Cold Cracking Delay Times for Single Pass Weld Metal

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ABSTRACT

The study of hydrogen assisted cracking in steels has received considerable attention although certain aspects still elude academia and industry. The aim of this paper is to communicate physical research undertaken at The University of Wollongong (UOW) and the subsequent analysis at BMT Fleet Technology in relation to small scale testing of single pass weld deposits. The work presented entails the use of 4-point bend testing to derive hydrogen weld metal cold cracking delay times. It is proposed that a maximum delay time before cracking, rather the prediction of a single delay crack time for one condition, may provide more realistic expectations.

Two loading regimes were investigated. The first constituted a 5mm/min incremental load increase and the second regime ensured that the crosshead displacement was maintained when a specified load was reached. The results from the mechanically strained samples suggest that hydrogen-dislocation interactions deliver a significant influence on the expected delay times. It is argued that these interactions will then contribute to the variations in delay times observed for specimens welded and tested in near identical conditions. Other possible reasons for variations in weld metal delayed crack times will briefly be discussed and images from scanning electron microscopy will be presented to show fractured surfaces.

1. INTRODUCTION

It is known that the dominant effect of hydrogen on both heat affected zone (HAZ) and weld metal (WM) mechanical properties is embrittlement. This means that the presence of hydrogen in either region will reduce the work needed to induce mechanical failure. If the failure occurs after the weld has cooled to room temperature, the process is often referred to as cold or delayed cracking. This form of cracking was traditionally associated with the HAZ, but improved steel making processes and an increase in weld metal alloying content is transferring cold cracking from the heat affected zone into the high strength steel weld metal [1]. Furthermore, the development and expanded use of high strength steels is arguably also enhancing the risk for WM cold cracking to occur. The immediate challenge therefore faced by industry and academia relates to evaluating the major variables responsible for cold cracking of the weld metal, as traditional methods are based on experiences related to the HAZ.

Significant benefits can consequently be gained by producing test methods aimed preferentially at the weld metal and which also provide close control over the influencing factors that lead to its occurrence. The broad theme of this work therefore relates to mechanical testing of high strength ferritic steel weld metal and the subsequent investigations to elucidate the interactions between complex and sensitive factors that lead weld metal cold cracking. Although the focus here will be on differences in cold cracking delay times, it is anticipated that further use of the mechanical test methods presented will improve our understanding of delayed cracking mechanisms in the weld metal studied. It will do this by incorporating three variables strongly associated the cracking phenomena. These variables are diffusible hydrogen, applied stress and time.

2. BACKGROUND

One method by which the transport of hydrogen occurs within the microstructure is believed to result from the movement of dislocations. It is known from literature that the moving dislocations provide hydrogen trap sites, which then due to their mobility, also function as a vehicle by which the hydrogen is transported.

Krom and Bakker [2] produced experimental data to show that plastic deformation increases the density of hydrogen trap sites. Beachem [3] and Birnbaum and Sofronis [4] further argue that the absorption of hydrogen into the weld metal facilitates the generation and/or the motion of dislocations. It is therefore maintained that diffusible hydrogen becomes trapped by the dislocations and will then be transported to other areas within the microstructure during the movement of dislocations.

Earlier investigations performed by Johnson and Hirth [5] confirmed that mobile hydrogen could be transported to deposition sites by a process known as dislocation dragging. It is understood that the delivery of hydrogen to the deposition sites will be proportional to the strain rate. An increase in dislocation velocity will then also be accompanied by an increase in the flow stress, as shown by the work of Birnbaum and Sofronis [4].

It is expected that mechanical test methods which impose a dynamic strain on test specimens induce dislocation motion. If hydrogen is present, hydrogen transport will conceivably also occur even before macroscopic yielding is reached, assuming that the binding energies of the moving dislocations exceed the energy that permits general lattice diffusion. According to Toribio and Kharin [6], the dislocation transport mode that arises through continuous plastic straining will further result in a non-equilibrium distribution of hydrogen. The overall result will subsequently be shown by a decrease in diffusivity, brought upon by the trapping associated with new dislocations. Combining the arguments proposed by the references above will be central to the results presented later on in an attempt to explain differences in weld metal cold cracking delay times.

