ELECTRIC POWER RESEARCH INSTITUTE

2011-102 BWR Vessel & Internals Project (BWRVIP)

May 17, 2011

Document Control Desk
U. S. Nuclear Regulatory Commission
11555 Rockville Pike
Rockville, MD 20852

Attention: Jonathan Rowley

Subject: Project No. 704 – BWRVIP Response to NRC Request for Additional Information on BWRVIP-138, Revision 1

Reference: Letter from Jonathan Rowley (NRC) to David Czufin (BWRVIP Chairman), "Request for Additional Information Re: BWRVIP-138, Revision 1: "BWR Vessel and Internals Project, Updated Jet Pump Beam Inspection and Flaw Evaluation Guidelines (TAC NO. ME2191)," dated June 2, 2010.

Enclosed are five (5) copies of the BWRVIP response to the NRC Request for Additional Information (RAI) on the BWRVIP report entitled "BWRVIP-138, Revision 1: BWR Vessel and Internals Project, Updated Jet Pump Beam Inspection and Flaw Evaluation Guidelines." The RAI was transmitted to the BWRVIP by the NRC letter referenced above.

Please note that the enclosed response contains proprietary information. A letter requesting that the response be withheld from public disclosure and an affidavit describing the basis for withholding this information are provided as Attachment 1. The response includes yellow shading to indicate the proprietary information. The proprietary information is also marked with the letters "TS" in the margin indicating the information is considered trade secrets in accordance with 10CFR2.390A.

Two (2) copies of a non-proprietary version of the BWRVIP response to the RAI are also enclosed. This non-proprietary response is identical to the enclosed proprietary response except that the proprietary information has been deleted.

If you have any questions on this subject please call Randy Schmidt (PSEG Nuclear, BWRVIP Assessment Committee Technical Chairman) at 856.339.3740.

Sincerely,

Dave Czufin
Exelon
Chairman, BWR Vessel and Internals Project

Additional copies sent to PM

G004
NRR

ELECTRIC POWER RESEARCH INSTITUTE

May 16, 2011

Document Control Desk
Office of Nuclear Reactor Regulation
U.S. Nuclear Regulatory Commission
Washington, DC 20555-0001

Attention: Jonathon Rowley

Subject: Request for Withholding of the following Proprietary Document:

BWRVIP Response to NRC Request for Additional Information (RAI) Regarding BWRVIP-138, Revision 1: BWR Vessel and Internals Project, Updated Jet Pump Beam Inspection and Flaw Evaluation Guidelines

To Whom It May Concern:

This is a request under 10 C.F.R. §2.390(a)(4) that the U.S. Nuclear Regulatory Commission ("NRC") withhold from public disclosure the information identified in the enclosed Affidavit consisting of the proprietary information owned by Electric Power Research Institute, Inc. ("EPRI") identified above (the "Report"). Proprietary and non-proprietary versions of the Correspondence and the Affidavit in support of this request are enclosed.

EPRI desires to disclose the Report in confidence to assist the NRC. The Report is not to be divulged to anyone outside of the NRC or to any of its contractors, nor shall any copies be made of the Report provided herein. EPRI welcomes any discussions and/or questions relating to the information enclosed.

If you have any questions about the legal aspects of this request for withholding, please do not hesitate to contact me at (650) 855-2164. Questions on the content of the Report should be directed to *Randy Stark* of EPRI at (650) 855-2122.

Sincerely,

Christine King

Christine King
Sr. Manager, Business & Operations

ELECTRIC POWER RESEARCH INSTITUTE

AFFIDAVIT

RE: Request for Withholding of the Following Proprietary Document:

BWRVIP Response to NRC Request for Additional Information (RAI) Regarding BWRVIP-138, Revision 1: BWR Vessel and Internals Project, Updated Jet Pump Beam Inspection and Flaw Evaluation Guidelines

I, Christine King, being duly sworn, depose and state as follows:

I am the Senior Manager of Business & Operations for the Nuclear Power Sector at Electric Power Research Institute, Inc. whose principal office is located at 3420 Hillview Avenue, Palo Alto, California ("EPRI") and I have been specifically delegated responsibility for the above-listed Report that is sought under this Affidavit to be withheld (the "Report"). I am authorized to apply to the U.S. Nuclear Regulatory Commission ("NRC") for the withholding of the Report on behalf of EPRI.

EPRI requests that the Report be withheld from the public on the following bases:

Withholding Based Upon Privileged And Confidential Trade Secrets Or Commercial Or Financial Information:

a. The Report is owned by EPRI and has been held in confidence by EPRI. All entities accepting copies of the Report do so subject to written agreements imposing an obligation upon the recipient to maintain the confidentiality of the Report. The Report is disclosed only to parties who agree, in writing, to preserve the confidentiality thereof.

b. EPRI considers the Report and the proprietary information contained therein (the "Proprietary Information") to constitute trade secrets of EPRI. As such, EPRI holds the Report in confidence and disclosure thereof is strictly limited to individuals and entities who have agreed, in writing, to maintain the confidentiality of the Report. EPRI made a substantial economic investment to develop the Report, and, by prohibiting public disclosure, EPRI derives an economic benefit in the form of licensing royalties and other additional fees from the confidential nature of the Report. If the Report and the Proprietary Information were publicly available to consultants and/or other businesses providing services in the electric and/or nuclear power industry, they would be able to use the Report for their own commercial benefit and profit and without expending the substantial economic resources required of EPRI to develop the Report.

c. EPRI's classification of the Report and the Proprietary Information as trade secrets is justified by the Uniform Trade Secrets Act which California adopted in 1984 and a version of which has been adopted by over forty states. The California Uniform Trade Secrets Act, California Civil Code §§3426 – 3426.11, defines a "trade secret" as follows:

"'Trade secret' means information, including a formula, pattern, compilation, program device, method, technique, or process, that:

(1) Derives independent economic value, actual or potential, from not being generally known to the public or to other persons who can obtain economic value from its disclosure or use; and

(2) Is the subject of efforts that are reasonable under the circumstances to maintain its secrecy."

d. The Report and the Proprietary Information contained therein are not generally known or available to the public. EPRI developed the Report only after making a determination that the Proprietary Information was not available from public sources. EPRI made a substantial investment of both money and employee hours in the development of the Report. EPRI was required to devote these resources and effort to derive the Proprietary Information and the Report. As a result of such effort and cost, both in terms of dollars spent and dedicated employee time, the Report is highly valuable to EPRI.

e. A public disclosure of the Proprietary Information would be highly likely to cause substantial harm to EPRI's competitive position and the ability of EPRI to license the Proprietary Information both domestically and internationally. The Proprietary Information and Report can only be acquired and/or duplicated by others using an equivalent investment of time and effort.

I have read the foregoing and the matters stated herein are true and correct to the best of my knowledge, information and belief. I make this affidavit under penalty of perjury under the laws of the United States of America and under the laws of the State of California.

Executed at 3420 Hillview Avenue, Palo Alto, California, being the premises and place of business of Electric Power Research Institute, Inc.

Date: May 16, 2011

Christine King

Christine King

BERTE A. DAHL
COMM. # 1926383
NOTARY PUBLIC - CALIFORNIA
SANTA CLARA COUNTY
MY COMM. EXP. MAR. 20, 2015

(State of California)
(County of Santa Clara)

Subscribed and sworn to (or affirmed) before me on this 16th day of May, 2011 by Christine King, proved to me on the basis of satisfactory evidence to be the person(s) who appeared before me.

Signature Berte A Dahl (Seal)

My Commission Expires 20th day of May, 2015

Non-Proprietary BWRVIP Response to NRC Request for Additional Information on "BWRVIP-138, Revision 1: BWR Vessel and Internals Project, Updated Jet Pump Beam Inspection and Flaw Evaluation Guidelines"

BWRVIP Response to NRC Request for Additional Information on "BWRVIP-138, Revision 1: BWR Vessel and Internal Project, Updated Jet Pump Beam Inspection and Flaw Evaluation Guidelines"

Each item from the NRC Request for Additional Information (RAI) is repeated below verbatim followed by the BWRVIP response to that item.

RAI 1

Has the BWRVIP considered the effect of surface condition on initiation of stress corrosion cracking? A paper from Foucault and Benhamou, "Influence of the Surface Condition on the Susceptibility of Alloy X-750 to Crack Initiation in PWR Primary Water" 1993, TMS 6th International Symposium of Environmental Degradation of Materials in Nuclear Power Systems-Water-Reactors is potentially relevant. This is not the boiling water reactor (BWR) environment, but it is possible that the surface condition is important in both BWR and pressurized water reactor (PWR) environment in terms of initiation.

BWRVIP Response to RAI 1:

The Alloy X-750 conclusions from the above Reference 1 for PWR environmental conditions are summarized as follows:

Alloy X-750 with high temperature solution annealing (1093°C [2000°F]) and single stage (704°C [1300°F]) HTH aging treatment is very resistant to SCC in high temperature primary water if any surface oxide layer due to heat treatment is removed by machining. However, this SCC resistance can be reduced by:

- The presence of sulfide inclusion clusters in the matrix
- The presence of certain surface layers built up during the last step of the manufacture such as the aging treatment

If the final step is the aging treatment, the nature of the atmosphere of the furnace has a direct influence on the behavior of the material when stressed in high temperature primary water. Treatment under vacuum can induce SCC. Heat treatment under nitrogen or machining after heat treatment leads to the best resistance to SCC. The exposure of these surfaces to primary water at high temperature modifies the surface conditions. Stress corrosion cracking may be then triggered when tensile stresses are applied. This phenomenon can be accounted for by the formation of grain boundary damage in the sub surface matrix during the aging heat treatment or during the exposure to primary water depending on the composition of the surface layer. The composition of the surface film may also have a direct influence on the initiation phase for SCC.

