Environmentally-Assisted Cracking in Austenitic Light Water Reactor Structural Materials

Final Report of the KORA-I Project

Hans-Peter Seifert and Stefan Ritter
Nuclear Energy and Safety Research Department
Laboratory for Nuclear Materials

Environmentally-Assisted Cracking
in Austenitic Light Water Reactor Structural Materials

Final Report of the KORA-I Project

Hans-Peter Seifert and Stefan Ritter

Paul Scherrer Institut
5232 Villigen PSI
Switzerland
Tel. +41 (0)56 310 21 11
Fax +41 (0)56 310 21 99
www.psi.ch
# TABLE OF CONTENTS

- ABSTRACT
- EXECUTIVE SUMMARY
- ABBREVIATIONS AND SYMBOLS
- INTRODUCTION
- PART I: CORROSION FATIGUE OF AUSTENITIC STAINLESS STEELS UNDER LIGHT WATER REACTOR CONDITIONS
- PART II: EVALUATION OF THE POTENTIAL AND LIMITS OF THE ELECTROCHEMICAL NOISE MEASUREMENT TECHNIQUE FOR THE EARLY DETECTION OF STRESS CORROSION CRACKING INITIATION IN AUSTENITIC STAINLESS STEEL UNDER BWR/NWC CONDITIONS
- PART III: ENVIRONMENTALLY-ASSISTED CRACKING BEHAVIOUR IN THE TRANSITION REGION OF AN ALLOY 182-SA 508 CL.2 DISSIMILAR METAL WELD JOINT IN SIMULATED BWR/NWC ENVIRONMENT
- ACKNOWLEDGEMENT
The following document is the final report of the KORA-I project, which was performed at the Paul Scherrer Institute (PSI) between 2006 and 2008 and was funded by the Swiss Nuclear Safety Inspectorate (ENSI). The three sub-projects of KORA-I covered the experimental characterisation of the effect of the reactor coolant environment on fatigue initiation and crack growth in austenitic stainless steels under boiling and pressurised water reactor conditions, the experimental evaluation of the potential and limits of the electrochemical noise measurement technique for the early detection of stress corrosion cracking initiation in austenitic stainless steels under boiling water reactor/normal water chemistry conditions, as well as the characterisation of the stress corrosion crack growth behaviour in the fusion line region of an Alloy 182–low-alloy reactor pressure vessel steel dissimilar metal weld. The main scientific results and major conclusions of the three sub-projects are discussed in three independent parts of this report.
EXECUTIVE SUMMARY

**Corrosion fatigue in austenitic stainless steels:** The corrosion fatigue behaviour of different low-carbon and stabilised austenitic stainless steels was characterised under simulated boiling water reactor (BWR) and primary pressurised water reactor (PWR) conditions by cyclic fatigue tests with pre-cracked and notched-only fracture mechanics specimens in the temperature range from 70 to 320 °C. The special emphasis was placed to BWR/hydrogen water chemistry (HWC) and PWR conditions at low corrosion potentials.

Under BWR/HWC and PWR conditions, relevant environmental acceleration of fatigue crack growth (and environmental reduction of fatigue initiation life) occurred in all investigated low-carbon and stabilised stainless steels for the combination of loading frequencies $\leq 0.1$ Hz (strain rates $\leq 0.1 \%$/s), temperatures $\geq 150$ °C ($\geq 100$ °C) and $\Delta K$ values $\geq 3$ MPa$\cdot$m$^{1/2}$ (notch strain amplitudes $\geq 0.3 \%$). If these conjoint threshold conditions were simultaneously satisfied, the environmental enhancement increased with decreasing loading frequency (strain rate) and increasing temperature. Material and water chemistry parameters usually only had a little effect. Sensitisation affected the corrosion fatigue behaviour under highly oxidising BWR/normal water chemistry (NWC) conditions only.

If the critical conjoint requirements were satisfied, the BWR/HWC and PWR environments usually resulted in acceleration of fatigue crack growth by a factor of five to 20 with respect to air and the environmental reduction of the genuine initiation life was even more pronounced. Under comparable loading conditions, the (short and long crack) corrosion fatigue crack growth rates under oxidising BWR/NWC conditions were usually a factor of two to five higher than under BWR/HWC or PWR conditions at low corrosion potentials. The genuine corrosion fatigue initiation life of solution annealed steels, on the other hand, was slightly smaller in BWR/HWC and PWR environments.

The current ASME Boiler and Pressure Vessel Code Section XI fatigue crack growth curves do not include environmental effects and are not conservative in light water reactor environments if the above mentioned conjoint threshold conditions were satisfied. A modification of Section XI should therefore be pursued. The JSME BWR curve conservatively bounds the overwhelming part of corrosion fatigue data in simulated BWR/NWC environment. The Japanese PWR curve proposal reasonably predicts corrosion fatigue crack growth under PWR or BWR/HWC conditions.

The fatigue design curves in Section III of the ASME Boiler and Pressure Vessel Code do not contain excess conservatism to cover environmental effects and are not conservative for some critical parameter combinations. The corrosion fatigue initiation studies confirmed the NUREG/CR-6909 approach of the new US NRC Regulatory Guide 1.207, which defines a procedure to include environmental effects in Section III. Nevertheless, this approach should be critically evaluated by additional tests under more realistic and plant relevant conditions, before it is applied to existing plants.

**Detection of stress corrosion cracking initiation by electrochemical noise measurements:** The electrochemical noise measurement technique is a promising tool for continuous, in-situ corrosion monitoring in technical systems and has the potential to detect nucleation and initiation of localised corrosion processes. The potential and limits of this technique for the detection of SCC initiation in austenitic stainless steels under simulated BWR conditions was investigated by constant extension rate and constant load tests with notched tensile and fracture mechanics specimens in autoclaves. Similar experiments were also performed in aqueous thiosulphate solution at room-temperature.

Interrupted stress corrosion cracking (SCC) initiation experiments and tests with independent on-line crack growth monitoring by the direct current potential drop method clearly indicated that early SCC detection by electrochemical noise is possible under stable and stationary lab conditions.
in oxygenated high-temperature water. Individual small intergranular semi-elliptical surface flaws with a surface crack length and crack depth of about 150 μm may be detected by electrochemical noise measurements. But on the other hand, only crack initiation and the subsequent surface or near surface growth may be detected by electrochemical noise measurements in high-purity water with low conductivity. Furthermore, a short distance between the specimen surface and reference or counter electrode is crucial to achieve a high sensitivity.

Characteristic potential and current transients, indicating SCC initiation, could be observed during experiments in aqueous thiosulphate solution at room-temperature. These transients were comparable to the ones measured in high-temperature water with a short distance of the reference or counter electrode to the specimen surface. The observed polarity of the potential and current signal changes/shifts and shapes of individual transients during SCC initiation suggest a mechanism, which involves film rupture/local anodic dissolution/repassivation according to the slip-dissolution mechanism. Similar electrochemical noise signals were observed for inter- and transgranular SCC in sensitised and solution annealed stainless steels, which indicates that similar electrochemical processes are involved in both cracking mechanisms.

It is recommended to further evaluate this promising technique in long-term tests under static load and to include further important and plant relevant parameters (e.g., dissolved hydrogen content, temperature, flow rate, cold-work), which were not explored so far.

**SCC in Alloy 182 weld metal and dissimilar metal welds:** Recent SCC incidents in the weld metal of bottom head penetration housing and core shroud support welds of BWRs have drawn the attention to the SCC behaviour in the transition region of Alloy 182–low-alloy steel dissimilar weld joints, and in particular, to the issue whether an intergranular/interdendritic SCC crack propagating in the weld metal could easily penetrate through the fusion line into the adjacent low-alloy reactor pressure vessel (RPV) steel. Therefore, the SCC crack growth behaviour in fusion line region of a simulated Alloy 182-SA 508 Cl.2 weld joint was investigated under simulated BWR/NWC conditions.

These investigations confirmed the high SCC susceptibility of Alloy 182 under BWR/NWC conditions and the adequacy of the Swedish disposition curve for this material. The SCC crack growth rates in the Alloy 182 dilution zone were similar to those of the bulk weld metal and tended to slightly decrease with decreasing distance to the fusion line. In high-purity water, cessation of SCC crack growth in the Alloy 182 weld metal was observed at the fusion boundary for $K_I$-values $< 50$ MPa·m$^{1/2}$ and no crack relevantly propagated into the adjacent low-alloy RPV steel, which basically confirms the service experience of such welds, where cracking was usually confined to the Alloy 182 weld metal. In case of increased chloride levels, on the other hand, the SCC cracks very easily crossed the fusion line at $K_I$-values $< 30$ MPa·m$^{1/2}$.

Major crack propagation into the RPV material is thus not expected as long as the number of plant transients is limited and severe and prolonged chloride excursions are avoided (i.e., if water chemistry is kept below the EPRI action level 1 limit). This important conclusion should be verified by further tests at higher $K_I$-values of 50 to 90 MPa·m$^{1/2}$ and with additional dissimilar weld joints representing different welding qualities and low-alloy steel base materials.
ABBREVIATIONS AND SYMBOLS

a  Crack length
A5  Uniform elongation
AISI American Iron and Steel Institute
ANL Argonne National Laboratory, USA
ASME American Society of Mechanical Engineers
ASME BPV ASME Boiler and Pressure Vessel Code
ASTM American Society of Testing and Materials Standards
ASTM E399 Test method for plane-strain fracture toughness of metallic materials
ASTM E647 Standard test method for measurement of fatigue crack growth rates
ASTM E1681 Standard test method for determining threshold stress intensity factor for environment-assisted cracking of metallic materials
BWM Bulk weld metal
BWR Boiling water reactor
BWRVIP Boiling Water Reactor Vessel and Internals Program
CERT Constant extension rate test(ing)
CF Corrosion fatigue
C(T) Compact tension specimen
CUF Cumulated fatigue usage factor
ΔK Stress intensity factor range of fatigue cycle, \( K = K_{I,max} - K_{I,min} \)
ΔtH Hold time at maximum load of fatigue cycle
ΔtD Decline time (down-ramp) of fatigue cycle
ΔtR Rise time (up-ramp) of fatigue cycle
da/dN Crack advance per fatigue cycle
da/dt Time-based crack growth rate = da/dN / ΔtR
DCPD Direct current potential drop crack length measurement method
DH Dissolved hydrogen
DL Disposition line
DLZ Dilution zone
DO Dissolved oxygen
ε Mechanical strain
EAC Environmentally-assisted cracking
ECG-COMON European Cooperative Group on Corrosion Monitoring of Nuclear Materials
ECN Electrochemical current noise
ECP Electrochemical corrosion potential
EN Electrochemical noise
<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>SEM</td>
<td>Scanning electron microscope</td>
</tr>
<tr>
<td>SHE</td>
<td>Standard hydrogen electrode</td>
</tr>
<tr>
<td>SS</td>
<td>Stainless steel</td>
</tr>
<tr>
<td>SSY</td>
<td>Small-scale yielding</td>
</tr>
<tr>
<td>T</td>
<td>Temperature</td>
</tr>
<tr>
<td>TF</td>
<td>Thermal fatigue</td>
</tr>
<tr>
<td>TG</td>
<td>Transgranular</td>
</tr>
<tr>
<td>UF</td>
<td>Fatigue usage factor</td>
</tr>
<tr>
<td>UTS</td>
<td>Ultimate tensile strength</td>
</tr>
<tr>
<td>VTT</td>
<td>Technical Research Centre of Finland, Finland</td>
</tr>
<tr>
<td>YS</td>
<td>Yield strength</td>
</tr>
<tr>
<td>WE</td>
<td>Working electrode</td>
</tr>
<tr>
<td>Z</td>
<td>Reduction of area</td>
</tr>
</tbody>
</table>
INTRODUCTION

The following document is the final report of the KORA-I project, which was performed between January 2006 and December 2008 and funded by the Swiss Nuclear Safety Inspectorate (ENSI). KORA-I consisted on three sub-projects as described in Table 1 and dealt with environmentally-assisted cracking in austenitic stainless steels and Nickel-alloy weldments in light water reactor coolant environments.

Table 1: Topics of the three sub-projects of the KORA-I research programme.

<table>
<thead>
<tr>
<th>Sub-project</th>
<th>Topic</th>
<th>Share</th>
<th>Report</th>
</tr>
</thead>
<tbody>
<tr>
<td>SP I</td>
<td>Environmental effects on fatigue initiation and crack growth in austenitic stainless steels under PWR and BWR/HWC conditions</td>
<td>50 %</td>
<td>Part I</td>
</tr>
<tr>
<td>SP II</td>
<td>Evaluation of the potential and limits of the electro-chemical noise measurement technique for the early detection of SCC initiation in austenitic stainless steels under BWR/NWC conditions</td>
<td>25 %</td>
<td>Part II</td>
</tr>
</tbody>
</table>
| SP III      | - SCC crack growth in Alloy 182 weld metal under BWR and PWR conditions (ICG-EAC-Round Robin)  
- SCC crack growth in an Alloy 182-SA 508 Cl.2 dissimilar metal weld (collaboration with Tohoku University) | 25 %  | Part III|

The scientific results of these sub-projects are summarised in three independent parts of this public report. Each individual part has its own table of contents, summary and reference list at the beginning and end of the part, respectively. The results of the Round Robin programme of the International Co-operative Group on Environmentally-Assisted Cracking of Water Reactor Materials (ICG-EAC) on stress corrosion cracking (SCC) in Alloy 182 weld metal are not included in this report and are made available for ENSI as confidential information only.
Part I

Corrosion fatigue of austenitic stainless steels under light water reactor conditions

Hans-Peter Seifert and Stefan Ritter
# TABLE OF CONTENTS

1.1 INTRODUCTION ........................................................................................................................3
   1.1.1 TECHNICAL BACKGROUND ..........................................................................................3
   1.1.2 LITERATURE SURVEY ..............................................................................................5
   1.1.3. OBJECTIVES OF THE PROJECT TASK ........................................................................6

1.2 MATERIALS AND EXPERIMENTAL PROCEDURE ..................................................................7
   1.2.1 INVESTIGATED MATERIALS ......................................................................................7
   1.2.2 EXPERIMENTAL PROCEDURE ..................................................................................10
      1.2.2.1 Environmental conditions ..................................................................................10
      1.2.2.2 Corrosion fatigue crack growth experiments .......................................................11
      1.2.2.3 Corrosion fatigue initiation experiments ..............................................................12

1.3 RESULTS AND DISCUSSION .................................................................................................14
   1.3.1 CORROSION FATIGUE CRACK GROWTH ..................................................................14
      1.3.1.1 BWR/NWC conditions ......................................................................................14
      1.3.1.2 BWR/HWC or PWR conditions .........................................................................16
      1.3.1.3 NWC vs. HWC and PWR conditions .................................................................21
      1.3.1.4 Fractography ..................................................................................................21
      1.3.1.5 Summary .........................................................................................................22
   1.3.2 CORROSION FATIGUE INITIATION ............................................................................24
      1.3.2.1 Crack initiation and short crack growth behaviour ...........................................24
      1.3.2.2 Effect of loading parameters ............................................................................26
      1.3.2.3 Effect of material parameters ..........................................................................27
      1.3.2.4 Effect of environmental parameters .................................................................29
      1.3.2.5 Comparison of short and long CF crack growth with CF initiation ...............32
      1.3.2.6 Comparison with strain-controlled LCF experiments and the NRC Reg. Guide 1.207 ..32
      1.3.2.7 Fractography ..................................................................................................34
      1.3.2.8 Summary .........................................................................................................35

1.4 PRACTICAL IMPLICATIONS AND CONSEQUENCES ...................................................36
   1.4.1 ADEQUACY AND CONSERVATISM OF FATIGUE DESIGN AND EVALUATION CODES ....36
      1.4.1.1 Fatigue design and evaluation according to ASME III and XI .......................36
      1.4.1.2 Conservatism of fatigue codes with respect to environmental effects ...........38
   1.4.2 CRITICAL SYSTEM CONDITIONS AND COMPONENT LOCATIONS ...........................41

1.5 SUMMARY AND CONCLUSIONS .........................................................................................43

1.6 REFERENCES ...........................................................................................................................45
1.1 Introduction

1.1.1 Technical background

In light water reactors (LWRs), corrosion-resistant, low-carbon (AISI 304L, 316L, 316NG) or Ti- or Nb-stabilised austenitic stainless steel (SS) grades (AISI 321 and 347) and their corresponding weld filler metals (e.g., AISI 308L and 309L) are widely used as a construction material for piping and vessels, which enclose the primary reactor coolant. Cast SS (e.g., ACI CF-3, CF-8 and CF-8M) are used for primary circuit components with complex shape (e.g., pump housings). These primary pressure boundary components (PPBC) are critical systems for plant safety and lifetime. Several cracking incidents occurred in such components during LWR service in the last decades, which could be attributed to stress corrosion cracking (SCC) or to low- and high-cycle fatigue (LCF and HCF) [1]. SCC was much more widespread than fatigue or corrosion fatigue (CF). Although SCC and CF are related phenomena, which are, at least in part, governed by similar processes, SCC of austenitic SS in LWR environments is out of the scope of this work and the reader is referred to references [1-4] for further reading.

PPBC are usually designed against fatigue and very critical component locations are now increasingly equipped with fatigue monitoring systems. It is thus not surprising that only few fatigue cracking incidents in PPBC occurred in recent years, which were mainly related to either HCF through flow-induced vibrations from power up-ratings (e.g., boiling water reactor (BWR) steam dryers, socket welded small diameter instrument lines) or to thermal fatigue (HCF and LCF) caused by complex thermal-hydraulic phenomena (thermal stratification and striping, turbulent mixing, see Figure 1), which were not included in the original design and only partially covered by existing codes. The normal operating cycles were not generally the source of significant fatigue damage in nuclear plants, since they were adequately covered by the original component design. The good service record thus confirms the conservative and adequate character of the applied fatigue design procedures under most operating circumstances. [1]

![Figure 1: Examples of thermal fatigue cracking incidents in LWRs.](image_url)
In contrast to the HCF damage due to high-frequency vibrations, where strain rates are too high for significant environmental effects, the LCF or combined LCF and HCF damage due to thermal loadings may have been aggravated by corrosion effects due to exposure to the reactor coolant. If environment had access to the crack enclave, a differentiation between pure fatigue and CF in case of austenitic SS is very difficult by metallographical and fractographical failure analysis, in particular at low electrochemical corrosion potentials (ECP) (BWR/hydrogen water chemistry (HWC) or pressurised water reactors (PWRs)). Since the thermo-mechanical loading conditions of the components are usually subjected to relevant uncertainties, it is thus difficult to decide, whether such a cracking incident was relevantly aggravated by environmental effects or not.

There is clear experimental evidence that, if certain conjoint threshold conditions were simultaneously satisfied, the fatigue life and fatigue crack growth in austenitic SS may be relevantly reduced and accelerated in BWR/normal water chemistry (NWC), BWR/HWC or PWR environment, respectively (Figure 2). There is no evident reason, why such environmental effects should not occur in the field. On the other hand, the accumulated excellent field experience of fatigue-designed SS primary pressure boundary components does not indicates any generic deficiencies in the current fatigue design procedures by the ignorance of environmental effects. This is related to a relevant part to the large degree of overall conservatism in current fatigue design and evaluation procedures and to the fact that for many plant transients one or several of the above mentioned conjoint threshold conditions are not satisfied and thus resulting in moderate environmental effects only. [1, 5]

![Figure 2: Effect of the high-temperature water environment on fatigue initiation [5] (a) and crack growth (b) in austenitic SS [6].](image)

Although the possibility of environmental effects on fatigue initiation in LWR environments is undisputed, there is no international consensus about the practical consequences. Since more than one decade there is an ongoing controversy and debate on whether environmental effects are significant for fatigue or not and if or how such effects should be considered in Code fatigue design and evaluation procedures.

With regard to the current revision of the ageing management guidelines of the Swiss Federal Nuclear Safety Inspectorate (ENSI), the increased age of the Swiss nuclear power plant fleet (23 to 39 years) and the plans of the Swiss industry to replace the old LWRs by new ones, the clarification of the questions above became an important topic for ENSI. In this context, ENSI sponsored a three year research project at PSI on environmental effects on fatigue in austenitic SS.
1.1.2 Literature survey

At the beginning of this project a detailed literature survey on environmental effects on fatigue initiation and crack growth in austenitic and cast SS in LWR environments was performed in the first year and summarised in a report [1]. This report covers the state-of-the-art in science and technology in this field and also includes a brief introduction on SCC and fatigue in austenitic SS PPBC of LWRs. The major conclusions can be briefly summarised as follows:

There is clear experimental evidence that fatigue life and fatigue crack growth in wrought austenitic and cast SS may be relevantly reduced and accelerated in BWR/NWC, BWR/HWC or PWR environment with respect to air, respectively. The decrease in fatigue life depends on strain rate, dissolved oxygen (DO) level in water, and temperature and is usually more pronounced under PWR and BWR/HWC conditions at low ECPs than under oxidising BWR/NWC conditions. The fatigue life is decreased significantly when three threshold conditions are satisfied simultaneously, i.e., when the applied strain range and service temperature are above a minimum threshold level ($\approx 0.3\%$ and $150\,^\circ\text{C}$), and the loading strain rate is below $0.4\%\cdot\text{s}^{-1}$. Environmental effects are moderate, e.g., less than a factor of two decrease in life, when any one of the threshold conditions is not satisfied. Similarly, relevant environmental acceleration of fatigue crack growth is observed both under BWR/NWC, and BWR/HWC or PWR conditions below a loading frequency of 0.1 to 1 Hz and above temperatures of $150\,^\circ\text{C}$ and certain $\Delta K_{CF}$ thresholds (which are not yet adequately known). The environmental acceleration is comparable under BWR/HWC and PWR conditions and usually a factor of two to five lower than in oxidising BWR/NWC environment and increasing with decreasing loading frequency. Material and environmental parameters usually have little effect on CF crack growth. The mechanism for the environmental effects on fatigue in austenitic and cast SS at low ECPs is unclear and an open issue.

Environmental effects on fatigue initiation are reasonably well established, although relevant and specific knowledge and data gaps still exist (e.g., no data with AISI 321 and 347). Different proposals (e.g., US NRC Regulatory Code 1.207 [7]) now exist for incorporating environmental effects into the fatigue design procedure according to Section III of the ASME Boiler and Pressure Vessel (BPV) Code. The practical application of these procedures is complex and also related to some uncertainties. Furthermore, there are relevant differences between lab tests and component conditions in the field. The adequacy of these procedures and the transferability of the underlying lab results are thus strongly questioned by the industry. Since there is no pressure for immediate actions (as shown by service experience), it is necessary to further evaluate and verify these approaches under more plant relevant conditions, before they should be applied to existing plants.

