

Microstructure and Creep Rupture Characteristics of MarBN Welded Joint

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Abstract

The welded joint of a new kind of martensitic heat resistant steel, namely MarBN, was made by using matching welding rods. Creep rupture tests were carried out in the condition of 650 °C/120–220 MPa. Microstructure before and after creep rupture tests was observed and hardness tests on the welded joint were performed. Results showed that the microstructure of CGHAZ was finer than those of FGHAZ, ICHAZ and base metal after PWHT. FGHAZ and ICHAZ inherited the coarse lath morphology from base metal but with fine grains in some areas. Before creep, CGHAZ had the highest hardness, while FGHAZ and ICHAZ had the lowest. It was also worth noting that hardness of weld metal was higher than that of base metal before creep. During the process of long-term creep, weld metal experienced the most severe degradation as a result of the most serious ripening of $M_{23}C_6$ and it eventually fractured in the form of cavity aggregation. Therefore, weld metal was the weakest region throughout the whole welded joint during creep and the welded joint exhibited no tendency towards type IV creep cracking.

1. Instruction

Nowadays, with the rapid growth of economy and fast development of technology, energy is becoming more and more important to human society. Thermal power, which stems from fossil energy, is one of the most common energy sources. However, Thermal power also brings out energy and environmental crises, because it increases the consumption of non-renewable energy and the carbon emission which could cause greenhouse effect. Under the ever-increasing energy demand and the calling for more economical, environmental-friendly, and sustainable development, the most direct and effective way to achieve higher productivity and efficiency in thermal power plants is to increase the operating parameters (temperature and pressure). Therefore, it poses a challenge to the properties of power plant used heat resistant steels^[1, 2].

Because of its high creep rupture strength and resistance to steam corrosion, high Cr Martensitic Heat Resistant Steel (MHRS) is an ideal material in thermal power plant construction. For 9Cr MHRS, There are 4 generations in the process of its development, the representative figures are T/P9 (the 1st generation), T/P91 (the 2nd generation), T/P92 (the 3rd generation) and MarBN (the 4th generation) respectively. As the representative figure of the 4th generation of 9Cr MHRS, MarBN is more suitable for components with operating temperature above 630 °C than T/P92 on account of its chemical composition of 9Cr-3W-3Co which can bring about higher creep rupture strength^[3–6]. Up to now, there are much research on the weldability of T/P92 (9Cr-1.8W-0.5Mo) steel, and 2 crucial problems are pretty much the agenda for researchers: low toughness of Weld Metal (WM) and softening of Heat Affected Zone (HAZ)^[7]. WM is a relatively vulnerable region because of its low initial toughness and aging embrittlement during service^[8–10]. On the other hand, HAZ softening of T/P92 might result in type IV creep cracking which would eventually lead to premature failure^[11, 12]. Compared with T/P92, MarBN steel has both higher strength and alloy content, and its weldability (especially for the mechanical compatibility between welded joint and base metal) has drawn wide attention. However, research on these issues is rarely reported. Recently, because there is no matching consumables, Ni based filler material has been used in the welding of SAVE12AD steel^[13] (a kind of Japan developed MarBN steel, and is named as P93 in ASME standard). This dissimilar metal welded joint is poor in heat resistant ability on account of the differences in physical and chemical properties (such as linear expansion coefficient and chemical affinity) between different metals.

To solve the above-mentioned problems, welded joint of MarBN steel casting with matching filler material was fabricated. Microstructure observation and hardness test were performed on samples both in as-welded and after creep states. Additionally, microstructure transformation and softening of the welded joint were analyzed, and the most vulnerable region was determined. Findings of this research would provide technical reference to the construction of high-parameter thermal power units.

2 Experimental

MarBN steel casting with the dimension of 140 mm×390 mm×25 mm was used. The heat treatment was normalizing (1200 °C × 10 h + air cooling) + tempering (750 °C × 10 h + furnace cooling). The welding rod was CHRMET933. Chemical compositions of Base Metal (BM) and Deposited Metal (DM) were listed in Table 1.

Table 1

Chemical compositions of the BM and DM(mass%)

	C	Si	Mn	S	P	Cr	W	Co	Ni	V	Nb	B	N	Fe
BM	0.11	0.36	0.62	0.003	0.019	8.96	2.61	2.98	0.17	0.25	0.07	0.012	0.02	Bal.
DM	0.10	0.20	0.59	0.01	0.01	8.96	2.68	3.26	0.37	0.25	0.036	0.006	0.046	Bal.

The dimension of welding groove was illustrated in Fig. 1. Preheating temperature was 200 °C, and interpass temperature was no more than 350 °C. Post Weld Heat Treatment (PWHT) was 735 ± 10 °C × 8 h. Shielded Metal Arc Welding (SMAW) was adopted for the whole welding process. The welding parameters were listed in Table 2.

Table 2

The welding process parameters

Welding Pass(es)	Process	Filler Metal		Welding Current (A)
		Classification	Diameter (mm)	
Backing bead	SMAW	CHROMET 933	Φ 3.2	100–140
Filling bead	SMAW	CHROMET 933	Φ 4.0	130–170
Capping bead	SMAW	CHROMET 933	Φ 4.0	130–170

Samples for creep rupture tests were cut from the welded joint, and their dimensions were illustrated in Fig. 2. As could be seen in Fig. 2, The welded seam was located in the middle of the sample. Testing temperature and stress were 650 °C and 120–220 MPa respectively.

After creep rupture test, the location of fracture was determined. Samples were etched in FeCl₃ solution before observed under ZEISS Axio-1 optical microscope(OM) for microstructure. A HX-1000A micro-hardness tester was adopted for hardness distribution of the welded joint, and the load and duration were 9.8 N and 10 s respectively. The degradation of lath structure, coarsening of precipitates and creep cavities in different regions of the welded joint were also investigated by using the Quanta 400 Scanning Electron Microscope (SEM).

