ELECTRIC POWER RESEARCH INSTITUTE

Repair and Replacement Applications Center: Temperbead Welding Applications

48-Hour Hold Requirements for Ambient Temperature Temperbead Welding

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Technical Update, December 2006

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ABSTRACT

The purpose of this white paper is to provide technical justification to allow for the 48-hour delay to begin following completion of the third temperbead weld overlay (WOL) weld layer applied to P-3 low alloy steel pressure vessel component. This represents an alternative to the current requirement of completing the entire weld and cooling to ambient temperature prior to starting the 48-hour hold. This paper provides further technical discussion/bases to that presented by the white paper written in support of action SGRRA -05-08 aimed at modifying the requirements in Code Case N-638. The industry and regulatory technical concerns related to this change are examined and the technical bases for changing the requirements for the 48-hour delay are presented. The specific focus of this document is to address four key concerns associated with the 48-hour hold as summarized below:

-Microstructural Issues
-Sources for Hydrogen Introduction
-Tensile Stress and Temperature
-Diffusivity and Solubility of Hydrogen in Steels

Each of the areas of NRC concern has been addressed in the body of the white paper. The following summarize the results:

- The microstructure in the P-3 material directly beneath the temperbead WOL consists of a tempered martensite or tempered upper bainite that has excellent toughness, combined with a modest maximum hardness (of the order of Rc 30 or lower).

- The microstructure at the toe of the temperbead WOL in the P-3 weld HAZ at the OD surface where tempering is somewhat limited may have a very small HAZ with a maximum hardness of the order of Rc 36, at a distance of approximately 40 mils from the toe of the WOL. At a depth of approximately 2 to 2.5 mm (80 to 100mils) beneath the toe of the WOL, the hardness is reduced to less than 294 KHN (29Rc), a hardness level well below that required to cause hydrogen cracking.

- Sources of hydrogen include moisture, poor shielding gas and contamination. It is noted that moisture in the shielding gas or high humidity is not a problem for GTAW temperbead welding. Contamination will affect the weld, and should be identified either during the welding process or during the subsequent NDE of the overlay. Good welding practice should eliminate this problem for the temperbead WOL.

- Tensile stresses should not be an issue for cold cracking as the thermal stresses diminish with each weld overlay layer. Following the final layer, it is expected that the maximum surface temperature at the toe of the WOL in the P-3 HAZ would reach temperatures only on the order of 400°F to 500°F. Slow cooling to ambient temperature from these temperatures would be expected to produce relatively small stresses.

- The diffusion rate for hydrogen is greater in ferritic materials than in austenitic materials, but the solubility of hydrogen in austenite is from five to seven times greater in the austenite than in the ferrite or martensite. Consequently, due to the temperatures expected during the welding of the temperbead layers, and during the welding of the non-temperbead WOL layers, the temperature should be sufficient for the hydrogen to diffuse out of the material, either escaping the structure or diffusing into the austenite, where it can be held in much greater quantities. Thus, even if hydrogen is produced, a large hydrogen inventory in the P-3 material is not expected.

In summary, this study has identified a small finite region at the toe of the temperbead WOL in the P-3 material on the OD surface where it is possible for the hardness to reach a lower threshold level for hydrogen cracking in these components. Even in the very unlikely case that cracking was to occur, it would not be structurally significant.

There appears to be no technical basis for waiting the 48 hours after cooling to ambient temperature before beginning the NDE of the completed weld. There should be no hydrogen present, and even if it were present, the temperbead welded component should be very tolerant of the moisture

Finally, as noted in this report, there have been more than 20 temperbead WOLs applied to nuclear power plant low alloy steel nozzles and more than 100 temperbead repairs to other low alloy steel components. The initial overlays have been in service more than 20 years. Of all the temperbead overlays applied in the field, there has been no indication of hydrogen cracking found from NDE performed after the 48-hour hold or from subsequent ISI.

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OBJECTIVE

The objective of this white paper is to provide technical justification to allow for the 48-hour delay to begin following completion of the third temperbead weld layer applied to P-3 low alloy steel pressure vessel component. This represents an alternative to the current requirement of completing the entire weld and cooling to ambient temperature prior to starting the 48-hour hold. This paper provides further technical discussion/bases to that presented in the white paper written in support of the action SGRRA -05-08 aimed at modifying the requirements in Code Case N-638. The industry and regulatory technical concerns related to this change are examined and the technical bases for changing the requirements for the 48-hour delay are presented.

The NRC has identified several concerns in response to SGRRA 05-08 and the supporting white paper associated with moving the beginning of the hold period from when the completed weld reaches ambient temperature to the completion of the third temperbead overlay layer. The issues sighted by the NRC can be summarized into four areas:

-Microstructural Issues
-Sources for Hydrogen Introduction
-Tensile Stress and Temperature
-Diffusivity and Solubility of Hydrogen in Steels

A temperbead background section is provided for a historical perspective of the development of the temperbead process used for repair/replacement activities. In the following sections of this paper, each of e the technical issues is defined and discussed, and relevant research work is identified and described which may provide insight as to the magnitude and effect of each issue. Where options are available to mitigate the issue, they are noted. A conclusion section endeavors to summarize and define the technical bases for the change.

2
TEMPERBEAD HISTORICAL BACKGROUND

The initial temperbead weld overlay (WOL) repairs were applied to two core spray nozzle to safe end joints at the Vermont Yankee boiling water reactor (BWR) in 1986 [1]. The repair was necessitated by suspected intergranular stress corrosion cracking (IGSCC) in the Alloy 182 butter welded to the nozzle. The buttered nozzle was then joined to the Alloy 600 safe end using Alloy 82 weld metal. The earliest temperbead weld overlays required preheat and a post weld bake, necessitating that the component to be repaired needed to be dry on the inside surface. The purpose of these weld overlay repairs has been to provide a replacement pressure boundary for low alloy steel and the affected joint by replacing the existing defected component location in its entirety. The WOL repair is welded using a machine gas tungsten arc welding (GTAW) technique. IGSCC resistant nickel based weld metal (currently Alloy 52/52M weld metal), and the weld overlay application covers the original dissimilar metal weld, the heat affected zones on each side of the weld, and a sufficient distance beyond the heat affected zones to provide for adequate load transfer from the original components through the WOL and back to the sound base metal. The repair design assumes that the underlying flaw is entirely through the original component wall, and extends completely around the circumference of the component. For design purposes no credit is taken for any remaining ligament in the repair location. This assumption is clearly quite conservative, and is consistent with the approach that has been used for weld overlay repairs of BWR recirculation and other IGSCC-susceptible material since 1985.

Because a portion of one of the potential repairs requires welding on the low alloy steel nozzle SA-508 Class 2 material (designated P3), a temper bead welding process is used in any such repairs applied to the P3 nozzle material, as the nozzle material is required to be sufficiently tempered so that no delayed hydrogen or cold cracking can occur, and so that the material toughness is adequate to carry both the static and dynamic loads that affect the component.

One key advantage of the temperbead weld overlay repair is that it provides an alternative to the traditional post weld heat treatment that had been required for low alloy steel materials, provided that certain activities are performed. During the earliest applications of the dissimilar metal WOLs, these actions have consisted of requirements such as preheat, post weld bake, and a 48-hour hold at ambient temperature following the completion of the welding and post weld bake. The initial American Society of Mechanical Engineers Boiler and Pressure Vessel Code (ASME Code) Case that allowed for temperbead repairs of pressure vessel nozzles was Code Case N-432 [2], which required the application of preheat, post weld bake, and a 48-hour hold at ambient temperature following the post weld bake.

In the years following the application of the initial temperbead WOLs, the excellent field performance of these WOL repairs have allowed for increased use of these repairs to dissimilar metal welds involving P3 materials. However, the increased personnel radiation exposure attendant with the dry welding, combined with the technical difficulties of draining the component, and the additional outage and repair costs, has argued for the development of a

method whereby the welding could be performed with water backing, at ambient temperature, and without the required post weld bake. As a result of these needs, during the past several years, EPRI [4,5] and other industry investigators have demonstrated that repairs to low alloy steel pressure vessel components could be made without need for the preheat or post weld bake. As a result of these investigations, dissimilar metal nozzle to safe end or pipe welds, joining P3 low alloy steel nozzle components to austenitic piping components have been performed recently in both BWRs and in PWRs using ambient temperature temperbead welding techniques that did not require preheat nor post weld bake.

During the past twenty years, tens of temperbead weld overlay repairs were successfully performed on BWR and PWR nozzle joints using this technique, as illustrated in Table 2-1 [3].

