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METALLURGICAL FACTORS THAT CONTRIBUTE TO CRACKING IN BWR PIPING

John R. Weeks

A Report to the
U. S. Nuclear Regulatory Commission Ad Hoc Task Force

MASTER

April 30, 1975

OFFICE OF REGULATORY ASSISTANCE
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UPTON, NEW YORK 11973

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A. Introduction and Summary of Observations

During the fall of 1974 and early winter of 1975, cracks have been discovered in the 4" bypass lines of the following Boiling Water Reactors (BWR's) in the United States: Dresden Nuclear Power Station, Unit #2; Millstone Nuclear Power Station, Unit #1; Monticello Plant; Peach Bottom Atomic Power Station, Unit #3, and Quad Cities Nuclear Power Station, Units #1 and 2. This history was described by Chairman Anders before the JCAE on February 5, 1975.(1) Further, similar cracks were discovered at two BWR's in Japan during the same period. More recently, cracks have been discovered in the core spray piping and in a furnace-sensitized "safe end" and adjacent "dutchman" at the Dresden Nuclear Power Station, Unit #2; Although inspections at all other U.S. BWR's have not disclosed further instances of cracking in core spray piping, leaking cracks have been found in the core spray piping of two BWR's overseas.

Metallurgical examinations of these cracks are not yet complete, although the results of investigations of the first cracks that developed in the 4" bypass lines at Dresden-2 and Quad Cities-2 have been submitted to the NRC by Commonwealth Edison(2) and General Electric.(3) Any conclusions drawn at this time must therefore be tentative, pending the outcome of future investigations. The following observations have been made to date. All cracks (except those in the furnace-sensitized safe end and dutchman) occurred in seamless type 304 stainless steel piping or in elbows fabricated from such piping, in the outer heat affected zone of either field or shop welds, in lines isolated from the main primary coolant flow during full power operation, except for the not yet examined cracks in the Monticello bypass lines. The cracks are exclusively intergranular, and occur in metal that has been lightly sensitized by the welding process, with only intermittent grain boundary carbides. They developed in the areas of peak axial residual stresses from welding rather than in the most heavily sensitized areas. No fatigue striations have been found on the fracture surfaces. The observations are similar to those made on the cracks that developed in the 6" bypass lines at Dresden-1 in 1966.(4) General Electric attributed these earlier problems to the high heat input in the block welding techniques used at this plant(5), although

the recent rash of cracking in the bypass lines of the new plants suggests that the welding heat input controls instigated by GE since the Dresden-1 reactor was built have not eliminated this problem. The evidence received to date strongly indicates that these cracks were caused by intergranular stress corrosion of weld-sensitized stainless steel by BWR water containing > 0.2 ppm oxygen. The possible role of fatigue or alternating stresses in this corrosion is not clear. Further, not all the cracks detected to date necessarily have occurred by the same mechanism.

B. Factors Known to Cause Intergranular Stress Corrosion (ISC) of Stainless Steels

The factors causing ISC of stainless steels by BWR coolants have been discussed in detail in two recent survey papers by Cowan and Tedmon[6] and Cowan and Gordon.[7] The principal factors are tensile stresses exceeding the yield stress of the material, some degree of sensitization of the material, and the presence of an oxidizing agent in the water.

1. Tensile Stress Levels Locally Equal to or Greater Than the Yield Stress of the Material at the Temperature of Exposure - These stresses, and the resultant (perhaps intermittent) yielding of the material, will lead to repeated rupture of the passivating oxide film and subsequent electrochemical dissolution attack in the depassivated areas, under certain environmental conditions. Possible sources of stress levels of this magnitude have been identified by E. C. Rodebaugh in his report to the task force.[8] Stress levels of this magnitude can only be achieved by superimposition of applied mechanical, pressure, thermal and thermal expansion stresses upon the residual axial stresses from welding and from grinding during weld preparation and post-welding cleanup. The near-yield residual stresses dominate: the cracks in the bypass lines are located in the zone of maximum residual tensile stresses from welding, not in the zone of maximum applied stress or sensitization. The cracks in the core spray lines, however, occurred in the more heavily sensitized regions, closer to the welds. Intergranular corrosion fatigue has also been observed under some conditions[9] in BWR environments.

