Effect of surface state of Alloy 182 weld metal on stress corrosion cracking initiation in high temperature high pressure water

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Abstract

Incidents of stress corrosion cracking (SCC) in Alloy 182 dissimilar metal welds in light water reactor (LWRs) service have not only seriously reduced plant availability, but also challenged integrity, safety and lifetime of LWRs This study briefly summarizes the impact of surface states on SCC initiation of Alloy 182 welded components in high temperature high pressure water simulating LWR environments, drawing insights from several years of laboratory experiments conducted at PSI in collaboration with international partners. Constant extension rate tensile tests and constant load tests with flat tapered tensile specimens were carried out in the temperature range of 270-320 °C in simulated boiling water reactor environment – both in normal water chemistry and in hydrogen water chemistry– as well as in primary pressurized water reactor environment. The study highlights the complexities of SCC initiation, emphasizing the need for further research to understand the interplay of surface states and operational histories in mitigating SCC in Alloy 182 components.

1. INTRODUCTION

Alloy 182 is widely used Nickel (Ni) -base weld filler and attachment pad metal in dissimilar metal welds (DMWs) between low-alloy and stainless steel or Ni-base wrought alloy components in light water reactors (LWRs) due to suitable thermal expansion coefficient. Incidents of stress corrosion cracking (SCC) in Alloy 182 DMWs in LWR service have not only seriously reduced plant availability, but also challenged integrity, safety and lifetime of LWRs [1-4]. This subcritical failure can mimic brittle behavior in the otherwise ductile material under specific tensile stress-environment conditions and is susceptible to hot cracking and inter-granular (IG) SCC both in hydrogenated and oxygenated high-temperature water (HTW) under typical plant-representative material and environmental conditions [1-4]. SCC rates can be quite high, particularly at high temperatures close to the Ni/NiO phase boundary, or in case of high plastic weld shrinkage strains, even in hydrogenated HTW [1-3]. Periodic in-service inspections are challenging and often underestimate the size of deep cracks, making SCC mitigation crucial for the safe long-term operation of LWRs.

The reasonable service record of Alloy 182 is primarily due to beneficial weld residual stress profiles, surface stress states, and the slow growth of shallow cracks, despite high SCC growth rates of long cracks under typical service conditions. However, investigations into SCC initiation [4-12] and short crack growth [13-15] are relatively few, primarily due to the complex and stochastic nature of these phenomena, and challenging, time consuming and costly laboratory research that often lack transferability to real components and accurate lifetime predictions. SCC initiation occurs through the cyclical rupture and re-formation of the protective Chromium (Cr)-rich oxide film at the substrate-environment interface, influenced by often unknown surface conditions viz. surface hardness and cold-work, surface/subsurface microstructure, residual stress, oxide films and deposits, roughness and scratches acting as stress concentrators and crevices, etc., all of which are affected by the surface machining/grinding processes used as essential component finishing steps. The plastic deformation in the machining/grinding-induced cold-worked surface-affected layer (SAL) is inhomogeneous, with a strong gradient in microstructure, hardness, and residual stress, along with a high concentration of

defects like dislocations and crack-like features that may act as crack initiators. The SAL consists of a severely plastically deformed nano/ultrafine grained (NUG) layer at the surface, followed by a strained subsurface layer with distinct grain structure. The high internal energy and defect density in the SAL provide a strong thermodynamic drive for accelerated microstructural changes and reactions with the environment under operating conditions. Hydrogen from corrosion and/or the environment can enhance plastic deformation via mechanisms such as hydrogen-enhanced local plasticity. The SAL can also accelerate diffusion, affecting oxide film formation, corrosion, and SCC initiation. Accelerated diffusion of Cr, e.g. along the multiple grain boundaries (GBs) in the NUG layer, could enhance formation of the protective Cr-rich thin oxide film whereas localized preferential oxidation along GBs or deformation/slip bands may act as precursors for SCC initiation. The physical IGSCC initiation is likely governed by local micro-stresses at specific grain boundaries, with subsequent short-crack growth influenced by macro residual and applied stress profiles. The higher strength/hardness also likely results in higher SCC rates under otherwise identical conditions [3]. Methods such as shot peening and mechanical stress improvement aim to introduce high compressive residual stresses as potential SCC mitigation strategies. Thus, while surface states can influence cracking, the extent to which they affect SCC initiation and growth in real-world conditions remains unclear, given the complexities of boundary conditions and operational histories.

