INFLUENCE OF HYDROGEN ON FRACTURE TOUGHNESS OF HIGH STRENGTH STEELS

Hideo Kobayashi*, Kazumi Hirano*, Hiroshi Kayakabe* and Hajime Nakazawa*

*Department of Physical Engineering, Faculty of Engineering, Tokyo Institute of Technology, Tokyo, Japan

ABSTRACT

Fracture toughness tests were carried out on hydrogen-charged high strength steels, AISI 4340 and 10B35, and the influence of hydrogen on fracture toughness was examined in detail on the basis of linear elastic fracture mechanics and fractography. The experimental results show that sub-critical flaw growth assisted by hydrogen occurs during fracture toughness testing. Accordingly, plane-strain fracture toughness, K_{Ic}, of hydrogen-charged material evaluated according to ASTM procedure E399 becomes apparently smaller than that of hydrogen-uncharged material due to this sub-critical flaw growth characteristics. Moreover, it makes clear that hydrogen has no influence on critical flaw growth characteristics.

KEYWORDS

Hydrogen embrittlement; hydrogen assisted cracking, fracture toughness.

INTRODUCTION

Considerable works on the problem of hydrogen embrittlement of high strength steels have been conducted to date, where fracture toughness is believed to be influenced by hydrogen (Carter, 1971; Stavros, 1970; Kawabe, 1974; Fukagawa, 1974; Ahmad, 1975; Thompson, 1977). However, exact and detailed studies concerning the influence of hydrogen on fracture toughness have not yet been made. A question arises whether hydrogen causes the decreasing of fracture toughness or not.

In this work, plane-strain fracture toughness tests were carried out on hydrogen-charged high strength steels, AISI 4340 and 10B35, according to ASTM procedure E399, and the influence of hydrogen on fracture toughness was examined in detail on the basis of linear elastic fracture mechanics and fractography.

MATERIALS

The materials used in this work were AISI 4340 and 10B35 steels. The chemical compositions are presented in Table 1. The heat treatment of specimens consisted of austenitizing for 20 min at 1123 K, quenching in oil, tempering for 2 hr at the appropriate temperature (373 ∿ 873 K), and quenching in water. The mechanical properties after heat treatments are presented in Table 2. The yield strength decreased with increasing tempering temperature, from 1569 MPa (tempering at 373 K) to 1049 MPa (tempering at 873 K) in 4340 steel, and from 1588 MPa (tempering at 473 K) to 706 MPa (tempering at 873 K) in 10B35 steel.

SPECIMENS

The specimen geometries were compact type (CT) with dimensions of W = 51 mm and B = 25.4 mm, 3-point bend type (3B) with dimensions of W = 10 mm and B = 5 mm, and self-stressed WOL type (WOL) with dimensions of W = 51 mm and B = 25.4 mm, as shown in Fig. 1. All the specimens were pre-cracked by fatigue at a sufficiently low load (maximum stress intensity factor $K_I \leq 18.6\ MPa\sqrt{m}$).

TABLE 1 Chemical Compositions (wt. %)

	C	Si	Mn	P	S	Ni	Cr	Mo	Cu	B
4340	0.36	0.30	0.77	0.017	0.020	1.85	0.73	0.18	0.14	
10B35	0.38	0.24	0.99	0.019	0.026	0.02	0.09		0.01	0.0012

TABLE 2 Mechanical Properties

Material	Tempering temperature (K)	0.2 % proof stress (MPa)	Tensile strength (MPa)	Elongation (%)	Reduction of area (%)
4340	373	1569	2118	11.0	
	473	1579	1932	10.2	40.5
	573	1412	1667	10.2	49.9
	673	1373	1510	9.0	
	773	1177	1236	12.0	
	873	1049	1118	18.2	54.3
10B35	473	1589	1912	12.1	46.1
	673	1147	1236	12.6	59.7
	873	706	804	21.5	66.0

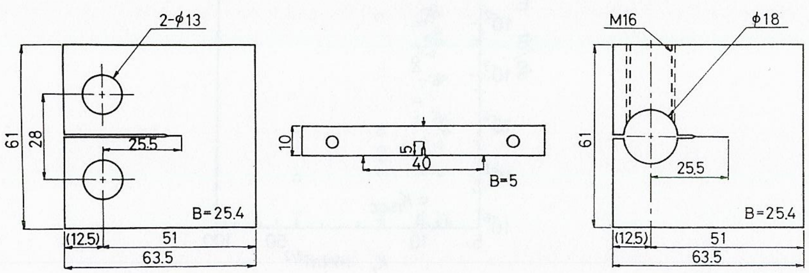


Fig. 1. Geometries and dimensions of specimens.

