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Stress corrosion cracking initiation and short crack growth behaviour in Alloy 182 weld metal under simulated boiling water reactor hydrogen water chemistry conditions

Juxing Bai[a], Stefan Ritter[a,*], Hans-Peter Seifert[a], Sannakaisa Virtanen[b]

[a] Paul Scherrer Institut, Nuclear Energy and Safety Research Division, Lab. for Nuclear Materials, 5232 Villigen PSI, Switzerland
[b] Friedrich-Alexander University Erlangen-Nürnberg, Germany

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ABSTRACT

The effect of dissolved hydrogen on the stress corrosion cracking initiation and short crack growth behaviour of Alloy 182 weld metal was evaluated in 274 °C hydrogenated high-purity water using accelerated crack initiation and growth tests with sharply notched fracture mechanics specimens and in-situ crack initiation and growth monitoring. A maximum in initiation susceptibility and crack growth rates was observed at the Ni/NiO phase transition line. Grain boundary misorientations and mismatch in Schmid factor were measured by electron backscattered diffraction along intergranular stress corrosion cracks and at crack initiation sites. Low-angle boundaries seem to be particularly resistant to cracking.

1. Introduction

The Ni-based Alloy 182 is widely used in light water reactors as weld filler metal and attachment pad metal to join the low-alloy steel reactor pressure vessel to both wrought Ni-based alloys (e.g., Alloy 600) and austenitic stainless steels (e.g., AISI 304L, 316L). In recent years, several intergranular stress corrosion cracking (IGSCC) incidents occurred in Alloy 182 dissimilar metal welds in both boiling water (BWR) [1–4] and pressurized water reactors (PWR) [5–7], which affected the safe and economic operation of nuclear power plants. In case of BWRs, components such as different nozzle safe ends [4], bottom head penetration housings [8] and core shroud support welds [9] have suffered from SCC. The cracking was usually confined to the weld metal and none of the SCC cracks significantly penetrated the adjacent reactor pressure vessel base material, which is consistent with the very high SCC resistance of low alloy steel under light water reactor conditions [10]. The possibility of SCC crack growth in Alloy 182 weld metals in high-purity BWR (neutral high-purity water, 274–288 °C, $pH_T = 5.6$) and PWR (slightly alkaline borated and lithiated water, 290–360 °C, $pH_T \sim 7$, 2–3 ppm dissolved hydrogen (DH)) environment has also been demonstrated by laboratory investigations. In PWR environment, SCC in Ni-based alloys is strongly influenced by the temperature (high Arrhenius activation energy for SCC of ~130 kJ/mol)) and the DH content of the high-temperature water. A peak in SCC initiation and crack growth susceptibility at the Ni/NiO phase transition line was observed at 320–360 °C [11,12]. This thermodynamic boundary is predicted to decrease from 2.3 ppm DH at 360 °C to 0.25 ppm DH at 274 °C (Table 1) [13,14]. In BWRs with moderate hydrogen water chemistry (HWC, ~100 to 300 ppb H_2) or noble metal chemical application (NMCA)-HWC (~20 to 40 ppb H_2), the DH concentration is close to this DH concentration (Table 2). Andresen et al. [15] confirmed the existence of this maximum in SCC crack growth rate for Alloy 182 at the Ni/NiO phase transition line at lower temperatures of 300 °C in BWR environment also. Richey et al. [16,17] measured the SCC initiation times as a function of DH for Alloy 600 at 360 °C and compared them with the reciprocal of crack growth rate data, and has shown that the DH effects on SCC initiation and crack growth rate were analogous. More recent results from the authors [18] indicate that the SCC initiation time is much longer in the NiO region than in the Ni region, so a lower DH level seems more effective to mitigate SCC initiation. The

Abbreviations: ASTM E1681, standard test method for determining a threshold stress intensity factor for environment-assisted cracking of metallic materials under constant load; ASTM E399, standard method of test for plane strain fracture toughness of metallic materials; BWR, boiling water reactor; C(T), compact tension fracture mechanics specimen; CSL, coincidence site lattice; DCPD, direct current potential drop; DH, dissolved hydrogen; EBSD, electron backscattered diffraction; ECP, open circuit free electrochemical corrosion potential; EDX, energy dispersive X-ray spectroscopy; EPRI, Electric Power Research Institute; FEGSEM, field emission gun scanning electron microscope; HAB, high-angle grain boundary; HWC, hydrogen water chemistry; IGSCC, intergranular stress corrosion cracking; LAB, low-angle grain boundary; NMCA, noble metal chemical application; OM, optical microscopy; ppb, part per billion, μg/kg; PWR, pressurized water reactor; SCC, stress corrosion cracking; SEM, scanning electron microscopy; UTS, ultimate tensile strength; YS, yield stress

* Corresponding author.
E-mail address: stefan.ritter@psi.ch (S. Ritter).

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Table 1
DH concentration and ECP (for high-purity water) at Ni/NiO transition line.

Parameter	220 °C	274 °C	288 °C	360 °C
H_2 Ni/NiO [ppb]	~50	253	362	2300
ECP Ni/NiO [mV_{SHE}]	−456	−535	−556	−645

Table 2
Different BWR water chemistries with corresponding hydrogen levels and the corresponding ECP and Ni oxidation stability regions (NWC: normal water chemistry, M-HWC: moderate hydrogen water chemistry, OLNC: on-line NMCA).