Table 2-1
Successfully Implemented Weld Overlay Projects for Dissimilar Metal Butt Welds [3]

Date	Plant	Component
Fall 2006	McGuire	Pressurizer Nozzles
Fall 2006	Byron 2	Pressurizer Nozzles
Fall 2006	Beaver Valley 2	Pressurizer Nozzles
Fall 2006	Oconee 1	Pressurizer Nozzles
Fall 2006	SONGS 3	Pressurizer Nozzles
2002	Calvert Cliff	~ 50 Heater Sleeve Repairs (Pads)
2003	Palo Verde 1	36 Heater Sleeve Repairs (Pads)
2004	Palo Verde 2	34 Heater Sleeve Repairs (Pads)
2005	Palo Verde 3	36 Heater Sleeve Repairs (Pads)
Fall 2005	SONGS 3	~ 29 Heater Sleeve Repairs (Pads)
Spring 2006	SONGS 2	~ 30Heater Sleeve Repairs (Pads)
2005	Waterford	Mid Wall heater sleeve repair
2005	ANO	Mid Wall heater sleeve repair
Fall 2005	TPC Kuosheng	N1 Nozzle
Spring 2006	Millstone 3	Pressurizer Nozzles
Spring 2006	SONGS 2	Pressurizer Nozzles
Spring 2006	Davis Besse	Hot and Cold Leg
Spring 2005	DC Cook Unit 1	Pressurizer Safety Nozzle
February 2005	Calvert Cliffs Unit 2	Hot Leg Drain and Cold Leg Letdown Nozzles
April 2004	Susquehanna Unit 1	Recirculation inlet and outlet nozzles
November 2003	TMI Unit 1	Hot leg and Surge line nozzle
October 2003	Pilgrim	Core spray nozzle CRD return nozzle
October 2002	Peach Bottom Units 2 & 3	Core spray, recirculation outlet, and CRD return nozzles
October 2002	Oyster Creek	Recirculation outlet nozzle
December 1999	Duane Arnold	2 Recirculation inlet nozzle
June 1999	Perry	Feedwater nozzle
Spring 2000	Nine Mile Point Unit 2	Feedwater nozzle
October 1999	Hope Creek	Core spray nozzle
March 1996	Brunswick Units 1 & 2	Feedwater nozzle
February 1996	Hatch Unit 1	Recirculation inlet nozzle
January 1991	River Bend	Feedwater nozzle
March 1986	Vermont Yankee	Core spray nozzle (Not ambient Temperature)

These results have led to the adoption of Code Case N-638 [6] by Section XI of the ASME Code that allows for GTAW temperbead welding on P3 materials without preheat and the post weld bake when conventional temperbead welding is impractical. Code Case N-638 has become widely used in the nuclear industry, and it has been demonstrated to be an effective and sound approach for temperbead welding of WOLs to low alloy steel pressure vessel nozzles.

Whereas the Code Case N-638 provisions allow for ambient temperature welding, and no post weld bake, the post weld 48 hour hold at ambient temperature has remained as a provision for such temperbead welding. This 48 hour delay between welding completion and cooling to ambient temperature and the final non-destructive examination (NDE) of the fully welded component is provided so that any delayed hydrogen cracking or cold cracking that may occur can take place prior to the final NDE of the weld overlay and surrounding area. The post weld 48 hour delay following cooling to ambient temperature has resulted in a substantial cost burden to the involved utilities. As there are significant economic advantages associated with starting the 48 hour clock immediately following the final temperbead layer of the WOL, it is important to determine the technical advantages and disadvantages of making such a change.

3
TECHNICAL ISSUES RELATED TO THE 48 HOUR DELAY

The reasons for performing the 48 hour delay following the completion of welding and cooling to ambient temperature is driven by the knowledge that alloy steels can become susceptible to cracking as a result of welding or heat treatment. Two primary degradation mechanisms are of concern for low alloy steel welding at ambient temperatures. These are cold cracking of high restraint geometries (weld shrinkage-induced) and hydrogen assisted cracking (hydrogen delayed cracking).

EPRI studies [4] have indicated that restraint cracking occurs under conditions of high geometrical restraint especially where low toughness HAZs are potentially present. Restraint mechanisms can occur either hot (resulting in intergranular or interdendritic cracking), or cold (resulting in transgranular cracking of material having marginal toughness). Proper joint design, appropriate welding procedures and bead sequences, and effective tempering are practical concerns that avoid critical restraint cracking conditions. The other form of cracking at ambient temperature is delayed hydrogen cracking. This cracking mechanism manifests itself as intergranular cracking of prior austenite grain boundaries and generally occurs within a period of 48 hours after completing the weld. It is produced by the action of internal tensile stresses acting on low toughness HAZs (generally characterized by inadequate tempering of weld related transformation products). The internal stresses will be produced from localized buildups of monatomic hydrogen. Monatomic hydrogen can be entrapped during weld solidification, and will tend to migrate, over time, to prior austenite grain boundaries or other microstructure defect locations. As concentrations build, the monatomic hydrogen will recombine to form molecular hydrogen, thus generating highly localized internal stresses at these internal defect locations. Monatomic hydrogen is produced when moisture or hydrocarbons interact with the welding arc and molten weld pool.

Issues associated with restraint cracking and with hydrogen delayed cracking have been identified by the technical community as noted in the Objective section of this White Paper. These issues are fundamental welding and heat treatment issues related to temperbead welding, requiring a technical resolution prior to modification of the current ASME Code Cases N-638 and N-740 by the ASME Code and the technical community. Specific concerns relate to the following issues:

-Microstructure

-Sources for Hydrogen Introduction

-Tensile Stress and Temperature

-Diffusivity and Solubility of Hydrogen in Steels

In the following sections of this report, each of these issues is addressed, and relevant research work is identified and described which may provide insight as to the magnitude and effect of each issue. Where options are available to mitigate the issue, they are noted.

3.1 Microstructure

3.1.1 EPRI Studies

EPRI has examined in detail the effects of welding on the hardening of low alloy steels. In the Reference [5] report, the author noted that two regions may become hardened as a result of welding. The first region is the high temperature region adjacent to the weld fusion line. The welding heat raises the temperature of a small zone of adjacent base material (or weld substrate) above a critical transformation temperature range (A_{C1}) unique to the material composition whose microstructure within this zone transforms to austenite. Upon cooling the material retransforms into martensite, upper bainite, lower bainite, pearlite and ferrite, and a percentage of retained austenite. Figure 3-1 presents a Continuous Cooling Transformation Diagram for SA-508 Class 2 low alloy steel taken from Reference 5. One notes from this curve that any preheat or post weld bake below 550°F will be below the martensite finish temperature and will have no effect on the structure [5]. The transformation products depend upon the hardenability of the material (specific composition) and the rate of cooling through the transformation temperature range. A high hardenability composition will favor martensite and upper bainite, and a low hardenability composition will favor bainite and pearlite microstructures. In the untempered state, the bainite and pearlite mixtures will be more tolerant to stress and weld distortion. When tempered the microstructure of the higher hardenability material is superior. The more rapid the cooling rate through the critical transformation temperature range, the transformation from austenite to martensite will be favored for any given composition. Therefore any action that slows the cooling rate will tend to favor bainite and pearlite in the microstructure. This microstructure mixture will have increased tolerance to stress cracking when untempered, but will be a less desirable in the tempered condition. The application of preheat is such an action because the rate of cooling is slowed over the no preheat condition. It is interesting to note that this is one reason that interpass temperature is strictly controlled in low alloy steels.