EXPERIMENTAL PROCEDURE

Hydrogen was charged into the specimens by cathodic charging method without loading in tap water. This cathodic charging was carried out at a constant current density of 16 A/m^2 for 700 hr. Hydrogen concentration was 3 ∿ 4 ppm (weight) regardless of tempering temperature in 4340 steel. After pre-charging, the WOL specimens were loaded by bolt and constant deflection delayed fracture tests were carried out. Hydrogen was charged continuously into the specimens during the tests. As a result of sub-critical flaw growth assisted by hydrogen, the WOL specimens were pre-cracked further at a sufficiently low stress intensity factor ($\approx K_{Iscc}$).

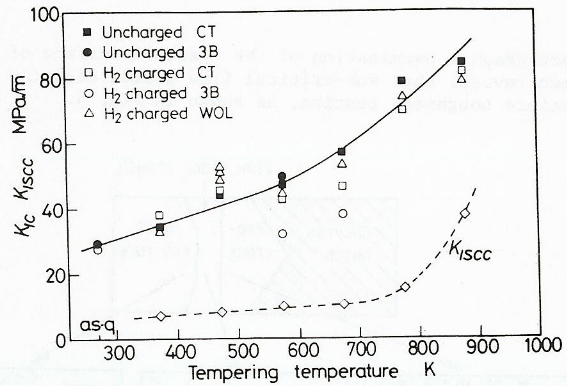


Fig. 2. Comparison between the fracture toughness, K_{Ic}, values of hydrogen-charged and -uncharged 4340 steel, as a function of tempering temperature.

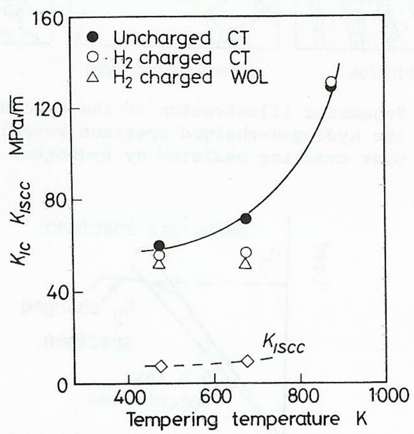


Fig. 3. Comparison between the fracture toughness, K_{Ic}, values of hydrogen-charged and -uncharged 10B35 steel, as a function of tempering temperature.

The plane-strain fracture toughness tests were carried out on hydrogen-charged and -uncharged materials in a laboratory atmosphere, according to ASTM procedure E399.

RESULTS AND DISCUSSIONS

Fig. 2 represents the results of the plane-strain fracture toughness, K_{Ic}, values obtained for the hydrogen-charged and -uncharged 4340 steel, as a function of tempering temperature. As shown by Fig. 2, the K_{Ic} value of the hydrogen-charged material becomes apparently smaller than that of the hydrogen-uncharged material, particularly in the case that the tempering temperature are 573 and 673 K. The similar results for the hydrogen-charged and -uncharged 10B35 steel are shown in Fig. 3.

However, a fractographic examination of the fracture surface of the hydrogen-charged specimen reveals that sub-critical flaw growth assisted by hydrogen occurs during the fracture toughness testing, as shown in Fig. 4.

