Effects of precipitation phases on the hydrogen embrittlement sensitivity of Inconel 718

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Abstract

We investigated the effects of precipitation phases on the hydrogen embrittlement (HE) sensitivity of Inconel 718 by means of tensile tests. Hydrogen was charged into the test specimens via a cathodic charging process prior to the tensile tests. Various heat treatments were applied to conventionally aged specimens to fabricate specimens with different precipitation conditions for the $\gamma^\prime$ phase and the $\delta$ phase. For each precipitation condition, we fabricated two specimens, one of which was charged with hydrogen before the tensile test. All specimens were tensioned under identical tensile conditions. The percent loss of the reduction of area (RA) caused by pre-charged hydrogen was used to assess HE sensitivity. Both the $\delta$ phase and the $\gamma^\prime$ phase were found to play significant roles in altering HE sensitivity of Inconel 718. When these phases were totally dissolved, the HE sensitivity of the alloy was very low. The percent loss of RA decreased along with a decrease in the fractional volume of $\gamma^\prime$. The $\delta$-free aged alloy had greatly enhanced HE resistance, the same level as that of conventionally annealed alloy, and its strength was equal to that of the conventionally aged alloy. Fracture origins noted on the specimens were located on the surface layers and displayed brittle cleavage when pre-charged hydrogen was utilized. Local transgranular cleavages initiated from the $\delta$/matrix were also observed in conventionally aged specimens, where there was a presence of pre-charged hydrogen. Therefore, the $\delta$ phase was considered to promote HE by initializing micro-cracks from $\delta$/matrix interfaces. Since the $\delta$-free aged alloy has both good strength and good ductility, we propose that it is advantageous for fabricating some hydrogen-containing parts.

Keywords: Hydrogen embrittlement; Inconel 718; Superalloy; $\delta$; $\gamma^\prime$; Cathodic hydrogen charging

1. Introduction

Inconel 718 is a nickel–iron-based superalloy that is strengthened by an ordered, body-centered tetragonal $\gamma^\prime$-phase, and an ordered FCC $\gamma$ phase [1–3]. It has been a mainstay superalloy for fabricating gas-turbine engines because of its good mechanical properties up to 923 K [4]. Since new applications for Inconel 718 in hydrogen-containing environments, such as liquid-hydrogen-fueled engines, have been unveiled, researchers have been studying HE of conventionally aged alloy [5–10], and decreasing the HE sensitivity [11–14].

Most research on the HE of Inconel 718 has been conducted to investigate the effects of hydrogen on mechanical properties or fracture structures [5–10]. It is widely accepted that both internal and external hydrogen causes serious ductility loss, and causes the fracture to change from a ductile micro-void coalescence (MVC) mode to a brittle cleavage mode. The ductility loss caused by HE has mainly been ascribed to the interaction between the absorbed hydrogen and the bulk material, i.e. hydrogen accelerates crack nucleation and propagation. It is also widely accepted that there is a correlation between HE sensitivity and carbide particles, as carbide/matrix interfaces are irreversible hydrogen trapping sites [15]. However, there are few reports about the effects of the primary strengthening phases (i.e. $\gamma^\prime$/$\gamma$) and the grain-boundary strengthening phase (i.e. $\delta$) on the HE susceptibility of this alloy. Hirose et al. reported that HE sensitivity was significantly decreased by dissolving the $\gamma^\prime$/$\gamma$ particles distributed in the thin surface layer of conventionally aged Inconel 718 by a laser surface annealing (LSA) process [11,12]. Since HE fractures generally start from the surface layer, Hirose’s findings suggest that the $\gamma^\prime$/$\gamma$ phases greatly increases HE sensitivity. Fukuyama et al. reported that cracks propagated along the $\delta$/matrix interfaces when Inconel 718
was tensioned in high-pressure hydrogen [5,6]. Use of higher solution temperatures above the δ solvus has also been reported to promote the alloy’s performance in high-pressure gaseous hydrogen [16,17]. However, there are no quantitative reports about the effects of these two phases on the HE sensitivity of alloys, or the possibility of enhancing HE resistance by adjusting their precipitation conditions.

