O. Raquet[1], E. Herms[1], F. Vaillant[2], T. Couvant[2]
[1] Commissariat à l'Energie Atomique (CEA), DEN/DPC/SCCME/LECA, 91191 Gif-sur-Yvette Cedex, France
[2] Electricité de France (EDF), Les Renardières, R&D/MMC – 77250 Moret sur Loing – France

SCC OF COLD-WORKED AUSTENITIC STAINLESS STEELS IN PWR CONDITIONS

Keywords: stress corrosion cracking, cold work, austenitic stainless steels, hydrogen embrittlement, primary water

INTRODUCTION

Susceptibility of sensitized austenitic stainless steels to Stress Corrosion Cracking (SCC) is a well known phenomenon identified in Boiling Water Reactor (BWR) oxidizing environments. Research associated with the case of BWR-type conditions have shown certain minimum levels of chloride and oxygen are required for SCC susceptibility [1 - 4]. The detrimental role of cold-work on the resistance of austenitic stainless steels to SCC was also clearly recognized in pure water BWR-type conditions in the presence of limited amounts of oxygen [5 - 9]. Cold-work was generally identified to increase Crack Growth Rates (CGR) and reduce the estimated crack initiation time for austenitic stainless steels tested in pure water containing oxygen at a temperature around 300°C. The electrochemical potential usually measured for austenitic stainless steels exposed to hydrogenated Pressurized Water reactors (PWR) conditions appears well below the critical potential recognized for SCC initiation and too low to promote SCC susceptibility of these materials in absence of a significant amount of specific impurities. Consequently, solution annealed austenitic stainless steels, essentially of type AISI 304(L) and AISI 316(L), are generally considered as immune to SCC in hydrogenated primary water and these materials are thus widely used in PWRs. Some rather recent data [10, 11] have however demonstrated that austenitic stainless steels specimens that were severely cold-worked were susceptible to SCC in hydrogenated PWR environment under CERTs conditions**.** Failures of pressurisers heaters rods by SCC constructed from cold-worked austenitic stainless steels were additionally observed in several French power plants (12). The resistance of austenitic stainless steels to SCC in PWR conditions should be therefore re-evaluated with regard to the potential specific effects of cold-work. The objective of this paper is to detail the results obtained on SCC of cold-worked austenitic stainless steels. The main focus of this program is to assess SCC initiation susceptibility of steels considering the respective effects of cold-working mode and environmental parameters derived from standard PWR conditions.

EXPERIMENTAL PROCEDURE

Material

Commercial purity AISI 304L plate material (30 mm in thickness) was the main material used for the study. The average grain size was 50 µm. The level of ferrite (measured by Ferriscope®) was between 5.5 % and 6.3 % in the middle of the plate and between 2% and 3% near the surfaces. Chemical composition of the 304L stainless steel used is given in Table I. Some tests were also performed on commercial AISI 316L stainless steels with the composition given in Table II. The samples were fully annealed at 1050°C for 30 minutes under vacuum then quenched by argon in order to homogenize the austenitic phase and erase the prior cold work associated with sample manufacturing. After annealing the level of ferrite measured on test samples was approximately 3%. The annealing treatment was carried out before application of any cold-working procedure.

Table 1. *Chemical composition 304L*

304L	**C**	**Cr**	**Ni**	**Mo**	**Mn**	**Si**	**P**	**S**	**N**
W%	0.026	19.23	9.45	0.24	1.49	0.52	0.027	0.002	0.064

Table 2. *Chemical composition 316L*

316L	**C**	**Cr**	**Ni**	**Mo**	**Mn**	**Si**	**P**	**S**	**N**
W%	0.027	17.2	12.15	2.34	1.76	0.48	0.23	0.001	0.064

PWR test procedure

All of the tests performed in the framework of this study were carried out inside dedicated static 316L autoclaves including direct measurement of hydrogen partial pressure via the use of in-situ Ag/Pd probes. Hydrogen concentrations were in the range of 25 to 35 cc/kgH_2O.STP in accordance with the usual PWR specification for primary water. Some specific tests were performed at low (1 cc/kgH_2O.STP) or high (100 cc/kgH_2O.STP) dissolved hydrogen concentrations. Pollutants (chloride, fluoride, sulfate) were measured after each test using ionic chromatography in order to verify that the concentration of each pollutant remained below the value of 50 ppb usually specified for PWR primary water.

Examination of the fracture surface was performed using Scanning Electron Microscopy (SEM) in order to determine the morphology of fracture surface and the average depth of SCC. When mentioned, the average SCC growth rate was defined and calculated as follows: average SCC growth rate (µm/hour) = maximum crack depth on fracture surface (µm) / time to failure (hours).

