Rapid Evaluation of Hydrogen Embrittlement Resistance for Spot-Welds of High Tensile Strength Steel Sheet by Slow Rate Tensile Shear Test under Hydrogen Charging Conditions*

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Automobile manufacturers are accelerating adoption of spot welding of Advanced High-Strength-Steels (AHSS) sheets to reduce weight of automobile bodies. Rapid evaluation of the hydrogen embrittlement (HE) resistance for the spot-welds of AHSS sheets is required, since it is worried that the HE resistance of the nugget will deteriorate compared to the base metal due to the difference in microstructure caused by rapid cooling and solidification during spot welding. However, evaluation of the HE resistance for the spot-welds has not been established. In this study, we prepared spot-welded specimens using AHSS sheets and performed tensile shear tests with varying tensile rates under hydrogen charging to evaluate the relationship between diffusible hydrogen content and tensile shear strength. As a result, the tensile shear strength of spot welds decreased as the amount of diffusible hydrogen increased. The quasi-cleavage fracture surface and intergranular fracture surface were observed at the nugget and inside the crack generated at the nugget-heat affected zone interface. Furthermore, as the results of crack growth behavior and hydrogen thermal desorption spectroscopy analysis, hydrogen embrittlement in spot welds can be attributed to the stress-induced diffusion of hydrogen and the hydrogen trapped in dislocation and vacancy clusters at crack tips. [doi:10.2320/matertrans.MT2021016]

Keywords: hydrogen embrittlement, spot-welds, tensile shear test, crack, diffusible hydrogen

1. Introduction

Automobile manufacturers are increasingly adopting 980 MPa-grade advanced high-strength steel (AHSS) sheets and 1500 MPa-grade hot-stamping steel to reduce the weight of automobile bodies.1) Spot welding, which is used for fabricating joints from conventional steel sheets, is also used to fabricate AHSS joints and hot-stamping sheets. However, molten zones (nuggets) in spot-welded AHSS sheets undergo rapid cooling and solidifying; in addition, AHSS sheets have a relatively low hydrogen-embbrittlement sensitivity. These factors result in changes in their crystalline structure and hardness,2) thus raising concerns on increased hydrogen-embbrittlement sensitivity around the nugget. It is known that when fusion welding (e.g. arc welding) is conducted in conditions in which the hardness of and residual stress in welds are high, cracking occurs due to hydrogen generation, moisture invasion from the ambient environment, and from the filler metal in the welding process.3) Several studies4-10) were conducted on hydrogen-induced cracking in the arc welds of high-strength steel plates in the past. It has been reported that the strength of steels, diffusible hydrogen content, and residual stresses in welds significantly influence their hydrogen-embbrittlement sensitivity. Although there are many studies on hydrogen embrittlement and the associated mechanisms during arc welding of thick plates, few studies have evaluated hydrogen-embbrittlement resistance and the mechanism of embbrittlement in spot welds on thin plates.11,12) When spot welding different material combinations, it is necessary to determine the appropriate welding conditions to satisfy design requirements. This is necessary to ensure that embbrittlement is not caused by hydrogen entering during the painting process or usage environment. Therefore, to determine the appropriate welding conditions to prevent hydrogen embrittlement in spot welds, an evaluation method that can rapidly and quantitatively evaluate the effect of diffusible hydrogen on the ultimate strength of welded joints is required.

U-bending,13-16) 4-point bending,17,18) and tensile tests17-19) are often used to test high-strength steel sheets. In particular, from the tensile data of AHSS sheet specimens containing hydrogen (from tests such as the slow strain-rate method (SSRT), which tests at a strain rate of $10^{-3}$/s or lower,20) conventional strain-rate test (CSRT)21) which tests at a tensile rate of $\sim 1$ mm/min, and the constant load test (CLT),22,23) which evaluates fracture/unbroken samples by holding at a constant load), it is possible to analyze the critical diffuse hydrogen content or the fracture limit diagram19,22-24) showing the relationship between fracture stress and diffusible hydrogen content. In tensile tests on general AHSS test pieces19) without a notch, the fracture limit diagrams obtained using SSRT and CLT were quite similar, which is considered to be relatively close to the actual usage environment, such as a corrosive atmospheric environment, and practical values of load applied to the product; thus, it was concluded that SSRT is the most effective rapid evaluation method.

