Effects of boron addition on the microstructure and creep properties of a Ni-Fe-based superalloy weld metal alloy

Wencai Xie1,*, Dong Wu1,∗ and Shanping Lu1,∗

1 Shenyang National Laboratory for Materials Science, Institute of Metal Research, Chinese Academy of Sciences, Shenyang, 110016, People’s Republic of China
1 School of Materials Science and Engineering, University of Science and Technology of China, Shenyang, 110016, People’s Republic of China
∗ Authors to whom any correspondence should be addressed.

E-mail: shplu@imr.ac.cn and wudong@imr.ac.cn

Abstract
The role of boron in the creep properties and grain boundary characteristics in a new Ni-Fe-based weld metal suitable for advanced ultra-supercritical (A-USC) coal-fired power plant applications has been investigated. Ni-Fe-based filler wires without boron and boron-doped (50 ppm wt% boron) were prepared for this study. Boron-doped weld metals exhibited longer rupture lives and lower steady creep rates during the creep rupture tests at 750 °C / 380 MPa and 750 °C / 210 MPa. This study explains the improvement mechanism of boron on creep resistance from the perspective of the effect of boron on M23C6. Boron increased the nucleation rate of M23C6 and participated in the formation of M23(C, B)6 type boron-carbides. After creep deformation, boron still existed stably in the M23(C, B)6. The higher density of discrete M23C6 particles due to boron addition could restrain grain boundary crack propagation and grain boundary sliding, and thereby improve the creep fracture resistance of the GH984G weld metal at 750 °C / (380 / 210 MPa).

1. Introduction

Coal power stations are currently the dominant plants for the generation of energy around the world [1]. It has been shown that increased steam temperature and pressure can improve the efficiency and reduce carbon emissions of coal-fired power stations [2, 3]. Therefore, it is hoped that advanced ultra-supercritical (A-USC) coal-fired power plants could operate at service temperatures above 700 °C and steam pressures higher than 35 MPa [4]. In this case, the superheater and reheater materials must possess excellent creep resistance and high-temperature oxidation resistance [5]. In China, a Ni-Fe-based alloy, named GH984G, has been designed for the superheater and reheater of the 700 °C (A-USC) power plants because of its excellent oxidation and creep resistance, outstanding workability, and low cost [6–8]. Because the main assembly process of superheater and reheater is fusion welding [9, 10], the development of supporting welding materials is very important. At present, some studies have proved that GH984G alloy has good welding adaptability [11, 12]. The corresponding Ni-Fe-based welding materials for GH984G are under development [13].

After the welding and assembling of the parts of the superheater and reheater, a high-temperature homogenization process is frequently limited. Therefore the weldment will be used with segregated microstructure, non-equiaxed grains, and non-equilibrium primary phases, which may cause the stress–rupture properties of the weld metal to be inferior to those of the base metal. Moreover, at the operating temperature of 700 °C ~ 750 °C, the grain boundary becomes the weak area of this alloy [14]. Therefore, improving the stress rupture property of the weld metal is a compelling issue in the design of the welding materials.

In polycrystalline superalloys, adding an appropriate amount of boron strengthen the grain boundaries and improve their high-temperature performance [15–17]. For instance, boron significantly improve the stress rupture properties at 750 °C / 430 MPa of K4750 alloy [18]. After boron addition, the stress rupture life and
elongation of Nimonic 105 superalloy both be increased [19]. Although most studies agree on the beneficial role of boron, the strengthening mechanism of boron is contentious. Clarifying the states of the boron in the alloys is the key to revealing its mechanism. Some studies indicated that boron atoms segregate at the grain boundaries to form covalent bonds thereby increasing boundary cohesion and inhibiting embrittlement [20, 21]. On the other hand, some studies proposed that boron could improve the stress rupture properties by decreasing the diffusivity of oxygen at the grain boundaries [22]. It has also been claimed that boron decreases the rate of void formation by occupying the diffusion channels on the grain boundaries, which can reduce the adverse effects of diffusion on high-temperature deformation [23, 24]. The above studies suggest that boron plays a role through the solid solution at grain boundaries. For materials requiring long-term service in a high-temperature environment, such as GH984G alloy, precipitates often form at the grain boundaries during high-temperature exposure, which makes the form in which the boron exists and its effect on the grain boundary more complex. Some studies proposed that boron improved the high-temperature properties by inhibiting the precipitation of the $\eta$ phase in U720Li alloy and in some kinds of Ni-Fe-based alloys [25, 26]. In addition, some studies proved that boron segregating at the grain boundaries could decrease the growth rate of M$_{23}$C$_6$ along the grain boundaries thereby improving the crack resistance of the grain boundaries [17, 27], and some studies proposed that boron optimized the distribution of precipitated phases at the grain boundaries [28].