The CF crack growth behaviour of austenitic and cast SS in BWR/NWC, BWR/HWC and PWR environment is much less investigated and relevant knowledge and data gaps were identified. The fatigue crack growth curves for austenitic SS in Section XI of the ASME BPV Code do not consider environmental effects and are significantly exceeded in BWR/NWC, BWR/HWC and PWR environment for certain critical parameter combinations. A modification should therefore urgently be pursued. Proposed CF crack growth curves for austenitic SS in BWR and in particular in PWR or BWR/HWC environments are based on rather limited data bases or specific materials with limited relevance (e.g., AISI 304). Further CF crack growth investigations in PWR and BWR/HWC environments are therefore urgently required to develop code crack growth curves for flaw tolerance evaluations and safety assessments in austenitic SS components.
1.1.3. Objectives of the project task

Based on the literature survey, the special emphasis of this project task was primarily placed to the quantitative experimental characterisation of the fatigue crack growth in austenitic SS under simulated LWR conditions with a focus to low ECPs (PWR, BWR/HWC). Environmental effects on fatigue initiation and short crack growth were covered by some few experiments and will be evaluated in more detail in a follow-up project. A primary task was the generation of reliable CF crack growth rate data as a function of material, loading and environmental parameters for flaw tolerance evaluations. Other major goals were the identification of the major factors of influence and parameter effects as well as of those critical system conditions and component locations, where strong environmental effects may occur during service. Finally, the conservatism and adequacy of the current ASME XI fatigue crack growth curves and of the Ford & Andresen model [8] shall be assessed and, if necessary, ideas and proposals for improved CF crack growth curves for implementation in codes be worked out.
1.2 Materials and experimental procedure

1.2.1 Investigated materials

Table 1 gives an overview on the type of the investigated austenitic SS. The chemical composition and mechanical tensile test properties of the different steels are shown in Tables 2 and 3. Eight heats of different types of low-carbon and stabilised austenitic SS seamless pipe materials were investigated. These materials are representative for primary piping in Swiss BWRs and PWRs. Additionally, a plate material of the Ti-stabilised SS AISI 321 with a very low austenite stability was included in the test program. Two rod materials of the high-carbon SS AISI 304, which can be easily sensitised and show a sufficiently high susceptibility to intergranular (IG) SCC, completed the test matrix. The materials were usually in the solution annealed condition. A sensitisation heat treatment at 620 °C for 24 h was applied to the high-carbon steel, which resulted in electrochemical potentiokinetic reactivation (EPR) values of 22 to 28 % (see Section 3).

Table 1: Overview on the investigated austenitic SS.

<table>
<thead>
<tr>
<th>AISI</th>
<th>Design.</th>
<th>Type</th>
<th>Product form</th>
<th>Rp0.2 [MPa]</th>
<th>Austenite stability</th>
</tr>
</thead>
<tbody>
<tr>
<td>304L</td>
<td>A</td>
<td>Low-carbon</td>
<td>Piping</td>
<td>262</td>
<td>High</td>
</tr>
<tr>
<td>304L</td>
<td>H</td>
<td>Low-carbon</td>
<td>Piping</td>
<td>298</td>
<td>High?</td>
</tr>
<tr>
<td>316L</td>
<td>B</td>
<td>Low-carbon</td>
<td>Piping</td>
<td>220</td>
<td>High</td>
</tr>
<tr>
<td>316L</td>
<td>G</td>
<td>Low-carbon</td>
<td>Piping</td>
<td>268</td>
<td>High?</td>
</tr>
<tr>
<td>321</td>
<td>C</td>
<td>Ti-stabilised</td>
<td>Piping</td>
<td>291</td>
<td>Medium</td>
</tr>
<tr>
<td>321</td>
<td>E</td>
<td>Ti-stabilised</td>
<td>Piping</td>
<td>198</td>
<td>Medium</td>
</tr>
<tr>
<td>321</td>
<td>F</td>
<td>Ti-stabilised</td>
<td>Plate</td>
<td>255</td>
<td>Low</td>
</tr>
<tr>
<td>347</td>
<td>D</td>
<td>Nb-stabilised</td>
<td>Piping</td>
<td>209</td>
<td>Medium</td>
</tr>
<tr>
<td>347</td>
<td>K</td>
<td>Nb-stabilised</td>
<td>Piping</td>
<td>221</td>
<td>Medium?</td>
</tr>
<tr>
<td>304</td>
<td>304-1 &amp; 2</td>
<td>High-carbon &amp; -sulphur</td>
<td>Rod</td>
<td>291</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 2: Chemical composition in wt.% of the investigated austenitic SS.

<table>
<thead>
<tr>
<th>AISI</th>
<th>Desig.</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Mo</th>
<th>Ni</th>
<th>Co</th>
<th>Cu</th>
<th>N</th>
<th>Nb</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>304L</td>
<td>A</td>
<td>0.029</td>
<td>0.20</td>
<td>1.77</td>
<td>0.020</td>
<td>0.004</td>
<td>19.0</td>
<td>0.122</td>
<td>10.35</td>
<td>0.034</td>
<td>0.186</td>
<td>0.0748</td>
<td>0.002</td>
<td>0.002</td>
</tr>
<tr>
<td>304L</td>
<td>H</td>
<td>0.024</td>
<td>0.35</td>
<td>1.49</td>
<td>0.026</td>
<td>0.005</td>
<td>17.9</td>
<td>0.247</td>
<td>10.00</td>
<td>0.088</td>
<td>0.305</td>
<td>0.0590</td>
<td>0.001</td>
<td>0.001</td>
</tr>
<tr>
<td>316L</td>
<td>B</td>
<td>0.021</td>
<td>0.26</td>
<td>1.69</td>
<td>0.033</td>
<td>0.003</td>
<td>17.5</td>
<td>2.150</td>
<td>11.14</td>
<td>0.093</td>
<td>0.273</td>
<td>0.0601</td>
<td>0.012</td>
<td>0.003</td>
</tr>
<tr>
<td>316L</td>
<td>G</td>
<td>0.021</td>
<td>0.20</td>
<td>1.75</td>
<td>0.002</td>
<td>0.014</td>
<td>17.7</td>
<td>2.488</td>
<td>13.20</td>
<td>0.291</td>
<td>0.124</td>
<td>0.0930</td>
<td>0.001</td>
<td>0.000</td>
</tr>
<tr>
<td>321</td>
<td>C</td>
<td>0.060</td>
<td>0.18</td>
<td>1.88</td>
<td>0.030</td>
<td>0.004</td>
<td>17.9</td>
<td>0.234</td>
<td>10.13</td>
<td>0.109</td>
<td>0.223</td>
<td>0.0107</td>
<td>0.015</td>
<td>0.431</td>
</tr>
<tr>
<td>321</td>
<td>E</td>
<td>0.073</td>
<td>0.20</td>
<td>1.87</td>
<td>0.021</td>
<td>0.004</td>
<td>18.0</td>
<td>0.334</td>
<td>10.20</td>
<td>0.055</td>
<td>0.182</td>
<td>0.0098</td>
<td>0.001</td>
<td>0.517</td>
</tr>
<tr>
<td>321</td>
<td>F</td>
<td>0.050</td>
<td>0.18</td>
<td>1.27</td>
<td>0.021</td>
<td>0.003</td>
<td>17.4</td>
<td>0.209</td>
<td>9.31</td>
<td>0.121</td>
<td>0.227</td>
<td>0.0120</td>
<td>0.000</td>
<td>0.430</td>
</tr>
<tr>
<td>347</td>
<td>D</td>
<td>0.055</td>
<td>0.27</td>
<td>1.74</td>
<td>0.023</td>
<td>0.006</td>
<td>17.9</td>
<td>0.351</td>
<td>10.48</td>
<td>0.102</td>
<td>0.225</td>
<td>0.0266</td>
<td>0.717</td>
<td>0.003</td>
</tr>
<tr>
<td>347</td>
<td>K</td>
<td>0.058</td>
<td>0.20</td>
<td>1.72</td>
<td>0.025</td>
<td>0.006</td>
<td>17.4</td>
<td>0.441</td>
<td>10.40</td>
<td>0.077</td>
<td>0.268</td>
<td>0.0450</td>
<td>0.571</td>
<td>0.001</td>
</tr>
<tr>
<td>304</td>
<td>304-1 &amp; 2</td>
<td>0.062</td>
<td>0.20</td>
<td>0.55</td>
<td>0.026</td>
<td>0.025</td>
<td>18.7</td>
<td>0.298</td>
<td>8.79</td>
<td>0.185</td>
<td>0.298</td>
<td>0.019</td>
<td>0.001</td>
<td></td>
</tr>
</tbody>
</table>
Table 3: Mechanical tensile test properties of the investigated austenitic SS at 25 °C.

<table>
<thead>
<tr>
<th>Material</th>
<th>Desig.</th>
<th>Rp0.2 [MPa]</th>
<th>Rm [MPa]</th>
<th>Ag [%]</th>
<th>A [%]</th>
<th>Z [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>304L</td>
<td>A</td>
<td>262 ± 3</td>
<td>576 ± 5</td>
<td>46.0 ± 1.5</td>
<td>62.0 ± 2.5</td>
<td>84 ± 1</td>
</tr>
<tr>
<td>304L</td>
<td>H</td>
<td>202 ± 2</td>
<td>540 ± 2</td>
<td>43.0 ± 0.5</td>
<td>58.0 ± 0.5</td>
<td>83 ± 1</td>
</tr>
<tr>
<td>316L</td>
<td>B</td>
<td>220 ± 5</td>
<td>523 ± 1</td>
<td>48.5 ± 2.5</td>
<td>63.5 ± 2.0</td>
<td>83 ± 1</td>
</tr>
<tr>
<td>321</td>
<td>C</td>
<td>281 ± 37</td>
<td>555 ± 5</td>
<td>46.5 ± 1.5</td>
<td>59.0 ± 2.0</td>
<td>78 ± 1</td>
</tr>
<tr>
<td>321</td>
<td>E</td>
<td>193 ± 4</td>
<td>542 ± 3</td>
<td>47.0 ± 0.5</td>
<td>60.5 ± 1.0</td>
<td>-</td>
</tr>
<tr>
<td>347</td>
<td>D</td>
<td>209 ± 3</td>
<td>556 ± 3</td>
<td>44.5 ± 1.0</td>
<td>57.5 ± 1.0</td>
<td>76 ± 1</td>
</tr>
<tr>
<td>347</td>
<td>K</td>
<td>221 ± 2</td>
<td>570 ± 1</td>
<td>38.0 ± 0.5</td>
<td>51.0 ± 0.1</td>
<td>75 ± 4</td>
</tr>
</tbody>
</table>

The cyclic plastic behaviour of many of these steels in isothermal LCF tests between 25 and 340 °C as well as in- and out-of-phase thermo-mechanical fatigue experiments between 100 and 340 °C in air is described in [9]. In spite of significantly different initial hardening, softening and secondary hardening behaviour, the cycle number to crack initiation N_i (“fatigue life”) was very similar for all materials for given strain amplitudes over a wide range of conditions and only slightly decreased with temperature from room-temperature to 340 °C. Distinct secondary hardening was increasingly observed with decreasing strain amplitudes at high temperatures > 200 °C due to dynamic strain ageing and at low temperatures below 100 to 150 °C in steels with low austenite stability due to strain-induced martensite formation. It is believed that the secondary hardening will relevantly affect the fatigue life in air and high-temperature water at small strain amplitudes.

The cyclic stress-strain curves at N_i/2 in air at 35 °C for selected materials are shown in Figure 3. In the saturation region the cyclic stress-strain curves of the solution annealed piping materials were very similar. A temperature increase from 35 to 340 °C resulted in small change of about 20 % only.

![Cyclic stress-strain curves](image)

**Figure 3:** Cyclic stress-strain curves for selected austenitic SS piping materials at room-temperature in air.
Figure 4: Representative micrographs of the investigated austenitic SS piping materials.
Typical microstructures of the different piping materials in the as-received solution annealed condition are shown in Figure 4. The Ti- and in particular the Nb-stabilised grades had a relevantly finer average austenite grain size than the low carbon grades. The mean grain size varied between 75 μm for material B (316L) and 9 μm for material K (347).

In spite of the relevant variations in different material properties like cyclic plastic behaviour, yield strength (200 to 300 MPa), mean grain size (9 to 75 μm), delta-ferrite content, twin boundary fraction, austenite stability, carbides, homogeneity and anisotropy, these parameters usually had little effect on the fatigue and CF behaviour over a wide range of conditions. Secondary hardening might relevantly affect CF close to thresholds and at small strain amplitudes. Sensitisation may affect the CF behaviour under oxidising BWR/NWC conditions or in case of high anion (sulphate, chloride) contents at low ECPs in PWR or BWR/HWC environments, although the effects are usually less severe than for SCC under static load.

1.2.2 Experimental procedure

1.2.2.1 Environmental conditions

The EAC tests were performed in SS autoclaves with electromechanical loading systems, which were attached to refreshing high-temperature water loops (Figure 5). During the experiments all important mechanical (load, pull rod stroke) and environmental parameters at inlet and outlet (DO, dissolved hydrogen (DH), κ, pH, T, p, flow) were recorded continuously. The ECP of the specimens and the redox potential (platinum probe) were continuously monitored with Cu/Cu2O/ZrO2-membrane electrodes. The crack advance was monitored using the reversed direct current potential drop (DCPD) method with a resolution limit of ca. 1 μm. After the tests all specimens were broken open by fatigue for fractographical analysis in the scanning electron microscope (SEM).

Figure 5: Schematic of the used high-temperature water loops with electro-mechanical tensile machine.
Most experiments were performed at low ECPs under BWR/HWC or PWR conditions. Tests under oxidising BWR/NWC conditions were included in the test matrix for comparison reasons. BWR conditions were mostly simulated with high-purity, hydrogenated (HWC) or oxygenated (NWC) water at a temperature of 288 °C. For HWC conditions, a DH content of 150 ppb was usually applied resulting in a redox potential and an ECP of -530 and -550 to -600 mV_{SHE}, respectively. For NWC conditions, usually a DO content of 0.4 ppm was applied (+120 mV_{SHE}). In some cases, mixtures of DO and DH were applied to check results. Additionally, tests under HWC conditions at 150 ppb DH were performed at a temperature of 150 and 220 °C, resulting in an ECP of -430 and -460 mV_{SHE}, respectively. PWR tests in neutral high-purity water were performed with 2.2 ppm DH at temperatures between 70 and 320 °C. The pH_T and ECP at temperature in high-purity water varied from 6.4 (70 °C) to 5.7 (320 °C) and from -410 (70 °C) to -630 mV_{SHE} (320 °C), respectively. Corresponding PWR experiments with H_2BO_3 and LiOH (usually 800 to 1000 ppm B and 2.5 to 3 ppm Li) were performed with a DH of 2.2 ppm at temperatures of 150 and 288 °C. The pH_T and ECP at temperature in water with B and Li varied from 6.4 (150 °C) to 7.1 (288 °C) and from -530 (150 °C) to -750 mV_{SHE} (288 °C), respectively. All tests were performed under low-flow conditions (4 to 9 autoclave exchanges per h) with a local flow rate of some few mm/s. In some cases chloride or sulphate was added to reduce crack initiation time or to accelerate crack growth.

1.2.2.2 Corrosion fatigue crack growth experiments

In case of the CF crack growth investigations, air fatigue pre-cracked 12.5 mm thick compact tension (0.5 T C(T)) specimens with 5 % side-grooves according to ASTM E399 were used. The specimens from the piping and plate materials were manufactured in the C-L or L-C and in T-L or L-T orientation, respectively. The specimens were pre-cracked by fatigue in air at room-temperature to a final pre-crack length with a specimen ligament ratio a_0/W of 0.4, using a load ratio R of 0.1. The maximal K_I at the final load step was ≤ 18 MPa·m^{1/2}.

Before the cyclic loading was applied, the specimens were pre-oxidised in the test environment at a small constant pre-load for at least one week. The subsequent cyclic loading in CF tests was performed under load control. Constant load amplitude loading with a saw tooth waveform (constant loading rate) was applied. At loading frequencies ν ≥ 2.5·10^{-3} Hz a symmetric saw tooth waveform was used (rise time Δt_R = decline time Δt_D). To save testing time, below this frequency a positive saw tooth waveform (slow loading, faster unloading with a decline time of 200 s in most cases) was applied. The frequency, ΔK level and load ratio R were varied over a very broad range form 3·10^{-6} to 2.5·10^{-2} Hz, 2 to 35 MPa·m^{1/2} and 0.05 to 0.9. With some few exceptions the K_{I,max} values were well below the ASTM E647 limit.

In most experiments, either the loading frequency was stepwise decreased at a constant load ratio R (and re-increased in some cases) or the load ratio R was stepwise increased at a constant maximum load P_{max} and constant loading rate dP/dt. In the load ratio increase tests, the step size was decreased with increasing load ratio.

For the different experimental phases, the ΔK and K_I values were calculated according to ASTM E399 by the measured load and the actual mean crack length derived by the DCPD and corrected by post-test fractography. In the plots and tables, the corresponding ΔK and K_I values at the beginning of the different experimental phases are usually given. In spite of the constant load amplitude loading, these values only slightly changed during such an experimental phase because of the moderate crack increments and the small crack length to specimen ligament ratio a/W of about 0.4. With very few exceptions (close to thresholds or in case of pinning phenomena), the crack growth rates in the individual experimental phases were constant. Cycle- and time-based crack growth rates da/dN and da/dt = (da/dN)/Δt_R were calculated for the different experimental phases by linear regression fits of the corrected crack length vs. time plots and were compared with the corresponding fatigue crack growth rates in air at the same temperature under otherwise identical conditions.
1.2.2.3 Corrosion fatigue initiation experiments

The CF initiation behaviour was evaluated with load controlled experiments with sharply notched 0.5 T C(T) specimens with 5% side-grooves according to ASTM E399. The specimens were fabricated from the piping materials and manufactured in the C-L or L-C direction.

The applied approach in this work is fundamentally different from those in other comparable literature investigations (e.g., in NUREG/CR-6909 [5]), which were based on strain-controlled LCF tests with smooth cylindrical specimens. In the case of the sharply notched C(T) specimens there is a multiaxial stress state (plane stress at surface of the notch-root, plane strain in the region of the peak stress) with a stress concentration $\sigma_{\text{max}}/\sigma_{\text{nom}}$ and stress gradient $d\sigma/dr$ in the notch-root region. The plasticity is limited to the direct vicinity of the notch-root. Crack initiation and short crack growth can be easily followed by the DCPD technique in this case, but a notch strain analysis is always required. In the strain-controlled LCF experiments, there is a uniaxial homogeneous stress state and a fully plastic specimen ligament. This kind of test simulates the conditions at the notch-root of components with notches. Only technical crack initiation can usually be followed by the load drop method, but on the other hand, these experiments provide detailed information on the cyclic plastic behaviour of the material.

Constant load amplitude CF initiation tests with symmetric and asymmetric (for rise times $\Delta t_R \geq 500$ s) saw tooth waveform were performed under different environmental, loading and material conditions with notched C(T) specimens at a load ratio $R$ of 0.05. The C(T) specimens had a notch radius $\rho$ of 0.1, 0.3 or 0.5 mm and a notch depth $a_0$ from the load line of 7.5, 10 and 12.5 mm ($a_0/W = 0.3, 0.4$ and 0.5), respectively.

This kind of loading resulted in a stress-strain hysteresis loop at the notch ground with a stress ratio $R_{\sigma}$ of about -1. A total elastic stress range $\Delta\sigma_{\text{LEFM}}$ at the notch-root of the C(T) specimen was estimated from the stress intensity factor range $\Delta K_I$ and the notch radius $\rho$ according to equations (1) and (2) by linear elastic fracture mechanics (LEFM) [10-12]. Together with the nominal stress range $\Delta S_n$ in the notched C(T) specimen (equation (3)), a theoretical elastic notch stress concentration factor $K_t$ can be derived according to equation (4).

\[
\Delta\sigma_{\text{LEFM}} = 2 \cdot S_{n,\text{LEFM}} = \frac{2}{\sqrt{\pi}} \cdot \frac{\Delta K_I}{\sqrt{\rho}} \quad (1)
\]

\[
\Delta K_I = \frac{\Delta P}{B \cdot \sqrt{W}} \cdot f\left(\frac{a}{W}\right) \quad \text{according to ASTM E399} \quad (2)
\]

\[
\Delta S_n = \frac{\Delta P}{B} \left(\frac{3\cdot(W+a)}{(W-a)^2} + \frac{1}{W-a}\right) \quad (3)
\]

\[
K_t = \frac{\Delta\sigma_{\text{LEFM}}}{\Delta S_n} \quad (4)
\]

Additionally, notch-root strain amplitudes $\varepsilon_{a,N}$ were estimated numerically according to the concepts of Neuber [13] (equation (5)) and Glinka [14, 15] (equation (6)) with $K_t$ and the experimentally-derived uniaxial cyclic stress-strain-curves of the materials (Figure 3), which were modified for plane strain conditions by simple plastic theory [14]. For the used notch configurations, the method of Glinka usually resulted in 15 to 20 % smaller notch strains only.

\[
\text{Neuber: } K_t = \sqrt{K_\sigma \cdot K_\varepsilon} \quad (5)
\]

with $K_\sigma = \frac{\sigma}{S_n}$, $K_\varepsilon = \frac{\varepsilon}{\varepsilon_n}$ and $\varepsilon_n = \frac{S_n}{E}$.
The elastic stress concentration factor $K_t$ of the different notches ($\rho = 0.1, 0.3$ and $0.5 \text{ mm}$, $a_0 = 7.5, 10$ or $12.5 \text{ mm}$) varied from $7.7$ to $21.2$. The elastic stress amplitude $S_{a,LEFM}$ ranged from $425$ to $1165 \text{ MPa}$, which corresponded to notch strain amplitudes $\varepsilon_{a,N}$ of $0.27$ to $0.97 \%$ (Glinka) and $0.32$ to $1.19 \%$ (Neuber). The rise time $\Delta t_R$ was varied from $25$ to $5000 \text{ s}$ covering a notch strain rate from $2.4 \cdot 10^{-2}$ to $2.4 \cdot 10^{-4} \%/s$.

The point of crack initiation and the subsequent short crack growth were determined by the DCPD technique. In contrast to notched tensile specimens in slow strain rate tests, the effect of the geometry change by plastic deformation on the DCPD potential drop in the sharply notched C(T) specimens under small scale yielding conditions in fatigue experiments is very moderate. The point of crack initiation derived by the DCPD in the later case corresponds to a very small average crack advance over the specimen thickness of about $5$ to $10 \text{ m}$. By using the estimated local strain amplitude $\varepsilon_{a,N}$ or elastic stress amplitude $S_{a,LEFM}$ at the notch ground, the CF initiation behaviour in load-controlled experiments with notched C(T) specimens in high-temperature water could be directly compared to the ASME III fatigue mean curve of austenitic SS, which is based on strain-controlled LCF tests with smooth specimens at room-temperature in air. In case of the elastic stress amplitude $S_{a,LEFM}$, the strain amplitude $\varepsilon_s$ at the ASME III air mean curve was multiplied by the Young modulus $E$. In control experiments in air at room-temperature with notched C(T) specimens, the local notch strain $\varepsilon_{a,N}$ estimations resulted in shorter predicted crack initiation lives than experimentally observed, whereas very good correlations were observed by using the local elastic stress amplitude $S_{a,LEFM}$ (Figure 6). Therefore, the later approach is used in the following Sections only.

![Figure 6](image_url): Comparison of fatigue initiation life in sharply notched C(T) specimen in a load-controlled test at room-temperature in air with the air mean curve from strain-controlled LCF experiments based on different notch strain concepts. The best correlation is observed with the LEFM approach.
1.3 Results and discussion

1.3.1 Corrosion fatigue crack growth

1.3.1.1 BWR/NWC conditions

Based on the relevant amount of literature data under BWR/NWC conditions, only a limited number of experiments were performed in this environment for comparison reasons. In the tested ΔK and load ratio range of 5 to 20 MPa-m^{1/2} and 0.5 to 0.8, relevant environmental acceleration of fatigue crack growth was observed in all investigated solution annealed or sensitised SS at low loading frequencies ≤ 0.1 Hz under BWR/NWC conditions (Figure 7a). Below this frequency, the crack advance per fatigue cycle da/dN was increasing with decreasing loading frequency down to the lowest loading frequency tested (3·10⁻⁶ Hz).

The IG CF crack growth rates in the sensitised SS were usually only slightly higher than the transgranular (TG) propagation rates in the corresponding heats in the solution annealed condition (Figure 7a). An increasing difference in CF crack growth rates between sensitised and solution annealed SS appeared at very low loading frequencies or under periodical partial unloading conditions, when SCC contributions became significant in the sensitised steels (Figure 8). Similarly, chloride increased the CF growth rates only under these loading conditions, otherwise they were comparable to those in high-purity water. [16]

![Figure 7](image)

**Figure 7:** Effect of the loading frequency on the cycle-based CF crack growth rate in a sensitised or solution annealed 304 SS under BWR/NWC conditions (a) and in a solution annealed 304L SS under BWR/HWC conditions (b).