3 Results

3.1 Creep rupture test

Figure 3 was creep stress against rupture time plot of the MarBN steel welded joint. As could be seen in the figure, the fracture time showed a monotonous downward trend with creep stress decreasing, and it was basically a linear relationship in logarithmic coordinates. Figure 4 exhibited the feature of each sample after fracture, and the welded seam was marked inside yellow lines. It was obvious that all samples fractured at WM regions when creep stress was below 220 MPa. On the other hand, the sample enduring 220 MPa fractured at BM. Moreover, necking was becoming less and less obvious with creep stress decreasing, and this unraveled that the creep plasticity decreased with stress decreasing and creep time prolonging..

3.2 Microstructure and hardness before creep

Figure 5 showed the microstructure of different sub-regions of the welded joint after PWHT. Figure 5a was an overall picture of the whole welded joint which consisted of WM, HAZ and BM, and the width of HAZ was about 5mm. Furthermore, based on the distance from fusion line, HAZ could be subdivided into Coarse Grain HAZ (CGHAZ), Fine Grain HAZ (FGHAZ) and Inter-Critical HAZ (ICHAZ), and their widths were 1.5mm, 1.5mm and 2mm respectively. Microstructure of the whole welded joint was tempered martensite but with different grain sizes in different sub-regions. The lath morphology was clear and fine in WM (Fig. 5b), while it was much coarser in BM (Fig. 5f). Compared with BM, the lath of CGHAZ was finer, and it was the result of diffusive Martensite (M) → Austenite (A) transformation at high temperature (Fig. 5c, the dark area was the phase transformation structure, and it will be discussed in section 4. Hereinafter, this structure will be referred as “fine grain structure”). Most areas of FGHAZ had maintained coarse structure, while some areas also had the “fine grain structure” (Fig. 5d). It could be also observed that the “fine grain structure” in ICHAZ was less than that in FGHAZ (Fig. 5e).

Figure 6 was hardness distribution of the whole welded joint after PWHT. As shown in the figure, the average hardness of WM was about 285 HV, and was higher than the one of BM which was about 268 HV. For all the sub-regions of HAZ, CGHAZ had the highest hardness, peaking at 340 HV. The lowest hardness was found at the border of FGHAZ and ICHAZ, averaging at 243 HV.

3.3 Microstructure and hardness after creep

Figure 7 showed the OM structure of the fractured sample enduring 120 MPa (the rupture time was 12596.7 h). As could be seen, the fracture location was in the middle of WM (Fig. 7a). Many macro- and micro- cracks were also observed in WM (Fig. 7c). On the other hand, there was no cracks or cavities found in HAZ and BM (Fig. 7b).

Figure 8 was SEM pictures of the 120 MPa/12596.7 h sample after fracture. It was obvious that there were a great number of micro-cracks in WM which resulted from the aggregation of creep cavities (Fig. 8a and 8b). Additionally, white and gray precipitates could be found in creep cavities (Fig. 8c). According to EDS analysis, the white precipitates were rich in W, so they should be Laves phase (Fig. 8d). On the contrary, the gray precipitates were rich in Cr, but with low W content, so they should be $M_{23}C_6$ carbides (Fig. 8e).

Figure 10 was hardness distribution of the welded joint after creep (120 MPa/12596.7 h). It could be seen that hardness of all regions showed a downward trend compared with the one before creep, and the difference between them became smaller. Notably, the hardness of WM (285 HV) was higher than the one of BM (285 HV) before creep. But after creep, the hardness of WM (230 HV) became lower than the one of BM (255 HV), and it was the lowest hardness throughout the whole welded joint. This was in good agreement with the fracture location (Figs. 4 and 7). It was also worth noting that there was a hardness trough in ICHAZ, but it was still higher than in WM.

4 Discussion

4.1 Why did not type IV creep cracking take place?

Type IV creep cracking is a kind of common premature failure in MHRs welded joint during service, and it is well established that the type IV creep cracking happens under the interaction between high temperature and low stress and is located in FGHAZ or ICHAZ. Up to now, much in-depth research has been conducted on its mechanism. Wang et al.^[14, 15] indicated that type IV creep cracking nucleated in ICHAZ of P91 welded joint, and was the result of Cr segregation in grain-scale. Pandey C et al.^[16] investigated the effect of PWHT on type IV creep cracking of P91 welded joint. They pointed out that although PWHT (760 °C + 2 h) could mitigate hardness unevenness throughout the welded joint, high temperature and pressure would bring it out again during service. They also unraveled that Post Welding Normalizing Treatment (PWNT) could basically eradicate the tendency of type IV creep cracking in P91 welded joint. Wang et al.^[17]

reported that the type IV creep cracking in P92 welded joint nucleated at the region with relatively high temperature (a little higher than A_{c3} , namely FGHAZ) and under low stress (lower than 120 MPa), and coarse Laves phase at grain boundaries, loss of lath morphology and low hardness were all stimuli to type IV creep cracking. All the research on type IV creep cracking above is focused on the 2nd and 3rd generation of 9Cr MHRS. As the representative figure of the 4th generation of 9Cr MHRS, MarBN steel is quite different from P91 and P92 in chemical composition, so its microstructure evolution during the welding process would be also quite different.

Recently, Japanese researchers^[18] found out that when B content reached 100 ppm in P93 steel, diffusive phase transformation in HAZ could be suppressed, so $M \rightarrow A$ transformation in the heating process would take place in the displacive form which could impede the formation of FGHAZ and ICHAZ. On the other hand, traditional FGHAZ or ICHAZ was the prerequisite for type IV creep cracking, so MHRS with high B content exhibited a low type IV creep cracking tendency. Based on the findings above, researchers^[19, 20] further pointed out that grain refinement in CGHAZ might result from the formation of so called “lath austenite^[21]” in the heating process. Not only would this “lath austenite” maintain certain orientation relationship with the original martensite, but it also retained the sub-structure of the martensite (lath and dislocation density) which could increase the stored energy. During subsequent heating process, because of the stored energy it inherited, the “lath martensite” would recrystallize, which could eventually lead to grain refinement.