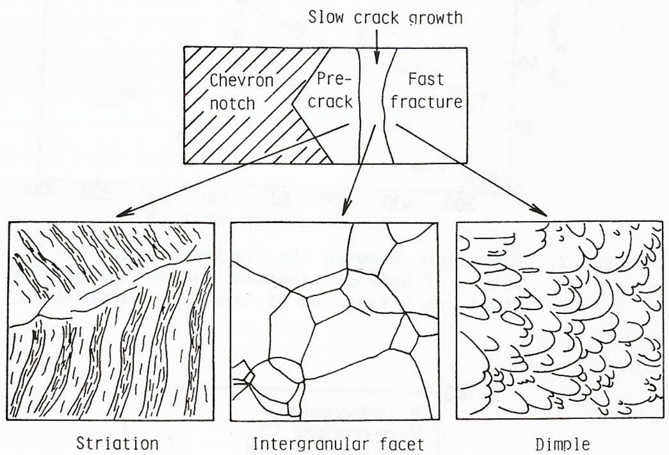


Fig. 4. Schematic illustration of the fracture surface of the hydrogen-charged specimen revealing intergranular cracking assisted by hydrogen.

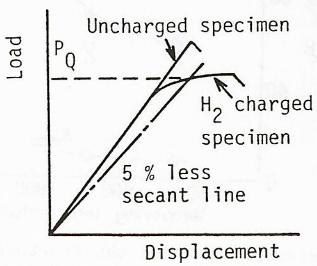


Fig. 5. Typical load-displacement recoads of the hydrogen-charged and -uncharged materials.

Typical load-displacement records of the hydrogen-charged and -uncharged materials are shown in Fig. 5, As shown by Fig. 5, the load-displacement plot of the hydrogen-uncharged material is linear to maximum load. On the other hand, the load-displacement plot of the hydrogen-charged material becomes nonlinear at a load substantially less than the maximum load. Thus, the K_Q value, determined according to the method of 5 % less secant line in ASTM procedure E399, of the hydrogen-charged material becomes apparently smaller than that of the hydrogen-uncharged material. Furthermore, the K_Q value of the hydrogen-charged material is judged to be the valid K_{Ic} value, because the specimen size and maximum load requirements are satisfied. Then, it can be expected that the nonlinearity is partly due to sub-critical flaw growth during the testing.

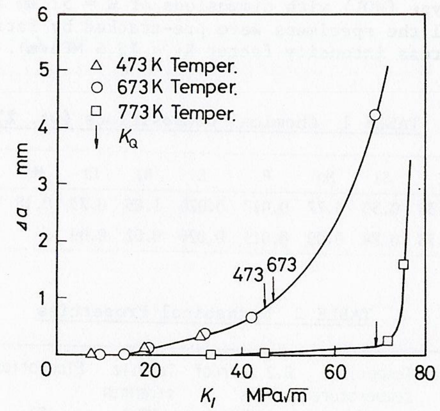


Fig. 6. Relations between the sub-critical crack extension values, Δa, and the K_I values (CT, 4340 steel).

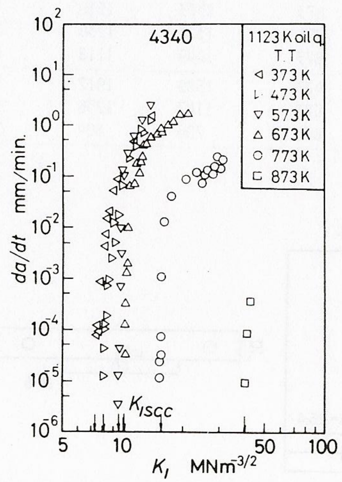


Fig. 7. Crack growth rate obtained in the constant deflection delayed fracture test (WOL, 4340 steel).

Several hydrogen-charged CT specimens were loaded to different displacement values, at which a fast fracture did not initiate, by the similar method of the fracture toughness test. Then, each specimen was unloaded and the crack front was marked by fatigue. The result shows that sub-critical flaw growth occurs at a sufficiently low load and the value of stress intensity factor, K_I, corresponding to the load is smaller than the K_{Ic} value obtained here. Fig. 6 represents relations between the sub-critical crack extension values, Δa, measured on the fracture surface of the unloaded specimens and the K_I values in 4340 steel. For the sake of comparison, the results of constant deflection delayed fracture tests using the WOL specimens for the material used here are shown in Fig. 7, where the hydrogen assisted crack growth rates, da/dt, are plotted as a function of K_I (Hirano, 1979). The sub-critical crack growth rate, da/dt, obtained here in the fracture toughness test is in excellent agreement with that obtained in the constant deflection delayed fracture test.