We conducted this research to investigate the effects of the γ′ and the δ phases on HE sensitivity. First, we fabricated several types of specimens with different precipitation conditions of these phases from conventionally aged alloys, using various annealing processes (two specimens of each kind). We then pre-charged one specimen of each kind with hydrogen via a cathodic charging process, before the tensile test. Finally, we tensioned two specimens of each kind to fracture under identical tensile conditions. The HE sensitivity was assessed based on the percent loss of RA that was caused by the pre-charged hydrogen. The effects of these phases on the fracture morphologies were studied via microscopic observations.

2. Material and experiments

2.1. Materials

The chemical composition of Inconel 718 employed in this research is shown in Table 1. Mitsubishi Materials Corporation supplied forged, 3 mm thick plates for use in the study. Before delivery, the plates were annealed at 1253 K for 1.2 ks, and were then water-quenched (W.Q.). Because the γ′ phase and γ phase were dissolved by the conventional annealing process, the matrix contained only δ and carbide particles. The hardness of the as-received alloy was measured to be about 225 Hv (load 1.96 N). The conventionally aged alloy was manufactured by a conventional double-aging process, i.e. furnace cooled to 993 K/28.8 ks and air-cooled to 894 K/28.8 ks. The γ′ phase and γ phase were precipitated in this aging process. The hardness of the conventionally aged alloy was measured to be about 450 Hv (load 1.96 N).

Since some δ-dissolved alloy (specimens) had the γ′/γ′ phases dissolved or precipitated, we refer to the δ-dissolved alloys (specimens) as δ-free annealed alloys (specimens) when they had no γ′/γ′ phases, or δ-free aged alloys (specimens) when the γ′/γ′ phases were precipitated by the conventional double-aging process. Conversely, we refer to those alloys (specimens) without any dissolution of the δ phase as conventionally annealed alloys (specimens).

2.2. Preparing specimens with various precipitation conditions

We fabricated several types of specimens with various precipitation conditions via different heat-treatment processes. The specimens and their corresponding heat treatment processes are listed in Table 2. The conventionally aged alloy, as stated in Section 2.1, was used in making conventionally aged specimens without any further processing. The specimens with other precipitation conditions of γ′ or δ phases were prepared by annealing the conventionally aged specimens using various heat treatments, as shown in Table 2. We annealed the conventionally aged specimens in order to obtain different precipitation conditions for the γ′ phases, rather than obtaining them by aging the conventionally aged specimens. The primary aim of this was to simulate the dissolution of the γ′ phase that occurs in the LSA processes [11–14]. We heat-treated two specimens of each kind at the same time to ensure that they had identical precipitation conditions. After polishing each specimen, we measured its hardness with a load of 1.96 N and a holding time of 15 s.

2.3. Cathodic hydrogen charging

Cathodic hydrogen charging was carried out as shown in Fig. 1. The molten salt consisted of NaHSO₄ and KHSO₄ at a composition ratio of 1:1 (mol/mol). The charging was conducted at 573 K for 108 ks with a fixed electric current.

| Table 1 | Chemical composition of Inconel 718 (wt%) |
|---------|----------------------------------------|
| C       | 0.03 | Mn       | 0.15 | Si       | <0.01 | P      | <0.001 | S      | 0.002 |
| Cr      | 18.54 | Ni      | 52.07 | Mo      | 3.04 | Nb + Ta | 5.35 | Ti      | 1.01 |
| Al      | 0.51 | Co      | 0.02 | B      | 0.004 | Cu     | 0.01 | Fe     | REM |

| Table 2 | Precipitation condition and corresponding heat treatment process |
|---------|---------------------------------------------------------------|
| Precipitation condition no. | Expected precipitation morphologies | Heat treatment processes |
| 1 | δγ′ free | Aging<sup>a</sup> | → 13 K/s, 1303 K, 0 s, W.Q.<sup>b</sup> |
| 2 | δ+γ′ (partly dissolved) + (γ′) | Aging | → 14 K/s, 1219 K, 0 s, W.Q.<sup>b</sup> |
| 3 | δ+γ′ (partly dissolved) + (γ′) | Aging | → 13 K/s, 1204 K, 0 s, W.Q.<sup>b</sup> |
| 4 | δ+γ′+(γ′) | Aged |
| 5 | δ-free, γ′(γ′) free | Aging | → 1313 K/3.6 ks, W.Q.<sup>c</sup> |
| 6 | δ-free, γ′+(γ′) | Aging | → 1313 K/3.6 ks, W.Q.<sup>c</sup> → Aging |

<sup>a</sup> Aging: 993 K for 28.8 ks, furnace cooling → 894 K for 28.8 ks, air cooling.
<sup>b</sup> Format: heating rate, peak temperature, holding time, quenching method.
<sup>c</sup> Format: temperature/holding time, water quenched.
of 0.5 mA/mm². Water vapor was bubbled into the molten salt to stir the molten electrolyte and to provide sufficient hydrogen ions.

