Thermal Aging Effects on Heat Affected Zone of Alloy 600 in Dissimilar Metal Weld

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1. Introduction

Dissimilar metal weld (DMW), consists of Alloy 600, Alloy 182, and A508 Gr.3, is now being widely used as the reactor pressure vessel penetration nozzle and the steam generator tubing material for pressurized water reactors (PWR) because of its mechanical property, thermal expansion coefficient, and corrosion resistance [1].

Among several key components in primary circuit, primary water stress corrosion cracking (PWSCC) was first observed in the early 1990s at the vessel head penetration weldment which is fabricated from DMW. After that accident PWSCC has become a generic problem for DMW components in PWR plants [2].

Especially, the heat affected zone (HAZ) on Alloy 600 which is formed by welding process is critical to crack. According to G.A. Young et al. crack growth rates (CGR) in the Alloy 600 HAZ were about 30 times faster than those in the Alloy 600 base metal tested under the same conditions [3]. And according to Z.P. Lu et al. CGR in the Alloy 600 HAZ can be more than 20 times higher than that in its base metal [3].

To predict the life time of components, there is a model which can calculate the effective degradation years (EDYs) of the material as a function of operating temperature [4]. However, even though EDYs were low enough to safe, some problems occurred on the material such as PWSCC. That means some other factors which are not included in this model should be considered, for example long-term thermal aging effect.

The main goal of this experiment is to investigate how the long-term thermal aging affects the mechanical properties of Alloy 600 HAZ in point of view on Vickers hardness and residual strain.

2. Experimental

2.1 Materials and specimens

By joining Alloy 600 and A508 Gr. 3 with Alloy 182 serving as dissimilar filler metal, representative mockup sample was fabricated with welding process which was qualified by the American Society of Mechanical Engineers (ASME) [5]. The chemical compositions of each metal is shown in Table 1.

The heat treatment condition for thermal aging simulation was determined based on diffusion equation with activation energy of chromium through grain boundary. Since chromium is known to do important role to material properties. And also chromium precipitate affects the property of material a lot [6].

Heat treatment simulates the aged DMW which is used during 15 years and 30 years in the commercial nuclear power plant (operating at temperature about 320℃). Both aging temperature and aging time are calculated under the diffusion equation in Figure 1.

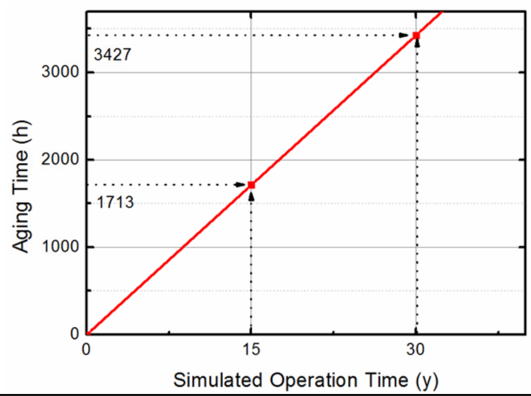
$$\frac{t_{aging}}{t_{ref}} = exp\left[-\frac{Q\left(\frac{1}{T_{ref}} - \frac{1}{T_{aging}}\right)}{R}\right]$$

t_{aging} = Aging time [hr]
t_{ref} = Simulated operation time [hr]
T_{aging} = Aging temperature [K]
T_{ref} = Simulated operation temperature (320°C) [K]
R = Gas constant [kJ/mol]
Q = Activation energy for Cr diffusion
(about 180kJ/mol)

Figure. 1. Diffusion equation to calculate the relation between temperature and time of reference case and aging case.