Possible alternating stresses that could be acting on the bypass lines include high-cycle, low amplitude vibrational stresses and low-cycle thermal and thermal cycling stresses during heatup and cooldown of the recirculation and bypass lines.

The absence of fatigue striations from the fracture surfaces suggests that high-cycle vibrational stresses are not the primary cause of these cracks. The major portion of the time to failure comes in the initiation of cracking; once a crack has initiated, stress concentrations around the resultant notch will tend to increase the rate of crack growth. The low-cycle alternating stresses may be responsible for initiating the cracks, but the steady applied stresses, concentrated by the notch due to the cracks, may be responsible for crack propagation.

Since the residual axial tensile stresses from welding only exist in the inner half of the pipe thickness, once cracks have been initiated, they would be likely to penetrate laterally along the inner surface of the pipe more rapidly than through the wall, until stress intensification from the notch effect causes through-wall crack growth to continue. The general appearance of the cracks observed to date is consistent with this growth sequence: a non-through-wall crack 3 inches long on the inner surface, and less than 1/4 inch deep was found in the Quad Cities-2 Loop B (Fig. 5b of Ref. 2).

2. Some Degree of Sensitization of the Material - Sensitization is caused by precipitation in the grain boundaries of chromium carbides and other compounds containing N, P and Si. Nitrogen in the 0.03 to 0.06% range promotes intergranular attack in a Strauss test of a 0.03% C stainless steel containing 11-12% Ni, but not of one containing 9-10% Ni.[6] Apparently, increasing the nickel and nitrogen content decreases the carbon solid solubility in the steel, and promotes chromium carbide precipitation. Once the passivating film has ruptured, electrochemical cells between these grain boundary precipitates and/or the chromium depleted zones adjacent to the grain boundary and the grain surface itself tend to promote continued localized corrosion.

Walker[10] has shown that the response of a Type 304 stainless steel specimen to sensitizing heat treatments often reflects the thermal history in

the mill. Whether or not the mill thermal history of small diameter seamless pipe, in which these cracks have exclusively been found, differs significantly from that of large diameter welded pipe in which no cracks have been found, is not known. Vermilyea et al.[11] confirmed Walker's results and showed that a specimen furnace-cooled following an annealing heat treatment is less prone to sensitization than an identical specimen water-quenched after being annealed. They found the delayed sensitization resulted from slower development of grain boundary carbides, and suggested that this occurs because the slow cooling results in a decreased quenched-in vacancy concentration, which reduces the effects of vacancy supersaturation on carbide particle nucleation and growth. The thermal stresses during quenching of the pipe could also produce defects that promote carbide nucleation. The thicker wall on the 28" pipes may cause them to cool somewhat more slowly in the quenching process and therefore be somewhat less prone to sensitization during welding. The large diameter welded pipe also normally has a somewhat lower nickel content than the seamless pipe.

3. The Presence of an Oxidizing Agent in the Solution - Oxygen dissolved in the water, or oxidizing agents such as nitrates, are known to promote this attack. Oxygen levels as low as 0.2 ppm can cause the phenomenon, although the median time to failure at 550°F under otherwise identical conditions decreases by a factor of 40 to 100 as the oxygen concentration is increased from 0.2 to 8 ppm.[12] Oxide films deposited on stainless steel surfaces by crud transport mechanisms are known to chemisorb chloride and fluoride ions, the presence of which can promote depassivation and initiate intergranular stress corrosion. These deposits can also lead to oxygen starvation type concentration cells that increase the probability of localized depassivation and crack initiation.

Isolation of the bypass lines from the primary recirculation lines by closing the bypass valves before high power levels are reached (as was the practice at all plants except Monticello), can trap BWR coolant containing relatively high oxygen concentrations in these lines. The oxygen level in the BWR coolant is controlled by a balance between contact with air during shutdown, radiolysis of the water, and removal of oxygen and hydrogen by the boiling process. At a given temperature and pressure, the solubilities of

hydrogen and oxygen in mols/liter are approximately equal.[13] Since 2 mols of hydrogen are produced by radiolysis for every 1 mol of oxygen, a slight excess of oxygen over hydrogen is left behind in the coolant by the boiling process. At startup, the coolant is believed to contain ~ 8 ppm O_2, which is reduced by the boiling process to a steady state level of 0.2 - 0.3 ppm. How effectively the boiling process lowers the oxygen level in the coolant away from the boiling area, as in the recirculation lines or in bypass lines with the valve closed, is not known. Similarly, the core spray lines may be exposed to coolant - or steam - containing higher oxygen levels than the 0.2 - 0.3 ppm believed to be present in the vessel. Therefore, the possibility must be considered that the environment in the bypass and core spray lines is more conducive to stress corrosion cracking than that in the main primary and recirculation piping.