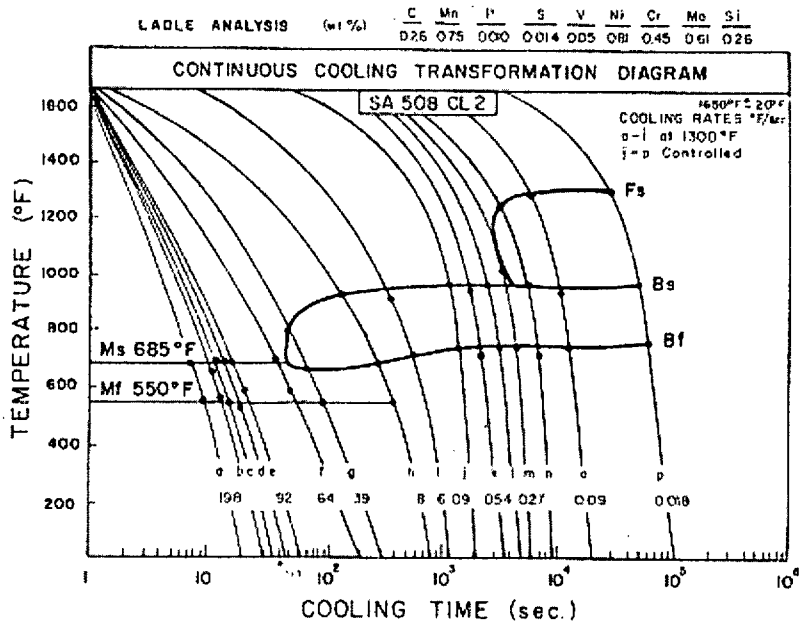

Figure 3-1
Continuous Cooling Transformation Diagram of SA-508, Class 2 Steel (5,15)

Untempered martensite and upper bainite microstructures are very hard and brittle structures. Carbon and other interstitial atoms are locked within the microstructure preventing deformation mechanisms to take place and resulting in increasing internal stress. This produces strength, but unfortunately, also produces low ductility and thus low toughness. When heat is applied to the structure in the range of 1250 to 1325 ^{0}F the carbon will begin to precipitate as carbides and the microstructure will toughen very rapidly. Tempering is both time and temperature dependent. Tempering is rapid at the upper temperatures, but drops off very rapidly as temperature decreases. Very little tempering occurs at traditional stress relief temperatures between 1050 and 1150 ^{0}F, because carbon is less mobile and diffusion is slow at lower temperatures.

The temperbead process is designed to provide a sufficient heat inventory so as to produce the desired tempering for high toughness. For this reason, the high martensite microstructure is desired because of the improved toughness that will be developed. The toughness and ductility of the HAZ typically will be superior to the base material substrate, because the quenching rate will be faster than the original product form. The machine GTAW temperbead process provides precision bead placement and control of heat. Therefore the resulting structure will be tempered and preheat is not desired [5].

EPRI and the utility industry performed extensive research in the 1980's and 1990's related to temperbead welding on P-3 materials and the effects of that welding on the microstructure, material hardness, and toughness of this material [4,5]. Large test coupons and overlaid nozzle mockups were examined and evaluated so that the extent of the weld heat affected zone could be determined [4].

An additional study was performed as part of the Reference 3 activity, in order to evaluate the effect of welding preheat temperatures on the subsequent hardness of SA 508 Class 2 low alloy steel (P-3). In this study, five individual weld beads (a single layer of five adjacent beads) were performed on 4-inch thick, SA508, Class 2 steel coupons using the machine GTAW process. Welding parameters for the GTA weld beads are presented in the Reference 3 report. Preheat temperatures for the five preheat conditions ranged from room temperature (75°F) to 500°F. The hardness results are shown in Figure 3-2. As expected, the room temperature (no preheat) weld bead resulted in the highest peak hardness values, while the highest preheat temperature resulted in the lowest peak hardness values. A substantial difference was observed between the 75°F and 500°F preheat values. The 75°F weld bead HAZ resulted in peak hardness values of up to 480 Knoop hardness (KHN) (47Rockwell-C (Rc)) compared to the 500°F preheat maximum hardness of 360 KHN (36Rc).

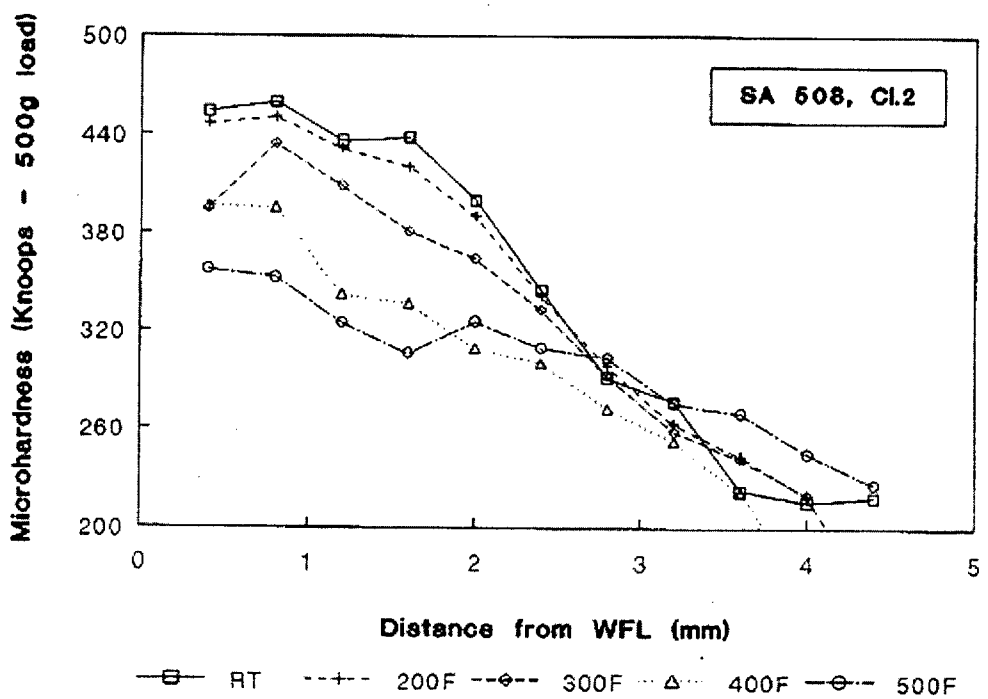

Figure 3-2
Influence of Preheat on HAZ Hardness [4].

It is noteworthy, however, that within approximately 2.5 mm (100 mils) beneath the surface, within the heat affected zone, the hardness values for each of the heat treatment conditions converge and remain essentially the same, independent on the heat treatment condition at a hardness value of approximately 300 KHN (30Rc). This is noteworthy since the ASME Code, Section III allows for welding on cladding or buttering or on dissimilar metal welds without requiring postweld heat treatment providing the remaining buttering or dissimilar metal weld thickness is greater than 3.2 mm (1/8 inch) prior to the welding repair [7]. In addition, even in

the 75°F preheat condition, the peak hardness falls below a hardness of 38Rc at a distance of 2 mm (80 mils) beneath the surface. The peak hardness also occurs at roughly 1 mm (40 mils) beneath the weld in the HAZ for these five heat treatment conditions.

As part of the Reference 4 study, EPRI specifically investigated the effect of dissimilar metal temperbead welding on P-3 materials, with and without preheat and post weld bake on the microstructure and on the hardness of the underlying P-3 material. In one of the earliest studies, the micro-hardness was determined beneath a single weld bead, and beneath three adjacent beads. Figure 3-3, taken from Reference 4, presents the results of this investigation. One notes from the figure, that for the single bead, the peak hardness is quite high, reaching a level of great than 520 KHN (approximately 52Rc). For the three adjacent bead weld, the peak hardness measured directly below the vertical centerline of the three bead single pass weld was of the order of 380KHN (or around 38Rc), illustrating bead to bead tempering within the initial layer. It is also noteworthy that the peak hardness occurred in the heat affected zone, approximately 1 mm (40 mils) from the weld fusion line (WFL).