The charging time was determined based on Eq. (1), allowing the hydrogen diffusion to reach the center of the specimen.

$$\sqrt{2\pi t} \approx H/2$$  \hspace{1cm} (1)

Here, $t$ is the charging time, and $H$ is the thickness of the tensile specimen, viz. 3 mm. Since hydrogen was charged from both surfaces of the specimen, the diffusion depth of hydrogen from each surface was half the specimen’s thickness, i.e. $H/2$.

$D$ is the diffusion coefficient of hydrogen in Inconel 718, calculated based on Eq. (2) [18].

$$D \text{ (m}^2/\text{s}) = 4.06 \times 10^{-7} \exp \left(-\frac{48.63 \times 10^3}{RT}\right)$$  \hspace{1cm} (2)

Here, $R$ is the gas constant, and $T$ is the charging temperature, i.e. 537 K in our research.

At 537 K, the time needed for hydrogen to diffuse through the half thickness of the specimen was calculated to be 75.6 ks. We used a longer charging time, i.e. 108 ks, in order to counteract the concentration gradient effects in the charging process. A longer time was also essential to obtain a homogeneous hydrogen distribution along the depth direction.

After charging the specimens with hydrogen, we cleaned them with distilled water and then stored them in liquid nitrogen until the tensile tests were to be conducted. The hydrogen content in each specimen was measured after the tensile tests, using a hydrogen analyzer (Horiba EMGA-521) and 4 mm wide samples cut from the broken halves. Some of these samples were equally polished on both surfaces to reduce their thickness to 1 mm. The hydrogen concentration in each sample was then measured and compared with that of the corresponding unpolished specimen. There were no obvious hydrogen concentration differences in the above comparisons, confirming that hydrogen had reached the specimen centers under these charging conditions. The results also suggested that, there was no obvious inhomogeneous distribution of the hydrogen in the depth direction in each specimen. An identical hydrogen concentration of about 20 wtppm was obtained for all specimens under different precipitation conditions. This ensured that the differences in the HE sensitivity among the different specimen types were not from the difference in hydrogen concentration.

2.4. Tensile tests

Tensile specimens with a planar part $40 \times 7 \times 3$ mm in size were used in the tensile tests. All specimens were tensioned in air at room temperature, with a slow crosshead speed of $1.67 \times 10^{-3}$ mm/s. The RA and the tensile strength were measured to assess the mechanical properties of each alloy. The percent loss of RA defined by Eq. (3) from the pre-charged hydrogen was used to evaluate HE sensitivity.

$$\text{Percent loss of RA} = \frac{R_{A0} - R_{AH}}{R_{A0}} \times 100$$  \hspace{1cm} (3)

Here, $R_{A0}$ and $R_{AH}$ are the RAs of the hydrogen-free and hydrogen-pre-charged specimens.

2.5. Microstructure observations

We studied the precipitation of the large carbide particles and the δ particles using scanning electron microscopy (SEM), and used transmission electron microscopy (TEM) to observe the precipitation of γ' and γ. SEM samples were prepared by mechanical polishing, and etching utilizing 10% oxalic acid. Thin foils for TEM observations were prepared by electro-polishing 3 mm discs in a Tenupol-II jet polisher, with 20% perchloric acid and 80% ethanol at 233–243 K. Fracture appearances were observed by SEM.

3. Results and discussions

3.1. Precipitation in tensile specimens

Typical microstructures of the conventionally aged alloy observed by SEM and TEM are shown in Figs. 2 and 3 [13]. As seen in these figures, the precipitate phases were mainly needle-shaped or rod-shaped δ, block-shaped carbide, and
disc-shaped $\gamma''$. Although $\gamma$ should also be present, it generally has much less fractional volume (about 5%) than the $\gamma''$ phase (about 13%) [1–3]. Furthermore, the $\gamma''$ phase exhibits very minimal strengthening effects with this alloy when compared to the $\gamma''$ phase [2,3]. Therefore, we can rationally have the $\gamma''$ phase represent the coherent phases ($\gamma''$ and $\gamma$) when we study their effects on HE.

