Improvement in Resistance to Disbonding of Stainless Steel-overlaid 2 1/4Cr-1Mo Steels

Jun-ichi SHIMOMURA, Yoshifumi NAKANO¹⁾, Syozaburo NAKANO¹⁾ and Syuzo UEDA¹⁾

Analysis and Material Science Research Center, Kawasaki Steel Corporation, Kawasaki-cho, Chiba, Chiba-ken, 260 Japan.
and Steel Research Laboratories, Kawasaki Steel Corporation, Kawasaki-cho, Chiba, Chiba-ken, 260 Japan.

¹⁾ Iron

(Received on September 5, 1990; accepted in the final form on November 16, 1990)

The effects of V on the disbonding characteristics of the 2 1/4Cr-1Mo steels overlaid by austenitic stainless steel were metallurgically studied. The addition of V to the 2 1/4Cr-1Mo steels significantly increases the resistance to disbonding. The scanning transmission electron microscope and energy-dispersive X-ray spectrometer observation showed that the V-modified and ordinary 2 1/4Cr-1Mo steels had different fine precipitates in terms of morphology, distribution and composition. The analyses of hydrogen diffusion and hydrogen evolution indicate that the fine precipitates in the V-modified steels trap hydrogen to increase the solubility of hydrogen and also decrease the diffusion of hydrogen. A theoretical calculation of hydrogen distribution during cooling from the atmosphere of elevated temperature and high pressure hydrogen indicates that both diffusion and solubility of hydrogen in the base metal affect the concentration of hydrogen at the interface between the base metal and the stainless steel overlay. It is concluded that the fine precipitates in the V-modified 2 1/4Cr-1Mo steel increase the resistance to disbonding by reducing the concentration of hydrogen at the interface during cooling.

KEY WORDS: pressure vessel steel; Cr-Mo steel; overlay disbonding; overlay welding; hydrogen diffusion; hydrogen trapping; hydrogen solubility.

1. Introduction

The operating temperature and hydrogen partial pressure of pressure vessels in oil refineries are currently 454°C (850°F) and 13.8 MPa (2 000 psi), respectively. They are expected to be raised to 482°C (900°F) and 20.7 MPa (3 000 psi) in near future. A 2 1/4Cr-1Mo steel which can be used in such severe operating conditions, therefore, is designed. Since these conditions exceed the safety limits for the 2 1/4Cr-1Mo steels currently used, as indicated in Nelson diagram,¹⁾ the steels for future pressure vessels should have high resistance to hydrogen attack.

It has been already reported that the addition of V is effective to improve the resistance to hydrogen attack²⁻⁴⁾ and the 2 1/4Cr-1Mo steel containing 0.25-0.35 mass% V is superior to the ordinary 3Cr-1Mo steel in the resistance to hydrogen attack.^{5,6,12)} In the operating conditions exceeding 454°C, furthermore, the creep rupture strength determines the allowable stress of 2 1/4Cr-1Mo steels. The addition of V also increases the creep rupture strength.^{7,8)} Since the favorable effect of V in the pressure vessel steels was recognized, MPC (The Materials Properties Council) under ASME and ASTM have been testing the V-modified 2 1/4Cr-1Mo steel to accept it in the ASME code.⁹⁾

The concentration of hydrogen dissolved in the pressure vessel wall increases as the operating temperature and hydrogen pressure are raised. Consequently a possibility of disbonding increases at the

interface between the base metal of Cr-Mo steel and the stainless steel welded overlay during cooling. Since disbonding is caused by the anomalous concentration of hydrogen at the interface due to the diffusion of hydrogen from ferrite phase to austenite phase after the shutdown, it can be avoided by preventing the concentration of hydrogen at the interface by lowering the cooling rate during the shutdown practice.¹⁰⁾ Since slow cooling requires longer time for shutdown, however, it is economically unfavorable. The resistance to disbonding can be controlled to some extent by introducing delta ferrite into the stainless steel welded overlay.⁴⁾ To avoid disbonding during a rapid cooling after shutdown, however, the improvement of weld metal alone is not sufficient and, thus, the improvement of base metal is also necessary. The addition of V to Cr-Mo steels changes the diffusion coefficient of hydrogen.²⁾ Consequently, the sensitivity to disbonding is expected to be changed. It has been reported, furthermore, that a 3Cr-1Mo-V-Ti-B steel has more resistance to disbonding than the conventional 2 1/4Cr-1Mo steel.^{7,9-11)} With the above-mentioned facts in mind, the disbonding characteristics of V-modified 2 1/4Cr-1Mo steel was studied.