Notched specimens

In order to analyze the effect of stress triaxiality on the susceptibility of the materials to SCC, some tests were performed on smooth specimens with a circumferential V-notch perpendicular to the tensile axis. The stress triaxiality coefficient induced by the notch was evaluated to be 1.55 (as per Bridgman formula). These specimens were fully annealed after machining the V-notch and thus no residual cold-work was initially present in material before testing in the water environment. The tensile/relaxation test was performed in nominal primary water conditions at 360°C.

CERTs on specimens cold-worked by fatigue

The specimens cold-worked by fatigue were tested by CERT at 360°C and $1x10^{-7}s^{-1}$. All specimens were manufactured from AISI 304L of the composition mentioned above. These specimens were extracted from 8 mm diameter, 16 mm long samples previously cold-worked by fatigue at ambient temperature. The characteristics of the cyclic pre-loading procedure applied in the first stage of sample preparation were the following: tensile-compressive loading cycle, total strain $\pm$ 0,024 , extension rate $4x10^{-3}s^{-1}$, 50 cycles. A small size tensile specimen was subsequently machined from inside the gage length of the initial fatigue sample. As a result of this procedure, the whole thickness of the 4 mm diameter tensile specimens extracted from the initial fatigue samples was cold-worked. The resulting surface hardness was 320HV (1.96N) and Ferriscope® measurements indicated the formation of 6% of martensite phase from austenite in the cold-worked material. These specimens were then tested in CERTs at 360°C at a constant extension rate of $1x10^{-7}s^{-1}$. In order to obtain information on the role of martensitic phases formed during cold-working regarding the susceptibility of the material to SCC, some specimens were extracted from samples previously cold-worked by fatigue at 200°C. At this temperature, the formation of martensite by cold-working is not anticipated. In this latter case, the initial surface hardness of the material was lower compared to specimens cold-worked at laboratory temperature: 255HV (1.96N) instead of 320HV (1.96N) due to the absence of martensitic transformation.

CERTs on specimens cold-worked by shot-peening

Shot-peening was selected to reproduce practical cases of superficial cold-working. The procedure for shot-peening provided a high initial surface hardness of 474HV (0.49N). The material hardness decreased with depth as shown in Figure 1 (profile of micro-hardness measurements). The cold-worked layer had a total depth of 250 µm. The maximum residual stresses in surface of the material were in the range of 850 MPa (compressive). During a CERT test, the evolution of stress versus the applied strain was rather different inside the outer cold-worked layer compared to the stress variation evaluated in the bulk of the specimen. The increase in stress value was rapid for low values of strain in the cold-worked layer and during the majority of the CERT test duration; further evolution of stress inside the cold-worked layer was very slow after the first initially observed increase (see Figure 2). In order to study the specific effect of the initial surface hardness on material susceptibility to SCC, dedicated specimens were prepared from shot-peened samples to obtain progressively lower surface hardnesses. Electropolishing was used to eliminate a part of the outer hard cold-worked layer resulting from shot-peening and finally obtain test specimens with initial surface hardnesses of 300HV (0.49N) and 270HV (0.49N) respectively. All specimens were made from AISI 304L of the composition given earlier and were tested by CERT at 360°C and $1x10^{-7}s^{-1}$.

With the aim to determine the effects of the dissolved hydrogen on the susceptibility of the material to SCC, some specimens that had been cold-worked by shot-peening (initial hardness 474HV) were tested by CERT in primary water of nominal composition at 320°C with 3 different dissolved hydrogen levels: 1 cc H_2/kgH_2O.STP, 30 cc H_2/kgH_2O.STP and 100 cc H_2/kgH_2O.STP.

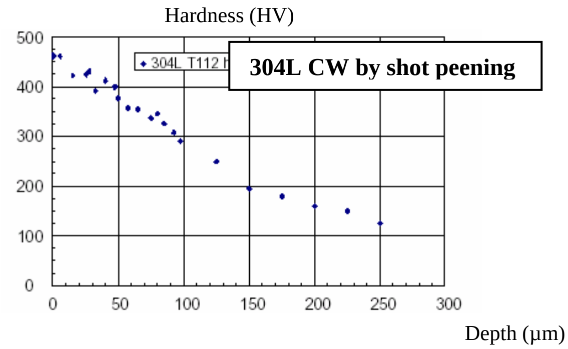

Fig. 1. profile of micro-hardness measurements on specimen cold-worked by shot-peening

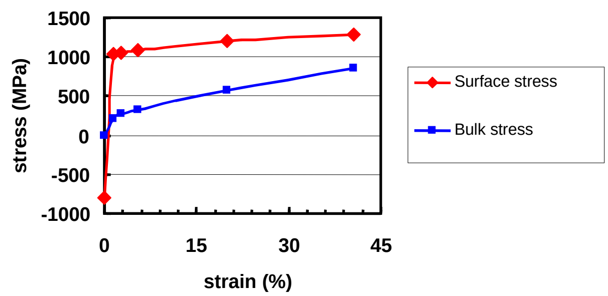

Fig. 2. Schematic evolution of stress inside a specimen cold-worked by shot-peening during a tensile test

