Crack Propagation of High Strength Steels in Oxygen-doped Hydrogen Gas

By Morihiko Nakamura* and Ei-ichi Furubayashi**

Crack propagation of high strength steels with the tensile strength of 1500 or 2000 MPa was investigated at room temperature in an oxygen-doped gaseous hydrogen atmosphere. The doped oxygen gas suppressed the crack propagation in the range of stress intensity factor $K$ below a critical value, at which a hydrogen induced crack began to propagate in a steady state under a load relaxation test i.e. a rising $K$ test. The steady crack propagation rate was hardly affected by the content of doped oxygen gas. The critical value of $K$ was dependent on both hydrogen and oxygen pressures and not determined only by the oxygen concentration in hydrogen. The effect of oxygen was explained by the assumption that oxygen gas decreased the number of absorption or adsorption sites of hydrogen on the crack tip surface and consequently the permeation of hydrogen flowing into a specimens decreased with increasing oxygen pressure.

(Received December 3, 1986)

Keywords: high strength steels, hydrogen gas, oxygen gas, crack propagation rate, stress intensity factor, permeation of hydrogen

I. Introduction

The susceptibility of steels to hydrogen environment embrittlement (HEE) is an important technical problem. It is well known that austenitic stainless steels\(^\text{(1)-(5)}\) and other alloys\(^\text{(6)(7)}\) are embrittled by hydrogen gas as in the case of high strength steels\(^\text{(8)-(14)}\). The decreases in the reduction in area of the unnotched specimens and in both the reduction in area and the tensile strength of the notched specimens tested in a hydrogen gas have been used to measure the degree of HEE of low strength steels. Pre-yield fracture as well as the decrease in the reduction in area is often observed for unnotched high strength steels tested in tension in a hydrogen gas\(^\text{(15)}\).

The influence of stress intensity factor $K$ on the crack propagation behavior in a pure hydrogen gas has been studied using precracked specimens of various high strength steels, e.g. maraging steels\(^\text{(16)(17)}\), AISI 4000-series steels\(^\text{(18)-(22)}\), etc.\(^\text{(20)(23)-(25)}\). The effect of alloying elements and tensile strength on the threshold stress intensity factor $K_{th}$ were reported for AISI 4340 steel\(^\text{(19)}\) and high strength steels\(^\text{(24)}\), respectively. The effect of Mn and C on the crack propagation rate was also studied for high strength steels in a previous paper\(^\text{(25)}\).

On the other hand, it has been reported that the crack propagation behavior of high strength steels is affected by addition of various impurity gases to hydrogen gas\(^\text{(26)-(30)}\). The crack extension in precracked specimens was accelerated in H\(_2\)S-doped hydrogen gas\(^\text{(28)(29)}\).

Hancock and Johnson reported that a crack in a stressed precracked specimen began to grow by exposing the specimens into the flow of pure hydrogen, and was arrested by doping oxygen into the hydrogen gas\(^\text{(26)}\). The threshold stress intensity factor $K_{th}$ in oxygen-doped hydrogen is reported to be proportional to the oxygen concentration\(^\text{(30)}\). Their study on this suppressing effect of oxygen is, however, considered to be limited only in a narrow range of hydrogen pressure. The present work is, therefore, conducted to clarify the effects of wider range pressure of oxygen-doped hy-
drogen on the crack propagation behavior of the high strength steels used in the previous work\(^{(25)}\).

**II. Experimental**

Six steels were vacuum melted in 10 kg heats using electrolytic iron, nickel, etc. Table 1 shows their chemical compositions. They were hot forged, hot rolled to 18 mm thick plates, then annealed at 1125 K for 1.5 ks in a salt bath, followed by oil quenching and tempering at temperatures between 473 and 673 K for 3.6 ks. Modified CT specimens (Fig. 1) and round bar tensile specimens with a diameter of 4 mm and a gage length of 22 mm were machined from the plates so that their tensile axis was parallel to the rolling direction. The CT specimens were precracked by fatigue, where the maximum applied stress intensity factor \( K \) was held less than 10 MPa m\(^{1/2}\).