Although stress corrosion failures are comparatively rare, the consequences can be severe. Consequently, considerable effort is being focused on evaluating the effect of operational variables and in developing an improved basis for structural integrity assessment. The current paper discusses and compares some of the main (and only partially published) phenomenological results of SCC initiation test campaigns conducted at PSI during the several years [13, 16-20] and some selected results from MEACTOS (EURATOM Horizon 2020 [21]) that was proposed and conducted by a consortium of 15 partners from 12 countries (Spain, France, Finland, Czech Republic, Belgium, Germany, Slovakia, Romania, UK, The Netherlands, Slovenia, and Switzerland), comprising research laboratories (VTT, SCK CEN, CVR, CIEMAT, PSI, JRC, RATEN, ZAG), universities (University of Manchester, STUBA), nuclear component suppliers (NAMRC, ENSA), utility (EdF), an engineering company (Jacobs) and plant designer (Framatome GmbH). The SCC initiation behavior of Alloy 182 weld metal was investigated by constant extension rate tensile (CERT) tests and constant load (CL) tests with flat tapered tensile (FTT) specimens. Tests were carried out in simulated boiling water reactor (BWR) environment – both in normal water chemistry (NWC) and in hydrogenated water chemistry (HWC) – as well as in primary water reactor (PWR) primary coolant. The effect of surface states - ground, electropolished, machined and shot peened surfaces – are briefly discussed. It is inherent in such an overview paper that most experimental and material aspects cannot be captured in detail and the readers are referred to the corresponding references for further information.

2. EXPERIMENTAL PROCEDURE

2.1 Material

The Alloy 182 test welds used in the present study were fabricated according to nuclear welding specifications by multipass shielded metal arc welding involving (1) BW A: an as-welded 100 mm thick butt weld in a very large 220 mm thick RPV plate on which an additional post-weld heat treatment (PWHT) at 620 °C for 24 h was performed on cutout segments [1, 13, 16, 17], and (2) WOL C: a post-weld-heat-treated (PWHT) (580-600 °C for approximately 1.5 h) 20 mm thick weld overlay on a 75 mm thick carbon steel plate [18, 19, 21]. In spite of different fabrication and heat treatment conditions, all the welds had similar chemical compositions and mechanical tensile properties with a yield stress (YS_T) in air at 274 and 350 °C of approximately 400 and 360 MPa, respectively. The strength of weld WOL C was slightly lower than in BW A. The welds had slightly different microstructures (grain structure and GB distribution, precipitates), plastic weld shrinkage strain distributions (that varied with location in the weld) and thus, potentially different SCC susceptibilities.

2.2 Sample preparation

FTT specimens were designed to ensure that the stress in the largest and smallest cross-sections reached the yield strength (YS_T) and ultimate tensile strength (UTS) at maximum load, respectively. Consequently, exposure time above a certain plastic strain and strain rates varied with location. More