Constant load tests

These tests were performed on AISI 304L samples of the composition mentioned earlier. The procedure for specimen cold-working was based on those used for CERTs. The objective of the long term constant load tests described here was to assess the susceptibility of austenitic stainless steel material under static conditions. One constant load test was performed on a specimen cold-worked by shot-peening (initial surface hardness 474HV (0.49N)). The axial stress applied was 550 MPa which corresponds to a total deformation of 18%. The second test was performed on a cylindrical smooth specimen previously cold-worked by fatigue at ambient temperature (same procedure as above for cyclic pre-loading and subsequent manufacturing of a tensile specimen from within the gage length of the fatigue sample). An additional circumferential V-notch was machined around the specimen to increase the severity of the test. The initial surface hardness before testing was 340HV (1.96N). The load applied corresponds to 80% of the maximum value reached during previous CERT tests on the circumferential V-notched specimens previously mentioned. The constant load tests were carried out at 360°C in nominal primary water conditions.

Pre-charging tests in primary water conditions

Hydrogenated primary water of PWRs will cause progressive hydrogen charging of metallic materials exposed for long periods to PWR environment. The dissolved hydrogen present in primary water of PWRs is in equilibrium with 30 appm hydrogen in austenitic stainless steel [13] as per Sievert's law. According to some references, the practical levels of atomic hydrogen present inside austenitic alloys after long exposure to PWRs conditions could be significant: 750 appm is a value reported as typical for austenitic stainless steels irradiated at very low fluence [12] and the values reported appear independent of irradiation fluence. Brittle cracking associated with hydrogen-induced embrittlement is however generally observed in the low temperature domain. Therefore, in order to check the possibility of embrittlement on materials exposed to primary water conditions, the selected testing procedure consisted firstly of exposing (no applied stress) 304L austenitic stainless steels samples to hydrogenated primary water for long periods at high temperature (320°C) and then tensile testing the pre-exposed specimens at lower temperature in air. The fracture surfaces were analyzed after testing to identify any evidence of embrittlement. An initial CERT test was carried out at 20°C in air on a sample pre-exposed to primary water during 2000 hours. The deformation rate selected was $5x10^{-5}s^{-1}$. As the observations of fracture surface were conclusive, two other CERT tests were carried out on 304L specimens after exposure for 2000h at 360°C. The tensile tests were achieved at 140°C in primary water at two distinct deformation rates. In this latter case, cold-working was performed by cold rolling (50% of thickness reduction) with the aim of confirming results obtained above using a second cold-working mode. In the case of SCC, it is well known that the susceptibility of metallic materials increases when the deformation rate decreases. The first test was performed with a deformation rate sufficiently fast ($1x10^{-6}s^{-}1$) to eliminate the possibility of SCC and promote generation of cracks by hydrogen-induced embrittlement mechanism. The second test was performed at lower deformation rate ($5x10^{-8}s^{-1}$) in an attempt to promote susceptibility to SCC and discriminate the respective effects of H-induced embrittlement (1st test) and SCC (this 2nd test).

RESULTS AND DISCUSSION

The main obtained results are summarized in the following Table 3.

Table 3. *Results of SCC tests carried out on cold-worked 304L and 316L – (CL= constant load test), CERT: Constant Elongation Rate tests*

Test	CW state	Material	T (°C)	Results
CERT	V-notch, annealed after machining	304L	360	IGSCC
CERT	Shot-peening	304L	320, 360	TGSCC
CERT	Fatigue	304L	360	TGSCC
CL	Shot-peening	304L	360	TGSCC initiation
CL	Fatigue + notch	304L	360	TGSCC initiation

SCC susceptibility

Tensile testing of austenitic stainless steels in conditions promoting significant strain/stress triaxiality (notch) appeared to have a decisive effect on material susceptibility to cracking even in the absence of noticeable cold-work. Smooth specimens circumferentially notched and then annealed revealed considerable susceptibility to cracking by SCC when tested in tensile/relaxation conditions. A typical view of a fracture surface obtained after testing these specimens is given on Figure 3. The initiation of cracks is located inside the notch and the mode of fracture obtained is intergranular with a shallow transgranular initiation area. The average crack growth rate based on fracture surface examination was close to 0.3 µm/h (crack depth around 500 µm). The mentioned transition of fracture mode could possibly be correlated with the variation of stress intensity factor K associated with crack propagation during the test. Strain localization associated with stress/strain triaxiality combined with slow strain rate testing conditions appears therefore sufficient to give rise to a noticeable susceptibility of annealed austenitic stainless steel to cracking by SCC under PWR primary water conditions. This conclusion is of importance when considering the possible presence of surface defects of various origins on components exposed to PWR conditions.

