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**High cycle fatigue behavior of a dissimilar metal welded joint in ultra-supercritical steam turbine rotor**

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**Abstract**

The high cycle fatigue (HCF) property and microstructure of a dissimilar metal welded joint (DMWJ) for ultra-supercritical steam turbine rotor were systematically investigated in this paper. The DMWJ was fabricated using narrow gap submerged arc welding (NG-SAW) technique with buttering layer. Conditional fatigue strength of 30Cr1Mo1V (BM-1), buttering layer (BL), weld metal (WM) and 30Cr2Ni4MoV (BM-2) based on S-N curve was obtained by HCF tests at room temperature. The microstructure and second-phase carbide as well as fracture appearance of BM-1, BL, WM and BM-2 were characterized by optical microscopy (OM), scanning electron microscopy (SEM) and transmission electron microscope (TEM). The results show that the BL has lower fatigue strength, which are related to more content of the soft ferrite and less content of the hard carbide. The acicular ferrite represents better fatigue properties than massive ferrite due to barrier effect. Additionally, dislocation density sharply increases near the carbides, grain boundaries, and lath structures resulting in dislocation tangle after HCF test, which helps to improve the fatigue strength of the welded joint. The carbide precipitated, aggregated, and grown in the grain boundaries promoted crack nucleation, reduced the crack propagation rate and affected the crack propagation path.

**1. Introduction**

With the rapid development of the economy, human demand for electric energy is increasing, and carbon dioxide emissions have also led to increasingly serious environmental pollution [1]. To this end, there is an urgent need for the power industry to develop in the direction of high efficiency, low energy consumption and low pollution. Therefore, ultra-supercritical (USC) generation technology has been vigorously developed in some countries because it can increase energy conversion rate and reduce carbon dioxide emissions [2]. The USC technology is primarily based on improving the steam parameters such as higher temperature and pressure. As the enhancement of steam parameters, it is necessary to develop new materials and technologies for the rotor to improve the thermal efficiency of steam turbine [3].

In the past few decades, dissimilar metal welding techniques has been gradually applied to the manufacture of some important large-scale equipment, such as nuclear, thermal power rotor and pressure vessels in petrochemical industry [4–7]. But there are big challenges to get excellent dissimilar metal welded joint (DMWJ) due to the differences in compositions and properties between different metals [8, 9]. For large thick-walled turbine rotor steels, narrow-gap SAW and TIG welding techniques are commonly adopted based on multi-layer multi-pass welding processes due to their high deposition efficiency and stable weld quality compared to conventional welding methods. Du et al [10] reported the fatigue properties and microstructure characteristics of welded nuclear power rotor, which proved that narrow-gap SAW (NG-SAW) could be used to fabricate large size rotor. Guo et al [11] mentioned that the connection of dissimilar metals in the turbine rotor was mostly realized by using multi-layer multi-pass welding technology and NG-SAW, because of the good toughness of the welded joint. Yang et al [12] studied that creep behavior of CrMoV multi-pass weld metal. They found that
multi-pass weld used by NG-SAW included two regions, equiaxed grain zone (EGZ) and columnar grain zone (CGZ). The EGZ had higher tensile strength and creep rupture stress than the CGZ. Recently, the method of adding a buttering layer in dissimilar metal welds has been studied for applications. Javadi et al. [13] deemed Cr-Mo materials were usually prepared for field welds through the application of a buttering layer because their steels cannot normally enter service without undergoing post weld heat treatment (PWHT). Especially, redistribution of residual stress should be paid attention to during welding of buttering layer [14]. In addition, Rathod et al. [15] studied the influence of the buttering layer on the tensile, impact toughness and fracture toughness properties of the DMWJ in nuclear power, and found that the properties were superior of welded joint using buttering layer. It was instrumental in reducing residual stress, controlling carbon migration, and obtaining equivalent fatigue performance compared to the welded joint of without buttering layer [16–18]. Our previous work showed the microstructure and tensile strength of a DMWJ in stream turbine rotor, which displayed that buttering layer and multi-layer multi-pass NG-SAW techniques could be employed to manufacture large-size turbine rotor [19].

