Materials Science & Engineering A 685 (2017) 87–94

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea

Fracture toughness and hydrogen embrittlement susceptibility on the interface of clad steel pipes with and without a Ni-interlayer

L. Jemblie[a,*], H. Bjaaland[a,1], B. Nyhus[b], V. Olden[b], O.M. Akselsen[a,b]

[a] Department of Engineering Design and Materials, NTNU, 7456 Trondheim, Norway
[b] SINTEF Materials and Chemistry, 7456 Trondheim, Norway

ARTICLE INFO

Keywords:
Clad pipe
Cladding
Hydrogen embrittlement
Fracture toughness
Electroplating
Interface

ABSTRACT

The objective of the present work has been to study the fracture properties of the interface between clad and base material of two 316 L austenitic stainless steel - X60/X65 carbon steel hot roll bonded clad pipes; with and without a Ni-interlayer. Fracture mechanical tests were performed in air and under in situ electrochemical hydrogen charging to establish crack growth resistance curves and fracture initiation toughness for both systems. The results revealed that an electroplated Ni-interlayer reduces the fracture initiation toughness for testing in air, while it raises the fracture initiation toughness for testing in hydrogen environment. The samples with a Ni-interlayer revealed little influence of hydrogen on the fracture resistance, with a reduction in the fracture initiation toughness of 20%, attributed to crack propagation mainly occurring in the nickel layer. The samples without a Ni-interlayer revealed a strong influence of hydrogen on the fracture resistance, with a reduction in the fracture initiation toughness of 85%. An alternating crack path was proven, shifting between the dissimilar interface and the base material adjacent to the interface.

1. Introduction

Steel pipelines represents an important part of the subsea infrastructure for transport of unprocessed oil and gas. In areas facing high content of corrosion inducing products, there is a demand for pipes with a corrosion resistant interior, able to withstand environmental degradation and cracking during full service life. Composite pipes, where a corrosion resistant alloy (CRA) is internally bonded to a conventional carbon steel pipe, has become increasingly popular as an economical viable option for corrosion management, combining the mechanical properties of the structural steel with the corrosive properties of the CRA. This however offers new challenges with respect to integrity management and degradation assessment, due to an inhomogeneous material combination and a complex interface region.

Clad steel pipe refers to pipes where the bond between the base material (BM) and the CRA is metallurgical, as opposed to lined pipes, where the bond is mechanical. The principle manufacturing method combines hot rolling of clad plates to form the metallurgical bond, followed by bending into the shape of a pipe in a press bending process [1,2]. During production, due to the wide difference in chemical composition between the base material and the clad layer, considerable transport of elements across the interface may occur. This is especially prominent for carbon, considering the element's high mobility. The resulting interface, while highly dependent on the production parameters, is microstructural complex with increased hardness and carbide precipitates on the clad side and carbon depletion followed by grain growth in the ferritic base metal [3–8].

The presence of inter metallic compounds, hard zones and residual stresses may significantly degrade the dissimilar interface, making it prone to hydrogen induced cracking. Recently a series of failures on cathodically charged subsea dissimilar welds have been attributed to hydrogen embrittlement (HE), where the presence of carbide precipitates at the interface resulted in a microstructure particularly sensitive to hydrogen induced failures [9,10]. Hydrogen induced degradation of mechanical properties is a well recognized threat in subsea structures and pipelines, with several reported incidents. It manifests as loss in toughness, which may result in unexpected and premature catastrophic failures. The basic mechanisms responsible for HE are still under debate, however two theories have advanced as the more accepted ones for the case of hydrogen degradation in steel: Hydrogen Enhanced Decohesion (HEDE), in which interstitial hydrogen reduces the bond strength and thus the necessary energy to fracture [11,12]; and Hydrogen Enhanced Localized Plasticity (HELP), in which atomic hydrogen accelerates

* Corresponding author.
E-mail address: lise.jemblie@ntnu.no (L. Jemblie).
[1] Present address: Technip Norge AS, Grønøra Industriområde, 7300 Orkanger, Norway.

http://dx.doi.org/10.1016/j.msea.2016.12.116
Received 26 September 2016; Received in revised form 15 December 2016; Accepted 28 December 2016
Available online 31 December 2016
0921-5093/

dislocation mobility through an elastic shielding effect which locally reduces the shear stress [13,14]. Today it is seemingly recognized that no single mechanism can comprehensively explain all the phenomena associated with HE, rather it appears that a combination of mechanisms is more likely in many cases.

The objective of the present study has been to investigate the fracture susceptibility and fracture behaviour of 316 L austenitic stainless steel - X60/X65 carbon steel hot roll bonded clad pipes, both in air and with respect to hydrogen degradation. Bjaaland et al. [7] reported that the presence of a Ni-interlayer between the base material and the clad layer limits carbon diffusion across the interface, thereby preventing the formation of carbide precipitates and hard zones in the clad. The effect of an interface Ni-layer on the fracture susceptibility and degradation is therefore of special interest. Compact tension (CT) fracture mechanical testing has been performed in order to establish the crack growth resistance curves and fracture initiation toughness. For comparative reasons, values of the critical fracture toughness has also been determined. The fracture surfaces and fracture surface profiles has been investigated and related to the fracture toughness results.