According to the results of this research, another mechanism of why type IV creep cracking did not take place was put forward. As could be seen in Fig. 5d, although FGHAZ had inherited the coarse martensitic morphology from BM, some areas (the dark “fine grain structure”) still showed the sign of diffusive $M \rightarrow A$ transformation. Therefore, high B content (0.012% in wt. in this research) could not suppress diffusive $M \rightarrow A$ transformation completely, which meant the diffusive $M \rightarrow A$ transformation might still happen in MarBN if higher driving force (higher temperature) for phase transformation could be reached. On the other hand, obvious grain refinement was observed in CGHAZ (Fig. 5b). As mentioned above, some researchers ascribed the grain refinement in CGHAZ to recrystallization, but there was another possibility: With temperature increasing, the driving force would be enough to make diffusive $M \rightarrow A$ transformation take place completely in CGHAZ, thereby refining the grains. Firstly, the as-delivered state of MarBN was normalizing + tempering. It was well known that tempering could basically eliminate the residual stress caused by normalizing and significantly reduce the dislocation density in martensitic laths simultaneously, so stored energy stemming from the dislocation retained in “lath austenite” might not be adequate to make recrystallization possible. Secondly, B content in this experimental MarBN could not suppress diffusive $M \rightarrow A$ transformation completely. As for the reason why MarBN welded joint fractured at WM rather than FGHAZ or ICHAZ (type IV creep cracking) below 220 MPa, it might be caused by the strength reduction in WM (compare Fig. 6 and Fig. 10). In other words, if WM could maintain high strength during creep, type IV creep cracking was still likely to take place in MarBN steel welded joint.

4.2 Why did fracture happen at WM below 220 MPa?

For the as-delivered state of MarBN steel welded joint, the hardness of WM was higher than that of BM (Fig. 6). In the process of high stress creep (220MPa), there was no obvious degradation taking place in WM because of short rupture time. Thus the strength of WM was still higher than that of BM, thereby making the fracture happen in BM. But in the condition of low stress creep (below 220MPa), the rupture time was long and therefore both WM and BM would experience evident degradation such as $M_{23}C_6$ coarsening, dislocation recovery and sub-grain growth. Nevertheless, the degradation in WM was more serious than that in BM (Fig. 9a and 9b), thereby leading to a faster hardness reduction in WM, so the fracture would happen in WM under this circumstance. Abe^[22] pointed out that when B content was more than 100 ppm in 9Cr-3W-3Co, the coarsening of $M_{23}C_6$ could be suppressed. Actually, in most cases, B content in WM was much lower than that in BM because of burning loss during welding process, and it was just only a half of B content in BM in this research (60 ppm in WM versus 120 ppm in BM according to Table 1). This would lead to a faster $M_{23}C_6$

coarsening rate in WM than that in BM, which meant more serious degradation in WM. This could also be proved by a more significant hardness reduction in WM compared with BM during creep (see Fig. 6 and Fig. 10 for comparison).

5 Conclusion

1. For MarBN steel welded joint after PWHT, microstructure of CGHAZ was finer than that of base metal, while FGHAZ and ICHAZ inherited coarse lath structure from base metal but with fine grains in some areas. FGHAZ and ICHAZ had the lowest hardness.
2. In the creep condition of 650 °C and below 220 MPa, weld metal of MarBN welded joint experienced the most significant softening because of the most serious ripening of $M_{23}C_6$ and therefore fracture took place at weld metal. In the range of testing temperature and stress, the welded joint showed no tendency towards type IV creep cracking.
3. In CGHAZ of MarBN steel welded joint, diffusive martensite to austenite transformation happened during the welding heating process. Because of short residence time at high temperature, it was difficult for grains to grow, thereby leading to grain refinement.

Declarations

Conflict of Interest

The authors declare that there exists no competing financial interest or personal relationships that could have appeared to influence the work reported in this paper.

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Figures

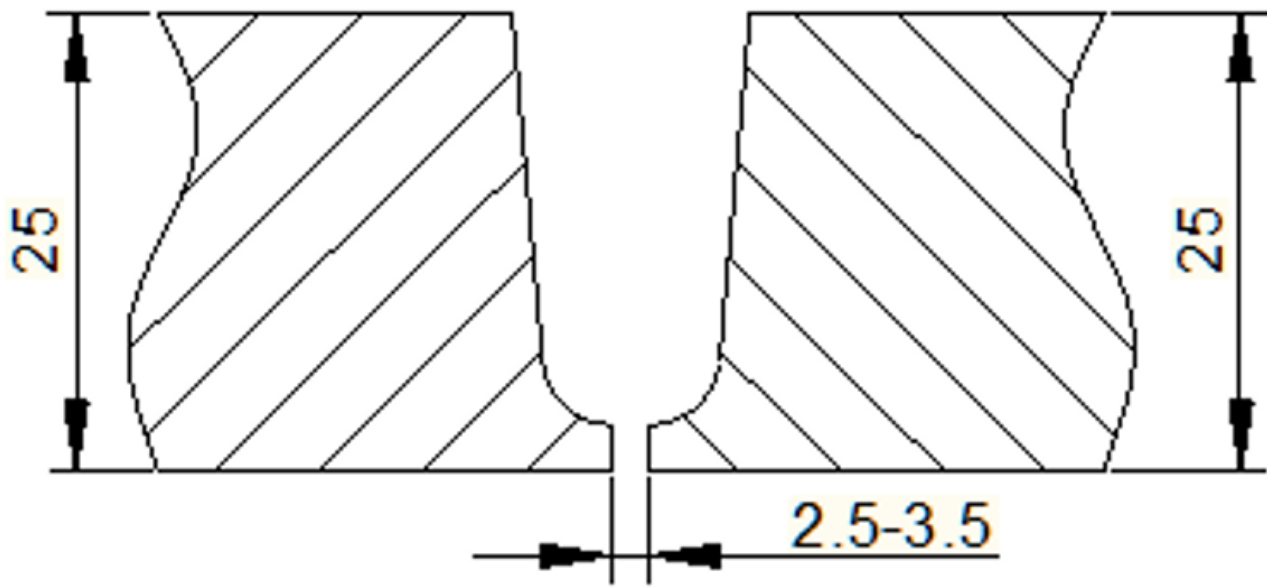


Figure 1

The dimension of welding groove (in mm)