Smooth AISI 304L specimens previously cold-worked by fatigue are strongly susceptible to cracking during CERTs. Fracture surface of specimens previously cold-worked by fatigue at ambient temperature showed large zones of SCC propagation after CERT tests in primary water at 360°C. The morphology of cracking was mixed (intergranular and transgranular) with a dominant transgranular fracture mode. The measured average crack growth rate was over 1.6 µm/h (crack depth generally around 1 mm). Specimens previously cold-worked at 200°C and tested by CERT in primary water at 360°C showed that there was no major effect of the pre-existing martensite phase on SCC susceptibility in these experiments. The extension of cracking is similar on specimens containing no martensite (cold-worked by fatigue at 200°C before CERTs) with respect to specimens containing martensite (previously cold-worked at ambient temperature). A very limited role of the martensite phase on SCC susceptibility was similarly underlined by Andresen [8] after testing austenitic stainless steels in pure water under oxygenated and hydrogenated conditions.

Specimens cold-worked by shot-peening and tested by CERT in primary water at 320°C and 360°C showed strong susceptibility to cracking. As mentioned above, the existence of a residual compressive stress state before tensile tests in primary water promote significant susceptibility to SCC. Noticeable SCC cracking was identifiable on the specimen surfaces (see Figure 4). The SCC propagation mode was transgranular and the measured average crack growth rate from fracture surface was greater than 1 µm/h (crack depth close to 1 mm). A strong influence of the initial surface hardness on SCC susceptibility was observed. Figure 5 shows the evolution of the average crack growth rate depending on initial surface hardness measured before the CERTs. The average crack growth rate increase significantly at high surface hardness. A criterion of 300 ±10HV (0.49N) can be proposed as a minimum hardness required to promote SCC cracking susceptibility of austenitic stainless steels (this value of hardness corresponds to an initial yield strength of 850 MPa for a 304L stainless steel). This criterion is quite similar to that was proposed by Tsubota and col. [7] for the SCC of austenitic stainless steels in pure water containing oxygen at 288°C. These authors observed that SCC of 304L in creviced bent beam conditions (CBB tests) when immersed in pure water saturated with oxygen occurred in specimens with a critical surface hardness of 270HV. The values of hardness mentioned above are clearly beyond the values authorized by usual codes (ASME or RCC-M) relevant to PWRs but could in practice originate from various causes like improper surface treatment (grinding).

Influence of mechanical loading conditions

The specimens tested under constant load in primary water at 360°C did not show significant susceptibility to SCC. Slight initiation was only observed in the case of the two modes of cold-work previously applied: shot-peening and fatigue, after long periods of tests. Figure 6 show the SCC initiation cracks observed by SEM inside the notch surrounding the specimen. This sample was cold-worked by fatigue at ambient temperature, notched and then tested under constant load in primary water for a total exposure of 17,000 hours. The procedure applied during sample preparation represents the most severe conditions identified during CERTs tests described above. In the absence of any failure and in order to determine the depth of the identified initiated SCC cracks, this specimen was opened by fatigue in air and the fracture surface observed by SEM. These latter observations showed that the maximum crack depth identified on the fracture surface was lower than 20 µm. The morphology of cracking was fully transgranular. The same behavior was observed on a shot-peened specimen tested under constant load: initiation of SCC cracks was observed after 4000 hours of test but no additional propagation of these initiated cracks was reported after 7,000 hours total test duration. Observation of cross sections of the tested sample showed that the maximum depth of the initiated crack was less than 20 µm. This slight susceptibility of cold-worked austenitic stainless steels to SCC propagation under constant load is astounding considering the strong tendency of these materials to cracking in CERT. Taking into account the specific role of the deformation rate on SCC phenomena, the absence of crack propagation must be associated with the very low deformation rate expected during constant load tests. Measurements performed during testing indicate that strain rate decreased progressively with time in the range of 10^{-9}-$10^{-10}s^{-1}$. Austenitic stainless steels are characterized by very low creep rate in the temperature range 200-450°C [19]. The deformation rate resulting from constant load testing conditions appears too low to promote crack propagation.

Compared to constant load tests, the susceptibility of austenitic stainless steels to SCC under constant deformation is questionable. No significant cracking was observed in the case of the previously cold-worked AISI 304L or 316L samples after a total test duration of 9,000 hours in primary water at 360°C. Slight, widely dispersed initiation of surface defects was only observed in the case of AISI 304L and 316L previously cold-worked by shot-peening after 5,600 hours of test but no further propagation was reported. No initiation was seen on notched specimens cold-worked or not. It is important to note that a slight tendency to defect initiation was identified only in the case of heavily cold-worked specimens showing highest initial surface hardness among all of those resulting from the cold-working procedures evaluated during constant deformation tests.