3. UOW - CONTROLLED MECHANICAL STRESS TEST

The work being presented is part of a larger project that was undertaken at The University of Wollongong. The aim of the larger project was to induce cold cracking in high strength ferritic weld metal and to determine whether a new test method could accurately predict delay times. This test relied on quantifiable levels of stress being applied to bead on plate specimens that contained estimated quantities of diffusible hydrogen. These estimations were based on performing diffusible hydrogen testing on matching test weld specimens.

The welded specimens contained two levels of diffusible hydrogen, which were achieved by introducing 2% H_2 and 5% H_2 to the prescribed CO_2 shielding gas. The applied load was converted into stress by using a 95% confidence interval to predict the weld bead cross sectional area. This was done by examining the cross sectional areas of numerous welds produced under similar test conditions. The statistical prediction of a standardized test weld profile was then used to determine the applied stress, as the applied stress was a function of the specimen geometry and loading conditions. A specific requirement of the mechanical testing was to produce hydrogen assisted crack initiation and propagation that was principally influenced by the existing microstructure, the applied load and the level of hydrogen. No artificial stress raisers, such as notches or pre-cracks were therefore introduced into the test specimens.

A single run of high-strength ferritic weld metal (720 MPa yield) was deposited on a quenched and tempered martensitic base material (690 MPa yield) via an automated flux cored arc welding process. A heat input of 1.5 kJ/mm was maintained for all welding trials. Storage of the specimens in liquid nitrogen prior to testing was conducted using a fixed procedure, as this arrested loss of hydrogen from the weld. The subsequent mechanical testing was performed once the samples were brought back to room temperature.

The primary means of mechanical testing was achieved by 4-point bending, in agreement with Pussegoda et al [7] and Graville [8]. These authors demonstrated that bend testing can be used as a valid platform from which the weld metal's susceptibility to cold cracking can be gauged. Their specimen geometry had a charpy type v-notch placed in the selected area of the weld, whereas the four point bend specimen used during the current investigation did not contain a notch. The use of 4-point bending would furthermore evaluate a greater volume of weld metal, when compared with 3-point bending. The 4-point bend arrangement consisted of two rollers on the load span pushing against two rollers on the support span, see Fig. (1). The support span (S) was positioned on the deposited weld bead and the load span (L) was positioning behind the weld on the base material. This orientation then induced a tensile stress in the weld bead and a compressive stress on the base material. The shorter load span was subjected to a uniform bending moment and would produce uniform surface stresses across the weld metal (excluding variations related to weld surface profile).

Calculation of the second moment of area and the imposed bending moment was required to define the stresses in the weld bead. This was achieved by assuming that the bead approximated half an ellipse, positioned on top of the rectangular base material. The moments of inertia and centroid positions of the base and the weld could then be calculated. The final stress and strain at the top of the weld was calculated by determining the imposed bending moment. The effects of plasticity were not included during derivation of the bend test results.

Use of this test method would however provide a close approximation of the applied stress and will therefore related the stress and the level of introduced diffusible hydrogen to weld metal cold cracking delay times. As an example, an applied bending stress of 900 MPa, an approximated diffusible hydrogen content of 15ml per 100g of deposited weld metal and a high strength ferritic weld bead would translate into a delayed crack time of 100 seconds (see Fig.3). This pattern was reproduced for different loading and hydrogen conditions and typical results are presented in Figs. (2, 3). The results are for both rising load (5mm/min) bending, shown in Fig. (2) and deflection controlled bending, shown in Fig. (3). For the latter, a defined stress was applied by first straining the specimen to a prescribed load, then maintaining that position while still recording the load. Supplementary details to describe the test process and the stress calculations are available in Reference [9].

