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Initial Evaluation of the Heat-Affected Zone, Local Embrittlement Phenomenon as It Applies to Nuclear Reactor Vessels

D. E. McCabe

ABSTRACT

The objective of this project was to determine if the local brittle zone (LBZ) problem, encountered in the testing of the heat-affected zone (HAZ) part of welds in offshore platform construction, can also be found in reactor pressure vessel (RPV) welds. Both structures have multipass welds and grain coarsening along the fusion line. Literature was obtained that described the metallurgical evidence and the type of research work performed on offshore structure welds. It was determined that other than having multipass welding and coarse prior-austenite grains along the fusion line, there was little else upon which to connect RPV weldments to the LBZ problem in offshore structures. However, coarse prior-austenite grains make RPV steels susceptible to temper embrittlement, an entirely different embrittlement mechanism. Experiments were undertaken to evaluate the potential for temper embrittlement problems in RPV HAZ materials. Enlarged prior-austenite grains were made in bulk using laboratory furnace treatments and in weld cycle simulations using the Gleeble. Five commercially made RPV steels were used in the present study. Embrittlement aging was performed at various temperatures and times. RPV steels with large prior-austenite grain sizes were found to suffer significant temper embrittlement with thermal aging. Because of high carbon equivalent and routine postweld heat treatment that was standard manufacturing practice for RPV production, this project could not make a case that an LBZ type of problem exists in reactor vessels. A final phase of this investigation is to examine the effect of simultaneous irradiation and thermal aging at 288°C (550°F), which can create a critical damage situation. The irradiation exposure is currently in progress.

BACKGROUND

In 1989, Combustion Engineering proposed to the U.S. Nuclear Regulatory Commission (NRC) that consideration be given to replacing the fracture-toughness-based dynamic K_{IR} curve with the quasi-static K_{Ic} curve for setting allowable temperature/pressure limits during normal and upset conditions.* The idea was accepted for consideration pending the development of clear evidence that there can be no unanticipated trigger sources that would introduce running cleavage cracks.

Such trigger sources could be either crack pop-ins or possibly local brittle zones (LBZs) that reside along the fusion line of multipass weldments. The existence of LBZs was suggested as one of the possibilities for reactor pressure vessel (RPV) weldments.

The objectives of the present project were to (1) review the available LBZ information, (2) determine its applicability to nuclear vessel fabrication and service conditions, and (3) conduct experiments in support of the evaluations.

REVIEW OF EXISTING LBZ INFORMATION

LBZs in multipass weldments have so far been identified only in offshore platform weldments.^{1,2} At first, offshore platforms were constructed with carbon/manganese grades of steel. Multipass welds in these materials require postweld heat treatment (PWHT) to obtain satisfactory heat-affected zone (HAZ) fracture-toughness properties. Later, in the early 1980s, the preferred plate material shifted to low-carbon microalloyed steels.³ The advantages were clear. Microalloyed steels are given controlled hot-rolling practices to achieve outstanding fracture-toughness properties while retaining good strength properties. At temperatures as low as -40°C (-40°F), these materials are on the upper shelf of fracture toughness as measured by the Charpy V-notch (CVN) tests, and the energy is typically 260 to 280 J (190 to 205 ft-lb).⁴ Additionally, the low-carbon content, usually below 0.1 wt %, promotes excellent weldability. The greatest

*R. K. Nanstad, Travel to Rockville, Maryland, for meeting at the U.S. Nuclear Regulatory Commission, January 9, 1990.

advantage is that no PWHT is required, resulting in highly significant construction cost savings.

The petroleum industry performs fracture-toughness determinations by a crack tip opening displacement (CTOD) test method.⁵ This is a pseudo-fracture-mechanics test method in which a linear displacement measurement at the tip of a fatigue precrack is assumed to have a linear relationship with the fracture-mechanics-based J-integral. A fatigue-precracked bend bar specimen is used. For LBZ evaluations, the precrack is meticulously positioned to be in a grain-coarsened region along the weld fusion line. In replicate tests, some specimens suffer only ductile deformation and consequent slow-stable crack growth.

Two petroleum industry organizations that implement this test procedure in their standards are the American Petroleum Institute⁶ and the Engineering Equipment and Materials Users Association.⁷ The standards are used to qualify a steel manufacturer as a certified producer of LBZ-resistant microalloyed steel. All valid LBZ specimens according to these methods must have a minimum of 15% coarse grain material along the crack tip. Coarse grain size is generally indicated by American Society for Testing and Materials (ASTM) to be 4.5 or numerically lower. Not all coarse grain material is embrittled material so that proof of a material's susceptibility to LBZ failures in HAZs is never absolute. Weld joint preparation is either "K" type or "half-K" type, with the intention of making the straightest possible HAZ for test specimen positioning.⁸

Research projects designed to evaluate LBZ damage mechanisms generally use electric resistance heating of specimen blanks to simulate material that has been through a weld pass cycle.^{9,10,11} Brittle-zone microstructures are created in sufficient bulk to develop reliable Charpy transition curves that represent the various microstructures identified within the LBZ region.

