

A STUDY OF HYDROGEN EMBRITTLEMENT (HE) ON TEMPERED CARBON STEEL

Astuty Amrin

and

Yusri Abu Bakar

Department of Production and Industrial Engineering

Faculty of Mechanical Engineering

Universiti Teknologi Malaysia

ABSTRACT

The influence of tempering temperature and environment on hydrogen embrittlement (HE) behavior of low carbon steel has been investigated. A range of tempered conditions was examined in air and exposed to corrosive media for observing the behavior of HE attack. Scanning electron microscopy technique has been used to examine the fracture mode associated with HE. It was revealed that the HE attack is more pronounced in specimens tempered at low temperature with the production of brittle fracture. The features indicating hydrogen induced cracking (HIC) on the fracture morphology, that is, micropores and ductile hairlines are also evident. The present results also showed that there is a strong link between temper embrittlement and hydrogen embrittlement in that changes in tempering treatment can give large varieties of HE resistance.

1.0 INTRODUCTION

It is well established that high strength steels are susceptible to embrittlement as a result of dissolved hydrogen and generally exhibit brittle fracture along prior austenite grain boundaries when stressed in hydrogen or hydrogen producing environment [1]. The embrittling effect of hydrogen on steels is based on the fact that it is located interstitially between the lattice atoms and causes localized distortion. The mobility of dislocations is restricted, and hence, the ability for plastic deformation to occur is markedly reduced.

Hydrogen embrittlement (HE) is especially prevalent in steels because of the restricted slip capabilities in the BCC structure (the number of slip systems is lower). Hydrogen also tends to concentrate strongly at suitable sites in the metal and these are referred to as traps. Such traps may be dislocations, grain boundaries, voids and non-metallic inclusions. The principal effects of HE in steels are a decrease in tensile strength and ductility when tested under static loads or low strain rates.

The microstructural state of the metal is particularly important since it controls the properties such as strength, toughness, ductility and susceptibility to HE. For a given environment, steels with different metallurgical structures response differently to hydrogen. In high strength steels tempering at higher temperature lowers the strength and improves the toughness and this would help to enhance the resistance to HE. The very severe effects of hydrogen on the performance of steels can be attributed to its role in allowing brittle fracture at ambient temperature. The damage due to hydrogen tends to be more severe at the highest strength level. This may be associated with the restricted plastic deformation of the stronger material which facilitates the transition to brittle fracture. Various mechanisms have been proposed for HE including hydrogen pressure, surface adsorption, decohesion and enhanced plastic flow [3, 4]. Bearing in mind that the concentrations of hydrogen in steels are very low, of the order of 1 ppm, it is very difficult to see how such small amounts of hydrogen can modify the fracture properties so dramatically and hence, know the exact mechanism involved.

Hydrogen pressure has long been thought to build up at preferred sites such as dislocations and second-phase interfaces and thus weaken the material [3]. Surface adsorption and decohesion models [3, 4] are based on the fact that either hydrogen decreases the strength of metal-metal bond, thereby facilitating brittle fracture by reducing the energy necessary to form brittle surface or that hydrogen can increase the stress required to emit dislocations from the crack tip, thereby making ductile fracture more difficult [5]. Another model of hydrogen induced cracking supported by considerable experimental evidence suggests that localized enhanced plastic deformation accounts for HE [6].

The major objective of the present work reported here is to examine the environmental and metallurgical aspects of the HE susceptibility of a medium plain carbon steel.

2.0 EXPERIMENTAL PROCEDURE

A commercial AISI 1050 steel was used in this investigation. The chemical composition of the steel was: 0.5 %C, 0.6 % Mn, and 0.3 % Si. Tensile test specimens were made from the bar, conformed to BS 18:1987, and had a gauge length of 185 mm and a minimal width of 7.5 mm. Charpy V-notch impact specimens 10 mm x 10 mm in cross section were also made from the steel bar. The specimens were austenized for 1 hour in a flowing argon atmosphere at 850⁰C and then water quenched. Decarburisation was avoided and a fully martensitic structure was obtained. Austenized specimens were tempered at temperatures of 200, 300, 400 and 500⁰C.

Tensile tests were carried out on an Instron 4206 universal testing machine and charpy impact tests were performed at room temperature on an Avery impact machine. Fracture surfaces were examined in a Philips XL40 scanning electron microscope operated at 30 Kv. The effect of the corrosive environment on the HE susceptibility was investigated by conducting salt spray tests in a saturated sodium chloride (NaCl) solution of 0.001 M with a pH of 6-7 for 6 hours at 35⁰C and seawater containing about 3.5 % NaCl. After cleaning and weighing, the heat treated specimens were immersed in seawater for a duration of 20 days. After this duration the specimens were taken out, cleaned, dried and then tested for tensile and impact and the fracture surfaces were examined in a scanning electron microscope for hydrogen embrittlement.