4. Temperature - The role of temperature is not clear. This phenomenon was observed at the BNL HFBR in a weld-neck flange that never operated above 150°F[14], as well as in commercial BWR's at much higher temperatures.

5. Time - The time for these cracks to initiate and propagate has ranged from over six years at Dresden-1[4] and the HFBR[14] to less than a year after startup at Beach Bottom-3.

C. Detectability and Growth Rate of Cracks

The detectability of tight intergranular cracks by inservice UT inspection is difficult to evaluate. An instance has been cited during the recent inspection of the Dresden-2 core spray piping of a leaking crack that went undetected by UT techniques until after the leak was found. Frequently, when the reactor is shut down, relaxation of the applied tensile stresses will cause the crack surfaces to close tightly, and therefore to transmit rather than reflect pressure waves. Further, the roughness of the external surface has a strong effect on the ability of the transducer applied to it to detect flaws. For these reasons, estimates of the crack growth rates are not possible from the recent inspection results. Also, as noted above, crack growth rates are believed to increase with time, and to differ substantially

in the lateral and through-wall directions. Therefore, from the inspection data, it is impossible to estimate a typical "worst case" rate of crack growth that can be used in establishing a reasonable inspection interval. The plant can be operated safely, however, since, if leakage occurs, the leak detection system will warn of a possible crack while it is still small.

D. Alternate Materials

Since the oxygen levels in the BWR coolant cannot readily be reduced below current levels and residual stresses from welding (although perhaps reducable somewhat by changes in welding techniques) cannot be removed without furnace-sensitization of the steel, consideration needs to be given to adoption of alternate materials that are less sensitive to ISC than Types 304 and 316 stainless steels. Because of the recent experience, these need to be considered especially for bypass and core spray piping.

General Electric has had an in-house research program to examine the compatibility of alternate materials for BWR coolants for several years, under the direction of G. M. Gordon of the Vallecitos Nuclear Center.[6,7,12,15,16] This program includes both in-house laboratory studies and exposures of bent-beam specimens in the Dresden-1 and Humboldt Bay BWR's. Other programs to examine alternate materials have been performed at the U.S. Steel Applied Research Center[17], Battelle Columbus Laboratories[18] and the Ohio State University.[19]

Classes of materials that have been tested include alternate stainless steels such as Types 309 and 316 SS and the carbon-stabilized grades of Types 347 and 321; the low-carbon grades Type 304-L, Type 308-L (weld metal) and Type 316-L; a nitrogen-strengthened low carbon steel Type 304-LN; higher nickel alloys such as Incoloy-800 and Inconel-600; ferritic and martensitic steels such as 2¼ Cr-1% Mo, 5% Cr-½% Mo, Airco E-Brite 26% Cr-1% Mo, Climax 19 Cr-2 Mo, and Carpenter Custom 450, and austeno-ferritic duplex alloys such as Creusot-Loire Uranus 50 and Sandvik 3RE60. The nominal compositions of these alloys are given in Table 1. The advantages and disadvantages of adopting these materials for BWR piping are discussed below.

