Microstructural Evolution of 9CrMoW Weld Metal in a Multiple-Pass Weld

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Abstract: 9CrMoW steel tubes were welded in multiple passes by gas-tungsten arc welding. The reheated microstructures in the Gr. 92 weld metal (WM) of a multiple-pass weld were simulated with an infrared heating system. Simulated specimens after tempering at 760 °C/2 h were subjected to constant load creep tests either at 630 °C/120 MPa or 660 °C/80 MPa. The simulated specimens were designated as the over-tempered (OT, below A_C1, i.e., WT-820T) and partially transformed (PT, below A_C3, i.e., WT-890T) samples. The transmission electron microscope (TEM) micrographs demonstrated that the tempered WM (WT) displayed coarse martensite packets with carbides along the lath and grain boundaries. Cellular subgrains and coarse carbides were observed in the WT-820T sample. A degraded lath morphology and numerous carbides in various dimensions were found in the WT-890T sample. The grain boundary map showed that the WT-820T sample had the same coarse-grained structure as the WT sample, but the WT-890T sample consisted of refined grains. The WT-890T samples with a fine-grained structure were more prone to creep fracture than the WT and WT-820T samples were. Intergranular cracking was more likely to occur at the corners of the crept samples, which suffered from high strain and stress concentration. As compared to the Gr. 91 steel or Gr. 91 WM, the Gr. 92 WM was more stable in maintaining its original microstructures under the same creep condition. Undegraded microstructures of the Gr. 92 WM strained at elevated temperatures were responsible for its higher resistance to creep failure during the practical service.

Keywords: 9CrMoW; weld metal; reheated microstructure; creep resistance

1. Introduction

Increasing the steam pressure and operating temperature of a boiler in a coal-fired power plant can reduce the CO_2 emissions and save energy. Advanced 9–12 Cr ferritic steels are often applied to manufacture boilers in ultra-supercritical fossil-fired power stations [1]. The Gr. 92 steel is applied extensively to replace T/P91 steel in steam pipes, headers, reheaters and superheaters [1]. M_{23}C_6 carbides precipitate mainly at the prior austenite grain boundaries (PAGBs). The MX precipitates composed of V, Nb, C, and N are dispersed in tempered martensite laths of normalized and tempered T92 steel [2]. Moreover, after a long period of service, the microstructures of the Gr. 92 steel deteriorate, and the mechanical properties are degraded. After creep of Gr. 92 steel, the laths and M_{23}C_6 carbides increase in size and the dislocation density decreases near the crack tip, as compared with those of the virgin material [3]. Aging P92 weld metal (WM) at 625 °C/1000 h causes a decisive reduction in the impact energy, which is ascribed to the coarse solidified microstructure, Laves phase and M_{23}C_6 carbide [4].
Complex transformation proceeds in the heat-affected zone (HAZ) of an arc weld. Creep failure in the HAZ of advanced ferritic steel welds is identified as Type IV cracking [5–8]. The cracking of the Gr. 92 steel is attributed to the lack of precipitation-strengthening at PAGBs and block boundaries [9–11]. The coarsening of M$_{23}$C$_6$ carbides at the PAGBs can be inhibited by adding B at 130 ppm to 9CrWCo steel [12]. Moreover, increasing B and decreasing N contents can suppress grain refinement after welding through a larger pinning effect imposed by B-stabilized M$_{23}$C$_6$ carbides [13]. Among the different zones of the P92 weld, the fine-grained HAZ (FGHAZ), heated just above the temperature of $A_{C3}$, shows the highest degree of creep damage [14]. As reported in international journals, the microstructural evolution of an advanced ferritic steel weld generally focuses on the creep failure of the simulated HAZ, and less attention is paid to the WM. The creep growth rate of the P92 WM is faster than that of the base metal (BM) but slower that of the FGHAZ [15]. With the same tempering at 760 $^\circ$C/2 h, re-austenization at 1050 $^\circ$C can significantly increase the Charpy energy of the Gr. 92 WM [16]. Despite the finer structure of electron and laser beam welds, as compared to gas-tungsten arc welds, the variation in hardness in the Gr. 92 WM with the three welding processes is not significant [17].

