CRACK EXTENSION TESTING OF 316L STAINLESS STEEL UNDER SLOW STRAIN RATE AND HYDROGEN CHARGING

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The combined action of cathodic hydrogen charging and slow mechanical loading of precracked samples of 316 L stainless steel is examined in this work. As to the macroscopical failure mechanism crack extension testing of this steel and simultaneous hydrogen embrittlement showed no significant differences from the behaviour in air. For the loading rates considered, multi-cracking damage due to hydrogen slightly accelerates blunting and sub–notching of the crack tip prior to plastic collapse, but the damaged material does not concentrate in such a way that a mechanism of crack extension could appear.

INTRODUCTION

316 L steel is an extraordinarily ductile austenitic stainless steel which reaches a maximum uniform elongation of 60% in simple tension. It is known as a material of low hydrogen diffusivity and low sensitivity to hydrogen embrittlement. However, tensile tests of thin samples with previous or simultaneous hydrogen charging have shown moderate losses of strength and ductility which were attributed to the surface damage that hydrogen produces in the form of shallow cracking (1). This effect of hydrogen on the mechanical behaviour of 316L steel might restrict the use of this material for structural applications where hydrogen embrittlement could occur. So the phenomenon is worth examining under the most unfavourable conditions for the material, namely, stress and strain concentrations, slow strain rate and hydrogen charging by cathodic polarization. This is the aim of the present work in which cracks are used as stress and strain concentrators whereas in a previous work (2) notches were used for this purpose. This previous research shows that hydrogen damage consists in a localized and superficial multi-cracking process that produces the same effect on

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the mechanical behaviour of the specimen as a geometric enlargement of the notch, even for notch profiles with a curvature radius as small as 0.2 mm.

MATERIALS AND EXPERIMENTAL PROCEDURE

The steel used in the research was a commercially produced AISI 316L stainless steel whose chemical composition, mechanical properties and heat treatments are given in (2). It was supplied as a 3 cm thick plate with a final grain size of roughly 60 μm and an entirely austenitic microstructure.

The aim of the experiments entailing the most severe condition in which to study the effects of hydrogen on fracture resistance of 316 L steel, crack extension tests were carried out in air and hydrogen environments following the same methodology as in J-integral testing (3). Standard side-grooved compact tensile specimens with a gross thickness of 12.5 mm were fatigue precracked and tensile loaded by controlling the rate of the crack mouth opening displacement (CMOD) at the load line. Partial unloadings were performed during the tests in order to infer the crack extension from the compliance measurements. After testing and before being broken open, the specimens were subjected to cyclic loading to produce fatigue postcracking and facilitate their subsequent scanning electron microscope (SEM) fractographic analysis. This SEM analysis was performed to observe the fracture mechanisms and the influence on them of hydrogen .

The tests in air were continued beyond the CMOD gauge range and the CMOD was extrapolated from the cross head displacement. In spite of being precracked the specimens showed a such extraordinary ductility that a maximum load was not attained and no external sign of tearing appeared in them before a crack opening of half the specimen width. Since in these conditions the compliance variations are due both to large geometrical changes and to crack growth, the elastic compliance was not measured beyond the CMOD gauge range (7 mm). However, during an additional test the crack tip region was periodically photographed for visual evidence of the events that took place at this critical region.

The tests in hydrogen environment were performed by loading mechanically the specimens and at the same time charging them cathodically with hydrogen by immersion in an 1N H_2SO_4 aqueous solution containing Na As O_2 against hydrogen recombination, and by application of a cathodic potential of -1000 mV versus saturated calomel electrode (SCE). The potential was controlled by a potentiostat through a three electrode arrangement, the cathodic current density measured for that potential and for the solution previously de-aerated being 0.1 A/cm^2. A standard servohydraulic testing machine was used but the specimens were loaded horizontally so that the crack tip and the ligament of the specimen were immersed in the aggressive solution throughout the test. To this purpose a device was designed and constructed that allowed the vertical loads and displacements produced by the testing machine to be transformed into horizontal ones capable of being measured and used as test control signals. All the tests in hydrogen were carried out at CMOD rates ranging from 1 to 10 μm/min and were finished at a CMOD value of about 5 mm.

RESULTS AND DISCUSSION

The load–CMOD curves in air and hydrogen at CMOD rates of 10, 5 and 1 μm/min. are plotted in Figure 1. The effect of hydrogen and loading rate is gradual and moderate and does not indicate drastic changes in the fracture process. The compliance measurements in both air and hydrogen presented oscillations comparable with the differences to be recorded, which led to their substitution by the trend shown in each test as expressed by an interpolating function of the applied CMOD. The smooth curves J–Δa obtained by applying the method ASTM E 813 to these interpolated compliance values are shown in Figure 2 and imply no crack growth, since they are situated above the blunting line as defined in the aforementioned standard.