Water chemistry	Feed water H_2 220 °C [ppb]	Reactor water H_2 274 °C [ppb]	*ECP* [mV_{SHE}]	*ECP-ECP*$_{Ni/NiO}$ [mV_{SHE}]	Stability region
NWC	0	5 to 40	+200	+750	NiO
M-HWC	1000 to 2000	100 to 300	−525	-5 to +20	Ni/NiO
OLNC	150 to 400	20 to 40	−480	+40 to +60	NiO

advantageous effect of low DH levels on SCC initiation is also confirmed in [19,20], where three-point-bend specimens of weld metal Alloy 182 and Alloy 600 were exposed up to 4.5 years (32600 h) to simulated PWR primary water with DH levels varying from 0.45 to 2.23 ppm. During the exposure experiment, periodic visual inspections were performed to examine the ratio of cracked specimens. For the environment with the lowest DH content (0.45 ppm), the crack initiation time is about 50% longer than in the other two environments (1.34 and 2.23 ppm), which suggests that the effect of hydrogen on the initiation of PWSCC is the same for Alloy 600 wrought and Alloy 182 weld metal. In a more recent work conducted by Molander et al. [21], SCC initiation of Alloy 600 was studied in simulated PWR environments with hydrogen contents from 0.45 up to 6.26 ppm. At the high hydrogen content of 6.26 ppm, the time to crack initiation was approximately doubled compared to the initiation time at 2.68 ppm, the resistance to SCC initiation increased significantly at low hydrogen levels. The minimum in initiation time and the maximum in crack growth rate were both located in proximity of the NiO/Ni phase transition line.

So far, the focus of the experimental investigations was placed to SCC crack growth with Alloy 600 under PWR conditions. The Alloy 182 weld metal has never been used for crack initiation tests under BWR conditions before, although it has been proved to be susceptible to IGSCC in crack growth rate tests [15]. Furthermore, the very limited existing data of tests with Alloy 182 conducted under PWR conditions did not reveal a clear trend with regard to DH effects on SCC initiation, because of the inherent large scatter and significant variations in test parameters in the individual experiments. Mechanistic studies of IGSCC in Ni-based alloys have increased in recent years and several mechanisms have been proposed such as slip-dissolution [22], selective internal oxidation [23], hydrogen assisted cracking/hydrogen embrittlement, and vacancy condensation [24], however, the SCC mechanism is still under very controversial discussion. A convincing mechanistic explanation for the effect of DH is still missing. Other studies focused to the effect of grain boundary structure on intergranular SCC and corrosion susceptibility in austenitic stainless steels and Ni-alloys revealed that twin boundaries are highly resistant to SCC cracking and represent a strong barrier for IGSCC [25–28]. Attempts were done to increase the IGSCC resistance by increasing the share of more resistant grain boundaries and to break-up the connectivity of susceptible grain boundaries with resistant ones by specific thermo-mechanical treatments ("grain boundary engineering") [29]. Some studies also showed that the twin boundary structure can be damaged by cold working induced during fabrication and installation of components [30–32] that increases the SCC susceptibility.

The aim of this study is to evaluate the unexplored effect of DH content on the SCC initiation and subsequent short crack growth behaviour in Alloy 182 weld metal under BWR/HWC conditions, and thus to identify optimal operation DH levels to mitigate SCC in BWRs. In BWR/HWC, the DH content at different locations in the reactor strongly varies depending on the H_2 injection rate in the feedwater, reactor design, downcomer γ dose rates, recirculation flow etc. In BWRs with moderate HWC, the DH concentration could be close to the peak SCC susceptibility region (Table 2).

In this work, accelerated SCC initiation and short crack growth tests under simulated BWR environments at different DH levels were performed with sharply notched fracture mechanics specimens. The crack initiation sites, crack path and fracture mode were characterized by optical and scanning electron microscopy (SEM). Grain boundary misorientations and mismatch in Schmid factor were measured by electron backscattered diffraction (EBSD) along intergranular stress corrosion cracks and at crack initiation sites. The present paper summarizes the most important results and conclusions of these investigations. In a related work, we are also investigating SCC initiation with smooth tapered tensile specimens in slow strain rate tests and the results are published in [56].

2. Material and experimental procedure

2.1. Material and specimens

2.1.1. Alloy 182 weld

The Alloy 182 test weld was fabricated according to nuclear welding specifications by filling an U-shaped groove in a large quenched and tempered SA 508 Cl. 2 low-alloy steel plate (1 m × 1 m × 0.22 m). This plate is from the forged lower cylindrical shell of the Biblis C PWR reactor pressure vessel which was never commissioned. In a first step, a weld butter layer (double layer at the groove root, single layer on the groove flanks) was produced by shielded metal arc welding (110–115 A and 22 V, pre-heating temperature 125–140 °C), which was then grinded and post-weld heat treated at 620 °C for 9 h 15 min in air. The groove was then filled by multipass SMAW (115–140 A and 22–24 V, interpass temperature 47–130 °C) without subsequent post-weld heat treatment. The bulk weld metal is thus in the as-welded "solution-annealed" condition, in contrast to the butter layers. The chemical composition of the Alloy 182 weld metal was inspected by inductive coupled plasma optical emission spectroscopy and by combustion analysis by the hot gas extraction infrared absorption method (Table 3). The mechanical tensile test properties of small round tensile specimens (3 mm diameter, 18 mm gauge length) at 25 and 274 °C in air at a strain rate of 10^{-3} s^{-1} are summarized in Table 4.

Fig. 1 shows a cut-out section of the multipass shielded metal arc welding with the main orientation of the plate, welding direction and location of the tensile and fracture mechanics specimens. The typical multipass SMAW weld microstructure on the T-S and L-S surface is shown in Fig. 2. The microstructure with a fibre-like texture consists of large columnar grains of several mm in length that may stretch over

Table 3
Chemical composition of the Alloy 182 weld metal in wt.%.

C	Si	Mn	Cr	Mo	Ni	Nb	Al	Co	Fe	N	Ti
0.027	0.580	6.19	15.9	0.172	69.1	2.36	0.0302	0.0172	5.46	0.024	0.0924