3.0 RESULTS AND DISCUSSIONS

Metallographic and fractographic investigations by SEM examination were carried out on a series of specimens, with and without corrosion environment, paying particular attention to the fracture surface characteristics as function of tempering temperature. Fractographic observations revealed more clearly the influence of the tempering temperature. Figure 1 shows that the fracture in the steel tempered at 200⁰C was predominantly intergranular. By contrast, the fracture mode in steel tempered at a higher temperature, i.e. at 500⁰C was predominantly transgranular, Fig. 2. The features shown by HE on the fracture surfaces include ductile hair lines and micropores at the grain boundaries⁷ as shown schematically in Fig. 3. These characteristic features of micropores and ductile hairlines were observed in the present study as shown in Fig. 1. Similar

findings were also observed with the presence of a corrosive environment of both seawater and saturated sodium chloride, that is an intergranular mode of fracture at low tempering temperature and the crack path following the boundary of prior austenite phase, Fig. 4. As the tempering temperature is increased a mixed mode of fracture was observed. Such a change in fracture mode with tempering temperature has been found in low alloy steels [8, 9].

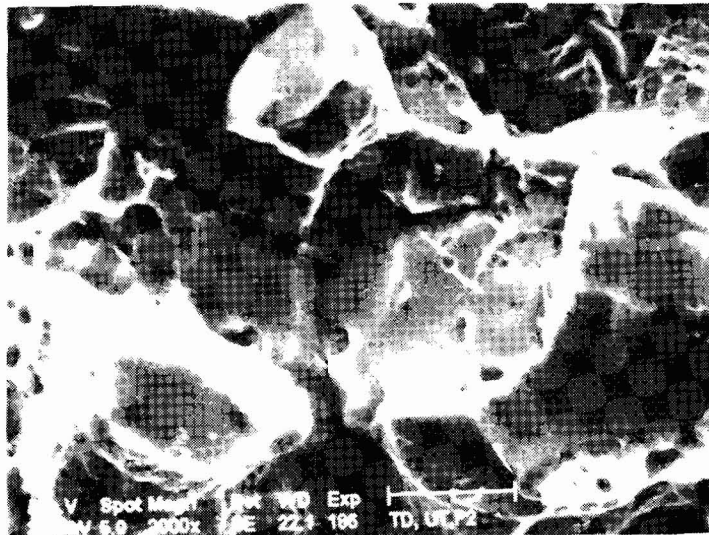


Fig. 1 SEM Fractograph of Tempered Steel at 200°C Showing a Predominantly Intergranular Fracture with Some Fine Dimples



Fig. 2 SEM Fractograph of Tempered Steel at 500°C Showing a Predominantly Transgranular Fracture

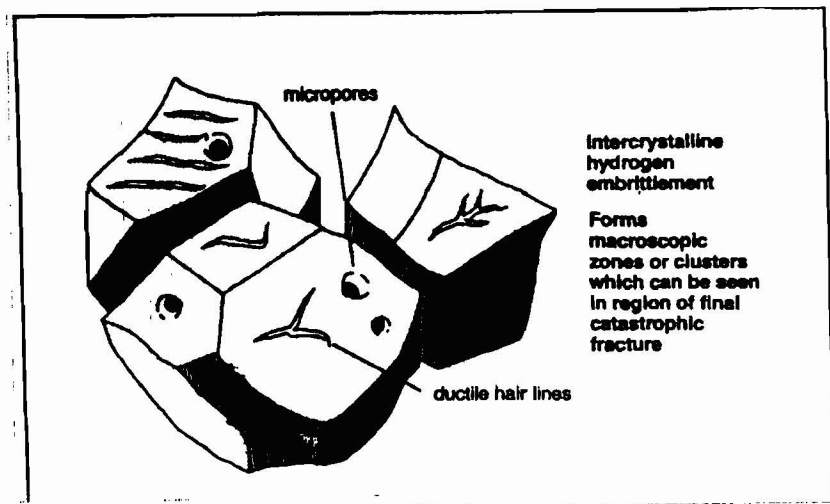


Fig. 3 Characteristic Features of Hydrogen Embrittlement (from Ref. 7)