An interpretation of the physical events taking place at the end of the crack can be given from the sequence of photographs in which the crack tip can be clearly distinguished by its brightness. Up to the point near maximum load marked at Figure 3 as separating blunting and sub-notching parts of the curve, the crack tip is largely stretched and suffers a blunting process with no localized incorporation of new surface. After this point, new surface is locally added to the original crack surface, but hardly has it appeared when blunting extends over it, so that the process becomes notch enlargement by sub–notch formation instead of crack extension. Eventually, at the point well beyond maximum load indicated as initiation of crack extension in Figure 3, creation of new surface dominates over blunting and fracture propagates in the form of a growing crack. So, the specimen fails by plastic instability as a consequence of the reduction of the bearing area due to the notch enlargement involved in the blunting and sub–notching processes. According to the curves in Figures 1 and 2, hydrogen does not change this failure mechanism but seems to accelerate it increasingly as the loading rate decreases.

The SEM analysis confirms this and provides details remarkably similar to that reported in [2]. The photographs in Figure 4 belong to three specimens tested in air and hydrogen at CMOD rates of 10 and 1 μm/min and whose curves load–CMOD are plotted in Figure 1. The macrographs at the top show the crack tip after finishing the test, postcracking the specimen by fatigue and breaking it open. In the three cases the crack tip appears very blunted, the main difference between the three specimens being the state of damage to the notch tip into which the original crack tip has been transformed. This damage is attributable to hydrogen for its absence in the specimens tested in air, and consists of the increasing surface multicracking process that can be seen in the micrographs at the bottom of Figure 4 and that appears in the macrographs as shallow cracks weakening the tip of the blunted crack. Then the material damaged by hydrogen does not concentrate at the root of the blunted crack tip and damage propagation does not occur as crack extension

CONCLUDING REMARKS

Tensile testing and simultaneous hydrogen charging of precracked specimens do not alter the failure mechanism of 316 L steel reported in (2) for notched specimens. For the loading rates considered, its extraordinary ductility dominates over the hydrogen damage which is confined to the crack tip, so it accelerates slightly the plastic collapse of the specimen but does not produce a dominant crack extension mechanism. The hydrogen effect increases with time through the applied loading rate, but without significant repercussions on the failure mechanism.

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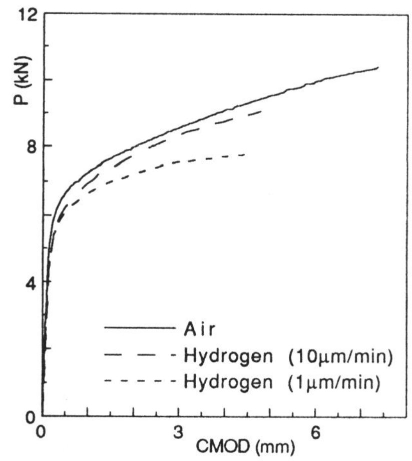

