Slow Strain Rate Tensile Test Properties of Iron-Based Superalloy SUH660 in Hydrogen Gas

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To investigate dependence of strain rate of tensile test for iron-based superalloy SUH660 (A286), tensile tests were conducted for the specimens in 70 MPa hydrogen gas and air at 150°C. Nominal stress-nominal strain curve of each strain rate in 70 MPa hydrogen gas showed same behavior to maximum load via yield point in comparison with that in air, however, each elongation at breaking point in 70 MPa hydrogen was a little shorter than that in air. The values of tensile strength didn’t depend on the strain rate in 70 MPa hydrogen as well as those in air. In addition, the difference in tensile strength wasn’t observed between that in 70 MPa hydrogen gas and that in air for the strain rate. However, it’s proved that relative reduction of area in 70 MPa hydrogen to that in air was significantly affected by strain rate of tensile test. Those values were 80%, 51%, and 32% in the case of strain rate 5.0×10⁻⁵, 7.5×10⁻⁶, and 1.25×10⁻⁶ s⁻¹, respectively. The morphology of fracture surface also changed from dimple to quasi-cleavage (QC), with a decrease in strain rate. Simulation of hydrogen gas diffusion from surface to inside during experiment showed that the hydrogen diffusion layer of specimen with QC fracture surface (RRA 51%, strain rate 7.5×10⁻⁶ s⁻¹) was only 0.25 mm in depth. That implies that hydrogen content at crack tips is much higher than that of simulation due to hydrogen concentration by a couple of defects. That tendency seems to become stronger with a decrease in strain rate.

KEY WORDS: SUH660; A286; slow strain rate tensile test; relative reduction of area; reduction of area; hydrogen embrittlement; quasi-cleavage fracture; dimple.

1. Introduction

In recent years, as one of solutions of global warming, expectations for Fuel Cell Vehicle (hereinafter referred to as “FCV”) without emitting any greenhouse gas at all is growing. Toward the spread of FCV, in order to enable cruising distance of about 650 km without sacrificing the indoor space, it is necessary to install hydrogen in the state of high pressure gas of about 70 MPa. Also, to fill it in about the same 3 minutes as a gasoline-powered vehicle, we have to cool down the hydrogen to about −40°C. Therefore, 70 MPa class hydrogen stations capable of quickly filling hydrogen gas by precooler have already been installed in more than 90 locations mainly in the four major metropolitan areas (Kanto, Chukyo, Kansai, Kitakyushu) in Japan.2,3) Along with that, FCV sold commercially in December 2014 also increased to more than 2,000 units in FY2017.4)

At these hydrogen stations, austenitic stainless steels with high nickel equivalent value,2,3) which are considered to have little mechanical influence by hydrogen, are currently mainly used. In the NASA database, Safety Standard for Hydrogen and Hydrogen Systems,5) as a criterion for evaluating the influence of hydrogen on metal, the ratio of the tensile strength in the hydrogen gas to the tensile strength in the helium gas obtained in the notched specimen, $\sigma_{B,H}/\sigma_{B,He}$ (relative tensile strength), and the ratio of reduction in area in hydrogen gas to helium gas, $\phi_{A,H}/\phi_{A,He}$ (relative reduction in area) obtained with a smooth test specimen.7) Also, in Japan, the materials with or more than 80 percent of a relative reduction in area ($\phi_{A,H}/\phi_{A,He} \times 100$) are classified as materials with less influence of hydrogen, by a low strain rate tensile test of a smooth test specimen under a hydrogen gas to the air.8) For example, austenitic stainless steels SUS 316 and SUS 316 L with or more nickel equivalent values ($= 12.6 \% \text{ C} + 0.35 \% \text{ Si} + 1.05 \% \text{ Mn} + [\% \text{ Ni}] + 0.65 \% \text{ Cr} + 0.98 \% \text{ Mo}, [\%]$ represent mass%)9) of 28.5 mass% are said to be usable in a 70 MPa class hydrogen station in Japan under a high pressure hydrogen gas of −40°C or more.8–10) However, since high pressure resistance is required in the hydrogen station, a material with high strength is required for thinning and weight reduction. SUH 660 which is an iron-based superalloy is a precipitation hardening type heat resistant alloy having higher strength than austenitic stainless steel such as SUS 304. In order to develop high strength, SUH 660 is obtained by precipitating a $\gamma'$ phase (Ni₃(Al,Ti)) by performing aging treatment at 720°C for 16 hours after solid solution heat treatment at