To evaluate the static strength of spot welds in thin sheets, cross-tension, coach peel, and tensile-shear tests are often conducted.25) Recent investigations demonstrated that during the coach peel test, the failure criterion of resistance of spot welds was lower under cathodic hydrogen charging than in air.11) However, no other major test method has been investigated to analyze the effect of hydrogen on the failure resistance of spot welds.

Therefore, in this study, tensile-shear tests were conducted on specimens fabricated from spot-welded AHSS sheets. During the test, diffusible hydrogen was introduced to quantitatively evaluate the relationship between diffusible hydrogen content and tensile-shear strength. In addition, tensile-shear tests were performed in different hydrogen-

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charging conditions and tensile rates to evaluate the effect of tensile rate on diffusible hydrogen content and tensile-shear strength. Furthermore, we studied the fracture morphology and fracture process to evaluate the hydrogen-embrittlement mechanism in the spot welds.

2. Experimental

2.1 Fabrication of spot-weld specimens

AHSS sheets (1500 MPa grade) with a known hydrogen-embrittlement sensitivity were used in this study. Their chemical composition is shown in Table 1.

| C   | Si | Mn | P  | S  | Cr | Al | Ti | V  | Nb |
|-----|----|----|----|----|----|----|----|----|----|
| 0.17| 0.49| 1.58| 0.010| 0.002| 0.03| 0.042| 0.005| 0.008| 0.013|

To evaluate only the influence of tensile speed during tensile-shear testing, two types of test equipment (which will be described later in this report) and two types of specimens (Figs. 1 and 2) were used. Specimen 1 (Fig. 1) is similar to specimens generally used in tensile-shear tests, but there is a concern that it may be affected by twisting or bending depending on the ability of the tester during setting. Meanwhile, with specimen 2 (Fig. 2), it is expected that the operator's influence is reduced because the specimen is supported with an insulating pin (Ø10 mm). The total length and shape of the chucking part in the two specimens were different, but the joints were similar in shape. To fabricate Specimen 1, 30 (width) × 100 (length) × 1.2 (thickness) mm³ AHSS sheets were used, while for specimen 2, 30 (width) × 60 (length) × 1.2 (thickness) mm³ sheets were used. Both specimens were resistance spot-welded using an electrode with a tip diameter of 6 mm, with a 30 mm overlap between the sheets. All welding operations were conducted at a constant force of 3.5 kN and current of 5 kA for 60 ms. The resulting nominal weld nugget diameter was ~3.1 mm.

2.2 Microstructure observation and Vickers hardness measurement

The microstructures of spot welds in the fabricated specimens were observed; each specimen was cut to include the nugget, embedded in epoxy resin, mirror-polished, and etched with 2.5% nital for metallographic observation by optical microscopy (GX53, OLYMPUS Corp.) and scanning electron microscopy (SEM, S-3600N, Hitachi High-Tech Corp.). Subsequently, specimen microhardness was measured using a Vickers microhardness testing machine with a 4.9 N load.

2.3 Evaluation of hydrogen-embrittlement resistance

The failure criterion of spot welds was evaluated by a tensile-shear test. To analyze the influence of diffusible hydrogen on the tensile-shear strength of the specimens, pre-hydrogen charging was conducted in advance for hydrogen saturation in the specimens, after which tensile-shear tests were conducted at different tensile speeds under the same hydrogen-charging conditions.

