Influence of intermediate postweld heat treatments on the fracture toughness of CrMoV steel welds

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Abstract. In the course of welding high thickness 2.25Cr1Mo0.25V steel plates, used in the fabrication of pressure vessels for the petrochemical industry, hard and brittle bainitic regions are produced in the heat affected zone and weld metal that can endanger the subsequent steps in the process of manufacturing these vessels. The application of a low temperature intermediate heat treatment has been studied as a substitute for conventional intermediate heat treatments with the aim of reducing costs and delivery time. The microstructure, tensile properties and fracture behavior (K_{Ic} and/or J_{Ic}) of the weld metal were analyzed in the as-welded condition, after the intermediate treatment and after the typical post-weld heat treatment. Finally, the capacity of this treatment to ensure a safe manufacturing process was demonstrated by means of an application example.

Introduction

As the service conditions of hydrogen conversion reactors in the petrochemical industry are shifting to higher temperatures and higher hydrogen pressures, plates with higher thicknesses (over 100 mm) are now being used and conventional 2.25Cr1Mo and 3Cr1Mo pressure vessel steels are being progressively replaced by stronger, more resistant steels. These include vanadium-modified low alloy steels such as 2.25Cr1MoV, 3Cr1MoV and 9Cr1MoV, in which the addition of vanadium in conjunction with a suitable heat treatment allows the precipitation of stable fine dispersed vanadium-containing carbides which increase tensile strength, creep rupture strength and the resistance to hydrogen assisted cracking [1]. 2.25Cr1Mo0.25V steels are currently being used in continuous services at 480ºC, 24 MPa and 17 MPa of hydrogen partial pressure [1].

Submerged arc welding of high thickness CrMoV plates is difficult; even when using preheating, inter-pass temperatures of around 200ºC and de-hydrogenation treatments at 350ºC, a brittle weld metal and a brittle heat affected zone are produced. Moreover, under the usual restraint conditions existing in industrial welds, cold cracking of these welds may occur. In order to avoid this problem in the case of the more geometrically restrained joints, intermediate heat treatments (680ºC, 4 hours) are recommended to achieve tempering and stress relaxation. These heat treatments are performed in huge industrial furnaces with the resulting significant increase in the final cost of the reactor [2]. On the other hand, the final strength of the product may be affected when these treatments are repeated several times throughout the manufacturing process [3]. Moreover, a post-weld heat treatment at 705ºC for 10 hours must be always applied at the end of the process of manufacturing the whole reactor.

An interesting possibility to reduce not only costs but also manufacturing time would be to replace the intermediate heat treatment with low temperature treatments which could be directly applied in the workshop by means of preheating torches, without the need for industrial furnaces. The aim of these treatments would be to apply a low tempering to the bainitic microstructures produced in the weld thermal cycle in order to attain a toughness level capable of preventing cracking in the following steps of manufacturing process [3]. It is therefore necessary to assess whether the low temperature heat treatment applied as an intermediate heat treatment during the manufacture of high thickness CrMoV pressure vessels increases the toughness of the welds to a high enough level or not.

Experimental procedure

Material

A 108 mm thick plate of 2.25Cr1Mo0.25V steel (SA 542 Grade D-Class 4) was used. The base metal was normalized at 950ºC, quenched in water from 925ºC and tempered at 720ºC for 3 hours. The chemical composition of the steel is shown in Table 1. A weld coupon with a length of 1300 mm and a width of 600 mm was produced using a maximum gap of 30 mm by means of a submerged arc welding procedure using alternating current, a 4 mm diameter Thyssen Union S1 CrMo2V consumable and a heat input of 2.2 kJ/mm (29-32 V, 425-550 A and 45-55 cm/min). A minimum preheat temperature of 205ºC and a maximum inter-pass temperature of 250ºC were likewise employed. Table 1 also gives the chemical composition of the weld metal.