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980°C. Utilizing the high mechanical strength, thinning of piping and fittings, and reduction of heat capacity are desired. The environment for use in a 70 MPa class hydrogen station is from −40°C to 50°C around a dispenser unit, but when used around a compressor unit, due to heat generation of compressed hydrogen gas accompanying pressurization usually from −10°C to 200°C in the environment.

About SUH 660, generally known as A 286 (AMS 5732), the relative reduction in area of the slow strain rate tensile test is described as 98% and 101% respectively in the database of NASA mentioned above and the database of AIST and is reported as a material not affected by hydrogen. This is in good agreement with the fact that the nickel equivalent value of SUH 660 is higher than that of austenitic stainless steel such as SUS 316 L in terms of nickel equivalent one. On the other hand, it has been reported that SUH 660, which has been charged with hydrogen, shows a reduction in area in the tensile test. Therefore, it does not agree with the result of slow strain rate tensile test under external high pressure hydrogen environment, and data of slow strain rate tensile test under high temperature hydrogen environment is not enough. At the present High Pressure Gas Safety Act in Japan, low strain rate tensile test under external high pressure hydrogen environment is one of the evaluation methods of influence of hydrogen on metallic materials. Therefore, in this study, the strain rate is changed, the slow strain rate tensile test in the high temperature hydrogen gas was performed, and its dependency was investigated. The internal hydrogen concentration was also discussed by comparing the fracture observation result and mechanical properties with those of the hydrogen charge material.

2. Experimental

2.1. Materials

The piping material SUH 660 (nominal diameter 9/16 inch: external diameter φ 14.15 mm, wall thickness t 3.75 mm), utilizing for 2 years around dispenser unit of hydrogen station, was used as a test material. The chemical composition of the test material is shown in Table 1. Figure 1 shows the macrostructure and the microstructure of the radial cross section of piping material. Somewhat inclusions were observed, but the microstructure was a homogeneous austenitic structure, and it was confirmed that the size of the crystal grain was not different between the outer side of the pipe, the center part, and the inner side. Also, the hardness was measured at ten places at equal intervals in the radial direction at a measuring load of 9.8 N using a Vickers hardness tester. As a result, as shown in Fig. 2, it was found that the value was almost constant in the cross-sectional direction (average Vickers hardness was 330.9 Hv, standard
deviation was 7.9 Hv). It was confirmed that SUH 660 was subjected to uniform aging treatment.

2.2. Hydrogen Content Measurement
Before preparing the tensile test specimen, the hydrogen content in the specimen was measured. A piping material SUH 660 was cut into a round shape with a thickness of about 3 mm, and it was cut in half lengthwise as a measurement sample. The temperature was raised from room temperature to 800°C at 10°C/min. using gas chromatography (ULVAC ST-200P special type) as the measuring apparatus. As a result of the measurement, no obvious emission peak was observed, and it was confirmed that the hydrogen content of the test material was hardly contained as 0.16 mass ppm.

2.3. Preparation of Tensile Test Piece
To prepare slow strain rate tensile specimens, the piping material (wall thickness t 3.75 mm) used as the test material is in the longitudinal direction of the pipe. The shape of the specimen is shown in Fig. 3. The parallel portion was 20 mm, and the diameter Φ of the test piece was 2.5 mm. Further, the parallel portion of the tensile test piece was mirror-finished with α-alumina.