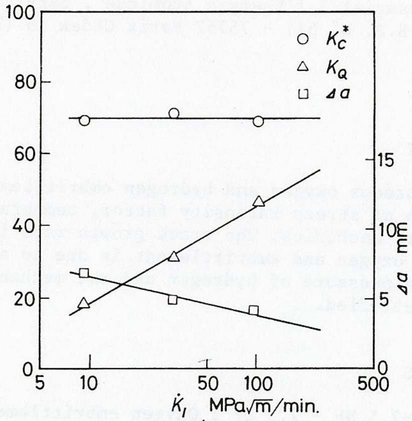


Fig. 8. The K_Q values, the $K_c{}^*$ values and the Δa_{max} values, as a function of the increasing rate of stress intensity factor, $\dot{K}_I$ (CT, 4340 steel).

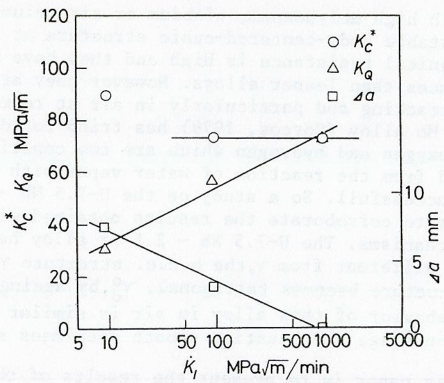


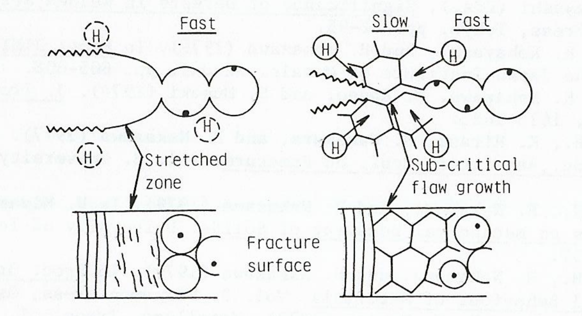
Fig. 9. The K_Q values, the $K_c{}^*$ values and the Δa_{max} values, as a function of the increasing rate of stress intensity factor, $\dot{K}_I$ (CT, 10B35 steel).

Besides, the plane-strain fracture toughness tests were carried out on the hydrogen-charged CT specimens at a varying loading rate. Fig. 8 represents the K_Q values, the $K_c{}^*$ values corresponding to the point of slow-to-fast fracture transition, and the maximum values of sub-critical crack extension, Δa_{max}, measured on the fracture surface of the broken specimen, as a function of the increasing rate of stress intensity factor, $\dot{K}_I = dK_I/dt$, in 4340 steel, where the K_Q value is judged to be the valid K_{Ic} value according to ASTM procedure E399 in the case that $\dot{K}_I$ = 93 MPa√m/min. As shown by Fig. 8, the K_Q value decreases and the Δa_{max} value increases as the $\dot{K}_I$ value decreases. On the other hand, the change of $\dot{K}_I$ has little influence on the $K_c{}^*$ value. The results on 10B35 steel are similar to those in the case of 4340 steel, as shown in Fig. 9.

Fig. 10 represents the results of fractographic examination. In the case of the hydrogen-uncharged specimen and of the hydrogen-charged specimen tested at high $\dot{K}_I$, there exists a stretched zone between fatigue pre-crack and fast fracture regions. This means that crack tip plastic blunting occurs and then fast fracture initiates (Kobayashi, 1977 and 1979). In this case, the K_{Ic} value can be evaluated according to ASTM procedure E399. On the other hand, in the case of the hydrogen-charged specimen tested at low $\dot{K}_I$, sub-critical flaw growth takes place instead of crack tip plastic blunting. In this case, the K_{Ic} value can not be evaluated according to ASTM procedure E399. Difference between these two cases due to $\dot{K}_I$ should be concerned with the diffusivity of hydrogen.