2.3.1 Hydrogen charging

The cathodic hydrogen-charging method was used for pre-hydrogen charging and hydrogen charging (Table 2) during tensile-shear tests. In condition A, tests were conducted in air at 23 ± 2°C without hydrogen charging. In order to control the diffusible hydrogen content while preventing specimen corrosion, hydrogen charging was performed by two methods – (1) specimens were subjected to a constant polarization potential of 1 V vs. Ag/AgCl in aqueous 0.1 M NaOH (condition B) and (2) specimens were subjected to a constant polarization current at a current density of 1 mA/cm² with 3% aqueous NaCl in NH₄SCN (0–30 g/L, corresponding to condition C–F) using a potentiostat (HA-151B, Hokuto Denko). NH₄SCN was added as a hydrogen-invasion promoter.

To analyze the hydrogen-embrittlement resistance of only the spot-welded region, the specimens were covered with masking tape and masking rubber, except in the overlap region (part a in Figs. 1 and 2). Hydrogen charging was performed only on the unmasked part (18 cm²). Per test piece, 1 L of the hydrogen-charging solution was required. A lead wire was spot-welded to the test piece and connected to a potentiostat for hydrogen charging.
2.3.2 Tensile-shear testing with hydrogen charging

After pre-hydrogen charging Specimen 1 and 2 for 24 h, tensile-shear tests were performed. Tests were conducted at tensile rates of 1 and 0.005 mm/min. The former corresponds to the condition generally used in tensile-shear tests and the latter corresponds to the rate used in SSRT tests to evaluate delayed fracture in steel sheets.19) In this study, the test conducted at 1 mm/min is called C-TST (conventional tensile-shear testing), and that conducted at 0.005 mm/min is called S-TST (slow displacement rate tensile-shear testing). Figure 3 shows a schematic illustration of the S-TST (in-house designed and manufactured). In this test, the specimen (Fig. 2) and test jig were installed in a tank containing the test solution. After installing the pre-hydrogen-charged specimen in the test jig, the test solution was poured into the tank. After connecting the test piece, reference electrode, counter electrode, and potentiostat, tensile-shear tests were conducted under hydrogen charging (similar to pre-hydrogen-charging conditions). C-TST was completed within 10 min and hence, it was considered that the effect of hydrogen diffusion was minimal in this test. Therefore, after pre-hydrogen-charged test pieces (Figs. 1 and 2) were installed in the test jig, C-TST was conducted without hydrogen charging. The test was conducted on a universal testing machine (AG-50KNI, Shimadzu Corporation) with a servo pulsar (NJ-30kNV-50, Shimadzu Corporation). After the tests were completed, the specimens were immediately placed in liquid nitrogen to prevent the dissipation of diffusible hydrogen. After S-TST and C-TST tests, thermal desorption analysis (TDA) was conducted to measure the amount of diffusible hydrogen; the fracture surface and crack generation/growth behavior of the spot welds were observed to understand the correlation between diffusible hydrogen content and the mechanism of crack initiation and propagation.

2.4 Thermal desorption analysis of hydrogen

The nugget-containing specimens stored in liquid nitrogen were cut to ~7 × 30 mm² to perform TDA. A gas chromatograph-type hydrogen analyzer (SGHA-P2, FIS) was used for TDA testing. During testing, the temperature was increased from room temperature to 200°C at 2°C/min in an Ar stream (400 cc/min) to evaluate the hydrogen thermal-desorption spectrum. The spectrum thus obtained was integrated in the range of 25–150°C to evaluate the amount of diffusible hydrogen occluded in the specimens. In addition, higher resolution hydrogen TDA was conducted on a high-frequency heating-type thermal-desorption spectrometer (IH-TDS1700/ESCO) to investigate trap sites for diffusible hydrogen occluded in the spot weld after testing. In this case, the temperature was increased from −50 to 200°C at a rate of 2°C/min in high vacuum.

2.5 Fracture-surface morphology and crack generation/growth behavior (fractography)

The morphology of fracture surfaces obtained at the end of tensile-shear testing was observed by scanning electron microscopy (SEM). In the case of S-TST specimens tested under condition C, welded structures (in the cross-section perpendicular to the welding direction), whose testing was interrupted at a predetermined load, fracture surfaces were observed with an optical microscope to investigate crack generation/growth behavior.