2. Experimental Procedures

A V-modified and an ordinary 2 1/4Cr-1Mo steels melted in a 5 t vacuum induction furnace were hot-rolled into 50 mm thick plates. Their chemical compositions are shown in Table 1. Each plate was

Table 1. Chemical compositions of plate. (mass%)

	C	Si	Mn	P	S	Cr	Mo	V	Nb	Ti	B	Al	REM*	N
V-modified	0.125	0.09	0.49	0.005	0.001	2.35	1.00	0.29	0.027	0.009	0.001	0.019	0.008	0.0043
Ordinary	0.140	0.10	0.53	0.004	0.003	2.45	1.3	—	—	—	—	0.020	—	0.0037

* REM: Rare earth metals

Table 2. Chemical compositions of SUS347 welding material. (mass%)

C	Si	Mn	P	S	Ni	Cr	Nb	N
0.021	0.18	2.36	0.024	0.004	11.22	21.49	0.88	0.0235

heated to 1 000°C followed by cooling at a rate of 0.25°C/s which corresponds to that at the middle position of the through-thickness of a 300 mm thick plate during water quenching. The plate was then overlaid by ESW (Electroslag welding) with an SUS347 welding material whose chemical compositions are shown in Table 2 and was subjected to PWHT (Post-weld Heat Treatment) at 690°C for 43.2 ks (12 h). The disbonding test specimens whose geometries are shown in Fig. 1 were exposed to hydrogen at 400–500°C and the pressure of 9.8–24 MPa in an autoclave for 172.8 ks (48 h), followed by cooling in air according to a pattern shown in Fig. 2. The disbonding at the interface between the base metal and the welded overlay was examined by an ultrasonic C-scan method.

The diffusion coefficient and evolution of hydrogen in the base metal were measured to study the relation between them and disbonding. The diffusion coefficient of hydrogen was measured at 25°C by a cathode charging hydrogen permeation method using a 1 mm thick specimen whose one side was plated with Ni. Hydrogen was charged into the specimen from the non-plated surface in a (0.1N-NaOH+0.1N-NaCN) solution. The amount of hydrogen permeated through the specimen was measured in terms of the ion current in a 1N-NaOH solution. The change in hydrogen permeation with time was compared with the theoretical permeation curve to determine the diffusion coefficient of hydrogen. Hydrogen evolution during heating was measured in an apparatus shown in Fig. 3 using a 2 mm thick, 12 mm wide and 30 mm long specimen. Hydrogen was charged into the specimen at 25°C using the same solution as that for hydrogen diffusion coefficient measurement. The specimen was heated in vacuum at rates of 0.2 and 0.05°C/s to measure the evolution rate in terms of pressure difference in the tube. Before the measurement of diffusion coefficient and evolution of hydrogen, the specimen surface was polished with an emery paper and then electrochemically to remove the worked surface layer.

The shape, distribution and composition of precipitates in the base metal were examined by a 200 kV scanning transmission electron microscope (STEM) equipped with an energy-dispersive X-ray spectrometer (EDX) and an electron energy loss spectrometer (EELS) using carbon extraction replica specimens.

Fig. 1. Geometry of specimen for disbonding test.

Fig. 2. Cooling pattern after exposure to H₂ in disbonding test.

Fig. 3. Apparatus for hydrogen evolution test.

3. Experimental Results

3.1. Disbonding Test

The specimens of V-modified and ordinary 2 1/4 Cr-1Mo steels were removed from an elevated temperature and pressure hydrogen atmosphere (450°C, 14.7 MPa) in an autoclave and then air-cooled to room temperature. The microstructures near the interface between the base metal and the welded overlay are shown in Fig. 4. In both specimens, a C-concentrated layer which formed during PWHT is observable as a dark band at the interface. Disbonding

Fig. 4. Microstructure of the interface between base metal and austenitic stainless steel overlay weld metal.

is clearly observed along the coarse austenite grain boundary existing nearly parallel to the interface in the ordinary steel specimen, while no disbonding is observed in the V-modified steel specimen. Disbonding occurs within 7.2 ks (2 h) after the specimen is held at room temperature. It should be noted that disbonding occurs not in the C-concentrated layer but in the austenitic phase near the interface.