2.4. Slow Strain Rate Tensile Test and Fracture Surface Observation
The tensile test was carried out in air or in 70 MPa hydrogen gas while keeping the strain rates at 5.0 × 10⁻⁵, 7.5 × 10⁻⁶, and 1.25 × 10⁻⁶ s⁻¹, with the test temperature set to 150°C. In the hydrogen gas test, after replacing the inside of the system with nitrogen, hydrogen of 5N purity was replaced three times and then raised to the specified pressure. After elevating the temperature to 150°C, a tensile test was immediately conducted at each strain rate. Oxygen concentration is kept under 1 ppm according to ASME BPVC Section VIII, Division 3, Article KD-10. In addition, using a scanning electron microscope (SEM), the fracture surface of the specimen after the tensile test was observed.

3. Experimental Results and Discussion
3.1. Stress Strain Characteristics
When the strain rate is changed to 5.0 × 10⁻⁵, 7.5 × 10⁻⁶, and 1.25 × 10⁻⁶ s⁻¹ with the tensile test temperature set to 150°C, in air or 70 MPa hydrogen gas, the nominal stress-nominal strain curves are shown in Fig. 4. The material with a strain rate of 5.0 × 10⁻⁵ s⁻¹ showed almost the same curve until yield point, maximum load point and fracture in air and hydrogen gas (Fig. 4(a)). At the strain rates of 7.5 × 10⁻⁶ and 1.25 × 10⁻⁶ s⁻¹, those yield points and those maximum load points are almost the same but the behaviors after those maximum loads are different, there was a tendency that those elongations at break in hydrogen gas were slightly shorter than those in air (Figs. 4(b), 4(c)). Tensile strength whose test temperature is 150°C, the strain rate is changed to 5.0 × 10⁻⁵, 7.5 × 10⁻⁶, and 1.25 × 10⁻⁶ s⁻¹, in a 70 MPa hydrogen gas and in air is shown in Fig. 5. As a result, the influence of the strain rate on the tensile strength observed in high purity iron and high strength high carbon chromium bearing steel SUJ216,17) was not recognized in low strain rate tensile test of SUH660 in air and 70 MPa hydrogen gas, both of which were almost constant in this test range. There was also little difference in tensile strength between in air and in 70 MPa hydrogen gas. This is different from the behaviors of the austenitic stainless steel SUS304 and the like in which the strength decreases in the

Fig. 3. Shape and dimensions in mm of tensile test specimen of SUH660.

Fig. 4. Nominal stress-nominal strain curves at 150°C in air and 70 MPa hydrogen gas at strain rates of (a) 5×10⁻⁵, (b) 7.5×10⁻⁶, and (c) 1.25×10⁻⁶ s⁻¹. (Online version in color.)
hydrogen gas due to the formation of the strain-induced martensite, which is considered to be due to the stability of the austenitic phase. It was also different from the phenomenon that the tensile strength slightly increased when hydrogen was charged to SUH 66015 or SUS 316.18)

These slow strain rate tensile test results were summarized in Table 2. These relative elongation (elongation in hydrogen gas/elongation in air × 100) and relative reduction of area (reduction of area in hydrogen gas/reduction of area in air × 100: hereinafter referred to as RRA) values were also shown in Table 2. Tamura et al. examined the relative elongation values of SUS 304L, SUS 316, and SUS 316L under 45 MPa hydrogen gas at room temperature and 85°C under strain rate from 3 × 10⁻⁶ to 8 × 10⁻⁵ s⁻¹. As a result, it was reported that the influence of the strain rate was small.7) In this study of SUH 660, at a strain rate of 5.0 × 10⁻⁵ s⁻¹, the relative elongation value was 100%. But as the strain rate slowed to 7.5 × 10⁻⁶ and 1.25 × 10⁻⁶ s⁻¹, it was found that the relative elongation value became about 90%.