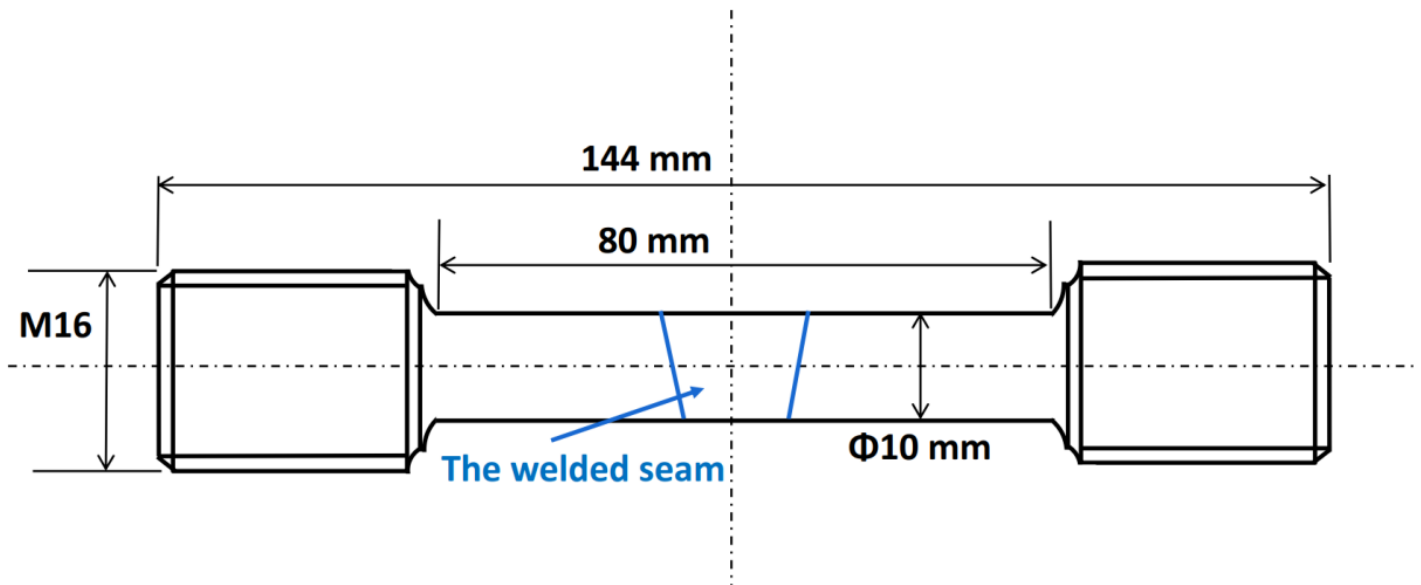


Figure 2

Dimension of creep rupture test sample

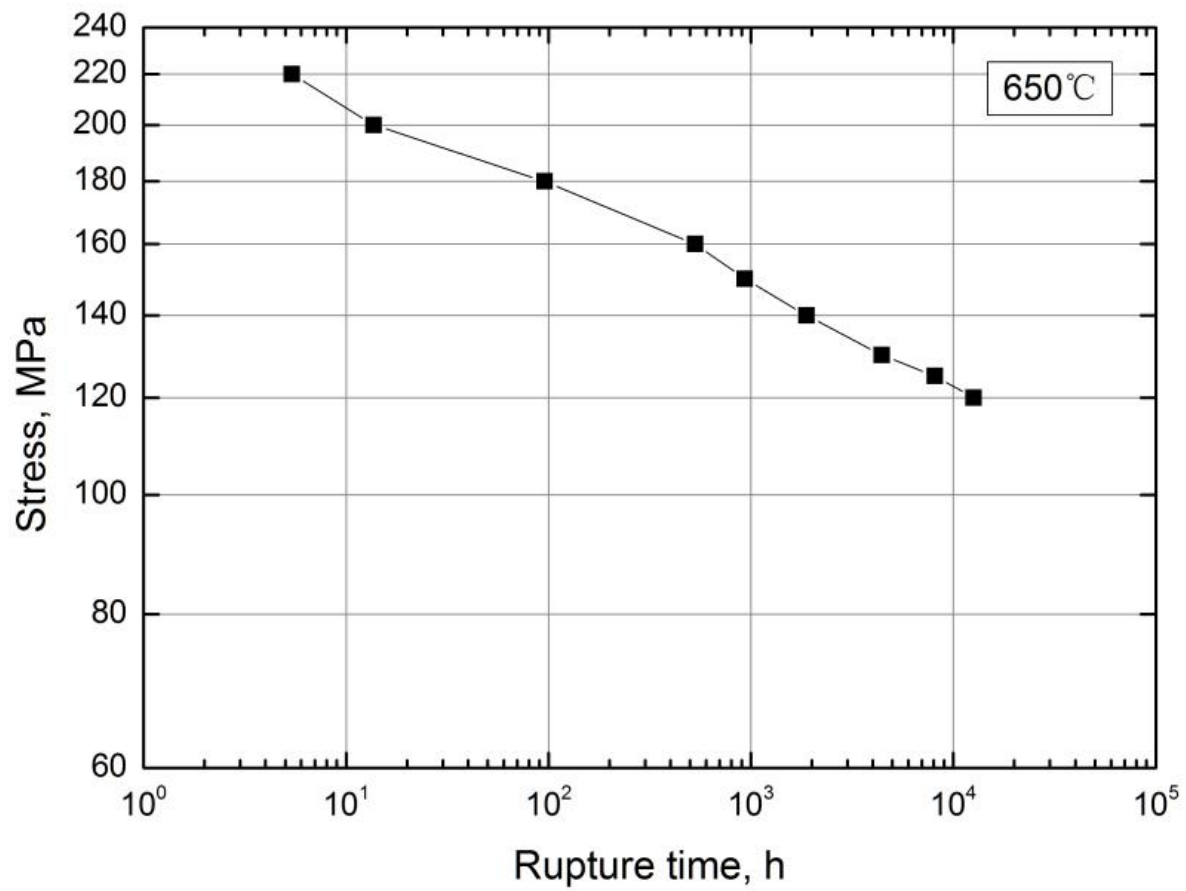


Figure 3

Stress versus rupture time plot of MarBN steel welded joint



Figure 4

