FORMATION AND CHARACTERISTICS OF QUASI-CLEAVAGE CRACK IN HYDROGEN DAMAGED LOW STRENGTH STEEL

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A study has been made on the critical behaviour of crack growth in low strength pipe line steels subjected to sustained load in hydrogen environment. Formation of quasi-cleavage (QC) crack occurs in the same manner as that of high strength steels though the material has a low yield strength level of about 600-700MPa. However, no contribution of Intergranular (IG) cracks to making fracture surface has been found in the low strength steel. Further SEM investigation of the fracture surface revealed a development of several numbers of the QC crack facets at subsurface defects such as non-metallic inclusions, followed by growth in the radial direction resulting in penny shaped crack morphology. From these it may conclude that the development of the QC crack is essential in the delayed fracture of structural steel rather than the formation of the IG crack irrespective of the strength level. Critical discussion was also made on the fracture susceptibility to hydrogen using a new parameter defined as the ratio of σ_{th} (delayed fracture) to σ_y (yield stress).

INTRODUCTION

Hydrogen induced brittle fracture has long been thought to be a critical phenomenon in high strength steels which becomes increasingly noticeable with an increase in strength level. Thus, it is natural for a designer to use lower strength steel for structural components such as bolts intended for use in an aggressive environment to assure structural integrity of a system. However, there still remains a critical issue that hinders the total understanding of fracture behaviour in high strength steels subjected to hydrogen attack; namely, the crack growth behaviour associated with the penetration of hydrogen into a material and the condensation of hydrogen at the crack tip.

The critical behaviour of crack growth has long been left unsolved owing to the fact that such hydrogen induced cracks develop in the subsurface, thus making it difficult to detect during tests. Previous studies investigate a critical condition for crack growth observed in the delayed fracture of very high strength steels employing unnotched specimen geometry (1)-(2). These results suggest that a quasi-cleavage (QC) crack first initiated at the subsurface consequently triggers the unstable growth of a crack which consists of an intergranular (IG) growth followed by a micro-void coalescence (MVC) mode. This crack growth behaviour is characterised by a period of subcritical growth of the QC crack and by a presence of the critical condition for the onset of growth of an unstable crack consisting of IG crack and dimple crack in later stage.

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Up to now, the hydrogen induced fracture of low strength steels has attracted little attention of researchers except for those interested in structural integrity under aggressive environments (3) or in the deformation characteristics associated with hydrogen induced plasticity (4). Though a number of questions in hydrogen induced fracture phenomena of low strength steels have so far been solved, no particular discussion has been focused on the similarities and/or dissimilarities between the critical behaviour of crack growth in low and high strength steels. Thus, a question arises as to whether the essential mechanism of hydrogen induced crack growth behaviour is different between high and low strength steels. In this paper, a relationship between the strength level of steel and the aspects of crack formation and propagation under hydrogen environments is studied on a X-65 type low-carbon pipe line steel possessing a martensitic structure. Furthermore a discussion has been made on the essential nature of the fracture susceptibility of structural steel to hydrogen degradation.

EXPERIMENTAL PROCEDURE

X-65 low-carbon pipe line steel was machined into smooth specimens with a gauge length and diameter of 10mm and 5mm respectively. Specimens were then heat treated to obtain a martensitic structure with an average prior austenite grain diameter of 10μm possessing a variety of strength levels (series A ~ D) as shown in Table 1. These specimens were then mechanically polished followed by vacuum annealing before mechanical testing. Conventional delayed fracture tests were carried out under a sustained load with concurrent hydrogen charging. Details of the delayed fracture test and the method of hydrogen charging are given in the authors' earlier papers (1)-(2).