The thermal history and consequent morphology of HAZ welds are extremely complex. Each individual weld pass produces a four-zone microstructural gradient depicted in Fig 1. The coarse-grain HAZ (CGHAZ), schematic (D), immediately adjacent to the

fusion line, consists of enlarged prior-austenite grains from the subsequent weld pass that grow rapidly due to being at a temperature close to the melting point. Upon cooling, the prior austenite can transform to martensite or a lower bainite/martensite mix, depending on the carbon equivalent (CE) of the base metal. At slow cooling rates, low-CE steels will transform to a lower-toughness microstructure, pearlite and/or upper bainite. With low CE, the cooling rate must be quite rapid to form martensite and lower bainite. Lower bainite and/or martensitic steel is more brittle as cooled, but these microstructures are converted to high-toughness steel after PWHT. The second zone, fine-grain HAZ (FGHAZ), schematic (C), from the fusion line in Fig. 1, is fine-grain prior austenite resulting from slower austenitic grain growth at lower temperatures. The third zone, intercritical, coarse-grain HAZ (ICCGHAZ), schematic (B), is material heated into the A_{c3} and A_{c1} (intercritical) transformation zone. A mixture of ferrite and austenite has precipitated along prior-austenite grain boundaries, which embrittles the material in this local region upon cooling to room temperature.^{9,10} These small volumes of material, ICCGHAZ, have been identified as the LBZ trigger-point locations. It is understood that LBZ material is created by the reheating of coarse-grain prior austenite from the subsequent weld passes that are adjacent to the fusion line. It therefore follows that there can be an appreciable amount of coarse-grain prior austenite along a fusion line that is not brittle crack initiation material. The fourth zone is the subcritically reheated coarse-grain zone of the (CGHAZ), which undergoes very little changed property.

The fundamental laboratory studies on LBZs usually produce ICCGHAZ material in bulk using electrical-resistance heating in a weld-cycle-simulation machine such as a Gleeble. It should be noted that some of the data sources reviewed had unintentionally produced damaged material, giving misleading results. It has been perhaps accurately noted that the temperature adjacent to the fusion line can approach 1590°C during a weld pass. The assumption that prior-austenite microstructures can be developed in simulation using such a temperature can too easily result in what is known as “burned steel.”⁴ Steel can be burned when heating and holding at a temperature above 1260°C (2300°F). Burned steel is irrecoverably-damaged steel, usable only as remelt stock. However, data reported from destroyed steel occasionally finds its way into the literature. An interesting outcome of the fundamental studies on microalloyed steels has

Figure 1. Schematic representation of microstructures local to the HAZ of multipass welds.

been the demonstration that ICCGHAZ microstructures can cause about a 15 to 20°C transition-temperature increase over that of unsensitized material.¹⁰ When these same materials undergo PWHT at 450°C (842°F), the fracture toughness of ICCGHAZ material is restored, and in some cases is slightly improved over that of the unsensitized condition.

Assessment of Commonality

The paradoxical feature of the LBZ problem is that brittle cleavage fractures have only been encountered in CTOD tests made with three-point bend specimens. There have been no reports of offshore platforms suffering failures or damage resulting from LBZs.¹ Hence, the extensive research work and material-qualification testing reported in the literature has provided answers that are in search of a problem. Nevertheless, cleavage trigger sources are considered dangerous when they occur in strain-rate-sensitive steels, and all failure-critical components deserve serious consideration.

Both offshore-platform welds and RPV weldments are multipass. Welds in offshore structures are principally shielded metal-arc welds (SMAWs),³ while RPV weldments are principally submerged arc welds (SAWs). The heat input is greater in the SAW process that can promote the development of coarse prior-austenite grains.

Transformation properties differ significantly between microalloyed steels and RPV steels. The CE is important from the standpoint of the transformation products formed on cooldown. Several important alloying elements contribute to CE, as indicated from the following equation:¹²

$$CE = C + \frac{Mn}{6} + \frac{Cu + Ni}{15} + \frac{Mo + V + Cr}{5} \quad (1)$$

The CE controls austenite stability and hence promotes transformation to low-temperature transformation constituents, such as lower bainite and martensite.

The typical CE in low-carbon microalloyed steel is about 0.37; the typical RPV steel has CE of about 0.62. Hence, the cooling rate after each weld pass is more critical in the low-CE steels. Shielded metal-arc welding is performed in offshore structures because of the lower heat input and consequent faster cooling rate. Increased CE generally decreases weldability due to increased propensity for cracking. The transformation products tend to become brittle as cooled. However, postweld tempering converts the brittle microstructures into a steel with superior fracture-toughness properties. Tempering also improves fracture toughness of microalloyed steels, but this obvious remedy would eliminate much of the economic gain from their use in an as-welded condition.

Another important factor is service condition. Offshore platforms are expected to function at ambient temperatures down to -40°C . These low service temperatures do not alter the basic metallurgical condition of the steel. LBZ tests have produced crack initiation at test temperatures typically between -10 and -60°C . Had the CTOD testing been performed at room temperature, this very likely would have been within the upper-shelf temperature range for the embrittled LBZ materials and LBZ failures might not have been encountered. Cleavage crack initiation would have been replaced with ductile tearing. On the other hand, RPV steels operate between room temperature and 288°C (550°F). Hence, a case has not been clearly made for the classical LBZ embrittlement model of microalloyed steels to apply to the RPV applications where service temperatures are not low. However, if end-of-life annealing were used, the service temperature cycle could include a 168-h excursion to 482°C (900°F) for a one-time cycle. Therefore, the concern for grain coarsening in RPV welds arises from the potential for metallurgical degradation by temper embrittlement. Only the HAZs are at issue because RPV steel plate and forgings usually have fine prior-austenite grains, and as a consequence, the base plate materials are not considered to be vulnerable to temper embrittlement.