This effect of the surface layer on the SCC initiation explains the beneficial effect of polishing the control rod tube guide pins in the French PWRs and suggests that a better resistance may be obtained by oxidizing the surface in a well defined environment to create a surface layer enriched in Cr that protects the sub surface matrix against internal oxidation.

The BWR testing experience shows similar surface effects results albeit these studies only examined the effect of surface condition on subsequent general corrosion performance in BWR environments rather than SCC response. However, the identical effect/benefit of optimum surface condition would also be anticipated relative to SCC resistance.

Numerous BWR environmental general corrosion studies [2-5] have been performed on Alloy X-750 corrosion films in recognition of the fact that Alloy X-750 components (e.g., fuel springs) have a high corrosion rate in the highly oxidizing (O_2/H_2O_2) BWR core environment [6]. The inhibition of ^{58}Co and ^{60}Co directly generated from the release of ^{58}Ni and ^{59}Co, respectively, from Alloy X-750 general corrosion would be an effective method to reduce radiation exposure.

As was the case for PWR environments, it was believed that pre-oxidation of Alloy X-750 is a method to mitigate its subsequent aqueous corrosion in the BWR environment and, thus, reduce radiation dose rates while still maintaining structural margin. The creation of a thinner, more adherent and more ductile film would also increase Alloy X-750's resistance to IGSCC.

K. Tada, et al. examined air oxidation as a candidate surface treatment for mitigation of Alloy X-750 general corrosion in BWR type environments [2]. Oxide layers on Alloy X-750 were characterized to determine their relationship with subsequent corrosion using solution heat treated (SHT) Alloy X-750 in a vacuum followed by either air oxidation at 500-700°C (932-1292°F) for a maximum of 25 hours or autoclaved at 390°C (734°F) for 13 hours after aging as a control.

The results of the study revealed that the oxides varied in thickness from 30 nm (as-autoclaved) to 270 nm (as-oxidized plus immersion) and there were no obvious differences in oxides on pre-oxidized Alloy X-750 before and after short term 280°C (536°F) immersion. However, the oxide layer consisted of a thin film plus oxide particles on the film. The film was the spinel $FeNiFeO_4$ plus Cr_2O_3 and NiO while the oxide particles were also $FeNiFeO_4$. Specimens pre-oxidized for 10 hours at 500-700°C (932-1292°F) had 30-90% lower nickel release levels than the non-pre-oxidized material, Figure 1, and the nickel release level decreased with increasing pre-oxidation temperature and pre-oxidation time, Figure 2.

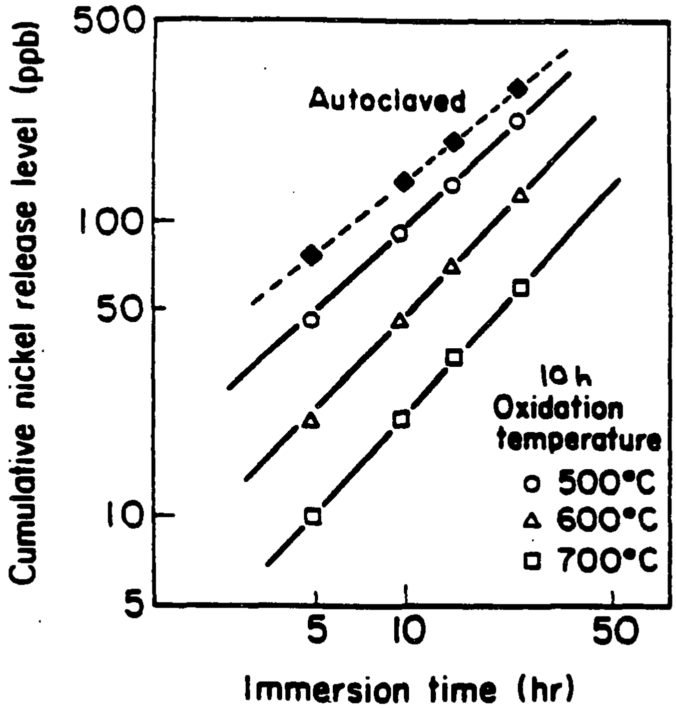

Figure 1. Cumulative Nickel Release from Autoclaved and Pre-oxidized Alloy X-750 [2]

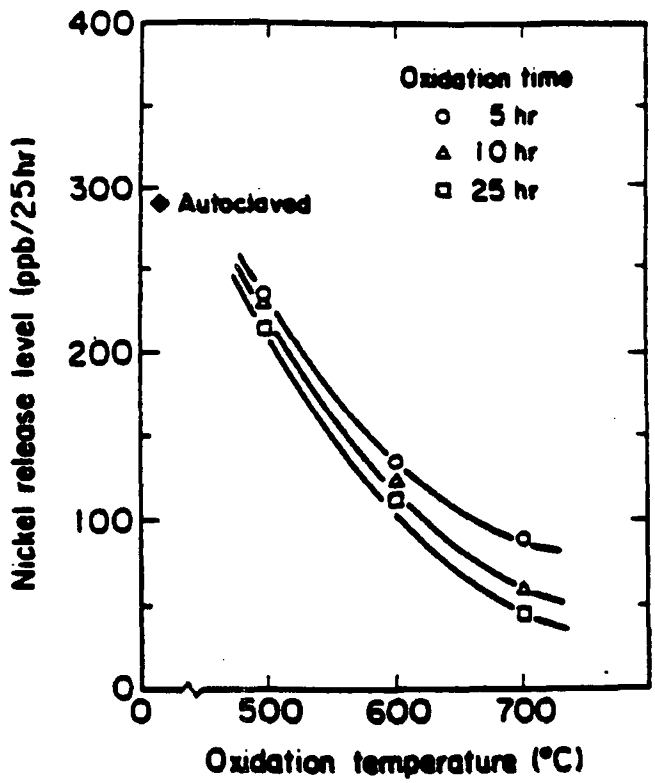

Figure 2. Effect of Pre-Oxidation Temperature on Nickel Release [2]

Y. Hemmi, et al. [3, 4] performed studies designed to compare the properties of Alloy X-750 oxide films grown in steam (390°C [734°F]/13 hours) with that grown in air (700°C [1292°F]/5 hours) relative to subsequent nickel release rate and oxide structure. The specimens were exposed to a BWR core type environment of 400 ppb dissolved oxygen, 60 ppb dissolved hydrogen and 200 ppb hydrogen peroxide at 270°C (518°F) with ^{60}Co gamma ray irradiation.

Y. Hemmi, et al. discovered that the highly protective oxide produced by air oxidation of Alloy X-750 consists of two distinct tightly adherent layers as illustrated in Figure 3:

- ~80% thermodynamically stable $FeNiFeO_4$ plus ~10% Cr_2O_3 and ~10% TiO_2 in the outer layer.
- ~65% Cr_2O_3 plus ~10% $FeNiFeO_4$, ~10% NiO and ~10% TiO_2 in the inner layer oxide film.