TABLE 1- Mechanical Properties of specimens

Code	U.T.S. (MPa)	0.2%Proof Stress (MPa)	Micro-Vickers Hardness (P=25g)
A	881	709	359
B	852	640	322
C	790	693	241
D	678	622	206

RESULTS AND DISCUSSION

Sensitivity of Delayed Fracture in Low Strength Steel

Delayed fracture tests were carried out on all series of specimens under a sustained load. Despite the fact that gauge geometry has no concentration of stress, almost all specimen series fractured in a brittle manner at stress levels below yield strength. Typical results of this test are shown in Fig.1. The threshold stress was estimated from a particular stress level at which the fracture did not occur after 10^4 min (~ a week) loading in a similar manner to that in previous high strength steel tests (1)-(2).

Furthermore, by defining delayed fracture sensitivity as the ratio of threshold stress to yield stress(σ_{th}/σ_y), we enable consideration of fracture susceptibility of structural steel to hydrogen against the full range of material strength level. The magnitude of this parameter for the present material is plotted in Fig. 2, where open marks are reference data obtained from the results of a similar delayed fracture test on high strength steel, such as AISI 4340, ASTM A-490, employing unnotched specimen (1)-(2). A noticeable difference of fracture resistance to hydrogen can be observed between low and high strength steels, where the fracture resistance increases in accordance to the yield strength level. Significance of an appreciable drop of the single plot in lower range of the strength level in the figure will be discussed later as a key issue to understand the fracture process in low strength steels.

Differences in Fracture Characteristics between Low and High Strength Steel

To investigate the reason to such remarkable differences in the fracture resistance to hydrogen between low and high strength steels, a fractographic analysis was carried out on all series of specimens, and features of the fracture surface of these specimens were compared to those of high strength steel. In unnotched high strength steel specimens subjected to a sustained-load delayed fracture test, a crack will normally starts at a subsurface non-metallic inclusion, resulting in the development of a QC crack, followed by IG and MVC (micro-void coalescence) cracks as shown in a schematic illustration in Fig. 3-(a) (1)-(2). Fracture characteristics of low strength steel specimens, in contrast, differ remarkably from those of high strength steel. The SEM fractograph revealed that numerous unusual QC facets, each of which had grown into a large radial pattern, were observed among dimples in the fracture surface. Remarkably, no trace of any IG facets, known to be the most typical fracture pattern in the hydrogen embrittlement of high strength steel, were observed. Fractographs illustrating this phenomenon are given in Fig. 4 with a schematic illustration in Fig. 3-(b). Assuming from the results and considerations obtained from fracture analysis in high strength steel, QC cracks in low strength steel develop in the subsurface and propagate in a stable manner until the onset of fatal crack growth begins with a coalescence of dimples. The comparison of fracture surfaces between low and high strength steel specimens has revealed differences in the size of QC facets, which are extremely large in the low strength steel, and the existence of IG facets in high strength steel which cannot be seen in low strength steel. The formation of an IG crack in high strength steel placed in a hydrogen environment requires a high local stress zone to develop at either the QC crack, fatigue precrack (1), or sharp notch (1) as shown in Fig. 5. Thus, if no sufficient high local stress is built up such as in the case of a blunt notch, the formation of additional QC cracks occurs within the specimen prior to the onset of growth of an IG crack (1), as shown in Fig. 5-(b). This implies that in low strength steels, no sufficient high local stress is built up, due probably to plastic blunting, to trigger an IG crack ahead of the QC crack. For further consideration of this matter, observations of the geometrical features of the tip of a QC crack were carried out in the cross section of the tested specimen.

The development of QC cracks which do not contributed to the fracture surface can also be expected in the subsurface of fractured specimens of low strength steel, similar in manner to those observed in high strength steel specimen (2). Following this assumption, SEM observations of the cross-sectional area were made on a series A specimen which had fractured after 487 min of loading under the applied stress of 550 MPa corresponding to about 75% of yield stress. Expected cracks were found in the subsurface as shown in Fig. 6-(a); these corresponding to similar QC cracks in high strength steel as shown in Fig. 6-(b) (2). This comparison has revealed that the tip of a QC crack in a low strength steel specimen is extremely stretched compared to that of high strength steel. This difference in crack tip geometry between high and low strength steel could be due to a difference in yielding behaviour at the respective crack tips. This should result in a different state of stress at the crack tip and bring about a difference in hydrogen behaviour, both diffusion and condensation (5). We assume that in the case of low strength steel, hydrogen condensation is not sufficient enough to trigger an IG crack ahead of a QC crack. The consideration extracted from these results may well explain the behaviour of the QC crack in low strength steel which had grown to a larger length without changing its mode of growth from QC to IG.