1. Alternate Classes of Piping Alloys

a) Stainless Steels - Types 316 and 304 SS appear to be equally susceptible to intergranular stress corrosion in the furnace-sensitized condition.[18] As a general rule, stainless steels that are less prone to sensitization are more resistant to intergranular stress corrosion cracking than Types 304 and 316 SS. These include the "L" grades and the Cb and Ti stabilized grades. It is possible to furnace-sensitize the "L" grades, however, since the solid solubility of carbon in the steels is considerably lower than the 0.03 wt% maximum allowable in these steels. The Mo in Type 316-L, however, tends to reduce the degree of sensitization after a given time in the sensitizing temperature range.[6] Also, there appear to be synergistic effects between carbon and the Ni, N, Si and phosphorus contents of the steels that affect the degree of sensitization of the material under a given set of conditions, as revealed by Strauss tests or Huey tests.[6,7] The Cb and Ti stabilized grades are susceptible to weld sensitization in a narrow band adjacent to the weld deposits, where the heat from welding tends to dissociate the more stable carbides of Cb or Ti.[6] The subsequent rapid cooling causes the carbon to recombine with the more abundant chromium to produce a network of grain boundary carbides of Cr and an adjacent Cr-depleted zone, where intergranular stress corrosion can occur in what has been described as a "knife-edge" attack.

b) Austeno-Ferritic Steels - The austeno-ferritic alloys, especially those containing 15-20% ferrite, appear to be highly resistant to intergranular stress corrosion by oxygenated water, and even to have improved resistance to chloride-induced stress corrosion cracking.[12,15,17] Similarly, weld deposits made of Type 308, which commonly contain 5-15% delta-ferrite, appear to be resistant to intergranular stress corrosion. Stress corrosion cracks have commonly been observed to stop at the interface of the steel and the weld deposit. Apparently, the ferrite in these alloys provides an electrochemical protection for the austenite, and prevents the intergranular cracks from growing. Weld deposited 308 has been successfully used in BWR safe ends which unavoidably become sensitized during heat treatment of the ferritic steel pressure vessel.

c) Higher Nickel Alloys - The higher nickel alloys, such as Incoloy-800 and Inconel-600, show some improved resistance over Type 304 SS to stress corrosion cracking in oxygenated water in the furnace-sensitized condition[18], although Van Rooyen[20] has suggested that "sensitization" of Inconel-600 may improve its resistance to intergranular attack by high temperature, deaerated water, but possibly not in the presence of dissolved oxygen. Intergranular stress corrosion cracking of Inconel-600 in creviced areas in oxygenated water has been reported[21], and occurred on the primary side of the Obrigheim PWR steam generator tubes during a period when a N_2 rather than a H_2 overpressure was used on the primary circuit.[22,23] None of these materials, however, performed well under the simulated BWR environment tests performed by Clarke et al.[15]

d) Ferritic and Martensitic Steels - The ferritic steels, such as carbon steel and the low chromium steels, are resistant to stress corrosion cracking by oxygenated waters. Their use has been discouraged by the relatively high crud release rates and resultant high radioactivity levels throughout the external piping attributed to these materials. While these rates are higher in PWR coolants with a hydrogen overpressure, the oxygen in BWR coolants tends to minimize the differences in source terms between the low alloy steels and the stainless steels.[24,25] The higher Cr, martensitic alloys tend to be brittle and difficult to weld, and less resistant to stress corrosion cracking. Dissimilar joints at the weld-o-lets between the stainless steel 28" piping and the 4" ferritic steel bypass lines would be subject to additional stresses due to differential thermal expansion between the two piping materials.

Table 2 gives the results of severe screening tests performed by Clarke et al.[15] on a number of alternate materials. In the "most resistant" materials for which substantial work is available are found the austeno-ferritic materials, a low carbon stainless steel, a carbon-stabilized stainless steel, and a high Cr-Mo stainless steel. Adoption of one of these materials may well be the long-term solution to the BWR materials problem. The German BWR's use a carbon-stabilized stainless

steel, which is similar to our Type 347, except that the carbon levels are very low (~ 0.02%).

2. Advantages and Disadvantages of Alternate Alloys

Any alternate alloys selected for BWR piping, even if only as a replacement for the 4" bypass lines, must be readily available, fabricable, and a Code-approved material for Class 1 nuclear construction. The more exotic materials may therefore be excluded for short-term bypass line replacement materials, although their adoption should be seriously considered in the design of advanced, future BWR's, for which Code Cases could be developed as required.

