Effect of Aging Heat Treatment on the Microstructure of Gas Tungsten Arc-Welded Inconel Superalloy 718

G. Ravi, N. Murugan, R. Arulmani and Gopal Magadi

Abstract The precipitation hardenable nickel-based superalloys are widely used in aerospace, power plants, oil, and gas industries due to their excellent high-temperature strength and remarkable hot corrosion resistance. Though it is weldable, the required properties such as hardness, resistance to fracture, corrosion, etc., depends upon the weld microstructure which is to be obtained by various heat treatment methods. In the present investigation, the influence of aging heat treatment on gas tungsten arc-welded microstructure of alloy 718 was studied. The aging heat treatment was carried out at 780 °C for 1 h, 4.5 h and 8 h and air cooled to enhance the properties of the weld metal. After heat treatment, the welded samples were characterized using an optical microscope (OM), field emission scanning electron microscope (FESEM), transmission electron microscopy (TEM), and microhardness tester. The effect of heat treatment time on the microstructure was evaluated. Gamma prime (γ'), gamma double prime (γ'') and the primary carbides of MC, $M_{23}C_6$ were observed in the HAZ and grain boundaries. It was observed that with the increase in aging time microhardness was significantly increased. The mechanical bend test was carried out to evaluate the ductility of the weld metal in the heat-treated conditions. On aging at 780 °C for 8 h, the ductility of the weld metal was found to be good and the microstructure depicted the reduction of laves phases and NbC in weld fusion zone and higher amount of γ'' resulting is higher strength and hardness.

Keywords Inconel 718 · Gas tungsten arc welding (GTAW) · Aging heat treatment · Microstructure and vickers microhardness (HV)

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1 Introduction

Nickel-based superalloy alloy 718 is mostly used in power and valve industries owing to the excellent corrosion and wear resistance at high temperatures. It also exhibits good resistance to stress corrosion cracking and strain age cracking. It is mainly strengthened by gamma prime γ' Ni (Al, Ti), and gamma double prime γ'' (Ni_3Nb) precipitates which are formed at aging temperature range of 650–850 °C on the aging heat treatment process [1–6]. Therefore the aging temperature should not exceed 810 °C to obtain optimum strength in Ni–Fe based alloy [5]. The amount of γ'' precipitate formed at a temperature range of 700–850 °C is 4 times higher than that of γ' phase resulting in enhanced strength [7, 8]. On increasing the aging time at above 850 °C aging temperature, γ'' will be changed into δ phase, which will reduce mechanical strength and microhardness [6]; above this temperature, the precipitates are dissolved in the material.

Numerous types of welding processes are applied to join nickel-based superalloy namely, gas tungsten arc welding (GTAW), gas metal arc welding (GMAW), laser beam welding, electron beam welding, etc. Laser and electron beam welding processes are used to join Inconel 718 alloys especially in the aerospace and automotive industries to have reduced heat input, distortion, and heat-affected zone (HAZ) area [9]. Most of the works reported were on studies on microstructures, weld defects, and mechanical properties of Inconel 718 alloy [10–12]. It was reported that use of pulsed current (PC) in gas tungsten arc welding (GTAW) of solution treated and aged Inconel 718 alloy resulted in reduction of % of volume fraction of laves phases and Nb segregation in the weld metal causing improved mechanical properties compared to that joined by the constant current (CC) GTAW [13]. The amounts of Nb segregation and laves phases in the weld metal were less in electron beam welded 718 alloys compared to that of GTA welded after the cyclic post-weld heat treatment (PWHT) process [14].

It can be summarized that most of the research work carried out on the aging heat treatment process was focused on analyzing the mechanical properties or microstructures at the temperature range of 500–750 °C, aiming to increase strength in the high-temperature alloy. To the best of the knowledge of the authors, no broad investigation on the effect of aging heat treatment at a temperature of 780 °C at different aging timings have not yet reported in GTA-welded Ni-based superalloy Inconel 718 in aged condition. In the present research work, 6 mm thick Inconel 718 plates were welded by gas tungsten arc welding and aged at different aging timings. The effect of aging heat treatment on the metallurgical characterization and mechanical properties of GTA welded Inconel 718 has been studied.

