–0.006</td>
</tr>
<tr>
<td>HR3C</td>
<td>0.06</td>
<td>24.63</td>
<td>/</td>
<td>/</td>
<td>0.49</td>
<td>20.29</td>
<td>1.24</td>
<td>0.012</td>
<td>0.001</td>
<td>0.39</td>
<td>0.24</td>
<td>/</td>
<td>/</td>
<td>/</td>
</tr>
<tr>
<td>ASME SA-213</td>
<td>≤0.10</td>
<td>23.00–27.00</td>
<td>/</td>
<td>/</td>
<td>0.20–0.60</td>
<td>17.00–23.00</td>
<td>≤2.00</td>
<td>≤0.030</td>
<td>≤0.030</td>
<td>≤1.50</td>
<td>0.15–0.35</td>
<td>/</td>
<td>/</td>
<td>/</td>
</tr>
<tr>
<td>TP310HcBN</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>/</td>
</tr>
</tbody>
</table>

Heat treatment conditions:
T92: 1050 °C × 20 min (normalizing) + 760 °C × 60 min (tempering).
HR3C: solution-treated at 1110 °C minimum.

**Fig. 1** Metallographic microstructures of tested base materials (a) T92, 1500× (b) HR3C, 200×.
Table 2 Chemical compositions of welding wire ERNiCr-3 (wt%)

<table>
<thead>
<tr>
<th>Elements</th>
<th>C</th>
<th>Mn</th>
<th>Fe</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cu</th>
<th>Ni</th>
<th>Ti</th>
<th>Cr</th>
<th>Nb+Ta</th>
</tr>
</thead>
<tbody>
<tr>
<td>ERNiCr-3 welding wire</td>
<td>0.030</td>
<td>2.90</td>
<td>1.30</td>
<td>0.004</td>
<td>0.001</td>
<td>0.04</td>
<td>0.01</td>
<td>72.5</td>
<td>0.31</td>
<td>20.0</td>
<td>Nb 2.40</td>
</tr>
<tr>
<td>ASME SFA-5.14 (AWS) ERNiCr-3</td>
<td>≤0.10</td>
<td>2.5–3.5</td>
<td>≤3.0</td>
<td>≤0.030</td>
<td>≤0.015</td>
<td>≤0.50</td>
<td>≤0.50</td>
<td>≥67.0</td>
<td>≤0.75</td>
<td>18.00–22.00</td>
<td>2.0–3.0</td>
</tr>
</tbody>
</table>

Table 3 Tensile test results of T92/HR3C dissimilar steels welded joints

<table>
<thead>
<tr>
<th>Sample No.</th>
<th>Tensile strength (σs, MPa)</th>
<th>Rupture position</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>707</td>
<td>T92 base material</td>
</tr>
<tr>
<td>2</td>
<td>699</td>
<td>T92 base material</td>
</tr>
<tr>
<td>T92 specification</td>
<td>≥620</td>
<td>/</td>
</tr>
<tr>
<td>HR3C specification</td>
<td>≥655</td>
<td>/</td>
</tr>
</tbody>
</table>

Table 4 Bending test results of T92/HR3C dissimilar steels welded joints

<table>
<thead>
<tr>
<th>Bending style</th>
<th>Sample No.</th>
<th>Test condition</th>
<th>Result</th>
</tr>
</thead>
<tbody>
<tr>
<td>Face bending</td>
<td>1</td>
<td>D = 4T, α = 180°</td>
<td>Qualified</td>
</tr>
<tr>
<td></td>
<td>2</td>
<td>D = 3T, α = 90°</td>
<td>Qualified</td>
</tr>
<tr>
<td>Back bending</td>
<td>1</td>
<td>D = 4T, α = 180°</td>
<td>Qualified</td>
</tr>
<tr>
<td></td>
<td>2</td>
<td>D = 3T, α = 90°</td>
<td>Qualified</td>
</tr>
</tbody>
</table>

'D' denotes the bending diameter; 'T' denotes the material thickness; 'α' denotes the bending angle.

