Creep behavior and in-depth microstructural characterization of dissimilar joints

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Abstract
The 700 °C power plants currently under development will utilize Ni-base alloys such as alloy 617 for components to be operated at temperatures >650 °C. Due to economic reasons for components or parts of components which are subjected to temperatures <650 °C, 2% Cr or 9–12% Cr steels is used, depending on the required mechanical properties. This makes the dissimilar joining of Ni-base alloys and Cr steels a necessity in these plants. Experimental investigations show that these joints have to be identified as weak points with regard to damage development under creep and creep-fatigue loading. The present investigation focuses on welds between the alloy 617 and 2% Cr steel. Under creep load the fracture occurs near the fusion line between the 2% Cr steel base metal and alloy 617 weld metal. To explain the reasons for this fracture location, the microstructure of this fusion line was investigated using TEM and FIB techniques after welding and after creep loading. The TEM investigations have shown a small zone in the weld metal near the fusion line exhibiting chromium depletion and clearly reduced amounts of chromium carbides, leading to a weakening of this zone.

Keywords: dissimilar welds, high temperature materials, transmission electron microscopy

1. Introduction

In the development of modern steam power plants, it is the aim to increase the steam parameters from the current optimum of about 620 °C at 28 MPa up to 700 °C at 35 MPa in order to improve the rate of efficiency and thereby reduce CO₂ emission [1]. According to current knowledge, these steam conditions cannot be reached with ferritic–martensitic alloys, which are currently the state of the art material for steam power plants [2]. Nickel-based alloys are used for the higher temperature regimes. However, due to economic reasons (Ni comes at about ten times the financial cost than high temperature steels), the Cr alloyed steels are still used wherever possible (9–12% Cr steels for up to 620 °C, 2% Cr steels for up to 550 °C). Therefore the joining of these steels with the nickel-based alloys, and mainly with alloy 617 for the 700 °C power plant, will be a crucial point in the material design.

Even similar welds introduce relatively weak regimes in the structural material. During welding, different microstructural zones form, with different mechanical properties. The weld metal itself was of course solidified from the melt, commonly exhibiting a dendritic microstructure. Adjacent to the weld metal is the so-called heat-affected zone I (HAZ I), which in the case of steels consists of coarse prior austenite grains and has a high hardness along with a low toughness. The next region is the HAZ II with small prior austenite grains (in this case, the stabilizing carbides were not dissolved during welding), and finally the HAZ III or intercritical zone, where only a partial phase transformation between ferrite and austenite occurs during welding. This zone typically exhibits the lowest hardness values and creep strength, which leads to a strain localization in this area.
The different microstructure zones also lead to a multiaxiality of the stress state [3]. The combination of these effects can lead to reduced creep strength by an amount up to 20% [4].

For dissimilar welds the situation near the weld becomes significantly more complicated. One problem is often that the two joined materials require different heat treating conditions for optimum performance. For example many Ni base alloys such as alloy 263 or alloy 740 require a heat treatment to precipitate strengthening gamma prime particles at 800 °C, but usually steels should be tempered at 730–780 °C. In addition to the different microstructures of the different materials, one has to consider chemical differences and directional process between the different zones. When two materials with a different chemical composition are joined, there exists a driving force for diffusion from the side rich in one element to the other side of the joint. This diffusion can lead to a local change in the chemical composition along with altered properties, but also to the formation of intermetallic phases or precipitates which would normally not be present in the materials. For small atoms such as carbon, uphill diffusion also has to be considered [5]. A further problem of joining different materials is their difference in mechanical properties such as strength or toughness. This leads to discontinuities of stress and strain states, and can yield localized stress–strain parameters which are quite different from the conditions on the component. All these points make dissimilar joints potential weak links in component design.