2 Experimental Procedure

2.1 *Material*

Inconel superalloy 718 received in the aged condition in the form of a circular disc of diameter 250 mm and 6 mm thickness was used as the base material. Three plates having a width of 70 mm were prepared from the base disc and in each plate, a V-groove with 70° groove angle and depth of 5 mm was machined by wire cut EDM process. Alloy 718 was deposited in the V-grooves by gas tungsten arc welding (GTAW) process using a filler rod ERNiFeCr-2 of size 2.4 mm diameter and length of 1 m. The chemical composition of the base metal and filler rod and the GTAW process parameters are presented in Tables 1 and 2, respectively. The same GTAW process parameter was used to deposit weld metal on the V-Grooves of three plates using pure argon shielding and the welded plates are shown in Fig. 1. The quality of the welded plates was tested with 100% X-ray radiography test and no defects were found.

Table 1 Chemical composition of the base metal and the filler metal (in wt%)

Elements	Ni+Co	Cr	Fe	Nb+Ta	Mo	Ti	Al	Cu	Si
Base metal	52.94+0.15	18.12	Bal	4.87	2.92	0.95	0.34	0.063	0.20
Filler rod	53.56	18.65	17.9	5.05	2.95	0.97	0.59	0.08	0.001

Table 2 The GTA welding process parameter

Sl. no	Current (A)	Voltage (V)	Welding speed (cm/min)	Shielding gas flow rate (lit/min)	Purging gas flow rate (lit/min)
1	150	14	10	12	5

Fig. 1 GTA-welded Inconel 718 plates
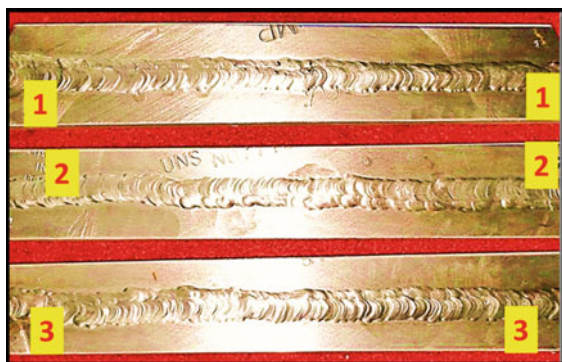
2.2 Heat Treatment

The heat treatment setup consisted of an electrical muffle furnace for heating to a maximum temperature of 1500 °C. The time and temperature were controlled by a Eurotherm controller and R-type thermocouples. The three GTA welded plates were heated up to a temperature of 780 °C as recommended by Special Metal Corporation [15] and the first plate (S_1) was held for 1 h, the second plate (S_2) for 4.5 h, and third plate (S_3) for 8 h, and all the plates were separately cooled to room temperature in air. The heat treatment cycle adopted for the three plates is shown in Fig. 2.

2.3 Metallographic Examination

After heat treatment, the welded plates were cut transverse to the welding direction and samples of size 15 mm $\times$ 10 mm $\times$ 6 mm were prepared from each plate for macro and microstructural analysis. The cut samples were prepared as per the standard metallographic procedure using different abrasive grade papers (from 400 to 1500 SiC) and diamond paste up to 1 μ to obtain mirror finish surface. The samples were etched with a Super Kalling's (5 ml H_2O_2 + 60 ml HCL + 6 g $CuCl_2$) etchant to reveal the microstructure under optical microscope (OM), field emission and scanning electron microscope (FESEM) and to identify various elements present in the phases using EDS. The etched sample of the welded specimen was presented in Fig. 3. Samples were prepared as per standard procedure from each welded plates for transmission electron microscopy (TEM) analyses.