Fig. 5 shows the dependency of disbonding sensitivity on temperature and hydrogen pressure. In the case of ordinary steel, higher temperature and/or higher hydrogen pressure increase the sensitivity to disbonding. Disbonding occurs when the specimen is cooled from the conditions of 450°C and 14.7 MPa which are nearly same as the current operating conditions of hydrotreating reactors in oil refineries. On the contrary no disbonding occurs in the case of V-modified steel even after cooling from the conditions of 500°C and 19.6 MPa. These results indicate that the V-modified steel has much higher resistance to disbonding than the ordinary one.

3.2. Microstructure

High concentration of hydrogen at the interface between the base metal and the welded overlay is necessary for disbonding. The C-concentrated layer at the interface probably affects the sensitivity to disbonding. As already mentioned in Sec. 3.1, the sensitivity to disbonding depends on the composition of the base metal even if welding conditions are the

Fig. 5. Disbonding characteristics of V-modified and ordinary 2 1/4Cr-1Mo steels.

same. This indicates the possibility that the chemical composition of the base metal affects the sensitivity to disbonding through the composition of the interface.

Fig. 6 shows the one-dimensional composition profiles of C and Cr across the interface between the welded overlay and the base metal for the V-modified and ordinary steels. The maximum C concentration is slightly higher for the ordinary steel specimen than for the V-modified steel specimen. However, the thickness of C-concentrated layer is almost same for both steels. No significant difference in the C-concentrated layer exists for both specimens. Furthermore, the position of the maximum C concentration is located at the position of average Cr concentration for the welded overlay and the base metal and is different from that of disbonding in the welded overlay.

Therefore, the difference in sensitivity to disbonding between both steels is not attributed to the difference in the C-concentrated layer but to the difference in the concentration of hydrogen which diffuses from the base metal into the welded overlay during cooling from elevated temperature and high pressure hydrogen. It is considered that the difference is affected by microstructure *via* diffusion and solubility of hydrogen in the base metal.

In ferritic steels diffusion of hydrogen at low temperatures is retarded by the trapping effects of various lattice defects, among which the precipitate-matrix interface is noted as a large, effective trap site. The precipitates in the steels were observed by an STEM and an EDX using carbon extraction replica specimens.

Fig. 7 shows the STEM images and EDX spectra of the precipitates in the V-modified and the ordinary steels. In the ordinary steel, coarse granular carbides of 1–2 μm in diameter exist both inside grains and along grain boundaries and fine, 0.5 μm long needle-like carbides inside grains. The former are carbides with metallic compositions of Cr and Fe and are divided into two types with different concentrations of Cr and Fe. One is a Cr-rich Cr-Fe carbide of M_7C_3 type. The other is a Fe-rich Cr-Fe carbide of $M_{23}C_6$ type.⁴⁾ The fine needle-like carbides have metallic compositions of Cr and Mo with a crystal structure of M_2C type. They precipitate inside the

Fig. 6. EPMA analysis of elements at the interface between base metal and austenitic stainless steel overlay weld metal.

Fig. 7. STEM images and EDX spectra of precipitates in V-modified (a) and ordinary (b) 2 1/4Cr-1Mo steels.

grains and precipitation-free zones (PFZ) are observed along grain boundaries.

In the V-modified steel, on the other hand, fine granular precipitates of 0.2 μm in diameter and coarse granular carbides are observed. The quantitative analysis of metallic composition of the coarse granular carbides using the EDX indicated that the amount of V in the coarse granular carbides was as low as 10 mol% in spite of the addition of V. Consequently, the difference in coarse granular carbides between the V-modified and the ordinary steels is rather small. Fine granular precipitates containing V, Mo and Cr exist even near grain boundaries and no PFZ is observed. The analysis of the precipitates using an EELS indicated that the precipitates containing V, Mo and Cr contained N.^{5,6)} This is not the case for the fine precipitates containing Cr and Mo in the ordinary steel, where the concentration of N is below the detection limit of EELS. However, neither Nb, Ti, B nor rare earth metals (REM) were recognized in the EELS or EDX spectrum from the fine precipitates in the V-modified steel. These results conclude that the addition of V significantly changes the composition, shape and distribution of fine precipitates.