LBZ in RPV Steels

The temper-embrittlement susceptibility of grain-coarsened RPV steels was clearly demonstrated in work at AEA-Technology, Harwell, United Kingdom.¹³ Eleven laboratory heats of steel with typical RPV chemistry were made, and three of these with varied

copper and phosphorus contents were selected for temper-embrittlement studies. The sensitivity to embrittlement was enhanced by heat treating to obtain huge prior-austenite grain size (about 0 to 1, ASTM Standard E 112). The austenitizing temperature was 1200°C (2200°F) with a 30-min soak followed by an oil quench. All were given PWHT at 615°C (1140°F) with a 24-h soak, again followed by an oil quench. In contrast, coarse grains in offshore-structure LBZ studies were about 4 to 5 ASTM size. The AEA temper-embrittlement aging was 450°C (840°F) for 2000 h to be fully aged in order to maximize the degree of embrittlement obtainable. A second aging cycle was 475°C (890°F) to simulate an end-of-life recovery anneal cycle. One other exposure was irradiation to 12.2 mdpa damage, presumably at 288°C (550°F). Table 1 is a condensed representation of the AEA results. It is important to keep in mind that ΔT_o from temper embrittlement is extremely grain size dependent, and with 0 to 1 grains, these results in Table 1 do not truly represent the coarse grain situation in RPV welds.

ORNL Evaluation Followup

Five commercially produced RPV steels were selected for the Heavy-Section Steel Irradiation (HSSI) evaluation program. Selection was based on the reported chemical elements shown in Table 2. An objective was to include at least one material with a

Table 1. Midtransition temperatures and shifts measured by AEA with CVN specimens

Chemistry variable (wt %)		Aging (h/°C)	T _o (°C)		ΔT_o (°C)
Cu	P		Initial	Aged	
0.01	0.007	2000/450	−86	32	118
0.01	0.017	2000/450	−55	119	174
0.16	0.017	2000/450	−95	114	209
0.16	0.017	^a	−95	−26	69

^aIrradiated 12.5 mdpa.

Table 2. Commercial materials selected

Material	Code	Content (wt %)			
		P	Ni	S	Cu
A 302 grade B	Maine Yankee	0.015	0.2	0.017	0.14
A 508 class 2	Midland	0.010	0.76	0.015	0.02
Modified A 302 grade B	GE (Z5)	0.016	0.62	0.015	0.17
Modified A 302 grade B	GE (Z7)	0.007	0.53	0.014	0.16
A 533 grade B	HSST Plate 01	0.018	0.75	0.013	0.18

phosphorus level as high as the material used in the AEA-Technology experiment (Table 1). High copper, by the AEA definition, is almost always present in the commercial steels. The high phosphorus, on the other hand, was not easily obtained (see Fig. 2).¹⁴

Confirming chemical analyses and tensile properties were obtained during the course of the present investigation, and these properties are reported along with the as-received values in Table 3. The initially reported chemistries were not entirely reliable, as can be

Figure 2. Cumulative number of chemical samplings made on RPV base metals vs the phosphorus content.

Table 3. Summary of chemical analyses

Material	Condition ^a	Composition (wt %)												
		C	Mn	P	S	Si	Ni	Cr	Mo	V	Nb	Cu	As	Sn
A 302 B	AR	0.21	1.20	0.015	0.017	0.28	0.20	0.24	0.60	0.004	0.006	0.14	0.015	0.012
	AEA	0.21	1.15	0.014	0.016	0.31	0.21	0.22	0.57	0.005	0.005	0.14	0.018	0.011
	AS	0.22	1.12	0.014	0.015	0.31	0.18	0.21	0.56	0.005	0.003	0.15	0.17	0.010
A 508 class 2	AR	0.20	0.63	0.010	0.015	0.19	0.76	0.37	0.60	<0.01	—	0.02	<0.01	<0.01
	AEA	0.22	1.04	0.013	0.010	0.28	0.73	0.21	0.57	0.002	0.004	0.05	0.015	0.006
	AS	0.22	0.66	0.013	0.009	0.27	0.71	0.34	0.58	0.002	0.003	0.05	0.015	0.006
Mod A 302 B	AR	0.16	1.27	0.014	0.015	0.17	0.62	0.10	0.50	0.002	0.004	0.17	0.013	0.017
	AEA	0.20	1.12	0.015	0.018	0.19	0.61	0.09	0.47	0.003	0.005	0.18	0.018	0.016
	AS	0.19	1.10	0.016	0.017	0.19	0.54	0.09	0.46	0.003	0.003	0.19	0.017	0.015
Mod A 302 B	AR	0.26	1.47	0.007	0.014	0.15	0.53	0.09	0.52	0.002	0.003	0.16	0.014	0.018
	AEA	0.25	1.42	0.010	0.013	0.18	0.55	0.09	0.48	0.002	0.004	0.17	0.021	0.015
	AS	0.27	1.42	0.011	0.016	0.18	0.49	0.09	0.50	0.002	0.004	0.19	0.021	0.015
A 533 B	AR	0.24	1.34	0.018	0.013	0.23	0.75	—	0.53	—	—	0.18	—	—
	AEA	0.23	1.15	0.010	0.012	0.28	0.74	0.10	0.55	0.003	0.004	0.12	0.013	0.010
	AS	0.22	1.49	0.010	0.012	0.27	0.65	0.10	0.55	0.003	0.004	0.20	0.020	0.015

^aAR = As reported; AEA = AEA + 2000 h at 450°C; AS = as-received and aged.