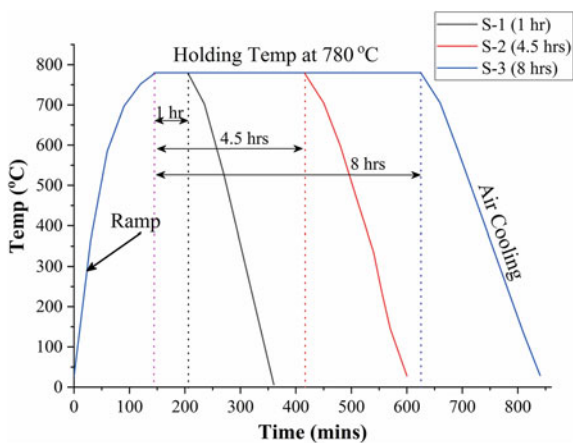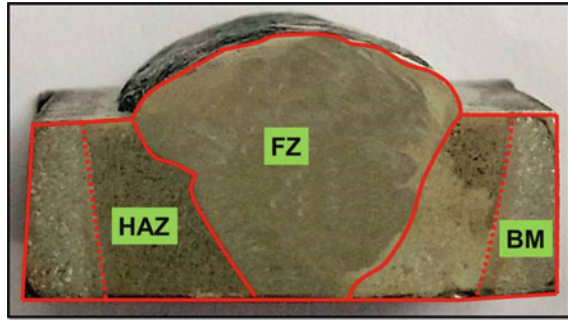**Fig. 2** Heat treatment cycles applied for welded plates

Fig. 3 Typical weld cross section of Inconel 718

3 Result and Discussion

3.1 Microstructure of Aged GTA Welded Inconel 718 Alloy

After aging heat treatment, the microstructures of heat-affected zone (HAZ) and weld metals were obtained for the three samples using optical microscope and presented in Fig. 4. From the figure, it is found that the grains are coarser when the holding time increases. The measured average grain sizes of HAZ of the heat-treated samples were ASTM #5.5, #4.5, and #4, respectively for S_1, S_2, and S_3. There were no microfissures on HAZ and fusion line in GTA welded plates. The average volume fractions of laves phases present in dendritic and interdendritic regions were calculated with the help of Image processing software Metal Plus and presented in Table 3. It is also observed from Fig. 4 that laves particles are finer with the increase in aging time at the aging temperature of 780 °C.
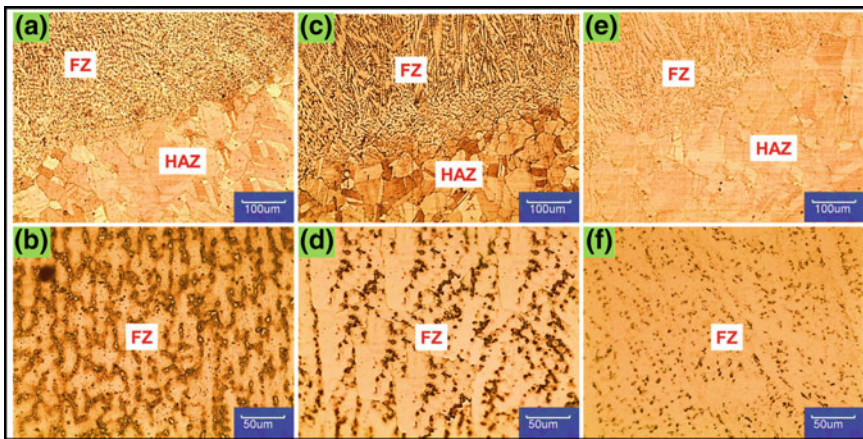

Fig. 4 Optical microstructures of fusion zone (FZ) and HAZ of heat-treated alloy 718: **a**, **b**—S_1 (1 h), **c**, **d**—S_2 (4.5 h) and **e**, **f**—S_3 (8 h)