Under BWR/NWC conditions, the current ASME XI fatigue crack growth curves for these materials were increasingly exceeded with decreasing loading frequencies below 0.1 Hz for all investigated SS (Figure 9a). These curves are therefore not conservative under the corresponding loading and environmental conditions. The CF curves for SS and BWR conditions in the corresponding Japanese JSME Code [17, 18] conservatively cover the observed CF crack growth rates under BWR/NWC conditions (Figure 9).

The extension of the slip-dissolution model, originally developed by Ford & Andresen for SCC in SS, to cyclic loading conditions [8] realistically describes the CF crack growth in SS over a wide range of conditions (Figure 10). This indicates that this cracking mechanism might also be active for CF crack growth in SS at high ECPs.
Figure 8: Effect of sensitisation on time-based CF crack growth rate under BWR/NWC and HWC conditions (PPU = periodical partial unloading, trapezoidal waveform).

Figure 9: Comparison of CF crack growth rates under BWR/NWC conditions with ASME XI and JSME BWR Code [17, 18] (a) as well as under BWR/HWC conditions with ASME XI and Japanese PWR Code proposal [19] (b).

Figure 10: Comparison of CF crack growth under simulated BWR/NWC conditions [1] with predictions of the slip-dissolution model of Ford & Andresen [8].
1.3.1.2 BWR/HWC or PWR conditions

In contrast to BWR/NWC at high ECP, the relevant acceleration of CF crack growth in SS under BWR/HWC or PWR conditions at low ECP is rather surprising at a first glance. This behaviour is also quite different from that of carbon and low-alloy steels [20].

Effect of loading parameters: Similarly to NWC conditions, relevant environmental acceleration of fatigue crack growth was observed in all investigated solution annealed or sensitised SS at low loading frequencies \( \leq 0.1 \text{ Hz} \) under BWR/HWC (Figure 7b) or PWR conditions up to high load ratios \( R < 0.9 \) and small \( \Delta K \geq 3 \text{ MPa-m}^{1/2} \). Below this frequency, the crack advance per fatigue cycle \( \frac{da}{dN} \) increased with decreasing loading frequency down to the lowest loading frequency tested \((3 \times 10^{-6} \text{ Hz})\). Below 0.1 Hz a power law relationship between the crack advance per fatigue cycle \( \frac{da}{dN} \) and the loading frequency \( \nu \) (\( \frac{da}{dN} \propto \nu^{-n} \)) with an exponent \( n \) in the range of 0.25 to 0.4 was observed (about 0.28 at 288 °C). In preliminary load shedding experiments (Figure 11) a disappearance of the environmental acceleration was observed at high load ratios \( R \geq 0.9 \) and small \( \Delta K < 3 \text{ MPa-m}^{1/2} \). Further tests with smaller load steps are necessary to determine whether these values represent a true CF threshold behaviour.

![Figure 11: Apparent \( \Delta K_{CF} \) threshold of about 3 MPa-m\(^{1/2}\) in load shedding experiments with increasing load ratio in HWC environment (a: 304L, b: 347).](image-url)

Apart from the threshold region, \( \Delta K \), loading frequency \( \nu \) and load ratio \( R \) effects were covered by a single curve in the time-domain (Figure 12). Although load ratio \( R \) and \( \Delta K \) effects were not systematically studied, the behaviour in the time-domain allowed to derive certain conclusions about these effects (Figure 13). Based on this analysis, the load ratio \( R \) only has a moderate effect on CF crack growth, but this might be different close to CF thresholds or in case of possible SCC contributions (for cold-worked SS). Above the CF thresholds \( \Delta K_{CF} \), the cycle-based CF growth rates \( \frac{da}{dN} \) increase with increasing \( \Delta K \) at intermediate \( \Delta K \) levels according to a power law (\( \frac{da}{dN} \propto \Delta K^m \)). In this region, the effect of \( \Delta K \) on CF crack growth was smaller than on fatigue
crack growth in air (power law exponent m of ≈ 2.5 in water and 3.3 in air). The environmental acceleration of fatigue crack growth thereby increased with decreasing ΔK and frequency ω and increasing temperature (Figures 7b, 12, 13, 17b).

Figure 12: Time-domain analysis of CF for different load ratios (a) and ΔK values (b).

Figure 13: Effect of ΔK, load ratio (a) and loading frequency (b) on cycle-based CF crack growth rates. The curves were derived by the regression curves from Figure 12.

Effect of material parameters: Within the investigated parameter range under HWC or PWR conditions, no significant effect of the material conditions on CF crack growth was observed (Figures 8 and 14). Neither the heat treatment (sensitised vs. solution annealed, Figure 8) nor the carbon content (304 vs. 304L), steel grade (304L vs. 316L, stabilised vs. non-stabilised), grain size, yield strength nor the product form (piping, rod, plate) had a relevant effect. It is expected that larger differences between materials may primarily appear at low loading frequencies and close to stress intensity thresholds like ΔK_{CF} (R, ω). In another study from literature [21], cold-work had also very little effect on CF crack growth. This is not unexpected, since a cyclic hardened plastic zone is also formed at the crack-tip in solution annealed steels under cyclic load. It is stressed that at very low loading frequencies, high load ratios or under periodical partial unloading conditions, cold-work might result in higher growth rates because of possible SCC contributions in cold-worked SS.
Effect of environmental parameters: Similarly, apart from temperature environmental parameters had no significant effect on CF crack growth under BWR/HWC and PWR conditions (Figures 15 and 16). The CF rates were very similar over the whole ECP range, which is of relevance for BWR/HWC (-500 to -200 mV_{SHE}) and PWR (-800 to -600 mV_{SHE}) conditions. Within the investigated parameter and ECP range, neither the DH, sulphate and chloride concentration nor the B/Li content [21, 22] had a relevant effect. On the other hand, a clear effect of temperature on CF crack growth was observed (Figures 17 and 18). Relevant acceleration of fatigue crack growth occurred above a temperature of 100 to 150 °C. This threshold seemed to be shifted to lower temperatures with decreasing loading frequency. In high-purity water (pH_{T} = 5.7), an Arrhenius activation energy of 20 kJ/mol was observed, which is about 30 % lower than in similar investigations in lithiated water by Tice [21]. This difference might be related to the different pH at temperature (5.7 vs. 7.8 at 300 °C). The CF rates were increasing by a factor of 5 to 10 with increasing temperature from 150 to 320 °C. Under comparable loading conditions in air, the fatigue growth rates increased only by a factor of 1.4. Furthermore, pre-oxidation temperature and period had no relevant effect on CF crack growth rates at other temperatures (Figure 19). Long-term pre-oxidation at 288 °C, e.g., resulted in similar CF crack growth rates, when changing to 150 °C as in experiments, where specimens were pre-oxidised at 150 °C.

Figure 14: Small effect of material parameters on CF in BWR/HWC environment.

Figure 15: Comparable CF crack growth rates over the full ECP range of HWC.
**Figure 16:** Small effect of environmental parameters on CF crack growth in PWR or BWR/HWC environment (including literature data from [21-23]).

**Figure 17:** Effect of temperature and loading frequency on CF crack growth rates (a) and on environmental acceleration (b) of fatigue crack growth in simulated PWR environment.

**Figure 18:** Effect of temperature on CF crack growth (Arrhenius plot).
Comparison with different crack growth and code curves: In the time-domain, the observed CF crack growth behaviour under BWR/HWC and PWR conditions are reasonably well predicted by the Bettis PWR [22] and Japanese PWR curves [19] or the moderate DO (200 ppb) curve developed by Argonne National Laboratory (ANL) for sensitised SS in oxygenated high-temperature water [24] (Figures 14 to 16 and 20).

Figure 19: Example of CF crack growth in an experiment with temperature changes (a) and comparison of the resulting crack growth rates with those of purely isothermal tests (b).

Comparison with the Ford & Andresen model: In contrast to high ECP, the slip-dissolution model developed by Ford & Andresen [8] significantly underestimates the CF crack growth at low ECP under BWR/HWC or PWR conditions in its present form (Figure 20). This indicates that...
another mechanism might be operative under these conditions. The observed material and environmental parameter trends let us suppose that CF under these conditions might be somehow related to the higher general corrosion rate and less protective oxide film in these materials at low ECP [25].

1.3.1.3 NWC vs. HWC and PWR conditions

Under comparable loading conditions, the CF crack growth rates under oxidising BWR/NWC conditions at \( \nu > 5 \times 10^{-4} \) Hz were a factor of 2 to 5 higher than under BWR/HWC conditions at low ECPs (Figures 8 and 21). At very low loading frequencies this difference may be higher because of possible SCC contributions, in particular in sensitised SS (Figure 8). The mitigation effect of HWC or noble metal chemical addition is thus less pronounced on CF than on SCC crack growth, but still remarkable. In contrast to genuine LCF initiation from smooth surfaces, where environmental reduction of fatigue life is more pronounced at low ECPs, the CF crack growth rates of short and long cracks are higher at high ECPs, although the extent of environmental acceleration of fatigue crack growth is still very remarkable at low ECPs (see Section 1.3.3).

![Figure 21: Comparison of time-based CF crack growth rates da/dt\(_{H2O}\) under BWR/NWC and HWC conditions in a time-domain plot.](image)

1.3.1.4 Fractography

The metallo- and fractographical observations of the CF crack growth tests at PSI can be summarised as follows:

A TG CF crack path is usually observed under BWR/HWC or PWR conditions or at high loading frequencies \( \nu \) under BWR/NWC conditions (Figure 22). An IG (or mixed-mode) CF crack path is observed under oxidising BWR/NWC conditions, in particular in sensitised materials, but also in solution annealed materials at low loading frequencies or high load ratios. IG cracking under BWR/HWC or PWR conditions is possible in some specific cases, e.g., in case of a preceding NWC phase or of cold-work and low loading frequencies. Isolated IG islands may be observed in case of BWR/HWC or PWR conditions and very low loading frequencies in some materials. [1]

Fatigue striations perpendicular to the local crack growth direction were usually observed both on the IG and TG fracture surfaces. The individual TG facets were rather smooth indicating growth along specific crystallographic planes. With decreasing loading frequency and increasing environmental acceleration the surface roughness became higher (and contained more steps and tear ridges) and the crack front more uneven because of local crack pinning. The striations were perpendicular to the local crack growth direction. The local crack growth direction was increasingly
deviating from the macroscopic propagation direction with increasing environmental acceleration of fatigue crack growth. At very low loading frequencies or high load ratios, the fracture surface may become striation-free. General corrosion of the fracture surface during prolonged exposure periods in high-temperature water or crack closure/fretting can also destroy original striations. Furthermore, it is stressed that the TG fracture surface appearance can relevantly differ from material to material even in case of similar loading conditions and comparable environmental acceleration of fatigue crack growth. This may be explained by different dominating plastic deformation mechanisms (dislocation structure, planar vs. wavy slip, twinning, martensite formation, dynamic strain ageing, etc.) which can relevantly change with temperature, strain amplitude and rate, microstructure and chemical composition. [1]

Apart from a larger spacing of the striations and a significantly higher surface roughness, the fracture surface appearance under BWR/HWC and PWR conditions was not that fundamentally different from that of fatigue in air under similar loading conditions. Since local loading conditions (e.g., for acting $\Delta K$ or strain rates) of field components are subjected to relevant uncertainty in some cases, this has practical implications for failure analysis, in particular on the decision whether environmental effects may have played a relevant role in a specific cracking incidents or not. Similarly, striation-free IG CF at very low loading frequencies under oxidising BWR/NWC conditions is difficult to differentiate from IG SCC.

![Typical TG CF fracture surface with fatigue striations from a fatigue test under BWR/HWC conditions. High similarity to fatigue in air.](image)

**Figure 22:** Typical TG CF fracture surface with fatigue striations from a fatigue test under BWR/HWC conditions. High similarity to fatigue in air.

### 1.3.1.5 Summary

- Relevant environmental acceleration of fatigue crack growth was observed at loading frequencies $\leq 0.1$ Hz and could be sustained down to the lowest investigated loading frequencies of $3 \cdot 10^{-6}$ Hz in BWR/NWC, BWR/HWC and PWR environment.

- In BWR/HWC and PWR conditions, relevant environmental acceleration of fatigue crack growth occurred in all investigated low-carbon and stabilised SS for the combination of loading frequencies $\leq 0.1$ Hz, temperatures $\geq 150$ °C and $\Delta K$ values $\geq 3$ MPa·m$^{1/2}$. If these conjoint threshold conditions were simultaneously satisfied, the environmental enhancement increased with decreasing loading frequency and $\Delta K$ and increasing temperature. Otherwise environmental effects were absent or moderate.

- Both, under BWR/NWC and under PWR or BWR/HWC conditions, there is usually little effect of material and environmental parameters. The environmental acceleration is comparable under HWC and PWR conditions and usually a factor of 2 to 5 lower than in oxidising BWR/NWC environment.
• The current ASME BPV Code Section XI fatigue curves are increasingly exceeded with decreasing loading frequency below 0.1 Hz for BWR/NWC, BWR/HWC and PWR conditions and therefore not conservative. A modification of these curves should therefore be pursued!

• The JSME BWR curve conservatively bounds the overwhelming part of CF data in simulated BWR/NWC environment. The Japanese PWR curve proposal reasonably predicts CF crack growth under PWR or BWR/HWC conditions.

• From an engineering point of view, the CF crack growth in the time-domain in different austenitic SS can be described with reasonable accuracy by a single curve for BWR/NWC or PWR and BWR/HWC conditions, respectively. The time-domain analysis is therefore the best way to develop modified fatigue crack growth curves in case of possible environmental effects.

• The slip-dissolution model developed by Ford & Andresen reasonably predicts CF crack growth under BWR/NWC conditions, but significantly underestimates it under BWR/HWC or PWR conditions. This indicates a different mechanism for CF at low ECPs. The environmental enhancement of fatigue crack growth at low ECPs might be related to the increased general corrosion rate and less protective oxide film at low ECPs.
1.3.2 Corrosion fatigue initiation

The CF initiation behaviour was evaluated with load-controlled experiments with sharply notched 0.5 T C(T) specimens with 5% side-grooves. The overwhelming part of experiments was performed under BWR/HWC or PWR conditions.

1.3.2.1 Crack initiation and short crack growth behaviour

Figure 23 exemplarily shows the evolution of crack length after genuine CF initiation. Apart from the tests with severely sensitised specimens under BWR/NWC conditions, a transition phase with apparent slowly accelerating crack growth was always observed after crack initiation before the steady-state short crack growth regime was reached. Depending on test and notch parameters, this transition typically occurred over a crack advance of 50 to 300 μm. The subsequent stationary short CF crack growth rates were always exactly in the same range as the results from tests with pre-cracked specimens with long cracks of several mm to cm generated within this project [10] (Figure 24). It is believed, that this transition phase is mainly related to crack engagement over the whole specimen thickness from several semi-elliptical surface cracks and, to a lower extent, to the formation of mature occluded long crack crevice chemistry.

![Figure 23: Evolution of CF crack growth from a sharply notched 0.5T C(T) specimen during a fatigue test in BWR/HWC environment.](image)

![Figure 24: Comparison of short and long CF crack growth rates in PWR and BWR/HWC environment.](image)
The initiation and short crack growth behaviour in air at room-temperature and HWC environment at 288 °C under otherwise identical conditions is shown in Figures 25 and 26.

![Figure 25:](image1) Comparison of fatigue initiation and short crack growth behaviour in air at room-temperature and HWC environment at 288 °C.

![Figure 26:](image2) Reduction of fatigue initiation life in solution annealed low-carbon SS in simulated BWR/HWC environment.

The effect of BWR/HWC environment on the genuine initiation (average crack advance from notch-root of about 10 μm) appears to be drastic at a first glance. For this specific case with a notch strain rate of $2 \times 10^{-2} \%/s$, the environment resulted in a reduction of the fatigue initiation life by a factor of almost 30 with respect to corresponding experiments in air at room-temperature under otherwise identical conditions! From a practical point of view the effects are less severe. Technical crack initiation is related to the non-destructive testing-limit of the periodic in-service inspection, which is typically in the range of 5 to 10 % of the wall thickness. This corresponds to a crack advance of 0.5 to 1 mm for a pipe wall thickness of 10 mm. As shown in Table 4, the environmental effects are less severe for such crack advances and are close to the environmental acceleration of the short fatigue crack growth rate. The later parameter is thus more relevant for practice and, as will be shown in Section 1.3.2.5, excellently correlates with the environmental factors suggested in NUREG/CR-6909 [5].

-25-
Table 4: Comparison of genuine and technical fatigue initiation and stationary short crack growth in air at room-temperature and BWR/HWC environment at 288 °C.

<table>
<thead>
<tr>
<th></th>
<th>Ni ( \Delta a = 10 , \mu m ) [-]</th>
<th>( \frac{\Delta a}{dN_{\text{steady state}}} ) ( [\mu m/c] )</th>
<th>Ni ( \Delta a = 500 , \mu m ) [-]</th>
</tr>
</thead>
<tbody>
<tr>
<td>HWC, 288 °C</td>
<td>125</td>
<td>4.44E-1</td>
<td>1561</td>
</tr>
<tr>
<td>Air, 25 °C</td>
<td>3173</td>
<td>7.08E-2</td>
<td>12808</td>
</tr>
<tr>
<td>( N_{\text{Air}} / N_{\text{HWC}} ) [-]</td>
<td>25.4</td>
<td>-</td>
<td>8.2</td>
</tr>
<tr>
<td>( \frac{\Delta a}{dN_{\text{HWC}}} / \frac{\Delta a}{dN_{\text{Air}}} ) [-]</td>
<td>-</td>
<td>6.3</td>
<td>-</td>
</tr>
</tbody>
</table>

The typical scatter for CF initiation and short crack growth is shown in Figure 26 based on four tests on identical conditions. The observed scatter is in the same range as for fatigue in air. It is expected that for small strain amplitudes and strain rates the scatter can be relevantly larger.

Figure 26: Typical scatter of CF initiation (a) and short crack growth (b).

1.3.2.2 Effect of loading parameters

The effect of strain rate on CF initiation and subsequent stationary short crack growth under simulated BWR/HWC conditions is exemplarily shown in Figure 27a. Below a strain rate threshold of about 1\( \times 10^{-1} \) %/s, the environmental reduction of fatigue initiation life and the environmental acceleration of fatigue crack growth increased with decreasing strain rate, whereas the number of cycles to CF initiation and the absolute value of the CF crack growth rate decreased. Above this strain rate threshold the environmental effects disappeared. The effect of the environment on genuine fatigue initiation was a factor of 5 to 10 stronger than on subsequent stationary short crack growth. The environmental acceleration of long crack fatigue crack growth showed the same trends with respect to loading frequency (Figure 27b).
Figure 27: Effect of strain rate on environmental reduction/acceleration of fatigue initiation life/crack growth (a) and of loading frequency on long crack growth (b).

Figure 28 shows the effect of notch-root stress amplitude $S_{a,LEFM}$ on CF initiation for a solution annealed 304L SS in simulated HWC environment at 288 °C at a strain rate of $2 \times 10^{-2}$ %/s. It seems that environmental effects disappear at very high and very small strain/stress amplitudes and are maximal in the intermediate range. The rather limited amount of data suggest that environmental effects on the fatigue initiation from notched surfaces disappear below a strain amplitude threshold of 0.3 %, which is in good agreement with corresponding results from strain-controlled LCF experiments with smooth specimens [5]. It is stressed that this threshold might be significantly lower in case of positive mean stresses [26, 27] or at temperatures below 250 °C, where secondary hardening by dynamic strain ageing is absent (see Figures 34 and 35 in Section 1.3.2.4).

Figure 28: Effect of notch stress amplitude on CF initiation life.

1.3.2.3 Effect of material parameters

As shown in Figures 29a, 29b and 30a, in PWR or BWR/HWC environment at intermediate strain amplitudes, the environmental reduction of fatigue initiation life and environmental acceleration of the stationary short fatigue crack growth of sensitised and solution annealed SS was comparable for the high- and low-carbon grades 304, 304L and 316L as well as for the stabilised grades 321 and 347. More pronounced material effects might appear at small strain amplitudes, where the materials may reveal different secondary hardening behaviour (dynamic strain ageing, martensite formation). A similar behaviour was also observed in tests with long cracks at PSI (Figure 14, Section 1.3.1.2) [10].

-27-
In contrast to PWR or BWR/HWC conditions, much shorter CF initiation lives were observed in sensitised than in solution annealed SS under highly oxidising BWR/NWC conditions. In severely sensitised specimens, the crack even initiated during the first few fatigue cycles. In BWR/NWC the CF lives of solution annealed austenitic SS were slightly higher than in PWR or BWR/HWC environment.

After initiation, the subsequent stationary short CF crack growth rates under PWR or BWR/HWC conditions were comparable for sensitised and solution annealed SS (Figure 31) and behaved similar as long cracks. Under highly oxidising BWR/NWC conditions, the CF crack growth rates in sensitised materials were only slightly higher (typically a factor of two or less) than in solution annealed steels for loading frequencies $> 5 \times 10^{-4}$ Hz, which in turn were usually a factor of 2 to 5 higher than under PWR or BWR/HWC conditions (Figure 8) [10].

In summary, strong effects of sensitisation are thus observed on CF initiation at high ECPs only! At very low loading frequencies or high load ratios, sensitisation may also have a strong effect on short (and long) CF crack growth because of a possible SCC contribution. Otherwise the effects of sensitisation are moderate.
1.3.2.4 Effect of environmental parameters

The effect of temperature on CF crack initiation and short crack growth is exemplarily shown in Figures 32a and 33 for experiments with a solution annealed 316L SS under simulated BWR/HWC conditions. Relevant environmental effects on fatigue initiation and short crack growth were observed above 100 and 150 °C, respectively. Above this threshold the environmental reduction of fatigue life and acceleration of fatigue crack growth increased with increasing temperature and were more pronounced than in air, where temperature usually has little effect on LCF life between 25 and 350 °C. The temperature effects on CF crack initiation and short and long growth were very similar (Figure 32). The effect of the environment again was more pronounced on fatigue initiation than on subsequent crack growth.

Figure 31: Effect of sensitisation on short CF crack growth in PWR and BWR environments.

Figure 32: Effect of temperature on environmental reduction of fatigue initiation life and environmental acceleration of short fatigue crack growth in notched specimen (a) and of temperature on environmental acceleration of long fatigue crack growth in pre-cracked specimens (b).
At small strain/stress amplitudes, an opposite temperature trend was observed in preliminary experiments (Figure 34), which was also observed in [26, 27]. The different behaviour at small strain/stress amplitudes is probably related to the distinct secondary hardening at small strain amplitudes and high temperatures due to dynamic strain ageing (Figure 35).