As-received tensile properties

Material	Yield strength		Tensile strength	
	(MPa)	(ksi)	(MPa)	(ksi)
A 302 B	538	78.1	689	100.1
A 508 class 2	394	57.1	569	82.6
A 533 grade B	470	68.2	620	89.9
Mod A 302 B Z5	400	58.1	538	78.1
Mod A 302 B Z7	472	68.5	632	91.7

seen. One material, A 302 grade B, received in the as-normalized condition, was not available in the heat-treated condition desired. Consequently, the normalized microstructure was converted into a quenched-and-tempered product using the heat-treatment cycles summarized in Table 4.

The cycles in Table 4 simulated the heat treatment reported on the low-phosphorus modified A 302 grade B steel listed in Table 2. This particular modified A 302 grade B(Z7) came from a commercial reactor nozzle cutout. The end result was suitable from a mechanical-properties standpoint, but it was later determined that the retreated A 302 grade B quenched-and-tempered microstructure had a prior-austenite grain size of ASTM 5, which was large in comparison to the other commercially made steels (Table 5).

Table 4. Heat-treatment cycles

Treatment	Temperature (°F)	Time (h)	Quench
Austenitize	1700	6.5	Water
Austenitize	1625	6.5	Water
Temper	1200	6.5	Water
Temper	1125	6.5	Water

Table 5. Grain sizes of the commercially made RPV base metal in their as-received condition

Material	Size by E 112	Average grain diameter (μm)
A 302 grade B ^a	5	65
A 508 class 2	8	22
A 533 grade B	8.5	19
Mod A 302 grade B Z5	10	11
Mod A 302 grade B Z7	9	16
^a Reprocessed from normalized to quenched and tempered.		

Program Plan

The general plan to evaluate LBZ potential in RPV welds was structured to be in three phases, starting with the AEA-Technology large grain concept with the difference that commercially produced RPV steels were used. All three phases are defined as follows.

Phase I—Repeat the AEA austenitization and temper-embrittlement aging protocol (namely, temper-embrittlement cycles), using commercial steels.

1. Age 2000 h at 450°C (840°F).
2. Age 168 h at 490°C (914°F).

Phase II—Use a simulated SAW cycle to develop prior-austenite grains of ASTM 4 to 5. Then apply the AEA-Technology aging protocol.

1. Age 2000 h at 450°C (840°F).
2. Age 168 h at 450°C (840°F).

Phase III—Expand the aging-cycle evaluations starting with the simulated weld cycle austenitization used in Phase II.

1. 168 h at 482°C (900°F).
2. 168 h at 399°C (750°F).
3. Irradiate to 1×10^{19} n/cm² (>1 MeV at 288°C).
4. Irradiate to 1×10^{19} n/cm², then age 168 h at 450°C.
5. Age in air at 288°C (550°F).

Not mentioned in the above plan are the additions of baseline CVN data generated to provide the unaged baseline information needed to quantify transition-temperature shifts and to supplement the metallurgical evaluations. Temper-embrittlement damage is evaluated by standard CVN specimen transition-curve shift.

Results

Phase I—All of the commercially produced steels that were given the AEA-Technology austenitic cycle had microstructures ranging from tempered martensite to lower bainite. Grain size according to ASTM Standard E112 was generally between 0 and 00, meaning that the typical austenite grain diameter was on the order of 360 to 500 μm (Fig. 3). The microstructure after PWHT is represented by Fig. 4.

Transition-temperature CVN energy, E_J , data were least-squares fitted to the following hyperbolic tangent equation:

$$E_J = A + B \tanh [(T - T_o)/C] , \quad (2)$$

where

$$A = (USE + LSE)/2,$$

$$B = (USE - LSE)/2,$$

$$C = \text{width of the transition},$$

$$T_o = \text{midtransition temperature}.$$

Parameter T_o was selected to be the Charpy transition-curve temperature level of reference. T_o is a point halfway between the lower-shelf energy (LSE) and upper-shelf Charpy energy (USE). This point was selected because it tends to be a position that is least affected by changes in curve shape. The LSE used in fitting data to Eq. (2) was held constant at 2.7 J.

Table 6 lists the T_o temperatures of the five commercial RPV steels in the as-received condition, both before and after embrittlement aging. These results can be compared with the embrittlement from AEA-Technology austenitizing cycles and aging, Table 1. As-received RPV steels with small grains on the order of ASTM 8 to 11 showed insignificant embrittlement. The A 302 grade B exception was caused by larger grains, on the order of 5.

Figure 3. Example of the as oil-quenched microstructure after the AEA-Technology austenitization and quench cycle, A 533 grade B, 100x.

Figure 4. Example of the typical microstructure after the AEA-Technology austenitization and PWHT, A 533 grade B, 100x.