2. Materials and methods

2.1. Materials

Two different clad steel pipes are investigated in this study, presented in Table 1 as Sample A and Sample B respectively, with the main difference being the presence of a nickel interlayer between clad and base material for Sample A. The chemical composition and tensile properties are presented in Table 2.

The clad layer is bonded to the pipeline steel plate through hot rolling followed by quenching and tempering, before bent into the shape of a pipe in a press bending manufacturing process. Specimens for investigation were extracted in the longitudinal direction of the pipes in as supplied condition. The interface microstructure of both samples are presented in Fig. 1. For Sample A, the 30–35 µm thick nickel interlayer is clearly visible, while nearly no carbide precipitates are visible. Sample B displays a continuous area of carbide precipitates (~200 µm wide) on the clad side, attributed to carbon diffusion across the interface during production.

Microhardness profiles across the interface of both samples are presented in Fig. 2, measured in the heat affect zone, 700 µm from the pipe girth weld. Sample B displays a hardness peak adjacent to the interface on the clad side and an abrupt drop in hardness adjacent to the interface on the BM side before a minimum value is attained, confirming carbon diffusion across the interface. For Sample A, no apparent hardness peak is visible on the clad side of the interface, while a minimum value on the BM side indicates some carbon diffusion towards the Ni-interlayer. The microhardness profile of Sample A is consistent with results by Missori et al. [4] and Dhib et al. [8] on hot roll bonded carbon steel-austenitic stainless steel clad plates. Bjaaland et al. [7] found that carbon diffusion mainly take place during the production process, with only minor contribution from the welding, confirming the validity of these results also for the un-welded case. A detailed interface characterization has previously been performed, and is reported in ref. [7,15].

Table 1
Investigated samples.

	BM	Clad	Ni-interlayer	Clad thk. [mm]	Pipe wall thk. [mm]
Sample A	X65	316 L	Yes	3.0	16.0
Sample B	X60	316 L	No	3.0	15.7

2.2. Fracture mechanical testing

2.2.1. Experimental testing

Constant load rate CT fracture mechanical testing was performed in air and under cathodic protection (CP), in order to establish CTOD (Crack Tip Opening Displacement)-R curves and values for crack initiation. For comparative reasons, values of critical CTOD were also determined.

CT specimens were machined with the notch tip at the dissimilar metal interface to an initial crack length to width ratio a_0/W of 0.5, using electro-discharge machining (EDM), as it was deemed impossible to produce a fatigue crack propagating exactly along the dissimilar interface. Details of the specimen geometry and dimensions are given in Fig. 3. Prior to machining, the samples were lightly etched in 2% Nital to better reveal the dissimilar interface.

A constant loading rate of 0.74 N/min was applied, corresponding to a stress intensity rate of $6.8 \cdot 10^{-4}$ MPa $m^{1/2}$/s . This is in accordance with the work by Lee and Gangloff [16] on hydrogen assisted cracking of ultra-high strength martensitic steel, making the resulting fracture toughness independent of the loading rate. For testing in hydrogen environment, the specimens were immersed in a 3.5% NaCl solution with an applied cathodic potential of -1050 mV_{SCE}. Prior to test initiation, the specimens were hydrogen pre-charged in the test rig for 24 h at -1050 mV_{SCE} in a 3.5% NaCl solution. The charging conditions were maintained throughout the entire pre-charging and test period. Pre-charging time was decided by diffusion calculations in the BM, based on the thick plate solution of Fick's law. Using a diffusion coefficient equal to $2.50 \cdot 10^{-10}$ m^2/s experimentally measured on X70 pipeline steel, a hydrogen concentration level above 1.27 wppm was estimated throughout the BM, deeming 24 h pre-charging time sufficient. All testing was performed at room temperature.

For determination of CTOD-R curves, a multiple specimen procedure was applied where the specimens were unloaded at different CTOD values in order to establish points on the curve for various crack extensions. The extent of stable crack growth was marked with heat tinting. After testing the samples were cracked open in liquid nitrogen, and the crack length and crack extension was measured at 5 equally spaced points across the sample, obtaining the original crack length a_0 and the mean crack extension Δa according to

$$a_0 = \frac{1}{4}\left(\frac{a_1 + a_5}{2} + \sum_{1=2}^{1=4} a_i\right) \tag{1}$$

$$\Delta a = \frac{1}{4}\left(\frac{\Delta a_1 + \Delta a_5}{2} + \sum_{1=2}^{1=4} \Delta a_i\right) \tag{2}$$

where a_1 and a_5 refers to the two measurements at the outer points. Due to non-uniform crack growth, the maximum crack extension for each sample was also measured.

2.2.2. Analysis of test data

During testing, the load and the machine displacement were recorded. For testing in air, a machined clip gage, made to fit the small size of the specimen, was used to measure the Crack Mouth Opening Displacement (CMOD) at the knife edges. Due to the design of the fracture mechanical testing rig, it was not possible to use clip gages for testing under CP, where the specimen was immersed in a 3.5% NaCl solution. Rather, the average ratio between plastic CMOD (V_p) and plastic displacement (d_p), resulting from testing in air, was used to obtain the plastic CMOD under CP

$$V_{p,\mathrm{CP}} = \left(\frac{V_{p,\mathrm{air}}}{d_{p,\mathrm{air}}}\right) \cdot d_{p,\mathrm{CP}} \tag{3}$$

at the point of maximum load for determination of critical CTOD and at the point of unloading for determination of the fracture resistance curves.