For welded rotor that operate at high pressure and high-speed rotating for a long time, high cycle fatigue (HCF) performance must be addressed. Wu et al. [20, 21] reported the HCF behavior of 9%Cr/CrMoV dissimilar welded joint in nuclear power rotor at different temperatures. It was found that the fatigue fracture location appeared in the base metal (BM) at lower temperature, the fracture location was in the heat-affected zone at moderate temperature, and the fracture occurred in the interface of fusion line at higher temperature. The uneven distribution of the second-phase carbides at high temperatures are the causes of this phenomenon. Zhang et al. [22] studied very high cycle fatigue behavior of dissimilar welded joints of 9%Cr and CrMoV steels at 500 °C. It found that fatigue strength of the welds was lower than base metals. The heat-affected zone of CrMoV steel softened and micro-defects were responsible for decreasing fatigue strength. They suggested that soft zone and micro-defects would be considered in fatigue design and welding material selection. Shao et al. [23] discovered that HCF fracture mode for 9Cr-1Mo dissimilarly welded joint at different temperatures. However, there have been few reports on the welding and fatigue performance of 30Cr1Mo1V and 30Cr2Ni4MoV steels.

In this paper, the HCF behavior of 30Cr1Mo1V/30Cr2Ni4MoV welded joint used in ultra-supercritical stream turbine rotor was investigated. The novel DMWJ was welded using narrow gap submerged arc welding (NG-SAW) technique with buttering layer. Fatigue strength and cycles were obtained including two base metals (BMs), weld metal (WM) and buttering layer (BL). Micro-hardness values were obtained along the welded joint. Optical microscopy (OM), scanning electron microscopy (SEM) and transmission electron microscopy (TEM) were applied to explore the effect of microstructure on HCF behavior. The work of these fatigue properties for the various characteristic zones in the DMWJ is really significant for its reliability evaluation.

2. Experimental work

2.1. Materials and weld manufacture

Figure 1 is a schematic of geometric diagram dimensions and structural details of the DMWJ in stream turbine rotor. For convenience, the axial, circumferential and radial directions of the welded rotor are denoted by A, C and R, respectively. The two base metals 30Cr1Mo1V and 30Cr2Ni4MoV are represented by BM-1 and BM-2, respectively. Buttering layer and multi-layer multi-pass welding processes based on narrow gap SAW and TIG welding methods are used in the fabrication of welded rotors. First, multi-layer multi-pass SAW is used to form a
Table 1. Chemical composition of base metals and filler wires (wt%).

| Material          | C          | Mn         | Ni         | Cr          | Mo          | V        |
|-------------------|------------|------------|------------|-------------|-------------|---------|
| 30Cr1Mo1V         | 0.25 ~ 0.36| 0.66 ~ 1.04| <0.53      | 1.00 ~ 1.40 | 0.98 ~ 1.32 | 0.20 ~ 0.30 |
| 30Cr2Ni4MoV       | <0.37      | 0.17 ~ 0.43| 3.18 ~ 3.82| 1.45 ~ 2.05 | 0.22 ~ 0.62 | 0.06 ~ 0.16 |
| SAW wire          | <0.10      | 1.30 ~ 2.25| 2.00 ~ 2.80| <0.08       | 0.30 ~ 0.80 | /       |

Table 2. Mechanical properties of the base metals and deposited metal.

| Materials         | Yield strength (MPa) | Ultimate tensile strength (MPa) | Elongation percentage (%) | Charpy impact energy/ J (25 °C) |
|-------------------|----------------------|--------------------------------|---------------------------|---------------------------------|
| 30Cr1Mo1V         | 637                  | 789                            | 26                        | 56                              |
| 30Cr2Ni4MoV       | 834                  | 928                            | 35                        | 145                             |
| SAW wire          | 770                  | 863                            | 30                        | 89                              |

Table 3. Welding parameters for submerged arc welding.