**Table 6. Revisit of AEA cycles used on five commercially made RPV steels;
Phase I transition temperatures, T_o (°C)**

Condition	Aging time (h)	Aging temperature (°C)	A 302 B	A 533 B	A 508 class 2	Modified A 302 B	
						High phosphorous	Low phosphorous
As received	Initial	—	0	20	0	−9	−9
As received	2000	450	64	38	10	−4	10
AEA austenitize ^a and PWHT ^b	Initial	—	10	−67	4	−23	−73
AEA austenitize ^a and PWHT ^b	2000	450	45	78	41	34	42
AEA austenitize ^a and PWHT ^b	168	490	44	65	16	16	−3

^a30 min at 1200°C; oil quench.
^b24 h at 615°C; oil quench.

Focusing on the AEA austenization cycle data for 2000 h age at 450°C, it appears that steels with the best initial fracture toughness (lowest T_0 temperatures) tend to have the greatest transition-temperature shift. Figure 5 illustrates this point. If the AEA-Technology data from Table 1 are included, the extent to which copper and phosphorus seemed to control ΔT_0 appears to be second-order when the results are viewed within the effect shown in Fig. 5.

The simulation of aging for 168 h at 490°C was almost as effective as the 2000 h at 450°C in some cases. This indicates that the aging rate is asymptotic. However, this should not be regarded as a firm observation because ΔT_0 rate was not the subject of this work.

Figure 5. Relationship between initial T_0 temperature and ΔT_0 after 2000-h aging at 450°C.

In summary, the results from the Phase I repeat of the AEA-Technology experiments using commercially made steels in most cases did not quite measure up to the ΔT damage level of the AEA-Technology experiments. The conversion to large prior-austenite grains by the AEA-Technology austenitizing treatment by itself did not always cause significant change in fracture toughness from the as-received fracture toughness. For example, comparing rows 1 and 3 in Table 6, there were small increases in T_o for two materials, while the other three changes in T_o were negative, with two of these being substantial. The ΔT_o shift from the subsequent temper-embrittlement aging seems to be strongly influenced by the initial post-AEA austenitized T_o values as shown in Fig. 5. The grain-size dependence of ΔT_o on prior-austenite grain size was clearly demonstrated in four of the five materials tested and weakly demonstrated in the fifth. However, the grain sizes that resulted from the AEA austenitizing treatment are considerably larger than those expected in actual ICCGHAZ regions, suggesting that these Phase I results do not truly represent the coarse-grain situation that we wished to evaluate in commercial RPV welds.

Phases II and III—The electrical-resistance heating and cooling equipment (Gleeble) that is designed specifically for simulation of weld cycles was applied to austenization of the same five materials tested in Phase I. Because the intercritical reheating cycle, identified as the source of LBZ behavior in low-carbon microalloyed steels, was determined to be of no relevance to RPV steels, only a single simulation cycle was needed in the present experiment to simply create the simulated coarse prior-austenite grains in RPV HAZs. A report¹⁵ based on research funded by the NRC was used to decide upon the most appropriate thermal cycle. Peak temperature and time were selected to develop ASTM grain size in the range of 4 to 5. The peak temperature was limited to 1260°C (2300°F), specifically to be below the burning temperature for steels of typical RPV chemistry. The thermal cycle used is shown in Fig. 6. Specimen blanks prepared for the Gleeble were cylindrical rods about 76 mm long and 14.35 mm in diameter (3 by 0.564 in.). About 5 mm (0.2 in.) of material at the center of each rod was determined to be representative of the weld cycle material. An example photomicrograph of the resulting microstructure appears in Fig. 7, showing a mixture of martensite and lower bainite.

Figure 6. Gleeble-generated weld thermal cycle to develop ASTM 4.5 grain size.

Figure 7. Photomicrograph of microstructure after Gleeble austenitization, high-phosphorus modified A 302 grade B, nital etch, 100×.

Table 7 lists the Gleeble weld simulation and aging data developed to date: Phase II, on lines 1, 3, and 5, and part of Phase III on lines 2 and 4. The PWHT cycle used was identical to that used in the AEA-Technology simulation, namely, a 24-h soak time at 615°C (1140°F) followed by an oil quench. Note that the as-tempered fracture toughness reflects a favorable microstructural condition that in all probability resulted from the fast cooling of the weld cycle simulation. Aging time at 168 h represents the expected hold time that could be used in an RPV annealing procedure. There are three aging temperatures that represent the lowest to highest practical annealing temperatures that have been under consideration. At 399°C, there appears to be no evidence of temper embrittlement in four materials. This could also be true for the high-phosphorus modified A 302 grade B steel (Z5) that for a lack of sufficient material was not included in this aging exposure. Embrittlement develops and accelerates with temperatures between 450 and 482°C (842 and 900°F). Hence, these two annealing conditions affect temper embrittlement damage, provided there is coarse-grain HAZ material. However, Fig. 8 shows that none of the transition-temperature end points, after annealing for 168 h, exceeded room temperature. After 2000 h at 450°C, only two of the five materials had T_0 temperatures that exceeded room temperature.

Table 7. Materials austenitized by Gleeble, PWHT, and aged commercially made RPV steels (Phase II and first part of Phase III combined)

Austenitize method	Aging temperature (°C)	Aging time (h)	Transition temperature, T_0 (°C)				
			A 302 B	A 533 B	A 508 class 2	Modified A 302 B	
						High P ^a	Low P
Gleeble	Initial	0	-87	-67.0	-102	—	-67
Gleeble	399	168	—	-67.1	-98.6	-89.4	-71.1
Gleeble	450	168	-44	-28	-80	—	-47
Gleeble	482	168	—	7.4	-72.6	-47.4	-33.6
Gleeble	450	2000	43	66	-46	-14	-4

^aP = phosphorous.