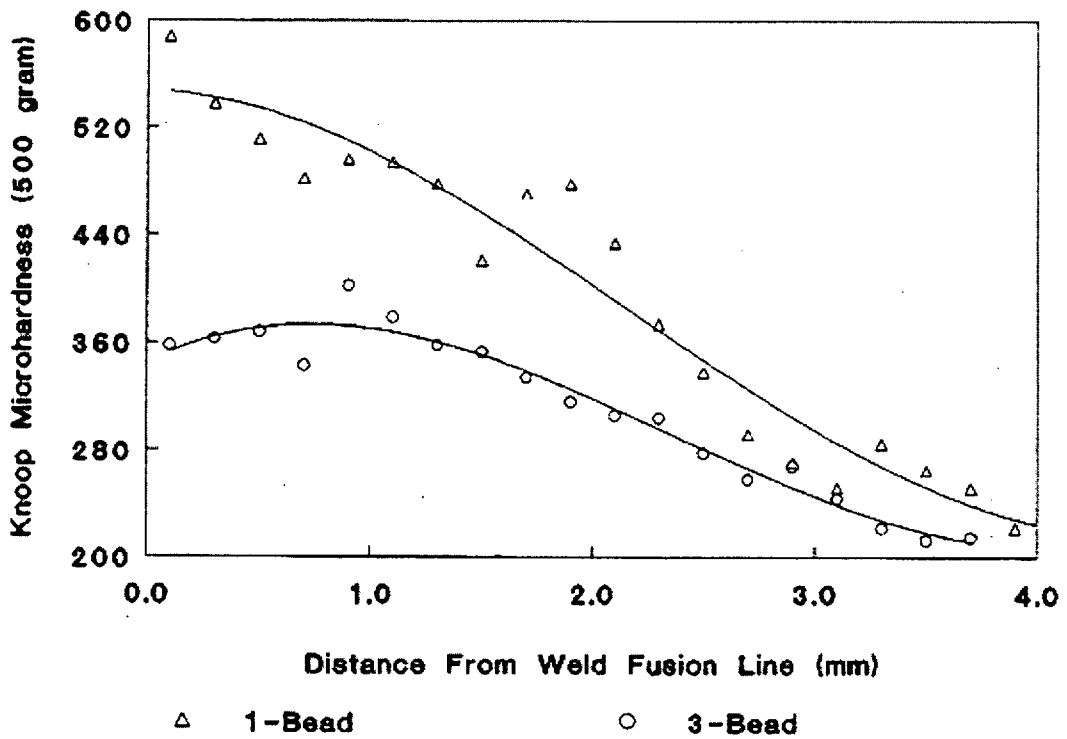

Figure 3-3
Comparison of HAZ Hardness Curves for a Single Weld Bead and a Three Weld Bead Influenced by Adjacent Beads [4]

Following the completion of the basic study comparing a single bead to three bead temperbead weld, EPRI investigated the maximum achievable properties that could be obtained following postweld heat treatment. Two 3-layer pads were deposited on a 3-inch thick SA 508 Class 2 low alloy steel material (P-3) that had been used in an earlier EPRI study. Following welding each of the pads was sectioned. The sectioning included removal of a 1-inch wide coupon from each weld pad which was utilized for metallography and hardness testing. The sectioned coupons represented the as-welded condition.

The remainder of each pad was then postweld heat-treated. The first weld pad was heat-treated at 1125°F for three hours. The second weld pad was heat-treated at 1250°F for three hours. The 1250°F temperature is the maximum allowable temperature recognized for postweld heat-treatment of P3 materials. Comparisons of the as-welded and postweld heat-treated specimens were accomplished using micro-hardness data. Hardness plots are shown in Figures 3-4 and 3-5 for each of the conditions. The peak hardness beneath these welds in the as welded condition was of the order of 370 KHN (37Rc), and this peak occurs within approximately 0.5 to 1.0 mm (20 to 40 mils) below the weld fusion line. The hardness dropped below 300 KHN (30Rc) within 2.5 mm (100 mils) of the weld fusion line. Figures 3-4 and 3-5 also illustrate that the 1125°F postweld heat-treatment provided only a small improvement in the hardness values, whereas, the 1250°F provided a decrease in average hardness of approximately 40KVN (4Rc). In either case, it appears that the lowest peak hardness achievable for postweld heat-treatment of this particular heat of SA508, Class 2 steel is slightly below 320KHN (32Rc).

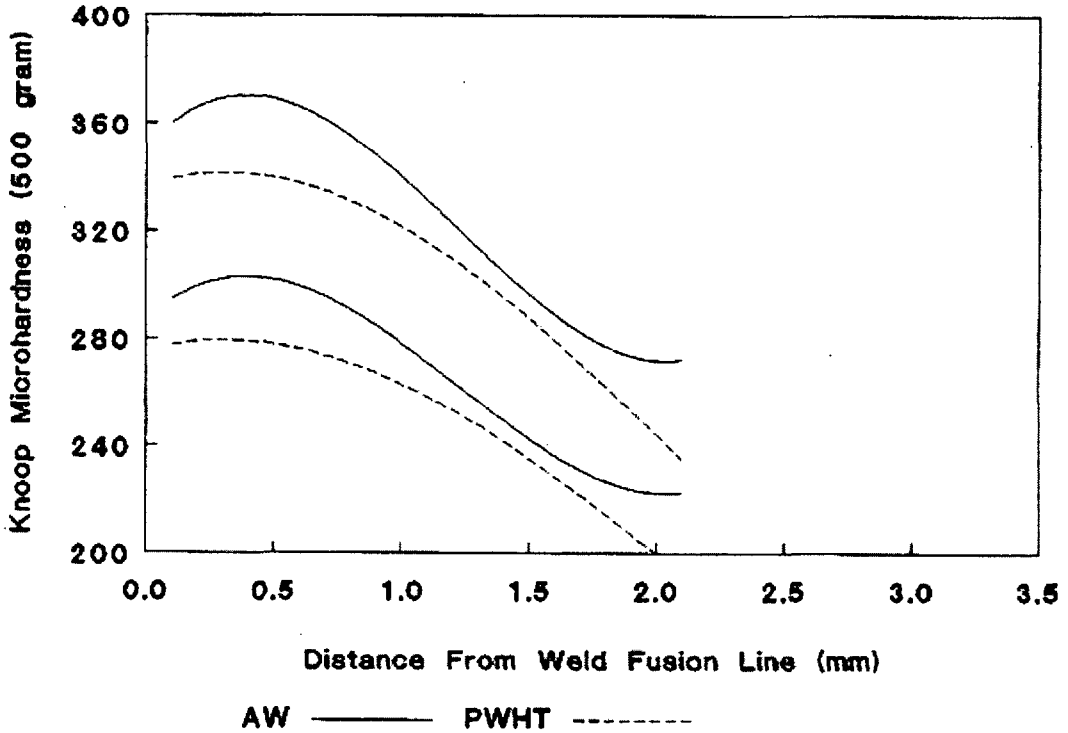

Figure 3-4
Comparison of HAZ Hardness Curves of a 3-Layer Weld Pad in the As-Welded and 1125°F Postweld Heat Treated Condition [4].

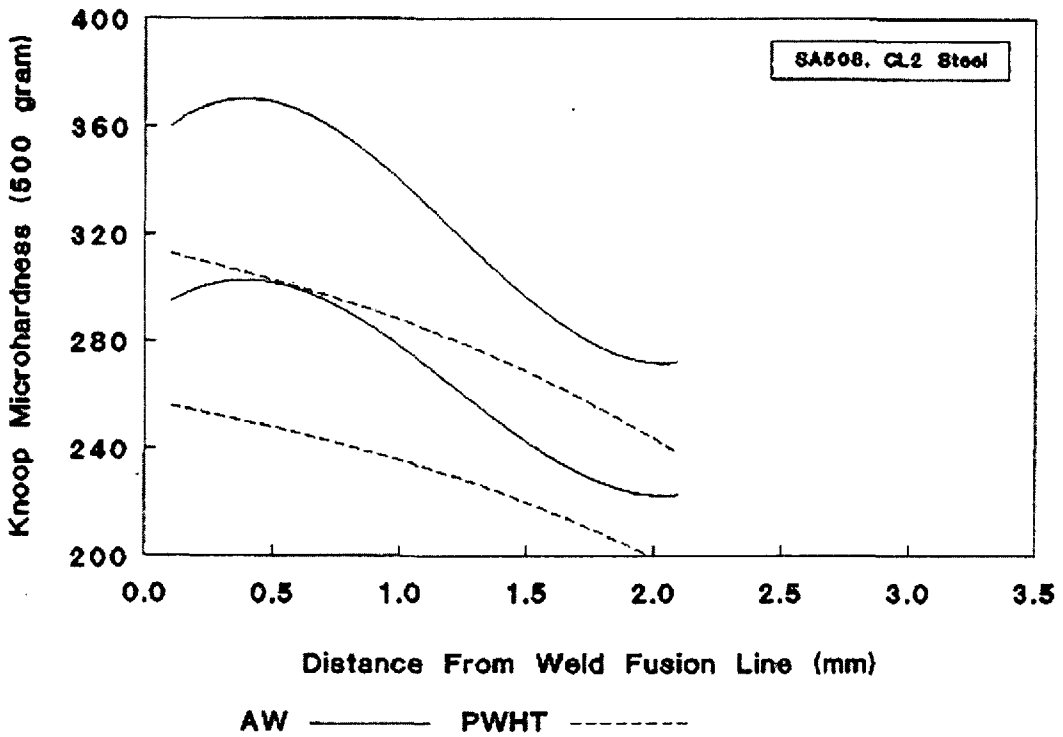

Figure 3-5
Comparison of HAZ Hardness Curves of a 3-Layer Weld Pad in the As-Welded and 1250°F Postweld Heat Treated Condition [4].