| Wire type          | Wire diameter (mm) | Welding current (A) | Arc voltage (V) | Welding speed (mm/s) | Preheating temp. (K) | Inter-pass temp. (K) |
|--------------------|--------------------|---------------------|-----------------|----------------------|----------------------|----------------------|
| E12015-G           | 3.2                | 400–500             | 28–32           | 6–8                  | 473–573              | 279–281              |

buttering layer (BL) on the cross-section of BM-1. Then the BL and BM-2 are welded to build the weld metal (WM) by multi-layer multi-pass SAW and TIG. The post weld heat treatment (PWHT) is performed on the BL at 670 °C for 8 h and on the WM at 590 °C for 8 h to relieve residual stress and improve the toughness of the DMWJ. It should be noted that BL and WM use the same SAW wire. The macro photograph in figure 1 clearly shows the details of the welded joint. The BL and WM constituted weld region of the DMWJ, and their average widths are 5 mm and 21 mm, respectively. The thickness of welded rotor is 98 mm including 86 mm SAW and 12 mm TIG. It can be seen that there are four characteristic regions on the DMWJ, which are BM-1, BL, WM and BM-2, respectively. Tables 1 and 2 display the chemical compositions (wt%) and mechanical properties of base metals (BM-1 and BM-2) and SAW wire (BL and WM). The amount of Cr and Mo in the BMs is higher than that of the SAW wire. The strength of SAW wire is higher than that of BM-1, and lower than that of BM-2. The plasticity and toughness of SAW wire are between values of BM-1 and BM-2. The weld region plays a transitional role in composition and performance from BM-1 to BM-2. Most of BL and WM are welded by SAW because of its high deposition rate. Table 3 shows the SAW welding parameters. With the methods of the preheating before welding, controlling the inter-pass temperature and the PWHT, the welds that satisfy the quality specifications can be available.

2.2. HCF test

Figure 2 shows the sampling method and specimen size of the HCF tests. The fatigue tests were performed on BM-1, BL, WM and BM-2, respectively, and the load direction was along the axial direction of the welding rotor. Because TIG is only a small part of the weld, it is not evaluated in this study. Figure 2(a) displays a sampling of the BL and WM regions, with their geometric centers as the center of the samples. The HCF specimen was designed according to the ASTM E466–07 [24], using hourglass-shaped specimen with a diameter of 5 mm at the thinnest position, as shown in figure 2(b). A sinusoidal waveform was selected with a stress ratio $R = -1$ ($R$ is the ratio of the minimum peak stress to the maximum peak stress). The frequency was about 130 Hz, while the data acquisition interval was 300 s. The tests were performed at room temperature. During the tests, the cycle and fatigue strength of the fracture specimen were recorded. When the cycle was $1.0 \times 10^7$, the test was stopped and the fatigue strength was recorded.

2.3. Microstructure analysis

Metallographic specimen of $60 \times 10 \times 10$ mm was cut from the welded joint to investigate microstructure and micro-hardness of the BM-1, BL, WM and BM-2, a solution of 4% HNO3 and ethanol solution were used to etch the samples. Optical microscope (OM, Nikon: ECLIPSE MA200) was employed to observe the microstructure of BM-1, BL, WM and BM-2. Besides, scanning electron microscopy (SEM, FEI, 25 V) and transmission electron microscope (TEM, JEM-2100F, JEOL) were also used to observe second-phase carbides...
and fracture surface. Micro-hardness was measured using a hardness tester (HXD-1000TMC) by using a test load of 200 g (1.961 N) and a test time of 15 s.

3. Results and discussion

3.1. HCF properties

Figure 3 shows the results of HCF test for the BM-1, BL, WM and BM-2. Points with arrows mean the specimens that didn’t fail at $10^7$ cycles. The relationships between maximum stress and cycle (S-N curves) were fitted using the Basquin’s equation [25]:

$$\sigma_{\text{max}} = AN_f^{b}$$  \hspace{1cm} (1)