![Figure 33](image1.png)

**Figure 33:** Effect of temperature on CF initiation life and crack growth rate.

![Figure 34](image2.png)

**Figure 34:** Opposite temperature trends on CF initiation at high and small notch stress amplitudes.

![Figure 35](image3.png)

**Figure 35:** Isothermal LCF at 340 (a) and 100 °C (a) in air. Distinct secondary hardening at small strain amplitudes and high temperatures only [9].
Figure 36 exemplarily shows the effect of ECP on CF initiation and subsequent stationary short crack growth for a solution annealed 304L austenitic SS at a temperature of 288 °C. The typical ECP ranges of SS under stationary PWR, BWR/HWC and BWR/NWC operation conditions are also shown. For the given loading conditions, a maximum environmental reduction of fatigue life was observed in high-purity HWC environment at a low ECP of -560 mV$_{SHE}$. In the HWC ECP range of -550 to -200 mV$_{SHE}$, the CF initiation life did not remarkably depend on ECP or DH content in high-purity water (Figure 36). Under highly oxidising BWR/NWC conditions, the environmental reduction of fatigue initiation life was smaller than under HWC conditions. This is totally different in case of sensitised SS (see Section 1.3.2.3). Similarly, the environmental reduction of the fatigue initiation life in the PWR range at lower ECPs was smaller than in high-purity BWR/HWC water. This is very probably related to the slightly higher pH$_{288 °C}$ of the PWR water of 7.1 with respect to 5.7 for high-purity water. The effect of ECP on the environmental acceleration of the subsequent stationary short CF crack growth, on the other hand, was comparable from -750 to -200 mV$_{SHE}$ in good agreement with corresponding investigations of the growth of long cracks (Figures 15 and 16 in Section 1.3.1.2) [10]. Under highly oxidising BWR/NWC conditions, where the environmental reduction of fatigue initiation life was smallest, the environmental acceleration of the stationary short CF crack growth was a factor of 2 to 5 higher than under PWR or BWR/HWC conditions at low ECPs! This is clearly related to the aggressive occluded crack crevice chemistry, which is formed under oxidising BWR/NWC conditions.

![Figure 35](image1.png)

**Figure 35:** Effect of ECP on CF initiation and subsequent short crack growth.

![Figure 36](image2.png)

**Figure 36:** Effect of DH content on CF initiation life in high-purity water.
The observed effect of ECP, temperature and pH under PWR and BWR/HWC conditions suggest that environmental effects on fatigue initiation might be related to the general corrosion behaviour (and the stability of oxide films of these steels), which increases with increasing temperature and decreasing ECP and Cr content.

1.3.2.5 Comparison of short and long CF crack growth with CF initiation

Stationary short crack CF crack growth rates after crack advances of 50 to 300 μm from the notch-root were in the typical range of corresponding results from long crack as shown in Figure 24. They also showed exactly the same response to changes of the different system parameters.

Similar parameter effects on CF initiation as on CF crack growth were observed in most cases (temperature, strain rate, etc.), but in some few cases, the trends were different. The environmental acceleration of short CF crack growth under highly oxidising BWR/NWC conditions, for example, was more pronounced than in PWR and BWR/HWC environment, whereas the environmental reduction of fatigue initiation life was smaller. Although very remarkable for both cases, the effect of the environment on the fatigue initiation was often stronger than on the subsequent short crack growth.

1.3.2.6 Comparison with strain-controlled LCF experiments and the NRC Reg. Guide 1.207

In Figure 37, selected PSI test results from load-controlled experiments with the sharply notched C(T) specimens are compared with the ASME III design and mean curve of the ASME BPV Code as well as with new NRC Regulatory Guide 1.207 [7], which considers environmental effects. The NRC Regulatory Guide 1.207 is based on strain-controlled LCF tests with smooth cylindrical specimens as described in NUREG/CR-6909 [5]. The genuine CF initiation life may be smaller than predicted by the ASME III design curve for certain critical parameter combinations. As discussed below and in Section 1.3.2.1, this is not as critically as it may appear at a first glance, since technical initiation involves a significant amount of crack growth, which drastically reduces the extent of environmental effects. As shown in Figure 37, environmental effects on the genuine fatigue initiation (Δa ≈ 10 μm) in load-controlled experiments with the sharply notched C(T) specimens were much more pronounced than predicted by the NUREG/CR-6909 methodology. The reason for this discrepancy is that in the strain-controlled LCF tests with smooth specimens, a 5 or 25 % load drop criteria for crack initiation is used (= “technical initiation”), which very roughly corresponds to a 5 or 25 % reduction in the cross section of the fatigue specimen by the CF crack growth. For the case of homogeneous crack growth around the circumference and a specimen radius of 5 mm, this would correspond to a crack advance of 125 and 670 μm. If this aspect is adequately considered, the apparent discrepancy disappears as shown in the next paragraph.
Figure 37: Comparison of selected CF test results with the ASME III design and mean curve and the NUREG/CR-6909 approach [5] of the NRC Regulatory Guide 1.207 [7].

As in NUREG/CR-6909, environmental factors were defined for the PSI test results for the environmental reduction of the genuine initiation life (equation (7)) and acceleration of the subsequent stationary short crack growth (equation (8)).

\[ F_{env}^i = \frac{N_{Air}^i}{N_{H_2O}^i} \]  
\[ F_{SCG,env}^i = \frac{da/dt_{H_2O}}{da/dt_{Air}} \]

In Figure 38, these environmental factors are compared to those of NUREG/CR-6909 for selected conditions. A very good correlation of environmental acceleration of short crack growth in PSI tests with the environmental reduction of fatigue life in LCF tests at ANL is observed over a wide range of system conditions, which is thus a good qualitative confirmation of the NUREG/CR-6909 approach. Very similar parameter effects were observed in the load-controlled experiments with the sharply notched C(T) specimens as in strain-controlled LCF tests. On the other hand, environmental effects on genuine initiation were much more severe than on (short) crack growth or on LCF initiation life, and the difference was most pronounced under those system conditions, which result in distinct environmental effects (e.g., at slow strain rates or high temperatures).

Figure 38: Comparison of environmental factors for genuine crack initiation and short crack growth from PSI tests with notched specimens with environmental factors of NUREG/CR-6909 [5] at different strain rates (a) and temperatures (b).
Tests with sharp and blunt notch C(T) specimens are an alternative and simple way to systematically study environmental effects on LCF and HCF initiation. In contrast to strain-controlled LCF tests, a better separation of environmental effects on the initiation and subsequent short crack growth process is possible. Furthermore, the mechanical conditions are more representative for an operating component. On the other hand, the notch strain has to be estimated based on notch concepts or by Finite Element modelling. The strain-controlled LCF experiments, on the other hand also provide information on the cyclic plastic behaviour of the material.

1.3.2.7 Fractography

Apart from sensitised SS in BWR/NWC environment, the fracture mode for genuine crack initiation and short crack growth in sharply notched C(T) specimens was always TG, most likely along crystallographic planes, leaving behind relatively smooth facets. Fatigue striations perpendicular to the local crack growth direction were usually observed at temperatures > 150 °C everywhere on the fracture surface apart from the first 5 to 10 μm behind the notch-root (Figure 39). The fractographical appearance was very similar as for long crack growth (see Section 1.3.1.4). There was little effect of material and environmental parameters on the general fracture surface appearance in the investigated parameter range. In case of severely sensitised SS in BWR/NWC environment, the fracture mode for genuine crack initiation and short crack growth was IG with fatigue striations on the IG facets.

Figure 39: Typical appearance of TG corrosion fatigue in 316L under BWR/HWC conditions at 288 °C. 5 to 10 μm behind the notch-root, fatigue striations can be seen everywhere on the fracture surface.
1.3.2.8 Summary

Relevant environmental reduction of fatigue initiation life was observed in all SS at strain rates $\leq 0.1 \%$/s in BWR/NWC, BWR/HWC and PWR environment. Subsequent to initiation, the stationary short crack CF crack growth rates after crack advances of 50 to 300 $\mu$m from the notch-root were in the typical range of corresponding results from tests with long cracks (pre-cracked specimens) and also showed the same system parameter response. The effect of environment on the genuine initiation process ($\Delta a \approx 10 \mu$m) was relevantly stronger than on the subsequent stationary short crack growth.

Both, under BWR/HWC and PWR conditions, a relevant environmental reduction of fatigue initiation life occurred for the combination of temperatures $\geq 100 ^\circ$C, notch strain rates $\leq 0.1 \%$/s and notch strain amplitudes $\geq 0.3 \%$. If these conjoint threshold conditions were simultaneously satisfied, the environmental enhancement increased with decreasing strain rate and increasing temperature. Material and water chemistry parameters usually only had a little effect. Sensitisation affected the CF behaviour under highly oxidising BWR/NWC conditions only.

If the critical conjoint requirements were satisfied, the BWR/HWC and PWR environments usually resulted in acceleration of short fatigue crack growth by a factor of 5 to 20 with respect to air and the environmental reduction of the genuine initiation life was even more pronounced. Solution annealed steels showed slightly shorter CF initiation lives, but also lower stationary short CF crack growth rates under BWR/HWC and PWR conditions with low ECPs than under highly oxidising BWR/NWC conditions.

The fatigue design curves in Section III of the ASME BPV Code were not conservative for some critical system parameter combinations. The CF initiation studies qualitatively confirmed the NUREG/CR-6909 approach of the new US NRC Regulatory Guide 1.207, which defines a procedure to include environmental effects in Section III. A very good correlation of environmental acceleration of short crack growth in PSI tests with the environmental reduction of fatigue life in LCF tests at ANL was observed over a wide range of system conditions.
1.4 Practical implications and consequences

1.4.1 Adequacy and conservatism of fatigue design and evaluation codes

1.4.1.1 Fatigue design and evaluation according to ASME III and XI

The fatigue design and flaw tolerance evaluation procedures for class 1 components of LWRs are described in Section III and XI of the ASME BPV Code. The basic ideas of the methods are described in Figures 40 to 42.

![Fatigue crack growth](image1)
![Fatigue initiation](image2)

**Figure 40:** Overview and comparison of fatigue design and flaw tolerance evaluation procedures.

The ASME Code Section III fatigue design procedures are described in NB-3200, “Design by Analysis,” and NB-3600, “Piping Design”. For each stress cycle or load set pair, an individual fatigue usage factor is determined by the ratio of the number of cycles anticipated during the lifetime of the component to the allowable cycles. Appendix I to Section III of the ASME BPV specifies the fatigue design curves that define the allowable number of cycles as a function of applied stress amplitude. The cumulative usage factor is the sum of the individual usage factors, and ASME Code Section III requires that the cumulative usage factor at each location must not exceed 1. Possible effects of the reactor coolant environment on fatigue life are not directly considered. The US NRC Regulatory Guide 1.207 from 2007 [7] and the related NUREG/CR-6909 report [5] describe a procedure, how environmental effects can be considered in the current ASME Section III fatigue design procedure, e.g., for the case of license renewal (lifetime extension) or the construction of a new plant, by applying environmental factors $F_{env}$. 

-36-
• Fatigue usage factor $UF = \frac{n}{N}$ for each stress range
• Cumulated fatigue usage factor $CUF = \sum \frac{n}{N} < 1$ at each location

**Figure 41:** Fatigue design and evaluation according to Section III of the ASME BPV Code.

Appendix C “Evaluation of Flaws in Austenitic Piping” of Section XI of the ASME BPV Code describes a fatigue flaw tolerance evaluation procedure in combination with periodic in-service inspection (ISI) by non-destructive testing. The anticipated fatigue crack growth $\Delta a$ from an initial crack $a_0$ by the load cycles during power operation shall not result in a final crack length at the end of the inspection interval $\Delta t_{ISI}$, which is larger than the maximum allowable crack length $a_f$. If this condition is not satisfied with the current inspection interval, the component has to be repaired or replaced or the inspection interval has to be reduced. The maximum allowable crack length $a_f$ for very ductile austenitic alloys is calculated by a plastic limit load analysis or elastic plastic fracture mechanics under normal operation and accident conditions. The fatigue crack growth is calculated by LEFM by the fatigue crack growth curves given in Appendix C. The curves for austenitic alloys do not consider environmental effects. A corresponding revision to consider environmental effects for Ni-base alloys is currently in preparation [28] and also planned for SS.

**Figure 42:** Fatigue flaw tolerance evaluation according to Section XI of ASME BPV Code.
1.4.1.2 Conservatism of fatigue codes with respect to environmental effects

The emphasis of the following discussion is primarily placed to fatigue design according to Section III, but many similarities exist for Section XI.

The conservatism in the ASME Section III Code fatigue evaluations may arise from (a) the fatigue evaluation procedures and/or (b) the fatigue design curves. The overall conservatism in ASME Code fatigue evaluations has been demonstrated in fatigue tests on components in air, which showed that the design margin for cracking exceeds 20, and for most of the components it is greater than 100, although for welds it may be far below 20. However, some isolated other studies also indicate that the code fatigue design procedures do not always ensure large margins of safety. [5, 29, 30]

a) The sources of conservatism in the procedures include the use of design transients that are significantly more severe than those experienced in service (see Table 5 and Figure 43), conservative grouping of transients, and use of simplified elastic-plastic analyses that result in higher stresses/strains [30]. With respect to real transients, the significantly higher strains of design transients usually fully compensate the smaller environmental reduction of fatigue life because of the higher strain rates [31, 32] (Figure 43). However, the ASME Code permits new and improved approaches to fatigue evaluations (e.g., Finite Element analyses, fatigue monitoring and improved plastification factors $K_e$) that can significantly decrease the conservatism in the current fatigue evaluation procedures.

**Table 5:** Comparison of reactor and lab test conditions and the applied design approach [33].

<table>
<thead>
<tr>
<th>Actual plant fatigue cycling</th>
<th>Cycling as analyzed in ASME Code analyses</th>
<th>Fatigue cycling in most EAC Test Programs</th>
</tr>
</thead>
<tbody>
<tr>
<td>Combined pressure and temperature transients</td>
<td>analyzed based on elastic behavior</td>
<td>All the tests are based on mechanical load cycling alone.</td>
</tr>
<tr>
<td>Temperature varies during much of the stress cycling</td>
<td>Temperature effects are included in the analysis</td>
<td>Almost all the tests are done at constant temperature.</td>
</tr>
<tr>
<td>During a cool-down highest stresses occur at lower temperatures</td>
<td>The concurrence of low temperatures and high thermal stresses is not addressed.</td>
<td>Effect of lower temperatures during tensile stress is not addressed (isothermal tests).</td>
</tr>
<tr>
<td>The ramp times for the different transients are not known.</td>
<td>Most thermal transients are analyzed as step changes</td>
<td>Ramp time (1/frequency) is an important variable</td>
</tr>
<tr>
<td>Weld residual stresses exist but are not quantified.</td>
<td>Weld residual stresses are not explicitly considered in ASME code fatigue analysis.</td>
<td>Residual stresses affect the $R$ ratio and crack growth rates.</td>
</tr>
<tr>
<td>Peak to peak ranges cycling come from transients that could be separated by months</td>
<td>Fatigue usage is computed by combining the extreme cases.</td>
<td>Cycling in the test is well defined (not separated by months), uniform, in sequence.</td>
</tr>
<tr>
<td>Water Chemistry (conductivity, sulfate and chloride, oxygen) varies during plant operation</td>
<td>Water chemistry is not a consideration in ASME Code fatigue analysis.</td>
<td>Autoclave water chemistry is an important factor in fatigue initiation and crack growth.</td>
</tr>
</tbody>
</table>
Figure 43: The very steep temperature profile, which is assumed in many design transients, can easily result in strain amplitudes or ΔK values, which are a factor of two higher than in real transients. The given example with arbitrary chosen parameters is based on the environmental factors in NUREG/CR-6909 and the PSI CF crack growth curves and shows that the higher strain or ΔK level of the design transient may fully compensate the stronger environmental effects of real transients because of slower strain rates.

b) The ASME Code fatigue design curves, given in Appendix I of Section III, are based on strain-controlled tests of small polished specimens at room-temperature in air. The design curves have been developed from the best-fit curves to the experimental fatigue-strain vs. life (ε–N) data that are expressed in terms of the Langer equation. The code fatigue design curves are obtained from the best-fit curves of the experimental data by first adjusting for the effects of mean stress on fatigue life (according to Goodman) and then reducing the fatigue life at each point on the adjusted curve by a factor of two on strain (or stress) or 20 on cycles, whichever is more conservative. [5, 29]

The factors of two and 20 were intended to cover the effects of variables that can influence fatigue life but were not investigated in the tests which provided the data for the curves. The contributions of four groups of variables, namely material variability and data scatter, differences in surface condition and size/geometry between the test specimens and actual components and loading history (load sequence effects), must be considered in developing the fatigue design curves that are applicable to components. These factors are thus not safety margins but rather adjustment factors that should be applied to the small specimen data to obtain reasonable estimates of the lives of actual reactor components. An analysis of the literature data by ANL showed that these factors are quite reasonable to account for differences and uncertainties in fatigue life, and do not contain excess conservatism that can be assumed to account for the effects of LWR environments. [5, 30]

As shown in the previous Sections, there is clear experimental evidence that, if certain conjoint threshold conditions were simultaneously satisfied, the fatigue life and fatigue crack growth in austenitic SS may be relevantly reduced and accelerated in BWR/NWC, BWR/HWC or PWR environment, respectively. There is no evident reason, why such environmental effects should not occur in the field.

Systematic CF cracking, as it might be expected based on the severity of environmental effects in the lab, was not observed in LWR service so far. Only few isolated fatigue cracking incidents in PPBC occurred in recent years under very specific circumstances. These cracking incidents were mainly related to either HCF through flow-induced vibrations from power up-ratings
or to thermal fatigue (HCF and LCF) caused by complex thermal-hydraulic phenomena (thermal stratification and striping, turbulent mixing), which were not anticipated during the design phase and thus not included in the original design [1, 33, 34]. In contrast to the HCF damage due to high-frequency vibrations, where strain rates are too high for significant environmental effects, the LCF or combined LCF and HCF damage due to thermal loadings might have been aggravated by corrosion effects due to exposure to the reactor coolant.

This apparent discrepancy between lab results and field experience is mainly related to the large degree of conservatism in the fatigue evaluation procedures mentioned before (a). As shown in Figure 43, the significantly higher strains of design transients often fully compensate possible environmental effects of real transients. Furthermore, for most transients one or several threshold conditions are not satisfied and environmental effects on fatigue are thus usually moderate and less severe than it may appear at a first glance. Transients with very strong environmental effects (i.e., with slow strain rates) are usually not very damaging with respect to fatigue damage accumulation because of their small strain amplitudes in combination with the rather limited cycle numbers during the whole lifetime.

The excellent field experience of fatigue-designed SS PPBC does not indicate any generic deficiencies in the current ASME III fatigue design procedure by the ignorance of environmental effects. It is thus well-approved and seems to be conservative and adequate under most operating circumstances, even in case of environmental effects. There is thus no pressure for immediate actions, i.e., for an immediate application of the US NRC Regulatory Code 1.207/NUREG/CR-6909 approach to existing plants. Since there are relevant differences between lab and component conditions in the field [33, 35] (Table 5 and Figure 44), this approach should be further verified under more plant relevant conditions (e.g., small strain amplitudes, thermal mechanical fatigue, mean stress, long periods with static load, etc.) before it should be applied to existing plants. Nevertheless, in particular in the context of plant life extension, environmental effects and the lab observations have to be considered in an adequate way, otherwise the current trend of applying improved fatigue evaluation approaches (replacement of design by real transients, fatigue monitoring systems), that reduce the degree of conservatism, might result in a non-acceptable reduction of the safety margins! [5, 29, 30]

**Experiments:** Uniaxial isothermal strain-controlled constant amplitude & strain rate LCF experiments at $R_e = -1$ with smooth and polished base metal specimens with plastic yielding of the full specimen ligament in air or in simulated stationary LWR environment

**Reality (e.g., turbulent mixing, thermal striping):** Biaxial thermally-induced variable & small strain amplitude fluctuations at high mean stress $R_\sigma > 0$ (pressure, residual stress, etc.), which are interrupted by long periods of moderate static stress (stationary power operation). Stress gradients. Real surfaces (roughness, cold-work, residual stress, pitting, etc.) with defects (weld defects, notches) and weld/weld HAZ materials. Sometimes in connection with different water chemistry (start-up/shut-down, ...)

![Figure 44: Comparison of lab test conditions with conditions for thermal fatigue in case of turbulent mixing or thermal striping.](image-url)
Although still being rather conservative, the flaw tolerance evaluation according to Appendix C of Section XI of the ASME BPV Code probably contains fewer margins than the fatigue design according to Section III. The temperature dependent air fatigue crack growth curves for austenitic SS in Appendix C accurately predict the fatigue crack growth in the Paris range with only slight conservatism [1, 36, 37]. They represent an upper bound behaviour of different materials. There is only little material to material variability (usually a scatter band of a factor of two), as the different material parameters of austenitic and cast SS only have little effect on fatigue crack growth rates in the Paris range far away from the thresholds [36, 37]. Since the uncertainties for the transfer of lab results on large components in service are much smaller for fatigue crack growth than for fatigue initiation, the reduced margins on the material behaviour side seem to be justified. The calculation of the maximum allowable crack length involves some conservatism on the material and load side (usually a safety factor of three (normal operating) and 1.5 (accident conditions) on the load). In the ductile austenitic SS, these maximum allowable crack lengths are very large, giving a lot of margin for fatigue crack growth during service (this situation may be significantly different for certain susceptible cast SS, where thermal ageing can result in relevant embrittlement and reduction of toughness during service and thus in a drastic reduction of the allowable crack length). The largest amount of conservatism is introduced, if design transients are used to calculate the $\Delta K$ values.

The flaw tolerance evaluation according to Appendix C of Section XI of the ASME BPV Code can be applied if a crack indication is found during the periodic in-service inspection in order to decide, whether the component has to be immediately repaired or replaced or continued operation for one further fuel cycle is allowed. If all threshold conditions are satisfied, environmental effects can fully compensate the safety margins from the use of design transients, drastically reducing the total margins. Therefore, a modification of the current ASME XI fatigue crack growth curves to include environmental effects should be urgently pursued.

1.4.2 Critical system conditions and component locations

Since no systematic CF incidents were observed in austenitic SS PPBC of LWRs, no direct conclusions with respect to critical components can be derived from service experience. Some very specific cases with thermal fatigue (HCF and LCF) [1, 34], which were caused by complex thermal-hydraulic phenomena, might have been aggravated by corrosion effects due to exposure to the reactor coolant, but a dominant or relevant role of the environment is not proven.

Both under BWR/HWC and PWR conditions, a relevant environmental reduction of fatigue initiation life (and relevant acceleration of fatigue crack growth) only occurs for the combination of temperatures $\geq 100 \, ^\circ C$ ($\geq 150 \, ^\circ C$), strain rates $\leq 0.1 \, %/s$ (loading frequencies $\leq 0.1 \, Hz$) and strain amplitudes $\geq 0.3 \, \%$. ($\Delta K$ values $\geq 3 \, MPa-m^{1/2}$) in all low-carbon and stabilised austenitic SS. If these conjoint requirements are satisfied, the environmental effects increase with increasing strain rate (loading frequency) and increasing temperature. Furthermore, it has to be considered that the fatigue damage accumulation is roughly proportional to the cycle number and that both fatigue and CF damage accumulation per cycle (and crack growth increment per fatigue cycle) strongly increase with increasing strain amplitude (and $\Delta K$ values).

The above mentioned threshold conditions and aspects can be used as screening criteria to identify components and transients, where environmental effects could play a relevant role. The extent of environmental effects on fatigue initiation and crack growth can be estimated by NUREG/CR-6909 [5] and the corrosion fatigue curves in [17] (BWR/NEC) and [19, 22] (PWR or BWR/HWC), respectively. By the strain rate threshold, e.g., environmental effects can be excluded for flow-induced or mechanical vibrations. Strain rates and amplitudes can vary over a wide range during service. The $\Delta K$ values additionally depend on the crack length and depth. The strain rates, for example, can range from $1 \cdot 10^{-7} \, s^{-1}$ (for certain stratification phenomena) up to $0.1 \, s^{-1}$ (severe thermal shock) or even higher values for mechanical vibration phenomena.
Even if relevant environmental effects might occur, the normal operating pressurisation and thermal cycles (e.g., start-up/shut-down with limitation of heating/cooling rate to < 50 °C/h) are not generally the source of significant fatigue damage accumulation in nuclear power plants, since they are adequately covered by the original component design. The pressurisation cycle may be relevant for fatigue crack growth, in particular for deep and long cracks.