A number of additional weld pads were fabricated using varying welding conditions and several metallurgical sections and micro-hardness tests were performed during this very extensive study of the effects of temperbead welding on low alloy steel P-3 materials [4]. The studies each produced results that were similar to those presented in Figures 3-3 through 3-5 above. In all cases the peak heat affected zone hardness for the single layer through three layer weld overlays was of the order of 380KHN (38Rc) or less, for the worst case, at a typical distance of 1 mm or less (40 mils) beneath the WFL. The hardness dropped rapidly with distance from the WFL reducing to near the base metal hardness within 3 to 4 mm (120 to 160 mils) of the WFL in all cases (See Figure 3-2).

In addition to the micro-hardness results reported here, most of the test programs reported in Reference 4 were supported by detailed metallography of the microstructure associated with the hardness measurements. These supporting results are not presented here, but are included in the Reference 4 report. The results confirm that the microstructures are consistent with the hardness values..

3.1.2 Temperbead WOL Application at Vermont Yankee

As noted earlier in this report, the first application of a temperbead WOL to a nozzle safe end configuration was applied to a core spray nozzle safe end weld in 1986 at the Vermont Yankee BWR station and described in EPRI report on application of temperbead weld overlays to nozzle safe end joints [1]. Two overlays were applied to repair suspected intergranular stress corrosion cracking (IGSCC) located in the Alloy 182 butter applied to the end of the nozzle. The buttered nozzle had been joined to the Alloy 600 safe end using Alloy 82 weld metal. The WOL was applied using the 300°F minimum preheat, and the 450-550°F post weld bake that was specified by Code Case N-432 [6] which was the controlling code case at the time for temperbead welding.

Since this was the first of a kind application of the temperbead WOL to an operating BWR, in addition to the standard weld qualification testing performed, additional supplementary testing was conducted on a representative full scale mock-up simulating the safe end to nozzle weld. The supplemental testing involved substantial metallographic studies, micro-hardness measurements, and compositional studies of the effect of the temperbead weld overlay on the low alloy steel properties.

In this investigation, micro-hardness measurements were performed at representative sections of the groove weld, the dilution layer study specimens, and the Alloy 600 weld overlay mockup. The results of these measurements are included in the Reference 1 report and are not repeated here, with the exception of one series of measurements. The toe region of the weld overlay on the P-3 side of the overlay produced a modestly hard P-3 heat affected zone at the overlay edge. Figure 3-6 presents a schematic representation of the overlay bead layout as well as the region where micro-hardness measurements were taken and Table 3.1 presents these microhardness results in a tabular format.

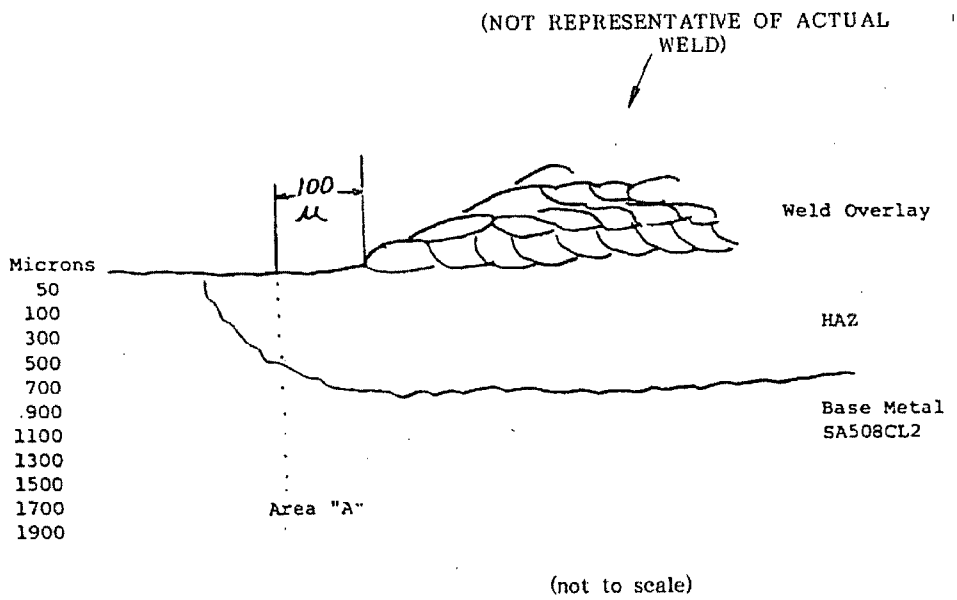

Figure 3-6
Location of Microhardness Traverse at Overlay Toe in SA 508 Class 2 Heat Affected Zone [1].

Table 3-1
Microhardness Results in P-3 Heat Affected Zone Near Toe of Alloy 600 Weld Overlay, Area A. [1]

Radial Distance From Nozzle Outside Surface (μ)	Vickers Hardness (500 gm)	Equivalent Rockwell C Hardness
50	335	34
100	354	36
300	343	35
500	319	32
700	348	35.5
900	294	29
1100	268	25.5
1300	252	22.5
1500	240	20.5
1700	232	HRB 96.5
1900	199	HRB 91

One observes that a hardness maximum of 354 KHN (36Rc) is observed in the P-3 heat affected zone. This contrasts to hardness levels of no greater than 320 KHN (32Rc) in the heat affected zone beneath the weld overlay. It is believed that this increase in hardness at the overlay extremity is the result of marginal bead overlap of the second and third overlay layers at the toe of the overlay. It is also noteworthy that the hardness peak occurs less than 1 mm (40 mils) beyond the toe of the weld, and is approximately 354 KHN (36Rc). At a depth of approximately 2 to 2.5 mm (80 to 100mils) beneath the surface in the HAZ, the hardness is below 294 KHN (29Rc), a hardness level unlikely to be capable of supporting hydrogen cracking.

3.2 Sources for Hydrogen Introduction and Effects on Temperbead WOLs

EPRI performed extensive studies during the 1990's investigating the diffusion of hydrogen into low alloy pressure vessel steels [4, 5]. In one of these studies [4], the investigation included a major literature review of sources of hydrogen and hydrogen induced cracking. The study described the sources by which hydrogen is introduced into the weld metal primarily as a result of flux constituents and poor gases utilized for shielding. These discussions and results are presented in the Reference [4] report and will not be repeated here.

Sources of hydrogen introduction into the weld puddle during the GTAW process was also reviewed in this study [4]. One investigator, Linnert, indicated that "welding with a gaseous shield of welding grade argon, helium, or carbon dioxide does not add any significant amount of hydrogen-bearing compounds into the weld metal" [8]. Though reduced, the potential of hydrogen-induced cracking may still be present as a result of other sources [9]. Such sources include:

1. Moisture.
2. Poor grade shielding gas mixtures with hydrogen which has been intentionally added to enhance penetration and arc characteristics
3. Drawing compounds on welding filler wire.
4. Base metals (improved steel making processes have reduced hydrogen levels of most steels)
5. Lubricants, coatings, or paints.

Based upon the literature review, and the observation that very little information had been published on diffusible hydrogen levels associated with the GTAW process, EPRI resolved to generate experimental data on developing definitive information on the levels of diffusible hydrogen associated with GTAW welding. The experimentation included three individual sets of diffusible hydrogen tests as follows:

1. Severe Environment Tests
2. Diffusible Hydrogen levels At Various Shielding Gas Dew Point Temperatures
3. Off-the-Shelf Filler Wires.

4. Contamination from Lubricants, Coatings, Paints or other sources.

3.2.1 Severe Environment Tests

The initial battery of severe environment tests performed as part of the Reference 4 study included tests conducted in a purge chamber shown in Figure 3-7, where the atmosphere was maintained at 120°F and at a 95% (or greater) humidity throughout the diffusible hydrogen welding tests. In essence, this environment produced a fog-like welding condition. A second set of samples was tested at 150°F. A total of four samples comprised a complete test set.

The significance of the measurements obtained from the tests discussed above is that the GTAW process limits the hydrogen to extremely low levels. These results demonstrate that diffusible hydrogen levels, on the order of 1.0 ml/100g H_2, should be expected when welding at a preheat temperatures between 150°F and 300°F. However, the hydrogen levels prior to the test were expected to be dependent primarily upon the moisture content of the shielding gas and possibly the filler wire employed or the surface cleanliness of the component surface being welded [4].

The heat from the GTAW welding arc will quickly vaporize any surface moisture ahead of the welding arc. The shielding gas covers the arc and effectively dissipates the vaporized moisture away from the arc before it can react and form monatomic hydrogen to mix with the molten weld pool. Therefore, any effect of surface moisture on the substrate being welded is minimized with the GTAW process and is considered an inherent characteristic of the process [5].

