Mechanism of Stress Relief Cracking in a Granular Bainitic Steel

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Abstract Coarse-grained heat-affected zone (CGHAZ) of a low alloyed, granular bainitic steel T24 was simulated in a Gleeble apparatus. The stress relief of the CGHAZ was analyzed by annealing the samples. The morphology and behavior of the microstructure near the grain boundaries during stress relief were investigated by means of focused ion beam, in situ tensile testing, transmission electron microscopy, scanning electron microscopy and electron back-scatter diffraction. It was observed that there were large martensite/austenite islands distributed along the grain boundaries of CGHAZ. During stress relief at elevated temperature, the retained austenite at the grain boundaries decomposed into M_3C carbides and a ferrite forming softening zone. Together with the stress relief, piled up of dislocations occurred within the ferrite in the area adjacent to the ferrite/ M_3C interface, which resulted in high level of stress accumulation and caused crack initiation along the grain boundaries. These results indicate that the stress relief induced the grain boundary crack is controlled by other mechanisms rather than the creep-like grain boundary sliding.

KEY WORDS: Granular bainitic steel; Heat-affected zone; Stress relief cracking; Martensite/austenite island; Behavior of grain boundary

1 Introduction

Stress-relief cracking (SRC) is a major cause of weld failures in creep resistant, precipitation strengthened materials, such as ferritic steels, stainless steels, and Ni-based super alloys [1]. The cracking usually occurs in the heat-affected zone, which is the most susceptible area for crack initiation, due to solid-state metallurgical reactions during both heating and cooling thermal cycles [2]. SRC

and premature cracking in these materials are the types of intergranular cracking that occur following postweld heat treatment and during exposure to elevated temperatures ($\sim 550^\circ\text{C}$).

To date, many investigations have been performed to understand the SRC mechanisms. Typically, there are two well-known cracking mechanisms. The first is the formation of a precipitation free zone, resulting in local weakening and strain localization. The cracking mechanism was explained by the combination of a precipitation strengthened matrix and a soft precipitation free zone, formed adjacent to prior austenite grain boundaries [3–5]. The second is the creep-like grain boundary (GB) sliding that causes micro-void formation along the GBs and the triple point junctions. For example, the segregation of impurities to the prior austenite GBs may lower the cohesive strength of GBs and cause intergranular cracking through GB sliding [6–8]. It has also been reported that intergranular cracking takes place by the traditional creep failure

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mechanism in which cavities open along GBs and then coalesce to form a continuous crack [9–14].

Although many SRC mechanisms have been widely explored, the topic still remains debatable. There is still considerable work that needs to be done on boundary morphologies and corresponding behavior to avoid SRC problem. The alloy T24 is a type of 2.25Cr-1Mo granular bainitic steel and a promising candidate for applications in membrane water walls in ultra-supercritical (USC) coal-fired power plants [15]. However, T24 is susceptible to SRC, which limits its applications range. In this study, the mechanism of SRC in a T24 alloy was investigated through thermal simulation, stress relaxation testing and detailed microstructural characterization. The findings are beneficial to better understanding the SRC mechanisms.

2 Experimental

The commercial steel T24 was supplied in a form of tubes by Vallourec & Mannesmann (V&M), Germany. All the specimens were cut from the as-received tubes. The chemical composition of T24 is listed in Table 1. As-received T24 was heat treated as follows: austenitization at 1000 °C for 20 min, followed by air cooling; the subsequent tempering was conducted at 750 °C for 2 h.

The coarse-grained heat-affected zone (CGHAZ) simulation and stress relaxation testing were performed using Gleeble techniques. A Gleeble 3800 apparatus was used for these experiments. The peak temperature was set to be at 1350 °C and the cooling time ($t_{8/5}$) to 40 s. The curve of the temperature as a function of time during thermal simulation is presented in Fig. 1. For the specimen following the welding thermal simulation, the scratch lines of 20 μm in length across GB were marked using FEI NOVA 200 Nano Lab scanning electron microscope (SEM) with a focused ion beam (FIB), as observed in Fig. 2. Then, the specimen with scratch lines was used for compressive stress relaxation testing, as illustrated in Fig. 3. The time of stress relief was up to 100 min at 550 °C. Additionally, for studying the microstructural evolution, one sample was annealed at 550 °C for 400 h in a furnace. Hardness and microhardness Vickers of the specimens was performed using a load of 98 and 0.098 N, respectively, with a dwell time of 10 s in both cases.

Table 1 Chemical composition of T24 (wt%)

C	Si	Mn	P	S	Cr	Mo	V	Ni	N	Nb	Ti	B	Fe
0.06	0.33	0.68	0.012	0.0016	2.29	0.97	0.24	0.11	0.008	0.054	0.08	0.003	Bal.