Samples after creep rupture

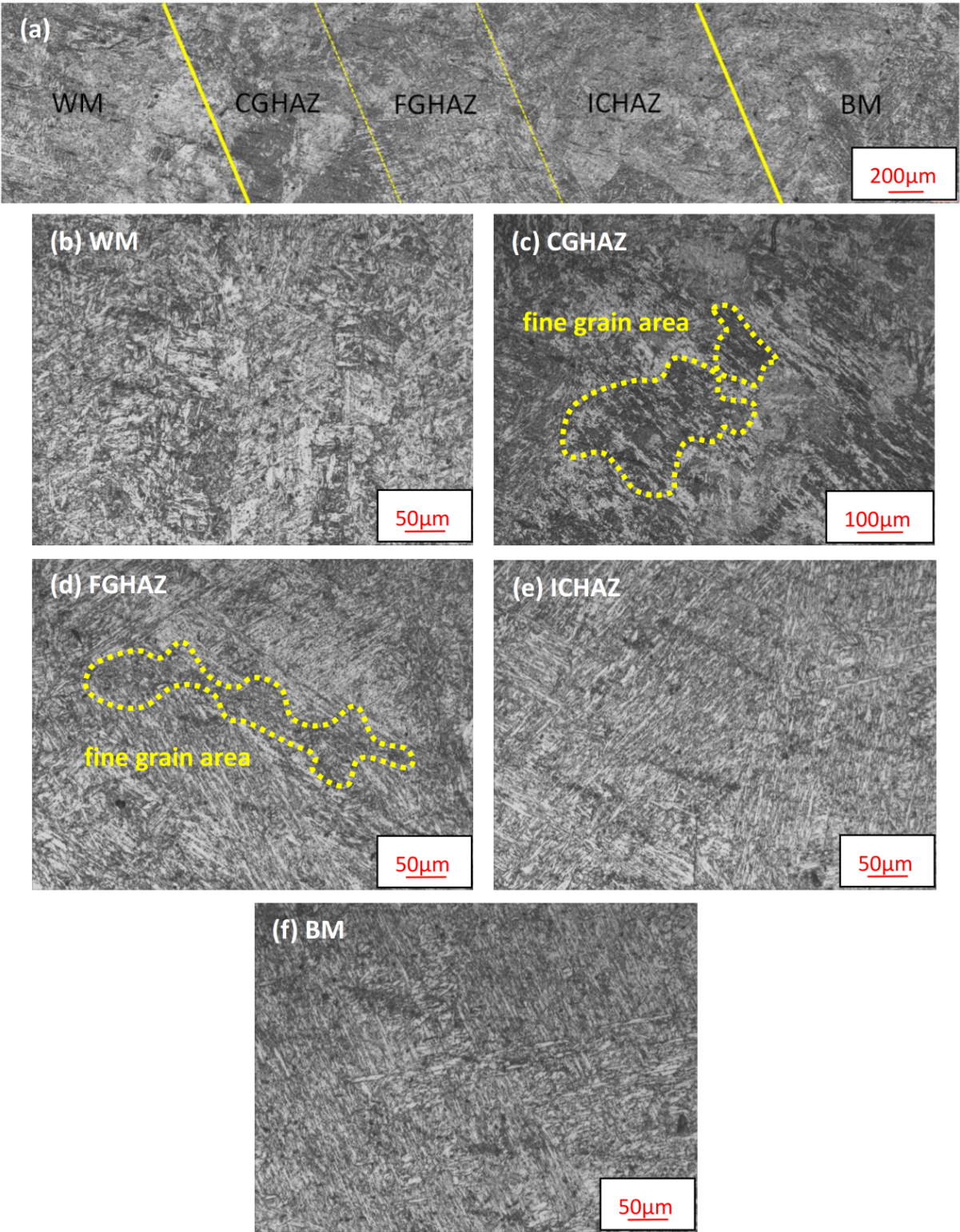


Figure 5

Microstructure of MarBN welded joint after PWHT

(a) Welded joint, (b) WM, (c) CGHAZ, (d) FGHAZ, (e) ICHAZ, (f) BM

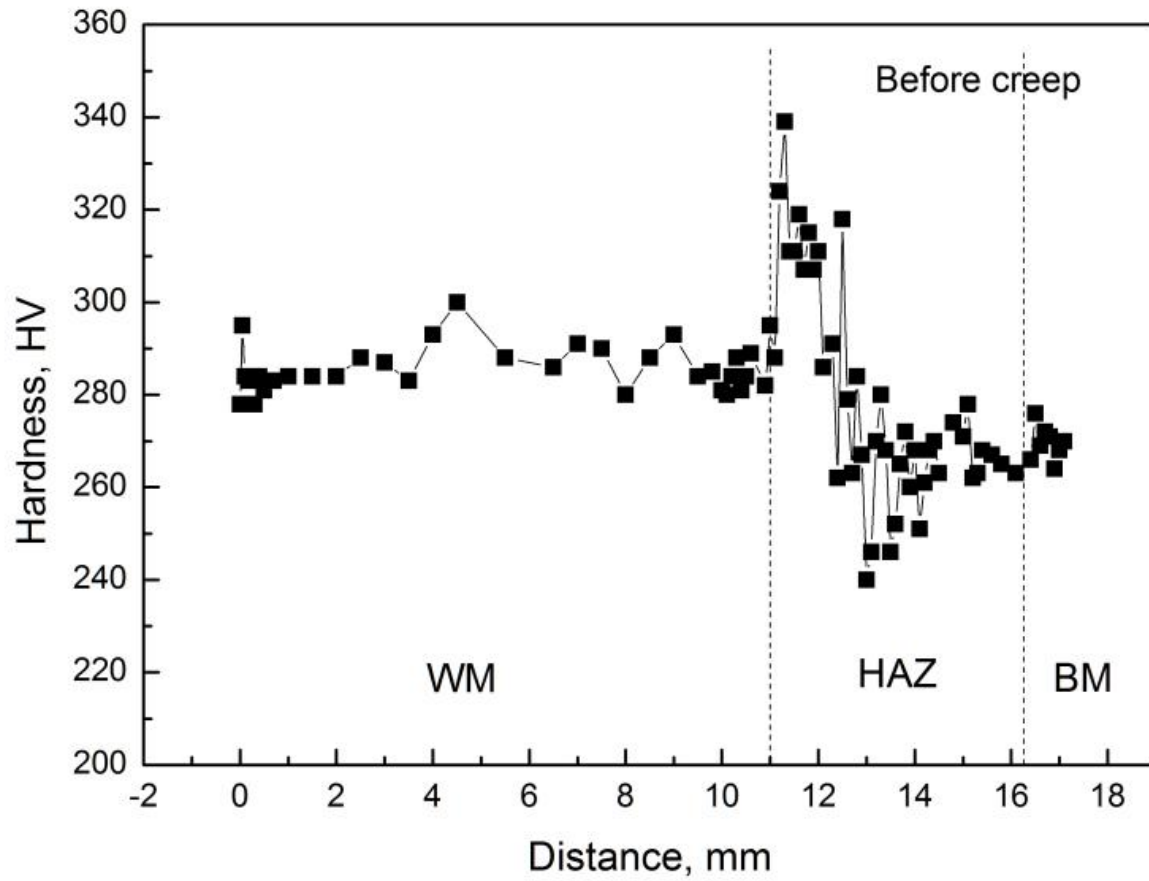