The annealing processes for dissolving the $\gamma''$ particles (1–3 in Table 2) were designed to avoid the transformation zone of the $\delta$ phase (based on the TTT diagram [19]), so that the phase transformation of the $\delta$ phase did not occur during the heating treatments. Therefore, the differences of HE sensitivity among these specimens were merely caused by differences in the precipitation conditions of the $\gamma''$ phase. Thus, we could evaluate the effects of the $\gamma''$ phase on HE sensitivity by correlating HE-induced ductility loss with the precipitation conditions of the $\gamma''$ phase, without considering the effects of other precipitates. For comparison, we measured the hardness of these specimens and listed the results in Table 3. To calculate the volume fraction of the $\gamma''$ phase in these specimens, we utilized the relationship between the volume fraction of the $\gamma''$ phase and the hardness of the alloy with an assumption that the number density and the aspect ratio of the disc-shaped $\gamma''$ particles remain constant [14]. The volume fraction of the $\gamma''$ phase was 0 in the conventionally annealed alloys and 13% in aged alloys.

$$f = f_0 \left( \frac{H_V - H_{V0}}{H_{V\max} - H_{V0}} \right)^{3/2}$$

Here, $f_0$ is the volume fraction of the $\gamma''$ phase in the conventionally aged Inconel 718, viz. 13%. $H_{V\max}$ and $H_{V0}$ are the hardnesses of the conventionally aged and annealed alloys, i.e. 450 and 225 Hv. $H_V$ is the hardness of an alloy that was annealed from the conventionally aged alloy via a particular heat-treatment process, such as those given in Table 3.

Annealing at 1313 K for 3.6 ks completely dissolved the $\delta$ particles (strengthening $\gamma''/\gamma'$ phases was also totally dissolved). We refer to these specimens as $\delta$-free annealed specimens, as shown in Fig. 4.

### 3.2. Effects of the precipitation particles on HE sensitivity

#### 3.2.1. Effect of the $\gamma''$ phase

The tensile strengths and the RAs of the specimens with various $\gamma''$ precipitation conditions are shown in Fig. 5(a) and (b). Regardless of whether hydrogen charging took place or not, the tensile strength was enhanced and the RA was decreased by increasing the volume fraction of the $\gamma''$ phase. For all specimen types observed, pre-charged hydrogen decreased the RA (that is, all specimens exhibited HE sensitivity). Furthermore, the RA of conventionally annealed alloy decreased from 25.5 to 11.6%, which was the RA of conventionally aged alloy. To evaluate the correlation between HE sensitivity and the precipitation condition of the $\gamma''$ phase, we plotted the percent loss of RA caused by pre-charged hydrogen for each specimen type in Fig. 6 with the calculated volume fraction of the $\gamma''$ phase as the abscissa. Because the $\gamma''$ particles maintain homogeneous distribution in the matrix and coherent characteristics regardless of their size or volume fraction [12,13], we can correlate the HE sensitivity of these specimens with the volume fraction of this phase. The percent loss of RA was calculated using Eq. (3); the fractional volume of the $\gamma''$ phase is from Table 3. As shown in Fig. 6, the percent lose of
RA decreased almost linearly with increasing volume fraction of the \( \gamma' \) phase. Therefore, we can conclude that the HE sensitivity of Inconel 718 can be decreased by dissolving the \( \gamma' \) phase.

Fracture surfaces of all specimens, with or without pre-charged hydrogen, are shown in Fig. 7. Fig. 7(b), (d), (f), and (g) show the morphologies of the fracture origins of the hydrogen-charged specimens. All hydrogen-free specimens fractured via the ductile MVC mode. While the dimples of the conventionally aged specimens were small and had sharp tear ridges (Fig. 7(a)), the dimples of the conventionally annealed specimens were large and had blunt tear ridges (Fig. 7(d)). When hydrogen was pre-charged, brittle cleavage morphologies could be observed at the fracture origins located on the surfaces, although the main part of the fracture surface in either of the hydrogen-pre-charged specimens still exhibited the MVC mode. Clavel [20] and Hicks [7,8] found that the microcleavage facets were \{111\}s. However, differences could be discerned among the fracture characteristics of the specimens with different precipitation conditions during the \( \gamma' \) phase. There were more distinct cleavage planes and steps at the crack origin in the conventionally aged specimen, than in other specimens that have fewer \( \gamma' \) particles. Furthermore, transgranular cleavage features were frequently observed in the central part of the fracture surface of the conventionally aged specimens when there was pre-charged hydrogen (Fig. 8). Such features became less apparent in the partly annealed specimens, and almost all conventionally annealed specimens were in the MVC mode, except for the fracture origin, when there was pre-charged hydrogen.