Cracking at or close to dissimilar welds is a persistent problem in joining technology [6, 7]. The fracture of these joints can often be related to the formation of carbides near the weld [8]. The present paper focuses on the mechanical testing, microstructural characterization and simulation of a dissimilar weld between a 2% Cr steel (in wt%: 0.23C, 2Cr, 0.8Mo, 0.7Ni, 0.6W, 0.3V; Fe balance) and alloy 617, a solution strengthened Ni-based alloy (in wt%: 22Cr, 11.5Co, 0.8Mo, 0.7Ni, 0.6W, 0.3V; Fe balance) and alloy 617, a Nickel balance). The dissimilar welds were produced by using tungsten inert gas welding. Alloy 617 was used as the weld material. The welds were heat treated at 670 °C for 10 h, which is the equivalent of a typical heat treatment in rotor components. A schematic of the creep test samples is given in figure 2. The weld metal was located at the center of the crossweld creep specimen.

Samples were tested at different load levels at 500 and 550 °C with testing times of up to 10,000 h. For the data acquisition for the simulation, gleeble specimens for simulated heat affected zones I–III were also tested, as well as the base material. The strain was measured discontinuously.

2. Experimental details

2.1. Sample material and creep testing

The dissimilar welds were produced by using tungsten inert gas welding. Alloy 617 was used as the weld material. The welds were heat treated at 670 °C for 10 h, which is the equivalent of a typical heat treatment in rotor components. A schematic of the creep test samples is given in figure 2. The weld metal was located at the center of the crossweld creep specimen.

The fracture surfaces after creep testing were investigated using a JEOL JSM 6400. The sample preparation for the TEM analysis by focused ion beam technique was done in a Zeiss Auriga Crossbeam Workstation using the ‘lift-out’ technique [9]. The same microscope was also utilized for EDX mapping and scanning TEM (STEM).

Conventional TEM investigations were performed using a JEOL JEM 2000 FX operated at 200 kV to determine microstructure and especially the precipitation state near the fusion line. For analytical measurements the TEM is equipped with an EDX detector (Kevex Sigma I from NORAN, capable of detecting elements with an atomic number ⩾11) and with a Gatan energy filter. The latter allows elemental maps or electron energy loss spectra to be acquired. A detailed description of EFTEM measurements can be found, for example, in [10, 11]. EDX line scans over the weld were recorded in the TEM with a beam diameter of about 10 nm. Microprobe linescans were performed on a JEOL JXA 8600 to get the distribution of C, Cr, Fe and Ni on a larger length scale.
2.3. Simulation

For the numerical simulations, the weld was divided into five zones: alloy 617, HAZ I-III, and 2% Cr steel base metal. To simulate all three creep states, a modified Graham–Walles creep law has been applied [12]. For the tertiary creep behavior, an effective stress concept with a damage variable \( D \) was used. The formulation of the modified creep law is given below:

\[
\frac{dD}{dt} = 10^{A_{D1}} \sigma^{n_{D1}} \varepsilon^{m_{D1}} + 10^{A_{D2}} \sigma^{n_{D2}} \varepsilon^{m_{D2}},
\]

\[
\frac{d\varepsilon}{dt} = 10^{A_1} \left[ \frac{\sigma}{1-D} \right]^{n_1} \varepsilon^{m_1} + 10^{A_2} \left[ \frac{\sigma}{1-D} \right]^{n_2} \varepsilon^{m_2}.
\]

The exponents \( A_1, n_1, m_1, A_2, n_2, m_2, A_{D1}, n_{D1}, m_{D1}, A_{D2}, n_{D2} \) and \( m_{D2} \) are material-dependent parameters which are fitted phenomenologically to uniaxial creep data separately for each of the five zones.
estimate of the 2% Cr steel minus 20%. This was found at both testing temperatures. For the experiments at 500 °C the results of the simulated heat-affected zones are also given. The samples taken from simulated material according to HAZ I (peak temperature 1300 °C) and HAZ II (peak temperature 1050 °C) did exhibit a higher creep strength than the base material, whereas the HAZ III (peak temperature 840 °C) did show worse creep behavior than the weld samples. Most of the specimens show multiple damage in, e.g., the fusion line and HAZ 3, designated as FL+HAZ3. Therefore the fracture locations of the welded samples are also given (FL = fusion line; BM = base material). Figure 3 shows the longest available testing times at the date of this publication. A further drop in creep strength at testing times higher than 10 000 h is possible, but no indications of that behavior has been observed so far.