The $\Delta K$-threshold is easily exceeded in all components containing cracks and quite high values can be reached for deep cracks or superimposed residual stresses, e.g., in weld heat-affected zones. The crack-tip strain rates of pressurisation and thermal cycles during start-up and shut-down are in a range, where relevant acceleration of fatigue crack growth occurs. The low number of such transients limits the total crack advance, in particular, in the view of the large maximum allowable crack length for ductile austenitic SS. Thermal stresses are often restricted to near surface regions and cannot produce through wall fatigue cracking alone.

Primary design stresses are limited to values below the high-temperature yield strength. The strain amplitude threshold is thus only exceeded in those components with high secondary stresses (e.g., by thermal stratification) or at notches, which act as local stress raisers. For most thermal transients there is a competition between the increase of environmental effects by decreasing strain rates and the decrease of fatigue damage accumulation by the decreasing strain amplitudes [32] and it may be expected that there is a maximum for the combined effect somewhere at intermediate strain rates. Thermal transients with large strain amplitudes (e.g., thermal shock), which are very efficient with respect of fatigue damage accumulation per fatigue cycle, are usually related to high strain rates and (very) rare events. Because of the high strain rates, environmental effects do not play a role here. Thermal transients with slow strain rates and strong environmental effects are usually related to small to medium strain amplitudes and infrequent events (e.g., thermal stratification during plant start-up or hot standby). In spite of the strong environmental effect, the total accumulated fatigue damage is moderate. Transients with large cycle numbers are usually related to small strain amplitudes and medium to high strain rates (thermal striping or turbulent mixing of cold and hot water in T-joints). But in superposition with thermal stratification and pressurisation and thermal cycles from plant start-up/shut-down, the combination of high cycle numbers and moderate environmental effects might eventually result in a relevant total (corrosion) fatigue damage accumulation here.

The most critical austenitic SS components with respect to CF are probably those, which may experience high cumulated fatigue usage factors close to one during the plant life without considering environmental effects and where, at least partially, the strain rates of the fatigue relevant loadings are in the critical range for environmental effects.
1.5 Summary and conclusions

The CF behaviour of different low-carbon and stabilised austenitic SS was characterised under simulated BWR and primary PWR conditions by cyclic fatigue tests with pre-cracked and notched-only fracture mechanics specimens in the temperature range from 70 to 320 °C. The special emphasis was placed to BWR/HWC and PWR conditions at low ECPs.

Relevant environmental acceleration of fatigue crack growth (and reduction of fatigue initiation life) was observed in all steels at loading frequencies \( \leq 0.1 \text{ Hz} \) (strain rates \( \leq 0.1 \%/\text{s} \)) and could be sustained down to the lowest investigated loading frequencies (strain rates) in all three environments. Subsequent to initiation, the stationary short crack CF crack growth rates, after crack advances of 50 to 300 \( \mu \text{m} \) from the notch-root, were in the typical range of corresponding results from tests with long cracks (pre-cracked specimens) and also showed the same system parameter response. The effect of environment on the genuine initiation process (\( \Delta a \approx 10 \mu \text{m} \)) was relevantly stronger than on the subsequent stationary short crack growth.

At low ECPs under BWR/HWC and PWR conditions, relevant environmental acceleration of fatigue crack growth (and environmental reduction of fatigue initiation life) occurred in all investigated low-carbon and stabilised SS for the combination of loading frequencies \( \leq 0.1 \text{ Hz} \) (strain rates \( \leq 0.1 \%/\text{s} \)), temperatures \( \geq 150 \text{ °C} (\geq 100 \text{ °C}) \) and \( \Delta K \) values \( \geq 3 \text{ MPa}\cdot\text{m}^{1/2} \) (notch strain amplitudes \( \geq 0.3 \% \)). If these conjoint threshold conditions were simultaneously satisfied, the environmental enhancement increased with decreasing loading frequency (strain rate) and increasing temperature. Material and water chemistry parameters usually only had little effect. Sensitisation affected the CF behaviour under highly oxidising BWR/NWC conditions only.

If the critical conjoint requirements were satisfied, the BWR/HWC and PWR environments usually resulted in acceleration of fatigue crack growth by a factor of five to 20 with respect to air and the environmental reduction of the genuine initiation life was even more pronounced. Under comparable loading conditions, the (short and long crack) CF crack growth rates under oxidising BWR/NWC conditions were usually a factor of two to five higher than under BWR/HWC or PWR conditions at low ECPs. The genuine CF initiation life of solution annealed steels, on the other hand, was slightly smaller in BWR/HWC and PWR environments.

The CF behaviour under BWR/NWC conditions may be rationalised by the slip-dissolution model developed by Ford & Andresen. The observed cracking behaviour under BWR/HWC or PWR conditions suggests that a different mechanism might be active at low ECPs, which might be related to the increased general corrosion rate and less protective oxide films of austenitic SS at low ECPs.

The current ASME BPV Code Section XI fatigue crack growth curves do not include environmental effects and are not conservative in LWR environments if the above mentioned conjoint threshold conditions were satisfied. A modification of Section XI should therefore be pursued to include environmental effects in an appropriate way! The JSME BWR curve conservatively bounds the overwhelming part of CF data in simulated BWR/NWC environment. The Japanese PWR curve proposal reasonably predicts CF crack growth under PWR or BWR/HWC conditions.

The fatigue design curves in Section III of the ASME BPV Code do not contain excess conservatism to cover environmental effects and are not conservative for some critical parameter combinations. The CF initiation studies qualitatively confirmed the NUREG/CR-6909 approach of the new US NRC Regulatory Guide 1.207, which defines a procedure to include environmental effects in Section III. A very good correlation of environmental acceleration of short crack growth in PSI tests with the environmental reduction of fatigue life in LCF tests at ANL was observed over a wide range of system conditions. Nevertheless, this approach should be critically evaluated by additional tests under more realistic and plant relevant conditions, before it is applied to existing plants.
The possibility of environmental effects on fatigue under LWR conditions is undisputed, but there is no international consensus about the practical consequences. The accumulated excellent field experience of fatigue-designed SS PPBCs does not indicate any generic deficiencies in the current fatigue design procedures by the ignorance of environmental effects. This is related by a relevant part to the large degree of overall conservatism in current fatigue design and evaluation procedures and to the fact that for many plant transients one or several of the above mentioned conjoint threshold conditions are not satisfied and thus resulting in moderate environmental effects only. Furthermore, the very high ductility of austenitic SS results in a rather large critical crack size and a remarkable flaw tolerance in case of fatigue. There is thus no big pressure for immediate actions. The higher priority should be given to a modification of Section XI of the ASME BPV Code including SCC. In flaw tolerance evaluations, the contribution of SCC usually overwhelms that of fatigue under many conditions (e.g., BWR/NWC or cold-worked SS).
1.6 References


Part II

Evaluation of the potential and limits of the electrochemical noise measurement technique for the early detection of stress corrosion cracking initiation in austenitic stainless steel under BWR/NWC conditions

Stefan Ritter and Hans-Peter Seifert
# TABLE OF CONTENTS

2.1 INTRODUCTION................................................................................................................................. 51
  2.1.1 TECHNICAL BACKGROUND AND MOTIVATION ........................................................................ 51
  2.1.2 LITERATURE SURVEY .............................................................................................................. 51
  2.1.3 OBJECTIVES ............................................................................................................................ 51

2.2 MATERIALS AND EXPERIMENTAL PROCEDURE ...................................................................... 53
  2.2.1 MATERIALS AND SPECIMENS .............................................................................................. 53
  2.3 EXPERIMENTAL PROCEDURE ........................................................................................................ 54
    2.3.1 Room-temperature experiments in aqueous thiosulphate solution ........................................... 54
    2.3.2 Experiments in a simulated BWR environment ........................................................................ 55
    2.3.3 Electrochemical noise measurements ................................................................................. 56
    2.3.4 Direct current potential drop measurements ...................................................................... 58

2.4 RESULTS AND DISCUSSION...................................................................................................... 59
  2.4.1 ROOM-TEMPERATURE INVESTIGATIONS IN AQUEOUS THIOSULPHATE SOLUTION .......... 59
  2.4.2 INVESTIGATIONS IN A SIMULATED BWR ENVIRONMENT ......................................................... 61
    2.4.2.1 Combined electrochemical noise and DCPD measurements ............................................. 64
    2.4.2.2 Effect of reference electrode distance on the electrochemical noise signal .................... 68

2.5 SUMMARY AND CONCLUSIONS............................................................................................... 70

2.6 REFERENCES............................................................................................................................... 71
2.1 Introduction

2.1.1 Technical background and motivation

The current pressure of competition in the power generation industry requires implementation of optimised, strategic and proactive plant ageing and life management methods for cost-effective operation of nuclear power plants at high safety levels. The ongoing occurrence of stress corrosion cracking (SCC) in boiling water (BWRs) and pressurised water reactors (PWRs) worldwide clearly demonstrates a need for the development of advanced, non-destructive, continuous monitoring tools for the early detection of SCC initiation in the technical pre-crack stage.

The electrochemical noise (EN) measurement technique is a promising tool for continuous, in-situ corrosion monitoring in technical systems and has the potential to detect nucleation and initiation of localised corrosion processes [1, 2]. The potential and limits of this technique for the detection of SCC initiation in austenitic stainless steels under simulated BWR conditions was therefore investigated in experiments at PSI within a small feasibility study.

2.1.2 Literature survey

At the beginning of this project a literature survey on the detection of SCC by the EN technique was performed and summarised in a report [1]. This report covers the state-of-the-art in science and technology in this specific field. The major conclusions can be briefly summarised as follows:

The EN technique is a promising in-situ corrosion monitoring technique, which has a certain potential for the early detection of SCC initiation. Several independent studies, investigating the suitability of the EN technique to detect SCC initiation or propagation in room-temperature systems, have been performed. Electrochemical current and potential noise measurements indicated SCC initiation events, although in most cases the cracking was confirmed by post-test analysis only. Very few investigations could be found in literature, which aimed to demonstrate the feasibility of EN measurements during SCC tests in high-temperature water [3-6]. In spite of some promising results, no systematic studies were performed and no independent online SCC detection method was used to verify these results so far. Based on the published data, it was concluded that the EN technique is a promising online monitoring tool for the detection of SCC in room-temperature environments as well as in high-temperature water, but further systematic studies with independent online SCC detection methods are necessary to qualify and validate this technique as a corrosion monitoring tool.

2.1.3 Objectives

The primary objective of this project was to evaluate the potential and limits of the EN technique for the detection of SCC initiation in stainless steels under simulated BWR conditions within a small feasibility study. The project shall help to develop a tool/sensor for SCC initiation studies in high-temperature autoclave systems in the lab or material test reactors. In case of promising results, the method could then be further developed and qualified as a corrosion monitoring tool in BWRs or PWRs in a follow-up project.

For this reason, SCC initiation in austenitic stainless steels under simulated BWR conditions was investigated by constant extension rate (CERT) and constant load tests with notched tensile and fracture mechanics specimens in autoclaves. After testing and optimisation of the EN measurement setup in the autoclaves, the special emphasis was placed to the independent in-situ confirmation of SCC initiation by the use of the reversed direct current potential drop (DCPD) technique as a second online detection method. Because of the limited throwing power in low-conductive, high-purity water, the effect of distance between EN electrodes and the specimen (or cracking location) was also evaluated.
To support the complex and time-consuming investigations in high-temperature water, a PhD thesis in a simpler corrosion system (sensitised stainless steel in aqueous thiosulphate solution at room- or slightly increased temperature) was started and a corresponding room-temperature facility was constructed. In preparation of this PhD thesis some preliminary investigations were performed with this corrosion system at room-temperature. Since the same cracking mechanism and system parameter response is assumed to occur as in high-temperature water, it is expected that the gained insights can be transferred to BWR conditions at least in a qualitative way.
2.2 Materials and experimental procedure

2.2.1 Materials and specimens

For the SCC initiation tests, a rod material of the high-carbon austenitic stainless steel AISI 304 was used, since this steel could be easily sensitised and showed a sufficiently high susceptibility to intergranular (IG) SCC (Tables 1 and 2). This material was used either in the solution annealed state with a rather low degree of sensitisation (value from double-loop electrochemical potentiokinetic reactivation (EPR) tests ≤ 0.3 %; measured according to JIS G 0580-1986) and therefore with a low susceptibility to IG SCC, or it was sensitised. The applied heat treatment at 620 °C for 24 h resulted in a very high degree of sensitisation (EPR value ≈ 28 %). In a few cases specimens with an intermediate degree of sensitisation were used (620 °C for 1 or 9 h). The room-temperature and some of the high-temperature water tests were performed using round bar tensile specimens with a U-shaped notch in the centre of the gauge section (Figure 1a, gauge length = 36 mm, diameter = 6 mm, notch depth = 0.2 mm). In the tests with combined DCPD and EN measurements, 12.5 mm thick compact tension (0.5T CT) specimens with a machined notch (ρ = 0.1 mm) were used (Figure 1b). To investigate the effect of electrode distance to the specimen surface on the EN behaviour in high-temperature water, flat tensile specimens with a “gentle” notch on one side of the specimen were used (Figure 1c, gauge length = 18 mm, width/thickness = 3.5 mm, notch depth = 0.3 mm). All specimens were electrically insulated from the autoclave/test facility and from each other.

Table 1: Mechanical properties and heat treatments of the investigated material (WQ = water quenched, YS = yield strength).

<table>
<thead>
<tr>
<th>AISI</th>
<th>Design.</th>
<th>Type</th>
<th>Product form</th>
<th>YS&lt;sub&gt;air&lt;/sub&gt; 25°C [MPa]</th>
<th>UTS&lt;sub&gt;air&lt;/sub&gt; 25°C [MPa]</th>
<th>YS&lt;sub&gt;air&lt;/sub&gt; 288°C [MPa]</th>
<th>Solution annealing heat treatment</th>
<th>Sensitisation heat treatment</th>
<th>DL-EPR value</th>
</tr>
</thead>
<tbody>
<tr>
<td>304</td>
<td>304 C</td>
<td>High-carbon</td>
<td>Rod</td>
<td>291</td>
<td>686</td>
<td>175</td>
<td>1050°C/30min/WQ</td>
<td>--</td>
<td>&lt; 0.3 %</td>
</tr>
<tr>
<td>304</td>
<td>304 D</td>
<td>High-carbon</td>
<td>Rod</td>
<td>291</td>
<td>686</td>
<td>175</td>
<td>1050°C/30min/WQ</td>
<td>620°C/24h/WQ</td>
<td>≈ 28 %</td>
</tr>
</tbody>
</table>

Table 2: Chemical composition of the investigated material in wt.%.  

<table>
<thead>
<tr>
<th>AISI</th>
<th>Design.</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Mo</th>
<th>Ni</th>
<th>Co</th>
<th>Cu</th>
<th>Nb</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>304</td>
<td>304 C &amp; D</td>
<td>0.062</td>
<td>0.15</td>
<td>0.53</td>
<td>0.023</td>
<td>0.029</td>
<td>18.3</td>
<td>0.273</td>
<td>8.59</td>
<td>0.169</td>
<td>0.296</td>
<td>0.019</td>
<td>0.001</td>
</tr>
</tbody>
</table>
Figure 1: Round bar tensile specimen with U-shaped notch and schematic of the potential drop (PD) measurement locations (a); 12.5 mm thick compact tension specimen (0.5T C(T)) with machined notch and instrumented with DCPD and electrochemical potential noise measurement technique (b); flat tensile specimen with “gentle” notch on one side of the specimen (c).

2.3.2 Experimental procedure

2.3.2.1 Room-temperature experiments in aqueous thiosulphate solution

The SCC initiation tests at room-temperature were performed with a water loop at ambient pressure with an electro-mechanical tensile machine (Figure 2). 0.01 M Na₂S₂O₃ aqueous solution was pumped with a flow rate of ca. 1 l/h through a small test cell (Volcell = 0.2 l) mounted around the central section of a round bar tensile specimen. The EN measurement setup and a schematic of the test cell can be found in Section 2.3.2.3. Specimen and test cell were shielded by a Faraday cage. After about three days at a pre-load of 1 kN, the specimen was strained with a constant stroke rate of the pull-rod \( v_{\text{pull rod}} \) of \( 7.2 \times 10^{-9} \) m/s which corresponds to a nominal strain rate of \( 2 \times 10^{-7} \text{s}^{-1} \). After the tests the specimens were carefully examined in the scanning electron microscope (SEM) and the in-depth cracking and fracture mode (IG or transgranular (TG) SCC, ductile mechanical failure, etc.) were investigated after opening the specimen by mechanical overloading in air at room-temperature.
2.3.2.2 Experiments in a simulated BWR environment

Sophisticated, refreshing, high-temperature water loops with autoclave and integrated electro-mechanical loading system were used for the SCC initiation studies with EN measurements under simulated BWR conditions (Figure 3). This system was optimised for these kinds of measurement in many pre-tests. The tests were performed in oxygenated, high-temperature water (250 or 288 °C with 2 ppm dissolved oxygen (DO) content). The electrochemical corrosion potential (ECP) of the specimens and the redox potential of the environment (Pt-probe) were measured (during the pre-oxidation phase only) with a Cu/Cu₂O/ZrO₂-membrane reference electrode (RE). The ECP of the stainless steel was about +110 mV \text{SHE} (250 °C) or +160 mV \text{SHE} (288 °C) and the redox potential was +270 mV \text{SHE} (250 °C) or +300 mV \text{SHE} (288 °C), respectively. In most experiments, 50 ppb sulphate was added to the high-purity water to promote SCC in the stainless steel specimens. The specimens were pre-oxidised at a small pre-load for at least one week before constant pull rod stroke rates \( (v_{\text{pull rod}}) \) of \( 7.2 \times 10^{-9} \) m/s (round bar tensile specimens, nominal strain rate = \( 2 \times 10^{-7} \) s\(^{-1}\)), \( 2 \times 10^{-8} \) and \( 1 \times 10^{-8} \) m/s (C(T) specimens) or \( 3.6 \times 10^{-9} \) m/s (flat tensile specimens, nominal strain rate = \( 2 \times 10^{-7} \) s\(^{-1}\)) were applied. After SCC initiation and a certain amount of crack advance was detected by the EN measurement and/or DCPD, the control mode was switched from CERT to load control. The specimens were then either directly unloaded, or the load was kept constant until unloading at the end of the test. After finishing the experiments, the specimens were carefully analysed using a SEM. In most cases the specimens were opened by mechanical overloading (round bar and flat tensile specimens) or by fatigue (C(T) specimens) in air at room-temperature for analysis of the in-depth cracking and of the fracture mode (IG or TG SCC, ductile mechanical failure, etc.).
2.3.2.3 Electrochemical noise measurements

The electrochemical potential (EPN) and current noise (ECN) were recorded either by a GAMRY™ potentiostat (ESA400 software) with a sampling rate of 5 Hz, or by a newly developed EN measurement device (EcmNoise, IPS) with a sampling rate of 2 or 4 Hz. The EN measurement devices were qualified and characterised by suitable electronic circuits and “round-robin” tests (e.g., within the European Cooperative Group on Corrosion Monitoring of Nuclear Materials – ECG-COMON) [7]. The EN was recorded with a zero resistance ammeter (integrated into the GAMRY™ or EcmNoise equipment) between a C-ring, made out of the same steel as the specimen, and a round bar specimen (Figures 4a and 4b). The EPN was measured versus a Pt-wire (pseudo RE) which was aligned around the U-notch of a round bar specimen or inserted into the notch of a C(T) specimen and located as close as possible to the U-notch (Figures 4a and 4b) or notch-root (Figure 1b), respectively. In case of the tests with simultaneous use of EN and DCPD techniques, the ECN could not be measured (Figure 5). To investigate the effect of electrode distance to the specimen surface on the EN signal behaviour, two different facilities were used. During CERTs a “Contact Electric Resistance Autoclave” at VTT (Finland) was used to record EPN of a flat tensile specimen against a precisely adjustable Ir-tip (Figure 4c) and, in an autoclave at PSI, two EPN signals were measured simultaneously with two Pt-wire pseudo REs which were positioned in different distances to a flat tensile specimen (Figure 4d).
Figure 4: Schematic of the cell configuration and experimental setup for EN measurements with round bar tensile specimens in aqueous thiosulphate solution at room-temperature (a) and in simulated BWR water (b), and with flat tensile specimens in simulated BWR water in a “Contact Electric Resistance Autoclave” with adjustable RE (by a step-motor) (c) and with two REs at different distances from the specimen surface (d).

Figure 5: Schematic of the cell configuration and experimental setup for EPN and reversed DCPD measurements with C(T) specimens. The two techniques were combined in an autoclave.
Despite a pre-oxidation phase of more than one week before starting the loading phase in the high-temperature water experiments, the EPN (and partly also the ECN) signal was still drifting slowly; quasi-stationary ECPs and passive current density values on stainless steels are usually achieved only after several thousands of hours of exposure in high-temperature water. This slow change in ECP from oxide film growth has to be carefully considered in EN analysis. Therefore, in case of the high-temperature water tests, the DC part and the trend of the EPN (and ECN) signal were removed for evaluation of the data, resulting in a basic EPN (and ECN) level of around 0 V (and 0 A) when the loading phase started \((t = 0 \text{ h})\). In the room-temperature experiments (only) the EPN signal was treated in the same way before further analysis of the EN pattern.

### 2.3.2.4 Direct current potential drop measurements

In case of combined EPN and DCPD measurements, crack advance/initiation was continuously monitored, together with EPN, using the reversed DCPD method with a resolution limit of about 1 μm. Figure 5 shows a schematic of the DCPD measurement setup for the C(T) specimens. A current of 5 A was sent through the specimen (the polarity was reversed about every 3 s) and the potential drop was measured at the front face of the specimen. In the case of the round bar specimen a current of 5 A was also used and the potential drop was measured at four locations around the U-notch (Figure 1a).

The geometry change by plastic deformation of the round bar specimens and of the notch-root in the C(T) specimens during the CERT phase resulted in a change of the potential drop \(U\) before crack initiation occurred. A linear relationship between change in 'U' and increase of pull rod stroke 's' (or nominal strain \(\varepsilon\)) was observed during this phase, and this behaviour was also confirmed by Finite Element modelling (Figure 6). The potential drop change \(\partial U/\partial s\) was much higher (typically by a factor of 10) in the round bar specimens with plastic yielding of the whole cross section (SCC initiation was typically observed at nominal stresses slightly above the yield strength) than in the notched C(T) specimen, where plastic yielding was restricted to the vicinity of the notch-root (small scale yielding). The SCC initiation point (and potential drop at crack initiation 'U₀') during the CERT phase was assigned to the location where the U vs. s (or \(\varepsilon\)) curve started to deviate from the above mentioned linear relationship. The subsequent SCC crack growth was then calculated from \(U/U₀\) and the analytical Johnson equation [8] for the case of C(T) specimens, and by using our own calibration curve, which was derived by Finite Element modelling (Figure 6), for the round bar specimens. Although the DCPD has a higher sensitivity for crack growth \(\partial U/\partial a\) in the round bar than in the C(T) specimens, exact determination of the point of crack initiation was subjected to a larger degree of uncertainty because of the more pronounced plasticity effects.