Table. 1. Chemical composition (in wt.%) of dissimilar metal used in this experiment

Material	Composition									
	C	Si	Mn	P	S	Fe	Cu	Ni	Cr	etc.
Alloy 600	0.06	0.30	0.07	-	0.001	8.13	0.01	74.9	15.6	-
Alloy 182	0.05	0.37	7.42	0.01	0.01	4.48	0.01	70.9	14.9	1.77Nb+Ta 0.03Ti
A508 Gr.3	0.19	0.22	1.33	0.008	0.002	Bal.	0.02	0.91	0.19	0.47Mo, 0.02Al 0.003V



Specimen I.D.	t_{ref}	t_{aging}, T_{aging}
as-welded	-	-
HT400_Y15	15-yr	1713-hr at 400℃
HT400_Y30	30-yr	3427-hr at 400℃

Figure. 2. Aging time at the temperature 400℃ to simulate 15 years and 30 years at the temperature 320℃ in nuclear power plant respectively.

The heat treatment process is described in Figure 2. Heat treatment temperature was selected as 400°C. It can be said that higher temperature may reduce the simulating time, however, when heat treating temperature goes above 450°C, some un-wanted phase can occur. Therefore, 400°C was chosen in this experiment.

According to the abovementioned values for the diffusion equation, the finally calculated aging time for 15 years at 320℃ is 1713 hours at 400℃ and the aging time for 30 years at 320℃ is 3427 hours at 400℃.

2.2 Experimental procedure

The specimens cut from the as-welded mockup sample were polished with emery paper of 320grits, 600grits, and 800 grits, diamond paste of 6 µm, 3 µm, and 1 µm. And finally vibration polishing was conducted with 0.05 µm colloidal silica/alumina solution during 48 hours to minimize mechanical deformation on the surface.

After the polishing process, the specimens were washed with acetone in 2 minutes, ethanol in 2 minutes, and distilled water in 20 seconds.

The hardness of each specimen was measured by Vickers hardness tester using a load of HV 0.3 (i.e. 1 kg/mm^2). Hardness measurements were conducted from Alloy 182 to Alloy 600.

The residual strain of each specimen at the Alloy 600 HAZ was observed by kernel average misorientation (KAM) mapping, a function of electron backscattering diffraction (EBSD) with step size 2µm.

3. Results and Discussion

3.1 Hardness

Figure 3 shows the distribution of Vickers hardness measured along the two parallel lines. As each graph shows, there is an obvious tendency between the distance from the fusion boundary and Vickers hardness. The hardness peaked near the fusion boundary and also high through the HAZ. And then it decreased as distance from the fusion boundary. This result coincides with preceding research conducted by S. Yamazaki et al. [2]