Figure 8. Position of CVN T₀ temperatures after aging 168 h at 399°C, 450°C, and 482°C and 2000 h at 450°C (MA302B stands for modified A 302 grade B).

Six commercial welds were examined for maximum grain size along the fusion line (Table 8). The results of this small sampling of actual RPV welds suggested that the presumed coarse-grain size and uniform distribution characteristics do not always exist. The coarse grains that could be found were much smaller than ASTM 4.5, and the percentage of coarse grain along all fusion lines was found to be very low. Example photomicrographs taken at selected large-grain locations are shown in the Appendix.

Table 8. Largest coarse-grain sizes found in HAZs of six commercially fabricated RPV welds

Weld source	ASTM E 112 grain size
HSSI Fifth Irradiation Series, 72W	6.5
HSSI Fifth Irradiation Series, 73W	9.0
SNUPPS beltline	7.5
SNUPPS longitudinal	7.5
Midland beltline	10
Midland nozzle course	6.5

Measurement of Element Diffusion to Grain Boundaries

Measurement of element diffusion to grain boundaries was conducted as part of a collaboration with AEA-Technology. The segregation of critical elements to grain boundaries on four selected specimens was measured by scanning transmission electron microscopy. These results were then compared with theoretical calculations using the McLean model as well as the Murphy and Perks model.¹⁶ Experimental results are shown in Table 9. All specimens had been Gleeble austenitized to ASTM grain size of 4 to 5, followed by PWHT at 615°C (1140°F) for 24 h. The grain-boundary analyses also included Cu, Si, and Cr; none of which were found to be involved in the segregation process. Measurement replication was four times, and in some cases the variability was huge. Phosphorous segregation to grain boundaries was also calculated. The McLean

Table 9. Measured increased contents, percent increase of selected elements; grain boundary vs bulk contents

Material	Treatments	Percent increase at boundaries (at %)			
		Mn	Mo	P	Ni
Modified A 302 B (low phosphorous)	Gleeble austenitized and PWHT (unaged)	9	12	1	1
Modified A 302 B (low phosphorous)	Gleeble austenitized and PWHT (aged at 450°C, 2000 h)	19	16	10	1.5
Modified A 302 B (high phosphorous)	Gleeble austenitized and PWHT (aged at 450°C, 2000 h)	14	32	17	8
A 533 grade B (high phosphorous)	Gleeble austenitized and PWHT (aged at 450°C, 2000 h)	19	15	5	10

equation reasonably predicted segregated concentrations of all the modified A 302 B specimens but overestimated the A 533 grade B segregation. The problem could be that the bulk phosphorous given in Table 2 was used in the calculation, instead of the more recent phosphorous determinations that appear in Table 3.

The same Gleeble-austenitized and 450°C aged A 533 grade B steel analyzed by AEA was also submitted to Auger analysis at ORNL.¹⁷ In this method, a fresh fracture surface is created under vacuum and analyzed. Part of the surface is intergranular, exposing the grain boundaries (GBs), as shown in Fig. 9, and the other part of the surface is transgranular (TG) cleavage. Bulk phosphorous was determined on cleavage surfaces and on grain boundaries after 10Å were removed by sputtering (SP) (see Fig. 10). On average, the difference between bulk and GB phosphorous is a factor of 10. Additionally, intergranular fracture vanishes at upper-shelf test temperatures as seen, for example, in Fig. 11.

**Modified A302 Grade B
Low Phosphorus
200x**

**Modified A302 Grade B
High Phosphorus
200x**

Figure 9. Intergranular fracture of A 533 grade B steel after Gleeble austenitization, PWHT, and 200-h aging at 450°C. Charpy lower shelf test temperature.

Figure 10. Phosphorous concentrations in bulk material and grain boundaries.
 Auger analyses are shown after 2000 h, 450°C temper embrittlement.

**Modified A302 Grade B
Low Phosphorus
200x**

**Modified A302 Grade B
High Phosphorus
200x**

Figure 11. Fracture surface of material shown in Fig. 9 when tested at an upper shelf temperature.

Discussion

Evaluations made from a literature search on LBZs in multipass weldments of low-carbon microalloyed steels have revealed that the problem of LBZs, as described for microalloyed steels, does not transfer automatically to welds in RPVs. All of the published LBZ work addressed the problem associated with the welding of low-CE steels. The application is offshore structures for which operating temperatures can be as low as -40°C . Temperatures used in CTOD bend bar tests to find LBZs ranged from -10 to -60°C , which are well below temperatures of interest to RPV operations. PWHT can substantially alleviate LBZ effects in welds of microalloyed steels,⁴ but such a procedure is avoided if possible in offshore structures because of the difficulty and financial penalty to do so. All RPVs, on the other hand, are given PWHT. However, RPVs potentially could develop a brittle-zone problem because of a completely different embrittlement mechanism; namely, the susceptibility of RPV steels with enlarged prior-austenite grains to temper embrittlement.

Experiments at AEA-Technology have clearly demonstrated that steels with the typical RPV chemistry can be made to be highly sensitive to temper embrittlement provided the prior-austenite grain size is enlarged to more than $360\text{ }\mu\text{m}$ in diameter. The SAW process used in RPV fabrication is a high-heat-input welding procedure, and grains of substantial enlargement would seem quite likely; more so than in offshore SMAW weldments. Hence, there was a need to pursue more information on temper embrittlement. The scope of interest was the temper-embrittlement properties of commercially made RPV steels and the influence of grain-size and thermal-cycle effects.