Tensile tests were carried out using an Instron-type testing machine at a constant crosshead speed of 0.055 mm/s. Delayed fracture test was conducted using CT specimens in a hydrogen chamber which was fabricated from an austenitic stainless steel and fixed to the Instron-type machine. At first, the hydrogen chamber was filled with oxygen gas of a given pressure, then hydrogen gas was added to the oxygen gas through Pd film. Thus, the oxygen-doped hydrogen gas with a given pressure of oxygen gas was obtained in the hydrogen chamber. The pressure of hydrogen ranged from 98 to 748 kPa. The applied load was measured by the strain gages adhered to the pull-rod in the chamber, and the crack opening displacement (COD) at the notched surface was measured using a clip gage.

The compliance of the CT specimens with various crack lengths was compared with the predicted values which were calculated from the equation reported by Novak and Rolfe\(^{(31)}\). It was found that the calculated values coincided with the measured ones of the modified CT specimen. Thus, the applied \( K \) value and the crack length of the modified CT specimen were calculated using the measured values of applied load and COD. In this experiment both the applied load and the COD were measured as a function of testing time, so that the crack length, the crack propagation rate and the associated \( K \) value were calculated from the load and COD-time records. The applied \( K \) increased with the crack length because the load relaxation test was used\(^{(25)}\).

**III. Experimental Results**

The steels used had tempered martensitic microstructures. X-ray diffraction analyses showed that retained austenite was hardly observed except for 0.53C–0.83Mn and 0.43C–2.13Mn steels, where the content of retained austenite phase was about 3–4%. The mechanical properties of the steels and the associated tempering temperatures are listed in Table 2. The tempering temperatures were selected to produce the tensile strength of 1500 or 2000 MPa. The yield strength (0.2% offset stress) increased with increasing C content and decreased with increasing Mn content.

Figure 2 shows some examples of relationships between crack propagation rate \((da/dr)\) and applied \( K \) in oxygen-doped hydrogen. A
A propagating crack was sometimes decelerated and arrested in spite of the increase of applied \( K \) in a range of relatively low values of \( K \). In contrast to the case of oxygen-doped hydrogen, a propagating crack was never arrested with increasing \( K \) in pure hydrogen \(^{25}\). This fact suggests that the fresh surface of a tip of the propagating crack in oxygen-doped hydrogen may be covered with oxygen atoms, which suppresses the further growth of the crack in a relatively low \( K \) range. The crack propagation was, however, observed again as the applied \( K \) was further increased. This process was sometimes repeated in the same sample, and in a higher \( K \) range \( \frac{da}{dt} \) increased monotonically with the increase of applied \( K \), followed by steady crack propagation.

Figure 3 shows the effect of oxygen pressure on \( \frac{da}{dt} \) in hydrogen of 98 kPa in pressure for 0.43C–2.14Mn steel with the tensile strength of 2000 MPa. In this figure only the stage of the steady propagation is cited. The critical value of applied \( K \), \( K_{cv} \), above which a crack continued to propagate and reached a steady propagation, increased gradually with increasing oxygen pressure, excepting the case of oxygen pressure of 135 kPa which is similar to that in pure hydrogen as shown in Fig. 3. It can be seen in the same figure that once a crack began to grow, \( \frac{da}{dt} \) increased rapidly with increasing applied \( K \) and became independent of the oxygen pressure in the higher \( K \) range. Figure 4 shows the similar relationship between \( \frac{da}{dt} \) and applied \( K \) for 0.53C–0.84Mn steels with the tensile strength of 1500 MPa at oxygen-doped hydrogen of 294 kPa in pressure. Figure

---

**Table 2** Mechanical properties of steels used and tempering temperatures.

| No. | \( \sigma_{0.2} \) (MPa) | \( \sigma_{u} \) (MPa) | R.A. (%) | \( K_{IC} \) or \( K_{Q} \) (MPa m\(^{1/2}\)) | Tempering temperature (K) |
|-----|-----------------|-----------------|--------|-----------------|-----------------|
| 1   | 1223            | 1445            | 65     | 127*            | 473             |
| 2   | 1304            | 1479            | 64     | 125*            | 573             |
| 4   | 1339            | 1463            | 52     | 115*            | 673             |
| 6   | 1534            | 1959            |        | 79.5            | 473             |
| 7   | 1468            | 1965            |        | 81.4            |                  |
| 8   | 1392            | 1973            | 44     | 73.7            |                  |

\( \sigma_{0.2} \): 0.2% offset stress, \( \sigma_{u} \): tensile strength, R.A.: reduction of area, \( K_{IC} \) and \( K_{Q} \): valid and invalid fracture toughness, respectively.
960 Morihiko Nakamura and Ei-ichi Furubayashi

shows the relationship between $da/dt$ and applied $K$ for 0.42C-1.27Mn steel. The hydrogen pressure was changed from 98 to 784 kPa, while the fraction of oxygen pressure ($=oxygen\ pressure/(hydrogen+oxygen)\ pressure$) was constant for each hydrogen pressure. $K_{CV}$ was dependent on the hydrogen pressure, though the fraction of oxygen pressure was approximately constant. The effect of doped oxygen on $da/dt$ was considerable under the low hydrogen pressures, but little under the high hydrogen pressures. In the stage of steady crack propagation, however, $da/dt$ was nearly independent of the doped oxygen pressure and was affected by the hydrogen pressure.