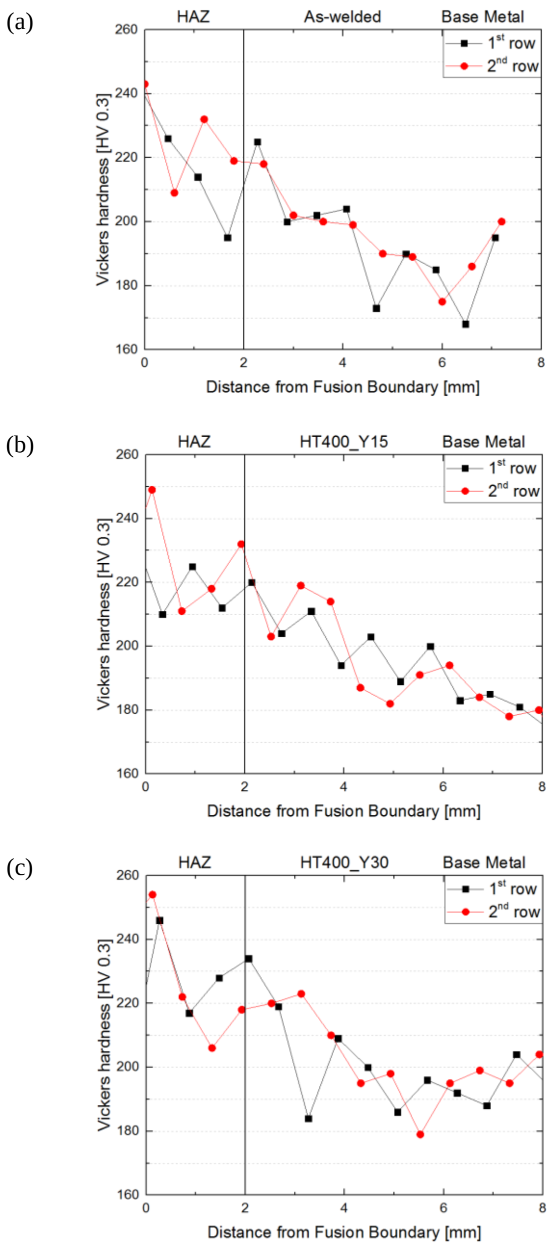


Figure. 3. Vickers hardness result from the fusion boundary between Alloy 182 and Alloy 600 to the Alloy 600 base metal. Each graph shows the result of (a) as-welded, (b) HT400_Y15, and (c) HT400_Y30 case respectively.

As is the case of precipitation-hardening process, when the metal is heated above the solvus temperature, such as welding temperature, grain boundary carbides dissolves. And dissolved precipitates are not formed because of the process quenching. As figure 4 (a), the SEM image of HAZ region of as-welded specimen, shows, the number of precipitates on grain boundary is low [7].

(a)
(b)
(c)

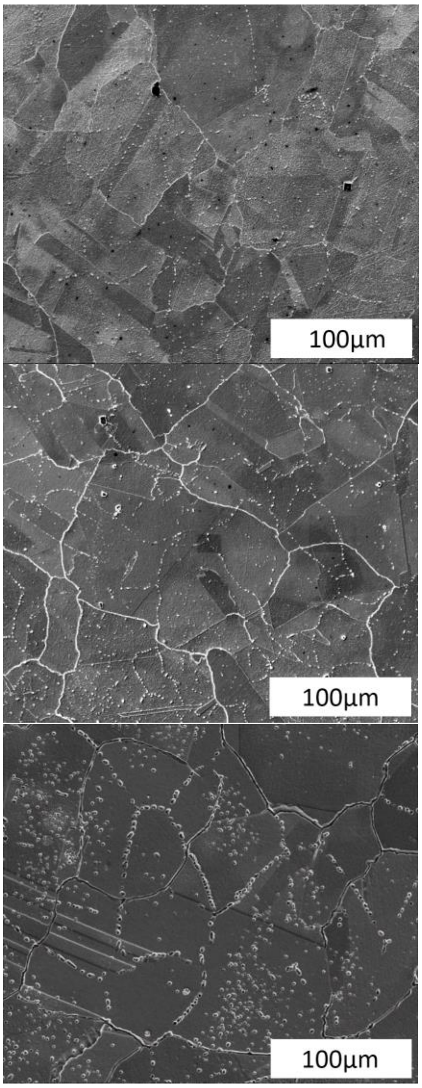


Figure. 4. Distribution of Cr precipitates on the surface of (a) as-welded, (b) HT400_Y15, and (c) HT400_Y30 specimen respectively pictured by scanning electron microscopy (SEM).

According to S.C. Yoo et al [8] chromium ion moves along grain boundary when the material heat treated. This migration of chromium can form the precipitates, chromium carbide such as Cr_7C_3 and $Cr_{23}C_6$ since it is stable phase when certain portion of chromium ion exists as the form of precipitate [8-9].

According to Figure 4 the number of Cr precipitates increases after thermal aging, therefore, it was expected that the hardness will increase after heat treated. However, even though the material was undergone heat treatment, there is no change among the specimens in point of the value of hardness and also the tendency of hardness.

This unexpected result may be caused by the change of residual stress. There is a process after welding named post weld heat treatment (PWHT) which reduces the residual stress formed by welding. Since residual stress can degrade the mechanical property of the material. Likewise, the thermal aging can also do the role of PWHT. To find out this phenomenon the kernel average misorientation (KAM) mapping was conducted.