details about the accelerated CERT testing technique using FTT specimens can be found in [17]. Specimens were fabricated using electrical discharge machining, with gauge surfaces subsequently prepared via specific finishing treatments. They were either harshly ground using P180 silicon carbide paper or electropolished with perchloric acid and ethanol/methanol solutions. Harsh grinding along the loading direction created a rough surface with residual stresses that facilitated the detection of small SCC microcracks, while electropolishing produced a smooth strain-free surface but with potential hydrogen uptake and chemical modifications in the surface layer. Large scratches might act a stress concentrator if oriented perpendicular to stress axis, and as crevices with occluded crevice chemistry in NWC environment. For the MEACTOS project [21], three industrial-grade techniques simulated the component surface: (1) conventional face milling, (2) advanced face milling with supercritical CO_2 cooling, and (3) shot peening. A reference surface (RS) was created by gentle grinding with P2000 SiC paper, leading to improved residual stresses but with variability due to the grinding process. Only the specimens for the MEACTOS project were quantitatively characterized in detail. Table 1 summarizes parameters of the industrial-grade surface finishes applied to the FTT specimens, including surface roughness, hardness, and residual stress values. The initial shot peening treatments had many defects and microcracks, while subsequent optimized peening resulted in a defect-free surface. The machined and SP specimen surfaces revealed a clear and very distinct cold-worked SAL that consisted of a gradient microstructure with a very thin NUG (sub-grains, recrystallized) surface layer with a thickness of 0.5-5 μm at the top and subsurface cold-worked deformed layer with a thickness ranging between 10-50 μm in all cases. Depending on type of surface treatment, there were either high tensile (STI: ~+100 MPa, SAM: ~+700 MPa) or compressive (RS: ~ -400 MPa) residual stresses in a very thin surface layer of 10-20 μm. SP resulted in high compressive residual stresses (~ -400 to -600 MPa) up to a depth of 0.5-1 mm. Similar residual surface stresses were measured by XRD on the specimens and plate. Further details about the surfaces can be found elsewhere [21].

2.3 Tests and characterizations

CERT and CL tests with FTT specimens were performed in state-of-the-art HTW loops with continuous monitoring and controlling of all relevant testing parameters (O_2, H_2, $pH_{25°C}$, conductivity, flow rate in inlet and outlet, T, pressure, ECP, redox potential, load, displacement, etc.). Tests compared in the present work were carried out at 274 °C in BWR/HWC water, 288 °C in BWR/NWC and in BWR/HWC water, and 320 °C in PWR primary water environment. All specimens were first pre-oxidized for one week in the test environment at a small constant pre-load well below the YS_T. Most specimen surfaces were thoroughly analyzed using scanning electron microscopy (SEM) before testing to identify any pre-existing defects that could be mistaken for IGSCC in subsequent CERT tests. The remaining surfaces were examined under a light optical microscope. After the CERT tests, initiated SCC cracks on the gauge surfaces of the FTT specimens were analyzed with a field emission gun-SEM. Shorter and fewer cracks were noted towards the wider gauge section, where the last "qualified" IG crack was identified for calculating the critical stress for crack initiation (σ_{th}). The critical stress σ_{th} was determined by dividing the peak load by the measured cross-section at the location of the last crack [16-19, 21]. The extrapolation of σ_{th} to zero strain rate provided an approximation for the SCC threshold stress (σ_{SCC}) under static load. An exponential fit function was typically employed to assess the strain dependence of σ_{th}. It was found that there was generally little difference between the extrapolated σ_{SCC} and the mean σ_{th} from multiple tests at strain rates up to $5*10^{-7}$ s^{-1}. It is assumed that SCC will not occur below σ_{SCC}; thus, higher SCC stress thresholds indicate lower susceptibility to SCC.

Table 1 Overview of industrial grade surface finish treatments applied to the FTT specimens

Surface finish treatment	R_a	R_t	Hardness	Residual surface stress
Face milled (STI)	0.5 μm	2.8 μm	260 HV_5 310 $HV0_{0.5}$	Tensile ~+100 MPa
Face milled with supercritical CO_2 cooling (SAM)	0.3 μm	2.5 μm	270 HV_5	High tensile ~+700 MPa
Shot peened (SP) ("poor quality")	3 μm	3 μm	400 HV_5 460 $HV0_{0.5}$	Very high compressive residual stress (~-400 to -600 MPa)
Gently ground (RS)	0.02 μm	0.1 μm	230 HV_5	High compressive (~-400 MPa)