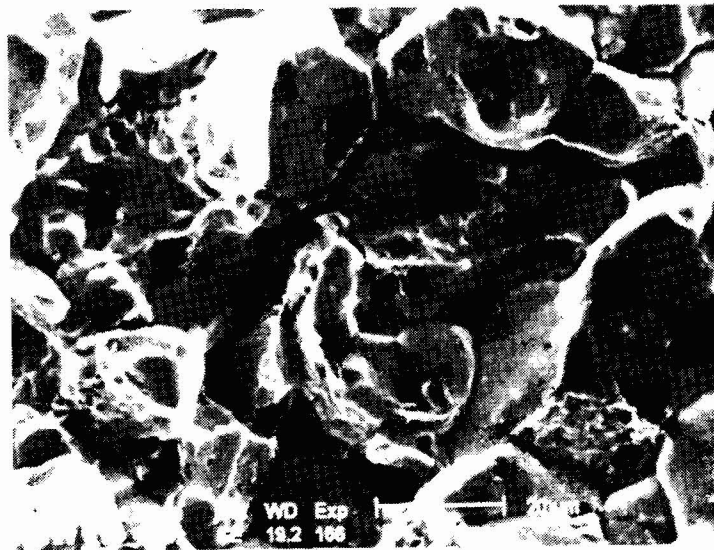


Fig. 4 SEM Fractograph of Tempered Steel at 200°C and Exposed to Saturated Sodium Chloride Solution Showing an Intergranular Fracture and the Crack Path Following the Prior Austenite Boundary

Previous work [10] on the permeation of hydrogen in tempered martensite structures showed the predominance of weak traps in the tempered martensite structures, thereby providing an increased susceptibility to HE. As shown in Fig. 1 the fracture mode in the low tempered martensitic structure was predominantly intergranular with fine dimples. These dimples are evidence of some local plasticity probably around a preferred site for hydrogen adsorption such as a non-metallic inclusion. The presence of ductile hairlines in hydrogen embrittled fracture surface, due to microplastic mechanisms [10] are also evidence of some local plasticity. In martensitic steels cracking occurs mainly along prior austenite grain boundaries, probably due to i) adsorption of hydrogen on grain boundary intersections such as triple points and ii) non-metallic inclusions located at prior austenite grain boundaries [11]. However, the dominance of the brittle intergranular fracture in low tempered martensitic specimens is clearly demonstrated and it seems more likely that the only crack initiation sites in tempered martensite structure are the prior austenite grain boundaries and that failure by HE is due to hydrogen adsorption at prior grain boundaries as shown in Fig. 5. A particularly important aspect of the microstructural state of the steel is the condition of the grain boundaries. Impurity segregation and carbide precipitation at the grain boundary may significantly modify its behavior. Thus, it is very likely that there is strong interaction between tempered embrittlement (TE) and hydrogen embrittlement (HE) [12]. Therefore, it is particularly important to obtain appropriate microstructure containing as many strong finely dispersed traps as is possible without any temper embrittlement.



Fig. 5 SEM Micrograph Showing the Prior Austenite Grain Boundary as the Likely Site for Hydrogen Embrittlement

With the presence of corrosive environment, the behavior of HE was unchanged but the corrosion rate increased due to the presence of chloride ions Cl^- . When combined with hydrogen, chloride ions (Cl^-) form an acidic environment thereby accelerating the corrosion phenomenon.

4.0 CONCLUSION

The present investigation has confirmed that the susceptibility to HE is enhanced in low tempered martensitic structures. HE structures fail in a brittle intergranular fashion due to the adsorption of hydrogen at the grain boundaries and that the presence of some fine dimples in the fracture surface are evidence of some localized plasticity due to hydrogen adsorption.

The behaviour of HE does not appear to be affected by the presence of a corrosive environment in the steel studied.

REFERENCES

1. Bandyopadhyay, N., Kameda, Jun and McMahan, J. C. Jr., Met. Trans., Vol 14A, 1983.
2. Zapffe, C. and Sims, C. , Trans. AIME, Vol 145, 1941.
3. Peteh, N. J. and Stables, P., Nature, Vol. 169, 1952.
4. Oriani, R.A. , Josephic, Acta Metall., Vol. 22, 1974.
5. Thompson, R. and Lin, I. H., Hydrogen Degradation of Ferrous Alloys, ed. R.A. Oriani et al. ,1984, p454.
6. Jones, D. A, Principles and Prevention of Corrosion, Prentice Hall, New York, 1992.

7. Engel, L. and Klingele, H., Atlas of Metal Samage, Prentice Hall, New York, 1981, p125.
8. Craig, B. D. and Krauss, G.; Met. Trans., Vol. 11A, 1980.
9. Banerji, S. K., et al., Mechanisms of Environment Sensitive Cracking of Materials, eds. Swann, Pr., The Metals Society, 1977.
10. Strecker, E, Ryder, D. A. and Davies, T., J. Iron & Steel Inst., 1969.
11. Cottis, R. A., Stress Corrosion Cracking of High Strength Steel, University of Manchester, 1989.
12. Hippsly, C. A., Material Science Technology, Vol. 3, 1987.