In recent years, it has been found that boron participates in the formation of grain boundary M$_{23}$C$_6$ in cast superalloys. For instance, Kang and Han [17] studied the effect of boron and zirconium on creep properties in a Nickel-based superalloy and proposed that boron had formed M$_{23}$C$_6$-type boron carbides thus improving the fracture resistance at high temperatures. In addition, boron was also found to be involved in the formation of M$_{23}$C$_6$ in the early study of weld metals for low alloy steel. For instance, Rejeesh et al. [29] studied the effect of boron addition on 9Cr-1Mo steel and showed that boron were found in M$_{23}$C$_6$ precipitates at the grain boundaries in boron-doped steels both in normalized and tempered specimens and also in creep-tested specimens. Honma [30] also found that M$_{23}$C$_6$-type boron-carbides formed in Cr-Mo low-alloy-steel weld metal. However, the role of boron in the weld metal superalloys has rarely been considered. Therefore, more studies are necessary to reveal the preferred precipitation sites of boron in superalloy weld metals to explain its influence on high-temperature properties; this is especially important for the alloys subjected to long-term exposure at elevated temperatures.

In this study, Ni-Fe-based filler wires designed for the welding of the GH984G alloy were prepared. It has been reported [31–33] that better results were obtained by adding about 50 ppm of boron to the base metal of the alloy we studied. Based on these literature reviews, we designed and prepared filling wires without boron and with 50 ppm boron to study the effect of boron on the microstructure and creep properties of weld metal. To investigate the effect of boron on the microstructure and creep rupture property, filler wires with no boron and with 50 ppm boron were produced. The role of boron in the microstructure and properties of weld metal of this alloy during stress-free long-term aging was studied by Wu et al. [13]. However, our present study specifically examined the effect of boron on microstructure and properties during the high temperature creep. The form of boron existing in the initial state of the weld metal and after creep deformation was characterized by scanning electron microscopy (SEM), high-resolution secondary ion mass spectroscopy (SIMS), and transmission electron microscopy (TEM). The effect of boron on the high-temperature creep properties of the weld metals was examined.

2. Experimental

2.1. Weld metal preparation and processing

After solution treatment at 1200 °C for 4 h, the customized alloy ingots with or without boron addition were forged into alloy rods with a cross-section of 40 × 40 mm by a billet hot forging process. The alloy rods were hot rolled into 8 mm diameter wire rods. The wire rods were annealed and cold drawn several times and finally processed into 1.2 mm diameter welding wire. Elemental analysis of the wires and ingots was performed using a plasma emission spectrometer, a carbon and sulfur analyzer, and an oxygen, nitrogen, and hydrogen analyzer. To avoid systematic errors, all welding wires were produced by the production process described above. The measured compositions of the ingots and welding wires used in this study with or without boron contents are shown in table 1. Gas tungsten arc welding (GTAW) was used for preparing weld metals with the main welding parameters as shown in table 2. To eliminate element dilution, the weld base metal and backing plate were also prepared from GH984G superalloy. The dimensions of the base metal and backing plate and weldment design are shown in figure 1(a). To induce the $\gamma'$ phase precipitation, the weld metal was subjected to heat treatment that held at 750 °C for 8 h and air-cooled to room temperature before processing into creep-test specimens. Creep rupture tests were performed at 750 °C with different stress levels 380 MPa and 210 MPa for the specimens with standard heat treatment.
2.2. Metallography

The metallography specimens were cut from the center of the cross-section of the weld metal, ground using 150–2000 grit abrasive SiC papers, and polished with 2.5 μm diamond paste. Fine polishing was performed using 1 μm diamond paste. Following polishing, the specimens were cleaned and dried. The microstructure of the specimens was revealed by electrolytic etching at 3 V for 5 s in 10% Methanolic hydrochloric acid solution (a solution of 10 ml hydrochloric acid and 90 ml methyl alcohol that could corrode both γ and γ′). Electron microscopy was performed on the samples with a Supra 35 SEM equipped with energy dispersive spectroscopy (EDS).