In the open literature, great concern has been paid to the degradation of creep resistance in the FGHAZ of the 9–12 Cr steel welds. It is noticed that few studies have clarified the effects of the solidified microstructure and/or welding thermal cycle on the creep resistance of Gr. 92 WM. In a multiple-pass weld, the prior deposits will be reheated by the following passes, producing a reheated microstructure. The microstructures of the reheated zone in the WM will depend on the peak temperature of the imposed thermal cycles. In this study, the reheated microstructure in a Gr. 92 WM was simulated by an infrared heating system. The goal of this study was to inspect the simulated over-tempered and partially transformed microstructures in the WM of a multiple-pass weld. Furthermore, the impacts of welding thermal cycles on the microstructures of the Gr. 92 WM were related to creep rupture life.

2. Material and Experimental Procedures

Gr. 92 tubes with 50 mm in diameter and 9.6 mm in wall thickness were applied for the butt joint in this work. The chemical composition of the tube was 0.003 B, 0.12 C, 8.89 Cr, 0.43 Mn, 0.47 Mo, 0.038 N, 0.05 Nb, 0.24 Ni, 0.014 P, 0.002 S, 0.23 Si, 0.003 Ti, 1.76 W, 0.20 V and balanced Fe in wt.%. The wire filler composition was 0.002 B, 0.09 C, 9.66 Cr, 0.65 Mn, 0.44 Mo, 0.037 N, 0.12 Nb, 0.42 Ni, 0.013 P, 0.002 S, 0.29 Si, 0.0002 Ti, 0.19 V, 1.46 W and balanced Fe. A single V groove was machined for the gas-tungsten-arc welding of the T92 tube in multiple passes. Due to the size restriction of the T92 tube, a Gr. 91 steel plate of 12.5 mm cut into a double V groove was used as the substrate for the preparation of the creep sample of all WM. The schematic dimensions of the tested sample sliced from the butt-welded groove weld are shown in Figure 1. A dilatometer preset at distinct heating and cooling rates was examined to determine the $A_{C1}$, $A_{C3}$, $M_s$ and $M_f$ temperatures of the WM for comparison with the T92 substrate [18]. The tempered WM (WT) was heated using an infrared heater to generate the reheated microstructures present in a multiple-pass weld. Reheated microstructures were produced by heating the tempered WM to 820 $^\circ$C (WT-820) or 890 $^\circ$C (WT-890) for 60 s. These temperatures were slightly below either the $A_{C1}$ or $A_{C3}$ temperatures of the WM, respectively. The simulated specimens were tempered at 760 $^\circ$C for 2 h and labeled the WT-820T or WT-890T, according to the prior heating temperature.
The sample hardness was obtained with a MVK-G1500 micro-Vickers hardness tester (Mitutoyo, Kawasaki, Japan) under 300 gf loading for 15 s. To assess the influence of the reheated microstructures on the creep resistance, simulated specimens after tempering were subjected to the short-term creep tests with a constant load at either 630 °C/120 MPa or 660 °C/80 MPa. As shown in Figure 1b,c, the tensile and impact specimens of 5 mm thickness satisfied the ASTM E8 and E23 specifications, respectively. The tensile strain rate was kept at $1.6 \times 10^{-4}$/s at room temperature. The creep tests were terminated if no cracking occurred after straining for 4000 h. The results shown in this work are the average of three samples for each test. Their microstructures were explored with the aid of a JSM-7100F scanning electron microscope (SEM, JEOL, Tokyo, Japan). The minute microstructures in the specimens were investigated using a Tecnai G2 F30 transmission electron microscope (TEM, FEI, Hillsboro, OR, USA). A NordlysMax2 electron backscatter diffraction (EBSD, Oxford Instruments, Abingdon, UK) detector was used to examine the grain size, high-angle and subgrain boundaries of the sample.