3. Results

3.1 Microstructure and Vickers hardness of the fabricated spot welds

The cross-sections of welded structures (in a direction perpendicular to the welding direction) were observed by SEM and optical microscopy, as shown in Fig. 4. The base material, heat-affected zone (HAZ), and nugget were mainly lath martensite structures. A columnar structure reflecting the solidification direction from the interface with HAZ toward the final solidification part was observed inside the nugget.
The area around the final solidification section was an equiaxed crystal. The average Vickers hardness of both the nugget and base metal was ~500 HV4.9. Further, as the HAZ was annealed during welding, its hardness was ~300 HV4.9 at the interface between the HAZ and base metal.

### 3.2 C-TST and S-TST analysis

Figure 5 shows the variation in the tensile-shear strength of spot welds tested by C-TST and S-TST at different hydrogen-charging conditions. In these tests, strength was calculated by dividing the maximum load by the area of the nugget after fracture. In C-TST, the tensile-shear strength obtained in conditions C and D was the same as that in air (condition A); however, the strength obtained in conditions E and F was lower than that in air. Meanwhile, in S-TST, the strength obtained in conditions B–F was lower than that in condition A.

Figure 6 shows the variation in the diffusible hydrogen content of the specimens occluded in C-TST and S-TST analysis in various hydrogen-charging conditions. From the figure, it can be noted that there was no significant difference between C-TST and S-TST specimens. More diffusible hydrogen was accumulated upon constantcurrent polarization in 3% NaCl than with constant potential polarization in 0.1 M NaOH. In addition, in the former case, the diffusible hydrogen content increased significantly when NH₄SCN was added as a hydrogen-invasion promoter.

Figure 7 illustrates the relationship between tensile-shear strength and diffusible hydrogen content in C-TST and S-TST specimens. In C-TST, strength did not decrease up to a hydrogen content of 0.5 wt. ppm, but beyond this value, specimen strength decreased. Meanwhile, in S-TST, tensile-shear strength decreased gradually up to a diffusible hydrogen content of 0.003 wt. ppm, but beyond this value, it decreased sharply. In C-TST, the trends of tensile-shear strength in specimens 1 and 2 were similar. Therefore, it is judged that the influence of specimen shape on both C-TST and S-TST was small. The strength of S-TST specimens, which were tested at a lower tensile rate, was significantly lower than that of C-TST specimens. In other words, S-TST specimens were more sensitive to hydrogen embrittlement than C-TST specimens.

### 3.3 Fractography

Figure 8 shows the SEM images of the macro-fracture surfaces of spot welds after C-TST and S-TST at different hydrogen-charging conditions. Except in the S-TST specimen produced in condition F, the vertical direction of the photograph represents the load direction and the upper end of the nugget is the stress-concentration region. In C-TST specimens, it was observed that in conditions A–E, interface fracture occurred at the nugget. Meanwhile, in condition F, the main crack generated from the stress-concentration region propagated inside the nugget while diverting in the direction...
of sheet thickness inside the nugget, resulting in an uneven fracture-surface morphology. In addition, a flat fracture surface was observed in condition A (without hydrogen charging) and conditions C and D (less than 0.5 wt. ppm diffusible hydrogen with no decrease in strength). Meanwhile, a large number of cracks were observed extending in the direction orthogonal to the fracture surface near the stress-concentration region and at the interface between the nugget and HAZ on the fracture surface in condition E; in this instance, the tensile-sheet strength of the specimen decreased. In condition F, many cracks were observed near the stress-concentration region and large cracks were confirmed inside the nugget. In S-TST specimens, it was observed that interface fracture occurred at the nugget in conditions A and B. Meanwhile, in conditions C–E, the main crack generated in the stress-concentration region propagated inside the nugget and diverted in the direction of sheet thickness, resulting in an uneven fracture-surface morphology. In condition F, unlike in other specimens, the base metal underwent fracture due to the main crack extending in the sheet-thickness direction from the interface between the nugget and HAZ.