where $\sigma_{\text{max}}$ is the maximum stress, $N_f$ is the number of cycle to failure, $A$ is the fatigue strength coefficient, and $b$ is fatigue strength exponent. The fitting curves for the four regions are shown in figure 3, and the corresponding fatigue parameters $A$ and $b$ are shown in table 4. It is clearly seen that the fatigue strength coefficients are higher for the specimen tested at base metals (BM-1 and BM-2) than weld regions (BL and WM), while the fatigue
strength exponent is just the opposite. Conditional fatigue strength (at $10^7$ cycles) is evaluated to be 338 MPa, 308 MPa, 346 MPa and 396 MPa in BM-1, BL, WM and BM-2, respectively. BM-2 has the highest fatigue strength, while BL has the lowest fatigue strength. The fatigue strength of BM-1 and WM are relatively close. According to the chemical compositions and mechanical properties of BMs and SAW wire in tables 1 and 2, tensile strength of the welding wire of BL and WM is between the tensile strength of BM-1 and BM-2. However, the fatigue test showed that the fatigue strength of BL was lower than that of BM-1 and WM, which indicates that BL becomes a fatigue weak region of the welded joint after welding and PWHT. The fatigue properties of the four regions showed unevenness. Based on the correlation between microstructure and mechanical properties, microstructures are the main reason for the uneven fatigue strength distribution of the DMWJ.

### 3.2. Microstructure and micro-hardness analyses

Figure 4 shows the OM microstructure of the BM-1, BL, WM and BM-2. The microstructures of BM-1 are ferrite and tempered sorbite, as shown in figure 4(a). It is well known that ferrite/sorbite in this kind of steel is beneficial to improve elevated temperature strength and creep property [26]. Figures 4(b) and (c) display the microstructure characteristics of BL and WM. It can be seen that their microstructures are granular bainite and ferrite, but the shape of ferrite is different. The massive ferrite is on the BL, while the acicular ferrite is on the WM. Acicular ferrite has better fatigue performance due to the barrier effect, which hinders the movement of dislocations to the grain boundary [27]. This is the main reason why the fatigue strength of WM is higher than BL. In addition, since the fatigue strengths of BM-1 and WM are comparable, it can be inferred that their microstructures have similar fatigue properties. As shown in figure 4(d), the microstructures of BM-2 are lath tempered martensite and bainite, which have high strength, good toughness and resistance to stress corrosion. It is commonly used in low-pressure rotor of nuclear power and thermal power [28]. Lath tempered martensite

| Test location | $A$ (MPa) | $b$ | $A_{10^7}$ cycles |
|---------------|-----------|-----|-------------------|
| BM-1          | 710.3     | −0.046 | 338              |
| BL            | 559.3     | −0.037 | 308              |
| WM            | 628.7     | −0.037 | 346              |
| BM-2          | 766.2     | −0.041 | 396              |
and bainite can hinder dislocation motion, thereby increasing the fatigue strength of the material, which is consistent with the higher fatigue strength of BM-2 in welded joint. Figure 5 shows the distribution of the micro-hardness along the welded joint. The micro-hardness test was performed along the ‘L’ line in the figure, and the macro image of the welded joint was used to indicate the different regions of the welded joint. It can be seen that the micro-hardness values in BM1, BL, WM and BM2 are relatively stable, which is related to their relatively uniform microstructure. The average values are 268 HV in BM-1, 224 HV in BL, 251 HV in WM, and 296 HV in BM-2, respectively. BL has the lowest hardness and is the softening zone in welded joint. BM-2 has higher hardness than others. The distribution of micro-hardness in the four regions is basically consistent with the distribution of fatigue strength. In fact, micro-hardness can indirectly reflect the strength of the material [29]. Moreover, it is worth noting that there are 3 fusion boundaries (FBs) and heat affected zones (HAZ) at the junction of the four regions. The micro-hardness has a local maximum values at FB and gradually decreases toward the base metal at HAZ, which related to the induced martensite during the welding thermal process. Owing to welding heat flow, the microstructure near FB regions will change considerably, especially the coarse-grained heat affected zone (CGHAZ) can form lath martensite [30]. This demonstrates that hardness and fatigue strength may increase near the interface regions (FB and HAZ).