3. Results
3.1. Determination of Transformation Temperatures

Table 1 lists the measured $A_{c1}$, $A_{c3}$, $M_s$ and $M_f$ temperatures of the Gr. 92 WM at selected heating/cooling rates and compared them with those of the T92 substrate [18]. Transformation temperatures of $A_{c1}$ and $A_{c3}$ increased slightly with an increase in the heating rates. For the heating/cooling rates of 15 °C/s, the $A_{c1}$ and $A_{c3}$ temperatures of the Gr. 92 WM were 862 °C and 898 °C, relative to 886 °C and 934 °C for the T92 substrate. The $A_{c1}$ and $A_{c3}$ temperatures of the Gr. 92 WM were 20–30 °C lower than those of the T92 substrate. This implied that the phase transformation of Gr. 92 WM would occur at a lower temperature range than that of the T92 substrate if multiple welding passes were performed. In contrast, the $M_s$ temperature was around 380 °C and the $M_f$ temperature dropped to about 250 °C at the high cooling rate (45 °C/s). Thus, martensite would be formed completely after the solidified WM cooled to room temperature.
Table 1. Transformation temperatures of Gr. 92 WM determined by a dilatometer at specific heating/cooling rates.

| Temperature (°C) | Rates (°C/s) | Heating Rate (°C/s) | Cooling Rate (°C/s) |
|-----------------|--------------|---------------------|---------------------|
|                 | 5            | 15                  | 30                  | 45                  |
| \(A_C1\)        | 851          | 862                 | 869                 | 894                 |
| \(A_C3\)        | 880          | 898                 | 904                 | 933                 |
| \(M_s\)         | -            | -                   | -                   | -                   |
| \(M_f\)         | -            | -                   | -                   | 388                 |
|                 |              | 385                 | 384                 | 377                 |

3.2. Microhardness Distribution in the Cross Welds

The cross-sectional appearance of a T92 weld and microhardness curves in the as-welded (AW) and tempered conditions are shown in Figure 2. The WM microhardness fell to a range of HV 380 to 420 in the AW condition. The maximum microhardness above HV 450 appeared in the site near to the fusion boundary (FB). In the AW condition, the HAZ showed high hardenings at a depth within 3 mm from the FB, and then a steep drop in microhardness further away from the FB. After tempering at 720 °C/2 h, the WM hardnness decreased to below HV 350 and the HAZ microhardness dropped obviously, as compared with that of the AW condition. With an increase in tempering temperature to 760 °C, continuous decreases in WM and HAZ microhardness were observed, but the softening effect was weaker. The results reveal that the BM microhardness was less affected, regardless of the tempering conditions. It was noticed that a narrow soft zone developed ahead of the BM and it was more obvious in the weld tempered at 760 °C. Overall, after tempering at 760 °C, the WM microhardness was a little higher than those of the BM and HAZ.

![Figure 2](image_url). (a) Macroscopic appearance of a T92 weld in cross-sectional view cut from the butt-welded tube, (b) microhardness curves of the welds determined from the centerline of the WM across the heat-affected zone (HAZ) to the base metal (BM) in the as-welded and tempered conditions.

3.3. Mechanical Properties of a Cross Weld

Table 2 lists tensile properties of the Gr. 92 cross weld and the impact toughness of the WM in the AW or tempered conditions at 25 °C. For the T92 BM cut from the steel tube, the ultimate tensile strength (UTS) and yield strength (YS) were about 711 and 554 MPa, respectively. The elongation of the T92 BM was as high as 29%. In the AW condition, the UTS of the cross weld (707 MPa) was about the same as that of the BM, but the YS of the weld (526 MPa) was about 30 MPa lower than that of the BM. Moreover, the elongation of the cross weld (17%) in the AW condition was obviously lower than that of the BM, which could be attributed to the minimal deformation of the hardened HAZ during straining. Moreover, the fracture location of the AW condition was at the
BM. In the cross weld tempered at 740 °C, the UTS and YS of the weld were 667 and 500 MPa, respectively, together with 21% elongation. Increasing the tempering temperature to 760 °C, the UTS and YS of the weld decreased to 659 and 472 MPa, respectively, and the elongation improved to 24%. After tempering, the fracture locations of all the cross welds were at the HAZ near to the BM. The results indicate that increasing the tempering temperature would reduce the tensile strength but improve the ductility of the cross weld, as compared to the AW condition. A sub-size Charpy V-notch specimen of 5 mm thickness was used to determine the impact energy of the WM. The impact energy of T92 BM was about 80 J, which was obviously higher than that of the WM (12 J) in the AW condition. After tempering at 740 °C, the impact energy of the tempered WM was about 54 J, which was slightly lower than that of the BM. The impact energy of the WM raised to 80 J after tempering at 760 °C. The results indicate that the WM had very high brittleness in the AW condition, whereas the WM tempered at 760 °C exhibited the same resistance to impact fractures as the BM.