The tested specimens were subsequently fractured in liquid nitrogen to expose the failed region and analyzed via scanning electron microscopy. Selected images from one fractured specimen are presented in Figs. (5-8). The images also include a microcrack produced under 5% H_2 that was situated approximately 2.5mm below the surface of the weld bead. All cracking was achieved by subjecting the weld metal to global post-yield strains. Strains lower than this not did result in surface breaking crack activity.

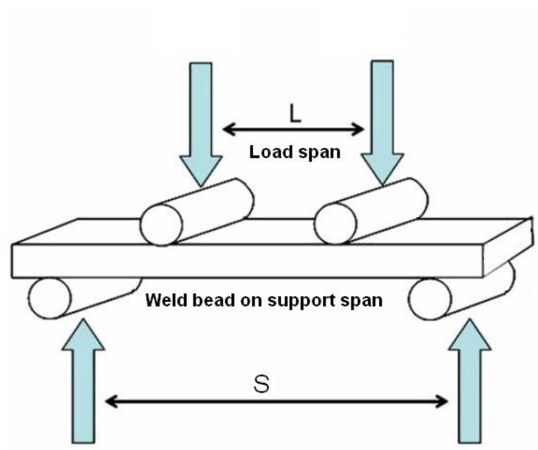

Figure 1:
Illustration of loading conditions in a 4-point bend test**.** The load span (L) pushes down on the support span (S) to induce uniform surface stresses across the load span

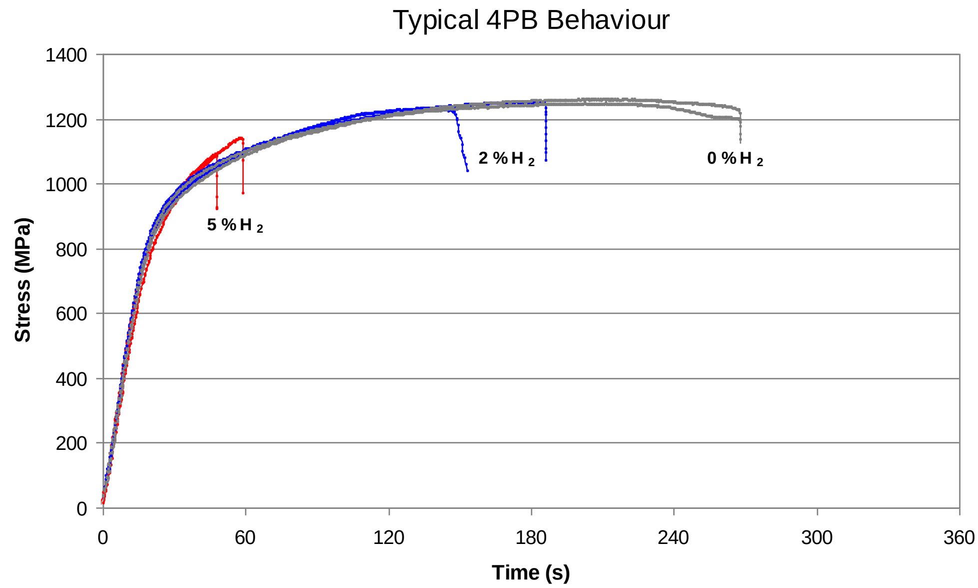

Figure 2: Typical 4-point bend mechanical behaviour (rising load) of welded test specimens which represent the additions of H_2 to the shielding gas. All specimens were welded and tested in near identical conditions. Note the increase in strength of the 5% H_2 specimens and the expected reduction in ductility.