3.3. Second-phase characteristics

In heat-resistant steels and their welded joints, the second phase strengthening is an important way to strength material performance, and also the main route to improve the fatigue durability. Figure 6 shows SEM images of BM-1, BL, WM and BM-2, and it can be seen that a large number of second phases are dispersed in the matrix. Alloying elements such as Cr and Mo in the BMs and weld region combine with carbon to form second-phase carbide during heat process such as welding thermal cycle and PWHT. These second-phase carbides are mainly $M_7C_3$, $M_{23}C_6$ in CrMoV and NiCrMoV steels [31, 32]. The second phases can fix the dislocations effectively, and improve fatigue strength of the welded joint at room temperature. As shown in figure 6(a), there are many carbides of different sizes in the grain interior and grain boundary of BM-1. Large-sized carbide may cause stress concentration during fatigue, which may affect the initiation and propagation behavior of fatigue crack. Figures 6(b) and (c) show the carbide distribution of BL and WM, and it can be seen that the carbides are mostly distributed in the grain boundary and have spheroidal characteristics. During high temperature tempering, the weld regions precipitated more carbides in the grain boundary [33]. For high cycle fatigue, the second phase with a larger size in the grain boundary will cause a large stress concentration due to dislocation plugging, which becomes a possible crack initiation and a factor affecting crack propagation path [34]. Figure 6(d) shows the carbide characteristic of BM-2, and it can be seen that the very fine carbides are distributed in the grain boundary and the lath martensite. The lath interfaces and carbides cause different degrees of dislocation accumulation, thereby strengthening the matrix. The fine carbides not only increase the strength of the BM-2, but also improve its toughness. Additionally, the number of carbide in BL and WM is less than two BMs, which means the second phase strengthening effect is weaker in weld regions. It can be seen from the chemical composition of table 1 that the sum of the contents of Cr and Mo of welding wire is less than that of BM-1 and BM-2. Fewer alloying elements are provided for BL and WM during the carbides
precipitate. This is one of the reasons why the hardness and fatigue strength of the weld region, especially BL, is low.

3.4. TEM observation before and after HCF
From the above analysis, it is known that in addition to the characteristic microstructure, the second-phase carbide is also an important factor affecting high cycle fatigue behavior of BM-1, BL, WM and BM-2. Therefore, it is necessary to further study the behavioral mechanism of the carbides in HCF. Figure 7 shows TEM photographs of BM-1, BL, WM and BM-2 before and after fatigue (10^7 cycles). Figures 7(a) and (b) display the changes of the second-phase carbide and dislocation in BM-1 before and after fatigue. The carbides are mainly rod-shaped second phases with length of 50–200 nm and have certain directionality. The morphology of the carbides don’t change during fatigue tests, but the dislocation density increased. Dislocation pile-up and tangle occur near carbides, resulting in the uneven distribution of dislocation density. Figures 7(c) and (d) show the change of the carbide and dislocation in BL before and after fatigue. It can be seen that spherical carbides have diameter of about 200 nm. However, some of the carbides have diameter of 400 nm and distribute along the grain boundary, which is consistent with that observed by SEM in figure 6. The dislocation density of BL increases after fatigue. More dislocations tangle near the carbides and sub-grain boundary and result in higher dislocation density. It can be seen from figures 7(e) and (f) that the bulk carbides of WM have sizes between 110 and 250 nm. The morphology of the carbide don’t change before and after fatigue, while the dislocation density increased. The dislocation density near the carbide and the grain boundary is higher. Figures 7(g) and (h) show the distribution of the carbides and dislocations of BM-2 before and after fatigue. The carbides with a size of approximately 100 nm is dispersedly distributed on lath martensites. Uneven dislocation accumulate among the martensite laths after fatigue. It means that the lath martensite affect the uneven distribution of dislocation density. In summary, during the HCF process, dislocation pile-up and tangle occur at the second-phase carbide and the grain boundary. Higher dislocation density increases the fatigue strength of the welded joint. However, the uneven distribution of carbides and dislocation densities may lead to non-uniformity of internal stress, which affects the fatigue crack nucleation and propagation behavior. The crack nucleation is mainly realized by dislocation accumulation. It is confirmed by literature that the dislocation plug-in can promotes the crack nucleation. Moreover, during high-cycle and very high-cycle fatigue damage, the uneven distribution of dislocations also leads to the formation of substructure refinement [35]. The fatigue crack propagation behavior is also affected by the second-phase carbide and grain boundary. When the crack tip emits dislocations,
high-density dislocations around the carbides and grain boundary may cause microvoids and microcracks due to stress concentration. Large microvoids and microcracks affect the direction of crack propagation, and small microvoids can absorb energy, so as to slow crack propagation rate.