Table 2. Typical mechanical properties of the Gr. 92 cross welds in the as-welded and post-weld tempered conditions.

| Specimen | Yield Strength (MPa) | Ultimate Tensile Strength (MPa) | Elongation (%) | CVN Impact Toughness (J) |
|----------|----------------------|---------------------------------|----------------|--------------------------|
| AW ¹     | 506                  | 707                             | 17             | 12                       |
| PT 1 ²   | 500                  | 667                             | 21             | 54                       |
| PT 2 ³   | 471                  | 659                             | 24             | 80                       |
| BM ⁴     | 554                  | 711                             | 29             | 80                       |

¹ AW: as-welded. ² PT 1: post-weld tempering at 740 °C/2 h. ³ PT 2: post-weld tempering at 760 °C/2 h. ⁴ BM: base metal.

3.4. Microhardness of the Simulated Sample

The micro-Vickers hardnesses of the WM specimens after different thermal treatments were determined. The peak WM microhardness in the AW condition (W sample) was about HV 450 and showed a certain degree of fluctuations in the microhardness of WM owing to the imposed thermal cycles of subsequent passes. After tempering at 760 °C/2 h, the microhardness of the WT sample was reduced to HV 266. The WT specimens reheated to 820 °C (WT-820) demonstrated little change in microhardness (HV 263) as compared with that of the WT sample. It was indicated that the short-term over-tempering at 820 °C induced less variation in the microstructures. Subjecting the WT-820 sample to tempering treatment, the specimen’s microhardness decreased slightly to HV 254 (WT-820T sample). In the case of the WT sample reheated to 890 °C (WT-890 sample), the sample microhardness rose to HV 354, which was the result of incomplete hardening. Tempering the WT-890 sample was found to lower its microhardness to about HV 259, i.e., the WT-890T sample. The results reveal that the WT, WT-820T and WT-890T samples were all similar in microhardnesses but might have different microstructures.

3.5. Microstructural Observations and IPF Identifications

SEM micrographs and the inverse pole figure (IPF) maps, illustrating individual grain orientations of the WT, WT-820T and WT-890T samples in different colors, are displayed in Figure 3. Different colors in the IPF map represent the different orientations of the grains. The martensite packets of the WT sample were very coarse and mostly aligned in the same direction (Figure 3a,b). Coarser martensite packets separated by grain boundaries were seen in the WT-820T sample (Figure 3c,d). Those observations indicate that the short-term over-heating had little effect on the Gr. 92 WM microstructure. However, dislocation recovery, break-down of the lath structure, polygonization and carbide coarsening can occur during over-tempering of Gr. 92 steel [18,19]. Moreover, significant changes in granular morphology occurred in the WT-890T sample (Figure 3e,f). The SEM micrograph (Figure 3e) revealed that the WT-890T sample seemed to be composed of many fine grains.
The IPF map showed that the coarse martensite packets in the WT and WT-820T samples were replaced by short and fine ones in the WT-890T sample (Figure 3f). The fine-grained zone in the WM could be found after reheating to below the A_C3 temperature in a multiple pass weld.

**Figure 3.** (a,b) WT, (c,d) WT-820T, (e,f) WT-890T. (a,c,e) Scanning electron micrographs; (b,d,f) the corresponding inverse pole figure (IPF) maps.

TEM micrographs of the Gr. 92 WM tempered under different conditions are shown in Figure 4. In the AW condition, the WM consisted of lath martensite with a high dislocation density and few precipitates in the sample (Figure 4a). After tempering at 760 °C, the dislocation density of the WT sample decreased, and the lath martensite boundaries and grain boundaries were decorated by precipitates (Figure 4b). The main precipitates present in the WT specimens were M_{23}C_6 carbides, as confirmed by the diffraction pattern. It has been reported that some fine precipitates, possibly carbides or carbonitrides (NbC, VC, and (NbV)(CN)), could be dispersed in the tempered martensite matrix [2]. For the WT-820T sample, the TEM micrograph showed a low dislocation density and fine cell subgrains instead of parallel lath martensite (Figure 4c), as compared to the WT sample. Furthermore, some carbides in the WT-820T sample aggregated to a very coarse size of about 400 nm and above. As shown in Figure 4d, the degraded lath morphology and the precipitation of numerous carbides of different sizes were seen in the WT-890T sample. The carbides in the WT-890T sample were greater in amount than those in the WT and WT-820T samples, but the exact reason for this is not known at this time.
3.6. Grain Boundary Characteristics