a) Low Carbon Stainless Steels - All research with these materials, especially Types 304-L and 316-L, shows them to have good resistance to intergranular stress corrosion cracking. J. Danko stated at the task force meeting with GE and several utilities on January 16, 1975 that laboratory test data on welded Type 304-L in accelerated corrosion tests look very good. Furnace-sensitized Type 304-L stressed bend specimens exposed in Dresden-1 for 42 months showed no evidence of intergranular attack.[15] Type 304-L is a Code-approved material, and is relatively available. Therefore, *a priori*, it is an obvious choice for an alternate material. Type 304-L was also extensively used in the Dresden-1 reactor[27]. Type 316-L has the additional advantage that Mo retards its sensitization.[6]

There are, however, several disadvantages to using these materials. They have somewhat lower Code-allowable design stresses than Types 304 and 316, which may require a redesign of the BWR piping system. At the 0.03% carbon level, they can be furnace-sensitized, and synergistic effects between N, Ni, P and Si levels can lead to intergranular attack under some conditions. Ideally, a Type 304-L should have .02% maximum carbon, be low in phosphorous (~ .01%) and silicon (~ .1%), and contain no more than 10% Ni, based on the information now available.[6] This is clearly *not* a readily obtainable material. Type 304-L fuel cladding has failed by intergranular (and transgranular) stress corrosion in the

VBWR[28,29], as did annealed Type 304 SS. The roles in this attack of neutron radiation, stresses from fuel pellet swelling, the in-core nascent oxygen and hydrogen, and crud deposits known to contain halide ions (Cl^- and F^-) are not clear. For these reasons, this experience may not be applicable to external piping materials.

Although mill ladle analysis of the heat of stainless steel from which the Peach Bottom-3 bypass piping was believed to have been made indicates a carbon content of 0.024%, carbon analyses of the actual piping (performed by GE and Philadelphia Gas and Electric) show carbon levels > 0.04%.

Use of the Type 304-LN stainless steel would alleviate the lower design allowable strength of Type 304-L, but may involve a problem of availability. Since the Code for Type 304 does not specify nitrogen levels (except in the Code Case which deals with Type 304-N) or minimum C levels, this material might readily be used with the design rules for Type 304 or 304-N, provided it meets the minimum required mechanical properties. I would suggest a maximum of 10% Ni be placed on this material because of the synergistic effects of N, C and Ni, described by Cowan and Tedmon.[6]

b) Cb and Ti-Stabilized Grades - The European experience with the low carbon grades of these materials, and the good test results with Type 347[14,17] suggest that these materials might be acceptable alternates. They are Code-acceptable materials.

The "knife-edge" attack in weld sensitized areas of Type 347[6] and the higher alloy 20 Cb-3[30], difficulties in field welding these alloys, and lesser availability and greater cost of Types 321 and 347 or 20 Cb-3 make them less attractive alternates than Types 316-L, 304-L, or 316 or 304-LN.

c) Austeno-Ferritic Steels - All the evidence to date[15,17] suggests that use of these materials, especially the austinite-base materials containing 15-25% ferrite, with their excellent stress corrosion resistance and high mechanical strength (~ 50% stronger than

Type 304) would be an acceptable long-term answer to the stress corrosion problems in BWR's. However, some of them are still in the developmental stage, and the others are not readily available in the quantities and variety of shapes needed for immediate adoption. Further, years of testing in actual BWR environments would be needed to prove their applicability for reactor construction before their adoption to BWR's. At this time, their adoption seems unlikely, although fabrication and installation of a bypass line of this material may not be unreasonable, provided Code Case approval can be obtained. Use of cast stainless steel piping for large diameter lines also seems attractive since cast stainless steel contains residual delta ferrite.

d) Higher Ni Alloys - In oxygenated water, adoption of the higher Ni alloys, such as Inconel-600, Inconel-690 and Incoloy-800, does not appear to be a satisfactory solution to the bypass line corrosion problems, in view of the experience with Inconel-600 in Obrigheim[22,23] and the poor experience of Clarke et al.[15] with Inconel-690, cited in Table 2.

e) Martensitic Steels - The embrittling martensite phase, difficulties in welding, and the susceptibility of these materials to stress corrosion cracking in high temperature water make these materials unattractive alternates to Type 304 stainless steels for BWR piping.