Figure 1 Load- CMOD curves in air and hydrogen

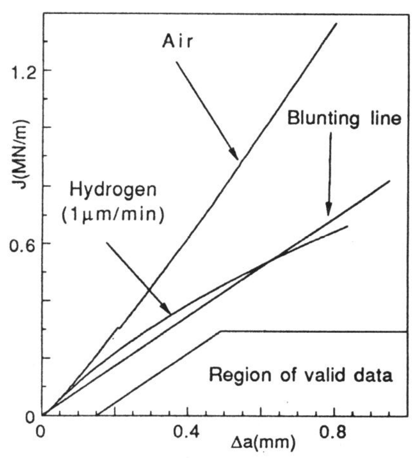

Figure 2 J–Δa curves as derived from the tests

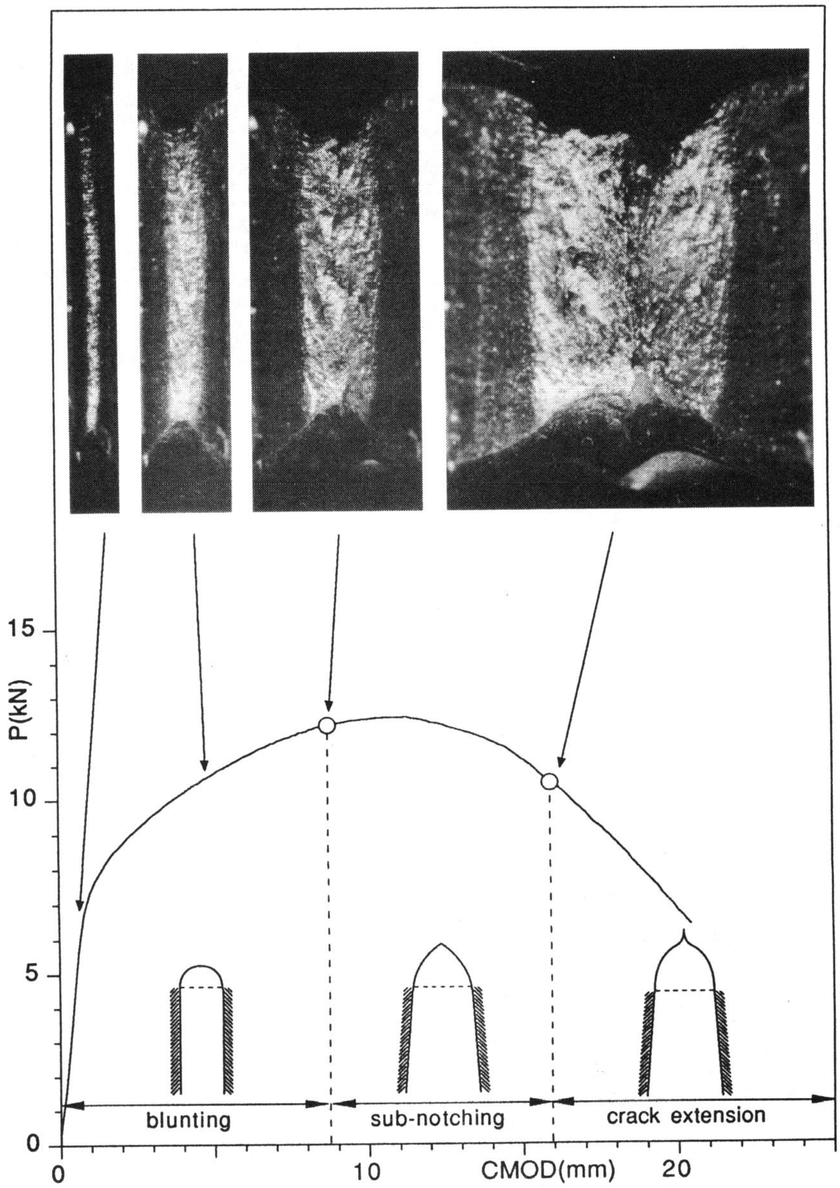

Figure 3 Tests in air: Load- CMOD curves and physical events at the crack tip

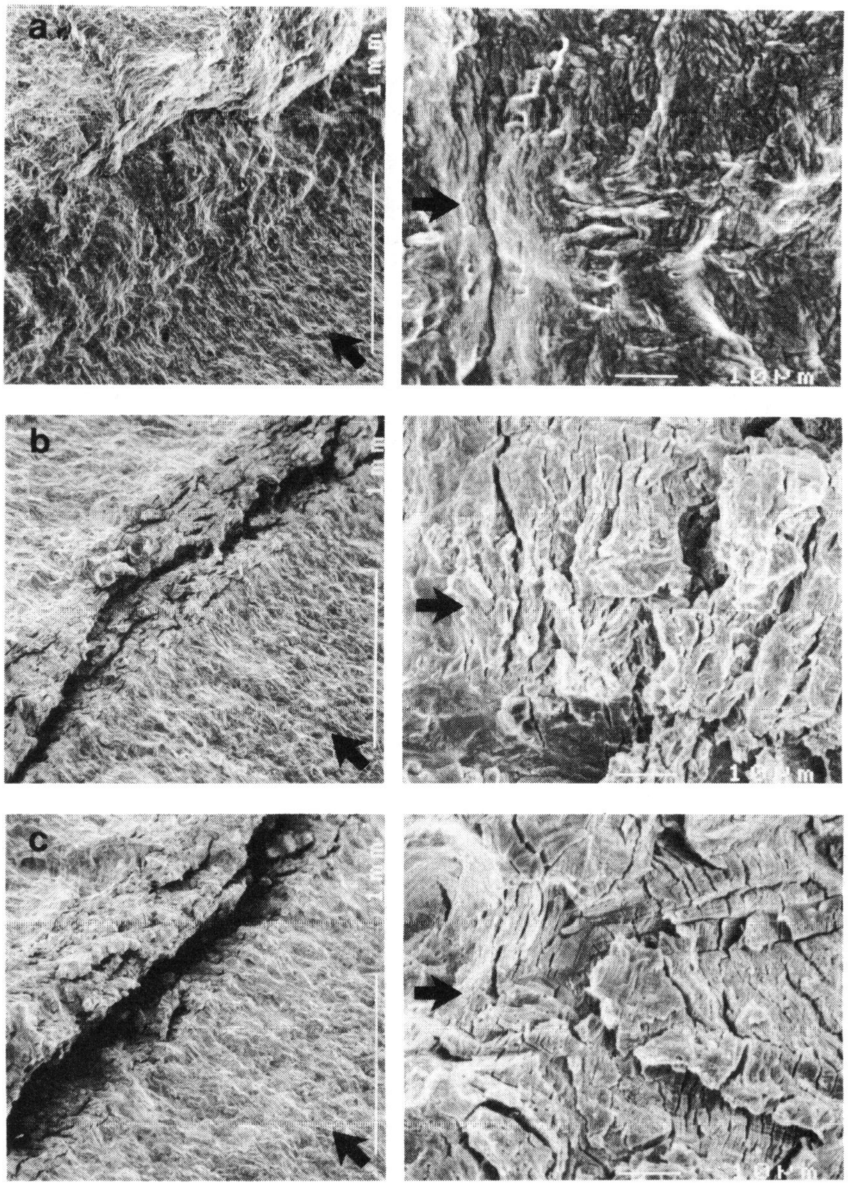

Figure 4 Macro and micrographs of the crack tip after crack extension testing: (a) in air; (b) and (c): in hydrogen, at CMOD rates of 10 and 1 μm/min