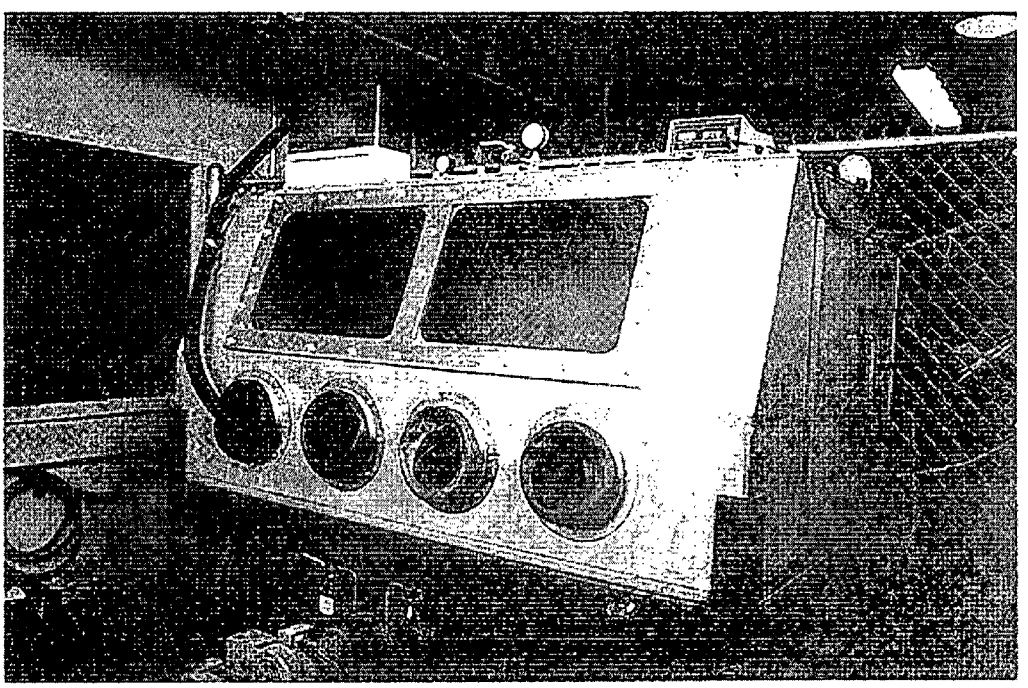

Figure 3-7
Purge Chamber Employed to Conduct Severe Environment Tests [4].

3.2.2 Diffusible Hydrogen levels At Various Shielding Gas Dewpoint

Temperatures

Diffusible hydrogen tests, included in the EPRI program [4], consisted of the injection of moisture directly into the shielding gas and then welding on SA508 Class 2 coupons. A broad range of dew point temperatures (temperature at which condensation of vapor begins upon cooling) were utilized, beginning with the bottle gas dew point (-70°F) and continued all the way up to that of a fog within the shielding gas (+50°F). The hydrogen levels were created by bubbling welding grade argon through a cylinder of water. A second cylinder and a needle valve were used to mix argon with the argon which had been bubbled through water to create the specific shielding gas dew points. The results indicated that high moisture content within the shielding gas was insufficient to develop even low levels of hydrogen and that hydrogen cracking should not be a concern provided some level of inert gas shielding is employed.

3.2.3 Off-the-Shelf Filler Wires

Modern techniques for producing quality GT AW filler wires have resulted in improved alloy compositional consistency and in lowering of diffusible hydrogen levels. The typical hydrogen content for the GTAW process is less than 1.0 ml/100g H_2. Table 3-2 provides measured hydrogen contents for several 80 series filler wires. Therefore the volumes of hydrogen available in the base material or the consumables are so low that they do not represent credible sources of hydrogen in sufficient quantities to support hydrogen delayed cracking.

Table 3-2
Measured Diffusible Hydrogen Contents for GTAW Filler Wires using a 150°F Preheat [4]

Filler Wire	Heat Number	Avg. Diffusible Hydrogen Content (ml/100 g H_2)
ER80S-B2	2301309	0.32
ER80S-B2	F5409	1.86
ER80S-G	9331612	0.73
ER80S-G	349866	0.45
ER80S-G	485936	0.52

The potential sources of hydrogen in welding would normally be found with flux shielding processes such as SMAW, SAW, and FCAW and not GTAW. With these processes the hydrogen source comes from high temperature reactions among hydrogen bearing compounds and from moisture adsorbed in the fluxes. Environmental moisture (humidity) is only considered a secondary source of hydrogen. As the weld pool cools, most of the monatomic hydrogen in the melt combines to form molecular hydrogen gas or react with other elements or compounds to form other low solubility gases that are vented to the atmosphere. It is noted that there are shielded metal electrodes that are produced today have extremely low hydrogen levels and in fact today's electrode producers formulate all coatings to have extremely low hydrogen. Finally, rules for handling coated electrodes and fluxes facilitate effective drying and handling of consumables.

3.2.4 Contamination from Lubricants, Coatings, Paints or other Sources

In another EPRI study, it was noted that hydrocarbons and moisture bearing contaminants are sources of hydrogen to welds. Typically, weld joints are suitably cleaned of such contaminants prior to welding. Normally, approved welding procedures control the joint cleanliness such that contaminants must be removed prior to welding. In addition, the heat of welding volatizes the hydrogen bearing contaminants that remain in front of the arc, and they are swept away from the weld area by convection with the cover gas [10].

The issue of contamination in the weld due to foreign media is a real issue, and will be addressed further in the discussion section of this report.

3.3 Tensile Stress and Temperature

Cold cracking or restraint cracking occurs under conditions of high geometrical restraint especially where a low toughness HAZ is present. Cold cracking generally occurs during cooling at temperatures approaching room temperature. As stresses build under a high degree of restraint, cracking may occur at defect locations according to fracture mechanics principals. High hardenability without tempering will promote cold restraint cracking [5].

The Reference 5 report notes that "untempered martensite and upper bainite microstructures are very hard and brittle structures. Carbon and other interstitial atoms are locked within the microstructure causing a tremendous degree of internal stress. This produces strength, but unfortunately, also produces low ductility and thus low toughness. When heat is applied to the structure in the range of 1250 to 1325 ^{0}F the carbon will begin to precipitate as carbides and the microstructure will toughen very rapidly. Tempering is both time and temperature dependent. Tempering is rapid at the upper temperatures, but drops off very rapidly as temperature decreases. Very little tempering occurs at traditional stress relief temperatures between 1050 and 1150 ^{0}F.

The temperbead process is designed to provide a sufficient heat inventory so as to produce the desired tempering for high toughness. For this reason, the high martensite microstructure is desired because of the improved toughness that will be developed. The toughness and ductility of the HAZ typically will be superior to the base material substrate because the quenching rate will be faster than the original product form. The machine GTAW temperbead process provides precision bead placement and control of heat. Therefore the resulting structure will be tempered and preheat is not desired."

The Reference 4 EPRI study examined the effect of the temperbead weld overlay and vee-groove repair of low alloy steels on the residual stresses and shrinkage associated with this weld repair. The program evaluated the stresses and shrinkage for these geometries using a 1-1/2Cr, ½ Mo (P-4) pipe. This material has increased hardenability over P-3 materials. The study evaluated several techniques to improve as-welded residual stresses. It was determined that bead placement along the toe area of the 2nd layer pass provided for some degree of reduction in residual stress. Residual stress at the weld fusion line was measured at approximately 65ksi, a reasonable value for the weld fusion line. Residual stress values drop rapidly to a value approximately zero at a position 0.35-inch (9mm) from the fusion line [4].

Residual stresses on the order of the yield strength of the substrate are to be expected with the temperbead process using conventional filler materials [4]. The key is to maintain the as- welded P-3 material HAZ in a crack tolerant, tough condition.

Another part of the Reference 4 study examined the peak temperature recorded for each deposited weld overlay layer beneath a weld block, monitoring the temperature at a given depth below the weld surface. Figure 3-8 presents a sketch of the three layer weld pad, the layout for the thermocouples is presented in Figure 3-9, and a temperature profile for the temperatures for each of the three layers is presented in Figure 3-10. Each of the transformation blocks consisted of a 3-inch thick section of SA 580 Class2 low alloy steel.