Two transmission characterization methods were used to determine the type and element composition of the phase at the grain boundaries. Scanning transmission electron microscopy (STEM) imaging was performed using an FEI Tecnai12; the EDS maps were observed using a Talos F200X. The thin foils used in the above two methods were first mechanically thinned to 50 μm, followed by twin-jet electropolishing in a 10% perchloric acid ethanol solution at −25 °C and 30 V. Electron back scatter diffraction (EBSD) was used to characterize the grain structures of the standard-aged (SA) specimens. Oxford Instruments HKL Channel 5 software was used to analyze the grain structure data and generate the orientation maps, inverse pole figures (IPFs), and grain boundary distribution figures.

### Table 1. Chemical compositions of experimental materials (wt%).

| Element | C    | Cr  | Fe  | Mo  | Nb  | Al  | Ti  | B    | Ni  |
|---------|------|-----|-----|-----|-----|-----|-----|------|-----|
| Ingots  | B free | 0.046 | 19.3 | 20.6 | 2.23 | 1.16 | 0.79 | 1.21 | <0.0005 |
|         | 50 ppm B | 0.045 | 19.41 | 20.27 | 1.2 | 0.8 | 1.22 | 0.0041 | Bal. |
| Wires   | B free | 0.047 | 19.51 | 20.6 | 2.22 | 1.19 | 0.76 | 1.18 | <0.0005 |
|         | 50 ppm B | 0.043 | 19.63 | 20.25 | 1.19 | 0.75 | 1.14 | 0.005 | Bal. |

### Table 2. Weld parameters.

| Parameter               | Value |
|-------------------------|-------|
| Welding current (A)     | 180   |
| Arc voltage (V)         | 14    |
| Welding speed (m min⁻¹) | 0.01  |
| Wire feed rate (m min⁻¹)| 0.1   |
| Shielding gas           | 99.999% Argon |
| Gas flow rate (L min⁻¹) | 15    |
| Interpass temperature (°C)| 20–60 |
SIMS analysis was performed with an ION TOF-SIMS 5 for the boron-doped specimens. SIMS was performed at 30 keV and 1.0 pA with Bi\(^+\) as the ion source. SIMS ion images showing boron and chromium distributions were acquired using \(^{11}\text{B}\)\(^+\) and \(^{52}\text{Cr}\)\(^+\) ions, respectively. The samples used in EBSD and SIMS that had been finely polished were ion milled for 1 h to clean the surface and achieve a high ion yield.

### 2.3. Creep testing

The creep resistance performance assessment of the specimens with or without boron additions was conducted by creep rupture tests. The dimensions and location of the creep-test specimens are illustrated in figure 1 (b). The dimensions of the specimens were determined in accordance with with GB/T 2039–2012 (a Chinese Standard for Uniaxial Creep Tests). All the creep specimens were produced using the same process to minimize errors caused by processing. Each creep specimen was sent for radiographic testing to ensure that the specimen was without cracks. Creep tests were conducted in air at 750 °C/380 MPa (high stress) and 750 °C/210 MPa (low stress) with an RDL50 Creep-testing machine. Before applying the constant load, the sample was preheated for 10 min at the test temperature.

### 3. Results

#### 3.1. Initial microstructure

Figures 2(a) and (b) show SEM images of the boron-free and boron-bearing weld metal, respectively, in the as-welded (AW) condition. It can be seen that the grain boundaries are free of phase precipitation in both the boron-free and boron-bearing alloy. The SIMS image in figure 2(c) shows that a continuous trace of boron is detected at the grain boundaries of boron-bearing weld metal; this implies that boron was segregated at the grain boundaries in the AW state.

Figure 3 shows the grain structure of boron-free and boron-bearing SA weld metal characterized by EBSD. The results show that the boron-free and boron-bearing samples possessed similar grain size distributions. The distribution of different angle grain boundaries is shown in figures 3(c) and (d). The effect of boron on the distribution and range of the grain boundary angles is not significant. The distribution of the range of the grain boundary angles of the SA weld metal is 5°~65°.

After standard heat treatment, some discontinuous phases precipitated along the grain boundaries both in boron-free and boron-bearing specimens. The morphologies of precipitates at grain boundaries with different misorientations have been observed based on the EBSD result, as shown in figure 4. To characterize the fragments of the precipitates, parameter \(\rho\) was defined as the ratio of the length of the grain boundaries decorated with precipitates and the total length of the grain boundaries in the chosen SEM images. Apparently, a larger \(\rho\) represents more precipitates in the grain boundaries. The values of \(\rho\) were 87%, 97%, and 98% for the grain boundaries in the range of 5°~25°, 25°~45°, and 45°~65° in the boron-bearing alloy, respectively, which were higher than those for the corresponding grain boundaries in the boron-free alloy. Additionally, boron promoted precipitation on the grain boundaries despite the angle of the grain boundaries.