Figure 6 shows the dependence of $K_{CV}$ on C content for the steels with the tensile strength of 1500 MPa. Figure 7 shows the dependence of $K_{CV}$ on Mn content for the steels with the tensile strength of 2000 MPa. In these cases both the hydrogen pressure and the impurity oxygen pressure were constant. Compared with the experimental results in the previous paper(25), these figures indicate that the effect of impurity oxygen on the crack propagation
Crack Propagation of High Strength Steels in Oxygen-doped Hydrogen Gas

Fig. 6  Dependence of the critical value $K_{cv}$ on C content.

Fig. 7  Dependence of the critical value $K_{cv}$ on Mn content.

Fig. 8  Fracture surfaces of 2.13Mn-1.43C steel tested in oxygen-doped hydrogen gas.
behavior was little for these steels which had a high crack propagation rate in a pure hydrogen gas.

Figure 8 shows the fracture surfaces scanning electron micrographs of 2.13Mn–0.43C steel tested in hydrogen under a pressure of 98 kPa. The front of the fatigue crack surface is indicated in the upper part of each photograph. Dimple regions were observed in front of the fatigue crack tip. These dimple regions were found to become larger with increasing oxygen pressure. As the value of \( K \), at which a crack began to propagate, increase with increasing oxygen pressure in hydrogen, the fracture surface at the fatigue crack tip changed from grain boundary surface to dimple pattern via cleavage facets. Since \( K_{CV} \) was still much lower than the normal fracture toughness values in these steels, these dimple regions may be due to the effect of oxygen gas in the hydrogen induced crack.

IV. Discussion

As shown in Figs. 3–6, a hydrogen crack, which began to grow in the relatively low applied \( K \) range, was often observed to be decelerated and arrested in spite of an increase in applied \( K \), in contrast to the case in pure hydrogen, where the steady crack propagation was observed in the same \( K \) range. This fact indicates that impurity oxygen in a hydrogen gas increased the critical value of applied \( K \) above which the steady crack propagation was observed. The value of \( da/dt \) of the steadily propagating crack was nearly independent of the impurity oxygen content, although it was dependent on the hydrogen pressure.

The models of crack propagation in a hydrogen gas have been proposed by some investigators\(^{(25)(32)(33)}\). A hydrogen crack propagates mainly along grain boundaries in high strength steels, where \( da/dt \) is determined by the amount of hydrogen flowing into the material near the crack tip. Hydrogen is thought to flow into a grain from a fresh surface produced by plastic deformation at the crack tip when precracked specimens are loaded. The amount of hydrogen flowing into a grain, therefore, increases with the applied \( K \), i.e., the crack tip opening displacement (CTOD) \( \delta \), because an area of the crack tip surface at a given \( K \) is represented by the product of \( \delta \) and the thickness. It is also thought that oxygen gas has a tendency to be adsorbed (physisorbed) or absorbed (chemisorbed) on the surface of steels more strongly than hydrogen gas, so that the number of adsorption (physisorption) or absorption (chemisorption) sites for hydrogen on the crack tip surface is decreased by the addition of impurity oxygen gas. The amount of hydrogen flowing into a specimen through the crack tip surface per unit time and per unit thickness at a constant temperature is given by the following equation as explained in the previous paper\(^{(25)}\).