On the other hand, as shown in Table 2, it became clear that the reduction of area and RRA under 70 MPa hydrogen gas decreased considerably more than the change in elongation at break as the strain rate became slower. Then, the dependency of RRA on strain rate was shown in Fig. 6. In the vicinity of the strain rate of 5.0 × 10⁻⁵ s⁻¹ currently used for investigating the influence of hydrogen under the high pressure hydrogen gas, the RRA value was about 80% which was judged to be less influenced by hydrogen. The RRA values decreased when the strain rate was slower, and the RRA value was 32% at the strain rate of 1.25 × 10⁻⁶ s⁻¹. It was found that RRA, which was regarded as one of the influence evaluation criteria for the influence of hydrogen of metallic materials, showed a very large strain rate dependency in the evaluation of SUH 660 at high temperature. Therefore, in order to investigate the causes, the fracture surface was observed and the hydrogen content in the slow strain rate tensile specimen was considered.

### 3.2. Observation of Fracture Surface and Hydrogen Content in Specimen

Figures 7(a)–7(f) show the fracture surfaces by SEM for the tensile test specimens in air and in 70 MPa hydrogen gas at 150°C. The fracture surfaces of the SUH 660 tensile test specimen (Figs. 7(a), 7(c), 7(e)) executed in air were ductile fracture surfaces showing the dimple pattern. Small dimple faces, which are considered to have formed along the maximum shear stress plane between large dimples and large voids formed due to inclusions, etc. were observed in each sample. The fracture surface of the test conducted at a strain rate of 1.25 × 10⁻⁶ s⁻¹ as shown in Fig. 7(e) was somewhat shallow with dimples.

Table 2. Tensile properties of SUH660 at 150°C in air and 70 MPa hydrogen gas.

| Strain Rate /s⁻¹ | Environment  | Tensile strength /MPa | Elongation /% | Relative elongation /% | Reduction of area /% | Relative reduction of area /% |
|------------------|-------------|----------------------|---------------|-----------------------|---------------------|-------------------------------|
| 5.0 × 10⁻⁵       | Air         | 1 069                | 22            | –                     | 54                  | –                             |
| 7.5 × 10⁻⁶       | 70 MPa H₂    | 1 059                | 22            | 100                   | 43                  | 80                            |
| 7.5 × 10⁻⁶       | Air         | 1 053                | 23            | –                     | 57                  | –                             |
| 1.25 × 10⁻⁶      | 70 MPa H₂    | 1 102                | 20            | 87                    | 29                  | 51                            |
| 1.25 × 10⁻⁶      | Air         | 1 078                | 23            | –                     | 47                  | –                             |

* Relative elongation and relative reduction of area correspond to the ratio of each property of specimen in 70 MPa hydrogen gas to the one in air.
became slower. The specimens with a strain rate of $5.0 \times 10^{-5}$ s$^{-1}$ showed mostly dimple fracture surfaces as in air, but quasi-cleavage cracks were observed in the portions considered to be crack initiation portions (Fig. 7(b) right side QC of site B). In the specimen with a strain rate of $7.5 \times 10^{-6}$ s$^{-1}$ shown in Fig. 7(d), quasi-cleavage cracks occupied a lot, and the dimple fracture surface was seen only in a part. Plastic bond area (tear ridge) was also observed in the quasi-cleavage crack (Fig. 7(d) site A arrow). Quasi-cleavage cracking is a brittle fracture which does not cause macro-

irregularities in morphology, but unlike cleavage to a (001) plane which is observed by low temperature brittleness. In a martensitic structure, it is reported that it is fracture at the sliding surface along the lath boundary in the (110) plane.$^{20-22}$ On the other hand, SUH 660 having an internal hydrogen content of 37 mass ppm and 67 mass ppm by hydrogen charging has been reported to exhibit intragranular fracture due to sliding surface separation along the (111) plane.$^{14,15}$ In the enlarged photograph (within the dashed white frame) of part A in Fig. 7(d), a sliding line pattern is
 acknowledged in the grain. In the test piece with a strain rate of 1.25 × 10^{-6} s^{-1} shown in Fig. 7(0), quasi-cleavage cracks almost occupied and no obvious dimple fracture surface was observed. Crack propagated along the grain boundary to a part of the quasi-cleavage crack, and a part of intergranular crack (IG) with a tear ridge was also found (Fig. 7(f) part A dashed line within white frame).