Figure 9 shows the SEM images of microfracture surface near the stress-concentration region (upper part of the nuggets in Fig. 8). Based on these observations, the macrofracture surfaces were classified into the following two regions in Fig. 8. The region in which ductile fracture was observed is surrounded by a red dotted line and the region in which grain-boundary and quasi-cleavage fracture were observed are surrounded by blue dashed lines. As dimples were found to extend in the load direction, ductile fracture occurred in the test piece whose tensile-sheet strength did not decrease. In both C-TST and S-TST specimens, ductile-fracture surfaces extended in the load direction inside nuggets in test pieces with similar strength (C-TST: condition E; S-TST: condition B). In addition, a combination of grain-boundary- and quasi-cleavage-fracture surfaces was observed inside cracks at the nugget/HAZ interface.

The above-described grain-boundary- and quasi-cleavage-fracture surfaces were observed in specimens experiencing a relatively large decrease in tensile-sheet strength (C-TST: condition F; S-TST: conditions C–E). In particular, the C-TST fracture surfaces tended to be dominated by the grain-boundary-fracture surface at the end of the nugget and quasi-cleavage fracture inside the nugget. In C-TST (condition F) and S-TST (conditions C and D), a ductile fracture surface (toward the bottom of the photograph) was observed inside the nugget. In addition, the ductile-fracture area decreased as tensile-sheet strength decreased. In S-TST (condition F), quasi-cleavage fracture occurred in the vicinity of the-stress concentration region where cracks occurred and grain-boundary fracture was observed in the base metal region.
In short, the micro- and macro-fracture-surface morphologies of test pieces with the same tensile-shear strength (e.g. C-TST specimens in condition E and S-TST specimens in condition B; C-TST specimens in condition F and S-TST specimens in condition C) were similar. Therefore, it can be said that the tensile-shear strength and fracture-surface morphology are strongly correlated.

3.4 Crack generation/growth behavior in S-TST

Figure 10 shows the typical tensile-shear stress–displacement curves of spot welds in S-TST specimens in conditions A and C. In condition A, after the specimen exhibited the maximum tensile shear strength, local deformation (region corresponding to stress reduction) occurred, leading to rapid fracture. Meanwhile, in condition C, the slope of the curve was smaller than that in condition A at low stresses (over 400 MPa). In both conditions A and C, after the specimens exhibited the maximum tensile-shear stress, local deformation and fracture occurred.

To understand the reasons for the lower tensile-shear strength in condition C when compared to that in condition A, crack occurrence and propagation were studied after stopping the test at three stress levels, as shown in Fig. 10. The three main features in this figure are as follows. Point a: Region in which the slope of the curve is smaller than that in condition A. Point b: maximum tensile-shear stress, and Point c: before rapid fracture. Figure 11 shows photos of the cross-sections of spot welds after stopping tensile-shear testing at different stress levels. At point a (Fig. 11(a)), initial cracks (main cracks) with a length of ~200 µm originated from the end of the nugget and extended in a 45° direction, which is the shear direction. At point b (Fig. 11(b)), the main crack grew along sheet thickness and reached the specimen surface at the maximum tensile-shear stress. At point c (Fig. 11(c)), where local deformation occurred, a crack that extends from the nugget interface to the HAZ, a crack that branches from the main crack, and cracks that occur from the blow hole in the center of the nugget could be observed.