3.3. Measurement of Diffusion Coefficient and Evolution of Hydrogen

To estimate the concentration of hydrogen at the interface between the base metal and the welded overlay during the cooling from the condition of elevated temperature and high pressure hydrogen, the diffusion coefficient and evolution of hydrogen in the base metal were measured. The hydrogen permeation transient curve at 25°C obtained by the cathode

charging permeation method was in good agreement with a theoretically calculated one. The apparent diffusion coefficients of hydrogen were calculated as 1.0×10^{-10} and $3.4 \times 10^{-10} \text{ m}^2/\text{s}$ for the V-modified and the ordinary steels, respectively.

The fact that the V-modified steel has a lower hydrogen diffusion coefficient than the ordinary steel is attributed to the fine precipitates containing V, Mo and Cr which have larger trapping effect of hydrogen. To clarify the large trapping effect, the evolution of hydrogen during heating was measured.

Fig. 8 shows the hydrogen evolution curves for the V-modified and the ordinary steels during heating in vacuum at the rate of 0.20°C/s after cathodic charging of hydrogen. Hydrogen trapped at various lattice defects is released during heating in order of lower interaction energy between hydrogen and lattice defects. The total amount of hydrogen dissolved in the specimen is given as the area under the evolution curve. The total evolution of hydrogen for the V-modified steel is $5.96 \times 10^{-8} \text{ Nm}^3/\text{g} \cdot \text{Fe}$ which is higher than the value of $1.01 \times 10^{-8} \text{ Nm}^3/\text{g} \cdot \text{Fe}$ for the ordinary steel. It is concluded, therefore, that the solubility of hydrogen for the V-modified steel at 25°C is 6 times as large as that for the ordinary steel. Since the amount of hydrogen dissolved in the lattice is low in the ferritic steel, the large difference in hydrogen trapped at the fine precipitates. Furthermore, the temperature at which the maximum evolution of hydrogen appears, T_c , is 436 K (163°C) for the ordinary steel and 488 K (215°C) for the V-modified steel. This indicates that the interaction energy between hydrogen and trapping site is higher for the V-modified steel than for the

Fig. 8. Evolution rate of hydrogen at the heating rate of 0.2°C/s in V-modified and ordinary 2 1/4Cr-1Mo steels.

Fig. 9. Evolution rate of hydrogen at the heating rate of 0.05°C/s in V-modified and ordinary 2 1/4Cr-1Mo steels.

ordinary steel.

Fig. 9 shows the hydrogen evolution curves obtained at the heating rate of 0.05°C/s . Other conditions are the same as those for Fig. 8. The fact that the solubility of hydrogen in the V-modified steel is 6 times as large as that in the ordinary steel is the same as that shown in Fig. 8. However, the values of T_c for the V-modified and the ordinary steels are 407 K (134°C) and 346 K (73°C), respectively. They are lower than those shown in Fig. 8. In all the cases nearly all hydrogen had already evolved before the steel was heated to 623 K (350°C). In the operating conditions of pressure vessels at 450°C , therefore, it is expected that hydrogen is not trapped at lattice defects. Thus, the solubility of hydrogen at the operating conditions is expected to be almost same for both steels.

The experimental data indicates that the fine precipitates containing V, Mo and Cr strongly trap hydrogen at room temperature in the V-modified steel and thus the apparent diffusion coefficient of hydrogen is lowered.

4. Discussion

Disbonding is caused by anomalous concentration of hydrogen at the interface between the base metal and the welded overlay during cooling from elevated temperature and higher pressure hydrogen and holding at room temperature. The facts that disbonding depends on time, that is, the sensitivity to disbonding increases with cooling rate¹³⁾ and AE (Acoustic Emission) is detected discontinuously for a while even after cooling to room temperature,¹³⁾ suggest that disbonding is affected by diffusion of hydrogen.