In the slow strain rate tensile test, the approximate time from the start of loading to the maximum load under a hydrogen gas for the strain rate of 5.0 × 10^{-5}, 7.5 × 10^{-6}, and 1.25 × 10^{-6} s^{-1} are, 74, 489, and 2,820 min, respectively. These times are greatly different depending on the strain rate. Therefore, I compared the amount of hydrogen diffusing into each specimen at 150°C under a hydrogen gas of 70 MPa for each time, and consider the relationship between the fracture surface morphology and the hydrogen content.

For calculation of hydrogen content, the method of Matsuoka et al.^{15} was followed. First, the equilibrium hydrogen content Cs can be expressed by the following formulas (1) and (2).

\[
Cs = Ks \sqrt{f} \quad \ldots \ldots \ldots (1)
\]

\[
f = p \exp \left( \frac{p_b}{RT} \right) \quad \ldots \ldots \ldots (2)
\]

Ks: solid solubility, f: fugacity, p: hydrogen gas pressure, R: gas constant (\(=8.314\) [J K^{-1} mol^{-1}]), T: Absolute temperature [K], b: Constant (using the value 1.584 × 10^{-5} m^3 mol^{-1} obtained in A 286 by Marchi et al.^{23})

Also, when determining the solid solubility Ks and the diffusion coefficient D, the experimental values,^{15} Ks: 31.7 molH2 m^{-3} MPa^{-1/2}, Es: 1.51 kJmol^{-1}, Ds: 2.0 × 10^{-5} m^2 s^{-1}, Es: 64.3 k J mol^{-1} were used to obtain from the Arrhenius Eqs. of (3) and (4) below.

\[
Ks = K_0 \exp \left( -\frac{E_s}{RT} \right) \quad \ldots \ldots \ldots (3)
\]

\[
D = D_0 \exp \left( -\frac{E_s}{RT} \right) \quad \ldots \ldots \ldots (4)
\]

As a result, the diffusion coefficient D of hydrogen of SUH 660 in 70 MPa hydrogen gas at 150°C was 2.29 × 10^{-13} m^3 s^{-1} and the equilibrium hydrogen content Cs was 50.7 mass ppm. Using these values, the hydrogen content distribution in the test piece when the maximum load was reached was determined. For calculation, when hydrogen enters from the surface of the semi-infinite plate, the hydrogen content distribution \(C_s(x,t)\) in the material is given by the following formula (5) was used.

\[
C_s(x,t) = C_s \left[ 1 - \text{erf} \left( \frac{x}{2\sqrt{Dt}} \right) \right] \quad \ldots \ldots \ldots (5)
\]

Where \(x\) is the distance from the semi-infinite plate surface, \(t\) is the time during which the semi-infinite plate surface is exposed to hydrogen gas, \(C_s\) is the saturated hydrogen content, \(D\) is the hydrogen diffusion coefficient, and \(\text{erf}\) is the error function.

The results are shown in Fig. 8. At a maximum load of strain rate of 5.0 × 10^{-5}, 7.5 × 10^{-6}, and 1.25 × 10^{-6} s^{-1}, for example, in 70 MPa hydrogen gas at 150°C, the distance from the surface layer reaching the hydrogen content of 20 mass ppm were 0.04 mm, 0.1 mm, and 0.24 mm, respectively. According to the study by Matsuoka et al.,^{15} when a tensile test (strain rate 5.6 × 10^{-4} s^{-1}) of a test piece in which 67.0 mass ppm of hydrogen is uniformly contained by hydrogen charge is conducted at room temperature, its reduction area has decreased to 30.5%, and the RRA compared with the uncharged material is 52%. In addition, Imade et al. conducted a tensile test (strain rate 4.2 × 10^{-5} s^{-1}) of a charge material SUH 660 with an internal hydrogen content of 37 mass ppm at 27°C. and reported that RRA was less than 60%.^{14} These RRA values are approximately the same as the RRA (51%) of the tensile test at a strain rate of 7.5 × 10^{-6} s^{-1} in high pressure hydrogen gas at 70 MPa and 150°C. However, as can be seen from the simulation results in Fig. 8, the diffusion depth of hydrogen from the surface layer of the specimen with a strain rate of 7.5 × 10^{-6} s^{-1} is only about 0.25 mm at the maximum. The region with the same hydrogen content of 37 mass ppm as the above charge material is slightly 0.04 mm from the surface layer. The diffusion rate at which hydrogen collects at the tip of the crack at the time of crack propagation and the crack propagation speed at the time of fracture formation seemed to coincide exactly. As a result, it is inferred that the crack tip continued to be kept at the same hydrogen concentration as the charge material (uniform internal hydrogen content), and the same fracture mechanism and reduction behavior were obtained.