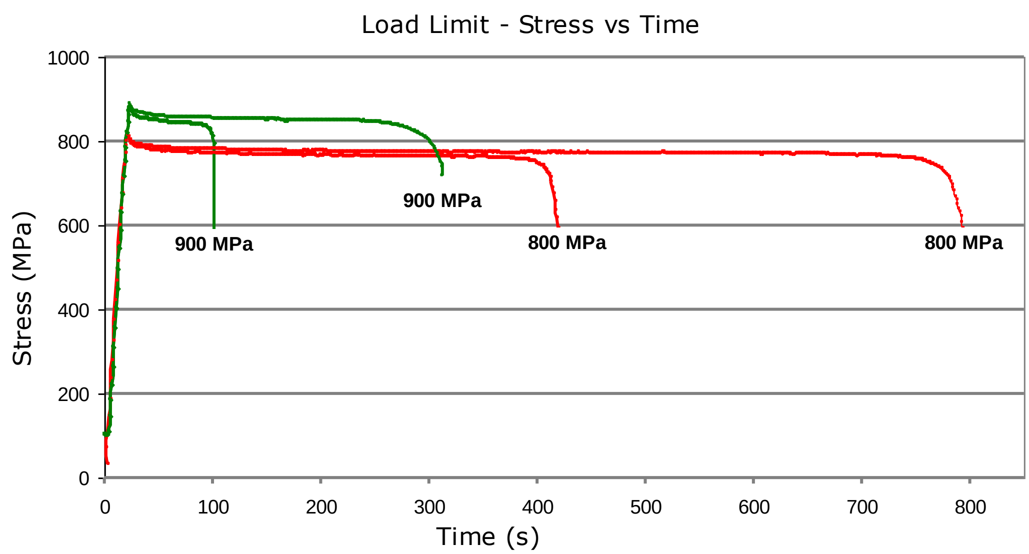

Figure 3: Example behaviour of stress controlled testing relating to specimens containing 5% H_2 in the shielding gas. For this type of testing, the crosshead displacement was maintained when a specified load was reached. The load at this displacement corresponded to specific values of stress, which was determined via normalizing the weld bead geometry using a 95% confidence interval. Note the reduction in stress with time, which later resulted in surface breaking cracking.

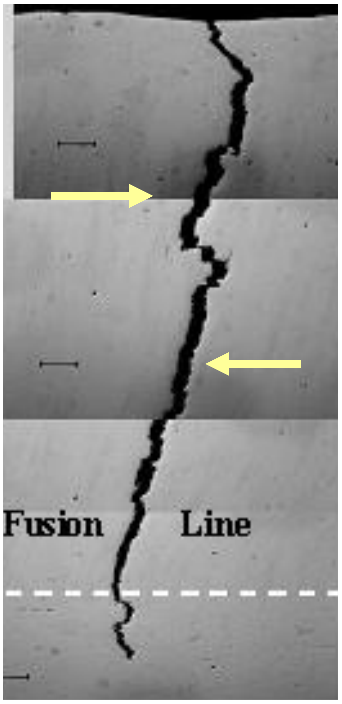

Figure 4: The image to the left represents the side profile of a specimen sectioned along the weld centre line. The crack appears to have formed through the linking of at least two earlier cracks, indicated by the arrows. All surface breaking cracks propagated throughout the weld and arrested in the HAZ.

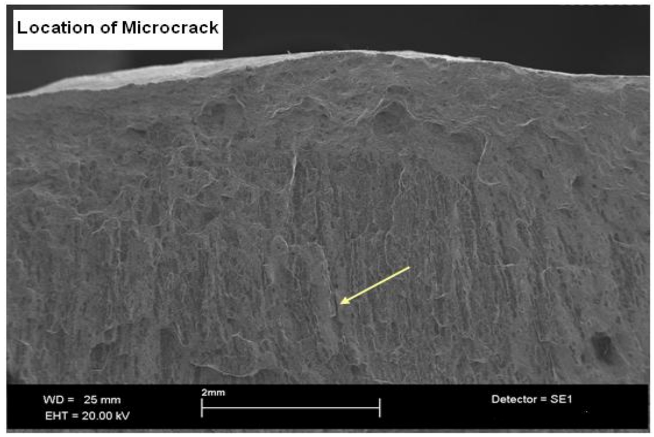

Figure 5: The image above to the right is a scanning electron image of a failed specimen that was subsequently fractured in liquid nitrogen. The arrow points towards a microcrack that appears to follow the path of solidification. Of note is that although the outer surface of the specimen was in tension, the orientation of the microcrack here suggests that multidirectional stresses contributed to the creation of this microcrack.