\[
q = M_H \delta S (1 - F_O) F_{H_2} \sqrt{P_{H_2}}, \quad (1)
\]

or

\[
q = M_H \delta S (1 - F_O) F_H \sqrt{P_H}, \quad (2)
\]

where \( F_O \) is the fractional coverage for absorption of oxygen at the crack tip surface, and \( F_{H_2} \) and \( F_H \) are those for adsorption and absorption of hydrogen respectively. \( S \) is the number of adsorption or absorption sites per unit area for hydrogen gas, \( P_{H_2} \) the hydrogen pressure, \( \delta \) CTOD, and \( M_H \) a constant containing the diffusion rate of hydrogen in the membrane which is assumed on the surface of the material. Equation (1) or (2) is used when the adsorption or the absorption is a rate controlling process, respectively. \( F_O, F_{H_2}, F_H \), and \( \delta \) are given by the following equations\(^{(34)}\):

\[
F_O = A_O \sqrt{P_O}/(1 + A_O \sqrt{P_O}), \quad (3)
\]

\[
F_{H_2} = A_{H_2} P_{H_2}/(1 + A_{H_2} P_{H_2}), \quad (4)
\]

\[
F_H = A_H \sqrt{P_H}/(1 + A_H \sqrt{P_H}), \quad (5)
\]

and

\[
\delta = m(1 - \nu^2)/(E \sigma_y) K^2, \quad (6)
\]

(for plane strain condition)

where \( P_O \) is the oxygen pressure, \( A_O \) the parameter for absorption of oxygen, and \( A_{H_2} \) and \( A_H \) are those for adsorption and absorption of hydrogen, respectively. \( E \) is Young's modulus, \( \nu \) Poisson's ratio, \( \sigma_y \) the yield strength and \( m \) a constant \((m < 1)\) which depends on the work hardening rate etc. \( F_{H_2} \)
Crack Propagation of High Strength Steels in Oxygen-doped Hydrogen Gas

or \( F_H \) in eq. (4) or (5) is further approximated by the following equation in a narrow range of hydrogen pressure, e.g. 98 to 748 kPa as reported in the previous paper:\(^{(25)}\)

\[
F_{H} = B P_{H}^{n} \quad (1/2 \leq n \leq 3/2),
\]

where \( B \) is a constant which decreases with increasing \( n \) up to 3/2. On the assumption that \( \frac{da}{dt} \) is proportional to the amount of hydrogen flowing into a specimen from the crack tip surface\(^{(18)(25)(34)}\), the following equation is obtained using eqs. (2), (5), (6) and (7).

\[
\frac{da}{dt} = Q F_H,
\]

\[
= B Q' M_H K^2 \left( 1 - \frac{A_O \sqrt{P_{O_2}}}{1 + A_O \sqrt{P_{O_2}}} \right) P_{H}^n,
\]

where \( Q \) is a constant which depend on the degree of hydrogen embrittlement of grain boundaries etc. and \( Q' = Qm(1 - \nu^2)/(E \sigma_s) S \). If the critical value of applied \( K \), \( K_{CV} \), is assumed to be that of \( K \) at which \( \frac{da}{dt} \) has a critical value, the relationship between \( K_{CV} \), \( P_{O_2} \) and \( P_{H} \) is obtained from eq. (8).

\[
\frac{1}{K_{CV}^2} = \frac{B Q' M_H}{V_0} \left( 1 - \frac{A_O \sqrt{P_{O_2}}}{1 + A_O \sqrt{P_{O_2}}} \right) P_{H}^n,
\]

where \( V_0 \) is a critical value of \( \frac{da}{dt} \). Equation (9) is further reduced to

\[
K_{CV}^2 = C_A \left( 1 + A_O \sqrt{P_{O_2}} \right) P_{H}^n,
\]

where \( C_A = \frac{V_0}{B Q' M_H} \). \( K_{CV} \) is, therefore, predicted to be a linear function of \( P_{O_2}^{1/2} \) for a given steel tested under a constant pressure of hydrogen.

Figure 9 shows the relationship between \( K_{CV}^2 \) and \( P_{O_2}^{1/2} \) for various steels. The experimental data may be explained by eq. (10) as shown in the figure. The predicted line of the relationship between \( K_{CV}^2 \) and \( P_{O_2}^{1/2} \) using the experimental one of 0.43C–2.13Mn steel tested under the hydrogen pressure of 98 kPa agrees well with the experimental value for the same steel tested under the hydrogen pressure of 392 kPa, where the value of \( n \) is assumed to be approximately equal to 3/4, the value obtained from the dependence of \( \frac{da}{dt} \) on \( P_{H} \) in pure hydrogen for 0.43C–2.13Mn steel\(^{(25)}\). Table 3 lists the relationships between \( K_{CV} \), \( P_{O_2} \) and \( P_{H} \) for the steels used, where \( C_A \) and \( A_O \) in eq. (10) were estimated from the experimental lines in Fig. 9. It can be seen that there is no notable difference in the value of \( A_O \), the parameter for absorption of oxygen on the crack tip surface among the steels used. The value of \( C_A \) is larger for 1.27Mn steel than for 2.13Mn steel and it is also larger for 0.53C steel than for 0.33C steel. It is a difficult problem to understand the meaning of \( C_A \), since \( C_A \) depends on various factors i.e. \( V_0 \), \( B \), the degree of grain boundary embrittlement, the diffusion rate of hydrogen in the membrane, etc.