RESULTS AND DISCUSSION

Mechanical tests results

As is clear from Table 3, the T92/HR3C dissimilar steels welded joints exhibit qualified tensile strength, and the T92 base material part is their weakest region under load stresses. Table 4 reveals the sign that the welded joints also present eligible toughness. In addition, no cracks were founded on the bended surfaces. Hardness survey results are displayed in Fig. 3, which indicates that the T92 base material and weld seam are the two low-hardness parts.

Metallographic microstructure inspection

As is well known, a dissimilar steels welded joint is often approximately divided into five regions, i.e. base material A, HAZ of A, weld seam, HAZ of B and base material B. In this paper, the metallographic microstructures of the five regions across the T92/HR3C dissimilar steels welded joint were inspected under OM.

Figure 4a presents the metallographic microstructure of T92 base material after welding, in which no signs of differential are detected when compared with that of original material in Fig. 1a. However, in the HAZ of T92,
obvious sorbitic microstructure with coarsened laths could be observed in Fig. 4b. The width of the sorbite lath is nearly two times that of martensite lath, which may result in an increase of hardness in HAZ of T92 as well as a decrease of toughness in this region simultaneously. The mechanism can be explained that coarser laths may block the growth of cracks under stresses, and eventually lead to brittle rupture in this region. Near the weld seam, a distinct boundary between HAZ of T92 and weld seam can be observed in Fig. 4c. Figure 4d shows the stripe-shaped austenitic microstructure of the weld seam, whose stripe width has already reached around 20 μm. Correspondingly, like Fig. 4c, Fig. 4e gives the boundary between weld seam and HAZ of HR3C. Average grain size of the austenite in HAZ of HR3C is about 6 (Fig. 4f), and carbides as M23C6 have precipitated at the grain boundaries. Compared with the average grain size of 7 in HR3C base material, the coarsened grains in HAZ of HR3C may also lead to increase in hardness in this region. Fig. 4g is the metallographic microstructure of HR3C base material, which shows no changes with its original status too. Collectively, the schematic diagram as well as the metallographic microstructures distribution across the T92/HR3C dissimilar steels welded joint is displayed in Fig. 4h.

Creep rupture test results

Table 5 lists the creep rupture test results under load stresses ranging from 110 to 180 MPa at 625 °C. This phenomenon that rupture positions vary with change in load stresses is in accordance with the results in the literature. Together with the tensile test results, it can be concluded that the T92 base material part is actually the weakest region of the T92/HR3C dissimilar steels welded joint under relatively higher stresses. Nevertheless, compared with the research data of Falat et al., the rupture time of our T92/HR3C dissimilar steels welded joints under 120 MPa at 625 °C actually even exceeds the counterpart value of P92/P92 ('P' denotes the word 'pipe', which contains the same chemical compositions of 'T', i.e. 'tube', but with a larger diameter and thickness) similar steels welded joint, 1174 h.

According to the classic theory about creep rupture test, a double logarithmic relationship between load stress σ and rupture time t, i.e. lgr versus lgt, can be expressed in Eq. (1), in which the letters A and B both denote the material parameters of the tested samples. Accordingly, Fig. 5 presents the plot of lgrσ versus lgt and the linear fitting result of it.

\[
\log t = \log A - B \log \sigma. \tag{1}
\]

Thus, the threshold steam stress for service of the T92/HR3C dissimilar steels welded joints exposed at 625 °C, which can be determined by extrapolation of the fitted line to \(10^5\) h, is 61.89 MPa. However, in practical applications, some other unpredicted factors may also affect the creep strength of welded joints. Hence, safety coefficient n is always adopted to modify the predicted threshold stress obtained from linear extrapolation. Consequently, permitted stress [σ] can be expressed as Eq. (2),
Fig. 4 Metallographic microstructures of the five different regions across the welded joint (a) T92 base material, (b) HAZ of T92, (c) boundary between HAZ of T92 and weld seam, (d) weld seam, (e) boundary between weld seam and HAZ of HR3C, (f) HAZ of HR3C, (g) HR3C base material, (h) schematic diagram of welded joint.
Table 5 Creep rupture test (625 °C) results of T92/HR3C dissimilar steels welded joints