Figure 5 presents grain boundary maps showing the grain boundary characteristics of distinct samples. The PAGBs are high-angle grain boundaries (HAGBs), whereas the lath boundaries and block/packet boundaries are low-angle grain boundaries (LAGBs). As shown in Figure 5, the LAGBs are indicated by red (1°–5°) and green (5°–15°) lines, whereas the HAGBs (15°–62.5°) are indicated by black. The IPF map (Figure 3b) demonstrates that the WT sample consisted of coarse martensite packets. The grain boundary map of the WT sample shows that the coarse martensite packets were outlined by HAGBs (Figure 5a). Furthermore, the appearance of the HAGBs within a coarse grain (Figure 5a) was attributed to different orientations of the martensite packets, which was supported by the IPF map. The grain boundary feature (Figure 5b) of the WT-820T sample was similar to that of the WT one; both had coarse granular structures and coarse martensite packets. Moreover, few fine grains were present at the HAGBs of the WT-820T sample (Figure 5b). It was found that the HAGBs of the WT-890T sample occupied a large portion of the map (Figure 5c). The coarse-grained structure of the WT sample was replaced by numerous irregular fine grains in the WT-890T sample (Figure 5c). This revealed that the WM reheated to the temperature slightly below the A_C3 temperature had a fine-grained structure. Those fine-grained structures were anticipated to deteriorate Gr. 92 WM creep resistance.
Figure 5. Grain boundary maps of the (a) WT, (b) WT-820T and (c) WT-890T samples.

3.7. Short-Term Creep Tests

Short-term creep tests were performed under the conditions either at 630 °C/120 MPa or 660 °C/80 MPa. Specimen elongation was also used to rank the creep resistance of the specimens. The WT samples did not fracture during the test period, despite the testing conditions. The WT-820T samples were resistant to rupture during the creep tests, but one of the samples fractured at 3472 h under the 630 °C/120 MPa conditions (Figure 6a). Moreover, the creep life of the WT-890T sample was much shorter than those of other samples under the 630 °C/120 MPa condition. As revealed in prior works [18–21], simulated thermal treatment will shorten the creep life of the sample, as compared to that of the original substrate, particularly in the case of the partially transformed samples. The results demonstrate that the reheated WM, which had been heated to the two-phase region, had the lowest creep resistance among the distinct zones of the WM. Under the 660 °C/80 MPa conditions, none of the samples fractured within the testing period, as shown in Figure 6b.

3.8. Fracture Features

The fracture features of the tensile and impact samples in the AW or tempered conditions were inspected, as shown in Figure 7. Before the tensile test, the sample was polished and slightly etched to ensure that the fracture location could be distinguished more accurately after tensile straining. The fracture of tensile specimens was located at the outer edge of the HAZ, regardless of the tempered conditions. In addition, the WM was resistant to deformation during straining (Figure 7). The slightly necked zone, symmetrically located with respect to the fracture site, was used to identify the weak zone in a cross weld. Tensile fracture of the weld in the AW condition was located at the BM (Figure 7a). The tempered weld was prone to rupture in the over-tempered zone just ahead of the BM (Figure 7c,e). The high thickness reduction in the tensile fractured zone was associated with extensive ductile dimple fracture (not shown here). The macro-fractured appearance of impact welds is shown in Figure 7b,d,f. The AW samples, as expected due to their low impact toughness, exhibited a flat fracture surface (Figure 7b). By contrast, the tempered welds showed a high distortion and change in sample profile (Figure 7d,f), which was related with the high resistance to impact rupture. The wide extent of the cleavage fracture was seen in
the impact-fractured sample in the AW conditions, whereas ductile dimples were seen extensively in the tempered samples.

Figure 6. Short-term creep tests of the WT, WT-820T, WT-890T samples at (a) 630 °C/120 MPa, (b) 660 °C/80 MPa.