3.2. Microstructural characterization

Almost all welded test samples did exhibit fracture at or very close to the fusion line. An example of a fractured creep sample is given in figure 4. The sample shows a circumferential crack starting at the fusion line of the weldment.

To clarify the reason for the failure location, the fusion line was characterized in the initial state and after creep testing. Figure 5 shows a STEM image of a TEM lamella. The formation of carbides near the fusion line within a distance of about 500 nm can be observed (dark particles in the steel area on the left-hand side of the image). Figure 6 shows the same sample in the TEM along with a superimposed EFTEM image. In the elemental map the chromium-rich particles near the fusion line (indicated in red) can be clearly identified.

Figure 7 also shows this sample, this time together with an EDX line scan of the matrix material over the fusion line. A further drop in creep strength at testing times higher than 10 000 h is possible, but no indications of that behavior has been observed so far.

Figure 8. OM image of a fractured sample after 8800 h testing time at 500 °C and 215 MPa. Most of the fracture occurred close due to fusion line.
Figure 9. TEM image of a fusion line after creep testing with corresponding EDX line scan. The findings correspond to the initial state, although the area of reduced chromium content in the weld is broader.

Figure 10. SEM image and EDX mapping of the chromium content (given in wt%) of a fusion line after creep testing. A zone of about 2 μm width with a reduced chromium content could be detected.

The chromium content decreased to about half the amount and only a very low density of 15 particles μm⁻² was observed (particle size was 23 nm). The chromium carbides on both sides of the fusion line were identified as M₂₃C₆ carbides. The microprobe measurements did show a slight increase in C content on the Ni-side of the weldment.

In figure 8, an optical microscopy image of a fractured creep sample is shown. The sample has been tested at 500 °C and 215 MPa, and did fracture after 8800 h. The crack most likely initiated at the fusion line and then grew in close proximity to the fusion line on the steel side. After the load became too high for the residual area, the crack followed the maximum shear stress with an angle of 45° to the loading direction and the last part of the fracture occurred parallel to the load in the HAZ. From the area of the fractured sample where both materials were still present at the fusion line, a TEM lamella was prepared using the focused ion beam lift out technique. Figure 9 shows a TEM image of the fusion line along with corresponding EDX data. The principal findings correspond to the initial state of the weld, with a decrease in chromium content in the weld metal close to the fusion line associated with the absence of fine chromium carbides, and the formation of coarser chromium carbides in the adjacent steel close to the fusion line. The area of chromium depletion was in this case about 2 μm wide. As this was basically the whole width of the TEM lamella, these findings were also checked by EDX mapping in the SEM (figure 10). These measurements could confirm the width of the chromium
depletion zone to 2 \mu m. The region of the coarse carbides in the steel was also broader than in the initial state, and was confined to a distance of about 1 \mu m from the fusion line.

### 3.3. Simulation

Figure 11 gives the results of the finite element modelling of the sample shown in figures 7–9 with respect to the strain of the sample. The mapping of the strain distribution indicates strain localization in the HAZ III near the specimen surface. The line segments give the evolution of the strain formation after 5000 h and at the time of fracture. The alloy 617 weld metal exhibited almost no strain during the test.

In figure 12 the evolution of the von Mises stress is given for this sample. Due to the higher strain in the HAZ III the stresses can be accommodated in this area. The highest stresses are found in the HAZ I near the fusion line, as the rigid nickel-based weld metal hinders deformation of the steel.

Due to the relatively high hardness values of the steel in the HAZ I, the material cannot reduce the stresses effectively with plastic deformation.