Table 1. Chemical composition of the SA 542 Grade D-Class 4 steel and weld metal [%wt]

	%C	%Si	%Mn	%Cr	%Mo	%V	%Ni
Base metal	0.15	0.09	0.52	2.17	1.06	0.31	0.19
Weld metal	0.08	---	---	2.28	0.93	0.24	0.03

All the work was performed on the weld metal in the following conditions:

- As-welded (AW), which includes a de-hydrogenation treatment of 4 hours at 350ºC. This treatment is always absolutely essential when dealing with this steel.
- After an additional low temperature intermediate tempering consisting in repeating the first treatment, 350ºC for 4 hours (LTIT) to be sure that the treatment can be directly applied in the workshop just after welding.
- After the conventional post-weld heat treatment, 705ºC for 10 hours (PWHT).

Microstructure and conventional mechanical tests

The hardness profile of the weld was measured at a depth of 2 mm from the top surface using a Vickers indenter and a load of 10 kg. Samples of the base metal, heat affected zone and weld metal were cut, ground, polished and finally etched with Nital to reveal their microstructure.
Circular tensile specimens with a diameter of 10 mm and a calibration length of 70 mm were machined from the weld metal under the three aforementioned conditions, AW, LTIT and PWHT. Tensile tests were performed according to the UNE EN ISO 6892-1 standard [4] on a 100 kN MTS machine. Two different tests were performed in each case.

Fracture tests

Fracture toughness tests were performed at room temperature using single edge notched bend (SENB) specimens with a TS orientation (L is the welding direction, T the transverse direction and S the thickness direction) and a crack length to width ratio, a/W=0.5. The specimen dimensions were 180x40x20 mm. Specimens were fatigue pre-cracked at ambient temperature to the required nominal a/W at an R-ratio of 0.1 and were subsequently side-grooved. Fracture tests were carried out at room temperature in accordance with the ASTM E1820 standard [5], using a load-line displacement rate of between 0.2 and 0.4 mm/min.

The single-specimen method, based on the use of the elastic unloading compliance technique, was used to determine the J-Δa resistance curves. The results thus obtained were corrected using the physical measure of the crack determined at the end of each test by means of a suitable low magnification microscope. The J-integral was determined after splitting up its elastic and plastic components. The elastic component was obtained from the stress intensity factor, K, as:

$$J_e = \frac{K_i^2(a_i)(1-\upsilon^2)}{E} \qquad (1)$$

while the plastic component is given by:

$$J_{pli} = \left[J_{pl(i-1)} + \left(\frac{\eta_{i-1}}{b_{i-1}}\right)\frac{(P_i + P_{i-1})(v_{pli} - v_{pli-1})}{2B_N}\right]\cdot\left[1-\gamma_{i-1}\frac{a_i - a_{i-1}}{b_{i-1}}\right] \qquad (2)$$

The values of η_i and γ_i are given in the ASTM 1820 standard [5]. P_i and v_i are the applied load and the load-line displacement, E and υ the elastic modulus and the Poisson ratio, B_N, W, a, the net thickness, width and crack length and $b_i=W-a_i$, respectively.
Two fracture tests were performed in the AW, LTIT and PWHT conditions.

Results

Hardness and microstructure

Figure 1 shows a general view of the welded coupon and a macrograph of a transversal cut of the weld revealing the large number of passes required to fill the joint, while Table 2 shows the average Vickers hardness measured in the weld metal, coarse-grained heat affected zone and base metal of the joint in the as-welded condition. A substantial increase in hardness was produced in the heat affected zone and weld metal with respect to the base metal, due to the high alloy content of the steel in conjunction with the high thickness of this joint (high cooling rate).