It must be underlined that, in the context of this study, SCC of austenitic stainless steels appeared strongly dependent on dynamic straining conditions (CERTs). Under static conditions or for an excessively low deformation rate corresponding to the creep rate calculated at 360°C, no propagation of SCC cracks was reported but slight crack initiation was detected. A consequence of these results is an expected susceptibility of cold-worked austenitic stainless steels to SCC in primary water under cyclic loading conditions. Most of results available in the open literature dealing with SCC of austenitic stainless steels in PWRs conditions and where evidence of significant SCC propagation is identified, are obtained under dynamic deformation conditions. Crack propagation in pre-cracked fracture mechanics CT-type specimens was for instance observed under cyclic loading (periodic unloading stages) [16, 17] or trapezoidal wave loading [18, 20]. A minimum deformation rate appears therefore, according to actual and published results, as a pre-requisite for SCC propagation on cold-worked austenitic stainless steels exposed to PWRs conditions.

Effect of dissolved hydrogen level on SCC

Dedicated CERTs were performed on 304L previously cold-worked by shot-peening in primary water at 320°C at 3 levels of dissolved hydrogen: 1 cc H_2/kgH_2O.STP, 30 cc H_2/kgH_2O.STP and 100 cc H_2/kgH_2O.STP. The results of these tests are given in Figure 7. SCC was observed at the 3 concentrations of dissolved hydrogen. No increase of the measured crack growth rate was seen at high dissolved hydrogen level but a significant reduction (10%) of the average crack growth rate was observed for the lowest dissolved hydrogen tested. The results obtained in the framework of this study confirm those produced by Arioka [14, 15] for a restricted range of dissolved hydrogen (Figure 7). Similar trend was observed after SCC tests on irradiated specimens of cold-worked austenitic stainless steels in PWR conditions [14-16].
The fracture mode was transgranular for 1 cc H_2/kgH_2O.STP and for 30 cc H_2/kgH_2O.STP. A transition of fracture mode was observed in the case of the highest level of dissolved hydrogen (100 cc H_2/kgH_2O.STP) with evidence of a mixed fracture mode (intergranular and transgranular) – see Figure 8.

Evidence of absorbed hydrogen embrittlement

Evidence of hydrogen embrittlement was observed after a tensile test performed in air at ambient temperature on a 304L specimen previously exposed to primary water conditions for 2000h. The 304L specimen was initially cold-worked by shot-peening (surface hardness 474 HV). Surface micro-cracking was observed after testing and was associated with H-induced embrittlement in the low temperature range. A dense cracking network was observed on the specimen surface over the entire gage length of the sample (Figure 9). Cracks were brittle in appearance and perpendicular to the main stress direction applied during tensile test. The depth of these cracks was small (<10µm) and the fracture mode fully transgranular. This type of damage is characteristic of hydrogen embrittlement on austenitic stainless steel. No surface cracking was observed on a specimen before tensile testing nor on a similar specimen tested in inert gas. It is clear that generation of this cracking was correlated with the presence of atomic hydrogen inside the material as a consequence of the pre-exposure period in primary water conditions. Austenitic stainless steels are usually considered resistant to hydrogen embrittlement phenomena as the diffusion rate of hydrogen is low and solubility high in the austenitic fcc phase. A limited form of embrittlement was however observed in this study after severe testing performed on strongly cold-worked 304L stainless steel at low temperature. Hydrogen concentrations measured after melting in the reference specimens exposed in the same autoclave than tensile specimens appeared to be very low: 1 to 3 ppm. The low level of hydrogen measured is compatible with the fugacity of dissolved hydrogen in primary water conditions; it may be assumed that localization of hydrogen in the cold-worked metallic network is the key parameter to explain susceptibility of 304L to cracking. Dedicated examinations of the specimens are in progress to examine this hypothesis. The data described above were confirmed by tensile tests performed at 140°C on 304L cold-worked by cold rolling and pre-exposed for 2000h to hydrogenated primary water at 360°C. The respective deformation rates were $1x10^{-6}s^{-1}$ and $5x10^{-8}s^{-1}$. The same type of cracking was observed after tests on the whole specimen surface. The fracture mode was transgranular (Figure 10) and the depth of cracking measured on the fracture surface remained low (<10µm). Consequently, no worsening of the SCC susceptibility was observed in the case of the test carried out at the lowest deformation rate ($5x10^{-8}s^{-1}$). One can conclude on the basis of this latter test that at 140°C the initiation of cracks by a hydrogen embrittlement mechanism does not promote subsequent propagation by SCC. Propagation of SCC cracks seems to be prevented by a temperature that is sufficiently low. The threshold of temperature for the occurrence of SCC is a critical parameter regarding both the application for cold-worked austenitic stainless steels in PWR primary water and understanding the SCC processes. Other experiments are in progress in order to detail the evolution of the cold-worked materials susceptibility to SCC with temperature.