Table 3 Estimation of volume fraction in weld fusion zone of heat-treated samples

Samples	Avg. laves particles (% of volume fraction)
S$_1$—1 h	22.5 ± 5
S$_2$—4 h 30 min	20.1 ± 2
S$_3$—8 h	18.5 ± 1

The microstructures of the heat-treated samples were observed under field emission scanning electron microscope (FESEM) to identify the phases like laves phases, Niobium, and gamma matrix in the weld fusion zone. Figure 5 shows the presence of laves particles in the weld fusion zone aged at different aging time. It was reported that the direct aging (DA) heat treatment had not produced any change in laves phase and Nb segregation in weld zone, but only on post-weld solution treatment (PWST) at above 980 °C laves phase and Nb segregation can dissolve not completely. In some case of PWST at a range of (860–955 °C), the δ precipitates were formed, after 955 °C solution treated and aging (STA) treatment is required to decrease the amount of δ phase precipitates by a cyclic treatment [12, 13]. The Nb segregation and laves particles are relatively less in sample S$_3$ compared to S$_2$ and S$_1$ samples. However, the Nb is a highly refractory element, which will segregate in the interdendritic region during the solidification process and undesirable phases like laves can form only in welding. Laves is a hexagonally close-packed phase (HCP) formed in the weld metals due to microsegregation of alloying elements like Nb, Ti, Mo, etc. The MC carbides are irregular with blocky morphology and located in the interdendritic region as shown in Fig. 5a, b. It was reported that the morphologies of M$_{23}$C$_6$ carbides were distributed evenly at grain boundaries, on increasing aging temperature and the amounts of M$_{23}$C$_6$ carbides were decreased and appeared nonuniformly at a temperature of 810–840 °C. γ′ phases are coarser with spherical morphology with increasing temperature and holding time and the γ′ particles were evenly distributed resulting in increasing the microhardness at a temperature of 810 °C [5]. The γ″ phase being the strengthening precipitate of alloy 718 has disc-like shaped coherent with γ′ precipitates. It was reported that the

Fig. 5 FESEM microstructures of fusion zone of heat-treated alloy 718: **a**—S$_1$ (1 h), **b**— S$_2$ (4.5 h) and **c**—S$_3$ (8 h)

respective volume fraction of the γ′ and γ″ precipitates and the clustering of Al/Ti and Nb suggested that γ″ nucleated prior to γ′ during aging at 706 °C for this alloy [3].

EDS spectrums analyses were performed at different points in the weld fusion zone in all three aged samples to identify laves phases, Niobium carbide (NbC) and gamma matrix and typical EDS spectrum and the corresponding elemental analysis are presented respectively in Fig. 6 and Table 4. The niobium segregation and laves phases present in the weld metal strongly influenced to reduce the strength of the GTA welded alloy compared to that of the EB and laser welded alloy 718 [12]. They are present in S_2 and S_3 samples less than S_1 leading to more γ″ in the weld fusion zone of the heat-treated alloy 718.

Figure 7 illustrates higher magnification TEM micrographs showing the γ″ precipitates present in both the heat-treated samples S_2 and S_3. However, it is found that on aging heat treatment at a temperature of 750 °C for 4 h the γ″ precipitates were formed on the γ' precipitates and increased in hardness [11]. It is reported that on longer duration of aging time for longer thermal exposure temperature from (700 to 900 °C) range, the γ″ phase seems to transform to the elongated structure of delta phases (δ) [2, 6].