**Figure 6:** Schematic of the course of the PD signal during CERT with point of crack initiation (left) and DCPD calibration curve calculated by Finite Element modelling (right).
2.4 Results and discussion

2.4.1 Room-temperature investigations in aqueous thiosulphate solution

To support the complex and time-consuming investigations in high-temperature water, a low pressure facility with a CERT unit for SCC experiments in aqueous thiosulphate solution at room-temperature was built [9]. Several CERTs with stainless steels with a different degree of sensitisation with EPN or simultaneous ECN and EPN measurements were performed in aqueous thiosulphate solution. Figure 7 shows the typical course of the ECN and EPN signals during a CERT with a severely sensitised stainless steel. Cathodic potential and anodic current transients were usually observed, which are related to oxide film rupture, anodic metal dissolution and repassivation events during crack initiation at different surface locations, as well as to surface crack growth of the resulting microcracks (slip-dissolution mechanism [10]). With further increasing strain the baseline current (ECN) increased and the baseline potential (EPN) decreased. This can be explained by the superposition of many transients caused by simultaneous multiple initiation of several cracks or fast (surface) crack growth. This is schematically described in Figure 8. Similar observations were also reported by Aballe et al. [11] and Watanabe et al. [12].

![Figure 7](image_url)

*Figure 7:* Typical current and potential transients and increase/decrease of the baseline ECN/EPN signal during a CERT with sensitised 304 stainless steel in aqueous thiosulphate solution (a). Shape of individual transients from (a) at higher resolution (b).
As it was not possible in the current project to verify these results with an independent method, further systematic testing was performed instead. CERTs were interrupted after the first appearance of relevant current and potential transients and the specimen was investigated in the SEM. An example is shown in Figure 9, where the test was finished after three small and one larger transients appeared. Careful post-test examination of the specimen in the SEM showed three small and one slightly longer crack with a total length of about 200 μm. This correlated very well with the four transients, which were observed. Further tests demonstrated that the detection of a single SCC initiation site with a crack area of about 100 times 40 μm by EN measurements is possible in this corrosion system.

To further verify the results described above, two experiments with solution annealed stainless steel were performed. The specimens were unloaded just after exceeding the nominal yield strength, where SCC was usually observed in the experiments with sensitised steel, but not with the less SCC susceptible solution annealed steel. As expected, neither transients (Figure 10) nor cracks (by SEM investigation) could be indentified. Only a very minor increase/decrease of the baseline current/potential signal appeared, which was probably caused by the plastic deformation of the specimen (see Section 2.3.2.4).
Figure 9: CERT with a sensitised 304 stainless steel in aqueous thiosulphate solution, which was interrupted after the first four relevant transients appeared in the EN signal (a). Four small cracks were found during post-test examination of the specimen in the SEM (b).

Figure 10: CERT with a solution annealed 304 stainless steel in aqueous thiosulphate solution, which was interrupted after reaching the nominal yield strength. No SCC cracks were found during post-test examination of the specimen in the SEM.

These results and the few investigations from literature clearly demonstrate that detection of SCC initiation in stainless steel in highly conductive, aqueous solution at room-temperature by the EN technique is possible. Nevertheless further systematic investigations are necessary to confirm these results, to optimise the measurement setup and to finally apply the EN technique as a SCC monitoring tool in the industrial field.

2.4.2 Investigations in a simulated BWR environment

In parallel to the room-temperature investigations, comparable experiments were performed in a simulated BWR environment in an autoclave. ECN and EPN were recorded during CERTs with stainless steel round bar tensile specimens using a similar setup as in the room-temperature investigations. Figure 11 shows a typical result of a CERT with the highly sensitised stainless steel. Around the nominal yield strength, where SCC initiation is expected, a clear increase of the baseline current and decrease of the baseline potential signal was observed. Additionally, some small anodic current and cathodic potential transients appeared. Post-test analysis in the SEM
revealed IG SCC in the notch of the specimen (Figure 11b). This behaviour of the EN signals was confirmed in many repeated experiments. The observed anodic (or cathodic) polarity and the shape of individual current (and potential) transients are probably related to oxide film rupture and repassivation events during crack initiation at different surface locations, as well as to surface crack growth of previously formed microcracks. The initiation process of IG SCC thus involves local anodic dissolution, which would be in line with a slip-dissolution mechanism [10] as stated before (Figure 8). The superposition of such current (and potential) signals from initiation events at different surface locations and the surface crack growth of microcracks under slow straining conditions with increasing plastic strain may result in a quasi-continuous increase in current (and drop of the potential). Furthermore, superimposed crevice currents and resulting potential changes to the differential aeration cell in the crack-mouth region, which vary with the crack-mouth opening, may further contribute to these signal changes. Individual transients therefore cannot be resolved in every case, in particular for large distances between specimen surface and counter or reference electrode in the high-purity water with extremely low conductivity.

Similar EN signals were observed for IG and TG SCC in sensitised and solution annealed stainless steels in high-temperature water, whereas in solution annealed materials the required strain for crack initiation and the subsequent crack growth rates were relevantly higher and lower, respectively (Figure 12). Based on these investigations, it was not possible to differentiate between IG and TG SCC, which indicates that similar electrochemical processes are involved in both cracking mechanisms.

![Figure 11: EPN and ECN in a CERT in high-temperature water with a sensitised stainless steel round bar specimen. The decreasing EPN and increasing ECN baseline signal as well as some transients are indicating SCC initiation and growth (a). The specimen revealed IG SCC (b).](image-url)
Figure 12: Comparison of the EPN during CERTs of a sensitised and solution annealed stainless steel with IG (b) and TG (c) SCC. The TG SCC in the solution annealed steel initiated at a relevantly higher strain and stress level and the resulting SCC crack growth rate was significantly lower than in the sensitised stainless steel (→ time to failure).

Similar to the room-temperature investigations, further CERTs were performed, which were interrupted after a relevant increase of the baseline current (and decrease of the potential). Then the specimens were carefully examined in the SEM. In Figure 13, a result is shown, where the specimen was unloaded briefly after an increase of the baseline current by ca. 100 nA (and decrease of the potential) was observed, thus indicating SCC initiation. Post-test analysis in the SEM revealed a few IGSCC surface cracks with a maximum depth of ca. 100 μm at the U-notch of the specimen (Figure 13b). The smallest crack, detected by EN measurements in high-temperature water so far, was one single, small, IG, semi-elliptical flaw with a surface crack length of about 150 μm and a maximum crack depth of about 120 μm, which corresponded to an increase of the current signal by ca. 50 nA.

Figure 13: EPN and ECN in a CERT with a sensitised stainless steel round bar specimen in high-temperature water, which was interrupted after a relevant increase of the current and decrease of the potential signal (a). A few surface cracks were found during post-test examination of the specimen in the SEM (b).
Experiments with solution annealed steel in high-temperature water, which were interrupted after exceeding the nominal yield strength, before SCC occurred (confirmed by post-test examinations in the SEM), showed only a minor increase/decrease of the current/potential signal (Figure 14). This increase/decrease was related to the plastic deformation of the specimen rather than SCC initiation (see Section 2.3.2.4).

![Figure 14: CERT with a solution annealed 304 stainless steel in high-temperature water, which was interrupted after exceeding the nominal yield strength. No SCC cracks were found during post-test examination of the specimen in the SEM.](image)

The investigations in simulated BWR environment showed that SCC could be detected by EN measurements at least under stable conditions in the lab. Although characteristic features in the EN could be clearly assigned to the onset of SCC by numerous systematic experiments, an unambiguous confirmation by an independent online method was still missing. Such investigations are described in the following Section. Another major issue of EN measurements in simulated BWR environment is the very low conductivity of the high-purity water, resulting in a limited throwing power of the current. Therefore the effect of electrode distance to the specimen surface was investigated by different experiments, described in Section 2.4.2.2.

### 2.4.2.1 Combined electrochemical noise and DCPD measurements

The DCPD method, which has been used for online crack growth measurements in C(T) specimens for many years, has been chosen as independent SCC monitoring tool to be simultaneously applied with the EN technique. The first combined EN and DCPD measurements in high-temperature water revealed that the DCPD technique interfered with the EPN signal. Figure 15 shows the disturbances in the EPN signal, which disappeared when the DCPD measurement was turned off. The steps in the EPN signal appeared synchronously with reversing the polarity of the current. This effect was probably caused by imperfect electrical insulation of the specimens and wires leading to very small leakage currents, which resulted in relevant ohmic potential drops in the low-conductivity electrolyte. Unfortunately the experimental setup could not be further improved and periodical interruptions of the DCPD measurements would have lead to a rather poor resolution limit for the detection of SCC initiation. A part of the information in the EN signal was thus lost due to these disturbances. To better see the global EPN signal trend the mean value of the EPN signal by averaging the EPN values over periods of 5 min was used in these kind of experiments. Individual potential (and current) transients, as they were observed during SCC initiation tests in high-temperature water are shown in Figure 15.

---

1. The term electrochemical potential noise (EPN) as used in this report also includes more “normal” measurement of electrochemical potential.
room-temperature and high-temperature water environment with notched round bar specimens without DCPD (Figures 7 and 11a), could therefore not be resolved by this approach, but the “global” change of the baseline potential was still visible.

**Figure 15:** EPN signal during the pre-oxidation phase of the first experiment with reversed DCPD turned on (left) and off (right). An interference of the DCPD to the EPN signal could be identified.

Figure 16 shows the result of the first SCC initiation test with combined EN and DCPD measurements using a heavily sensitised stainless steel C(T) specimen. After about eight hours of constant straining of the specimen, the EPN signal started to drop indicating SCC initiation. This was confirmed by the DCPD technique, which showed the onset of crack advance at the same time. Subsequent fractographic analysis by SEM revealed IG SCC initiation and growth (along the whole notch-root) in this specimen (Figure 16b). Due to the disturbances of the EPN signal, the expected individual potential transients in the EPN from single local film rupture/repassivation events, related to SCC initiation, could not be resolved. Investigations in aqueous thiosulphate solution at room-temperature, where a comparable cracking mechanism is believed to be active, revealed a similar behaviour of the EPN signal level if extensive SCC initiation occurred (see above). About three hours after SCC initiation was detected, the loading mode was switched to constant load and the crack continued to grow according to the DCPD signal (Figure 16a). Rather surprisingly at a first glance, the EPN signal rose back close to the original level. The reason for this is probably the absence of any further surface crack growth along the notch-root and the high-purity water electrolyte with a very low conductivity ($\kappa_{\text{inlet}}$ (at 25 °C) = 0.19 $\mu$S/cm). Once the SCC surface crack had spread out along the whole notch-root, the active crack-tip apparently grew in the depth direction only and moved away from the surface and RE. Thus the EN signal from the crack front accessible outside the crack became small, in good agreement with the limited throwing power of the current in the high-purity water. Therefore, only crack initiation and subsequent surface (or near-surface) crack growth appear to be detectable by EN measurements in low-conductivity electrolytes. Results from experiments in high-purity water with multiple REs investigating the effect of RE distance to the specimen surface (or “location of cracking”) seem to confirm this hypothesis (see below). After SCC initiation, the drop in the EPN signal (Figure 21a) and single potential transients (Figure 21b) could increasingly be resolved with decreasing distance to the specimen surface.
Figure 16: Mean values of the electrochemical potential noise (<EPN>) and crack advance measured by DCPD (a) in a combined constant extension rate (CERT)-constant load (CLT) test with a sharply notched, sensitised stainless steel C(T) specimen with IG SCC confirmed by post-test fractography in the SEM (b).

A second SCC initiation test was performed with sensitised (Figure 17) and solution annealed (Figure 18) stainless steel C(T) specimens at 288 °C and with a slower extension rate of the pull rod (v_{pull rod} = 10^{-8} m/s). This time, the test was interrupted shortly after switching to constant load. SCC initiation was detected in both specimens around 18 h after the start of the straining by a drop in the EPN signal and by crack advance recorded by the DCPD technique. In the sensitised stainless steel specimen, ca. 15 μm average crack advance over the whole specimen thickness was measured by the DCPD and the EPN signal decreased by ca. 5.5 mV, whereas in the solution annealed specimen a total crack advance of ca. 9 μm and a decrease of the EPN by ca. 3 mV were recorded. In both specimens, surface crack growth along the notch-root could be followed by the EN measurements. Post-test fractography revealed a few, localised, IG areas of SCC in the sensitised specimen, (Figure 17b) and very few, localised, TG areas of SCC in the solution annealed specimen (Figure 18b). The average extent of crack advance correlated reasonably well with the DCPD indications.

Figure 17: Mean values of the electrochemical potential noise (<EPN>) and crack advance measured by DCPD in a combined constant extension rate (CERT)-constant load (CLT) test with a sharply notched, sensitised stainless steel C(T) specimen (a) with localised, IG SCC confirmed by post-test fractography in the SEM (b).
Figure 18: Mean values of the electrochemical potential noise (<EPN>) and crack advance measured by DCPD in a combined constant extension rate (CERT)-constant load (CLT) test with a sharply notched, solution annealed stainless steel C(T) specimen (a) with very localised, TG SCC confirmed by post-test fractography in the SEM (b).

This second test also confirmed (see Section 2.4.2, Figure 12) that similar EN signals are observed for IG and TG SCC in sensitised and solution annealed stainless steels in high-temperature water, indicating that both cracking mechanisms involve similar electrochemical processes. Thus, differentiation between IG and TG SCC will hardly be possible based on EN measurements alone.

To further verify the very promising EN results gathered during many SCC initiation tests with round bar specimens (e.g., Figures 11 to 13), the EN and DCPD techniques were also applied in notched round bar specimens. The result of one of these tests is shown in Figure 19. After about 20 h of constant straining, the typical drop in the EPN signal was observed, whereas the DCPD signal indicated the onset of SCC some few hours later. The delayed indication of SCC might be explained by the more pronounced effect of plastic deformation on the DCPD signal in the round bar specimen (see Section 2.3.2.4) and by a greater sensitivity to small cracks for the EN technique.

Figure 19: Mean values of the electrochemical potential noise (<EPN>) and crack advance measured by DCPD during constant extension rate testing (CERT) with a notched, sensitised stainless steel round bar specimen (a) with IG SCC confirmed by post-test fractography in the SEM (b).
Additional SCC initiation tests with combined EN and DCPD measurements on C(T) and round bar specimens further confirmed the observations described above. These SCC initiation experiments with independent online crack growth monitoring by the DCPD technique thus confirmed the general ability of the EN technique to detect SCC initiation under simulated BWR conditions.

2.4.2.2 Effect of reference electrode distance on the electrochemical noise signal

The effect of the RE distance to the specimen surface (or cracking location) on the EPN signal behaviour in very low conductivity, high-temperature, high-purity water with very limited throwing power of the current was investigated with two different experimental setups which are briefly described in Section 2.3.2.3. With both setups, CERTs with notched flat tensile specimens were performed in a simulated BWR environment.

A “Contact Electric Resistance Autoclave” with electro-mechanical tensile machine (at VTT, Finland) was used to adjust the pseudo RE (Ir-tip) at different distances to the specimen surface by an electric step-motor. Because of several experimental problems (e.g., with temperature stability) only the two last experiments could be regarded as reliable and no definitive conclusion could be drawn with respect to the RE position aspect. Nevertheless, one test with a RE distance of 1 mm revealed cathodic transients in the EPN signal before the “global” drop of the baseline potential signal (Figure 20a). These transients were comparable to the ones observed in the room-temperature tests in aqueous thiosulphate solution (Figure 20b).

![Figure 20: EPN of a CERT in high-temperature water with a RE distance of 1 mm to the sensitised stainless steel flat tensile specimen surface (a). Comparable potential transients during a CERT with a sensitised stainless steel in thiosulphate solution at room-temperature (EPNRT) and oxygenated high-temperature water (EPNHT, detail from (a)) (b).](image)

To further investigate the effect of electrode distance, tests with multiple REs at different distances from the specimen surface were performed with notched flat tensile specimens. Figure 21 shows the result of such an experiment. The potential drop and amplitude of potential transients increased with decreasing distance to the specimen surface. Very similar potential transients as in the room-temperature tests at higher conductivities could be increasingly resolved with decreasing distance from the specimen surface (Figure 21b). There is a lower limit for this distance, where a different local water chemistry may arise because the evolution of an occluded chemistry in the crevice between the RE and specimen.
Figure 21: Effect of distance between RE and specimen surface on the EPN during a CERT with a flat tensile specimen and two pseudo REs (Pt-wires) (a). Example of an individual potential transient from (a) at higher resolution (b).

These experiments confirmed that for EN measurements the electrode distance plays an important role in low conductivity electrolytes. The detection sensitivity increases with decreasing distance between specimen surface and reference (or counter) electrode. It is concluded that with suitable positioning of the reference (or counter) electrode, early detection of SCC initiation by the EN technique is possible under stable and stationary lab conditions in oxygenated high-temperature water.
2.5 Summary and conclusions

The potential and limits of the EN technique for the early detection of SCC initiation in austenitic stainless steel under simulated BWR conditions was investigated by CERTs and constant load tests with notched tensile and fracture mechanics specimens in autoclaves. Similar experiments were also performed in aqueous thiosulphate solution at room-temperature.

The SCC initiation experiments with independent online crack growth monitoring by DCPD indicated that early SCC detection by EN is possible under stable and stationary lab conditions in oxygenated high-temperature water. Individual small IG semi-elliptical surface flaws with a surface crack length and crack depth of about 150 μm may be detected by EN measurements. But on the other hand, only crack initiation and the subsequent surface or near surface growth may be detected by EN measurements in high-purity water with low conductivity. Furthermore, a short distance between the specimen surface and RE is crucial to achieve a high sensitivity.

EN transients, indicating SCC initiation, could be observed during experiments in aqueous thiosulphate solution at room-temperature. These transients were comparable to the ones measured in high-temperature water with a short distance of the reference or counter electrode to the specimen surface.

The observed polarity of the potential and current signal changes/shifts and shapes of individual EN transients during SCC initiation suggest a mechanism, which involves film rupture, local anodic dissolution and repassivation according to the slip-dissolution mechanism. Similar EN signals were observed for IG and TG SCC in sensitised and solution annealed stainless steels, which indicates that similar electrochemical processes are involved in both cracking mechanisms.

It is recommended to further evaluate this promising technique in long-term tests under static load and to include further important and plant relevant parameters (e.g., dissolved hydrogen content, temperature, flow rate, cold-work), which were not explored so far.
2.6 References


Part III

Environmentally-assisted cracking behaviour in the transition region of an Alloy 182-SA 508 Cl.2 dissimilar metal weld joint in simulated BWR/NWC environment

Hans-Peter Seifert, Stefan Ritter, T. Shoji\textsuperscript{1}, Q.J. Peng\textsuperscript{1}, Y. Takeda\textsuperscript{1} and Z.P. Lu\textsuperscript{1}

\textsuperscript{1}Fracture and Reliability Research Institute (FRI), Tohoku University, Sendai, Japan
# TABLE OF CONTENTS

3.1 INTRODUCTION...................................................................................................................... 75

3.2 MATERIALS AND EXPERIMENTAL PROCEDURE................................................................. 77
   3.2.1 Test weld and specimens.............................................................................................. 77
   3.2.2 Experimental procedure .......................................................................................... 80
      3.2.2.1 Test facilities ....................................................................................................... 80
      3.2.2.2 Crack length monitoring and post-test evaluations .............................................. 80
      3.2.2.3 Environmental conditions .................................................................................. 81
      3.2.2.4 Loading conditions ............................................................................................ 82

3.3 RESULTS AND DISCUSSION................................................................................................. 83
   3.3.1 SCC crack growth behaviour perpendicular to the fusion line .................................. 83
   3.3.2 SCC initiation and crack growth behaviour parallel to the fusion line..................... 90
   3.3.3 SCC behaviour of the Alloy 182 bulk weld metal ..................................................... 95
   3.3.4 Corrosion fatigue behaviour of the Alloy 182 bulk weld metal .................................. 96

3.4 SUMMARY AND CONCLUSIONS......................................................................................... 97

3.5 APPENDIX ................................................................................................................................. 98
   3.5.1 Metallurgical and microstructural aspects of SCC in Alloy 182 weld metal .......... 98
   3.5.2 Metallurgical and microstructural aspects of SCC in the interface region
       between the Alloy 182 weld and LAS base metal ....................................................... 100

3.6 REFERENCES......................................................................................................................... 101
3.1 Introduction

Alloy 182 has been widely used as a weld filler metal to join the low-alloy steel (LAS) reactor pressure vessel (RPV) and pressure vessel nozzles to both wrought Nickel-base alloys (Alloy 600) and austenitic stainless steel (SS, AISI 304L, 316L, 316NG) components in light water reactors by manual shielded metal arc welding. In recent years, incidents of stress corrosion cracking (SCC) in Alloy 182 in both boiling (BWRs) [1, 2] and pressurised water reactors (PWRs) [2, 3] were reported in some countries. In the case of BWRs, components such as different nozzle safe ends [1], bottom head penetration housings [4] and core shroud support welds [5] have suffered from SCC (Figure 1). The cracking was interdendritic/intergranular (IG)¹ and usually confined to the weld metal. None of the SCC cracks significantly penetrated the adjacent RPV base metal, which is consistent with the very high SCC resistance of LASs under light water reactor conditions [6, 7]. Even after post-weld heat treatment (PWHT) significant residual stress is observed in such weld joints (see Appendix in Section 3.5.1) and, in particular in the context of repair welding, and has played a major role in many of these SCC incidents. Cold-work from surface grinding or weld shrinkage etc. also potentially affected the SCC initiation susceptibility. The weld shrinkage strain/residual stress profile and the solidification structure (grain/dendrite boundary chemistry and misorientation) were important factors controlling the SCC crack path and propagation in weld joints [8-12].

Boiling Water Reactors

Pressurized Water Reactors

Figure 1: Different examples of SCC cracking incidents in BWRs and PWRs.

¹ The SCC cracks usually grow along the high angle solidification grain boundaries (see Appendix). Because of the dendritic appearance of the corresponding fracture surface, this kind of cracking is called interdendritic in the following Sections although IG would be a more appropriate term.
The high SCC susceptibility of uncreviced bulk Alloy 182 weld material was also confirmed by laboratory investigations for both simulated BWR [8, 13-15] and primary PWR coolant environment [16-20]. The SCC crack growth behaviour of Alloy 182 weld material and parametric effects (for the effect of electrochemical corrosion potential (ECP), sulphate/chloride, cold-work, sensitisation, stress intensity factor, etc.) are quite similar to those observed in wrought Alloy 600 or SS, although, the higher high-temperature yield strength of Alloy 182 may make it more susceptible to SCC and low-temperature rapid crack propagation, and produce high weld residual stresses and stress intensity factors [8]. The different grain boundary chemistry of the weld metal may further increase its SCC susceptibility [8, 9].

In contrast to the relatively high susceptibility of Alloy 182 to SCC growth, lab investigations revealed that LAS and their heat-affected zones (HAZs) have a very high resistance to SCC growth in chloride-free BWR environment at temperatures of 270 to 290 °C up to very high stress intensity factor levels of 50 to 70 MPa·m\(^{1/2}\) (if excessive hardness > 350 HV5 is avoided by proper welding and PWHT parameters) [6, 7]. It is stressed that for some very specific combinations of Alloy 182 dissimilar metal weld and cracking configurations, the residual stress can shift the stress intensity factors in the interface region to such high levels that accelerated sustained SCC in the adjacent LAS in high-purity BWR/normal water chemistry (NWC) environment cannot be fully excluded, in particular in high-sulphur LAS. These situations thus also reflect a certain safety concern.

Although many investigations on the SCC behaviour of Alloy 182 bulk weld metal [8-20] and LAS [6, 7] have been performed, little attention has been paid to the SCC behaviour in the interface region between Alloy 182 and LAS, which has a complex microstructure with a potentially increased SCC susceptibility (see Appendix in Section 3.5). Most SCC investigations in Alloy 182 were related to the bulk weld metal [8-20] and to “homogeneous” specimens from simplified weld configurations or weld overlays and hardly involved the interface region between Alloy 182 and LAS. The environmentally-assisted cracking (EAC) susceptibility of dissimilar welds has been screened by slow strain rate tests with smooth tensile specimens [21-25]. It is not surprising that EAC was observed in all regions under adequate testing conditions: Transgranular (TG) strain-induced corrosion cracking [7], e.g., occurred in the LAS in oxygenated high-temperature water and its severity was strongly dependent on the steel sulphur content. As expected, IG/interdendritic SCC was observed in Alloy 182 weld metal in both BWR and PWR water chemistries. In several cases interface cracking along the fusion line, often in conjunction with high hardness of the interface region, has been reported, thus indicating a distinct SCC susceptibility of this region. Therefore, it can be anticipated that the SCC in the transition region plays an important role in the overall SCC behaviour in the dissimilar weld joint.