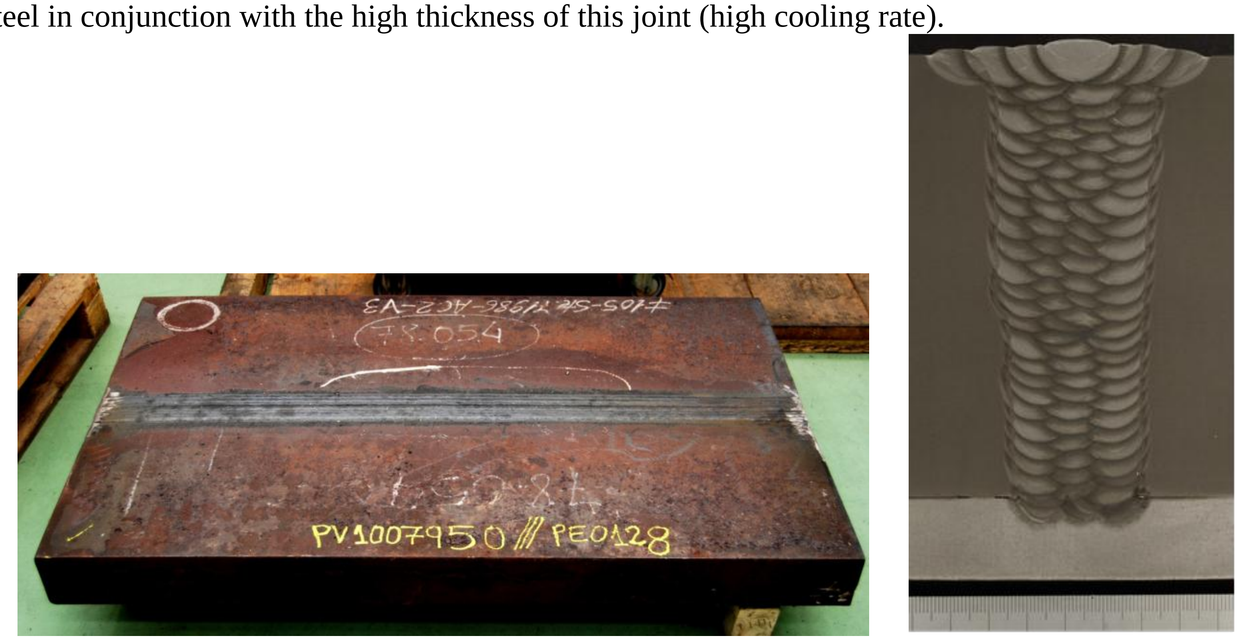
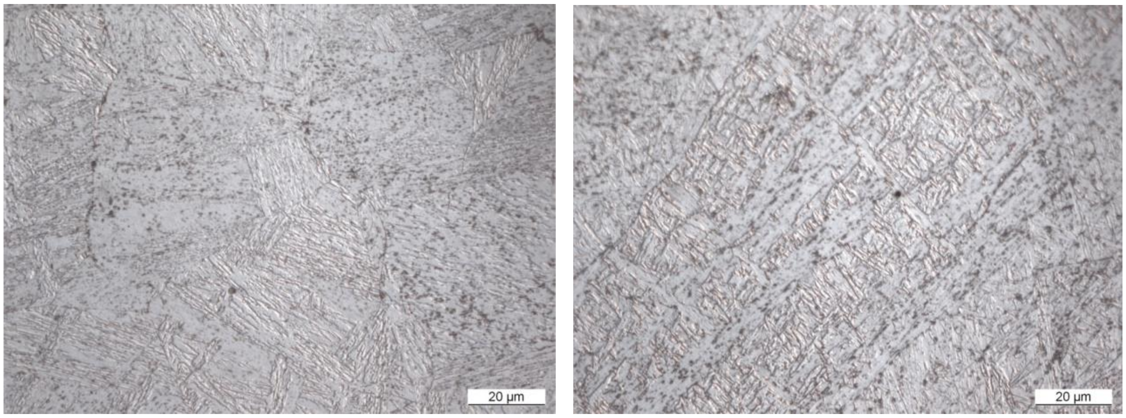


Figure 1. Welded coupon and macrograph of the weld.

Table 2. Vickers hardness of the different weld zones

HV10 [kg/mm^2]	BM	HAZ	WM
As-welded (AW)	224	365	334

Figures 2 a) and b) respectively show the microstructure of the coarse-grained heat affected zone (HAZ) and weld metal (WM) in the as-welded condition. Both regions are quite similar and are composed of fine randomly-oriented acicular bainitic packets, which justify their aforementioned high hardness.



a) b)

Figure 2 Microstructures of a) the coarse-grained heat affected zone (HAZ), and b) the weld metal (WM). As-welded condition.