Another factor affecting disbonding is the C-concentrated layer formed at the interface during PWHT. Its effect can be removed in some case by lowering

C content in the base metal.¹⁴⁾ However, the tensile strength decreases with lower C content. Furthermore, it is unfavorable to decrease the cooling rate during the shutdown from the economical point of view. Consequently, disbonding must be avoided through the improvement of microstructure of the base metal as well as the welded overlay. Since no difference in C concentration at the interface was observed between the two types of steel and disbonding occurred not at the interface but in the welded overlay in the present experiment, the difference in the resistance to disbonding is not due to the difference in C-concentrated layer. A significant improvement of the resistance to disbonding obtained by the addition of V to the base metal seems to come from the fact that the V addition affects the concentration of hydrogen at the interface.

As already described in Sec. 3.3, the hydrogen diffusion coefficient at room temperature is low and the evolution of hydrogen in the V-modified steel is 6 times as large as that in the ordinary steel. Since hydrogen was electrochemically charged at 25°C in the present experiment, the activity of hydrogen is different from that at the atmosphere of elevated temperature and high pressure hydrogen where actual disbonding occurs. It is not appropriate, therefore, to regard the evolution of hydrogen as the solubility of hydrogen. Since the electrochemically charged V-modified steel dissolves hydrogen in the quantity 6 times as large as the ordinary steel, however, it is reasonably assumed that the former also contains hydrogen in the quantity 6 times as large as the latter after cooling from elevated temperature and high pressure hydrogen to room temperature.

In the ferritic steel the amount of hydrogen being soluble in lattice is extremely low and a majority of hydrogen is trapped by lattice defects at low temperatures. Then, the trapping effects of lattice defect in

the V-modified and the ordinary steels are compared. As observed in Figs. 8 and 9, T_c is higher for the V-modified steel than for the ordinary steel. This fact indicates that the trapping sites in the former steel capture hydrogen more strongly.

The comparison of T_c between heating rates of 0.2 and 0.05°C/s indicates that T_c becomes lower with lower heating rate in both steels. Choo¹⁵⁾ reported that E_a (trap activation energy) can be calculated from the T_c values measured at different heating rates using Eq. (1).

$$A \exp(-E_a/RT_c) = E_a \phi / (RT_c^2) \dots\dots\dots(1)$$

where, ϕ : heating rate

A : constant

R : gas constant.

The calculated values of E_a for the V-modified and the ordinary steels using Eq. (1) are listed in Table 3. The value of E_a for the V-modified steel is 21 kJ/mol which is higher than that of 13 kJ/mol for the ordinary steel. This means that the trapping sites in the former steel more strongly capture hydrogen than in the latter steel.

Then, the state of trapping site should be considered. In the ferritic steel, major trapping sites are dislocations, grain boundaries and interfaces between precipitates and matrix. Choo has estimated the values of E_a for dislocation and grain boundary as 25.6¹⁶⁾ and 17.2 kJ/mol,¹⁵⁾ respectively. The comparison of the present experimental values of E_a shown in Table 3 with the above-mentioned values indicates that dislocations cannot be regarded as main trapping sites. Furthermore, the total length of prior austenite grain boundaries is the same for the V-modified and the ordinary steels in spite of the fact that the hydrogen dissolved in the former steel is 6 times as large in quantity as in the latter steel. It is concluded, therefore, that the major trapping site is the interface between precipitate and matrix.

The analysis of precipitates in the V-modified and the ordinary steels as described in Sec. 3.2 indicates that no difference exists in coarse carbides having main elements of Cr and Fe in both steels and the coarse carbides trap less hydrogen than the fine precipitates.¹⁷⁾ Consequently, the main trapping sites seem to be the fine precipitates containing V, Cr and Mo for the V-modified steel and those containing Cr and Mo for the ordinary steel. It is considered that the former precipitates have stronger interaction with hydrogen since they trap more hydrogen.

The above discussion makes it clear that the V-modified steel has a lower diffusion coefficient of hy-

drogen and can dissolve more hydrogen at room temperature. Then, the difference in hydrogen concentration at the interface between the base metal and the welded overlay is attributed to that of diffusion coefficient and solubility of hydrogen in the base metal. Therefore, the distribution of hydrogen in the disbonding specimen after cooling from elevated temperature and high pressure hydrogen was estimated by a one-dimensional difference calculus.