3 RESULTS AND DISCUSSION

The varying surface states (having different surface roughness, residual stress and plastic strain/hardness, oxide films and deposits, scratches, etc.) from surface finishing, machining or specific treatments like peening or PWHT may affect the SCC initiation behavior. These conditions strongly depend on the methods and parameters, cooling or lubricants, component thickness, etc., but also on the material itself. Resulting surfaces can vary strongly even for a given method and have very different surface hardness, roughness, residual stresses and plastic strains.

3.1 Effects of laboratory surface finishing techniques

Although they result in quite different surface states (surface residual stresses, cold-work/hardness, roughness, nature of protective oxide films), the different laboratory-level surface finishing techniques (grinding and electropolishing) had little impact on SCC initiation susceptibility in the strain-controlled CERT tests with significant plastic yielding. Moderately smaller (20-30 MPa) critical stresses σ_{th} were consistently observed in harshly ground surface than electropolished surfaces for all material and environmental conditions (Figure 1). A pre-existing, thin, air-formed, Cr-rich passive film resulting from the Cr enrichment at the surface during the electropolishing process might be the reason for this improvement.

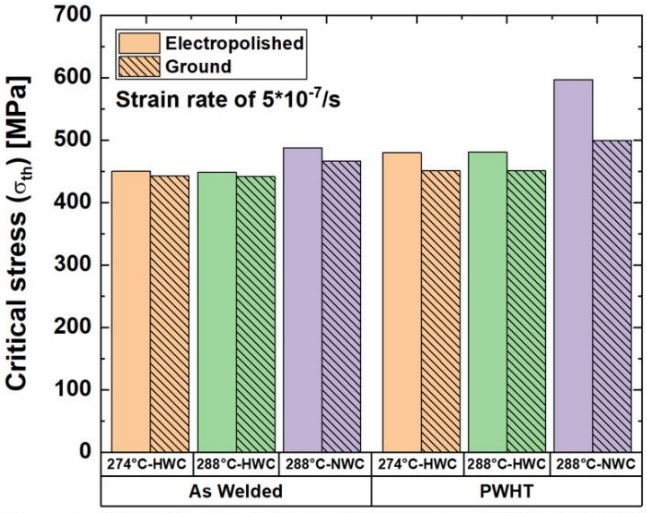

Figure 1 Effect of surface finish on critical stress σ_{th} of Alloy 182 BW A with different heat treatment conditions at a strain rate of $5 \cdot 10^{-7}$ s^{-1} exposed to hydrogenated HTW (at Ni/NiO phase boundary conditions) at 274 °C, hydrogenated (Ni/NiO) or oxygenated (2 ppm) HTW at 288 °C.

3.2 Effect of industrial surface finishing techniques

Despite lower roughness, hardness, and high compressive stresses, the RS surfaces showed inferior performance as compared to the other surfaces. SAM surfaces showed significantly higher (100 MPa in PWR environment, albeit with high scatter) or similar (BWR/NWC) extrapolated σ_{SCC} as STI or optimized gentle grinding (RS), indicating similar susceptibility (Figure 2(a-b)). This is unexpected, given that SAM specimens exhibited the highest tensile residual stresses and comparable hardness and roughness to STI. The reasons for this discrepancy remain unclear; accelerated grain boundary diffusion in the NUG layer of SAM may have facilitated Cr-rich barrier formation, though this is purely speculative. The advanced supercritical CO_2-cooled face milling technique offers additional benefits like higher cutting speeds and reduced lubricant-related pollution.