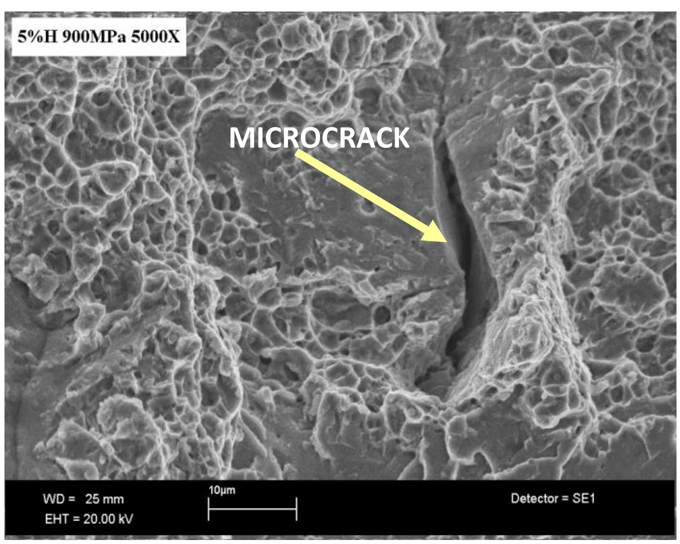

Figure 6: Scanning electron image of the crack face. This microcrack possibly initiated at a non-metallic inclusion, shown in Fig. (7). Both microvoid coalescence and quasi cleavage is noticeable in the proximity of the crack.

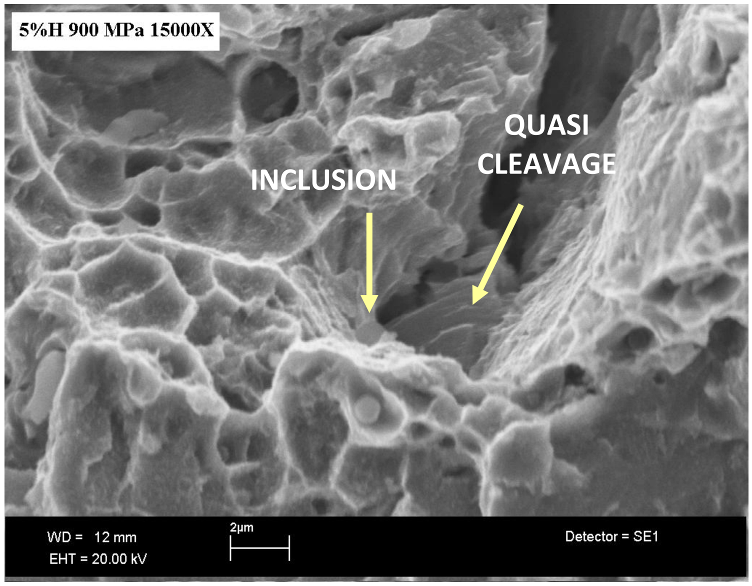

Figure 7: Scanning electron image of the bottom end of the microcrack shown in Fig. (5) which indicates the location of an inclusion as a possible crack initiation site. The quasi cleavage facet in which the inclusion is situated is likely to have been created by the interaction between the local microstructure and hydrogen.