TEM characterization was performed on the SA weld metal, and the results are presented in figure 5. The precipitates on the grain boundaries were identified as \(\text{M}_{23}\text{C}_6\) carbides that exhibited an incoherent relationship with one of the two grains along the grain boundary. The lattice constant of the \(\text{M}_{23}\text{C}_6\) phase is three times that of the austenite matrix, and the orientation relationships were identified to be \([011]_\gamma/\langle011\rangle_{\text{M}_{23}\text{C}_6}\). As shown in
Figure 3. EBSD IPF-coloring orientation maps and the distribution of the grain boundary angle of SA specimens: (a) and (c) boron-free specimens; (b) and (d) boron-bearing specimens.

Figure 4. SEM morphologies of the grain boundary microstructures of different grain boundary angles in the SA specimens: (a-c) boron-free alloy; (d-f) boron-bearing alloy.
figures 5(a) and (c), the width of the width of the $M_{23}C_6$ grains is about 50 nm in both boron-free and boron-bearing specimens. The spherical phase precipitated in intragranular was the $\gamma'$ phase, which is the main strengthening phase in the weld metal [14].

Figure 6 shows the SIMS and STEM-EDS results of the boron-bearing SA specimens. As shown in figure 6(a), the SIMS map displayed dispersed dots with high boron content distributed along the grain boundaries; this indicates that boron segregated significantly at the grain boundaries of the boron-bearing alloy in SA conditions. At the same time, STEM-EDS detected the presence of boron in the $M_{23}C_6$ precipitated at the grain boundaries, which indicates that boron had dissolved in the $M_{23}C_6$ carbides.
3.2. Creep performance

Figure 7 shows the creep strain and creep rate of the boron-free and boron-bearing weld metal at different stresses. It is obvious that the stress rupture life and elongation of the welded metals are both significantly increased and the creep rate decreased with the addition of 50 ppm boron. The stress rupture life increased from 5.2 h to 12.6 h and from 388 h to 556 h, at 380 MPa and 210 MPa, respectively. In the creep process, the weld metal first experienced a short primary creep and steady creep, and subsequently the tertiary creep acceleration stage which occupied most of the creep process. The minimum creep rate decreased from 0.51 h$^{-1}$ to 0.26 h$^{-1}$ and 0.0055 h$^{-1}$ to 0.0035 h$^{-1}$, at 380 MPa and 210 MPa, respectively, after the addition of 50 ppm boron. As can be seen in figure 7, along with the stress reductions from 380 MPa to 210 MPa, the stress fracture life of the boron-free and boron-bearing specimens increased from 5.6 h to 338 h and 12.6 h to 556 h, respectively, while the creep rate decreased by two orders of magnitude. Furthermore, at 380 MPa, almost no stable creep regime was observed, while at 210 MPa, a stable creep regime of about 120 h was observed for both boron-free and boron-bearing alloys.

3.3. Microstructures after creep rupture test

Figure 8 shows the fracture surface morphologies of the specimens after creep fracture test at 750 °C/380 MPa (210 MPa). Boron had little effect on the fracture morphology of the weld metal. All fracture surfaces show mainly exposed dendrites, with some small and shallow dimples in the fracture edge area. This indicates that intergranular fracture occurs predominantly in both boron-free and boron-containing specimens at 750 °C/380 MPa (210 MPa). The microscopic grain surface on the fracture at 380 MPa stress is smooth while that at 210 MPa stress is rough and accompanied by the presence of larger precipitated phases, as shown in figures 8(b),(e), (h),(k). This phenomenon is caused by the rapid expansion of cracks due to high stresses. From the SEM images of the longitudinal sections of the boron-free and boron-bearing fractured samples, which are located at the same distances from the corresponding fracture surfaces, obvious intergranular cracks were observed in the longitudinal section of the boron-free specimen, but only a few small cracks were observed at the edge of the boron-bearing specimen, as shown in figure 9.