Fig. 1 Curve of the temperature as a function of time during thermal simulation using a Gleeble 3800 apparatus

Fig. 2 SEM image showing the scratch lines of approximately 20 μm in length across GB, marked by a FEI NOVA 200 Nano Lab SEM/FIB

The specimen for in situ tensile testing was annealed at 650 °C for 10 h. The specimen had a thickness of 0.5 mm, as schematically demonstrated in Fig. 4. The in situ tensile testing was carried out in an FEI Quanta 400F SEM at room temperature, at a rate of $6.67 \mu\text{m s}^{-1}$. SEM and transmission electron microscopy (TEM) were employed for the observation on the microstructural features, especially the GBs. For electron back-scatter diffraction

Fig. 3 Plot of compressive stress as a function of time at 550 °C showing the stress relaxation

Fig. 4 Schematic drawing of the specimen, 0.5 mm in thickness, for in situ tensile testing

(EBSD) analysis, a boundary was defined as any two point pair that had a misorientation angle above 2°. TEM disks were mechanically ground to 100 μm and then perforated by a twin-jet electro-polisher at 40 V/18 mA and -30°C . The electrolyte consisted of 10% perchloric acid and 90% ethanol. TEM observations were conducted in a FEI Tecnai 20 microscope, operating at 200 kV.

3 Results and Discussion

3.1 Microstructures After Heat Treatment and Thermal Simulation

The microstructure of T24 following the heat treatment was that of the tempered bainite, as observed in Fig. 5. The matrix was comprised of lath bainitic ferrite (see Fig. 5a, b). $M_{23}C_6$ carbides were observed mainly at the GBs of prior austenite, as indicated by the arrows in Fig. 5b. The MX (MC type) precipitates were formed within the lathes and also along the lath boundaries. The observations are consistent with the results reported by previous studies [14–17].

The thermal simulation significantly affects the microstructure, as illustrated in Figs. 6 and 7. A number of martensite/austenite (M/A) islands were formed (see Fig. 6). The retained austenite was the major constituent of the M/A islands. The crystal structure of the ferrite and austenite is body-centered cubic and face-centered cubic,

Fig. 5 Initial microstructures of T24 following heat treatment. **a** Optical micrograph. **b** TEM micrograph. The MX precipitates at lath boundaries are indicated by arrows

respectively; thus, the two phases can be exclusively identified by using the EBSD technique. In Fig. 6, an EBSD image clearly shows the ferrite, marked by red color, and the austenite, marked by green color. The retained γ phase is primarily distributed along the lath boundaries and the prior austenite GBs. Also, the average size of the retained austenite became larger, approximately 200 nm, as observed in Fig. 7b. Finally, the significant

Fig. 6 EBSD image showing the microstructure of T24 following thermal simulation. The ferritic matrix (red color) and retained austenite (γ phase, green color) were primarily distributed along lath bainitic ferrite boundaries and the prior austenite GBs

grain growth occurred, due to the rapid heating of the specimen to 1350 °C.

3.2 Microstructural Evolution Near the Grain Boundaries After Stress Relief

Following stress relaxation testing at 550 °C for 30 min, the retained blocky austenite was observed to decompose into the ferrite and Fe_3C , as presented in Fig. 8. The newly formed ferritic phase was a softening zone with lower carbon content, where the dislocations were relatively easier to slip (Fig. 8b). Selecting the $(1\bar{1}2)$ reflection from austenite, the morphology of retained austenite is clearly illustrated in the dark field TEM image (see Fig. 8c). In Fig. 8d, the decomposed Fe_3C had an average size of approximately 60 nm.

After stress relaxation testing at 550 °C for 90 min, the retained austenite was continuously decomposed into ferrite and Fe_3C , as observed in Fig. 9. Compared to the specimen tested at 550 °C for 30 min, the decomposed carbides became larger, approximately 160 nm. The slipping of dislocations was blocked by the carbide at GB.

3.3 Vickers Hardness

As stated above, the simulated CGHAZ presented the microstructure of the granular bainite. The average macro Vickers hardness of CGHAZ was 366, and the microhardness of M/A islands at GB was 394, as illustrated in Fig. 10a. On the contrary, for the simulated CGHAZ specimen, following annealing at 550 °C for 400 h, the

Fig. 7 TEM micrographs of T24 following thermal simulation. **a** Bright field image showing the M/A islands at GBs. **b** Bright contrasted phases in dark field image are the retained austenite with a typical size of approximately 200 nm

Fig. 8 Microstructures of the specimen following stress relaxation testing at 550 °C for 30 min. **a** Bright field image, showing the ferrite and retained austenite. The white frame schematically indicates the area for electron diffraction pattern. **b** Enlarged image of the white frame of **a**, showing the morphology of Fe_3C at GB and dislocation slipping. **c** Corresponding dark field image of **a** showing the morphology of retained austenite. **d** Dark field image showing a Fe_3C particle at GB. **e** Overlapped electron diffraction pattern originated from the framed area in **a**. The austenite, ferrite and Fe_3C were indexed. The $(1\bar{1}2)$ reflection from the austenite was selected for taking dark field image **c**

average hardness of CGHAZ was 367, and the microhardness of M/A islands was 314, as presented in Fig. 10b. By comparing the two specimens, the hardness was maintained to be constant following 400 h annealing, and the M/A islands had a higher Vickers hardness than the overall hardness. However, the microhardness of M/A islands was significantly decreased following longer annealing time. It is believed that the longer the annealing time, the more austenite decomposes.