Where the gamma double prime (γ″) precipitate increases significantly the hardness also increases and improves the strength of the weld metals as shown in Fig. 8b, c. The hardness is increased in sample S_2 and S_3 due to the presence of gamma double prime (γ″) precipitates in alloy 718 after the aging heat treatment.
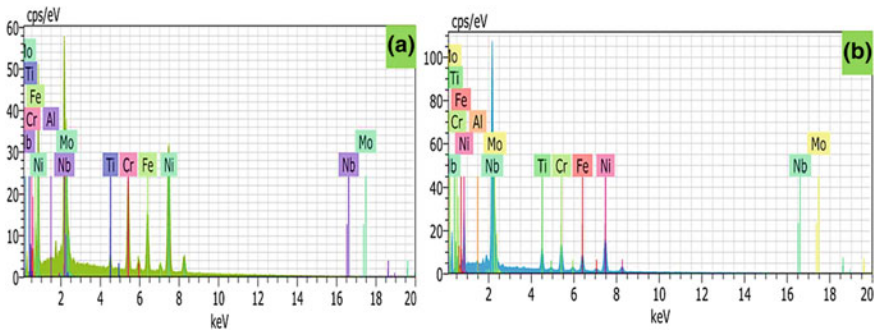

Fig. 6 Typical EDS spectrum of EDS analysis of GTA weld **a** Laves phases, **b** NbC

Table 4 Composition of phases obtained by EDS analysis of GTA-welded Inconel alloy 718

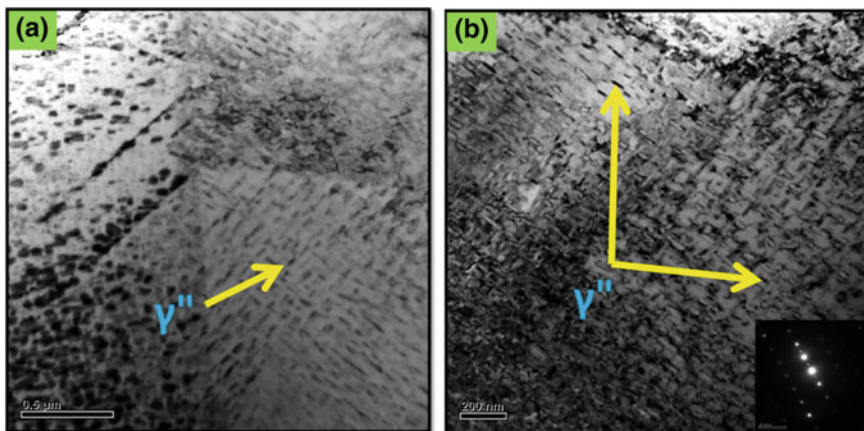| Spectrum	Elements (wt%)								
	Ni	Cr	Fe	Nb	Mo	Ti	Al	Si	C
Matrix	53.25	18.50	19.61	4.35	2.65	1.01	0.95	0.05	0.08
Laves	34.39	11.74	9.54	19.80	5.53	1.45	0.03	–	–
NbC	10.55	4.14	3.38	59.45	0.57	4.57	0.30	0.10	16.8
Dendrite	49.95	18.69	18.85	4.28	2.95	1.01	1.59	0.08	1.6

Fig. 7 TEM microstructures of fusion zone of heat-treated alloy 718: arrowhead showing the γ'' precipitates in samples: **a**—S_2 (4.5 h) and **b**—S_3 (8 h)

On sample S_2 the presence of γ'' precipitates are lower than that of sample S_3, where the aging time increases the amount of γ'' precipitates will increase and increase in the hardness values. From the result, the main strengthening phase of γ'' precipitates will occur at a temperature of 780 °C in short duration time of 4–8 h, respectively.