The fact that the main crack reached the surface of the specimen at the maximum tensile-shear stress when subjected to hydrogen charging suggests that diffusible hydrogen is involved in the generation and growth of main cracks and the decrease in tensile-shear strength. The reason for the occurrence and propagation of main cracks (Fig. 11(a)) in a direction angled at 45° from the nugget end can be explained by considering the plastic yield behavior of the stress-concentration region in Fig. 12. Figure 12(a) shows a schematic diagram of the initial cracks caused by stress concentration at the nugget end. In general, when stress...
concentration occurs at the end of the nugget, slip deformation occurs in the 45° direction, as shown in Fig. 12(b), and a plastic region is formed. A hydrogen-accumulation area is formed in the immediate vicinity of the nugget end during hydrogen absorption under load. As many dislocations accumulated at the intersection of the slip line and grain boundary, the latter becomes particularly brittle and grain-boundary fracture may occur. In an environment where diffusible hydrogen is continuously supplied to the stress-concentrated region, as grain boundaries experience hydrogen embrittlement continuously at the stress-concentration region and crack tip, fracture occurs at a lower stress level than when the samples are exposed to air. In addition, diffusible hydrogen is supplied at high stress levels even after the main crack reaches the specimen surface. Therefore, it is expected that another crack that occurs during local deformation grows along embrittled grain boundaries.

4. Discussion

It was found that C-TST and S-TST with the same tensile-shear strength exhibited the same fracture-surface morphology. Therefore, it can be concluded that the fracture modes of both tensile-shear tests are similar. However, because the tensile-shear strength obtained in S-TST is significantly lower than that obtained in C-TST at the same diffusible hydrogen content, the observed hydrogen-embrittlement susceptibilities were different. In this section, we shall discuss the reasons for such differences and analyze which method can quantitatively and rapidly evaluate the effect of diffusible hydrogen on tensile-shear strength.

According to the results described in Section 3, the relationship between diffusible hydrogen content, tensile-shear strength, and fracture-surface morphology can be summarized as shown in Fig. 13. As C-TST is performed at a tensile speed (strain rate) of ~1 mm/min, the test time is as short as ~1 min. As the diffusion rate of diffusible hydrogen cannot follow dislocation pile-ups, the amount of hydrogen accumulated in the stress-concentration region decreases. Meanwhile, when the tensile speed (strain rate) is slow (S-TST), the movement of diffusible hydrogen follows the generation and movement of dislocations, resulting in stress-induced diffusion. Therefore, even at a constant diffusible hydrogen content, specimen strength is lower at slower tensile speeds as more hydrogen accumulates in the stress-concentration region. It is considered that the same phenomenon occurred in the tensile-shear testing of spot welds in this study. In S-TST, because hydrogen accumulation occurred due to the stress-induced diffusion of diffusible hydrogen to the stress-concentrated region at the nugget end and to the crack tip and trapping of diffusible hydrogen in dislocations, tensile-shear strength decreased even at low diffusible hydrogen contents. In addition, quasi-cleavage- or grain-boundary fracture surfaces could be observed in the initial and secondary cracks.

Meanwhile, in C-TST, assuming that the cracks were broken before hydrogen accumulation occurred at the crack tips, the initial cracks broke down mainly due to lattic embrittlement due to diffusible hydrogen introduced uniformly during pre-hydrogen charging. The amount of hydrogen required to observe the effects of hydrogen embrittlement on the generation of initial cracks near the nugget end was ~1 wt. ppm. However, because diffusible hydrogen cannot follow crack growth, ductile fracture is likely to occur, except at the nugget end.

Meanwhile, in S-TST, hydrogen supplied from the test environment can be sufficiently diffused, so even if the average concentration is ~0.003 ppm, it is speculated that the accumulated hydrogen (~1 wt. ppm) in the stress-concentration region causes hydrogen embrittlement. In a previous report, hydrogen accumulation near a blunting crack tip was simulated based on Oriani’s theory using the known material properties of AHSS sheets. According to the authors, the proportion of hydrogen-trapped dislocations due to stress-induced diffusion increases by ~1000 times compared to that observed in the absence of load at the blunting crack tip. The amount of accumulated hydrogen at the crack tip was estimated as 3 wt. ppm. This result is not significantly different from the expected value, thus indicating the validity of this consideration. However, to