As stated in Section 2.3, the hydrogen concentration was about 20 wtppm in all specimens regardless of precipitation conditions in the \( \gamma' \) phase. Therefore, the differences of HE sensitivities among different specimens were not due to the differences in bulk hydrogen concentration. In Inconel 718, the binding energy of hydrogen to carbide particles ((Nb, Ti)C) is 77–87 kJ/mol, well in excess of the 58 kJ/mol [21] delineating the reversible and irreversible traps [22]. Therefore, carbide particles are irreversible hydrogen traps and cannot be expected to provide hydrogen for dislocation transport [23]. Moreover, for carbide precipitates, there were no differences in precipitation conditions among all types of specimens. Therefore, such sites should not affect the HE sensitivity differences among the specimens in this research.

Many theories have been suggested to explain the HE phenomenon, such as hydrogen-enhanced localized plasticity [24,25], hydrogen-induced de-cohesion [26,27], and...
hydrogen trapped by precipitates and second-phase particles causing cracking nucleation and propagation [22,28,29]. Regardless of the mechanisms, a high local hydrogen concentration and intensified stress concentration are essential for HE crack initiation, and fast transport of hydrogen for propagating crack tips is a prerequisite for crack propagation.

While the mechanism heightening the effect of the \( \gamma'' \) phase on HE sensitivity is not clear, one possible explanation is as follows. The \( \gamma''/\text{matrix} \) interfaces with a large lattice misfit of 2.86% can be expected to trap more hydrogen than the matrix. When the specimens were tensioned, especially when plastic deformation occurred, dislocations accumulated at these sites. Therefore, more hydrogen was likely to be dumped at these sites by dislocation transport. Micro-cracks could initiate from these sites when the hydrogen concentration reached a critical value. These micro-cracks could accelerate the propagation of the main crack by merging into it. Therefore, an increased volume fraction of the \( \gamma'' \) phase can accelerate the fracture process.

Compared to the \( \gamma'' \) phase, the \( \gamma' \) phase has both a much smaller volume fraction and a much smaller lattice misfit with the matrix. Therefore, the \( \gamma' \) phase should have much less effect on both the stress concentration and the trapping of hydrogen.

### 3.2.2. Effect of the \( \delta \) phase

We compared tensile strengths and RAs of the \( \delta \)-free annealed specimens and the \( \delta \)-free aged specimens (with and without hydrogen pre-charging) with the corresponding properties of the conventionally annealed and aged specimens, as shown in Fig. 9(a) and (b). There were only minimal losses of strength in the \( \delta \)-free annealed alloys compared to those with the conventional totally annealed alloys, both with and without hydrogen pre-charging. Also, minimal losses of strength were found in the \( \delta \)-free aged alloys, compared with the conventionally aged specimens, regardless of the hydrogen charging condition.

When we applied conventional double-aging (993 K/28.8 ks/furnace-cooling, followed by 894 K/28.8 ks/air-cooling) to \( \delta \)-free specimens, precipitation of \( \delta \) particles...
was not expected to occur because the aging temperatures (993 and 893 K) were far below the precipitation temperature range of the δ phase [19]. Thus, we were able to evaluate the effect of the δ phase on the HE sensitivity of Inconel 718 by comparing the HE sensitivity of the δ-free annealed alloy to that of the conventionally annealed alloy, and compare the HE sensitivity of the δ-free aged alloy to that of the conventionally aged alloy. Although the Nb released from dissolved δ particles might cause increased precipitation of the γ″ particles in the aging process compared to conventionally aged alloy, because the volume fraction of δ was very small compared to the original volume fraction of γ″ [30], the increase was negligible. In other words, the difference in the volume fraction of the γ″ phase between the δ-free aged alloy and the conventionally aged alloy could be neglected. Therefore, we were able to evaluate the effect of the δ phase on HE sensitivity by comparing the loss of RA between the δ-free aged alloy and the conventionally aged alloy.