Figure 10 shows the relationship between \( K_{CV}^2 \) and \( 1/P_{H} \) for 0.42C–1.27Mn steel tested in hydrogen containing a constant oxygen concentration, where the straight lines are estimated from the equation in Table 3. The experimental data are in relatively good agreement with the estimated lines. This fact is consistent with the observation that \( K_{CV} \) was not dependent solely on the oxygen concentration in hydrogen.

It may be reasonably deduced that a small

\[
\begin{array}{l}
\text{For } 0.84 \text{Mn–0.33C steel } \\
K_{CV}^2 = 2.65 \times 10^6 (1 + 22.9 \sqrt{P_{O_2}})^{-3/4} \\
\text{For } 0.84 \text{Mn–0.53C steel } \\
K_{CV}^2 = 6.55 \times 10^6 (1 + 22.9 \sqrt{P_{O_2}})^{-3/4} \\
\text{For } 1.27 \text{Mn–0.42C steel } \\
K_{CV}^2 = 1.02 \times 10^6 (1 + 17.5 \sqrt{P_{O_2}})^{-3/4} \\
\text{For } 2.13 \text{Mn–0.43C steel } \\
K_{CV}^2 = 2.18 \times 10^6 (1 + 17.9 \sqrt{P_{O_2}})^{-3/4}
\end{array}
\]

UNIT: MPa m\(^{1/2}\) for \( K_{CV} \) \\
kPa for \( P_{H} \) and \( P_{O_2} \)
amount of oxygen gas hardly affects the crack propagation rate, $da/dt$, when the crack tip run away before oxygen is absorbed on the crack tip surface. This consideration may lead to the concept of a critical value of $da/dt$, $V_0$. Since $V_0$ was $5 \times 8 \times 10^{-7}$ m/s (Figs. 3–5) and $da/dt$ was considerably low, $3 \times 4 \times 10^{-6}$ m/s, for $0.21C-0.84Mn$ steel tested in pure hydrogen, it appears that $K_{CV}$ did not exist and a small amount of oxygen gas could suppress the crack propagation in $0.21C-0.84Mn$ steel.

The fact that dimple regions were observed near the fatigue crack tip for specimens fractured in oxygen-doped hydrogen (Fig. 8) may be explained as follows. Oxygen may not be absorbed uniformly on the crack tip surface. Hydrogen may hardly permeate through some parts of the crack tip surface where oxygen was absorbed densely, and consequently a hydrogen crack can hardly extend there. Meanwhile, hydrogen may permeate through the other parts where oxygen was scarcely absorbed and a hydrogen crack may extend easily. The ligaments between hydrogen cracks will be fractured by dimples when the applied $K$, i.e. applied stress, is increased.

V. Conclusion

Crack propagation of high strength steels with the tensile strength of 1500 or 200 MPa was studied at room temperature in the environment of oxygen-doped hydrogen gas. Modified CT specimens were tested using an Instron-type testing machine under a condition where the applied stress intensity factor $K$ increases with increasing crack length. The results obtained are as follows.

1. Doped oxygen gas suppressed the crack propagation in the range of applied $K$ below a critical value $K_{CV}$ where the hydrogen crack began to grow in a steady state under a rising $K$ test. The rate of the steady crack propagation, however, was hardly affected by the amount of doped oxygen gas.

2. The critical value $K_{CV}$ decreased with decreasing C content from 0.53 to 0.33 wt%, but it could not be measured for $0.21C-0.84Mn$ steel in this experiment because of its much higher value. It decreased monotonically with increasing Mn content up to 2.13 wt%.

3. $K_{CV}$ was dependent on both the hydrogen and the oxygen pressures, and not only on the oxygen concentration in a hydrogen gas.

4. The fracture surface mainly consisted of intergranular facets, but dimple regions were also observed near the fatigue precrack tip.

5. The effect of doped oxygen in hydrogen was explained by the assumption that oxygen gas decreased the number of adsorption or absorption sites of hydrogen on the crack tip surface, and the hydrogen flowing into a specimen decreased with increasing oxygen pressure.