![Fig. 13](image-url) Relationship between diffusible hydrogen content, tensile shear strength and fracture surface.
exponentially validate hydrogen accumulation, quantitative hydrogen analysis should be carried out in local regions and this aspect will be considered in our future study. Therefore, the variation in hydrogen-embrittlement susceptibility with respect to tensile rate is closely related to the accumulation of hydrogen trapped in trap sites such as dislocations in the stress-concentration region. If the diffusion of hydrogen follows the movement of dislocations and increases in the stress-concentrated region, tensile-shear strength decreases even at low diffusible hydrogen contents. Therefore, special attention should be paid to locations where stress concentration is likely to occur, such as the nugget end and crack tip of the spot weld because hydrogen accumulation is likely to occur locally.

To further analyze the dependence of hydrogen accumulation with respect to tensile speed, in Fig. 14, we show the temperature-desorption spectra of the nugget and base material of C-TST and S-TST specimens under condition C. No differences were observed in terms of shape, peak temperature, and peak intensity of the hydrogen thermal-desorption spectra of the base metal, but the shape of the spectra in the nugget varied depending on the test method. The thermal desorption spectrum of C-TST exhibited two peaks, one at the same temperature as the base metal (~60°C), and the other a sharp peak at about 80°C. In S-TST, a relatively broad peak was observed on the high-temperature side (80°C) of the nugget spectrum when compared to the spectrum of the base metal. The peak on the low-temperature side (~60°C) corresponds to diffusible hydrogen trapped in weak trap sites such as interstitial sites and vacancies, and the peak on the high-temperature side (~80°C) corresponds to hydrogen trapped in vacancy clusters.

In C-TST, most of the diffusible hydrogen was trapped in weak sites such as interstitial spaces and vacancies as the tensile speed was high and the number of vacancy clusters could not be increased. Meanwhile, in S-TST, because the tensile speed is low, there is sufficient time for stress-induced diffusion. In addition, because the diffusible hydrogen can follow the movement of dislocations, vacancy clusters are generated at the crack tip and a large amount of diffusible hydrogen is trapped there. Thus, the thermal desorption spectrum shown in Fig. 14 was obtained.

The effect of diffusible hydrogen on the reduction in tensile-shear strength and fracture-surface morphology can be explained by the diffusible hydrogen trap sites identified by thermal-desorption analysis. Therefore, it can be understood that the dependence of tensile-shear strength on tensile speed with respect to hydrogen embrittlement of the spot welds in AHSS sheets is similar to that of hydrogen embrittlement in general high-strength steels. In addition, when evaluating the hydrogen-embrittlement resistance of spot welds by tensile-shear tests, S-TST is more rapid and effective than C-TST.

5. Conclusions

In this study, tensile-shear tests were conducted at different diffusible hydrogen contents and tensile speeds to evaluate the hydrogen-embrittlement resistance of spot welds in AHSS sheets. The major conclusions are as follows.

1. The tensile-shear strength of AHSS specimens decreased as the diffusible hydrogen content increased. In specimens with low tensile-shear strength, quasi-cleavage- and grain-boundary-fracture were observed inside cracks and on the surfaces.

2. During the tensile-shear testing of spot welds, the lower the tensile speed, the lower is the obtained tensile-shear strength, even at small diffusible hydrogen contents. In other words, because hydrogen-embrittlement sensitivity is highly evaluated, it can be said that the hydrogen-embrittlement resistance of spot welds can be evaluated to ensure a high safety margin.

3. Tensile-shear strength is defined as the strength at which the initial crack occurs and propagates. In addition, in specimens showing the same tensile-shear strength at different tensile speeds, both the macro- and micro-fracture surfaces exhibited similar morphologies.

4. It could be concluded from fractography and hydrogen trap-site analysis that differences in the hydrogen-embrittlement susceptibility of the specimens at different tensile speeds are due to the migration and accumulation of diffusible hydrogen and dislocations. It is presumed that the hydrogen embrittlement of spot welds of AHSS sheets follows the mechanism in general high-strength steels. In other words, hydrogen embrittlement of spot welds occurs due to the stress-induced diffusion of hydrogen to the crack tip and accumulation of dislocations and diffusible hydrogen trapped in vacancy clusters.