Significant plastic yielding alters the thin SAL layers (< 50 µm, with a NUG layer < 5 µm) and the associated surface residual stress from machining and surface finishing, ultimately likely having minimal impact on SCC initiation. Concerns about plastic shake-down due to fatigue are also relevant. The long-term stability and durability of these very thin surface layers in HTW during prolonged operation with various plant transients is questionable. Small pre-existing surface defects that extend beyond the thin compressive stress layer may initiate SCC cracks. Effective mitigation or adverse effects due to peening or subpar machining/grinding can only be anticipated when there is sufficient penetration depth (approximately 0.5-1 mm) of compressive surface residual stress, or in instances of tensile residual stress combined with high surface hardness, respectively.

The CERT tests, which involve increasing plastic straining across the entire ligament, likely underestimate the actual SCC mitigation effects of peening—particularly in welds where the residual stresses from welding dominate and significant (surface) plastic deformation during service can be discounted. The SP specimen is a sandwich composite featuring a hard surface (highly cold-worked, with high yield stress and compressive residual stress - the absence of which increases SCC susceptibility) and a soft bulk (normal yield stress of DMW, with moderate tensile residual stress).

Sufficient plastic straining in such tests can entirely reverse the surface stress state from compressive to high tensile stresses [22], which may account for the behaviour observed in the "poor quality" SP specimens. Other test methodologies, such as using C-rings with welds followed by peening, may prove more suitable. Furthermore, plastic shake-down due to fatigue can diminish compressive stresses [23,24]. Cyclic plastic deformation may also reduce surface cold-work/hardness through recovery processes, including initial cyclic softening in cold-worked stainless steel [23,24].

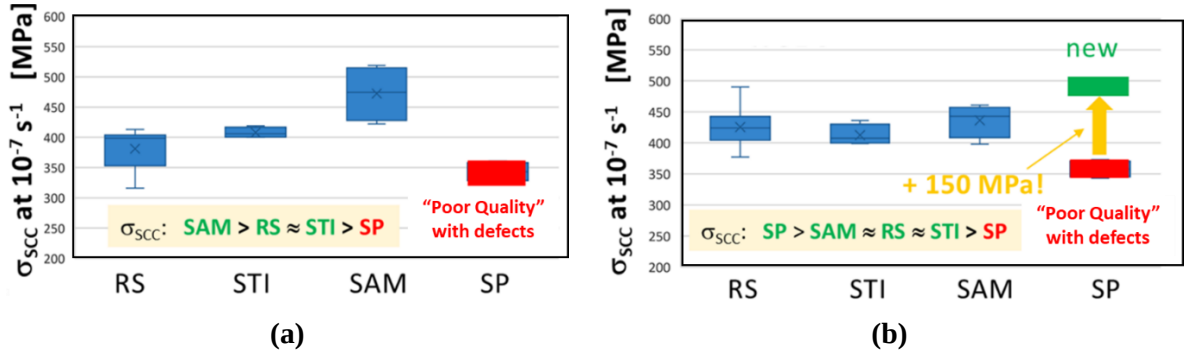

Figure 2 Critical stress σ_{th} (~σ_{SCC}) at 10^{-7} s^{-1} in WOL C for RS, STI, SAM or SP in (a) primary PWR environment at 320 °C and (b) BWR/NWC environment at 288 °C, both at (at the Ni/NiO phase boundary) [18-19, 21].