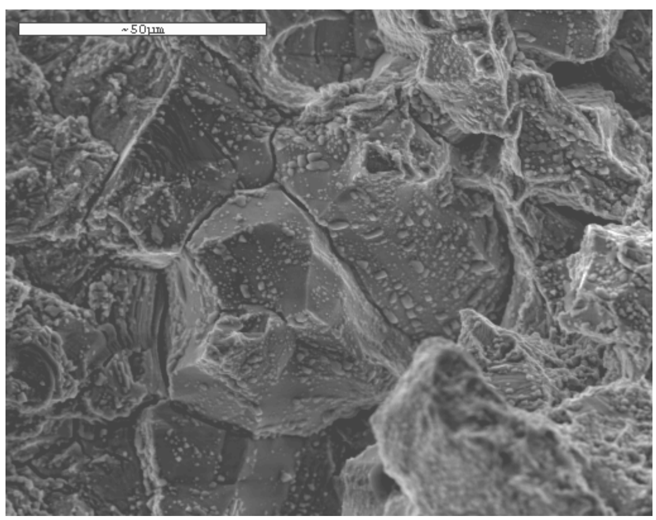

Fig. 3. Fracture surface obtained after CERT on a notched specimen (full annealing performed after machining of the notch); note the intergranular nature of SCC without cold-work, 304L, PWR primary water 360°C, time to fracture 2000 h

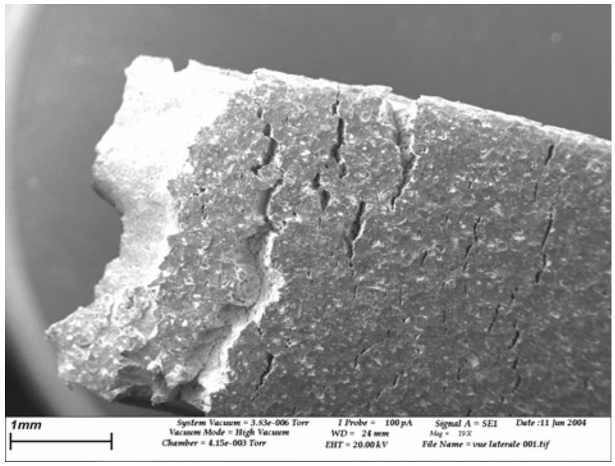

Fig. 4. Specimen surface after CERT ($1.5 \times 10^{-7} s^{-1}$) of a specimen previously cold-worked by shot-peening, 304L, PWR primary water 320°C, time to fracture 300 h

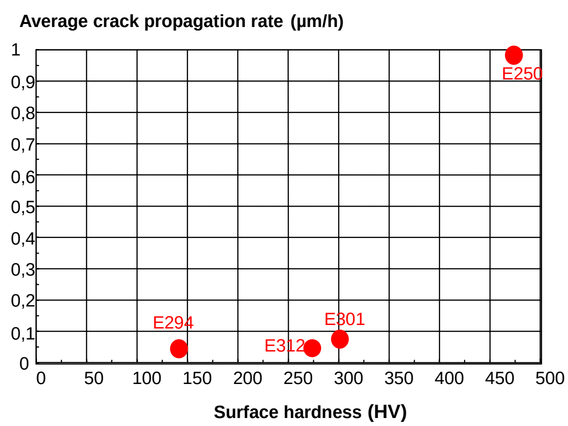

Fig. 5. Evolution of the average crack growth rate with respect to the initial surface hardness, cold-worked specimens (shot-peening) tested by CERT in primary water, $1.5 \times 10^{-7} s^{-1}$, 360°C

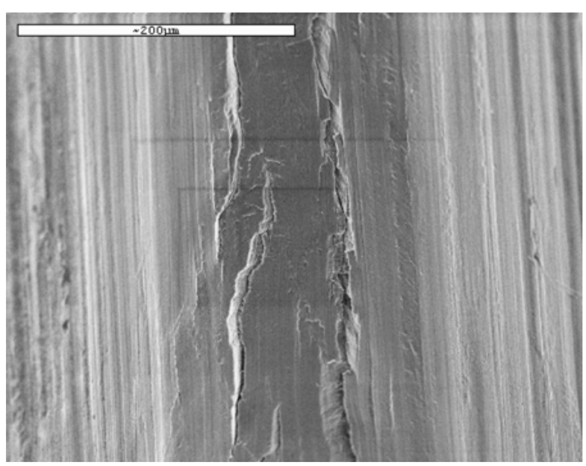

Fig. 6. Crack initiation observed inside the notch on a specimen initially cold-worked by fatigue and then tested under constant load (850 MPa); cross sections show that maximum crack depth is less than 20 µm, PWR primary water, 360°C, 17,000 h testing