<table>
<thead>
<tr>
<th>Sample No.</th>
<th>Load stress (σ, MPa)</th>
<th>Rupture Time (h)</th>
<th>Rupture position</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>180</td>
<td>378</td>
<td>T92 base material</td>
</tr>
<tr>
<td>2</td>
<td>160</td>
<td>865</td>
<td>T92 base material</td>
</tr>
<tr>
<td>3</td>
<td>150</td>
<td>930</td>
<td>T92 base material</td>
</tr>
<tr>
<td>4</td>
<td>140</td>
<td>1640</td>
<td>T92 base material</td>
</tr>
<tr>
<td>5</td>
<td>130</td>
<td>2787</td>
<td>HAZ of T92</td>
</tr>
<tr>
<td>6</td>
<td>120</td>
<td>3164</td>
<td>Weld seam</td>
</tr>
<tr>
<td>7</td>
<td>110</td>
<td>4182</td>
<td>Weld seam</td>
</tr>
</tbody>
</table>

Fig. 5 Double logarithmic plot of load stress versus rupture time for the welded joints.

and the value of $n$ is generally ranging from 1.2 to 1.65.22

$$[\sigma] = \frac{\sigma_{625{^\circ}C}}{n}.$$  

In this case, we take $n = 1.5$, and therefore the permitted stress $[\sigma]$ of T92/HR3C dissimilar steels welded joints can be determined from Eq. (3), whose result value 41.26 MPa also exceeds the steam stress of USC conditions.

$$[\sigma] = \frac{61.89}{1.5} = 41.26 \text{ MPa}.$$  

In terms of service life prediction for high-temperature components in boilers, the Larson–Miller equation, seen in Eq. (4), is always applied to estimate the allowable stresses under specific steam conditions:32

$$P = T(C + \lg t_r),$$  

where $T$ is the absolute temperature in K, and $t_r$ is the rupture time in hours. Actually, $C$ is a constant depending on different materials. Owing to the fact that T92 base material is the weakest part of T92/HR3C dissimilar steels welded joint, the value of $C$ can be determined from T92 matrix material. As for T92, the value of $C$ differs from 19.5 to 36 according to a variety of references involved.33,34 However, thanks to sufficient practical application experiences of T92 in the past decade, the value of $C$ has been modified according to lots of firsthand data. Hence, a value of 25 is now commonly adopted for $C$ to predict the service lives of T92 for better accordance with the actual conditions. Consequently, in this case, the rupture data for the T92/HR3C dissimilar steels welded joints can be plotted in terms of $\lg \sigma$ versus Larson–Miller parameters (LMP) in Fig. 6, where the LMP is expressed as given in Eq. (5).

$$LMP = T(25 + \lg t_r).$$  

Also, the $\lg \sigma$ versus LMP curve displayed in Fig. 6 is then polynomial fitted for the convenience of service life prediction for the welded joints under different stresses. As for USC boilers, service steam stresses commonly range from 30 to 40 MPa, whose LMP values can then be easily read from Fig. 6. Table 6 lists the LMP values and their corresponding rupture times under stresses of 30, 35 and 40 MPa at 625 °C.

Table 6 LMP values and rupture times under stresses of 30, 35 and 40 MPa at 625 °C

<table>
<thead>
<tr>
<th>Service stress (MPa)</th>
<th>LMP value ($\times 10^{-3}$)</th>
<th>Rupture time (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>30</td>
<td>27.15</td>
<td>168060</td>
</tr>
<tr>
<td>35</td>
<td>27.01</td>
<td>119198</td>
</tr>
<tr>
<td>40</td>
<td>26.89</td>
<td>87631</td>
</tr>
</tbody>
</table>
Fractograph observation of creep ruptured samples

According to the creep rupture test results, the rupture positions of the T92/HR3C dissimilar steels welded joints under different load stresses, mainly including three different parts, seen in Table 5. Hence, fractograph of samples under three representative load stresses: 110, 130 and 180 Mpa, whose rupture positions were respectively weld seam, HAZ of T92 and T92 base material, were then concretely inspected by means of SEM. Fracture positions of joints under 110, 130 and 180 MPa located in three different positions. Figure 7 was the macroscopic morphology of them three. Then, samples were cut from the rupture positions to observe the micromorphologies of their cross-sections under SEM.