If the diffusion rate of hydrogen to the crack tip is lower than the crack propagation speed during the low strain rate tensile test, the dimple fracture surface will be obtained without much being influenced by hydrogen (In the case of the strain rate of 5 × 10^{-5} s^{-1}, the maximum diffusion depth of hydrogen into the interior is about 0.1 mm and it shows...
the dimple fracture surface).

On the other hand, if the crack propagation speed related to the strain rate during the test is enough slow and the diffusion of hydrogen to the crack tip is sufficiently carried out (in the case of the strain rate of $1.25 \times 10^{-6} \text{ s}^{-1}$), the reduction of area decreases to 15%, and the RRA decreases to 32%. This is apparently thought to be the result of being influenced greatly by hydrogen than 50.7 mass ppm charge material which is saturated at 150°C and in 70 MPa hydrogen gas. Since the strain rate is so slow, it is considered that hydrogen above the equilibrium state also gathers at grain boundaries and effects other than the stress concentration part of the crack tip, thereby decreasing the reduction of area. Crack propagation along the grain boundary was also observed in part of the quasi-cleavage crack.

Matsumoto et al.\textsuperscript{24}) discusses the difference between the hydrogen charge material and the material tested in the high pressure hydrogen environment with respect to the SUS 316L series material. From the fracture surface morphology after the tensile test, they mentioned the possibility that the hydrogen concentration at the crack tip generated during the slow strain tensile test in 90 MPa hydrogen gas is higher than the hydrogen charge of about 80 mass ppm. This is because in a high pressure hydrogen gas, the passive film on the surface of the test piece is destroyed due to tensile deformation, the newly formed metal surface is exposed, and impurities such as oxygen are hardly present, so that the passive film does not regenerate. Since the newly generated surface is maintained at the tip, the amount of penetration of hydrogen is remarkably increased.\textsuperscript{25}) It is considered that the hydrogen concentration at the crack tip becomes higher as compared with the case where hydrogen is charged in a static state. In addition, it is confirmed by hydrogen micro-print method that hydrogen is accumulated in the stress concentration part like notch bottom.\textsuperscript{26,27}) Microscopically, it is considered that dislocation density increases due to strain, defects such as new vacancies are generated, hydrogen is trapped by them in the state of atoms or molecules, and the hydrogen content increases.\textsuperscript{28–31}) Macroscopically, it is conceivable that hydrogen accumulates in grain boundaries with a high diffusion rate, weakening the strength of grain boundaries, and inducing grain boundary cracking. In this study grain boundary cracking accompanied by tear ridge was observed in a part of the case with the slowest strain rate ($1.25 \times 10^{-6} \text{ s}^{-1}$).