In order to improve the understanding of the SCC behaviour in the transition region, in particular the SCC behaviour at the fusion line perpendicular and parallel to the fusion boundary, the SCC crack growth behaviour in the transition region of a simulated Alloy 182-SA 508 Cl.2 weld joint was investigated under simulated BWR/NWC and PWR conditions within the framework of a collaboration between the Fracture and Reliability Research Institute (FRI) of Tohoku University and PSI. A special emphasis was placed to the question whether a fast growing interdendritic SCC crack in the highly susceptible Alloy 182 weld metal can easily cross the fusion line and significantly propagate into the adjacent low-alloy RPV steel or if crack arrest or interface cracking do occur at or along the fusion boundary. The following part of the report summarises the most important BWR/NWC results of this joint project. In the Appendix (Section 3.5) some important metallurgical and microstructural aspects of SCC in Alloy 182 weld metal and the interface region between the Alloy 182 weld metal and adjacent LAS are briefly summarised for the reader who is not familiar with this topic.
3.2 Materials and experimental procedure

The EAC crack growth behaviour under cyclic and static loading conditions in the interface region between the Alloy 182 bulk weld metal and the unaffected RPV base material of a simulated dissimilar metal weld joint was investigated in oxygenated high-temperature water at 288 °C. The EAC crack initiation and growth behaviour was thereby studied parallel and perpendicular to the fusion line in different important microstructure regions, including the bulk and dilution zone of the Alloy 182 weld metal, the fusion line region, as well as the HAZ and unaffected LAS base metal.

3.2.1 Test weld and specimens

The Alloy 182-LAS part of a dissimilar metal weld was simulated by filling a rectangular groove in a forged quenched and tempered SA 508 Cl.2 RPV steel plate by multipass shielded metal arc welding (140 to 180 A and 22 to 26 V, interpass temperature < 177 °C) with Alloy 182 (Figure 2). After welding, a PWHT at 620 °C for 24.5 h was performed, which resulted in a sensitised microstructure with massive chromium carbide precipitation on the high-angle solidification grain boundaries in the weld metal. The chemical composition and tensile test properties of the Alloy 182 weld metal and LAS base metal are summarised in Tables 1 and 2. Typical chemical composition and micro hardness profiles perpendicular to the fusion line are shown in Figures 3 and 4.

![Figure 1: Schematic of simulated Alloy 182-SA 508 Cl.2 dissimilar test weld with welding direction of individual weld beads and the orientation of the primary dendrite growth axis in the bulk weld metal and in the fusion line region. The orientation of the MnS-inclusions and the main forming direction L of the forged LAS plate are also shown.](image-url)
Table 1: Chemical composition in wt.% of weld filler and low-alloy RPV steel of the dissimilar weld joint.

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Ni</th>
<th>Cr</th>
<th>Fe</th>
<th>P</th>
<th>S</th>
<th>Mn</th>
<th>Mo</th>
<th>Si</th>
<th>Ti</th>
<th>Nb+Ta</th>
<th>Al</th>
<th>Co</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>Alloy 182 Electrode</td>
<td>0.040</td>
<td>70.8</td>
<td>14.26</td>
<td>9.7</td>
<td>0.002</td>
<td>0.003</td>
<td>2.46</td>
<td>-</td>
<td>0.24</td>
<td>0.050</td>
<td>1.87</td>
<td>-</td>
<td>-</td>
<td>0.010</td>
</tr>
<tr>
<td>Alloy 182 PSI analysis</td>
<td>0.043</td>
<td>71.6</td>
<td>15.10</td>
<td>8.3</td>
<td>0.002</td>
<td>0.007</td>
<td>0.014</td>
<td>2.85</td>
<td>0.45</td>
<td>0.20</td>
<td>0.027</td>
<td>2.19</td>
<td>0.040</td>
<td>0.010</td>
</tr>
<tr>
<td>SA 508 Cl.2 PSI analysis</td>
<td>0.200</td>
<td>0.91</td>
<td>0.45</td>
<td>96.2</td>
<td>0.010</td>
<td>0.016</td>
<td>0.93</td>
<td>0.59</td>
<td>0.12</td>
<td>-</td>
<td>-</td>
<td>0.009</td>
<td>0.012</td>
<td>0.026</td>
</tr>
</tbody>
</table>

Table 2: Mechanical tensile test properties of the Alloy 182 bulk weld metal and the unaffected low-alloy RPV steel (YS = yield strength, UTS = ultimate tensile strength, A5 = uniform elongation, Z = reduction of area).

<table>
<thead>
<tr>
<th>Material</th>
<th>T [°C]</th>
<th>YS [MPa]</th>
<th>UTS [MPa]</th>
<th>A5 [%]</th>
<th>Z [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Alloy 182</td>
<td>288</td>
<td>345</td>
<td>547</td>
<td>28</td>
<td>50</td>
</tr>
<tr>
<td>SA 508 Cl.2</td>
<td>25</td>
<td>511</td>
<td>645</td>
<td>15.5</td>
<td>70.4</td>
</tr>
<tr>
<td>SA 508 Cl.2</td>
<td>288</td>
<td>440</td>
<td>608</td>
<td>12.4</td>
<td>70.5</td>
</tr>
</tbody>
</table>

Figure 3: Typical chemical composition profile across the fusion line.

Figure 4: Typical micro hardness profiles across the fusion line.
The bulk Alloy 182 weld metal revealed a yield strength at 288 °C of 345 MPa and a SCC resistance index SCRI value of 21.3 (see Appendix in Section 3.5). The sulphur content of the weld metal varied between 0.007 and 0.014 wt.%. The un-affected RPV base metal had a yield strength of 440 MPa and revealed a high sulphur content of 0.016 wt.% and a medium susceptibility to dynamic strain ageing. The dilution zone in the Alloy 182 weld metal close to the fusion line with a different chemical composition than the bulk weld metal had a thickness of 1.5 to 2.5 mm and a similar microstructure as the bulk weld metal. A peak micro Vickers hardness in the range of 280 to 320 HV0.5 was usually observed in the region of the fusion line. At some few locations, a very high peak hardness of up to 470 HV0.5 was observed in the dilution zone of the Alloy 182 weld metal in the absolute vicinity of the fusion line, which had a carbide or martensite-like appearance. The thickness of the weld HAZ in the LAS varied between 2 and 4.5 mm and the peak hardness in the LAS HAZ was usually close to the fusion line and below 300 HV0.5.

Two 25 mm and fourteen 12.5mm thick compact tension C(T) specimens with either a TS or ST orientation according to ASTM E399 [26] were cut from different locations in the test weld joint (Figure 5 and Table 3). This involved 12 heterogeneous (Alloy 182-LAS) and four homogenous (bulk weld metal) specimens. The specimens either had a sharp (notch radius \( \rho = 0.1 \) mm) or blunt notch (\( \rho = 1.5 \) mm). Four specimens were additionally fatigue pre-cracked in air at room-temperature with a load ratio of 0.1 or 0.2 and final \( K_{I,\text{max}} \) of 12 to 20 MPa\( \cdot m^{1/2} \).

![Figure 5: Schematic of location and orientation of the specimens in the test weld.](image)

The crack plane in the Alloy 182 weld metal was either perpendicular or parallel to the fusion line and parallel or perpendicular to the solidification direction (primary dendrite growth axis), respectively. The fatigue pre-crack-tip was either located in the bulk weld metal or in the dilution zone of the Alloy 182 at different distances from the fusion line. In the notched specimens, the notch-tip was either located at the fusion line or in the dilution zone. In all heterogeneous specimens, the MnS-inclusions were parallel to the crack front and crack plane (Figures 2 and 5).
**Table 3:** Overview on investigated heterogeneous dissimilar metal weld specimens (DLZ = dilution zone of Alloy 182 weld metal, BWM = Alloy 182 bulk weld metal, LAS HAZ = heat-affected zone of low-alloy steel, FL = fusion line, SD = solidification direction = primary dendrite growth axis).

<table>
<thead>
<tr>
<th>Specimen designation</th>
<th>Type of specimen</th>
<th>Type of notch</th>
<th>Location of notch ground or pre-crack-tip</th>
<th>Orientation of crack plane and growth direction with respect to FL and SD</th>
</tr>
</thead>
<tbody>
<tr>
<td>MN TS 1</td>
<td>0.5T C(T)</td>
<td>Air fatigue pre-crack</td>
<td>DLZ at 0.5 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN TS 2</td>
<td>0.5T C(T)</td>
<td>Air fatigue pre-crack</td>
<td>DLZ at 0.6 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN ST 3</td>
<td>0.5T C(T)</td>
<td>Sharp notch, ρ = 0.1 mm</td>
<td>DLZ at 0.8 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN TS 5</td>
<td>0.5T C(T)</td>
<td>Sharp notch, ρ = 0.1 mm</td>
<td>LAS HAZ</td>
<td>II to FL</td>
</tr>
<tr>
<td>MN ST 6</td>
<td>0.5T BNC(T)</td>
<td>Blunt notch, ρ = 1.5 mm</td>
<td>FL with specimen mid-plane at fusion interface</td>
<td>II to FL and ⊥ to SD</td>
</tr>
<tr>
<td>MN ST 7</td>
<td>0.5T C(T)</td>
<td>Sharp notch, ρ = 0.1 mm</td>
<td>FL with specimen mid-plane at fusion interface</td>
<td>II to FL and ⊥ to SD</td>
</tr>
<tr>
<td>MN ST 10</td>
<td>0.5T C(T)</td>
<td>Sharp notch, ρ = 0.1 mm</td>
<td>DLZ at 0.9 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN ST 11</td>
<td>0.5T C(T)</td>
<td>Sharp notch, ρ = 0.1 mm</td>
<td>DLZ at 0.9 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN TS 12</td>
<td>0.5T C(T)</td>
<td>Sharp notch, ρ = 0.1 mm</td>
<td>DLZ at 0.9 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN TS 13</td>
<td>0.5T C(T)</td>
<td>Sharp notch, ρ = 0.1 mm</td>
<td>DLZ at 1.5 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN TS 15</td>
<td>1T C(T)</td>
<td>Air fatigue pre-crack</td>
<td>DLZ at 1.5 mm from FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
<tr>
<td>MN TS 16</td>
<td>1T C(T)</td>
<td>Air fatigue pre-crack</td>
<td>BWM at 3 mm of FL</td>
<td>⊥ to FL and II to SD</td>
</tr>
</tbody>
</table>

### 3.2.2 Experimental procedure

#### 3.2.2.1 Test facilities

The EAC tests at PSI and FRI were performed in SS autoclaves with integrated electromechanical loading systems, which were attached to sophisticated refreshing high-temperature water loops (Figure 6). During the experiments all important mechanical (load, pull-rod stroke) and environmental parameters at inlet and outlet (dissolved oxygen (DO), conductivity, temperature, pressure, flow) were recorded continuously. The C(T) specimens were electrically isolated from the autoclave using zirconia or oxidised zirconium alloy sleeves/spacers. The ECP of the specimens and the redox potential (platinum probe) were continuously monitored with Cu/Cu$_2$O/ZrO$_2$-membrane or external Ag/AgCl/0.01 m KCl reference electrodes. Ionic impurities of the water (inlet and outlet) were analysed by Inductive Coupled Plasma – Atomic Emission Spectroscopy (ICP – AES) and Ion Chromatography (IC) once or twice during each test.

#### 3.2.2.2 Crack length monitoring and post-test evaluations

In all but one test, crack advance was continuously monitored using the reversed direct current potential drop method (DCPD) with a resolution limit of about 1 μm in homogenous specimens. In one experiment at FRI, the alternate current potential drop technique was used. In case of the DCPD, calibration curves were developed by a detailed two-dimensional Finite Element modelling of the DC potential field for the specific bimetallic weld joint specimen/DCPD system configuration [27], since the Johnson formula would significantly underestimate the real crack growth increment in the dissimilar metal welds. The DCPD crack growth results were linearly corrected with the average EAC crack increment measured on the fracture surface at 50 equidistant locations. In spite of the Finite Element calibration of the DCPD, the heterogeneous specimens involved correction factors of 10 to 50 %, which were related to the wavy shape of the fusion line,
uneven crack fronts, out-of-mid-plane cracking at the fusion line and in particular to the very high roughness of the fracture surfaces.

After the tests, the specimens were broken apart by fatigue in air for post-test evaluation. The fracture surface, cracking mode and position of the crack-tip were analysed by optical and scanning electron microscope (SEM) as well as with energy-dispersive X-ray microanalysis (EDX). The fatigue pre-crack/final crack length and EAC growth increments were measured on the fracture surface by optical microscope or SEM. For detailed fractographical analysis of the LAS part, the oxide film on the fracture surface of one specimen half was removed by galvanostatic reduction in an ENDOX-bath [28].

Cross sections perpendicular to the crack plane were cut from the specimens for micro-hardness measurements and detailed metallographical characterisation (microstructure, inclusions, crack-path, branching and secondary cracking, etc.) by optical microscope and SEM/EDX. The cross sections were polished and chemically and electrochemically etched by several different procedures to reveal the different microstructural features in the different regions of the weld joint.

Further metallographic specimens were directly cut from the weld joint for additional analyses of the microstructure, element-composition and micro-hardness across the fusion line.

Figure 6: Schematic of the refreshing water loop with autoclave and integrated electro-mechanical tensile machine.

3.2.2.3 Environmental conditions

The EAC tests were performed in oxygenated high-purity high-temperature water at 288 °C under low-flow conditions (4 to 8 autoclave exchanges per hour, local flow rate of some few mm/s). In several cases chloride (50 ppb Cl⁻ as NaCl) or sulphate (30 ppb SO₄²⁻ as H₂SO₄ or Na₂SO₄) was added to reduce the crack initiation time or to accelerate crack growth. A DO content of 0.2 and 0.4 ppm or an increased level of 2 and 8 ppm were used, respectively. The lower DO levels represent a realistic simulation of the total oxidant concentration in the reactor water, whereas the increased DO contents were used to simulate a realistic ECP at Alloy 182 attachment welds within
the RPV. The ECP of the heterogeneous specimens of +50 to +70 mV \text{SHE} (0.2 or 0.4 ppm DO) and +130 to +170 mV \text{SHE} (2 or 8 ppm DO) were between the typical values observed in homogeneous LAS and Ni-base alloys at PSI. The corresponding redox potentials of the environment were +250 to +270 mV \text{SHE} (0.2 or 0.4 ppm DO) and +290 to +300 mV \text{SHE} (2 or 8 ppm DO).

### 3.2.2.4 Loading conditions

Before applying the different loading sequences, the specimens were usually pre-oxidised in the test environment at a small constant pre-load of 0.5 kN ($K_I \leq 2$ MPa-m$^{1/2}$ in case of pre-cracked specimens) for at least one week in order to achieve stable environmental and electrochemical conditions.

In case of pre-cracked specimens with the pre-crack-tip in the bulk or dilution zone of the Alloy 182 weld metal, the following complex initial loading procedure was first applied to achieve a complete transdendritic $\rightarrow$ interdendritic transitioning along the whole crack front and to evolve a plastic zone, which is characteristic for a growing SCC crack:

1. Cyclic loading, stepwise increase of $R$ from 0.3 to 0.7 at 0.01 to 0.036 Hz,
2. cyclic loading at $R = 0.7$ with stepwise decrease of $\nu$ from 0.02 to 0.001 Hz,
3. periodical partial unloading (PPU) at $R = 0.7$, $\Delta t_{\text{hold}} = 9000$ s, $\Delta t_{\text{rise}} = \Delta t_{\text{decline}} = 500$ s,
4. constant load or near constant load (PPU with $\Delta t_{\text{hold}} = 24$ h) with a target $K_I$ or $K_{I,\text{max}}$ value of 25 to 35 MPa-m$^{1/2}$.

There were some slight variations in loading step 1 and 2 in some experiments, but this did not affect the crack growth results in the subsequent loading phases. After this initial loading procedure, constant or PPU loading was kept during the whole remaining test period until unloading in some experiments. In some other tests, this phase was followed by several sequences of cyclic, PPU and static loading to study the crack growth behaviour in different microstructure regions of the weld joint. In order to reduce the total testing time, the applied stress intensity factor was sometimes increased stepwise.

In case of experiments, which started with a sharp or blunt notch, the cyclic loading started with a low load ratio of 0.065 and a frequency of 0.036 Hz. The subsequent loading phases were identical to those described above.

The $K_{I,\text{max}}$-values were varied between 20 to 60 MPa-m$^{1/2}$. With few exceptions, these values were below the ASTM E647 limit [29] in case of cyclic loading and the ASTM E1681 limits [30] in case of PPU or constant loading conditions.

The SCC crack growth results were then compared with a Swedish disposition line (DL) for Alloy 182 [31] and the EPRI BWRVIP-60 SCC DL for LAS [32]. The corrosion fatigue crack initiation results were compared to the ASME III mean curve for austenitic alloys and LAS [33-35]. For this reason a pseudo elastic notch stress range $\Delta \sigma = 2S_a = \Delta K/\sqrt{\rho}$ was calculated for comparison purposes [36]. The corrosion fatigue crack growth behaviour was analysed in the time domain and compared to a PSI BWR/NWC curve for LAS [7, 37] and a curve developed by Argonne National Laboratory (ANL) for wrought Alloy 600 [39]. The corresponding air fatigue crack growth rates in the LAS and Alloy 182 were estimated based on [38] and [39], respectively.
3.3 Results and discussion

In the following paragraphs, the main observations of these complex investigations are summarised and illustrated by some selected test results. In the first two Sections, the SCC and EAC crack growth behaviour perpendicular and parallel to the fusion line is discussed. This is followed by a short summary of the SCC crack growth in Alloy 182 bulk weld metal. Finally, the corrosion fatigue initiation and growth behaviour in the Alloy 182 weld metal and at the fusion line is briefly outlined.

3.3.1 SCC crack growth behaviour perpendicular to the fusion line

The SCC behaviour perpendicular to the fusion line was investigated under PPU or constant load with fatigue pre-cracked 25 and 12.5 mm thick C(T) specimens at PSI and FRI in high-purity water and water with 30 ppb of sulphate or 50 ppb of chloride. The fatigue pre-crack-tips in the different specimens were located either in the bulk or the dilution zone of the Alloy 182 weld metal at a mean distance between 3 and 0.5 mm from the fusion line.

As expected, fast interdendritic SCC was observed in the Alloy 182 bulk weld metal parallel to the solidification direction in oxygenated, high-purity high-temperature water in the typical range of other investigations [8, 14, 40] (Figure 7) and slightly below the Swedish SCC DL for Alloy 182. The subsequent SCC crack growth rates in the Alloy 182 dilution zone were similar to those of the bulk weld metal and tended to slightly decrease with decreasing distance to the fusion line (Figure 8). The corresponding SCC crack growth rates in oxygenated high-temperature water with 50 ppb Cl⁻ or 30 ppb SO₄²⁻ were about a factor of 10 and 5 higher than in high-purity water and slightly higher than the Swedish SCC DL for Alloy 182 (Figure 27 in Section 3.3.2).

![Figure 7](image_url)

**Figure 7:** Similar SCC crack growth rates in the bulk and dilution zone of the Alloy 182 weld metal in the heterogeneous dissimilar metal weld specimens as in other comparable investigations with homogeneous Alloy 182 specimens under BWR/NWC conditions [8, 14, 40].
Cessation of SCC was observed in high-purity or sulphate-containing water in all tests under PPU or constant load conditions for those parts of the crack front, which reached the fusion line. As soon as the overwhelming part of the crack front has reached the fusion line, a drastic drop of interdendritic SCC crack growth rates (Figures 9 and 10) was observed even under aggressive environmental conditions (8 ppm DO and 30 ppb SO$_4^{2-}$) and crack arrest occurred at or close to the fusion line (Figures 10b, 11 and 12). Crack arrest at the fusion line was confirmed in one test with PPU, which was running for further 800 h after the DCPD had indicated cessation of crack growth close to the fusion line (i.e., da/dt$_{SCC} < 1 \times 10^{-11}$ m/s). In some specimens, minor crack growth (< 60 $\mu$m) into the HAZ of the RPV steels with subsequent crack arrest was observed at some few, very localised regions along the crack front. On the other hand, at several locations along the crack front crack branching and deflection of EAC crack growth (up to some few 100 $\mu$m) along the fusion line in the dilution zone of Alloy 182 in the direct vicinity of the fusion boundary was a quite common observation in most specimens, which indicates an increased SCC susceptibility of this region. It is stressed that this kind of interfacial cracking may be strongly favoured by the C(T) specimen, since the bending of the load arm results in a significant amount of tensile stress perpendicular to the interface (Figure 14) and to a relevant amount of mode I stress intensity factor, if a crack is growing perpendicular to the mid-plane.

**Figure 8:** Similar SCC crack growth rates in the Alloy 182 dilution zone and bulk weld metal of the weld joint.
Figure 9: Cessation of SCC crack growth at the fusion boundary under PPU conditions in specimens MN TS 1, MN TS 15 and MN TS 16. Comparison of the SCC crack growth rates in the interface region with a Swedish DL for Alloy 182 weld metals [31] and with the BWRVIP-60 SCC DLs for LAS [32].

Figure 10: Cessation of SCC under PPU in the Alloy 182-fusion line region in specimen MN TS 1 indicated by DCPD (a). In spite of the fact that the PPU was sustained for further 800 h after indication of crack arrest by DCPD, the SCC crack did not enter to the adjacent LAS as independently confirmed by post-test fractography in the SEM/EDX (b).
Figure 11: Confirmation of cessation of SCC at (a) or close to the fusion boundary (b) by etched cross sections of specimens MN TS 1 and MN TS 2. The “Nital” etching only attacks the LAS, but not the Ni-base weld metal.

Figure 12: Confirmation of cessation of interdendritic SCC in the fusion line region by post-test fractography of specimen MN TS 2 in the stereo light microscope and SEM.

After cessation of EAC crack growth under PPU at the fusion boundary in high-temperature water with 30 ppb of sulphate, the crack was forced to cross the fusion line by severe cyclic loading under aggressive environmental conditions in one specimen. Even in case of severe cyclic loading, the crack had problems to cross the fusion line and rather tended to grow in the Alloy 182 dilution zone along the fusion line before it finally entered into the RPV steel (Figure 13a). This again
indicates an increased EAC susceptibility of the interface region parallel and close to the fusion line in the dilution zone of the weld metal. After crossing of the fusion line under severe cyclic loading conditions, the EAC crack growth rates under PPU in the LAS HAZ were more than one order of magnitude lower than in the Alloy 182 bulk weld material. Suddenly after switching from PPU to constant load, cessation of SCC and crack arrest occurred in both the LAS HAZ and bulk LAS in spite of the high applied stress intensity factors and aggressive environmental conditions (Figure 15). The EAC crack growth in the un-affected LAS thus only grew under cyclic load (or during the slow rising load phase of the PPU) by corrosion fatigue (or strain-induced corrosion cracking), but not by SCC under pure static loading. The corrosion fatigue rates were thereby in the typical range for homogeneous LAS base metal and LAS HAZ specimens from other PSI investigations [7, 37] (Figure 16). The Alloy 182 bulk weld material and the bulk LAS of the bimetallic weld joint specimen therefore behaved exactly in the same manner as corresponding materials in homogeneous specimens.