Figure 10 shows the grain boundary microstructures of boron-free and boron-bearing specimens at 750 °C/380 MPa. After the creep rupture test, the $M_{23}C_6$ had coarsened slightly, and obvious dislocation clusters were discovered near the $M_{23}C_6$. The distribution of boron after the long-term creep test was analyzed using EDS maps obtained with the TEM. After the creep deformation at 750 °C/210 MPa, the $M_{23}C_6$ phases were...
obviously coarsened and boron was still present in the M23C6, as displayed in figure 11. After the creep rupture test, the $\gamma'$ phases were also coarsened, but it is stable.

4. Discussion

4.1. Effect of boron on microstructures

The low solid solubility of boron in the austenitic matrix facilitates its segregation into the grain boundaries or nearby areas; this phenomenon can be seen in the SIMS image (see figure 2(c)). The main precipitates at the grain boundaries of the weld metal are M23C6 carbides, and boron has no effect on the types of precipitates at the grain boundaries. Boron addition increases the coverage of M23C6 at grain boundaries. It is unlikely that the presence of boron drastically alters the thermodynamics of M23C6 formation, but the local increase of boron concentration may affect the kinetics of M23C6 carbide formation. The relationship between critical crystal
Figure 9. SEM images of longitudinal microstructures of ruptured SA specimens that were tested under 750 °C / 210 MPa: (a) boron-free specimen; (b) boron-bearing specimen.

Figure 10. The STEM micrographs of ruptured SA specimens that were tested under 750 °C / 380 MPa: (a) boron-free specimen; (b) boron-bearing specimen.

Figure 11. The STEM-EDS images of the longitudinal section of the boron-bearing weld metal after the creep rupture test at 750 °C / 210 MPa.
nucleus radius and grain-boundary free energy can be described by the following equation \([34, 35]\): 
\[
\gamma = \frac{2\gamma}{\Delta f_v}
\]
where \(\gamma\) is the critical crystal nucleus radius, \(\gamma\) is the grain-boundary free energy, and \(\Delta f_v\) is the free-energy change per unit volume caused by the formation of a crystal nucleus. Because of the low solid solubility of boron in the matrix, boron addition has little effect on \(\Delta f_v\). The segregation of boron at grain boundaries might fill in the voids as well as lattice defects, which decrease the grain-boundary free energy and therefore decrease the critical crystal nucleus radius. According to the above equation, boron could decrease the grain-boundary free energy \((\gamma)\) and therefore decrease the critical crystal nucleus radius, which means more crystal nuclei could reach the critical nucleus radius and grow into \(M_{23}C_6\) in the boron-bearing weld metal. This agrees with the SEM images of figure 4.

It has been reported that segregation competition exists between carbon and boron at grain boundaries since both the elements prefer to be located at grain boundaries \([36]\). It is expected that boron could diffuse more quickly than carbon based on the calculated result \([37]\). Therefore, the segregation of boron at the grain boundaries will reduce the carbon content at the grain boundaries. However, carbon is the forming element of \(M_{23}C_6\). From this view, the segregation of boron may decrease the nucleation of \(M_{23}C_6\). However, it must be noted that compared with MC carbides, the formation of \(M_{23}C_6\) needs fewer carbon atoms. Moreover, STEM-EDS detected the presence of boron in the grain boundary \(M_{23}C_6\) of the boron-bearing SA specimens, which indicates that boron can replace part of the carbon as the forming element of \(M_{23}C_6\). The formation of \(M_{23}C_6\) has also been found in other alloys \([17, 18]\). Hence, it can be concluded that the decrease of grain boundary carbon content caused by boron does not reduce the precipitation of \(M_{23}C_6\) at the grain boundaries, but promotes the precipitation of \(M_{23}C_6\). That is why boron segregation at grain boundaries could increase the grain boundary occupancy of \(M_{23}C_6\), as shown in figure 4.

### 4.2. Effect of boron on creep properties

As shown in figure 8, the creep rupture behaviors of the samples with or without boron addition were both intergranular rupture, which means that the grain boundaries were the weak area of the GH984G weld metal under the 750 °C test environment. This phenomenon is similar to that studied by Wu et al \([14]\). Wu et al proposed that intergranular fracture mainly took place at around 750 °C in the GH984G weld metal.