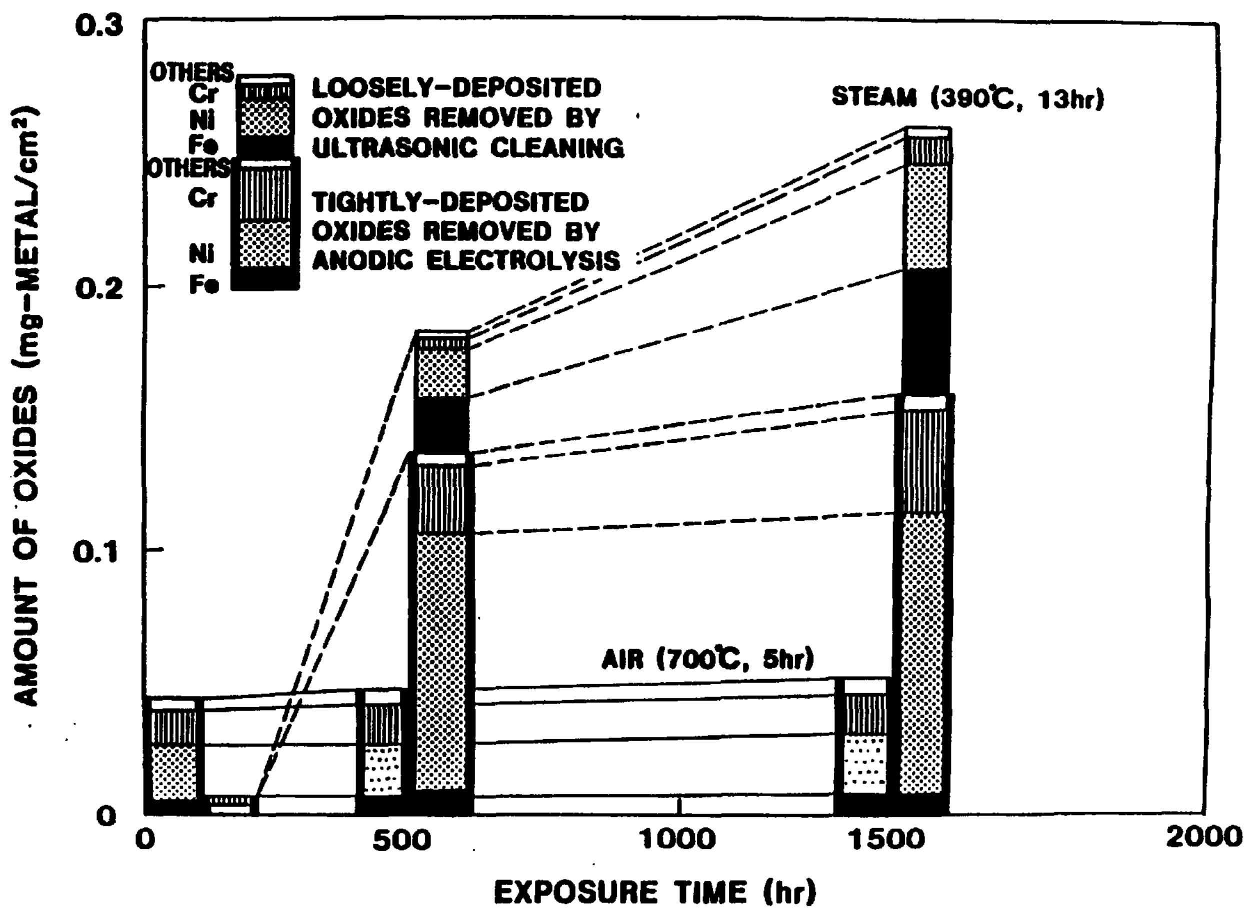

Figure 3. Comparison of Corrosion Oxides on Alloy X-750 [3, 4]

The less protective autoclave corrosion film also consists of loosely deposited assorted oxides of Cr, Ni and Fe plus tightly adherent, but more soluble ~60% NiO plus ~20% Cr_2O_3, ~10% $FeNiFeO_4$ and ~10% TiO_2, Figure 4.

The Alloy X-750 corrosion rate decreased with time with the thickening of the $FeNiFeO_4$ and Cr_2O_3 protective films after exposure to a simulated BWR environment, Figure 5. Due to its low solubility, Figure 6, the $FeNiFeO_4$ outer oxide layer restricts the dissolution of Alloy X-750 while the inner Cr_2O_3 layer restricts Ni/Co diffusion.

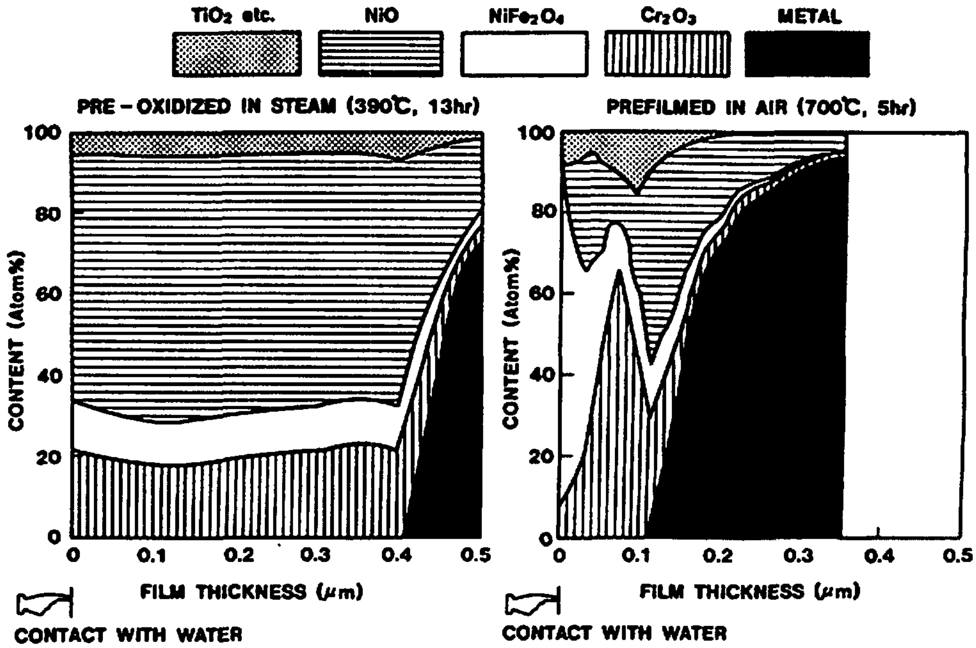

Figure 4. Distribution of Oxides formed on Pre-oxidized in Steam and Air on Alloy X-750 [3, 4]

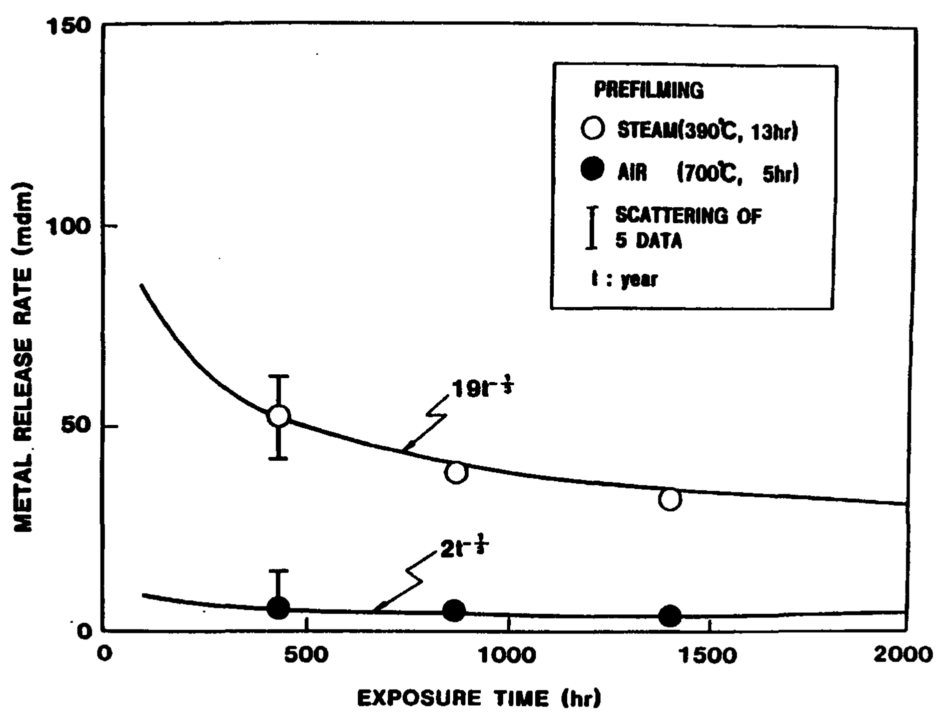

Figure 5. Metal Release Rate for Pre-oxidized in Steam and Air Alloy X-750 [3, 4]

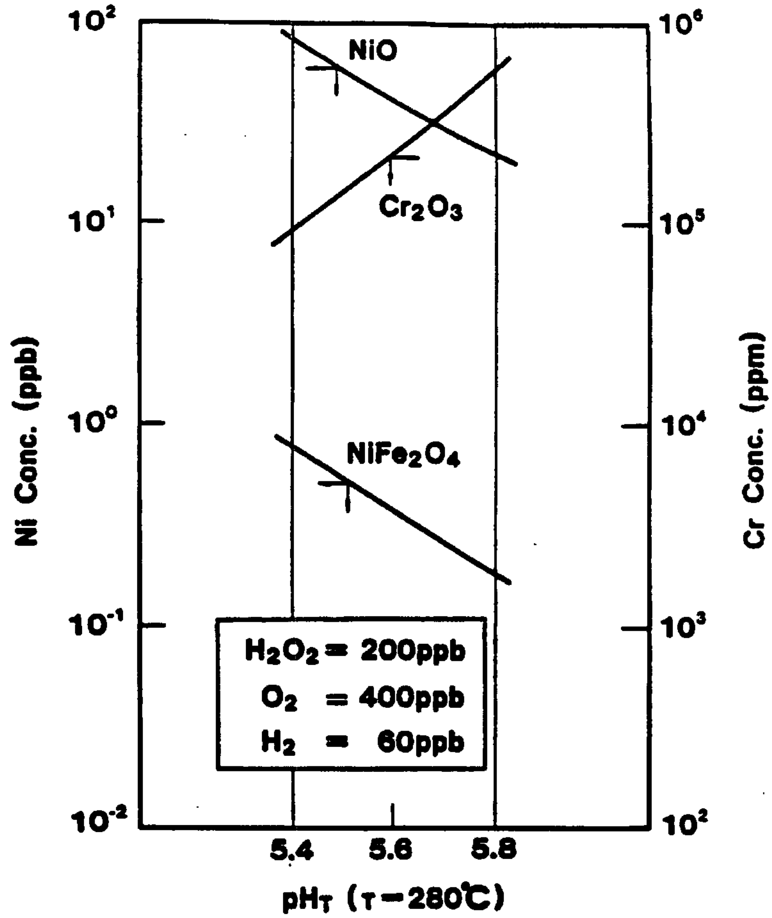

Figure 6. Solubility of Nickel and Chromium in BWR Water [3, 4]

E. Ahlberg and B. Rebensdorff examined the general corrosion properties of Alloy X-750 fuel spacer springs exposed to actual Swedish BWR and simulated BWR environments as a function of heat treatment and atmosphere [5]. Three types of fuel spacer springs were examined:

1. Type I. Aged at 705°C (1300°F)/20 hours in a vacuum (furnace initially air-filled during preheating) and air-cooled resulting in a varying thickness (~400 nm) oxide with low Cr and high Ni, Mn, Ti and Fe. The subsequent general corrosion rate in-core (seven years) and in-autoclave (200 ppb O_2, 10 ppm H_2O_2, 150°C [302°F]) was considered "acceptable."
2. Type II. Aged at 705°C (1300°F)/20 hours in a vacuum and helium cooled resulting in a shiny spring with a thin (~5 nm) oxide that was low in Cr, Fe and Ni but high in Al (35-40 atom %) and Ti. The subsequent general corrosion rate in-core (two years) and in-autoclave (200 ppb O_2, 10 ppm H_2O_2, 150°C [302°F]) was considered "poor." In fact, flaking of the Alloy X-750 was observed in-core.
3. Type III. Type II process followed by an additional age at 705°C (1300°F)/5 hours in 80% argon and 20% oxygen cooled resulting in an oxide of 200-300 nm of mostly Cr_2O_3 with low Ni and Fe and high Ti and Mn (no corrosion testing).