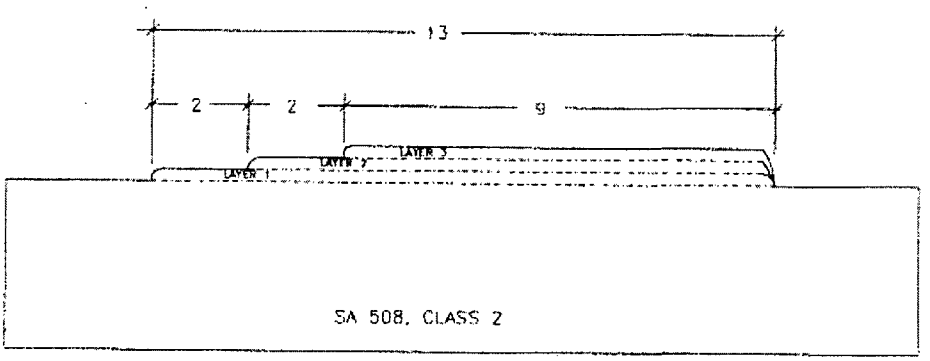

Figure 3-8
Transformation Block 1 Was Welded in Three Segments-One layer, Two Layers, Three Layers [4].

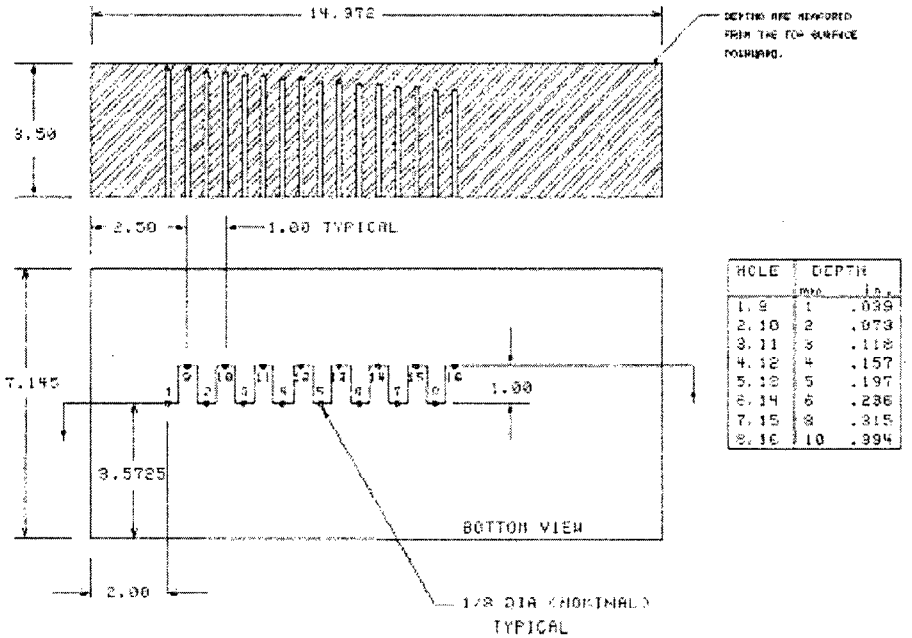

Figure 3-9
A Detained Drawing Providing the Thermocouple Locations for Each Transformation Block [4].

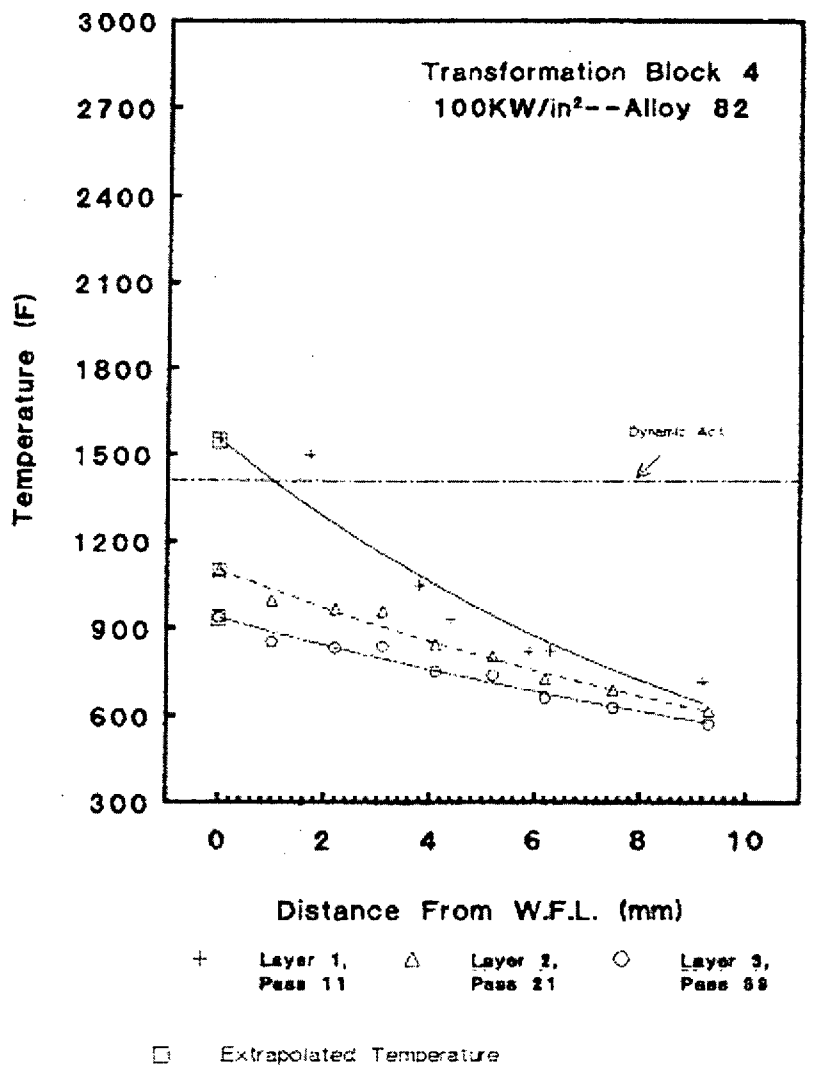

Figure 3-10
Graph Showing Peak Temperature Versus Distance From the Weld Fusion Line For Transformation Block 4 [4].

Transformation block 4, illustrated in Figure 3-10, is a relatively high heat weld input (100 KW/inch) pad application used to demonstrate the maximum temperatures observed beneath the welding torch for a conservative welding condition. Examining Figure 3-10 one observes that at a distance of 10 mm (0.25 inch) beneath the weld interface, the peak temperature observed during any pass is of the order of 800°F at any time during the welding. It can be further observed that at a distance of approximately 20-25 mm (0.5 to 0.625 inches) beneath the weld torch, the maximum temperature at any point on the nozzle surface during the WOL application is of the order of 600°F, or less. It is also noteworthy that very near the outside surface beneath the torch during the third temperbead layer, the maximum temperature observed anywhere was

of the order of 900°F. Extrapolating these results to the fourth layer, the first non-temperbead layer, it is expected that the peak temperature at any point on the surface should to be no greater than 750°F during the remainder of the WOL application.

These results suggest that after the completion of the temperbead weld layers, the OD surface of the P-3 nozzle would be expected to see relatively little in the way of severe temperature swings during each subsequent WOL layer, and consequently the shrinkage stresses are very small, reducing the likelihood of cold or restraint cracking during these non-temperbead passes.

3.4 Diffusivity and Solubility of Hydrogen in Steels

An important factor related to delayed hydrogen cracking in P-3 vessel materials is the diffusivity and solubility of hydrogen in low alloy steel as compared to that in austenitic stainless steel materials. It has been reported in the open literature that the solubility of hydrogen in steel is strongly dependent on crystal structure, temperature and composition. Hydrogen is much more soluble in austenitic materials than in ferritic materials [11]. Austenitic steels can hold up to five times more hydrogen than can carbon steels and about seven to eight times more than martensitic steels [12]. However, the diffusivity of hydrogen in austenitic steels is lower than that of ferritic materials thereby facilitating the removal of hydrogen by heating for ferritic materials [11].

Higher strength materials are also more susceptible to delayed hydrogen cracking than are cold worked materials. Cold worked or untempered alloy steels are more susceptible than quenched and tempered steels. Low alloy steels used for nuclear power plant vessels are quenched and tempered steels.

A discussion of the effect of the diffusion and solubility differences between low alloy steels and austenitic steels is presented in the discussion section of this report.