It should be emphasized that machining of specimens and their preparation can significantly alter the original component surface. The surface of flat laboratory specimens may differ considerably from that of actual components, particularly non-machined weld roots, which may contain various defects such as end craters, lack of fusion misalignment, protrusions, and hot cracks. These defects are critical as they are in contact with the environment from which cracks may initiate. The cutting of the 3 mm thin specimens may also significantly change the surface residual stress states compared to the original component surface, although X-ray diffraction (XRD) measurements in this study did not reveal strong differences. While the used test technique was not optimal for studying surface finishing/machining effects and results were not fully conclusive, additional challenges exist in producing controlled and reproducible surface modifications by machining/grinding, even on flat laboratory surfaces under well-controlled conditions. Variations in methods and process parameters, as well as moderate deviations from specifications, can lead to markedly different surface conditions. Such investigations necessitate detailed pre- and post-test characterization of surfaces, along with a sufficiently large number of specimens and surface areas, which poses challenges in terms of time and cost. Additionally, it is nearly impossible to alter one surface parameter without affecting others, complicating systematic studies of surface parameter effects. It is thus unsurprising that different studies sometimes report conflicting effects on SCC, e.g. regarding techniques such as electropolishing. The discrepancies between accelerated short-term SCC tests using small laboratory specimens and the actual loading and boundary conditions of large plant components are also noteworthy. Size effects, including the likelihood of defects and significant microstructural inhomogeneities, as well as the complete service history of components—such as prior operation conditions, temporary water chemistry (e.g., Cl^-, SO_4^{2-} in non-water-cooled environments, O_2 in PWRs), and mechanical or thermal transients—may all influence SCC initiation in the field. Furthermore, pre-service history factors such as pickling, heat treatments, atmospheric corrosion, and weld repairs also play a role in surface condition. These complexities severely limit the direct, quantitative applicability of laboratory results for predicting component behavior in terms of SCC initiation life.

3 CONCLUSION

The impact of thin surface deformation layers (< 10 μm) from surface finishing and machining on SCC initiation in field components is uncertain, and their properties cannot be reliably controlled for reproducibility. Machining/grinding weld-roots can be beneficial to remove defects and surface oxides/deposits, reducing stress raisers and crevices. Peening or low-quality machining/grinding may have clear mitigation or adverse effects only with sufficient penetration depths (~0.5-1 mm) of

compressive or tensile residual stress and high surface hardness, respectively. A quality control protocol for surface conditions and fabrication/finishing specifications is helpful but cannot completely prevent SCC in the field.