Quasi-cleavage Cracking Process in Low Strength Steel

Since the IG cracks shows no contribution to the fracture of low strength steel specimens subjected to a sustained load with hydrogen, the cause of delayed fracture should directly depend on whether or not a QC crack can develop in the fracture process. The primary parameter that permits the development of a QC crack in the low strength specimen may be distinguished from various parameters through an examination of the differences in the cracking process between series A and other less sensitive specimens (series B～D). Figure. 6-(a) suggests that in a series A specimen tested under the applied stress of 550 MPa, fracture occurs with the initiation and coalescence of several subsurface QC cracks and dimple cracks. However, series B～D specimens possessing a slightly lower strength level than that of a series A specimen (as shown in Table 1.) did not fracture at the same applied stress level of 550 MPa. This implies two possibilities for QC crack formation of subsurface cracking in series B ～ D specimens: (I) no QC cracks had initiated, or (II) QC cracks had initiated at an early stage of loading, but remained as non-propagating cracks. To obtain a reasonable explanation for subsurface crack development of series B～D specimens which had not fractured, SEM observations were conducted on these specimens in the same manner as the series A specimen. Many micro-QC cracks and voids were revealed as a result of the SEM observation. Figure. 6-(c), (d) respectively illustrate micrograph examples of micro-QC cracks and voids in the cross section of a series B specimen which had not fractured at 550 MPa. Micro-QC cracks approximately 20 μm in length showed relatively sharp crack tip geometry, whereas larger cracks of approximately 40-50 μm were mostly blunted and resulted in a void-like crack tip shape. From this observation we can understand that the tip of a micro-QC cracks blunts in accordance to the increase in length and thus changes into a void-like geometry, resulting in a reduction of stress concentration, which consequently prevents sufficient condensation of hydrogen to satisfy the condition for onset of growth of an IG crack to develop from the tip of a QC crack.

CONCLUSIONS

(1) The fracture of low strength steel charged with hydrogen begins at a QC crack similar to that observed in high strength steel. In low strength steel, fatal fracture occurs only by the coalescence of sufficiently grown QC cracks which result in dimple cracks, where as in high strength steel the QC crack triggers an IG crack at the final fracture.
(2) The condition which dominates delayed fracture in low strength specimens rely not only on the possibility of the development of a QC crack, but also on the possibility of subsequent growth of the QC crack while maintaining a sharp crack tip geometry.
(3) Since the tip of a QC crack in low strength steel is extremely stretched compared to that in high strength steel, no sufficiently high local stress ahead of the QC crack can result in a diffusion and condensation of hydrogen to a crack tip to trigger the onset of growth of an IG crack. Consequently, a QC crack can continue to grow, without triggering an IG crack formation, until the tensile fracture occurs with dimples in ligament regions.

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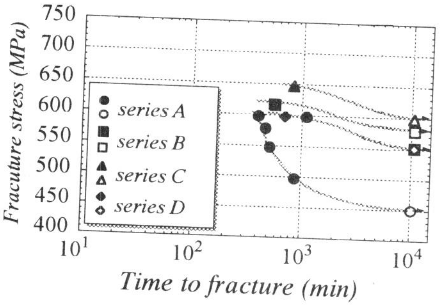