In this study, I found the possibility that the hydrogen concentration at the crack tip of SUH 660 can be greatly different depending on the strain rate. Therefore, from the results of this experiment, the relationship between the fracture surface morphology and the estimated crack tip hydrogen content depending on the strain rate and RRA during slow strain rate tensile test in 70 MPa hydrogen gas at 150°C is summarized in Fig. 9. At a strain rate of about $5.0 \times 10^{-5} \text{ s}^{-1}$, the hydrogen content at the crack tip is about 5 ppm and the RRA is 80%, indicating the dimple fracture surface, and the influence of hydrogen hardly appears. When the strain rate is about $5.0 \times 10^{-6} \text{ s}^{-1}$, the hydrogen content at the tip of the fracture from the comparison of the fracture surface morphology with the hydrogen charge material (Fig. 8) can be estimated to be about 10 times 50 ppm. Also, the RRA is about 50% and it shows quasi-cleavage cracks. When the strain rate is reduced from $5.0 \times 10^{-5} \text{ s}^{-1}$ to $5.0 \times 10^{-6} \text{ s}^{-1}$, and the exposure time becomes 10 times, the total amount of hydrogen inside the specimen given by Eq. (5) becomes about twice. In addition, it is inferred that the hydrogen content at the crack tip increased to about 10 times due to the movement of dislocation by the longer plastic deformation time and the increment of hydrogen concentration\textsuperscript{31}) at the crack tip by micro crack. When the strain rate of one digit is further delayed to about $5.0 \times 10^{-7} \text{ s}^{-1}$, the amount of hydrogen inside the specimen increases likewise, the plastic deformation time increases and the concentration of hydrogen also increases. Then the hydrogen content at the crack tip can be estimated to be about 500 ppm which is ten times larger, mostly indicating grain boundary cracking (RRA is about 20%). The analogous value of this hydrogen is higher than the surface hydrogen content obtained with the bulk material, but it can be considered as a value near the strain rate is one order of magnitude higher, as can be seen from the fracture surface morphology.

This difference in RRA and fracture surface morphology due to the strain rate apparently corresponds to the different behavior on the nominal stress-nominal strain curve in the air and high-pressure hydrogen gas in the region after about 90% after the maximum load point. However, when steel materials are used at hydrogen stations, since they are used in the low load region where the maximum load is divided by the safety factor, behavior analysis including the low load region is considered to be also important for practical application.

4. Conclusions

The iron based superalloy SUH 660 was subjected to a tensile test by changing the strain rate to $5.0 \times 10^{-5}$, $7.5 \times 10^{-6}$, and $1.25 \times 10^{-6} \text{ s}^{-1}$ under a 70 MPa high pressure hydrogen gas or air at 150°C, and the following findings were obtained.

(1) At each strain rate, the yield point and the nominal
stress-nominal strain curve up to the maximum load in the air and high pressure hydrogen gas is nearly equal, but the behaviors after the maximum load are different, the elongation at break tends to be slightly shorter in high pressure hydrogen gas than in air.

(2) The effect of strain rate on tensile strength was not observed under high pressure hydrogen gas and air. In addition, there was no difference in tensile strength under high pressure hydrogen gas and air.

(3) It was found that the reduction of area under high pressure hydrogen gas was greatly affected by the strain rate. When expressed by the RRA as the ratio of the reduction of area in high pressure hydrogen to that in air, it was revealed that it changed from 80% which is said to be not affected by hydrogen, to 32% to be greatly affected by hydrogen. To 32% to be greatly affected by hydrogen. The morphology of the fracture surface changes from dimple + partial quasi-cleavage, through quasi-cleavage + partial dimple, to quasi-cleavage + partial intergranular fracture.

(4) As a result of simulating the diffusion of hydrogen into the inside of the specimen during the tensile test, even at the sample (strain rate $7.5 \times 10^{-6} \text{ s}^{-1}$) showing almost a quasi-cleavage fracture surface, the region with the same hydrogen content as 37 mass ppm as the charge material was slightly only 0.04 mm from the surface layer at the maximum load. From this fact, it is considered that this is because defects such as new vacancies are generated at the tip of the crack at the time of crack propagation, hydrogen is trapped in the state of atoms or molecules, and the hydrogen concentration is increased. The degree of increase becomes more prominent as the strain rate becomes slower.

(5) The difference in these values of RRA was apparently a difference in the behavior after reaching the maximum load on the nominal stress-nominal strain curve, and it was also a behavior difference after 90% with respect to fracture elongation in air.

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