Figure 6

Hardness distribution of the MarBN steel welded joint after PWHT

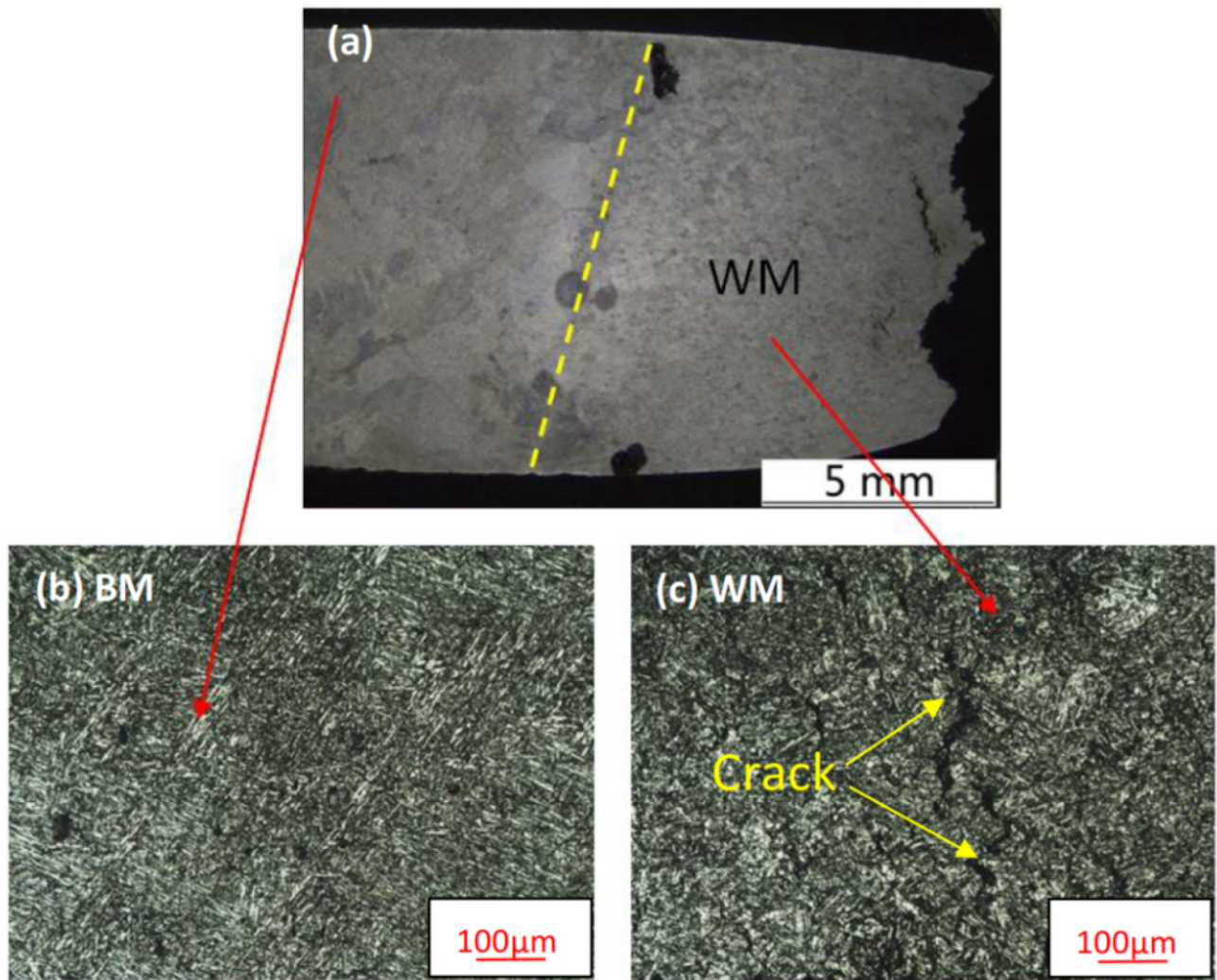


Figure 7

OM structure of the fractured MarBN steel welded joint sample under 120 MPa/12596.7 h