A reddish brown (TiO_2 and NiO) oxide located under a grayish oxide ($FeNiFeO_4$) was observed by X-ray diffraction on the more corrosion resistant Type I fuel spacers. The Type II fuel spacer was characterized by three layers of oxides with "occasional" amounts of the $FeNiFeO_4$ spinel. Additional Auger examination of the Type I and III spacers revealed that the concentration of chromium in Alloy X-750 is not large enough to form a continuous layer of Cr_2O_3. Instead a layered oxide is observed consisting of a sequence of Alloy X-750/Cr_2O_3/ mixture of spinels ($FeNiFeO_4$, $NiCrO_4$, and $FeCr_2O_4$).

The nickel release for oxidized fuel spacers as measured in the autoclave was approximately 20 ppb. This concentration agrees with the thermodynamic calculation for $FeNiFeO_4$ in the solid phase. The shiny Type II fuel spacer nickel release was as high as 90 ppb, approaching the equilibrium concentration of nickel in the presence of NiO.

The results of both the Japanese and Swedish BWR results clearly indicate that the improvement in corrosion resistance of pre-oxidized Alloy X-750 is produced by the formation of $FeNiFeO_4$ on the Alloy X-750 surface. Therefore, to minimize the general corrosion and ^{60}Co release from Alloy X-750 a high temperature pre-oxidation of the Alloy X-750 should be performed.

The oxidation process is conducted at the same temperature as is used for optimizing IGSCC resistance of Alloy X-750, i.e., the direct age cycle should be performed in air at a controlled dew point, rather than in vacuum. Since the pre-oxidation during the age hardening process produces a change in appearance from vacuum induced bright shiny to dull metallic/dark gray-green, the high temperature pre-oxidized components are readily apparent.

Summary

PWR and BWR studies demonstrate that, as would be anticipated, surface condition/surface films have an effect on the subsequent general corrosion and IGSCC propensities of Alloy X-750. The BWR industry, through the BWR Vessel and Internals Project, has recognized the potential issue of sensitivity of Alloy X-750 to SCC to surface effects and have published BWRVIP-84, "Guidelines for Selection and Use of Materials For Repairs to BWR Internal Components," as a guide for selecting BWR structural materials including Alloy X-750 [6]. This document provides guidelines for fabrication of Alloy X-750 components that include limits on bending and cold straightening (2.5%), limits on final machining that require that the final machining passes shall be limited to 0.254 mm (0.010-inches) and limits for fabricating threaded components, such that threads are fabricated only by machining or rolling with the final machining limited to 0.254 mm (0.010-inches) per pass. These suggested controls represent a portion of the BWR industry attempt to address the effects of surface condition on the SCC resistance of Alloy X-750.

RAI 2

Please expand the information in Table 2-1 to include the hydraulic load and preload in addition to the maximum stress for each design. It would help to summarize the maximum stress for each of the inspection locations in Table 2 or Table 3 of Appendix A for the Group 2 beam design and Table 2 of Appendix B for the Group 3 beam design.

BWRVIP Response to RAI 2:

The following information will be added to Table 2-1:

**Content Deleted -
EPRI Proprietary Information**

TS

RAI 3

Section 2.4 of the report states:

> "For Group 2 beams, the initial inspection interval (12 years) is warranted based on three factors: (1) Field experience (no beam with the 25 kip [kilopounds] preload has experienced cracking in the first ten years)..."

Clearly identify what the field experience is for Group 2 beams with the 25 kip preload. Including the information from RAI 2 in Table 2-1 and Appendix A and B will help answer this question.

BWRVIP Response to RAI 3:

Sections 2.3.3 and 2.3.4 provide some historical background on the changes to the heat treatment and fabrication processing of the Group 2 and Group 3 jet pump beam design, respectively. Since that time, there have been no in-service failures or confirmed cracking of either Group 2 or Group 3 beams having a preload of 25 kips. Thus, the statement in Section 2.4 was intended to reflect that the field experience for the revised designs and material heat treatment specification has been excellent. It should also be noted that 100% of jet pumps beams are inspected according to BWRVIP-138, Revision 1, Table 7-1 for Group 2 and Table 7-2 for Group 3 beams, respectively.

RAI 4

Section 2.4 of the report states:

> "Laboratory data for Alloy X-750 has shown a direct dependency of the time-to-initiation on the applied stress [2, 3].

Reference 2 includes design, procurement, fabrication and installation information for Alloy X-750, but no lab data while reference 3 is GE proprietary information. Provide the data discussed in the text or provide a reference document with this data.

BWRVIP Response to RAI 4:

Reference 2 is "BWR Vessel and Internals Project, Guidelines for Selection and Use of Materials for Repairs to BWR Internal Components (BWRVIP-84)" [6] and Reference 3 is M. F. Aleksey, R. A. Carnahan, A. A. Strod and L. M. Zull, "Improvements in Jet Pump Hold-Down Beam Service Life," NEDE-24362-1, December 1981 (GE Proprietary Information).

As is the case for all alloys, the microstructure of Alloy X-750 has a dramatic effect on its IGSCC propensities. For example, numerous laboratory and in-reactor studies have been performed over the last few decades to especially determine the effect of microstructure, (e.g., cold work, grain boundary composition and various heat treatments/aging steps) on IGSCC propensities of Alloy X-750 with special emphasis on identifying the optimum heat treatment for IGSCC resistance in BWR-type environments [7-16].

Table 1 summarizes the interrelationship among heat treatment, microstructure and IGSCC propensities as evaluated by the creviced bent beam (CBB) test in high dissolved oxygen (8 ppm) water at 288°C (550°F) as an example of this microstructural-IGSCC correlation.

Table 1. Microstructural and IGSCC Characteristics of Alloy X-750

Heat Treatment	Matrix	Grain Boundaries	IGSCC Factor	CBB IGSCC Depth (μm)
HTH[1] 1066°C (1950°F)/1 h + 704°C (1300°F)/20 h	TiC NbC γ' [$Ni_3(Al, Ti)$] (fine)	Globular $Cr_{23}C_6$ Cellular $Cr_{23}C_6$	Cr depletion due to cellular $Cr_{23}C_6$ and/or HSC	630
1066°C (1950°F)/1 h + 20% cold work + 704°C (1300°F)/20 h	TiC NbC γ' [$Ni_3(Al, Ti)$] (fine) $Cr_{23}C_6$ (at twins)	Globular $Cr_{23}C_6$ Cellular $Cr_{23}C_6$ (more)	Cr depletion due to CW enhanced cellular $Cr_{23}C_6$ and/or HSC	1240
1066°C (1950°F)/1 h + 60% cold work + 704°C (1300°F)/20 h	TiC NbC γ' [$Ni_3(Al, Ti)$] (fine) $Cr_{23}C_6$ (scattered, fine)	Globular $Cr_{23}C_6$	CW dispersed globular $Cr_{23}C_6$ in matrix and GBs	206
AH[2] 1066°C (1950°F)/1 h + 843°C (1550°F)/24 h + 704°C (1300°F)/20 h	TiC NbC γ' [$Ni_3(Al, Ti)$] (fine & coarse)	γ' [$Ni_3(Al, Ti)$] (fine) in PFZ η [Ni_3Ti] (needle-like) PFZ Si, B, P segregation	Grain boundary segregation?	1330

[1] HA in Figure 7
[2] HMA in Figure 7

All these studies have indicated that a high temperature anneal (1066°C [1950°F]/1 hour) followed by direct aging (704°C [1300°F]/20 hours) provides the optimum IGSCC resistance, but not immunity to IGSCC in BWR (and PWR) environments. In fact, some in-reactor, but not irradiated, testing of Alloy X-750 in this heat treatment has demonstrated excellent results. For example, no IGSCC has been observed on creviced and un-creviced Alloy X-750 constant load specimens after sixteen (16) years of in-reactor exposure at stress levels exceeding 120% of the 288°C (550°F) yield stress [15-16].