Fig.1 Delayed fracture test of low strength steel under sustained load

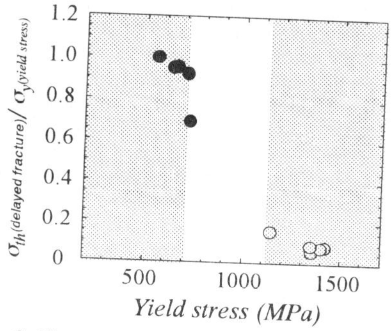
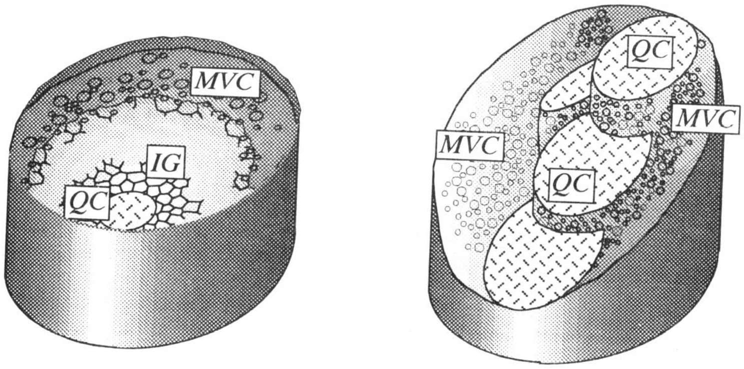


Fig.2 The relationship between delyea fracture sensitivity and yield stress



(a) Low strength steel (X-65)

(b) High strength steel (ASTM A-490)

Fig.3 Schematic fractograph illustrating the differences between low and high strength steels

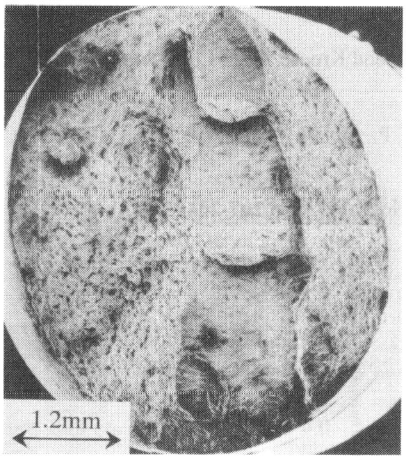


Fig.4 Characteristic of QC facet in fracture surface of low strength steel X-65

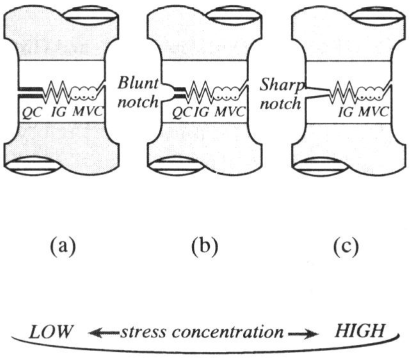
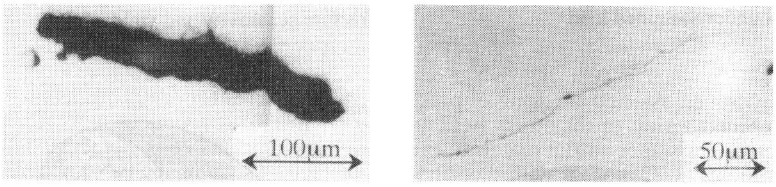
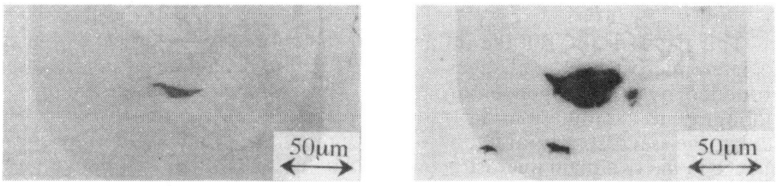


Fig.5 Characteristis of fracture process of hydrogen charged high strength steel



(a) Series A σ_{appl}=550MPa

(b) High strength steel (ASTM A-490)



(c)Series B σ_{appl}=550MPa (Micro-crack)

(d)Series B σ_{appl}=550MPa (Void)

Fig.6 Optical micrographs of cross-sections in low and high strength steels