4 Mechanical Testing

4.1 Vickers Microhardness

Microhardness measurements were taken in the transverses direction i.e. perpendicular to the welding direction in heat-treated samples S_1, S_2, and S_3 and the hardness profile is shown in Fig. 8. Table 5 presents the average microhardness values of three aged samples and the base metal microhardness was approximately

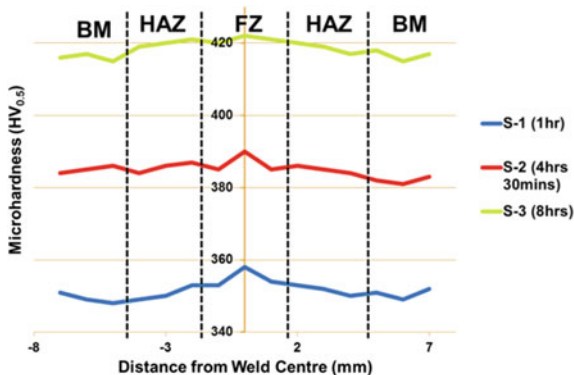**Fig. 8** Microhardness profile of PWH treated samples (S_1–S_3)

Table 5 Average Vickers microhardness values

Aging heat treatment timings	Fusion zone (FZ)	Heat-affected zone (HAZ)	Base metal (BM)
S_1—1 h	358 ± 4	355 ± 3	345 ± 2
S_2—4 h 30 min	390 ± 3	386 ± 4	380 ± 2
S_3—8 h	421 ± 3	418 ± 3	412 ± 1

378 HV received in aged condition. From the table, it is observed that microhardness of all the zones tends to increase with the aging time. The maximum hardness values attained in the S_3 sample is 421 ± 3, where the aging time increases simultaneously the amount of γ'' precipitates also increases. Also, it is found that the γ'' precipitates are not formed in the weld fusion zone. It was reported that with the increase in aging temperature the hardness decreases with the increasing aging time [5]. It is clear that the weld metal exhibits higher hardness in the weld fusion zone at a temperature of 780 °C for an aging time of 8 h.

4.2 Face Bend Test

Face bend test results are shown in Fig. 9. Cracks were observed in S_1 sample, which is due to the presence of large no. of Nb segregation and laves phases in the weld fusion zone that reduce the ductility of the metal as shown in Fig. 9a. After aging heat treatment of alloy 718, the more amount of γ'' precipitates present in S_2 and S_3 increased the hardness and ductility of the weld fusion zone and passed the bend test as shown in Fig. 8b, c. The ductility may be due to less amount of laves phase and NbC and less segregation of Nb on aging at the higher aging time.
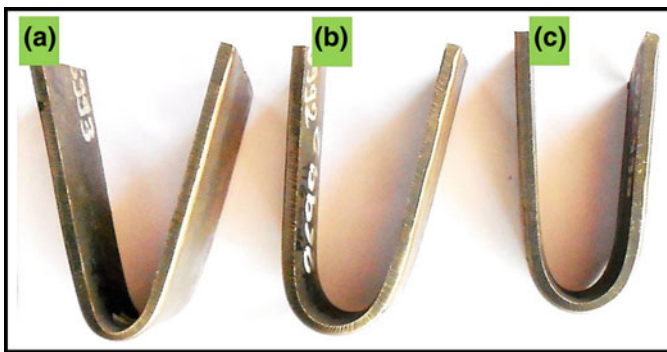

Fig. 9 Mechanical face bent samples: **a**—S_1 (1 h) **b**—S_2 (4.5 h) and **c**—S_3 (8 h)

5 Conclusion

In the present research work, 6 mm thick nickel alloy Inconel 718 was welded by GTAW and post-weld aging heat treatment was carried out at 780 °C for different aging time and the effect of aging time on microstructures on weld metal and HAZ was investigated. The following conclusions are drawn:

1. Higher the aging time up to 8 h at 780 °C, higher the microhardness and finer the weld metal microstructure resulting in adequate ductility. Higher the aging time up to 8 h less is Nb segregation and amount of laves particles in the weld metal resulting enhanced amount of γ'' precipitates leading to higher strength.
2. At 780 °C aging temperature, more γ'' precipitates occured in weld metal at 4.5 h and 8 h aging time and the microhardness was found to be (390 HV) and (421 HV) respectively compared to the base metal.

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