The objective of Phase I was to confirm the AEA-Technology findings, except in this case commercially made steels were used. This work clearly showed that the temper embrittlement seen with the AEA-Technology laboratory-made steels did in fact apply to the five selected commercially made steels. However, the ΔT_0 magnitude was somewhat less. An interesting new observation was that ΔT_0 due to embrittlement seemed to be strongly influenced by the initial T_0 temperature value. A material with extremely low initial T_0 temperature will tend to experience greater ΔT_0 . The phosphorus and copper

concentrations do not seem as significant when compared with the influence of the initial transition temperature.

Phase II introduced an anticipated more-representative weld HAZ coarse grain size, as well as the heating and cooling rates of a typical weld pass. Both are important new and relevant variables. Most RPV steels have CE of about twice that of microalloyed steels. Consequently, the transformation products formed with the high cooling rates of a weld pass are favorable for the development of high fracture toughness, provided postweld stress relief tempering is used. All of the commercial steels had exceptionally good (low-temperature) preaged transition temperatures after simulated weld cycles. Given a 168-h anneal at 450°C, the fracture toughness was degraded, but not to an extent that cancelled the fracture-toughness improvement from the austenitization and PWHT cycle.

Phase III expanded the study to look at other aging-condition variables. The austenitization, PWHT, and grain size are the same as in Phase II. Aging time is fixed at 168 h, but aging temperature is varied. It appears that there can be a lower-bound aging temperature for temper embrittlement. At 399°C (750°F) there was no detectable temper embrittlement effect in any of the steels. At 450 and 482°C (842 and 900°F), significant transition-temperature shifts could be seen. However, there were no postembrittlement (168 h) T_0 temperatures observed above room temperature.

The work in progress in Phase III will address the effects of irradiation and postirradiation annealing. The final issue to be determined is whether an RPV steel with demonstrated temper-embrittlement insensitivity at temperatures below 400°C can suffer embrittlement at 288°C when the thermal exposure is combined with neutron irradiation or when irradiation is followed by postirradiation heat treatments at 454°C for 168 h, intended to recover fracture toughness.

It is important to point out that the T_0 temperature used here serves only as a basis for embrittlement damage assessment. Engineering design-related temperatures can only be obtained using fracture-mechanics test methods supplemented with the technology

of the Master Curve. A reference temperature, T_{100} , has been determined that corresponds to a median fracture toughness of $100 \text{ MPa } \sqrt{\text{m}}$. Whenever sufficient material was left over after CVN specimens were taken, fracture-mechanics specimens (1/2T compacts) were made, and T_{100} temperatures determined. These results are compared to CVN T_o temperatures in Table 10, columns 3 and 6. The difference is nominally 90°C . A crude correlation that is known to exist between Charpy transition curves and T_{100} did not work well with these temper-embrittled steels. Columns 5 and 6 would have been reasonably close if the expected relationship had existed.

Table 10. Comparison of Charpy V T_o and fracture-mechanics-based T_{100} temperatures

Material	Aging treatment (h/ $^\circ\text{C}$)	CVN temperatures ($^\circ\text{C}$)		Fracture mechanics T_{100} ($^\circ\text{C}$)		
		T_o^a	T_{28J}^b	$T_{28J} - 18$	T_{100}^c	Error ^d
A 302 B	2000/450	45	14.2	-4	-30	26
	168/490	44	18.0	0	-38	38
A 508	2000/450	41	-30	-48	-84	36
	168/490	16	-24	-42	-36	-6
A 533 B	2000/450	78	48	30	5	25
	168/490	65	2.0	-16	-49	33
	2000/450 + 168/490	66	46	28	-21	49
Mod A 302 B	As received	9	-70	-88	-88	0

^aTemperature at 50% between CVN lower shelf and upper shelf.
^bTemperature at 28-J CVN energy level.
^cIn several cases only four replicate tests were made. Standard deviation on T_o uncertainty is about 20% greater than results obtained with six replicate tests.
^dDifference between CVN prediction of T_{100} , namely $(T_{28} - 18^\circ\text{C})$ and the actual value by fracture-mechanics tests.

CONCLUSIONS

The postulate that unirradiated RPV welds might contain LBZs could not be supported by the results of this study. Almost every aspect of the findings of the problems associated with welding low-carbon microalloyed steels in offshore structures could not be rationalized as directly applicable to the reactor vessel multipass welding situation. However, in view of the findings of the AEA-Technology study, it was necessary to take into consideration the potential for temper embrittlement in the coarse-grain region of the HAZ. The AEA findings were essentially verified using five commercially made RPV steels. Gleeble simulation of an HAZ weld cycle was used to simulate the real-case metallurgical conditions. Temper embrittlement was not found at an aging temperature of 399°C (750°F), so it appears that the RPV operating temperature of 288°C (550°F) may not create a problem unless synergistic interactions with neutron irradiation occur. The experiment to investigate this postulated mechanism is in progress, and its results will also include the possibility of what might happen if recovery annealing of RPVs were introduced in future years.