Figure 8 presents the micromorphologies of cross-section of the ruptured sample under 180 MPa. It can be seen from Fig. 8a that the diameter of the cross-section is about 2.5 mm. Compared with its original value of 6.0 mm, the reduction in area $\psi$ can be calculated as about 80%. As is shown in Fig. 8b, the cross-section is covered with densely distributed spherical creep voids, whose diameter ranges from 5 to 50 μm. The presence of creep voids, whose nucleation mechanisms still need to be further identified, commonly represents a good toughness of materials at high temperatures. In some region, three neighbouring voids with diameters of about 5 μm have coalesced into a larger wave-like one, whose axial width has already reached nearly 20 μm, as can be seen in Fig. 8c and d. This is typical evidence for ductile materials that creep rupture is led by the coalescence of microscopic cracks which are also generated by the sub-coalescence of creep voids.

Compared with the fractograph of ruptured sample under 180 MPa, the fractograph under 130 MPa displays...
an obvious brittle rupture morphology, seen in Fig. 9a. Lots of slim dissociation steps resulting from narrow sorbite laths can be found on the cross-section. Meanwhile, creep voids can also be observed in Fig. 9b. However, the amount of voids which have grown in a wide and shallow form is far less than that on the cross-section under 180 MPa. Although a small amount of dimples can be observed in Fig. 9b as well, they are not the dominant factors of rupture. This phenomenon may be attributed to the highest hardness and residual stress in HAZ of T92.

Figure 10 displays the fractograph of creep ruptured sample under 110 MPa. An obvious macroscopic dissociation step rather than macroscopic creep voids can be detected in Fig. 10a. Meanwhile, it can be learned that the reduction in area $\psi$ is less than 5%. Actually, magnified by 500 times, the zoom is filled with randomly distributed creep voids, seen in Fig. 10b. Furthermore, neighbouring voids also have the potential to coalesce into larger ones, as shown in Fig. 10c. However, at least three differentials including void depth, void area and distribution density can be determined among the voids generated, respectively, under 110, 130 and 180 Mpa. On the one hand, as is discussed above, the wide and shallow voids with diameter of about 10 $\mu$m (Fig. 10d) may be accounted for the low toughness of the weld seam. On the other hand, the larger distribution density of the creep voids under 110 MPa than that generated under 130 MPa may be accounted for the lower hardness of the weld seam. Thus, the fractograph of the ruptured sample under 110 MPa presents an intermediate micromorphology between that under 180 and 130 Mpa, which indicates its intermediate properties of weld seam.

**Finite element method results**

FEM is the most widely used computational simulation method for its superiorities as convenience, effectiveness, accuracy, etc., and is always applied in physical field analyses including thermal field, force field, magnetic field, etc. and their coupled fields. In this case, the residual
stress of the T92/HR3C dissimilar steels welded joint after welding was calculated by the FEM software ANSYS 10.0.

As the width of the welded joint is not sufficiently large, their effect on residual stress distribution is generally neglected. Thus, the three-dimensional (3D) thermal-stress coupled field analysis can be simplified as 2D axisymmetric problem. The 2D FEM model (after being meshed) is shown in Fig. 11a by using PLANE 13 2D coupled field solid element. Also, birth–death element was applied in the weld seam with 21 layers to simulate the welding procedure, seen in Fig. 11b, of all the layers that were initially dead, the lowest layer was firstly activated when being welded, then the rest ones would be activated layer by layer above the former ones sequentially. Displacement in horizontal and vertical directions of the left and the right borders of the welded joint set zero were the boundary conditions. The physical properties of T92 and HR3C base materials as well as ERNiCr-3 welding wire used in calculation are listed in Fig. 12 and Table 7, among which the shear modulus is commonly regarded as one-tenth of the elastic modulus.