f) Ferritic-Low Alloy Steels - The use of 2¼ Cr-1 Mo and 5 Cr-½Mo steels for the bypass lines appears to be entirely feasible at the present time. They are both resistant to stress corrosion cracking in BWR environments, and the steady state crud source term from these alloys appears to differ little from that due to stainless steels. Further, their lower Ni and Co contents would be beneficial in reducing the ^{58}Co and ^{60}Co activity in the crud. The major disadvantages of their adoption for the 4" bypass lines come from the difference in coefficient of expansion ($\sim 7 \times 10^{-6}/^{o}F$), compared with $\sim 10 \times 10^{-6}/^{o}F$ for the stainless steel 28" diameter piping, and the difficulties associated with oxidation during dry layup periods. These do not appear to provide

unsurmountable problems, however. Use of 5% Cr-½ Mo will lead to less pitting than observed with the lower Cr steels.[22]

E. Conclusions and Recommendations

1. The cracks that developed in the 4 inch bypass lines in BWR's and have been examined to date resulted from intergranular stress corrosion of weld-sensitized stainless steels by the oxygen-bearing BWR coolant. There is no reason to believe that examination of the remaining cracks in bypass and core spray lines will reveal anything substantially different from what is already known, although fatigue may have played a greater role in some plants than in others.

2. This phenomenon requires stresses greater than the yield stress, some degree of sensitization of the material, and an oxidizing agent in the coolant.

3. Stresses of the required magnitude arise from a combination of the residual stresses from welding and grinding with the applied mechanical, pressure, thermal and thermal expansion stresses.

4. The absence of fatigue striations on the fracture surfaces examined to date suggests that high cycle alternating (vibration) stresses were not the primary cause of the cracking. Alternating thermal and thermal expansion stresses during startup and shutdown of the reactors, however, may have contributed to the crack initiation process.

5. Some degree of sensitization will always be present in the heat-affected zones of welds in Type 304 stainless steel. Modification of welding techniques may reduce the amount of sensitization and the residual axial stresses, but cannot eliminate them. The more the stresses are relieved by thermal treatment, the greater the degree of sensitization will be.

6. The environment in the relatively stagnant bypass and core spray lines may be more conducive to intergranular stress corrosion cracking than that in the main primary and recirculation piping.

7. Crack growth rates can be relatively rapid once the cracks have been initiated. Further, the residual stresses from welding tend to promote lateral rather than through-wall propagation.

8. The probability of stress corrosion cracking of BWR piping, especially the lines less than 12" in diameter, would be greatly reduced if the Type 304 SS was replaced with a Type 304-L stainless steel containing a maximum of 10% Ni, or with a Type 316-L stainless steel. It would be eliminated if they were replaced with a 2¼% Cr-1% Mo or a 5% Cr-½% Mo steel. Other alloys, such as Type 304-LN and certain austeno-ferritic alloys, including cast stainless steel piping, look attractive for long term alternates, but may not be practical for adoption at the present time.

9. Tight, intergranular cracks are sometimes difficult to detect by ultrasonic techniques, especially after relaxation of the applied and thermal stresses causes the cracks to close tightly upon shutdown of the reactor.

10. Leak monitoring is therefore the most reliable means of detecting cracks before they can grow to the point at which failure or pipe rupture can occur during a design basis transient. Improved leak monitoring techniques should be installed in those areas of the piping where ISC has been found to occur. This should include local water collection and radiation detection equipment. Monitoring of leakage into the sump should be improved, and possibly the Technical Specification Limits for leakage from unidentified sources cut by 50% for some reactors.

11. The morphology of all the cracks in bypass and core spray lines needs to be assembled to assess the adequacy of the leak before break hypothesis.

12. There is no one piece of evidence offered to date that indicates why this phenomenon should be limited to the 4" bypass lines, the 10" core spray lines, or to other stagnant dead-leg lines. Therefore we must consider that this phenomenon may be occurring (or may occur at some future time) throughout the primary piping. However, the greater flexibility of the small diameter piping, the possibly greater residual stresses from welding it, and

the somewhat higher Ni content commonly found in seamless pipe than in seam-welded pipe, all suggest that conditions promoting ISC are less likely to develop in large (> 12") diameter welded piping than in smaller diameter seamless piping.