(a) Welded joint, (b) BM, (c) WM

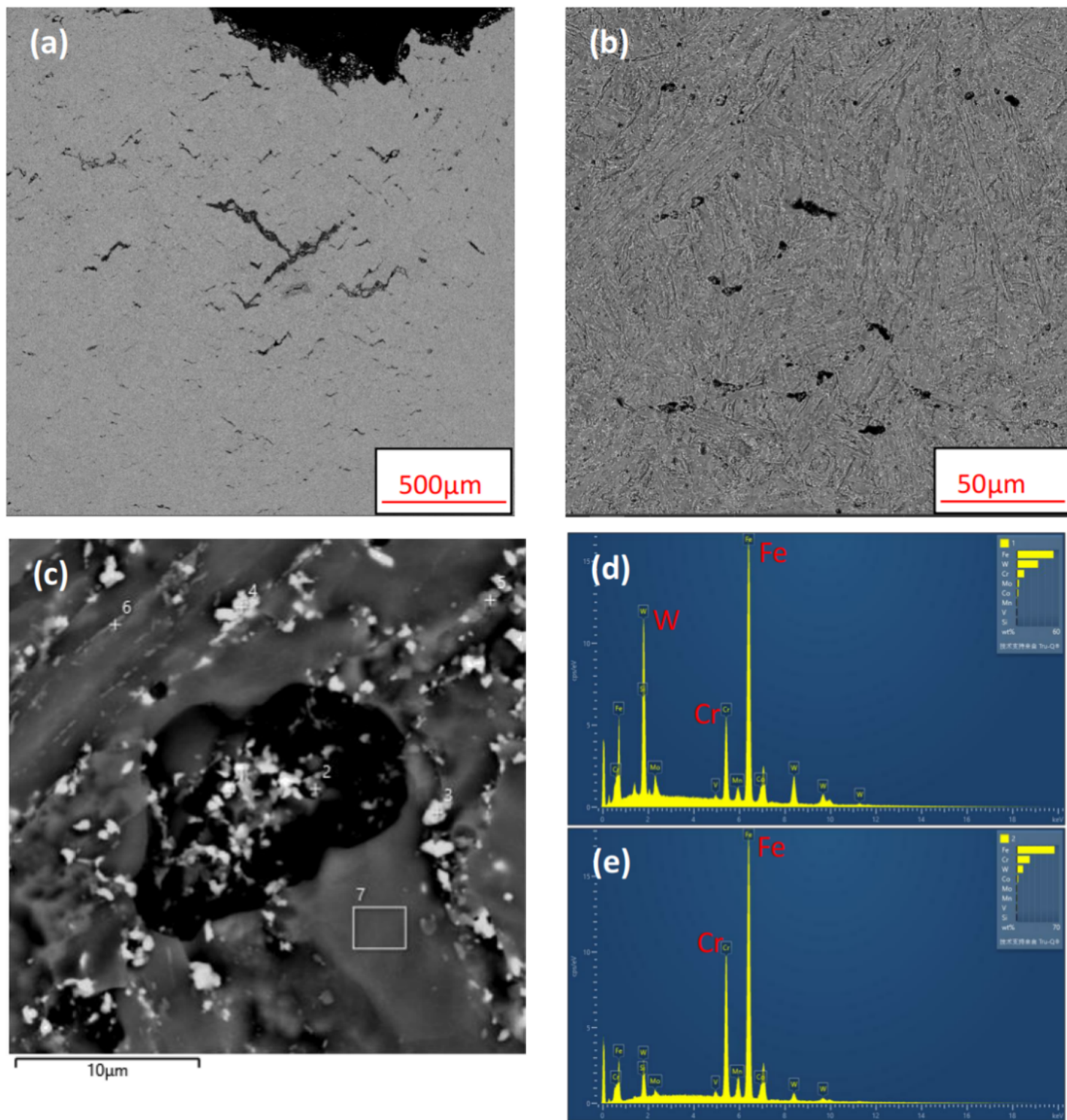


Figure 8

SEM structure of the fractured MarBN steel welded jointsample under 120 MPa/12596.7 h

(a) Micro-cracks near the fracture, (b) Creep cavities, (c) Precipitates inside creep cavity,