**Figure 7.** TEM images of BM-1 (a) and (b), BL (c) and (d), WM (e) and (f) and BM-2 (g) and (h) before and after HCF.
3.5. Fractography analysis

Figure 8 shows the HCF fractography of BM-1, BL, WM and BM-2. It can be seen from the macro fractures that there are three regions on the fracture surfaces, which are fatigue crack initiation zone, propagation zone and...
fracture zone. The fatigue crack initiation of the four regions occur at the surface of specimens. From the analysis of the microstructure, it is known that the grain boundary, the second-phase carbide, the martensite and the bainite of the welded joint are key microstructure factors affecting the crack initiation. Because they can cause uneven dislocation density, which may cause local stress concentration. In propagation zones, BM-1, BM-2 and BL and WM show different features. The propagation zones of BM-1 and BM-2 have many tearing. In particular, the propagation zone of BL and WM includes I and II zones. I zone is relatively flat with almost no tearing, and II zone has clear tearing marks. The I zone is a slow propagation zone, and the II zone is a fast propagation zone. The I zone is beneficial to restraining crack propagation, which is advantageous for the fatigue properties of the weld region. Figures 8(b), (d), (f) and (h) show a large multiple of the propagation zone images. Fatigue striations and a small number of secondary cracks can be seen on the propagation of BM-1 and BM-2. A large number of microvoids and secondary cracks can be seen in BL and WM, and the second-phase carbides are present in the microvoids. From the SEM and TEM analysis of BL and WM, it can be seen that the large spherical carbides in the grain boundary can accumulate a large number of dislocations during the fatigue process. This results in large stress concentrations on the carbides, and this causes fatigue cracks to propagate along the carbides. Therefore, a large number of microvoids are observed on the fatigue surface. The fracture analysis confirms that the second phase may cause the stress concentration and eventually affect the HCF properties. The crack nucleates near the carbides to form the main crack and the secondary crack [36]. Microvoids and secondary cracks absorb energy during crack propagation, thereby reducing crack propagation rate. This results in a slower crack propagation rate in the I zone. Although BL and WM have fewer carbides to strengthen the matrix, the slower crack growth rate in the early stage is beneficial to increase their fatigue properties.

4. Conclusions

The high cycle fatigue tests were performed in different regions of the dissimilar metal welded joint composed of BM-1, BL, WM and BM-2. The S-N curves and conditioned fatigue strength were obtained. Microstructure, micro-hardness, precipitation and fractured surfaces of different regions in the welded joints were investigated. The main conclusions were drawn as follows:

1. The fatigue strength was 338 MPa in BM-1, 308 MPa in BL, 346 MPa in WM and 396 MPa in BM-2, respectively, and the distribution was consistent with the average value of micro-hardness from BM-1 to BM-2.

2. The microstructures of BM-1 and BM-2 were ferrite and sorbite, martensite and bainite, respectively. The microstructures of BL and WM were mainly bainite and ferrite, but massive ferrite in BL, while acicular ferrite in WM. The acicular ferrite brought a more favorable effect on the fatigue properties than massive ferrite due to barrier effect.

3. There were more second-phase carbides in BM-1 and BM-2, and fewer carbides in BL and WM. The carbides of BL and WM were mainly distributed on the grain boundaries and are spherical. During the fatigue process, dislocation tangle occurred in the sub-structure such as the carbide, grain boundary, and lath structure, which increased the dislocation density and acted as a strengthening factor for fatigue properties.

4. The fracture surfaces of BL and WM had a large number of microvoids and secondary cracks in the initial stage of crack propagation, which indicated that the carbides in the grain boundary inhibited the crack propagation.

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