Figs. 10 and 11 show the diffusion coefficient and the solubility of hydrogen at 0.098 MPa (1 atm) for ferritic and austenitic steels, respectively. The relation between the diffusion coefficient of hydrogen, D , and the inverse of absolute temperature, $1/T$, was determined in the following way: As for D for the austenitic steel and ferritic steel at temperatures higher than 250°C, the data obtained by Johnson and Hill¹⁸⁾ were used. As for D for the ferritic steel at lower temperatures, the value at 250°C given by Johnson and Hill and that measured at 25°C were used to obtain a linear relation between D and $1/T$. The diffusion coefficients of hydrogen at 25°C were 3.4×10^{-10} m²/s (D1) and 1.0×10^{-10} m²/s (D2) for the ordinary and the V-modified steels, respectively.

As for the solubilities of hydrogen in austenitic and ordinary steels (C1), the data obtained by Phragmen¹⁹⁾ were used. According to Lange and Hofman²⁰⁾ a term proportional to the inverse of absolute temperature must be added to the solubility of hydrogen without voids for the steel containing voids. In the case of the hydrogen solubility for the V-modified steel (C2), the proportional coefficient of the additional term was determined as 0.6. Thus, the solubility of hydrogen for the V-modified steel at 25°C is 6 times as large as that for the ordinary steel. The hydrogen solubility at P atm was assumed to be $\sqrt{P}$ times as large as that at 1 atm according to Sievert law.

With these assumptions, hydrogen distribution in the disbonding specimen cooled from the current operating conditions for the heavy oil desulphurizing plants was calculated. The specimen was exposed to the hydrogen atmosphere of 450°C and 14.7 MPa (150 atm), cooled to the conditions of 25°C and 0.098 MPa (1 atm) hydrogen at the rate of 0.111°C/s (400°C/h), and then held at 25°C. Fig. 12 shows the hydrogen distribution in the disbonding specimens of the ordinary steel and the V-modified steel after being cooled and held at 25°C for 7.2 ks (2 h). The values of C1 and D1 were used as the solubility and diffusion coefficient of hydrogen for the ordinary steel, while those of C2 and D2 were for the V-modified steel. The hydrogen concentration peak in the austenitic phase is 0.0235 mass% for the ordinary steel, while it is 0.0140 mass% for the V-modified steel. This fact indicates that the difference in solubility and diffusion coefficient of hydrogen in the ferritic phase (2 1/4Cr-1Mo steel base metal) significantly affects the hydrogen concentration at the interface.

Fig. 13 shows the relation between maximum hydrogen concentration at the interface and time. In the case of the ordinary steel, the maximum hydrogen concentration increases steadily during holding at

Table 3. Hydrogen diffusion and trapping in V-modified and ordinary 2 1/4Cr-1Mo steels.

Steel	Diffusion 25°C (m ² /s)	Evolution			
		Amount (10 ⁻⁸ Nm ³ / g·Fe)	T_c (K)		E_a (kJ/ mol)
			0.05°C/s	0.2°C/s	
V-modified	1.0×10^{-10}	5.96	407	488	21
Ordinary	3.4×10^{-10}	1.01	346	436	13

Fig. 10. Relation between diffusion coefficient of hydrogen and temperature used to evaluate hydrogen distribution.

Fig. 11. Relation between solubility of hydrogen at 0.098 MPa (1 atm) H_2 and temperature used to evaluate hydrogen distribution.

25°C as well as during cooling. In the case of the V-modified steel, on the other hand, it increases with time during cooling but it hardly increases during holding at 25°C. To estimate the effects of solubility and diffusion coefficient of hydrogen separately, the change in maximum hydrogen concentration with time is also shown in the figure for the case where diffusion coefficient of hydrogen alone is low (the combination of C1 and D2) and for that where hydrogen solubility alone is high (the combination of

Fig. 12. Hydrogen distribution in the disbonding specimens after being cooled from 450°C and 14.7 MPa H_2 and held at 25°C.

Fig. 13. Change in maximum hydrogen concentration with time after the start of cooling.

C2 and D1). The values of maximum hydrogen concentration for both cases are almost same as the average value for the cases of the V-modified and the ordinary steels.