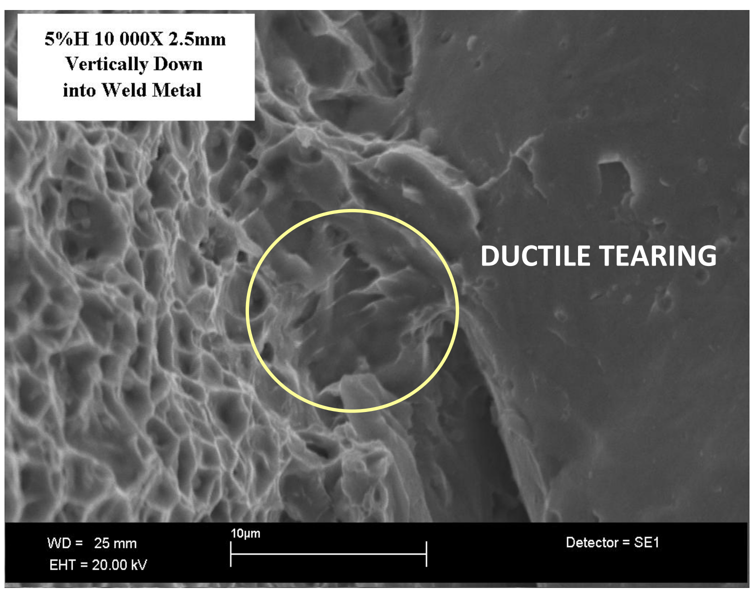

Figure 8: Scanning electron image at top end of the possible crack initiation site, i.e. the top of the crack in Fig. (5). The quasi-cleavage facet associated with the initiation site has been replaced by a ductile tearing mechanism. It is assumed that this is due to the propagating crack front escaping the hydrogen atmosphere.

5. DISCUSSION

The test methods used during this investigation were reliant on surface breaking cracks for the specification of delay times. It should be noted that these methods did not record crack initiation. The primary reason for the differences in delay times is therefore suspected to be associated with the distance that the propagating crack had to travel before it broke the surface. This means that a longer crack path before fracture of the outer fibres of the specimen would result in longer delay times.

It is also believed that even if crack initiation times for a given condition were similar, the subsequent fracture paths for different specimens would have to include variations in the location of crack initiation sites. A secondary reason for differences in delay times will therefore require closer evaluation of hydrogen trap sites, which is assumed to vary in weld metal specimens produced under similar conditions. It is most likely that hydrogen will influence the mobility of dislocations by altering the elastic interactions that exist between dislocations and other stress centres, as described by Sofronis and Robertson [10]. The dislocations may then provide trap sites which have the capacity to transport hydrogen to other regions in the lattice.

The two test methods (rising load and stress controlled) display two opposite effects of hydrogen-dislocation interactions when considering Figs. (2, 3). The effects can be seen as both hardening and softening of the material. Dislocation dragging can be seen in Fig. (2), where an increase in strength with an increase in the level of hydrogen present in the shielding gas is observed. In this region, the expected movement of dislocations is retarded due to its inability to break away from the hydrogen atmosphere, effectively anchoring the dislocations and causing a drag effect. The subsequent increase in applied load, without the required increase in plastic behaviour therefore translates into a strengthening mechanism.

In Fig. (3), the opposite is proposed when the applied strain is fixed and dislocations mobility is believed to continue without the application of further load. This can be seen by closer analyses of the stress versus time curves, where a decrease in gradient is observed with an increase time, i.e. a softening effect. In other words the moving dislocations constitute the decrease in stress with an increase in time. Here it is believed that the hydrogen swept along by the moving dislocations now facilitate the spawning and further mobility of dislocations, as proposed by Beachem [3] and Birnbaum and Sofronis [4]. The final fracture subsequently occurs when sub-surface cracks break through the outer fibres of the weld metal.

Whether stress assisted diffusion or dislocation motion resulted in the final fracture is not certain at this stage, as there is recognition that hydrogen can diffuse rapidly in ferritic steels to existing crack tips and then facilitate further extension without dislocation transport. It should however be noted that the applied stresses imposed in Fig. (3) is considerably above yield, implying the occurrence of advanced dislocation motion and therefore advanced hydrogen transport.

The proposed involvement of dislocations to explain the mechanical behaviour seen in this work is not conclusive and further in situ investigations on strained specimens with the aid of transmission electron microscopy will be useful. As an example, the investigations conducted by Sofronis and Robertson [10] can be used as a guide to determine what role dislocation transport of hydrogen will have on strained samples. It may then be possible to assess whether stress assisted diffusion or diffusion via dislocations dominate at elevated strains.

In conclusion, the results of this work show that differences in delay times for near identical conditions are to be expected. It is therefore proposed that a maximum delay time, rather than the prediction of a single delay crack time will provide more realistic expectations when conducting similar research.

6. ACKNOWLEDGEMENTS

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