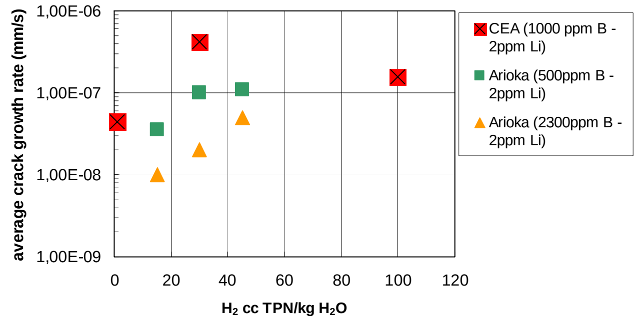

Fig. 7. Evolution of the measured average crack growth rate with respect to the dissolved hydrogen level – CERT tests on specimens cold-worked by shot-peening, comparison with results reported by Arioka [13]

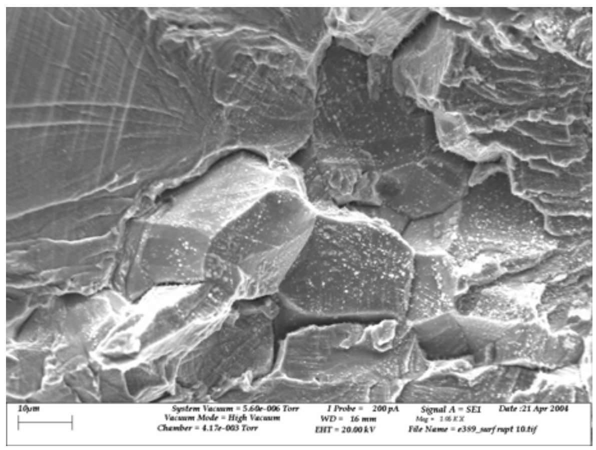

Fig. 8. Fracture surface of specimen previously cold-worked by shot-peening and tested by CERT($1.5x10^{-7}s^{-1}$), primary water at 320°C, hydrogen level 100 cc H_2/kgH_2O.STP, intergranular and transgranular fracture, time to fracture 500 h

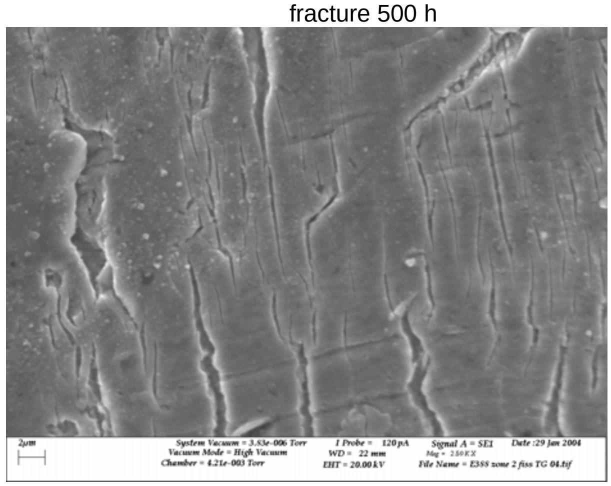

Fig. 9. Fracture surface showing an area of transgranular brittle fracture observed on a specimen previously cold-worked by cold rolling, pre-exposed during 2000h in hydrogenated primary water at 360°C and tensile tested ($5x10^{-8}s^{-1}$) in primary water at 140°C

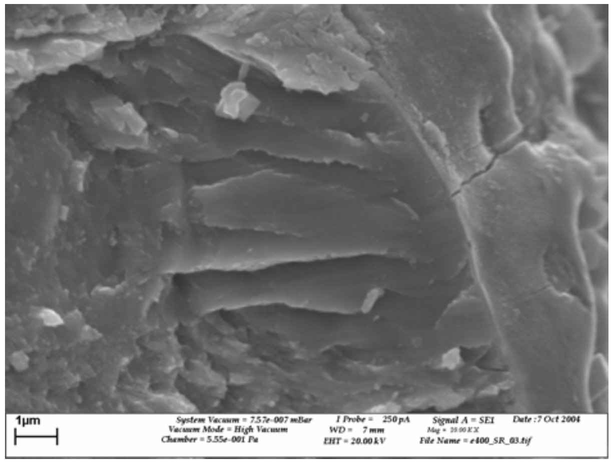

Fig. 10. Surface cracking observed on a specimen previously cold-worked by shot-peening, pre-exposed during 2000h in hydrogenated primary water at 320°C and tensile tested ($5x10^{-5}s^{-1}$) in air at laboratory temperature

CONCLUSIONS

1. A strong susceptibility for SCC of heavily cold worked austenitic stainless steels is observed in hydrogenated primary water typical of PWRs. This susceptibility to cracking increases with the extent of cold-work and/or localization of deformation. The levels of cold-work involved in this study are very high when compared to the maximum cold-work levels required by the usual international codes relevant to PWRs (ASME and RCC-M). These high values of cold-work could be eventually encountered on the surface of some components (grinding) or in the case of improper manufacturing.