The types and states of grain boundary precipitates have an important influence on the grain boundary strength and creep behavior. For instance, small precipitates such as \(M_{23}C_6\) carbides dispersed on grain boundaries are generally considered to be beneficial to the grain boundary properties. It has been accepted that discretely distributed \(M_{23}C_6\) carbides along the grain boundaries are beneficial to creep as they restrain grain boundary crack propagation and grain boundary sliding \([38]\). Moreover, Guo et al proposed that chain \(M_{23}C_6\) carbides could hinder grain deformation and reduce the steady-state creep rate, improving the creep resistance of materials \([39]\). On the contrary, the continuous precipitates and bulk precipitates at the grain boundaries may be harmful to the crack resistance of the grain boundaries. For instance, some studies have shown that continuous \(M_{23}C_6\) at the grain boundaries will damage the coordination of grain deformation, resulting in a large stress concentration on the grain boundaries and consequent cracking \([34, 40]\). It has been proposed that the bulk MC carbides and \(M_{23}C_6\) carbides at grain boundaries can easily cause stress concentration that leads to grain boundary cracking \([41, 42]\). It should be noted that dispersed granular \(M_{23}C_6\) carbides are the main grain boundary precipitates in the weld metal of this alloy, as shown in figure 4.

When investigating the influence of boron addition on the creep properties at 750 °C, two different stress levels (380 MPa and 210 MPa) were applied. These two test conditions resulted in generally similar creep behaviors. In all the creep processes of this study, the primary creep and steady creep stages were short, and the tertiary creep stage occupied most of the creep process. After boron addition, the minimum creep rate decreased significantly, and the tertiary creep stage was significantly prolonged. It is accepted that the minimum creep rate is related to intragranular deformation and grain boundary sliding, and the reason for the rapid increase of creep rate in the tertiary creep stage is the initiation and propagation of defects such as cracks and voids in the weak regions such as grain boundaries \([17, 19]\). Boron was segregated at the grain boundaries and promoted the precipitation of the discretely distributed \(M_{23}C_6\) carbides at the grain boundaries. The above analysis of grain boundary precipitates demonstrates that the discretely distributed \(M_{23}C_6\) carbides at the grain boundaries do not affect the coordination of grain deformation \([43]\). Moreover, the large number of discrete distributed \(M_{23}C_6\) carbides precipitated at the grain boundaries could restrain the grain boundary crack propagation and grain boundary sliding \([44]\).

In the boron-bearing weld metal, the higher density of discrete \(M_{23}C_6\) carbide particles could strengthen the grain boundaries and inhibit the propagation of grain boundary cracks \([38]\). However, in the boron-free weld metal, the precipitation of \(M_{23}C_6\) at the grain boundaries was rare, and the \(M_{23}C_6\) spacing was large, which resulted in insufficient grain boundary strengthening. Therefore, the cracks at the edge of the sample could propagate deeply along the grain boundaries into the sample of the boron-free weld metal; but on the other
hand, the propagation was impeded significantly in the boron-bearing sample, as shown in figure 9. This is also the reason why boron could prolong the tertiary creep stage. Meanwhile, as shown in figure 10, after the creep rupture tests, dislocations were plugging around M23C6, which indicates that during the creep process, the M23C6 that was precipitated on the grain boundaries effectively hindered the dislocation movement near the grain boundaries. Therefore, because the M23C6 obstructed the grain boundary slipping and the dislocation movement, the creep rate of boron-bearing weld metal was lower than that of boron-free weld metal.

The addition of boron improved the creep resistance of the weld metal, but it did not change the main failure mechanism of intergranular fracture of weld metal at 750 °C, namely, the fact that the grain boundaries are the weak areas of the weld metal at 750 °C [14]. These results suggest that, although boron did not change the main fracture mechanism, it effectively inhibited the propagation of grain boundary cracks, thus increasing intragranular deformation, which was also the reason for the increase of elongation after the addition of boron.

5. Conclusions

The effect of boron additions on the microstructure and creep performance of a new Ni-Fe-based weld metal has been studied. Based on the investigations the conclusions are summarized as follows:

a) Boron segregated at grain boundaries in Ni-Fe-based weld metals in the as-welded condition and dissolved in M23C6 after standard aging. After the long-term creep rupture test, boron was still dissolved in the M23C6.

b) The segregation of boron at the grain boundaries could decrease the grain-boundary free energy and therefore increase the nucleation rate of M23C6, and boron could also participate in the formation of M23(C, B)x type boron-carbides. Those could promote the formation of M23C6 at the grain boundaries.

c) The higher density of discrete M23C6 particles due to 50 ppm boron addition could restrain grain boundary crack propagation and grain boundary sliding, and thereby improve the creep fracture resistance of the GH984G weld metal at 750 °C / (380 / 210 MPa).

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

ORCID iDs

Shanping Lu ORCID: https://orcid.org/0000-0002-8346-9175

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