13. Any actions to reduce the applied stresses and make the environment less conducive to crack initiation would be desirable. These should include leaving the bypass valves open to reduce thermal stresses and improve circulation in the pipes, and improving piping supports to reduce dead load and vibrational stresses. These actions alone, however, may not be sufficient to eliminate the cracking problems.

14. It is not clear at present whether the chloride intrusion at Millstone-1 had any effects on the intergranular stress corrosion problems that have developed at that plant.

15. The Regulatory Position 4 given in Regulatory Guide 1.44 should be implemented for all future BWR's for that portion of the primary piping that cannot be isolated from the reactor vessel by a valve. This position is that, "material subjected to sensitizing temperature in the range of 800 to 1500°F, subsequent to solution heat treating .... should be L Grade material; that is, it should not have a carbon content greater than 0.03 percent. Exceptions are: (a) Material exposed to reactor coolant which has a controlled concentration of less than 0.10 ppm dissolved oxygen at all temperatures above 200°F during normal operation; or (b) material in the form of castings or weld metal with a ferrite content of at least 5 percent; or (c) piping in the solution annealed condition whose exposure to temperatures in the range of 800 to 1500°F has been limited to welding operations, provided it is of sufficiently small diameter so that in the event of a credible postulated failure of the piping during normal reactor operation, the reactor can be shut down and cooled down in an orderly manner, assuming makeup is provided by the reactor coolant makeup system only."

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TABLE I

Compositions (in wt.%) of Materials[15,18,25]

Material	Fe	Cr	Ni	C (max)	Mn	Mo	Si	Other
Austenitic Stainless Steels								
Type 304	bal.	19	9	.08	2.0 max	.2	1.00 max	
Type 304-L	bal.	19	9	.03	2.0 max	--	1.00 max	
Type 308-L*	bal.	20	11	.03	2.0 max	--	1.00 max	
Type 309	bal.	23	13	.20	2.0 max	--	1.00 max	
Type 316	bal.	17	12	.08	2.0 max	2.5	1.00 max	
Type 321	bal.	18	10.5	.08	2.0 max	0.4	1.00 max	Ti > 5xC; .7 max.
Type 347	bal.	18	10.5	.08	2.0 max	0.4	1.00 max	(Cb + Ta) > 10xC; 1.1 max.
Type 304-LN	bal.	20	9.1	.026‡	1.67	0.2	.46	
22-13-5	bal.	22	12	.04	5	2.2	.5	.2 (Cb + Ta); .3 N
18-3 Mn	bal.	18	3.7	.05	12.8	.2	.5	.3 N
20-Cb-3	bal.	20	34	.036‡	.23‡	2.25	.38	3.14 Cu, 0.79 Cb
Austeno-Ferritic Steels								
Uranus-50	bal.	21	8	.04‡	0.6	2.5	0.5	1.5 Cu
Sandvik 3RE60	bal.	18.5	4.5	.03‡	1.6	2.5	1.8	0.07 N
Higher Ni Alloys								
Incoloy-800	43.5	20.5	33.5	.04‡	.8	--	.25	.4 Ti
Inconel-600	7.4	14.9	bal.	.04‡	.2	--	.2	
Inconel-625	2.4	21.3	bal.	.02‡	.01	9	.2	3.65 (Cb + Ta)
Inconel-690	9	30	bal.	.02‡	.2	--	.1	.3 Ti
Ferritic & Martensitic Steels								
2¼ Cr-1 Mo	bal.	2.25	--	.15	1.0 max	1.0	.25	
5 Cr-½ Mo	bal.	5	--	.10	.5	0.5	.4	(Type 502)
E-Brite 26-1	bal.	26	.1	.001‡	.01	1.0	.2	
19 Cr-2 Mo	bal.	19	.1	.002‡	.4	2.0	.4	
Custom 450	bal.	15	6.5	.04‡	.4	0.8	.01	1.5 Cu; .6 (Cb + Ta)

* Weld pad (7)

‡ Actual

TABLE 2

Relative Ratings of Alternate Materials in G.E. Simulated BWR Environment Tests[15]

Most Resistant	Intermediate Resistance	Least Resistant
Uranus-50	Type 321	Inconel-690
E-Brite	Alloy 6X	Custom 450
304 LN	Sandvik 3RE60	18-3 Mn
22-13-5	Incoloy-800	
Type 347		