4
DISCUSSION OF TECHNICAL ISSUES RELATED TO THE 48 HOUR DELAY

In Section 3 of this report, the technical issues related to the 48 hour delay following the completion of welding and cooling to ambient temperature have been presented and some of the relevant research activities performed by EPRI and by the technical community has been cited. Key technical issues include the following:

4.1 Microstructure

The microstructure evaluations and hardness measurements discussed in this report and presented in the referenced reports [1, 4, 5, and 10] have described the effects of temperbead welding on the toughness and hardness of P-3 materials. The research activities highlighted in this report have illustrated that the microstructure in the low alloy steel (P-3) beneath the temperbead WOL in the weld HAZ consists of a structure that is tempered martensite or tempered bainite and has maximum hardness at a distance of 2 to 3 mm (80 to 120 mils) beneath the surface of the order of 280 to 300 KHN (28 to 30Rc) or lower, as illustrated in Figure 3-2 for a single layer overlay. This hardness level is far below what is believed to be the threshold for hydrogen cracking in these low alloy steels in aqueous environments, a level believed to be Rc 36 or greater. In addition, the maximum temperature observed beneath the weld overlay in the weld HAZ is 900°F at the completion of the third temperbead weld layer (Figure 3-10), well below the critical transformation temperature. These observations indicate that additional welding of the WOL will not cause a retransformation of the structure now present in the P-3 material. Consequently, the HAZ beneath the WOL in the P-3 material is not a concern for hydrogen embrittlement due to the tough microstructure and the low hardness.

However, the toe of the WOL in the P-3 material produced hardness readings approaching Rc 36 or slightly above in some of the temperbead WOLs at the OD surface of the nozzle, as presented in Section 3 and in Reference 4. In addition, the Section 3 investigation revealed that this region, at the toe of the weld overlay, was at or slightly above Rc 36 and occurred at a distance of about 1 mm (40 mils) from the weld fusion line at the OD surface, and extended to a depth into the P-3 material that was typically of the order of 1 or 2 mm (40 to 80 mils). This high hardness zone is extremely small, and would have no significant effect on the structural integrity of these components.

4.2 Sources of Hydrogen

The sources of hydrogen have been identified in Section 3.2 as resulting from moisture, poor grade shielding gas mixtures, drawing compounds, contaminated base metals, and lubricants, coatings or paints inadvertently in contact with the weld wire. The issues of moisture, poor shielding gas mixtures and various weld wires were discussed in Section 2 of this report. The

results demonstrate that introducing hydrogen is unlikely with the GTAW process and hydrogen cracking is extremely unlikely.

Additional welding issues may be present related to contamination by lubricants, coatings or paints. While these contaminants cannot be ruled out, if present, they would act on the weld and welding torch such that either the weld would be terminated to resolve the issue, or the post weld NDE of the weld metal itself would identify defects in the weld metal. This is really a quality control problem that is generally resolvable through the Owner's Quality Program. Good practice should eliminate this problem for the temperbead WOLs.

4.3 Tensile Stress and Temperature

Tensile stress and temperature were described in Section 3.3 of this report. Whereas these cyclic tensile stresses do occur during fabrication of the temperbead WOL, the effect of these stresses diminishes with each weld layer. As noted in Section 3.3, after the three layer temperbead layers are applied, the maximum temperature observed at the toe of the weld overlay, where the hardness may be elevated sufficiently so that cold cracking is possible, is no greater than 800°F, well below the austenitizing temperature so no phase change will occur and the subsequent stresses will be small. Experience has shown that thermal stresses produced during these small temperature cycles do not cause cold cracking at the toe of the WOL. Following the final layer, it is expected that the maximum surface temperature at the toe of the WOL in the P-3 HAZ would only be of the order of 400°F to 500°F. Slow cooling to ambient temperature from these temperatures would produce insignificantly small stresses.

4.4 Diffusivity and Solubility of Hydrogen in Steels

Section 3.4 of this report presented a brief discussion of the diffusion and solubility of hydrogen in ferritic, martensitic, and austenitic steels. It was noted that the diffusion rate for hydrogen was greater in ferritic materials than in austenitic materials, but that the solubility of hydrogen in austenite was from five to seven times greater in the austenite than in the ferrite or martensite. Consequently, due to the temperatures expected during the welding of the temperbead layers, and during the welding of the non-temperbead WOL layers, the temperature should be sufficient for the hydrogen to diffuse out of the material, either escaping the structure or diffusing into the austenite, where it can be held in much greater quantities. The diffusion rate is clearly from the ferrite or martensite to the austenite and whatever hydrogen remains will reside in the austenite, which has a much lower propensity to hydrogen related cracking.

5
CONCLUSION

The temperbead WOL has become an increasingly effective tool for performing repairs on dissimilar metal welds involving low alloy steel reactor pressure vessel nozzle (P-3) materials. The elimination of preheat and post heat bake, as allowed by ASME Code Cases N-638 and N740, has extended the usefulness of this process providing an increasingly cost effective and personal exposure effective process.

Whereas the Code Cases N-638 and N-740 provisions allow for ambient temperature welding, and no post weld bake, the post weld 48 hour hold at ambient temperature has remained as a provision for such temperbead welding. This 48 hour delay between welding completion and cooling to ambient temperature and the final non-destructive examination (NDE) of the fully welded component is provided so that any delayed hydrogen cracking or cold cracking that may occur can take place prior to the final NDE of the weld overlay and surrounding area. The post weld 48 hour delay following cooling to ambient temperature has resulted in a substantial cost burden to the involved utilities. As there are significant economic advantages associated with starting the 48 hour clock immediately following the final temperbead layer of the WOL, it is important to determine the technical advantages and disadvantages of making such a change.

This White Paper provides information developed by EPRI and other organizations addressing the issues associated with allowing the 48 hour delay to begin following completion of the third temperbead weld layer as applied to the P-3 low alloy steel pressure vessel component as an alternative to the current provision in Code Case N-638 and N-740. This document addresses the industry and regulatory concerns associated with this change, listed below:

-Microstructure
-Sources for Hydrogen Introduction
-Tensile Stress and Temperature
-Diffusivity and Solubility of Hydrogen in Steels

The technical data regarding these issues was examined and the results were presented as described below.

- The microstructure in the P-3 material directly beneath the temperbead WOL consists of a tempered martensite or tempered upper bainite that has excellent toughness, combined with a modest maximum hardness (of the order of Rc 30 or lower).

- The microstructure at the toe of the temperbead WOL in the P-3 weld HAZ at the OD surface where tempering is somewhat limited may have a very small HAZ with a maximum hardness of the order of Rc 36, at a distance of approximately 40 mils from the toe of the WOL. At a depth of approximately 2 to 2.5 mm (80 to 100mils) beneath the

toe of the WOL, the hardness is reduced to less than 294 KHN (29Rc), a hardness level well below that required to cause hydrogen cracking.

- Sources of hydrogen include moisture, poor shielding gas and contamination. It is noted that moisture in the shielding gas or high humidity is not a problem for GTAW temperbead welding. Contamination will affect the weld, and should be identified either during the welding process or during the subsequent NDE of the overlay. Good welding practice should eliminate this problem for the temperbead WOL.

- Tensile stresses should not be an issue for cold cracking as the thermal stresses diminish with each weld overlay layer. Following the final layer, it is expected that the maximum surface temperature at the toe of the WOL in the P-3 HAZ would reach temperatures only on the order of 400°F to 500°F. Slow cooling to ambient temperature from these temperatures would be expected to produce relatively small stresses.

- The diffusion rate for hydrogen is greater in ferritic materials than in austenitic materials, but the solubility of hydrogen in austenite is from five to seven times greater in the austenite than in the ferrite or martensite. Consequently, due to the temperatures expected during the welding of the temperbead layers, and during the welding of the non-temperbead WOL layers, the temperature should be sufficient for the hydrogen to diffuse out of the material, either escaping the structure or diffusing into the austenite, where it can be held in much greater quantities. Thus, even if hydrogen is produced, a large hydrogen inventory in the P-3 material is not expected.

In summary, this study has identified a small finite region at the toe of the temperbead WOL in the P-3 material on the OD surface where it is possible for the hardness to reach a lower threshold level for hydrogen cracking in this component. Even in the very unlikely case that cracking was to occur, it would not be structurally significant.

There appears to be no technical basis for waiting the 48 hours after cooling to ambient temperature before beginning the NDE of the completed weld. There should be no hydrogen present, and even if it were present, the temperbead welded component should be very tolerant of the moisture.

Finally, as noted in this report, there have been more than 20 temperbead WOLs applied to nuclear power plant low alloy steel nozzles and more than 100 temperbead repairs to other low alloy steel components. The initial overlays have been in service more than 20 years. Of all the temperbead overlays applied in the field, there has been no indication of hydrogen cracking found from NDE performed after the 48-hour hold or from subsequent ISI.

6
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