CONCLUSION

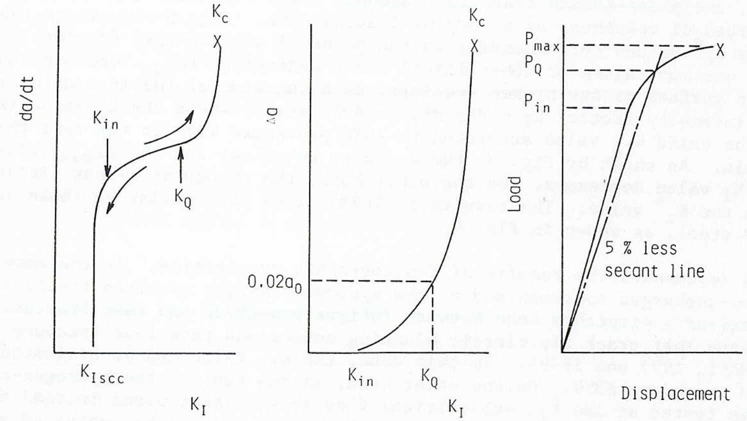
It is concluded that the K_{Ic} value of the hydrogen-charged material evaluated according to ASTM procedure E399 becomes apparently smaller than that of the hydrogen-uncharged material due to the occurrence of sub-critical flaw growth assisted by hydrogen during the testing. The K_Q value of the hydrogen-charged material is decided mainly by sub-critical flaw growth characteristics, as shown in Fig. 11. Moreover, it makes clear that hydrogen and $\dot{K}_I$ have no influence on critical flaw growth characteristics within the limits of this experiment. That is to say, hydrogen has no influence on fracture toughness of high strength steels.



(a) Hydrogen-uncharged specimen
Hydrogen-charged specimen (high $\dot{K}_I$)

(b) Hydrogen-charged specimen (low $\dot{K}_I$)

Fig. 10. Schematic illustrations showing the influence of hydrogen and $\dot{K}_I$ on fractographic features and fracture processes.



(a) Sub-critical flaw growth characteristics

(b) Sub-critical flaw growth during the K_{Ic} testing

(c) Nonlinearity of load-displacement plot due to sub-critical flaw growth

Fig. 11. Sub-critical flaw growth characteristics and non-linearity of load-displacement plot during the K_{Ic} testing.

REFERENCES

Ahmad, S. A., D. A. Ryder, and T. J. Davies (1975). Engng. Fracture Mech., 7, 357-365.

Carter, C. S. (1971). Metallurg. Trans., 2, 1621-1626.

Fukagawa, M., M. Ohyama, H. Okabayashi, and M. Higuchi (1974). In T. Kanazawa and A. S. Kobayashi (Eds.), Significance of Defects in Welded Structures. University of Tokyo Press, Tokyo. pp. 82-95.

Hirano, K., H. Kobayashi, and H. Nakazawa (1979). In Proc. JIMIS-2, Hydrogen in Metals. The Japan Institute of Metals, Sendai. pp. 605-608.

Kawabe, Y., K. Nakazawa, M. Kanao, and S. Muneki (1974). J. Iron & Steel Inst. Japan, 60, 1613-1623.

Kobayashi, H., K. Hirano, H. Nakamura, and H. Nakazawa (1977). In D. M. R. Taplin (Ed.), Proc. 4th Inter. Conf. on Fracture, Vol. 3. University of Waterloo. pp. 583-592.

Kobayashi, H., H. Nakamura, and H. Nakazawa (1979). In H. Miyamoto (Ed.), Recent researches on Mechanical Behavior of Solids. University of Tokyo Press, Tokyo. pp. 341-357.

Kobayashi, H., H. Nakamura, and H. Nakazawa (1979). In Proc. 3rd Inter. Conf. on Mechanical Behaviour of Materials, Vol. 3. Pergamon Press, Oxford. pp. 529-538.

Stavros, A. T., and H. W. Paxton (1970). Metallurg. Trans., 1, 3049-3055.

Thompson, A. W. (1977). In D. M. R. Taplin (Ed.), Proc. 4th Inter. Conf. on Fracture, Vol. 2. University of Waterloo. pp. 237-242.

ASTM Standards. "Standard method of test for plane-strain fracture toughness of metalic materials", E399-78 (1978).