Figure 7. Appearance of the fracture zone in the welds in the (a,b) AW, (c,d) PT 1 and (e,f) PT 2 conditions. (a,c,e) the side view of the tensile fractured samples; (b,d,f) impact fractured samples.
The typical fracture features of crept samples were examined with an SEM, as shown in Figure 8. The crept fracture appearance of the WT-890T sample macroscopically showed obvious plastic deformation before rupture (Figure 8a) and microscopically mainly displayed fine dimples under the 630 °C/120 MPa conditions (Figure 8b). The occurrence of intergranular fracturing (as indicated by the arrow) at the specimen corner of the WT-890T sample (creep life: 3472 h) was seen. By contrast, the fracture features of the WT-820T samples (creep loading) showed mixed mode fracturing under the 630 °C/120 MPa conditions (Figure 8c). Intergranular fracturing (indicated by the arrows in Figure 8d) was observed predominantly at the corners of the WT-820T specimen after creep-straining. Moreover, the fast fracture (FF) zone displayed predominantly ductile dimple fracture in the tested sample. With increasing the creep life of the tested sample, intergranular fracturing was more likely to be observed.

![Figure 8. Fracture features of the crept samples: (a, b) WT-890T; (c, d) WT-820T at 630 °C/120 MPa.](image)

Although not all of the tested samples fractured during the creep tests, creep defects could be also induced in the samples. The investigated specimen was cut from the gage section of the sample crept under the 660 °C/80 MPa conditions and subjected to metallographic preparation before inspection. The trace of carbides exhibited the profile of PAGBs. Figure 9 presents the microstructures of the samples before and after the creep loading. Before the creep-straining (Figure 9a,c,e), the precipitates in the original samples were very fine when inspected by SEM, especially in the WT sample. As compared with the initial state, a decrease in carbide density and an increase in carbide size occurred after creep-straining. Coarse precipitates in the WT-820T and WT-890T samples were observed after creep-loading, which implied the degraded creep strength (Figure 9d,f). The high density of carbides decorating the grain boundaries of the WT-890T sample revealed its fine-grained structure (Figure 9e,f). Under the creep-straining, microcracks were prone to be induced at the corners of the tested sample (Figure 9b,d,f), which propagated mainly along the grain boundaries. According to the carbide distribution, the coarse-grained structure was maintained in the WT and WT-820T samples (Figure 9b,d), but a fine-grained structure was found in the WT-890T sample (Figure 9f) after the creep tests.
Figure 9. Changes in the microstructures of the samples before and after the creep-straining at 660 °C/80 MPa: (a,b) WT, (c,d) WT-820T, (e,f) WT-890T samples.

4. Discussion

As mentioned previously, the $A_{C1}$ and $A_{C3}$ temperatures of the Gr. 92 WM were about 20–30 °C lower than those of the Gr. 92 tube. At the heating/cooling rates of 15 °C/s, the $A_{C1}$ and $A_{C3}$ temperatures of the Gr. 92 WM were 862 °C and 898 °C, respectively. This indicated that over-tempering or fully austenitic transformation of the Gr. 92 WM would occur at lower temperatures, relative to the T92 substrate, if welding thermal cycles were applied. The microhardness distribution showed that the HAZ ahead of the BM had the lowest hardness among the distinct regions of a T92 cross weld, which could be attributed to the over-tempering induced by welding thermal cycles. The WM in the AW condition had very low impact toughness (12 J) relative to that of the steel substrate (80 J). After tempering at 740 °C, the impact energy of the tempered WM (54 J) was slightly lower than that of the BM. The impact energy of the WM could be raised to that of the substrate after post-weld tempering at 760 °C/2 h. With an increase in the post-weld tempering temperature, a gradual reduction in the tensile strength but an increase in ductility of the cross weld occurred. After the post-welding tempering, the tensile fracture of a cross weld occurred in the over-tempering zone, which was slightly ahead of the BM. The over-tempering zone in the T92 cross weld accounted for the inferior tensile properties of the weld, as compared to the T92 substrate.
The grain boundary and IPF maps of the WT and WT-820T samples revealed that coarse martensite packets were inter-dispersed in a coarse-grained structure and outlined by HAGBs. By contrast, the WT-890T sample comprised numerous irregular fine grains. The fine-grained structure of the WT-890T sample was proven to degrade its creep resistance. The creep life of the WT-890T sample was found to be much shorter than those of other samples under the 630 °C/120 MPa conditions. The relatively high elongation of the WT-890T sample also confirmed its inherent low creep strength at elevated temperature. The crept fracture appearance of the WT-890T sample revealed obvious plastic deformation macroscopically and mainly fine dimples microscopically under the 630 °C/120 MPa conditions. Moreover, intergranular cracks were found predominantly at the corners of the WT-820T sample after creep-straining for 3472 h.