2. The specific cold-work procedures including a compressive stage (fatigue, shot-peening) strongly favors SCC susceptibility in PWRs conditions under dynamic deformation.

3. For a given cold-working procedure, SCC susceptibility of austenitic stainless steels materials increases with an increasing intensity of cold-work. A threshold of SCC susceptibility was identified in the case of the shot-peening procedure of cold-working for AISI 304L stainless steels through the value of the initial surface hardness before SCC testing (CERT). SCC crack propagation is thus only observed beyond 300±10HV on shot-peened specimens.

4. SCC initiation was observed but no propagation was noted under static conditions (constant load, constant deformation). Dynamic deformation conditions (e.g. CERTs, cyclic loading) appear as a prerequisite for SCC susceptibility of cold worked austenitic stainless steels in PWRs.

5. In the case of strongly cold-worked austenitic stainless steels previously exposed to hydrogenated PWR primary water, limited hydrogen embrittlement is observed after tensile testing in the low temperature domain. This form of hydrogen induced damage could be envisaged when considering the risk of cracking on cold-worked components exposed for long periods to PWR conditions.

REFERENCES

1. B.M. Gordon, "The effect of chloride and oxygen on the stress corrosion of stainless steels: review of literature", *Materials Performance*, 19 (4) 1980.
2. B.M. Gordon, "Corrosion in Boiling Water Reactors", ASM Metals Handbook vol. 13 Corrosion, (Metal Park, Ohio: ASM International, 1987).
3. M.O. Speidel, *Corrosion of stainless steels* (A.J. Sedricks, Wiley-Interscience publication) 1979.
4. M.O. Speidel, R. Magdowski, 9th International Symposium on Environmental of Materials in Nuclear Power Systems – Water Reactors (Edited by F.P. Ford, S.M. Bruemmer and G. Was, TMS, 1999), 325.
5. J. Kuniya, I. Masaoka, R. Sasaki, *Corrosion* 44 (1) (1988), pp. 21-28.
6. H. Hanninen, "Effect of sensitization and cold work on stress corrosion susceptibility of austenitic stainless steels in BWR and PWR conditions" (VTT Metals Laboratory Report 88, May 1981).
7. M. Tsuboda, Y. Kanazawa, I. Hitoshi, 7th International Symposium on Environmental of Materials in Nuclear Power Systems – Water Reactors, August 7-10, 1995, Breckenridge, Colorado.
8. P. Andresen, T.M. Angeliu, W.R. Catlin, L.M. Young, R.M. Horn, Corrosion 2000 conference, paper 203 (NACE International publications).
9. P. Andresen, T.M. Angeliu, L.M. Young, Corrosion 2001 conference, paper 228 (NACE International publications).
10. Z. Szklarska-Smialowska, S.W. Sharkawy, Z. Xia, *Journal of Nuclear Materials* 195 (1992), pp. 184-190.
11. Z. Szklarska-Smialowska, S.W. Sharkawy, Z. Xia, *Corrosion* 48 (6) (1992), pp. 455-462.
12. T. Couvant, P. Moulart, L. Legras, P. Bordes, J. Capelle, Y. Rouillon, T. Balon, Colloque International Fontevraud 6, Fontevraud, France, septembre 2006 (SFEN Publications).
13. E.P. Simonen, S.M. Bruemmer, Corrosion 97 conference, paper 106 (NACE International publications).
14. K. Arioka, Y. Kaneshima, T. Yamada, T. Terachi, 11th International Symposium on Environmental of Materials in Nuclear Power Systems – Water Reactors, Stevenson, WA, USA, 10-14 August 2003.

15. K. Arioka, Colloque International Fontevraud 5, Fontevraud, France, septembre 2002 (SFEN Publications).
16. M.L. Castaño, M.S. Garcia, G. de Diego, D. Gomez-Briceño, L. Francia, Colloque International Fontevraud 5, Fontevraud, France, septembre 2002 (SFEN Publications).
17. M.L. Castano, M.S. Garcia Redondo, G. de Diego Velasco, D. Gomez Briceno, 11th International Conference Environmental Degradation of Materials in Nuclear Systems – Water Reactors, Stevenson, WA, USA, 10-14 August 2003.
18. T. Shoji, G. Li, J. Kwon, S. Matshushima, Z. Lu, 11th International Symposium on Environmental of Materials in Nuclear Power Systems – Water Reactors, Stevenson, WA, USA, 10-14 August 2003.
19. S. Usami, T. Mori, Cryogenics, 40 (2) (2000), pp. 117-126.
20. F. Vaillant, T. Couvant, J.M. Boursier, C. Amzallag, Y. Rouillon, O. Raquet, 2004 Pressure Vessels and Piping Conference, San Diego, USA, 25-29 july 2004.