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REFERENCES

1. K. Saito: Materia Japan 53 (2014) 584–588.
2. T. Sadasue, S. Igi, K. Taniguchi, R. Ikeda and K. Oi: Q. J. JWS 32 (2014) 64–72.
3. H. Suzuki and N. Yurioka: Tetsu-to-Hagané 67 (1981) 1657–1669.
4. N.G. Alcantara and J.H. Rogerson: Weld. J. 63 (1984) 116s–122s.
5. P.H.M. Hart: Weld. J. 65 (1986) 14s–22s.
6. E. Takahashi and K. Iwai: J. JWS 48 (1979) 865–872.
7. T. Yatake, N. Yurioka, R. Kataoka and E. Tsunetomi: J. JWS 50 (1981) 291–296.
8. H.J. Kim and B.Y. Kang: ISIJ Int. 43 (2003) 706–713.
9. Y. Kitagawa, K. Ikeuchi, T. Kuroda, Y. Matushita, K. Suenaga, T. Hidaka and H. Takauchi: J. Mater. Sci. 43 (2008) 12–22.
10. N. Mukai, Y. Inoue, S. Sasakura and Y. Kinoshita: Q. J. JWS 38 (2020) 41–51.
11. K. Fujita, M. Sawai, M. Watanabe and S. Hashimoto: Preprints of the National Meeting of JWS 101 (2017) 182–183.
12. O. Schwedler, M. Zinke and S. Jütten: Weld. World 58 (2014) 339–346.
13. Y. Toji, S. Takagi, M. Yoshino, K. Hasegawa and Y. Tanaka: Tetsu-to-Hagané 95 (2009) 887–894.
14. Y. Shibayama, T. Hojo and E. Akitama: Tetsu-to-Hagané 105 (2019) 927–934.
15. S. Takagi, Y. Toji, K. Hasegawa, Y. Tanaka, N. Roessler, B. Hammer and T. Heller: Int. J. Automot. Eng. 1 (2010) 7–13.
16. S. Takagi, Y. Toji, M. Yoshino and K. Hasegawa: ISIJ Int. 52 (2012) 316–322.
17. T. Hojo, H. Waki and F. Nishimura: Tetsu-to-Hagané 100 (2014) 1306–1314.
18. S. Takagi, Y. Hagiwara, T. Hojo, W. Urushihara and K. Kawasaki: ISIJ Int. 56 (2016) 685–692.
19. A. Tsuji, T. Suzuki, G. Kitahara and T. Asada: Trans. JSME 85 (2019) 19-00168 (in Japanese).
20. M. Wang, E. Akiyama and K. Tsuzaki: Corros. Sci. 49 (2007) 4081–4097.
21. Y. Hagiwara, C. Ito, N. Hisamori, H. Suzuki, K. Takai and E. Akiyama: Tetsu-to-Hagané 94 (2008) 215–221.
22. S. Takagi, T. Inoue, T. Hara, M. Hayakawa, K. Tsuzaki and T. Takahashi: Tetsu-to-Hagané 86 (2000) 689–696.
23. S. Yamasaki and T. Takahashi: Tetsu-to-Hagané 83 (1997) 454–459.
24. T. Chida, Y. Hagiwara, E. Akiyama, K. Iwanaga, S. Takagi, H. Ohishi, M. Hayakawa, D. Hirakami and T. Tarui: Tetsu-to-Hagané 100 (2014) 1298–1305.
25. M. Ono: J. JWS 77 (2008) 745–751.
26. T. Asada, G. Kitahara, T. Suzuki and A. Tsuji: Trans. JSME 83 (2017) 17-00012 (in Japanese).
27. R.A. Oriani: Acta Mater. 18 (1970) 147–157.
28. K. Takai: Zairyo-to-Kankyo 60 (2011) 230–235.