δ-free annealed and δ-free aged alloys had RA values similar to the conventionally annealed and aged alloys, respectively, when hydrogen was not pre-charged. However, when pre-charged hydrogen was utilized, the RAs of the δ-free alloys were much larger than those of the conventionally annealed or aged alloys. Furthermore, the percent loss of RA caused by pre-charged hydrogen in the δ-free annealed alloy was only 6.83%, which was almost negligible when compared with the 44.32% percent loss of conventionally annealed specimens. For the δ-free aged specimens, the percentage of RA loss caused by pre-charged hydrogen was 29.41%, which was much lower than the 68.90% loss of the conventionally aged specimens. These results are summarized in Fig. 10. As the arrows in this figure indicate, the percent loss of RA in Inconel 718 caused by pre-charged hydrogen was decreased almost at a constant rate across the precipitation condition range of the γ″ phase by dissolving the δ phase in the alloy. The large decrease in HE sensitivity suggests that the δ particles significantly affect the HE of the alloy, and that the HE resistance of the alloy can be greatly improved by dissolving the δ particles.
in the alloy. The larger percent loss of RA for the δ-free aged alloy can be ascribed to the precipitation of the γ'/γ phases, which aggravated HE sensitivity.

Fracture surfaces of the δ-free specimens are shown in Fig. 11(a)–(d). Just as for conventional alloys, hydrogen-free specimens all exhibited ductile MVC fracture modes. The δ-free annealed specimen had much larger dimples than the conventionally annealed alloy; all these dimples originated from carbide particles (Fig. 11(a)). When the δ-free alloy was aged with the conventional double-aging process, the percent loss of RA caused by the pre-charged hydrogen increased because of precipitation of the γ'/γ phases. However, as shown in Fig. 10, the loss of RA of the δ-free aged alloy was still much less than that of the conventionally aged alloy. In the presence of pre-charged hydrogen, the RA of the δ-free aged alloy was at almost the same level as the conventionally annealed alloy, and much better than that of conventionally aged alloy. However, the percent loss of RA caused by pre-charged hydrogen in δ-free annealed specimens is very small, much smaller than that of the conventional annealed alloy, as shown in Fig. 9(b). This implies that the intrinsic HE of the nickel-based matrix, as well as the HE from carbides, is insignificant when neither δ nor γ⁰ particles exist.

Sites of transgranular cleavage that initiated from δ particles were frequently observed in conventionally aged specimens, as shown in Fig. 12. In the conventionally annealed specimens and the partly annealed specimen that had a hardness of 263 Hv, micro-cracks, including some intergranular micro-cracks, were also observed, as shown in Fig. 13. Such observations were in agreement with Fourrier’s observations wherein micro-cracks, including some intergranular ones, were reported to have originated from hydrogen-trapped δ particles [10]. In tensile experiments, moving dislocations could transport and dump more hydrogen into the δ/matrix interfaces. Micro-cracks could initiate from these sites when the hydrogen reached critical fugacity. These micro-cracks can merge into it and accelerate the propagation of the primary crack. Therefore,
the HE induced cracks might have a tendency to propagate along δ/matrix interfaces, in agreement with Refs. [5,6]. Since large δ particles are not coherent with the matrix and tend to distribute along grain boundaries, they are likely to have a larger effect on HE than the γ'/γ' particles. As shown in Fig. 9(b), the δ-free aged specimen had an HE sensitivity that was almost equal to that of the conventionally annealed specimen, i.e. dissolving the δ phase had almost the same effect on HE sensitivity as dissolving the γ'/γ' particles, although the δ phase has much smaller volume fraction than γ'/γ'.

Because the volume fraction of the δ phase is almost negligible when compared to that of the γ' phase and because dissolving the δ phase has almost no harmful effects on the strength of the alloy, we suggest manufacturing parts that have both good strength and good HE resistance for hydrogen environments using δ-free aged Inconel 718.

4. Conclusions

1. RAs for all specimens decreased when hydrogen charging was conducted. Therefore, Inconel 718 demonstrates HE sensitivity regardless of the precipitation conditions of the δ phase or the γ'/γ' phases. However, when all these precipitates were dissolved, the intrinsic HE sensitivity of the matrix, as well as that caused by the carbide phase, was found to be minimal.

2. The percent loss of RA caused by pre-charged hydrogen decreased almost linearly with decreasing volume fraction of the γ' phase. Therefore, applying certain
annealing processes to the conventionally aged Inconel 718 can decrease its HE sensitivity.