These promising results have been confirmed by laboratory creviced constant load tests in high oxygen (8 ppm) high conductivity (0.3 - 0.5 μS/cm) water where the direct aged Alloy X-750 (1066°C [1950°F]/1 hour + 704°C [1300°F]/20 hours) demonstrated high IGSCC resistance, compared to the HMA heat treatment in Figure 7 [8]. In other words, the two step aged AH (HMA in Figure 7) heat treated (1066°C [1950°F]/1 hour + 843°C [1550°F]/24 hours + 704°C [1300°F]/20 hours) Alloy X-750 was characterized by shorter times to failure. These results are a sample of open literature data on the effect of stress and heat treatment on the IGSCC propensities of Alloy X-750 in oxygenated BWR environments.

As shown in Table 1, the more susceptible AH (HMA in Figure 7) heat treated Alloy X-750 had γ' $Ni_3(Al,Ti)$ and η phase Ni_3Ti in the grain boundaries. The needle-like η phase was accompanied by a precipitate free zone (PFZ) along the grain boundaries, no precipitated chromium carbides and fine γ' particles precipitated in the PFZ. The more IGSCC resistant HTH (HA in Figure 7) heat treated Alloy X-750 had only chromium carbide $Cr_{23}C_6$. This suggests that the presence of η phase, PFZ and fine γ' particles in the PFZ have some detrimental effect on IGSCC resistance.

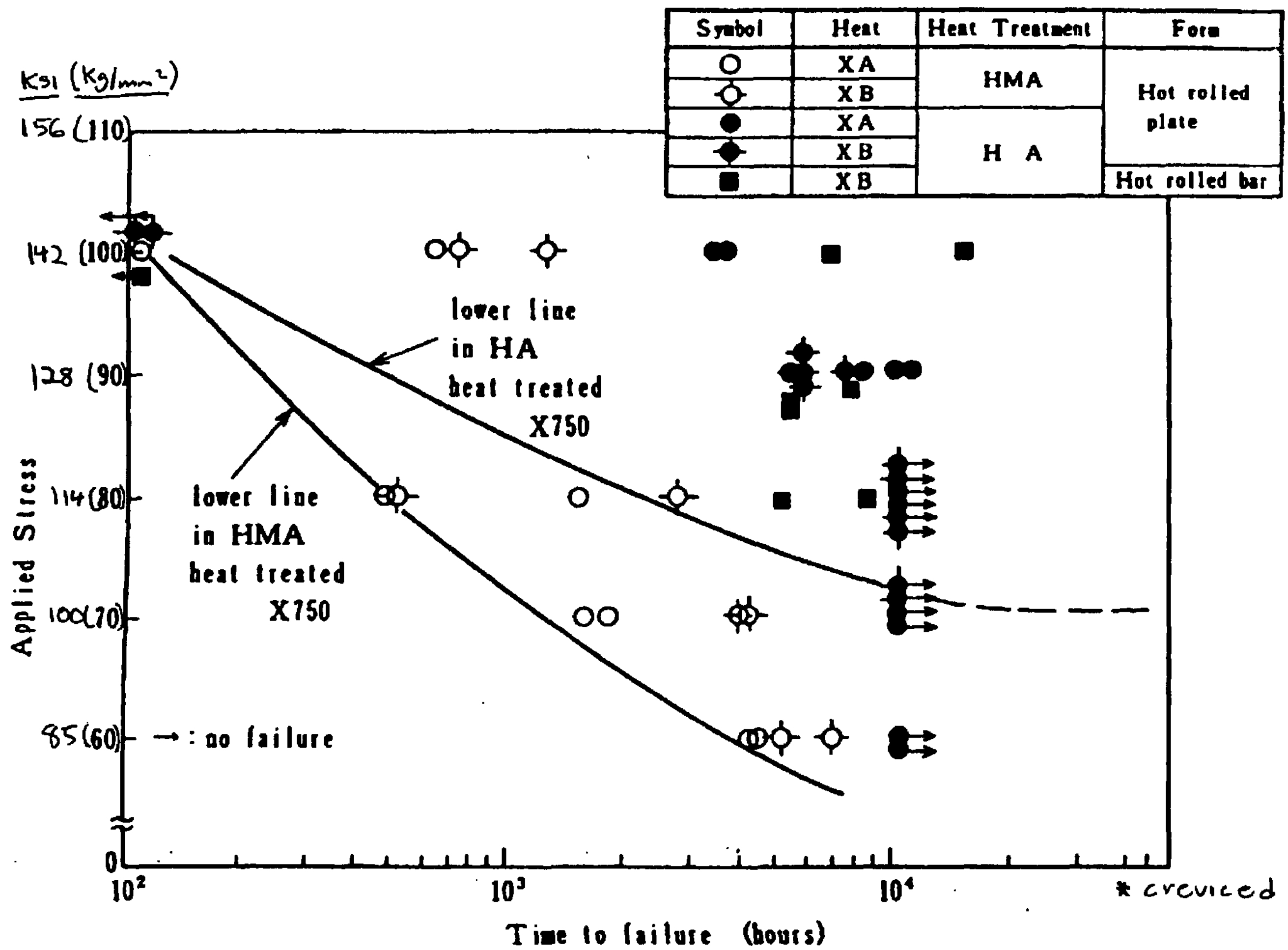

Figure 7. Stress versus Time to Failure for Alloy X-750 in 288°C (550°) Water in 8 ppm Dissolved Oxygen [8]

Summary

As a result of concerns related to the effect of stress on IGSCC of Alloy X-750, BWRVIP-84 has recommended that to avoid cracking in the BWR reactor internals environment, the use of Alloy X-750 requires diligent control over composition, heat treatment, design, and operational stresses [6]. In addition, BWRVIP-84 currently recommends that maximum allowable peak stress for Alloy X-750 shall be 80%[1] of the yield strength of the material at the intended operating

[1] A revision to BWRVIP-84 is in process to change the maximum allowable stress to 70% of the yield stress for non-threaded components and 78% for threaded components. This criteria applies to repair hardware only.

temperature, or a linear elastic fracture mechanics evaluation be performed, to calculate an applied K for comparison to appropriate data for the material in the BWR environment.

RAI 5

Section 2.4 of the report states:

> "For beams in the HTA [high temperature anneal and age] condition and under NWC [normal water chemistry] conditions, a statistical evaluation of the Group 2 and Group 3 beams (based on applied stress) has been used to quantify the significant differences in the initiation times in the jet pump beams. Based on available data, the mean time for SCC [stress corrosion cracking] initiation in the Group 2 beams is 40 years. Decreasing the stress ratio from 0.74 to 0.58 results in a six-fold increase in mean beam initiation time (see Table 2-2). As initiation is a statistical process, crack initiation in a small population of beams can be expected to occur somewhat earlier than the mean predicted time. To provide a high confidence (~99%) that cracking in the beam would be detected by inspection, a value of 3σ was selected to develop the appropriate inspection interval."

Provide additional explanation regarding how you arrived at such a high confidence level (~99%) based up on the available data. Include in the explanation the variability of material properties that can be expected for beams currently installed in BWRs and how they might be different from the materials used in the lab testing.

BWRVIP Response to RAI 5:

In order to determine a mean time to failure that is necessary to set the initial or first inspection interval, laboratory data where failure by SCC had occurred was used. The available data was fitted to a log normal distribution with its associated standard deviation calculated. Using this standard deviation, a probability of failure was calculated for the 3σ case, and for both Group 2 and Group 3 beams, the probability that crack initiation would not occur is greater than 99%. Therefore, the recommended initial inspection interval was set to incorporate this low probability of crack initiation.

Given recent field experience related to cracking in a core shroud tie rod assembly upper support bracket, further statistical evaluations have been performed by GEH to confirm the SCC behavior of Alloy X-750. The data set used included additional laboratory data, as well as the field experience for the material. Specifically, the in-service jet pump beams were included in the data evaluation, and were treated as no-failure data ("run out" data). The additional data confirm the previous evaluation, in that the mean time to failure is significantly greater than the initial inspection interval, and the probability of a crack initiating before the first inspection is very low (<1%).

Regarding laboratory data used to predict field behavior, it must be acknowledged that there would be more variability in material properties for actual hardware than for the limited amount of material used for the lab test samples. In order to minimize the differences and to minimize the variability in the actual material used in plant hardware, ASME specifications and GEH internal material design specifications have been used to procure the materials for the jet pump beams. Several studies have confirmed that one key factor in determining SCC initiation susceptibility is the ratio of the peak applied stress to the actual yield strength of the material. While other key factors including the specific material composition and each heat's actual thermo-mechanical processing history are also important, the direct relationship with peak stress was validated in the statistical studies performed. The development of the mean value of peak applied stress and the lower bound peak applied stress limit curves used the actual yield strength values of the lab materials. This led to an accurate value of the stress ratio imposed during testing. However, the evaluation of susceptibility used for the in-service jet pump beams was based on the lower ASME Code minimum yield strength in order to conservatively estimate the stress ratio to which the jet pump beams were subjected while in service. This approach addresses concerns about the variation in strength properties of the materials. A recent review of many of the material CMTR records for actual jet pump beam materials established that all heats exceeded the Code minimum value of the yield strength with the average high temperature yield strength being in excess of 110 ksi. This review supports the conservative nature of the assessment.

As previously mentioned, recent efforts to re-evaluate the behavior of Alloy X-750 SCC susceptibility led to a re-assessment by GEH of the stress ratio/time to initiate SCC. This new assessment included additional laboratory and a significant amount of field component data. Therefore it included many more heats of materials, thereby better addressing concerns regarding material property variability. As expected, this new assessment supports the earlier conclusions that have been presented in BWRVIP-138, Revision 1. Finally, consistent with the intent of BWRVIP-138, Revision 1, the jet pump beams will be inspected at specific intervals using damage tolerance methods to address concerns regarding undetected SCC initiation following service exposure.