Tensile properties

Table 3 shows the average tensile results (elastic modulus, yield strength, tensile strength, elongation and reduction of area) obtained on the weld metal in the three different studied conditions: as-welded (AW), after the low temperature intermediate heat treatment (LTIT) and after the post-weld heat treatment (PWHT).

Table 3. Average weld metal tensile results

	E [GPa]	σ_{ys} [MPa]	σ_{ts} [MPa]	E [%]	RA [%]
AW	220	1019	1120	17	58
LTIT	205	745	808	18	62
PWHT	208	533	631	23	70

Very significant tensile differences were obtained in the weld metal in the three studied conditions. The weld metal in the as-welded condition has very high yield and tensile strengths due to its non-tempered bainitic microstructure. Furthermore, both properties undergo a significant decrease when a heat treatment is applied, the effect of which increases with increasing tempering temperature and time (tempering of the bainite). It should be noted that the low temperature intermediate heat treatment (LTIT) is able to reduce both the yield strength and the tensile strength of the weld zone by approximately 300 MPa. The elongation and the reduction of area evolve in just the opposite way, as expected.

Fracture behavior

Figures 3, 4 and 5 present the load-displacement curves obtained in the fracture tests performed on the weld metal in the three studied conditions: as-welded, after a low temperature intermediate heat treatment and after a post-weld heat treatment.

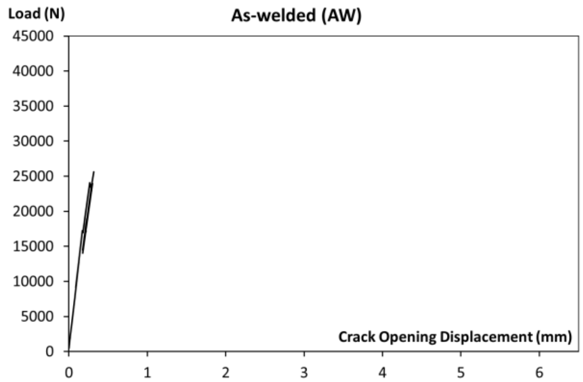


Figure 3. Load versus load-point-displacement. Weld metal, as welded (AW).

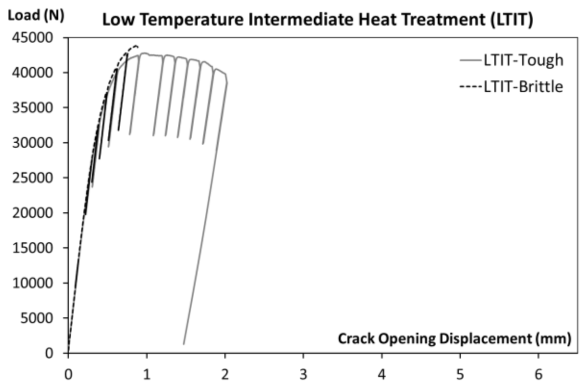


Figure 4. Load versus load-point-displacement. Weld metal, after an intermediate low temperature heat treatment (LTIT).