3. The percent loss of RA caused by pre-charged hydrogen was decreased to a large extent by dissolving \( \delta \) particles in the alloy. The \( \delta \)-free aged alloy had a much lower HE sensitivity than conventionally aged alloy. It exhibited the same level as conventional totally annealed alloy, while the strength was consistent with that of conventionally aged alloy.

4. For certain hydrogen-containing applications, parts with both good strength and low HE susceptibility can be fabricated by employing the \( \delta \)-free aged alloy.

References

[1] D.F. Paulonis, J.M. Oblak, D.S. Duvall, Trans. ASM 62 (1969) 611.
[2] J.M. Oblak, D.F. Paulonis, D.S. Duvall, Metall. Trans. 5 (1974) 143.
[3] M.C. Chaturvedi, Y.-F. Han, Metall. Sci. 17 (1983) 145.
[4] E.A. Loria, J. Metall. (1988) 36.
[5] S. Fukuyama, K. Yokogawa, M. Araki, H. Aoki, Y. Yamada, J. Soc. Mater. Sci. Jpn 40 (1991) 736.
[6] S. Fukuyama, K. Yokogawa, Y. Yamada, T. lida, Tetsu To Hagane 78 (1992) 860.
[7] P.D. Hicks, C.J. Altstetter, Metall. Trans. 21A (1990) 365.
[8] P.D. Hicks, C.J. Altstetter, Metall. Trans. 23A (1992) 237.
[9] J. He, S. Fukuyama, K. Yokogawa, A. Kimura, Mater. Trans. JIM 35 (10) (1994) 689.
[10] L. Fournier, D. Delafosse, T. Magnin, Mater. Sci. Engng A 269A (1999) 111.
[11] A. Hirose, Y. Arita, Y. Nakanishi, K.F. Kobayashi, Mater. Sci. Engng 219 (1996) 71.
[12] A. Hirose, L. Liu, K.F. Kobayashi, in: I. Miyamoto, K. Sugimoto, T.W. Sitterson (Eds.), Proceedings of SPIE, Saitama, Japan, 2000, pp. 236.
[13] L. Liu, K. Tanaka, A. Hirose, K.F. Kobayashi, Seventh International Welding Symposium, Kobe, Japan, 2001.
[14] L. Liu, A. Hirose, K.F. Kobayashi, Acta Mater. 50 (2002) 1331.
[15] B.G. Pound, Acta Metall. Mater. 38 (12) (1990) 2373.
[16] R.W. Staehle, Trans. Inst. Chem. Engng 47 (1969) T227.
[17] R.J. Walter, W.T. Chandler, in: I.M. Bernstein, A.W. Thompson (Eds.), Hydrogen in Metals, ASM, Metals Park, OH, 1974, p. 515.
[18] J. Xu, X.K. Sun, Q.Q. Liu, W.X. Chen, Metall. Mater. Trans. 25A (1994) 539.
[19] J.W. Brook, P.J. Bridges, Superalloys, The Metallurgical Society, 1988, p. 33.
[20] M. Clavel, A. Pineau, Metall. Trans. 4 (1973) 47.
[21] A. Turnbull, R.G. Ballinger, I.S. Hwang, M.M. Morra, M. Psaila-donbrowski, R.M. Gates, Metall. Trans. 23A (1992) 3231.
[22] G.M. Pressouyre, I.M. Bernstein, Acta Met. 27 (1979) 89.
[23] G.M. Pressouyre, I.M. Bernstein, Metall. Trans. 9A (1979) 1517.
[24] C.D. Beachem, Metall. Trans. 3 (1972) 437.
[25] S.P. Lynch, Met. Forum 2 (1979) 189.
[26] J.G. Morlet, H.H. Johnson, A.R. Troiano, J. Iron Steel Inst. 189 (1958) 37.
[27] A.R. Troiano, Trans. ASM 52 (1960) 54.
[28] G.M. Pressouyre, I.M. Bernstein, Metall. Trans. 9A (1978) 1571.
[29] G.M. Pressouyre, Acta Metall. 28 (1980) 895.
[30] W.C. Liu, Z.L. Chen, M. Yao, Metall. Mater. Trans. A 30A (1999) 31.

Fig. 12. Transgranular cleavage that originated from \( \delta \)/matrix interface in the center of the conventionally aged specimen when there was pre-charged hydrogen. (A: \( \delta \) particle).

Fig. 13. Micro-cracks in the hydrogen-precharged 263 Hv specimen. (A: an intergranular micro-crack).