Summary

SCC failure data obtained from laboratory testing and inservice/field experience was used to estimate the probability of crack initiation. A statistical evaluation showed that the mean time to failure is significantly greater than the initial inspection interval, and the probability of a crack initiating before the first inspection is very low (<1%). To minimize the variability in material properties of actual hardware, ASME specifications and GEH internal material design specifications have been used to procure the materials for the jet pump beams. Additionally, a re-assessment of the stress ratio/time to initiate SCC was performed which accounted for material property variability.

<u>RAI 6</u>

Section 4.3.1 of the report states:

> "As shown in Section 3, the majority of the failures have occurred in this region (BB-1). Based on the field experience and the stress analysis that indicates significant stresses are present in this location, continued inspection is warranted."

Can you clarify these statements given that no cracking of the Group 2 or 3 beams with the HTH heat treatment has been observed and all of the failures have been BWR/3 or Group 1 with the EQA heat treatment?

BWRVIP Response to RAI 6:

The following text is suggested, "As shown in Section 3, the majority of the observed failures have occurred in this region (BB-1). Although no failures have been observed in the Group 2 or Group 3 jet pump beam designs, stress analysis indicates significant stresses exist in this location for all beam designs. Based on field experience for Group 1 beams and the presence of high stresses in this region for all beam designs, continued inspection is warranted."

RAI 7

Section 5.1 of the report states:

"In order to predict the crack depth as a function of operating time, crack growth rates for the Alloy X-750 need to be developed. These crack growth rates or dependencies can be then used to assess the time for an initiated crack to reach a critical crack size. Most of the limited data was measured under high oxygen conditions in tests performed at GE NE [General Electric Nuclear Energy], GE GRC or available to GE NE. The effort must also make use of the extensive understanding of the behavior of other cold worked austenitic materials to benchmark the limited X-750 data and to support the projected benefit imparted by the effective HWC [hydrogen water chemistry] (HWC/NMCA [noble metal chemical addition]) environment that is present in many of the operating BWRs."

And Section 5.3 of the report states:

"While there is currently limited data on the crack growth rates in X-750 material, there are data on other austenitic materials in BWR type high temperature water environments. These materials include cold worked stainless steel, cold worked Alloy 600 and cold worked Alloy 182 weld metal. GE GRC has tested a variety of materials in both NWC and HWC environments. GE GRC has observed a definite dependence of rate with accompanying yield strength [11, 12]. Figure 5-3 displays data for these materials in a NWC environment. As the yield strength is increased toward 700 MPa [megapascal] (~100 ksi), the observed crack growth rate rises. This data is measured at levels similar to the other data on X-750 (shown as a range). The figure also shows that theory of SCC and the associated modeling is consistent with the measured data."

Elaborate on why data for Alloy 600 and austenitic stainless steels may or may not be applicable to Alloy X-750. What inputs were used to make these theoretical predictions and why is this trend important in this case given the fact that hold-down beams with EQA and high temperature heat treatments have nearly identical YS and are strengthened by precipitation instead of cold working as is the case for Alloy 600 and austenitic stainless steels? Does the theory support the extrapolation of crack growth rate data to applied stress-intensity factor (K) values an order of magnitude lower than the applied K values for the measured crack growth rates?

BWRVIP Response to RAI 7:

Figure 5-3 from BWRVIP-138 Revision 1 is reproduced here as Figure 8 [17]. The data in the figure suggests that since the Alloy X-750 crack growth rate data produced on a material with a yield stress of 700 MPa (~100 ksi) in a highly oxygenated environment is comparable to the crack growth rate data obtained on cold worked Types 304, 304L and 316L stainless steel plus Alloy 600, then the crack growth rate data produced on these cold worked non-precipitation hardened alloys can be used to support Alloy X-750 crack growth rates even though strengthening is produced by a different mechanism. BWRVIP-138, Revision 1 also presents a similar yield stress vs. crack growth rate plot for the hydrogen water chemistry (HWC) environment as Figure 5-4, which is reproduced here as Figure 9. However, the comparable Alloy X-750 crack growth rate data is not shown in this figure.

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EPRI Proprietary Information TS

Figure 8. Effect of Yield Stress on Crack Growth Rate for Various BWR Structural Materials in NWC-type Environments [17]

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EPRI Proprietary Information TS

Figure 9. Effect of Yield Stress on Crack Growth Rate for Various BWR Structural Materials in HWC-type Environments [17] (No Alloy X-750 data shown.)

As noted in Reference 18, the above results exhibit a broad consistency with prior work on yield strength effects on unsensitized stainless steel and Alloy 600. While these alloys have differences in composition that are both substantial (iron-base vs. nickel-base) and more compositionally subtle (Mo, Ti, Al, etc.) with a very pronounced difference in mechanism of yield strength elevation (cold work vs. precipitation hardening vs. irradiation), the growth rates are consistent with a universal effect of yield strength on unsensitized materials.

Precipitation hardening in Alloy X-750 produces a significantly higher yield strength than does 20% cold work in austenitic iron and nickel-base alloys. In turn, an associated but modest increase in the observed crack growth rate is generally observed in Alloy X-750 (age hardened or 20% cold worked) at both high and low corrosion potential compared to cold worked stainless steel and Alloy 600 [18].

At high corrosion potential, Alloy X-750 AH exhibit somewhat lower corrosion potential than the HTH or 20% cold worked alloy X750 [18]. At low corrosion potential, the low growth rates of Alloy X-750 in the HTH condition are well below that expected from their higher yield strength, and this may be related to grain boundary precipitates and/or composition. It has been noted in cold worked stainless steel that the growth rate at high corrosion potential is reduced by a factor of five when grain boundary carbides are present in the grain boundary, i.e., when no chromium depletion exists [19].

While the effects of most parameters on SCC are intertwined with other variables, as appears to be the case with age hardening treatments and grain boundary precipitates, there does appear to be a common underlying effect of elevated yield strength in most iron- and nickel-base materials in high temperature water. The origin of this commonality in the effect of yield strength is discussed elsewhere [20, 21].

Therefore, while there may be some discussion in the nuclear industry on the applicability of using crack growth rate data from other austenitic alloys to bound the crack growth rates of similar alloys even though the strengthening mechanisms are different (e.g., cold work vs. precipitation hardening vs. irradiation), the end result is that the crack growth rates for Alloy X-750 do indeed fall within the range of cold worked Types 304, 304L and 316L stainless steel and Alloy 600.

As presented in Figure 10, which is reproduced from Figure 5.2[2] from BWRVIP-138, Revision 1 [17] and is also available in the open literature [18], HWC reduces the crack growth rates of Alloy X-750 in a similar manner it reduced the crack growth rates of cold worked Types 304, 304L and 316L stainless steel and Alloy 600. More specifically for Alloy X-750 AH, the factor of improvement (FOI) in crack growth rate reduction with the application of HWC is between approximately 5 and 10 as shown in Figure 10. It should be noted that the factor of improvement in crack growth rate proposed for Alloy X-750 in the HWC environment in BWRVIP-138, Revision 1 is only a factor of 2 which is very conservative as compared to the results presented in Figure 10.

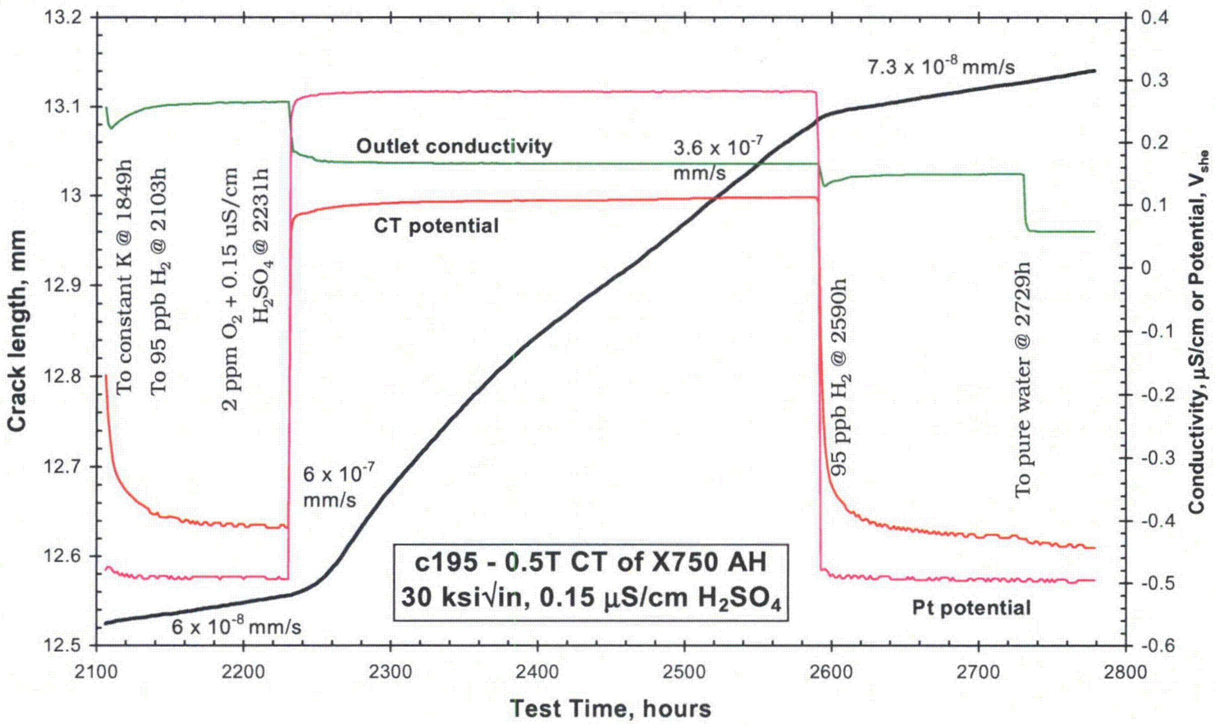

Figure 10. Crack Length vs. Time for Alloy X-750 AH Tested in 288°C Water [17, 18]

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[2] Subsequent to transmittal of BWRVIP-138, Revision 1 to the NRC, an error was noted that Figure 5-2 was a duplicate of Figure 5-4. Figure 10 contained in the response to RAI 7 is the corrected Figure 5-2. This corrected figure will be included in a revision to BWRVIP-138.