(d) EDS of white precipitates, (e) EDS of gray precipitates

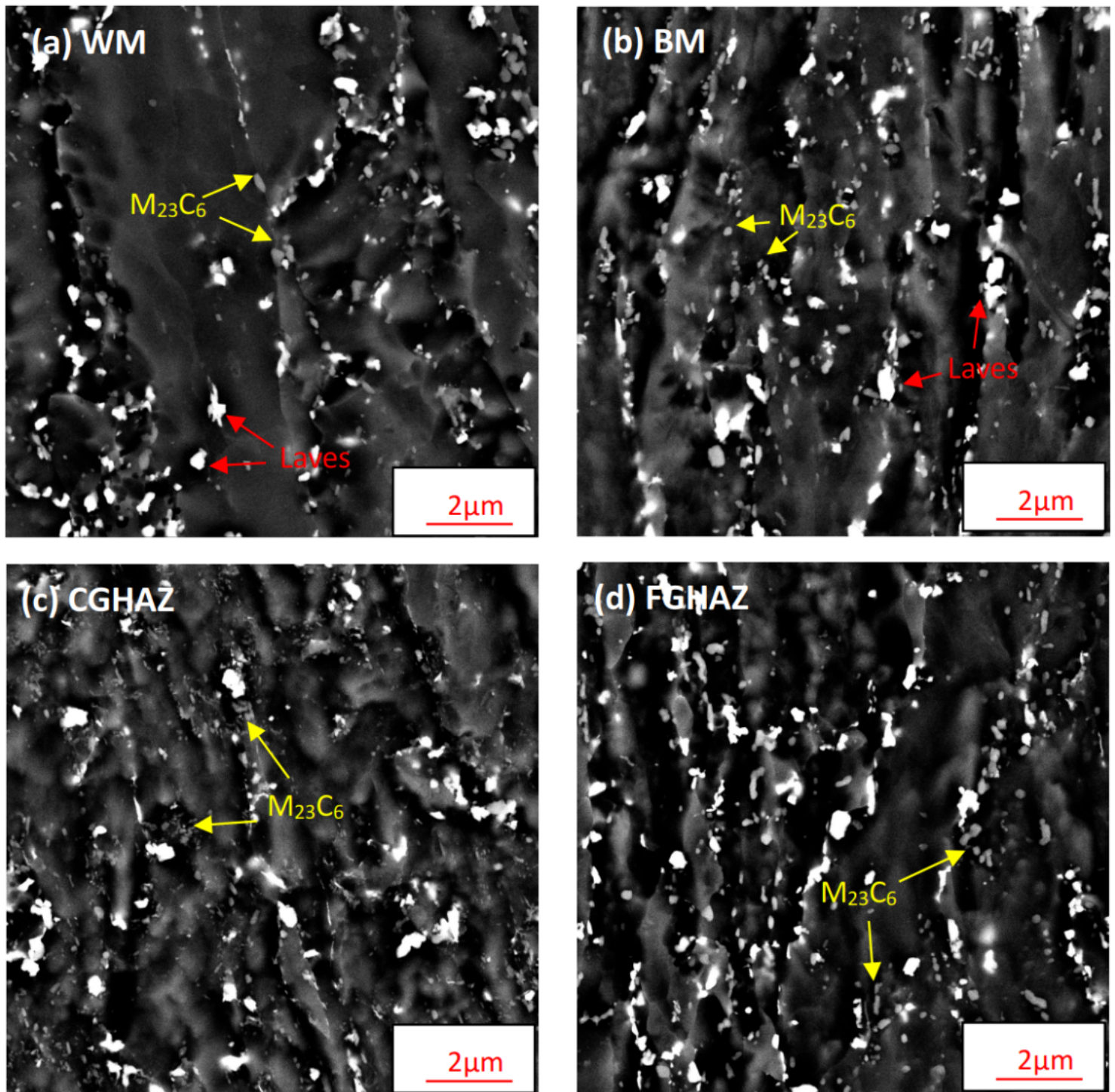


Figure 9

SEM pictures of precipitates in different regions of the MarBN steel welded joint after creep (120 MPa/12596.7 h)

(a) WM, (b) BM, (c) CGHAZ, (d) FGHAZ

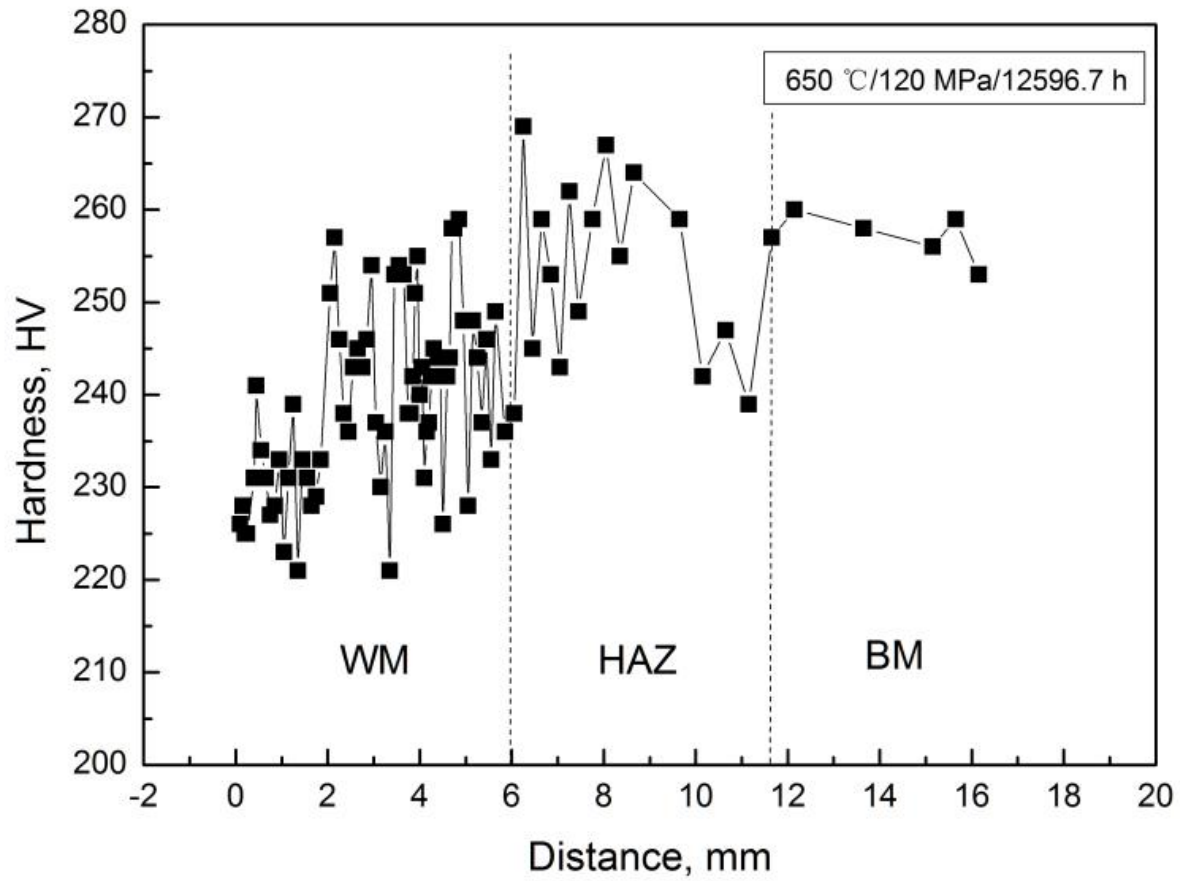


Figure 10

Hardness distribution of the MarBN steel welded joint after creep

(650 °C/120 MPa/12596.7 h)