The results of the computational analysis are presented in Fig. 13, which shows the Von Mises equivalent stress distribution across the welded joint. It can be obviously observed from Fig. 13a that residual stress mainly accumulates at the two HAZ regions. The largest residual stress occurs at the HAZ of T92 region, seen in Fig. 13b, which has already reached around 330 MPa. Meanwhile, the residual stress of the HAZ of HR3C is also up to around 250 MPa. This could be ascribed to the preferable ductility of austenite HR3C, as it is able to offset stresses through plastic deformation. The FEM result verified the fact that residual stresses usually accumulate at HAZ of the undermatched part, i.e. the ferrite HAZ, in ferrite/austenite dissimilar steels welded joints. A conclusion can then be put forward that the selection of welding wires with similar strength of ferrites could effectively relieve the residual stresses of the joints after welding; in other words, it can increase the comprehensive strength of the joints.

Comprehensive analysis

In terms of creep voids, they are commonly generated under the interaction among high temperature, load stress and ageing time. Till date, there are several classic controversial theories of the emergence of creep voids. By Greenwood, Argon, Raj and Ashby, the initiation positions of void nucleation were argued; by Hull and Rimmer, Needleman and Hancock, the controlling factors of void growth were debated; and by Stowell, Nicolaou and Semiatin, and Chokshi, the coalescence procedures of voids were discussed. In this paper, based on their works, a four-stage ‘nucleation to crack’ mechanism of creep voids was put forward to clearly understand the mechanisms of creep voids generation for the T92/HR3C dissimilar steels welded joints.

1 Nucleation

In this stage, the creep voids with irregular shapes are generated from the effect of grain boundaries sliding and/or grain matrix deformation under specific load stress and temperature in creep process. Accordingly, the voids commonly nucleate at the grain boundaries of the weakest part or the part with largest residual stress of tested material. As for the T92/HR3C dissimilar steels welded joints, the rupture positions i.e. the voids nucleation positions varied under different creep conditions, seen in Table 5. This may be accounted for the different properties of the five regions across the joint.

2 Growth

Creep voids continuously grow under constant temperature and load stress in this growth stage. However, the
ultimate void volumes may vary in a wide range of values according to different toughness of different materials. As for ductile materials like T92, the preferable toughness ensures a good freedom for the voids to grow, which may lead to a great amount of deep voids. However, in terms of brittle materials like HAZ of T92, voids’ growth is constrained in a wide and shallow form with only a small amount owing to the rigidness of matrix materials. Meanwhile, in the process of voids growth, the original voids with irregular shapes transform to uniformly spherical ones. This can be explained that voids with irregular shapes possess higher surface-free energy,
which will lead to a strong potential to minimize the surface area to reduce this surface-free energy. Thus, the spherical-shaped voids, which own minimal surface area, are produced from the original irregular-shaped ones by means of atoms diffusion. Consequently, the spheroidized voids decrease the extent of stress concentration and then increase the toughness of materials at high temperatures.

Coalescence

Coalescence is usually a familiar phenomenon in cavities initiating degradations for materials. With the enlargement of at least two neighbouring voids, they tend to be combined into a larger one, which may aggravate the damages on the materials. In creep process, neighbouring voids commonly coalesce orientate along the grains boundaries, which will act as the initial form of cracks.

Crack formation

When the voids orientate coalesce to a certain extent, micro cracks emerge. As for ductile materials, the growth of cracks is always alleviated by the spherical-shaped creep voids, which will finally result in the occurrence of dimples. But for brittle materials, cracks propagate quickly and then merge into macroscopic ones, which may finally lead to intergranular rupture. Thus, the fractograph of creep-ruptured samples for brittle materials always presents dissociation fracture.
morphology with obvious dissociation steps, while for ductile materials, it commonly shows a morphology covered with voids.