It should also be noted that the BWRVIP has a project underway to conduct crack growth rate tests on a representative heat of X-750. Results of this project are expected in 2012 and will be communicated to the NRC when available.

Summary

While cold worked austenitic stainless steels and non-precipitation hardened nickel alloys (e.g., Alloy 600) have a very pronounced difference in mechanism of yield strength elevation, i.e., cold work vs. precipitation hardening, than Alloy X-750, the relative growth rates are consistent with a universal effect of yield strength on unsensitized materials, i.e., the higher the yield strength, the higher the crack growth rate. Therefore, the crack growth rate data for austenitic stainless steels and Alloy 600 is applicable to Alloy X-750 since the end result is that the crack growth rates for Alloy X-750 do indeed fall within the range of cold worked Types 304, 304L and 316L stainless steel and Alloy 600.

RAI 8

Section 7.1 of the report states:

> "The flaw tolerance results at the limiting locations for each beam design are presented in Figures 7-1 through 7-4."

From the figures, the maximum stress values for Section IR of the Group 2 and 3 beams are not consistent with the maximum stresses shown in Table 2-1. In similar fashion, the maximum stress for Section C of the Group 2 and 3 beams are also not consistent with the max stresses shown in Table 2-1.

Justify why the peak stresses at each location are different from the maximum stress that is shown in Table 2-1.

This question can be withdrawn if there is a summary of maximum stress for each inspection location as requested in RAI 2.

BWRVIP Response to RAI 8:

The stresses listed in Table 2-1 represent the maximum stress along the longitudinal axis of the jet pump beam (at the beam bolt hole) for a 25 kip preload and without thermal relaxation. This table is intended to communicate the relative stresses for each beam design; it is not intended to be used as input to a detailed fracture mechanics analysis. The stress results summarized in Section 7.1 and in Appendices A and B are representative of the 99th percentile preload with thermal relaxation. The results in Section 7.1 and in Table 2-1 should be similar, but not identical since they represent slightly different conditions. Further, the results shown in Figures 7-1 through 7-4 are taken from a different orientation (Figure 7-1 and 7-3) than for Table 2-1 and at a different location (Figure 7-2 and 7-4). The stresses shown in Figures 7-1 and 7-3 are normal to the plane of the crack face; this plane is oriented at an angle with respect to the plane upon which the maximum longitudinal stress occurs and is reported in Table 2-1.

RAI 9

There appears to be a contradiction between Figure 31, "NWC, 99th Percentile Load, Group 2 Jet Pump Beam, Center Crack Evaluation, Crack Plane D," and the text on Page A-38 of the report with regards to the size of the assumed starting center crack. Which one is correct?

On Figure 7-4, for the three plots of Group 3 beams, the title of the plots says Section B while the Figure caption does not identify the Crack Plane. To compare with Figure 7-2 for Group 2 beams, the plots are all for Section C and the Figure caption also has Crack Plane C. The staff assumes that both Figure 7-2 and 7-4 should be for Crack Plane C. Is that correct?

In Appendix A and B, Sections 8.0 are both labeled "RESULTS" and staff assumes they should be "DISCUSSION". Is that correct?

BWRVIP Response to RAI 9:

The following clarification in bold and underlined type is proposed:

The residual life curve for crack plane D shows substantially longer life than the other fillet radius crack planes because of the low surface stress. An initial flaw size of 0.25 x 0.25 inches was used **for the corner crack and 0.1 x 0.1 inches for the center crack**, at Section D, because the upper surface stresses are so low that the residual life for a 0.01 x 0.01 inch flaw is so long that it makes the output file very large. The residual life and K curves for the center crack, **and corner crack**, at crack planes B, C1, C and D are shown in Figures 28 through **31, and Figures 16 through 19, respectively.**

The caption for Figure 7-4 is missing "B". The limiting section in the Group 3 beam occurred at crack plane B; whereas, for the Group 2 beam it occurred at crack plane C.

The staff is correct, Sections 8.0 of both Appendices A and B should be titled "Discussion" and will be revised to state such.

RAI 10

Why was the allowable flaw size in Section 7.2 of Appendix A based on the hydraulic load with a factor of safety instead of the upper-bound bolt preload with thermal relaxation that was used for crack growth in all of the crack growth analysis?

BWRVIP Response to RAI 10:

For the crack growth analysis, the assumption applied is that the beam retains sufficient stiffness that the preload has not been lost; therefore, the preload exceeds the hydraulic load and the preload is the appropriate load to consider. For collapse to occur the beam would experience significant distortion such that the preload is assumed to have been entirely relieved; therefore, the load acting on the jet pump beam which would drive collapse, is the hydraulic load.

RAI 11

Is irradiation-assisted stress corrosion cracking a potential aging mechanism for the jet pump hold-down beams?

BWRVIP Response to RAI 11:

PWR in-reactor testing of irradiated (e.g., 2.3 x 10^{20} n/cm^2) bolt-loaded compact tension (CT) fracture mechanics specimens have been performed in 360°C (680°F) hydrogenated (40 to 60 cc H_2/kg H_2O, [3.6 to 5.4 ppm]) PWR water to determine the IASCC behavior of HTH (HTH was undefined in the report, but assumed to be 1093°C [2000°F]/1 hour + 704°C [1300°F]/20 hours) Alloy X-750 (and direct-aged Alloy 625) [22]. This data confirmed some previous results showing that high irradiation levels reduce the IASCC resistance of Alloy X-750. Alloy X-750 heat-to-heat IASCC variability correlated with boron content, with low boron heats showing improved IASCC resistance.

Microstructural, microchemical and deformation studies were performed to characterize the mechanisms responsible for IASCC in Alloy X-750 [22]. The mechanisms under investigation were: boron transmutation effects, radiation-induced changes in microstructure and deformation characteristics and radiation-induced segregation (RIS). Irradiation of Alloy X-750 caused significant strengthening and ductility loss that was associated with the formation of cavities and dislocation loops. High irradiation levels did not cause significant segregation of alloying or trace elements in Alloy X-750.

In the non-irradiated condition, an IASCC susceptible HTH Alloy X-750 heat containing 28 ppm B showed grain boundary segregation of boron; whereas a non-susceptible HTH heat containing 2 ppm B did not show significant boron segregation [22]. To determine the distribution of transmutation-produced helium, these alloys were annealed at 816°C (1500°F) after irradiation to agglomerate the helium. Helium bubbles were observed at grain boundaries in the high boron heat, but little or no evidence of grain boundary helium was observed in the low boron heat. Based on these results, transmutation of boron to helium at grain boundaries, coupled with matrix strengthening, is believed to be responsible for IASCC in Alloy X-750. At fluences above 10^{19} n/cm^2, the in-reactor SCC resistance of HTH Alloy X-750 was significantly reduced. At low fluences (e.g., 10^{14} n/cm^2), there was only a slight degradation in SCC resistance that was attributed to radiolysis effects [22].

It is noted that since the jet pump beams are located in the middle of the downcomer, there is a fair amount of attenuation by water in the downcomer between the jet pump beams and the shroud OD. Therefore, the jet pump beams would not accumulate fluence as rapidly as the shroud wall. Also, the jet pump beams are at approximately 5/6 of the core height, i.e., the jet pump inlet is at 2/3 core height, the rams head is above the inlet and the jet pump beams are located on top of the rams head. Since radiation from the core decreases rapidly with distance above 2/3 core height, then the impact of fluence on the jet pump beams would be limited.

Finally, even if the boron to helium hypothesis is correct for PWR environmental conditions, since the BWR environment is characterized by significantly lower temperatures than the PWR, helium will not significantly migrate at BWR temperatures. There could be some radiation damage, which would include strengthening from dislocation loops and perhaps a very small amount of precipitate dissolution with some increase in yield strength, but it would be small compared to the yield strength of heat treated Alloy X-750. Similarly, RIS would also be small compared to the fact that many Alloy X-750 heat treatments already produce significant chromium depletion.

Summary

In summary, IASCC is not considered a potential aging mechanism for the Alloy X-750 jet pump hold-down beams, particularly since the helium migration is considered unlikely at BWR temperatures and the fluence is expected to be reasonably low.

Supplemental Information Provided by the BWRVIP

An error was noted that Figure 5-2 was a duplicate of Figure 5-4 in "BWRVIP-138, Revision 1. The correct Figure 5-2 is shown below. This figure will be incorporated in a revision to BWRVIP-138.