3.4 GB Sliding During Stress Relief

The specimen with the scratch lines across GBs was examined to identify the GB sliding behavior. Following stress relief testing at 550 °C for 30 min, the scratch lines still had a straight line and were continuous, as observed in Fig. 11. This implies that no significant GB sliding occurred. Thus, the creep-like GB sliding during stress relief can be excluded as a mechanism for SRC, and it can

Fig. 9 Microstructures of specimen following stress relaxation testing at 550 °C for 90 min. **a** Fe_3C located at GB. **b** Dislocation slipping, blocked by a Fe_3C particle

Fig. 10 Values of Vickers hardness and microhardness. **a** Simulated CGHAZ and M/A islands at GB without annealing. **b** Simulated specimen following annealing at 550 °C for 400 h

be considered that other mechanisms may dominate the SRC process.

3.5 In-Situ Tensile Test

Usually, the SRC occurs at temperatures above 550 °C. For the specimens annealing at 650 °C, the stress relief behavior is more pronounced. Thus, following thermal simulation, the specimen for in situ tensile testing was

annealed at 650 °C for 10 h. During in situ tensile testing at room temperature, the crack was observed to initiate adjacent a carbide at GB, as observed in Fig. 12a. Further increase of stress caused the crack propagation along GB, as shown in Fig. 12b. Tensile deformation causes plastic strain. However, dislocation slip was hindered by the $M_3\text{C}$ carbides, which resulted in the dislocation pile-up and stress concentration at GBs, typically, the triple point junctions. When the stress concentration is sufficiently

Fig. 11 SEM image showing the *straight shaped* and continuous scratch lines, as indicated by the white arrows, following stress relief testing at 550 °C for 30 min. This implies that no GB sliding occurred

high, it may result in crack initiation and subsequent propagation.

4 Discussion and Modeling

Most of the previous studies suggest that the MX phase precipitates intragranularly to strengthen the grain and that the sulfide precipitates intergranularly or S, as an impurity, aggregates along the GBs to induce the brittle during the process of the stress relief [1]. So, stress relief cannot take place within the grain but along the GBs, and the cracking mechanism is the creep-like GB sliding. However, in the present study, the mechanism of SRC is different. The SRC process is suggested as follows: following the thermal simulation (welding), most of the retained austenite is formed at GBs (see Fig. 13a). During stress relief at

Fig. 12 SEM images showing the crack initiation **a** and propagation **b**, as indicated by the *arrows*, adjacent to a triple point junction during in situ tensile testing at room temperature. The specimen was annealed at 650 °C for 10 h

Fig. 13 A schematic drawing showing the cracking mechanism. **a** Morphology of the retained austenite adjacent a triple point junction. **b** The retained austenite decomposes into ferrite and M_3C carbide. M denotes Fe, Cr, Mo, Nb or V. **c** Dislocations are initially generated within the relatively soft ferrite. Subsequent dislocation slipping is impeded by the M_3C adjacent the triple point junction, resulting in dislocation pile-up and stress concentration at the triple point junction

elevated temperatures (550 °C in the present study), the retained austenite decomposes into M_3C carbide and ferrite (see Fig. 13b). The new ferrite, formed by the decomposition of M/A islands, comprises a softening zone, and the dislocations may be generated in the ferrite adjacent to GBs. The subsequent dislocation slipping is hindered by the M_3C type carbides, which causes the dislocation pile-up and stress concentration at GBs, typically, at triple point junctions. When the stress concentration is sufficiently high, it may result in GB sliding and SRC (see Fig. 13c).

According to Taylor equation [18]:

$$\tau = \tau_0 + \alpha Gb\sqrt{\rho},$$

where τ is the flow stress, τ_0 is the stress required to move a dislocation in the absence of other dislocations, α is a dislocation strengthening constant, G is the shear modulus, b is the Burgers vector of slipping dislocation, and ρ is the dislocation density. As ρ increases, the flow stress increases as well, this results in the stress concentration. Fracture toughness K_{IC} is a material parameter to describe the resistance of the material against fracture, which is closely related to the applied stress and crack length, as follows [19]:

$$K_{IC} = Y\sigma\sqrt{a},$$

where Y is a dimensionless geometry factor, σ is the applied stress, a is the length of a crack. In the present study, the stress concentration due to microstructural changes facilitates the crack initiation and growth.

5 Conclusions

1. Large M/A islands were distributed along the GBs in the simulated CGHAZ specimen.
2. During stress relief, the retained austenite of the M/A islands along the GBs was decomposed into M_3C carbides and softening ferrite.
3. The creep-like grain boundary sliding was not observed during the stress relief process in the present study. It was indicated that the dislocations are generated within the ferrite, formed by the decomposition of M/A islands, adjacent the GBs and subsequent dislocation slipping was hindered by the M_3C carbides. The dislocation pile-up and stress concentration occurs at GBs, typically, at triple point junctions, finally caused the SRC.

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