3.2 Kernel average misorientation

According to Z. Lu et al. KAM value at the Alloy 600 HAZ is higher comparing with that of Alloy 600 base metal [10]. Furthermore, C. Ma et al. observed that a peak of KAM value at all grain boundaries is approximately 3 times higher than that of inside of the grain [11]. Summarizing these researches, the grain boundaries at Alloy 600 HAZ is the most vulnerable to the effect of residual strain.

(a)
(b)
(c)

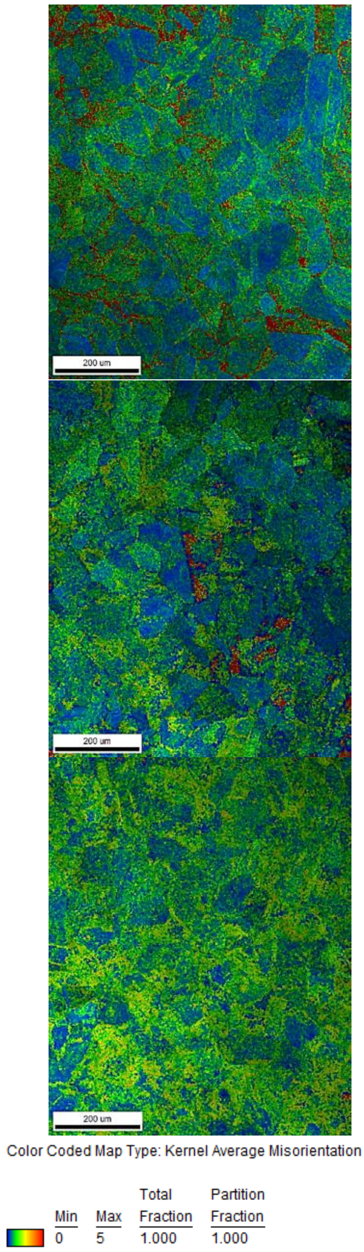


Figure. 5. KAM mapping on Alloy 600 HAZ distance from the fusion boundary about 700µm. Above three pictures show KAM distribution of (a) as-welded, (b) HT400_Y15, and (c) HT400_Y30 specimens.

According to the preceding researches [12-13], it was explained that post-weld heat treatment (PWHT) affects residual stress distribution on the material. When the heat applied to the material, residual stress relaxation occurs since heat can deform the material by decreasing yield strength of it [13]. This phenomenon is happened similarly when the material undergoes the thermal aging process.

Figure 5 (a) shows clearly that a lot of red dots are pointed at the grain boundaries. This means high residual strain exists at the grain boundaries of Alloy 600 HAZ. Figure 5 (b) and Figure 5 (c) show the distribution of KAM of 15 years simulated specimen and 30 years simulated specimen respectively. Comparing with Figure 5 (a), many red dots disappeared in Figure 5 (b). Moreover, in Figure 5 (c) it is hard to find the red dot from the picture. Therefore, it is obvious that thermal aging reduces residual stress, and this also affects hardness tendency of the material.

4. Conclusions

This study was conducted to investigate how thermal aging affects the hardness of dissimilar metal weld from the fusion boundary to Alloy 600 base metal and the residual strain at Alloy 600 heat affected zone. Following conclusions can be drawn from this study.

The hardness, measured by Vickers hardness tester, peaked near the fusion boundary between Alloy 182 and Alloy 600, and it decreases as the picked point goes to Alloy 600 base metal.

Even though the formation of precipitate such as Cr carbide, thermal aging doesn't affect the value and the tendency of hardness because of reduced residual stress.

According to kernel average misorientation mapping, residual strain decreases when the material thermally aged. And finally, in 30 years simulated specimen, the high residual strain almost disappears.

Therefore, the influence of residual strain on primary water stress corrosion cracking can be diminished when the material undergoes thermal aging.

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