These measurement of solubility and diffusion coefficient of hydrogen and the calculation of hydrogen distribution described above leads to the conclusion that the high resistance of the V-modified steel to disbonding is attributed to the low hydrogen concentration at the interface between the base metal and the welded overlay after cooling from elevated temperature and high pressure hydrogen. The low diffusion coefficient and high solubility of hydrogen at low temperature *via* the effect of strong interaction of hydrogen and fine precipitates containing V, Cr and Mo play important roles in the reduction of hydrogen concentration.

5. Conclusion

The mechanism for improving the resistance to disbonding at the interface between the base metal and the austenitic stainless steel overlay by an addition of V to 2 1/4Cr-1Mo steel was studied. The main results obtained follow:

(1) The diffusion coefficient of hydrogen is low and the solubility of hydrogen at room temperature is high in the V-modified steel compared with the ordinary steel.

(2) The composition, morphology and distribution of fine precipitates are different between the V-modified and the ordinary steels. The difference in solubility and diffusion coefficient of hydrogen between both steels can be explained by the difference in trapping effect of fine precipitates.

(3) A good resistance of the V-modified steel to disbonding is attributed to low diffusion coefficient of hydrogen and high solubility of hydrogen at room temperature and consequently low concentration of hydrogen at the interface between the base metal and the welded overlay.

REFERENCES

- 1) American Petroleum Institute: Steels for Hydrogen Service at Elevated Temperatures and Pressures in Petroleum Refineries and Petrochemical Plants, API Publ. 941 (1983).
- 2) T. Sakai, K. Asami, W. Kondoh and F. Hayashi: *Tetsu-to-Hagané*, **73** (1987), 372.
- 3) T. Ishiguro, K. Ohnishi and J. Watanabe: *Tetsu-to-Hagané*, **72** (1986), 70.
- 4) T. Imanaka, J. Shimomura, S. Nakano and K. Yasuda: *Kawasaki Steel Tech. Rep.*, (1985), No. 13, 109.
- 5) J. Shimomura, M. Miki, Y. Nakano and S. Ueda: Proc. 5th Japan-Czechoslovakia Joint Symp., ISIJ, Tokyo, (1987), 183.
- 6) J. Shimomura, E. Sugie, Y. Nakano and S. Ueda: *Atsuryoku Gijitsu*, **27** (1989), 22.
- 7) K. Ohnishi: Proc. 5th Tsukuba Symp. of Kogyo Gijutsuin, Tsukuba, (1984), 27.
- 8) T. Ishiguro, K. Ohnishi and J. Watanabe: *Tetsu-to-Hagané*, **71** (1985), 986.
- 9) T. Ishiguro: *Tetsu-to-Hagané*, **73** (1987), 34.
- 10) H. Okada, K. Naito, J. Watanabe, K. Ohnishi and R. Chiba: Proc. 1st Int. Conf. on Hydrogen Problems in Steels, Washington D.C., Am. Soc. Met., (1982), 331.
- 11) T. Ishiguro and K. Ohnishi: *Tetsu-to-Hagané*, **71** (1985), S1304.
- 12) J. Shimomura, Y. Nakano, S. Nakano, J. Tateishi and S. Ueda: Proc. 6th Int. Conf. on Pressure Vessel Technology, Beijing, Pergamon Press, (1988), 1013.
- 13) K. Naito, H. Okada, J. Watanabe and K. Ohnishi: *Atsuryoku Gijitsu*, **18** (1980), 47.
- 14) T. Saitou, Y. Otoguro, H. Itou, K. Hashimoto and Y. Nakada: *Tetsu-to-Hagané*, **70** (1984), S1483.
- 15) W. Y. Choo and J. Y. Lee: *Metall. Trans. A*, **13A** (1982), 139.
- 16) W. Y. Choo and J. Y. Lee: *Metall. Trans. A*, **14A** (1983), 1299.
- 17) J. Shimomura and T. Imanaka: Collected Abstracts of the 1983 Autumn Meeting of Japan Inst. of Met., JIM, Sendai, (1983), 391.
- 18) E. W. Johnson and M. L. Hill: *Trans. Am. Inst. Min., Metall. Pet. Eng.*, **218** (1960), 1104.
- 19) G. Phragmen: *Jernkont. Ann.*, **128** (1944), 537.
- 20) G. Lange and W. Hofman: *Arch. Eisenhüttenwes.*, **37** (1966), 391.