Compared to the experiments (figure 7) the critical locations HAZ I near the fusion line and HAZ III could be identified in the simulations. However, the exact damage mechanism consisting of multiple damage in the fusion line and HAZ III cannot be modeled, since the chronological sequence in both regions is unknown, i.e. whether fracture occurs in the fusion line first and affects damage in HAZ III, or vice versa.

### 4. Conclusions

The creep tests on the dissimilar welds between a 2% Cr steel and alloy 617 have shown weaker creep strength than expected (significantly below 80% of the creep strength of...
the base material). The fracture location of these samples was found to be at or close to the fusion line, and not only in the HAZ III of the Cr steel, which would be the volume with the lowest creep strength. Detailed microstructural analysis of the initial state and a creep tested sample by means of SEM and TEM did reveal a zone in the weld metal close to the fusion line with severely decreased particle density of fine chromium carbides connected to a significantly reduced amount of chromium in the same zone. This leads to a decrease in creep strength of the nickel-based weld metal. On the 2% Cr steel side of the fusion line, a zone with a high density of comparably large chromium carbides was observed. In the initial state, this zone was about 500 nm wide; in the sample after creep testing the zone was found to have a width of 1 µm. It is possible that the zone broadened during the creep test due to the diffusion of chromium from the weld metal to the steel, leading to the depletion of chromium in the weld metal and the nucleation and growth of chromium carbides in the steel. However the influence of the welding procedure on the width of these zones cannot be ruled out, as the position with respect to the weld passes might be slightly different.

The FEM simulations have shown that the region in the 2% Cr steel with the carbide formation coincides with the volume which has to endure the highest stress levels. As the fracture toughness of this zone is reduced due to the high amount of chromium carbides, the fracture location, as well as the comparably short lifetime of the creep samples, is understandable.

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References

[1] Tschauffon H 2005 COMTES700—on the way to 700 °C power plant Int. Workshop 'Performance and Requirements of Structural Materials for Modern High Efficient Power Plants' Paper B5, Darmstadt
[2] Helmirich A, Chen Q, Stamatelopoulos G and Scarlin B 2006 Materials development for advanced steam boilers Proc. 8th Liège Conf. on Materials for Advanced Power Engineering 2006 Vol II ed J Leconte-Beckers et al p 931
[3] Bauer M, Klenk A, Maile M, Roos E and Jochum C 2006 Investigations on optimization of weld creep performance in martensitic steels Proc. 8th Liège Conf. on Materials for Advanced Power Engineering 2006 Vol III ed J Leconte-Beckers et al p 1341
[4] Schmidt K, Klenk A and Roos E 2012 Qualifying materials for the 700/720 °C power plant—results from MARCKO 700 VGB POWERTECH Int. J. Electr. Heat Gener. 1/2 74
[5] Darken L S 1949 Diffusion of carbon in austenite with a discontinuity in composition Trans. AIME 180 430
[6] Nelson T W, Lippold J C and Mills M J 1999 Nature and evolution of the fusion boundary in ferritic–austenitic dissimilar weld metals: part 1. Nucleation and growth Weld. J. 78 329-s
[7] Nelson T W, Lippold J C and Mills M J 1999 Nature and evolution of the fusion boundary in ferritic–austenitic dissimilar weld metals: Part 2 On-cooling transformations Weld. J. 79 267-s
[8] Nicholson R D 1986 Creep rupture properties of nickel–base transition joints after long-term service Mater. Sci. Technol. 2 686
[9] Volkert C A, Heiland B and Kauffmann F 1999 Preparation of hard-to-make TEM samples using the FIB microscope Prakt. Metallogr. 40 193
[10] Hofer F, Warbichler P and Grogger W 1995 Ultramicroscopy 59 15
[11] Hofer F, Grogger W, Kohleitner G and Warbichler P 1997 Ultramicroscopy 67 83
[12] Graham A and Walles K F A 1955 Relationships between long and short term creep and tensile properties of a commercial alloy J. Iron Steel Inst. 179 104