In this study, aging at 450 and 482°C (842 and 900°F) for 168 h of soak time was investigated. Temper embrittlement was noted, but the transition temperatures of the initial conditions were so low that the CVN T_0 temperatures did not even reach room temperature. Additionally, these T_0 values were on average about 90°C (162°F) higher than the T_{100} transition temperatures related to in-service performance. Therefore, unless there are results to the contrary from the irradiation exposures, this investigation does not support the postulate of an LBZ problem in RPV welds.

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REFERENCES

1. D. P. Fairchild, "Fracture Toughness Testing of Weld Heat Affected Zones in Structural Steel," pp. 117–141 in *Fatigue and Fracture Testing of Weldments*, ASTM STP 1058, H. I. McHenry and J. M. Potter, Eds., American Society for Testing and Materials, Philadelphia, 1990.
2. C. Thaulow, O. M. Kselsen, M. Houge, B. Melve, A. J. Paauw, G. Rorvik, J. K. Solberg, and J. Troset, "Fracture Mechanics Testing of HSLA Steels for Offshore Applications," pp. 195–204 in *Proceedings of ASM Conference on Microalloyed Steels*, Vol. 8811-003, World Materials Congress, Chicago, 1988.
3. C. Thaulow, A. J. Paauw, and G. Rorvik, "Fracture Mechanics Testing of Weld Metal for Low-Carbon Microalloyed Steels," pp. 277–301 in *TMS Symposium on Welding Metallurgy*, Denver, February 1987.
4. C. Thaulow, A. J. Paauw, and K. Guttormsen, "The Heat-Affected Zone Toughness of Low-Carbon Microalloyed Steels," *The Welding Journal*, **9**, 266-s–279-s (September 1987).
5. The British Standards Institution, "Methods for Crack Opening Displacement (COD) Testing," BS 5762, 1979.
6. "API Specification for Production Qualification for Steel Plates for Offshore Structures," API RP2Z, American Petroleum Institute, Dallas, March 1987.
7. "Steel Specification for Fixed Offshore Structures," EEMUA Publication No. 150, Engineering Equipment and Materials Users Association, London, 1987.
8. A. J. R. Loureiro and A. A. Fernandes, "Toughness of CGHAZs of Welds in Q and T Steels," *The Welding Journal Supplement*, **9**, 225-s–232-s (September 1994).

9. A. M. Barnes, *The Effect of Intercritical Thermal Cycles on HAZ Microstructure and Toughness in C-Mn-Al-Nb Steel*, Report 402/1989, The Welding Institute, August 1989.
10. K. Arimochi, "CTOD Behavior of Steel Weld with 500 MPa Class Strength and HAZ-CTOD Characteristics of Newly Developed Low Al-B-N Steel," *Transactions of the Japan Welding Society*, **20**(2), 1989.
11. C. Thaulow and A. J. Paaw, "Materials Characterization with Respect to HAZ Local Brittle Zones," pp. 24-1 to 24-14 in *WELDTECH 88*, London, November 1988.
12. A. Barnes, "Local Brittle Zones in C-Mn Steel Multipass Welds," *The Welding Institute Bulletin*, September/October 1990.
13. R. J. McElroy, A. J. E. Foreman, G. Gage, W. J. Phythian, P. H. N. Ray, and I. A. Vatter, "Optimization of Reactor Pressure Vessel Surveillance Programmes and Their Analysis," contribution to IAEA CRP 3 Research Program, AEA-RS-2426, December 1993.
14. J. A. Wang, *Embrittlement Data Base, Version 1*, USNRC Report, NUREG/CR-5506 (ORNL/TM-13327), August 1999.
15. C. D. Lundin and S. Mohammed, *Effect of Welding Conditions on Transformation and Properties of Heat-Affected Zones in LWR Vessel Steels*, USNRC Report NUREG/CR-3873 (ORNL/SUB/78-7637/1), November 1990.
16. S. R. Ortner, J. A. Mace, J. Hyde, and C. A. English, "Grain Boundary Analysis in RPV Steels After HAZ-Simulating Treatment and Aging," AEAT-5292, March 1999.
17. D. E. McCabe and R. K. Nanstad, "Evaluation of the Potential of RPV Steels HAZ for Temper Embrittlement," *Proceedings of IGRDM Meetings, Nashville, Tennessee, January 18-22, 1999*.

APPENDIX

Examples of HAZ Coarse Grains Along Fusion Line, Fine Prior-Austenite Grain Zones, and Base Metal Grains from Three Selected Commercial Welds

Figure A.1. Midland Reactor, Unit 1, beltline weld. HAZ is on the right; all are at 200x.

Figure A.2. Midland Reactor, Unit 1, nozzle course weld. HAZ is on the right; one is at 20x, and two are at 200x; HAZs base metal; one is at 200x.

Figure A.3. HSSI Fifth Irradiation Series Weld 73W. HAZ is on the right; one is at 400x, and one is at 200x. Base metal is shown at 400x.

Figure A.4. HSSI Fifth Irradiation Series Weld 72W. HAZ is on the right; two are at 400 $\times$, fine-grain HAZ (FGHAZ, Fig. 1), and one is at 400 $\times$. Base metal is shown at 400 $\times$.

Figure A.5. SNUPPS longitudinal weld. HAZ is on the right; one is at 400x, fine-grain HAZ (FGHAZ, Fig. 1); one is at 400x. Base metal is shown at 400x.

Figure A.6. SNUPPS beltline weld. HAZ is on the right; two are at 200×, and one is at 100×.

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