4 REFERENCES

[1] H.P. Seifert, S. Ritter, H.J. Leber, and S. Roychowdhury, Corrosion, 2015, 71(4), 433-454.
[2] P. Andresen, "SCC of Alloy 182 and 82 Weld Metals in BWR Water", NACE - International Corrosion Conference Series, 2010.
[3] P.L. Andresen, "Understanding and Predicting Stress Corrosion Cracking (SCC) in Hot Water, Section 5 In ", in: Stress Corrosion Cracking of Nickel Based Alloys in Water-Cooled Nuclear Reactors - the Coriou Effect, D.S. Feron, R.W. (Ed.), Woodhead Publishing, UK, ISBN: 978-0-08-100049-6, pp. 169-238, 2016.
[4] U. Ehrnsten, "State-of-the-Art Report on Crack Initiation in Stainless Steel and Alloy 182 under LWR Conditions", Report No. MEACTOS DELIVERABLE D3.2, https://meactos.eu/documents/
[5] K. Norring and J. Engström, Energy Materials: Materials Science and Engineering for Energy Systems, 2008, 3(2), 113-118.
[6] E. Richey, D.S. Morton and R.A. Etien, "SCC Initiation Testing of Nickel-Based Alloys in High Temperature Water", In: 13th Int. Conference on Environmental Degradation of Materials in Nuclear Power Systems - Water Reactors, Canadian Nuclear Society, August 2007.
[7] F. Vaillant, J.-M. Boursier, C. Amzallag, C. Bibollet, and S. Pons, "Environmental Behaviour and Weldability of Ni-Base Weld Metals in PWRs", Report No. INIS-FR--08-0335, France, http://inis.iaea.org/search/search.aspx?orig_q=RN:39034050 .
[8] O. Calonne, M. Foucault, F. Steltzlen, and C. Amzallag, "Stress Corrosion Crack Initiation of Alloy 182 Weld Metal in Primary Coolant - Influence of Chemical Composition" INIS-FR-11-0625;TRN: FR1101461088772, France, Societe Francaise d'Energie Nucleaire (SFEN), 75 - Paris (France), 2011.
[9] M. Toloczko, Z. Zhai and S. Bruemmer, "SCC Initiation Behavior of Alloy 182 in PWR Primary Water", In: Proceedings of the 18th Int. Conference on Environmental Degradation of Materials in Nuclear Power Systems – Water Reactors, Springer International Publishing, Cham, ISBN: 978-3-030-04639-2, 2019.
[10] G. Troyer and et. al., "Dissimilar Metal Weld PWSCC Initiation Model Refinement for Xplr, Part I: A Survey of Alloy 82/182/132 Crack Initiation Literature", In: 17th Int. Conference on Environmental Degradation of Materials in Nuclear Power Systems – Water Reactors, Ottawa, Ontario, Canada, August 9-13, 2015.
[11] C. Amzallag, "Stress Corrosion Life Assessment of 182 and 82 Welds Used in PWR Components", In: 10th Int. Conference on Environmental Degradation of Materials in Nuclear Power Systems – Water Reactors, Lake Tahoe, NV, USA, August 5-9th, 2001.
[12] T. Couvant and e. al., "Initiation of PWSCC of Weld Alloy 182", In: Proceedings of the 16th International Conference on Environmental Degradation of Materials in Nuclear Power Systems – Water Reactors, Asheville, NC, USA, August 11-15th, 2013.
[13] J. Bai, S. Ritter, H.-P. Seifert, and S. Virtanen, Corrosion Science, 2018, 131, 208-222.
[14] P.L. Andresen, "Short Crack Response of Alloy 182 Weld Metal Undergoing Stress Corrosion Cracking in High-Temperature PWR Primary Water", Report No. 1020981, EPRI, Palo Alto, CA, 2010.
[15] P.L. Andresen, "Short Crack Response of Alloy 182 in PWR Primary Water", Report EPRI, Palo Alto, CA.
[16] J. Bai, S. Ritter, H.-P. Seifert, and S. Virtanen, Corrosion Engineering, Science and Technology, 2017, 52(8), 558-566.
[17] J. Bai, S. Ritter, H.-P. Seifert, M. Vankeerberghen, and R.-W. Bosch, "The Use of Tapered Specimens to Evaluate the SCC Initiation Susceptibility in Alloy 182 in BWR and PWR Environments", In: 18th Int. Conference on Environmental Degradation of Materials in Nuclear Systems - Water Reactors, Springer International Publishing, Part F11, Portland, OR, USA, ISBN: 23671181 (ISSN); 9783319684536 (ISBN), August 13-17, 2017.
[18] Z. Que et. al, Corrosion Engineering, Science and Technology, 2022, 57(5), 474-485.
[19] S. Ritter et al., "Effect of Surface Machining on the Environmentally-Assisted Cracking of Alloy 182 and 316L Stainless Steel in Light Water Reactor Environments – Results of the Collaborative Project MEACTOS", In: EUROCORR 2022, EFC, Berlin, Germany, August 28-September 1, 2022.
[20] S. Suman, H.P. Seifert, P. Spätig, and S. Ritter, "Stress Corrosion Cracking Behaviour of Thermally Aged Alloy 182", In: 20th Int. Conference on Environmental Degradation of Materials in Nuclear Power Systems – Water Reactors, Snowmass Village, Colorado, USA, July 17-21, 2022.
[21] M. Zimina and et. al., "Effect of Surface Machining on the Environmentally-Assisted Cracking of Alloy 182 and 316L Stainless Steel in Light Water Reactor Environments – Results of the Collaborative Project MEACTOS", Corrosion Reviews, 2023, 41.
[32] D. Kirk, "Effect of Plastic Straining on Residual Stresses Induced by Shot Peening", In: International conference of shot peening, Germany.
[33] R.C. McClung, Fatigue & Fracture of Engineering Materials & Structures, 2007, 30(3), 173-205.
[34] K. Dalaei, B. Karlsson and L.E. Svensson, Procedia Engineering, 2010, 2(1), 613-622.