Figure 13: Illustration of the fundamentally different EAC behaviour in the interface region perpendicular to the fusion line in high-temperature water with < 1 ppb (a) and 50 ppb chloride (b) by polished and etched cross sections.

Figure 14: Bending tensile stress as a part of the driving force for interfacial cracking.
In contrast to high-purity or sulphate containing oxygenated water, the interdendritic SCC in the Alloy 182 weld metal in high-temperature water with 50 ppb of chloride very easily crossed the fusion line straight forward on the same plane under PPU and constant loading conditions (Figures 13b, 17 and 18) with very minor cracking along the interface. In the LAS HAZ and LAS base metal the crack grew with an extremely high rate under constant load. The SCC rates were thereby in the typical range of similar PSI investigations with chloride impurities and homogeneous LAS specimens (Figure 19) [7]. The same observation was also made under cyclic load in chloride containing water [37]. The trans- or mixed inter-/transdendritic corrosion fatigue crack also very easily crossed the fusion line under cyclic load and further propagated along a TG path in the LAS HAZ. The corrosion fatigue crack growth rates were thereby in the typical range for LAS under similar system conditions [7, 37]. It was thus clearly demonstrated, that in contrast to high-purity or sulphate containing high-temperature water, a growing mixed trans-/interdendritic corrosion fatigue and interdendritic SCC crack in the Alloy 182 weld metal can very easily cross the fusion line under both cyclic or static load in case of chloride impurities in the reactor coolant.

Figure 15: Comparison of SCC crack growth rates at in the fusion line-LAS HAZ region and in the unaffected LAS base metal in specimen MN TS 15 after crossing the fusion line by severe cyclic loading (see Figure 13a) with SCC DLs for Alloy 182 [31] and LAS [32].

Figure 16: Comparable corrosion fatigue crack growth rates in LAS HAZ and base metal of specimen MN TS 15 with PSI results from homogeneous LAS specimens [7, 37].
Figure 17: Fracture surface of specimen MN TS 12 with interdendritic SCC in Alloy 182 and TG SCC in the LAS HAZ and base metal from a test in high-temperature water with 2 ppm DO and 50 ppb of chloride.

Figure 18: Comparison of the fundamentally different SCC behaviour in the interface region in high-temperature with sulphate and chloride.
3.3.2 SCC initiation and crack growth behaviour parallel to the fusion line

The EAC behaviour parallel to the fusion line was investigated with notched 12.5 mm thick 0.5T C(T) specimens, which either had a sharp ($\rho = 0.1 \text{ mm}$) or a blunt ($\rho = 1.5 \text{ mm}$) circular notch and which were pre-cracked in the test environment under cyclic load. The notch-root was located in the Alloy 182 dilution zone in the vicinity of the fusion line. The plane of the subsequent EAC crack growth was parallel to the fusion line and perpendicular to the solidification direction (and primary dendrite axis) of the Alloy 182 weld metal.

In case of the sharp notch, the crack initiated at the notch-root in the dilution zone of the Alloy 182 weld metal by fatigue with no (high-purity water) or only minor (50 ppb Cl) environmental effects (Figure 20). After a crack advance of about 200 $\mu \text{m}$, stationary “long crack” growth behaviour was observed. As shown schematically in Figure 21a, the crack grew transdendritically by corrosion fatigue parallel to the fusion line (perpendicular to the solidification direction). The environmental acceleration of corrosion fatigue increased with decreasing loading frequency (Figure 22). The corresponding corrosion fatigue crack growth rates were in the same range as the ANL-curve for Alloy 600 [39] and a factor of 5 to 10 lower than typical interdendritic corrosion fatigue rates perpendicular to the fusion line (parallel to the solidification direction) under identical loading and environmental conditions (see Figure 29 in Section 3.3.4). The transdendritic EAC crack growth rates under subsequent PPU confirmed the absence of any significant SCC contribution during the constant load periods (Figure 22). Briefly after addition of 30 ppb of sulphate, deflection of crack growth from its original path occurred under PPU and the crack grew interdendritically along the solidification direction towards the fusion line by SCC (Figures 21a, 22 and 23). The SCC crack growth rates were thereby in the typical range for Alloy 182 for this $K_I$-level and environmental conditions. When the crack front reached the fusion line, crack arrest occurred at most locations at the fusion boundary, but in some isolated regions the crack further grew along the fusion boundary, which further confirmed the previous observations (Figure 23).
Figure 20: Comparison of fatigue initiation in Alloy 182 dilution zone in oxygenated high-temperature water with ASME III air mean curve for austenitic SS [33-35].

Figure 21: Schematic of the EAC path in specimens MN ST 7 (a) and MN ST 6 (b) with crack planes parallel to the fusion line and notch-roots at the fusion boundary.
Figure 22: Comparison of the transdendritic corrosion fatigue crack growth rates in specimen MN ST 7 in the dilution zone of Alloy 182 weld metal parallel to the fusion line under cyclic and PPU in oxygenated high-temperature water with the ANL corrosion fatigue curve for Alloy 600 [39]. Furthermore, the subsequent interdendritic SCC growth rates perpendicular to the fusion line under PPU are compared with a modified ANL-curve with an additional SCC contribution.

Figure 23: Polished and etched cross section of Alloy 182 half of specimen MN ST 7 (see Figures 21a and 22).

The basic idea of the corresponding test with the blunt notch specimen ($\rho = 1.5$ mm) was that the crack can initiate in and easily grow along the most susceptible microstructure region. In this test, 50 ppb chloride was added from the beginning of the experiment to reduce the required crack initiation time. Although the mid-plane of the specimen was located in the dilution zone of Alloy 182 very close to the fusion line, the crack initiated in the HAZ of the LAS by corrosion fatigue (Figure 21b). This indicates a higher corrosion fatigue susceptibility of the LAS with respect to the Alloy 182 weld metal under these environmental and loading conditions, which is not surprising. It is stressed that the situation might be opposite under pure static loading conditions as shown by service experience, in particular in high-purity water without chloride. Unfortunately, typical SCC initiation times of one to several years make such investigations under lab/project conditions impossible. With respect to air environment the number of cycles to crack initiation was drastically reduced by the action of the test environment as it is expected for LAS under these conditions (Figure 24) [34].
Figure 24: Comparison of the numbers of fatigue cycles to crack initiation in the region of the HAZ of the low-alloy RPV steel in the blunt notch dissimilar metal weld specimen MN ST 6 in oxygenated high-temperature water with the ASME III mean and design curve for LAS [33, 34] (see also Figure 21b).

Subsequent to initiation, the corrosion fatigue crack growth rate in the LAS HAZ continuously accelerated. After a crack advance of about 400 μm, stationary TG corrosion fatigue crack growth with a rate in the range of the low-sulphur line of the Ford & Andresen EAC model [41, 42] was observed (which might be related to the fact that only few MnS-inclusions were intersected by the short crack enclave). With further crack growth and decreasing loading frequency the environmental acceleration of corrosion fatigue crack growth in the LAS HAZ increased and the corrosion fatigue crack growth rates shifted towards the typical PSI corrosion fatigue data of long cracks [7, 37] of homogeneous LAS specimens under NWC conditions (Figure 25). The corrosion fatigue crack grew parallel to the specimen mid-plane and therefore slowly approached the fusion line because of its wavy form (through the sequence of the individual weld beads). After switching to PPU the crack in the LAS HAZ grew by SCC with a high rate, which was in the typical range for LAS in chloride-containing, oxygenated high-temperature water (Figure 19). Close to the fusion line, constant load was applied and the TG crack further grew with a fast SCC rate. When the crack front reached the fusion line, it easily crossed this boundary and then further grew along the main dendrites in the dilution zone of the Alloy 182 weld metal parallel to the solidification direction (and therefore perpendicular to the fusion boundary), which caused a deflection of the SCC crack plane by an angle of 45 to 90° at the fusion boundary. The resulting mixed mode loading in combination with the high surface roughness then resulted in a crack closure-induced cessation of the SCC crack growth in this region at most locations of the crack front (Figure 26). After a load increase, the crack in the dilution zone of the Alloy 182 weld metal close to the fusion boundary re-started to grow. The SCC crack growth rate was thereby in the typical range for Alloy 182 under these test conditions (Figure 27).
Figure 25: Comparison of the corrosion fatigue and SCC crack growth rates in the LAS HAZ of the blunt-notched dissimilar metal weld specimen MN ST 6 under cyclic and PPU conditions in oxygenated high-temperature water with 50 ppb chloride with corresponding crack growth curves, which were developed for homogeneous LAS specimens [7].

Figure 26: Apparent cessation of interdendritic SCC growth in the dilution zone of the Alloy 182 weld metal of the blunt-notched dissimilar metal weld specimen MN ST 6 briefly after crossing the fusion boundary because of crack closure and re-establishment of sustained SCC with growth rates in the typical range for Alloy 182 weld metal after a small load increase.
3.3.3 SCC behaviour of the Alloy 182 bulk weld metal

The SCC crack growth rates in the Alloy 182 bulk weld metal parallel to the solidification direction in high-purity water were only slightly higher than in the Alloy 182 dilution zone and slightly below the current Swedish SCC DL for Alloy 182 and BWR/NWC conditions, which further confirms the adequacy and conservativeness of this DL (Figures 8 and 27). A plateau SCC cracking behaviour was observed above a $K_I$-value of 20 to 25 MPa·m$^{1/2}$, i.e., the SCC crack growth rates were independent of the applied $K_I$ (Figures 8, 27 and 28a). The SCC crack growth rates in the Alloy 182 bulk weld metal parallel to the solidification direction in oxygenated high-temperature water with 50 ppb Cl$^-$ or 30 ppb SO$_4^{2-}$ were about a factor of 10 and 5 higher than in high-purity water and slightly higher than the Swedish SCC DL for Alloy 182 (Figure 28a) [31]. The transdendritic EAC crack growth rates under PPU perpendicular to the solidification direction were more than a factor of 10 lower than the corresponding interdendritic EAC crack growth rates parallel to the solidification direction (Figure 28b).

Figure 27: Comparison of the SCC growth rates in the dilution zone of the Alloy 182 weld metal of the blunt-notched dissimilar metal weld specimen MN ST 6 with results from homogeneous and dissimilar metal weld specimens under similar testing conditions.

Figure 28: Interdendritic SCC crack growth rates in the Alloy 182 bulk weld metal in high-purity water and water with 30 ppb of sulphate or 50 ppb of chloride (a). Trans- and interdendritic SCC crack growth rates in the Alloy 182 bulk weld metal perpendicular and parallel to the solidification and primary dendrite axis, respectively (b).
3.3.4 Corrosion fatigue behaviour of the Alloy 182 bulk weld metal

A higher corrosion fatigue initiation susceptibility was observed in the LAS HAZ than in the dilution zone of the Alloy 182 bulk weld metal (see Section 3.3.2). In good agreement with wrought Alloy 600 and in contrast to LAS [34], only a very minor environmental reduction of fatigue life was observed in Alloy 182 under BWR/NWC conditions (Figure 20). On the other hand, significant environmental acceleration of fatigue crack growth of mechanically long cracks was observed in Alloy 182 weld metal under BWR/NWC conditions at loading frequencies below 0.1 Hz (Figures 22 and 29). The environmental acceleration increased with decreasing loading frequency. The transdendritic corrosion fatigue crack growth rates under cyclic load perpendicular to the solidification axis were a factor of 5 to 10 lower than the corresponding interdendritic corrosion fatigue crack growth rates parallel to the solidification axis and in a similar range for the wrought Alloy 600 and the ANL-curve for Alloy 600 [39] (Figure 29). Interdendritic corrosion fatigue crack growth rates in Alloy 182 under cyclic load were significantly higher (a factor of 10) than in wrought Alloy 600 and the ANL-curve for Alloy 600 (Figure 29).

![Figure 29: Comparison of trans- and interdendritic corrosion fatigue crack growth rates in Alloy 182 weld metal perpendicular and parallel to the solidification axis in oxygenated high-temperature water with the ANL-corrosion fatigue curve for wrought Alloy 600 [39] and a modified curve, which includes an additional SCC contribution at low loading frequencies or high load ratios.](image-url)
3.4 Summary and conclusions

Recent SCC incidents in the weld metal of bottom head penetration housing and core shroud support welds of BWRs have drawn the attention to the SCC behaviour in the transition region of Alloy 182-LAS dissimilar weld joints, and in particular, to the issue whether an IG/interdendritic SCC crack propagating in the weld metal could easily penetrate through the fusion line into the adjacent low-alloy RPV steel. In order to improve the understanding of the SCC behaviour in the transition region, the SCC crack growth behaviour in the transition region of a simulated Alloy 182-SA 508 Cl.2 weld joint was investigated under simulated BWR/NWC conditions at FRI and PSI.

The main conclusions of these experimental investigations can be summarised as follows [3, 7]: The observed EAC cracking behaviour correlates excellently with the field experience, where SCC cracking was usually confined to the Alloy 182 weld metal and no cases of SCC were observed in LAS primary pressure boundary components. The fusion boundary seems to represent a significant barrier for SCC crack growth from the Alloy 182 to the RPV steel, but minor crack growth into the adjacent RPV steel is not impossible. Under static loading conditions in chloride-free, high-temperature water, there seems to be little risk that a fast growing interdendritic SCC crack may cross the fusion line and significantly propagate into the adjacent low-alloy RPV steel. Even if a crack would cross the fusion line under static load, it is anticipated that cessation of SCC occurs in the LAS HAZ or at the latest in the unaffected LAS base material. Major EAC crack propagation into the RPV material is therefore not expected as long as the number of plant transients is limited and severe and prolonged chloride excursions are avoided (i.e., if water chemistry is kept below the EPRI action level 1 limit).

These important conclusions should be verified by further tests with additional dissimilar weld joints representing different welding qualities and different LAS base materials (steel sulphur content and susceptibility to dynamic strain ageing). Since the stress intensity factor in the interface region of dissimilar metal welds can reach quite high levels in some field situations, such investigations should also include experiments in the K_I-range of 50 to 90 MPa·m^{1/2} with chloride levels below 5 to 10 ppb, which requires the use of sufficiently large specimens. Because of the safety concern in case of SCC in the RPV, the special emphasis of these evaluations should be placed to a better establishment of the K_I-thresholds, where the onset of fast SCC in LAS may occur (Figure 30).

![Figure 30: Thresholds for SCC crack growth into LAS for different anion & steel sulphur contents.](image)

The observed interface cracking behaviour should also be verified by some few experiments under primary PWR water conditions at higher temperatures of 320 °C, although the situation is rather relaxed here, because of the very low SCC and moderate corrosion fatigue susceptibilities of LAS at low ECPs.
3.5 Appendix

3.5.1 Metallurgical and microstructural aspects of SCC in Alloy 182 weld metal

Manually welded multipass dissimilar weld joints usually have a complex welding configuration (e.g., safe ends with cladding and weld butter, Figure 31) and microstructure, which may significantly differ from weld to weld depending on welding parameters. During solidification, parallel bundles of dendrites with nearly identical crystallographic orientation form and grow into the melt along the direction of heat flux. The boundaries between these similarly oriented dendrites are often called solidification sub-grain boundaries and tend to have low angular mismatches, as well as low energy. Where different bundles of dendrites intersect or overlap, larger angular mismatches often occur between the grains of the bundles. In this case, the resulting high-angle grain boundaries, termed solidification grain boundaries, can be high energy random grain boundaries or low energy coincident site lattice boundaries. Solidification grain boundaries can migrate on cooling after solidification and during re-heating and result in a straighter, migrated grain boundary. [20]

Figure 31: Example of complex weld configuration of a safe end to nozzle dissimilar metal weld.

The general structure of an individual weld pass consists of elongated columnar dendrites at the edge of the melt pool, where heat transfer is highly directional, and more equiaxed cellular dendrites in the middle with slower cooling. During solidification, the dendrite arms grow towards each other and trap enriched (with S, P, Si, Mn, Nb) liquid in the interstices along with melt oxides and slag inclusions. Therefore, Mn and Nb segregation is often observed in the grains and the grain boundaries are usually decorated with precipitates (carbides) and enriched with S, P, and Si. Each weld pass heats up the layers below and can cause recrystallisation and migration of solidification grain boundaries, segregation, precipitation (e.g., carbides) or hot cracking and produce a wide range of microstructures potentially susceptible to SCC. [8, 20]

PWHT of the weld buttering is a standard practice, and is in some cases also performed on the weld filler metal following final welding. In addition, PWHT often occurs indirectly as a consequence of stress relief heat treatment performed on adjacent LAS components, but in this case the stress relief temperature is well below the optimum for the more creep/relaxation resistant Ni-base alloys. While stress relief treatment only slightly reduces the peaks in residual stress, it can produce some elevation in yield strength/hardness and potentially exacerbate Cr depletion/sensitisation by carbide precipitations, which both may further increase the SCC susceptibility of this region [8]. As-welded structures are thus usually less sensitised than welds, which were PWHT, although both behaved very similarly in lab SCC crack growth rate studies [8]. In-service thermal ageing might further exacerbate SCC susceptibility [2].
The grain boundary chemistry and misorientation in the weld metal as well as the S, P, Si and Cr content and the stabilisation ratio (Nb & Ti vs. C content) are material factors, which affect the SCC cracking behaviour in the Alloy 182 weld metal [2, 8-10]:

S and Si grain boundary segregation affects the anodic dissolution and repassivation behaviour at the grain boundaries and/or can result in decohesion and grain boundary weakening. The high-angle random boundaries are prone to segregation and precipitation, while low energy boundaries, such as low-angle boundaries, are not [10]. Furthermore, the strong mismatch in Taylor factor in high angle boundaries favours a higher strain incompatibility and grain boundary deformation and therefore further promotes IG/interdendritic cracking [11]. Low-angle boundaries are more resistant to SCC than high-angle coincident and random boundaries, the latter being the least resistant [10]. The different susceptibilities are also the main reason for crack pinning phenomena and uneven crack fronts in SCC lab testing.

SCC cracks in weld metals typically follow the higher energy random solidification grain boundaries and/or migrated grain boundary [20]. The orientation of the crack in the weld, i.e., relative to the weld’s columnar microstructure, has a strong influence on the crack growth rate. Cracks grow fastest along high energy grain boundaries in the direction of main dendrite growth, and next fastest along high energy grain boundaries perpendicular to the direction of dendrite growth, but parallel to the welding direction [20]. Cracks that grow perpendicular to the high energy grain boundaries, i.e., perpendicular to the columnar dendrites, grow significantly more slowly. In the interface region between the weld metal and the heavy-section LAS structures, the main dendrite growth direction is perpendicular to the fusion line according to the heat-flux during welding [20].

Cr strongly affects the SCC behaviour in Ni-base alloys since it significantly improves their corrosion resistance and repassivation behaviour. Cr has a high affinity to C and readily forms carbides such as Cr_{23}C_{6} or Cr_{7}C_{3}, which can precipitate at grain boundaries. This may result in Cr depletion in a narrow zone close to the grain boundary (sensitisation) and concurrent higher susceptibility to IG corrosion and IG SCC at high ECPs (BWR/NWC), although this effect is less pronounced than in SS [2, 8]. The problem of Cr depletion can be reduced by either increasing the Cr content or by introducing elements with high affinity to C like Nb and Ti, which form very stable carbides. This aspect is reflected by the SCC resistance index SCRI [43] in equation (1), which is used by the BWR vendors to characterise the SCC susceptibility of Ni-alloys.

\[
SCRI = Cr + (Nb + Ta) \cdot 5 + Ti \cdot 10 - 116.5 \cdot C
\] (1)

In wrought alloy 600, grain boundary carbide improves its IG SCC resistance, which is related to the reduction in grain boundary sliding. Such correlations are not observed in Alloy 182 weld metals [44].

Carbon strongly affects the mechanical properties and yield strength of the material by primary (MC, etc.) and secondary (M_{23}C_{6}, etc.) carbide precipitation. A higher C content usually results in a higher high-temperature yield strength, which may make the alloy more susceptible to SCC (steeper strain gradient and higher crack-tip strain rate) and low-temperature rapid crack propagation, and produce high weld residual stresses and stress intensity factors. On the other hand, high carbon content also favours sensitisation. Similarly to yield strength, cold-work (e.g., from surface grinding or weld shrinkage strain, etc.) also increases SCC crack growth rates and susceptibility at high (BWR/NWC) and low ECPs (PWR, BWR/HWC). [8]
3.5.2 Metallurgical and microstructural aspects of SCC in the interface region between the Alloy 182 weld and LAS base metal

In the transition region between the Alloy 182 weld metal and the LAS a very complex microstructure with varying chemical composition can evolve, e.g., due to dilution of alloying elements between the Alloy 182 weld filler and LAS base metal during welding or the migration of C from the LAS to the weld metal during PWHT: In the weld metal close to the fusion line, a few mm thick zone with different chemical composition from the mixing of the weld filler material with the molten LAS is observed. The reduced Cr (and Ni) content in this dilution zone close to the LAS tends to increase its corrosion and SCC susceptibility (e.g., by the effect of Cr on the repassivation behaviour) with respect to the bulk weld metal. The local formation of a very narrow and hard layer of martensite on cooling in the weld metal adjacent to the fusion boundary is possible, but requires a very high dilution and is thus much less frequently observed than in SS weld filler materials.

As a consequence of the C migration, PWHT can potentially produce a more distinct elevation in yield strength/hardness and more severe Cr depletion/sensitisation by carbide precipitations in the interface region compared to the bulk weld metal, which both may further increase the SCC susceptibility of this region. A further consequence of C diffusion is that the LAS may become slightly decarburised in a zone adjacent to the fusion boundary in severe cases. This decarburised zone may be softer and may have a lower yield strength than the parent LAS, and therefore a potentially higher susceptibility to strain-induced corrosion cracking during plant transients, but increased SCC resistance under stationary operating conditions. The LAS weld HAZ of dissimilar weld joints usually has a lower peak hardness (≤ 300 HV5) than in welds between LAS components (≤ 350 HV5).

The moderately different thermal expansion coefficient of Alloy 182 and LAS may produce thermal fit-up stress at the interface during start-up/shut-down and also at operating temperature and thus result in a change of the acting stress intensity factor in the field, although the situation is much less severe than with SS weld filler metals.

Significant galvanic effects are not expected, since the ECP of the LAS in BWR/NWC environment is only slightly lower (50 to 100 mV) than that of Alloy 182 and since the throwing power is very limited in high-purity (very low conductivity) water.
3.6 References


[27] K. Reichlin, "2-D FE calibration of DCPD for a bimetallic weld joint", Paul Scherrer Institute, Villigen, Switzerland, unpublished work.


[31] Personal communication with C. Jansson (SwedPower AB).


ACKNOWLEDGEMENTS

The financial support of this project by the Swiss Federal Nuclear Safety Inspectorate (ENSI) and Swiss Federal Office for Energy (BFE) is gratefully acknowledged. Special thanks are also expressed to B. Baumgartner, L. Nue, U. Ineichen, U. Tschanz, B. Gerodetti, and E. Groth† (all PSI) for their excellent experimental contribution to this work and to K. Reichlin (PSI) for the Finite Element calculations.

The results on the SCC behaviour in the fusion line region of the Alloy 182–low-alloy RPV dissimilar metal weld in Part III of this report were generated within a co-operation work between the Fracture and Reliability Research Institute (FRI) of the Tohoku University (Sendai, Japan) and PSI and were also a part of the international PEACE-II/III (FRI) program. The financial support of the FRI part by the PEACE-II/III programme by EDF, EPRI, Hitachi, JAPCO, KEPCO, MHI, SKI, Tohoku EPCO, TEPCO, TEPCO, Toshiba and IHI is also gratefully acknowledged. The excellent and fruitful collaboration with Prof. T. Shoji, Q.J. Peng, Y. Takeda and Z.P. Lu from FRI is highly appreciated.