Acknowledgments

The authors would like to thank Dr. S. Horibe for his helpful discussion. The authors also thanks Mr. T. Kato, J. Takahashi, J. Isobe and the members of Engineering Section etc., NRIM for preparation of specimens.

REFERENCES

(1) W. T. Chandler and R. J. Walter: ASTM STP 543, p. 170, ASTM, Pa, 1974.
(2) A. W. Thompson: Met. Prog., (1976), 30.
(3) R. R. Vandervoort: Met. Eng. Quart., 2 (1972), 10.
(4) R. M. Vennett and G. S. Ansell: Trans. ASM, 60 (1967), 242.
(5) A. W. Thompson: Mater. Sci. Eng., 14 (1974), 253.
(6) G. Garmong: Met. Trans., 8A (1977), 535.
(7) R. J. Walter and W. T. Chandler: Hydrogen in Metals, eds. I. M. Bernstein and A. W. Thompson, ASM, OH, (1974), p. 513.
(8) W. Hofmann and W. Rauls: Weld. J., Research
(9) P. Bowker and D. Hardie: Met. Sci., 9 (1975), 432.
(10) D. Hardie and P. Bowker: *Effect of Hydrogen on Behavior of Materials*, eds. A. W. Thompson and I. M. Bernstein, TSM AIME, N.Y., (1976), p. 251.
(11) D. Hardie and T. I. Murray: Metals Tech., May (1978), 145.
(12) J. B. Steinman, H. C. von Ness and G. S. Ansell: Weld. J., Research Suppl., (1965), 221S.
(13) K. Ohnishi, R. Chiba, K. Teshirogi and H. Kaga: J. Japan Inst. Metals, 40 (1976), 651.
(14) R. R. Vandervoort, A. W. Ruotola and E. L. Raymond: Met. Trans., 4 (1973), 1175.
(15) Y. Kawabe, S. Muneki and M. Hagiwara: *Hydrogen in Metals*, Trans. Japan Inst. Metals, 21 suppl. (1980), p. 453.
(16) R. P. Gangloff and R. P. Wei: Met. Trans., 8A (1977), 1043.
(17) S. J. Hudak and R. P. Wei: Met. Trans., 7A (1976), 235.
(18) D. P. William and H. G. Nelson: Met. Trans., 1 (1970), 63.
(19) S. Sandoz: Met. Trans., 3 (1972), 1169.
(20) J. P. Landes and R. P. Wei: Int. J. Fract., 9 (1973), 277.
(21) G. W. Simmons, P. S. Pao and R. P. Wei: Met. Trans., 9A (1978), 1147.
(22) P. W. Smith, M. R. Louthan, Jr. and R. P. McNitt: Scripta Met., 14 (1980), 199.
(23) A. W. Loginow and E. H. Phelps: Trans. ASME, J. Eng. Industry, (1975), 274.
(24) S. Hinotani, F. Terasaki and K. Takahashi: *Tetsu-To-Hagané*, 64 (1978), 899.
(25) M. Nakamura and E. Furubayashi: Met. Trans., 14A (1983), 717.
(26) G. G. Hancock and H. H. Johnson: Trans. Met. Soc. AIME, 236 (1966), 513.
(27) H. W. Liu, Ya-L. Hu and P. J. Ficalora: Eng. Fract. Mech., 5 (1973), 281.
(28) V. Srikrishnam and P. J. Ficalora: Met. Trans., 7A (1976), 1669.
(29) G. E. Kerns and R. W. Staehle: Corrosion, 34 (1978), 306.
(30) H. H. Johnson: *Stress Corrosion Cracking and Hydrogen Embrittlement of Iron Base Alloys*, eds. R. W. Staehle, J. Hochmann, R. D. McCright and J. E. Slater, NACE, Texas, (1977), p. 382.
(31) S. R. Novak and S. T. Rolfe: J. Mater., 4 (1969), 701.
(32) H. G. Nelson and D. P. Williams: *Stress Corrosion Cracking and Hydrogen Embrittlement of Iron Base Alloys*, eds. R. W. Staehle, J. Hochmann, R. D. McCright and J. E. Slater, NACE, Texas, (1977), p. 390.
(33) V. Sawicki and H. H. Johnson: Met. Trans., 2 (1971), 3496.
(34) N. J. Petch: Phil. Mag., 1 (1956), 331.