It is well-known that the strength of ferritic steels is commonly not as high as that of austenitic stainless steels. Especially their heat strength at elevated temperatures is not good. Thus, aged at 625 °C under relatively higher load stresses ranging from 140 to 180 MPa, the creep rupture positions of the T92/HR3C dissimilar steels welded joints concentrated at the T92 base material part. Moreover, the rupture mechanism was predominantly ductile rupture according to the densely distributed deep creep voids seen on the cross-section of the ruptured sample (Fig. 8b). This may be accounted for the better toughness of T92 base material for its fine grained narrow martensite laths. The mechanical test results that the rupture position of tensile test was T92 base material (Table 3) and

Fig. 13 Von Mises residual stress distribution of T92/HR3C dissimilar steels welded joint (a) total distribution, (b) magnification of HAZ.
the hardness of T92 base material was the second lowest part of all the five regions (Fig. 3) also verified this fact.

With decrease of load stresses in creep test, the heat strength differential between T92 base material and other regions of the welded joints did not perform as distinct as that under higher stresses like 180 MPa. Consequently, the residual stress may play an important role for the creep rupture property of the welded joints. According to the FEM analysis result, the largest residual stress accumulated in the HAZ of T92 region. Although the residual stress had been relieved a lot during the PWHT period, it still remained at relatively the largest value in HAZ of T92 across the welded joint. Therefore, the creep rupture position under 130 MPa was the HAZ of T92 region. Moreover, owing to its sorbitic microstructure with coarsened laths, which could result in the increase of hardness (Fig. 3) and alleviate the growth of creep voids, the rupture mechanism in HAZ of T92 was brittle rupture. Thus, any dimples could be seldom detected on the cross-section of the ruptured sample under 130 MPa, as can be seen in Fig. 9a. Meanwhile, the shape of the creep voids exhibited wide and shallow form, as shown in Fig. 9b.

Weld seam melted and recrystallized in the welding process, hence the grains of which could fully grow and became coarsened seen in Fig. 4d. Also, owing to its intermediate properties of toughness, the rupture mechanism of the weld seam under 110 MPa was predominantly brittle rupture like that under 130 MPa but the amount of creep voids was larger than that under 130 MPa. The former phenomenon could be attributed to the coarsened austenitic grains in weld seam while the latter one may be accounted for the low hardness and high toughness in this region, too.

CONCLUSIONS
1. Regular mechanical properties of T92/HR3C dissimilar steels welded joints were qualified at room temperature, among which the T92 base material part presented the lowest strength and the weld seam part exhibited the lowest hardness.

2. Aged at elevated temperature as 625°C, creep rupture positions of the welded joints accumulated at T92 base material part with ductile rupture mechanism due to its lowest heat strength in this region under relatively higher load stresses; while rupture positions changed to HAZ of T92 and/or weld seam regions with brittle rupture mechanism due to their low toughness and high residual stress in this two regions under relatively lower load stresses closer to the actual USC steam conditions.

3. Variation of the metallographic microstructures in the five regions across the welded joint greatly affected the heat strength of the joint. Coarsened sorbitic structure of HAZ of T92 as well as coarsened austenitic structures of weld seam and HAZ of HR3C led to the decrease of toughness in these three regions. However, the martensitic structure of T92 and austenitic structure of HR3C base materials did not show any sign of change after welding.

4. Based on the service life prediction curves obtained from creep rupture test, the T92/HR3C dissimilar steels welded joints could ensure safe service for duration of 105 h under 30 to 40 MPa at 625 °C, which is similar to the USC steam conditions, and their allowable stresses and service lives both exceed those values of the currently used steels like T/P 91.

5. FEM analysis results showed that the HAZ of T92 was the region with largest residual stress after welding, which was in accordance with the creep rupture test results under relatively lower load stresses.

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