Although none of the samples fractured within the testing period under the 660 °C/80 MPa conditions, creep defects still could be found in the tested samples. Creep voids were occasionally seen at the interior grain boundaries of the creep samples. The results indicate that the intergranular fracture was more likely to be seen in the crept samples with an increase in creep life (Figure 8d). It was deduced that the high surface strain combined with the high stress concentration at the specimen corners led to intergranular separation. It was deduced that the linking of surface creep voids into surface microcracks caused the creep fracture in the tested samples. Softening during the creep could result from two factors: the coalescence of subgrains and dislocation recovery, or coarsening of fine M23C6 carbides that hindered the subgrains and boundary motion. In prior work [19,21,22], the IPF maps and grain boundary maps around the fracture zones of creep-ruptured Gr. 91 and Gr. 92 samples exhibited a fine-grained structure and texture, which was related to dynamic recrystallization under straining at elevated temperature. As compared to the Gr. 91 steel or Gr. 91 WM, the Gr. 92 WM was more stable in maintaining the initial coarse microstructures under the same creep conditions, as shown in Figure 9. Undegraded microstructures of the Gr. 92 WM during straining at elevated temperature were responsible for its higher resistance to creep failure during the practical service.

5. Conclusions

An infrared heating system was used to simulate the reheated microstructures of Gr. 92 WM. The short-term creep life of the simulated samples was determined, and the results were related with the inherent microstructures of the samples.

(1) Despite the tempering condition, the HAZ ahead of the BM had the lowest hardness among the distinct regions of a T92 cross weld. The AW WM was very brittle and showed low impact toughness relative to that of the substrate. After tempering at 740 °C, the impact energy of the tempered WM was slightly lower than that of the BM. The impact energy of the WM could be increased to that of the substrate after tempering at 760 °C/2 h. When increasing the post-weld tempering temperature, a gradual reduction in the tensile strength, but an increase in the ductility of the cross weld, occurred. The over-tempering zone in a T92 cross weld accounted for the inferior tensile properties of the weld, as compared to the T92 substrate.

(2) The WT and WT-820T samples displayed coarse martensite packets inter-dispersed in a coarse-grained structure. By contrast, the WT-890T sample comprised numerous irregular fine grains. The creep life of the WT-890T sample was much shorter than those of the other samples under the 630 °C/120 MPa condition. Intergranular fracture was more likely to be found in the crept samples with increases in creep life. The high surface strain combined with the high stress concentration in the specimen corners assisted intergranular separation therein. By contrast, ductile dimple fracture was associated with the fast rupture zone in the sample. It was concluded that the fine-grained structure of the WT-890T sample played a crucial role in deteriorating its creep resistance. The formation of refined grains in the WM of the multiple-pass weld was not avoidable, but could be mitigated by lowering the heat input during welding.
(3) As compared to the Gr. 91 steel substrate or Gr. 91 WM, the Gr. 92 WM was more stable in maintaining the original microstructures under the same creep conditions. Undegraded microstructures of the Gr. 92 WM during straining at elevated temperatures were responsible for its higher resistance to creep failure during the practical service.

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Nomenclature

| Abbreviation | Description                  |
|--------------|------------------------------|
| WM           | weld metal                   |
| PT           | partially transformed        |
| OT           | over-tempered                |
| TEM          | transmission electron microscope |
| WT           | tempered weld metal          |
| PAGBs        | prior austenite grain boundaries |
| HAZ          | heat-affected zone           |
| FGHAZ        | fine-grained heat-affected zone |
| BM           | base metal                   |
| SEM          | scanning electron microscope |
| EBSD         | electron backscatter diffraction |
| AW           | as-welded                    |
| FB           | fusion boundary              |
| PT           | post-weld tempering          |
| UTS          | ultimate tensile strength    |
| YS           | yield strength               |
| IPF          | inverse pole figure          |
| HAGBs        | high-angle grain boundaries  |
| LAGBs        | low-angle grain boundaries   |
| FF           | fast fracture                |

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