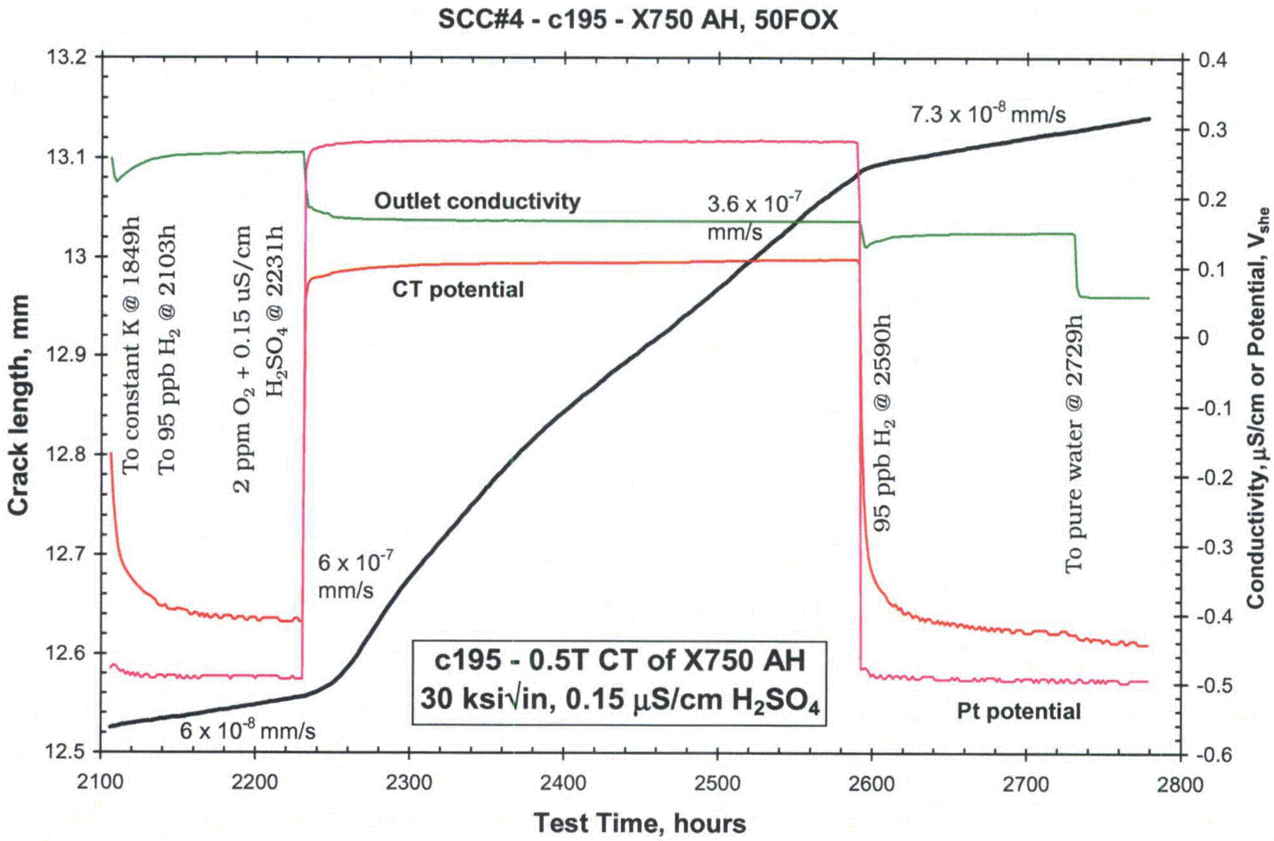

Figure 5-2. Recent Data from GE GRC on Alloy X-750 [10]

References:

1. M. Foucault and C. Benhamou, "Influence of the Surface Condition on the Susceptibility of Alloy X-750 to Crack Initiation in PWR Primary Water" paper presented at the 6th International Symposium of Environmental Degradation of Materials in Nuclear Power Systems - Water Reactors, August 1 – 5, 1993, San Diego, CA, published in proceedings of same, R. E. Gold and E. P. Simonen, editors, TMS, Warrendale, PA, 1993.

2. K. Tada, et al., "Control of Nickel Release from Inconel X-750 by Surface Treatment," paper presented at the 1988 JAIF Meeting International Conference on Water Chemistry in Nuclear Power Plants, Tokyo, Japan, April 1988.

3. Y. Hemmi, et al., "Control of Ni Release from Inconel X-750 by Surface Treatment (III)," January 10, 1989.

4. Y. Hemmi, et al., “Passivation Mechanism of Oxide Films Formed on Inconel X-750 in Simulated BWR Core Environment,” paper 29, presented at the Water Chemistry of Nuclear Reactor Systems 5, BNES, Bournemouth, October 1989.

5. E. Ahlberg and B. Rebensdorff, “General Corrosion of Alloy X-750 under BWR Conditions,” paper presented at the Water Chemistry of Nuclear Reactor Systems 6, BNES, Bournemouth, October 1992.

6. “BWR Vessel and Internals Project: Guidelines for Selection and Use of Materials for Repairs to BWR Internals (BWRVIP-84),” EPRI, Palo Alto, CA and BWRVIP: 2000. 1000248.

7. M. Tsubota, K. Hattori and T. Kaneko, "Study on SCC Susceptibility of Inconel X-750 in High Temperature Pure Water: Effects of Aging Temperature and Time," Corrosion, Vol. 44, No. 2, February 1988, p. 73.

8. J. Kuniya, et al., “Evaluation of Stress Corrosion Cracking Resistance of High Strength Ni Base Alloys,” paper 86 presented at Corrosion 92, Nashville, April 27, 1992.

9. K. Hosoi, et al., “Relationship Between Susceptibility to Stress Corrosion Cracking in High Temperature Water and Microstructure of Inconel Alloy X-750,” Proceeding of Int. Symp. on Environmental Degradation of Materials in Nuclear Power Systems - Water Reactors, NACE, Myrtle Beach, SC, August 22, 1983, p. 334.

10. K. Hattori, M. Tsubota and T. Okada, “Effect of Chloride on the Stress Corrosion Cracking Susceptibility of Inconel X-750 in High Temperature Water,” Corrosion, Vol. 42, No. 9, September 1986, p. 531.

11. S. Hattori, “Stress Corrosion Cracking of Age-Hardenable Nickel Base Alloy in High Temperature Water,” Annual Meeting, American Nuclear Society, Miami, FL, June 1981.

12. M. T. Miglin and H. A. Domain, “Microstructure and Stress Corrosion Resistance of Alloys X-750, 718 and A286 in Light Water Reactor Environments,” J. Materials Engineering, Vol. 9, No. 2, 1987, p. 113.

13. G. S. Was, “Grain Boundary Chemistry and Intergranular Fracture in Austenitic Nickel-Base Alloys - A Review,” Corrosion, Vol. 46, No. 4, April 1990, p. 319.

14. K. Hosoya et al., “Microstructural Role in Environmentally Assisted cracking in Ni-Base Alloys,” Corrosion, Vol. 44, No. 11, November 1988, p. 838.

15. L. G. Ljungberg and L. Unneberg, “In-Plant Test Data Related to the Integrity of Primary System Components,” paper presented at the Reactor Pressure Vessel and Internals Inspection and Repair Workshop, Charlotte, NC, July 16-18, 1991.

16. B. Bengtsson, et al., "Constant Load Tests of Age-Hardened Ni-Base Alloys in Oskarshamn 2 BWR," paper presented at the EPRI Workshop on Advanced High Strength Materials for LWR Vessel Internal Applications, Clearwater Beach, FL, March 12-13, 1986.

17. BWRVIP-138, Revision 1: BWR Vessel and Internals Project, Updated Jet Pump Beam Inspection and Flaw Evaluation Guidelines. EPRI, Palo Alto, CA: 2008. 1016574.

18. P. L. Andresen, et al., "SCC of High Strength Ni-base Alloys in High Temperature Water," paper 04675 presented at Corrosion 2004, NACE, New Orleans, LA, March 28 to April 1, 2004.

19. P. L. Andresen and G. S. Was, "SCC of Unirradiated Stainless Steels and Nickel Alloys in Hot Water," paper 4037 presented at the 17 International Corrosion Congress, NACE, Las Vegas, NV, October 6-10, 2008.

20. P. L. Andresen, L. M. Young, W. R. Catlin and R. M. Horn, "Stress Corrosion Crack Growth Rate Behavior of Various Grades of Cold Worked Stainless Steel in High Temperature Water", paper 02511 presented at Corrosion 2002, NACE, Denver, CO, April 7-11, 2002.

21. P. L. Andresen, "Similarity of Cold Work and Radiation Hardening in Enhancing Yield Strength and SCC Growth of Stainless Steel in Hot Water," paper 02509 presented at Corrosion 2002, NACE, Denver, CO, April 7-11, 2002,

22. R. Bajaj, W. J. Mills, M. R. Lebo, B. Z. Hyatt and M. G. Burke, "Irradiation-assisted Stress Corrosion Cracking of HTH Alloy X-750 and Alloy 625," paper presented at the 7th International Symposium on Environmental Degradation of Materials in Nuclear Power Systems - Water Reactors, Breckenridge, CO, August 7 – 10, 1995, published in proceedings of same, Eds. R. Gold, A. McIlree, published by NACE, Houston, TX, 1995, p. 1093.