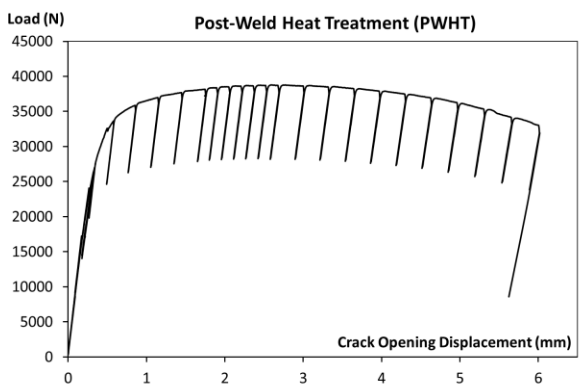


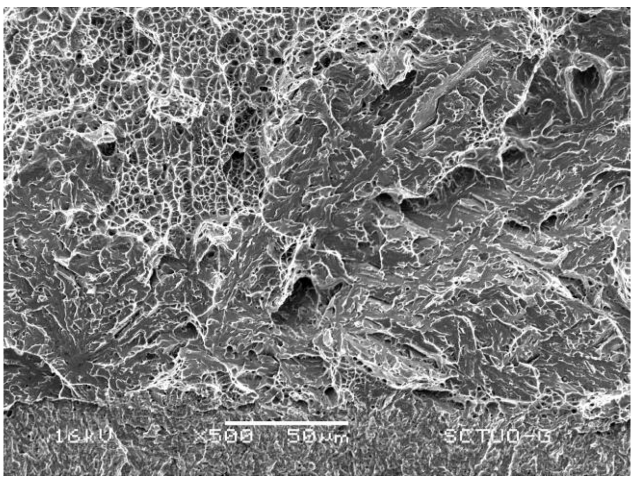
Figure 5. Load versus load-point-displacement. Weld metal, after a post-weld heat treatment (PWHT).

The weld metal in the as-welded condition showed a linear elastic behavior with a brittle, sudden catastrophic failure without any previous crack growth. The fracture behavior of this condition was characterized by means of the critical K parameter, which was calculated according to the ASTM E399 standard (it cannot be strictly denoted as the fracture toughness, since the P_{max}/P_Q <1.1 criterion was not satisfied) [6]. Table 4 shows the average result obtained.

On the other hand, when the weld metal is submitted to a low temperature intermediate heat treatment (LTIT: 350ºC, 4 hours), a tougher behavior was obtained with some stable crack growth during the fracture test. However, this behavior was different in the two tested specimens (Fig. 4): one exhibited stable crack growth until the end of the test (LTIT-Tough), while the other (LTIT-Brittle) gave way to sudden failure after only a very short crack growth. This behavior means that the transition temperature of this product may be around room temperature. The J-value at the onset of crack growth, J_Q, was determined in the case of the tough specimen (LTIT-Tough), but could not be determined for the brittle specimen (LTIT-Brittle), as crack growth was too low. In this latter case, the fracture toughness, K_{Ic}, could not be calculated either due to the high degree of plastification of the fracture specimen.

Finally, Fig. 5 shows the very tough behavior of the weld metal after the post-weld heat treatment (PWHT). This product exhibits a ductile elasto-plastic behavior with stable crack growth until the end of the fracture test, allowing the J-value at the onset of the crack growth, J_Q, to be determined. Table 4 gathers all these results. Eq. 1, which gives the relationship between the K intensity factor and the J-integral, was used to obtain the K_Q values corresponding to the specimens that exhibited ductile behavior in the fracture tests.

Figure 6 shows the brittle fracture of the specimen in the as-welded condition (although voids are also seen in some regions) and the ductile crack growth initiation micromechanism (nucleation, growth and coalescence of microvoids) observed in the LTIT-Brittle specimen, which finally broke by cleavage in a brittle manner after a crack growth of approximately 0.15 mm. All the other fracture specimens broke by means of a ductile micromechanism, as is also shown in the same figure.



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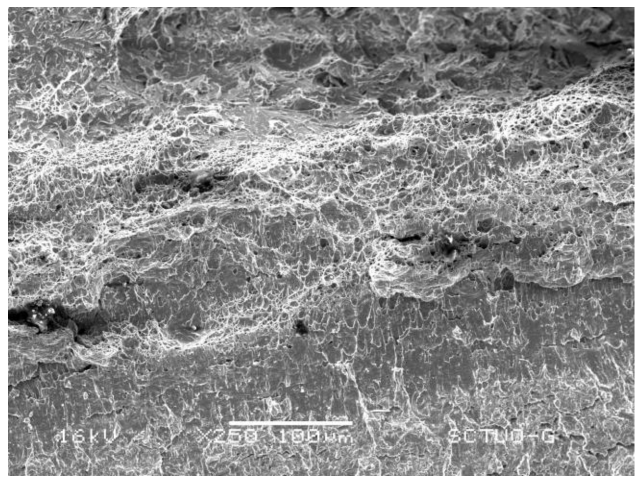

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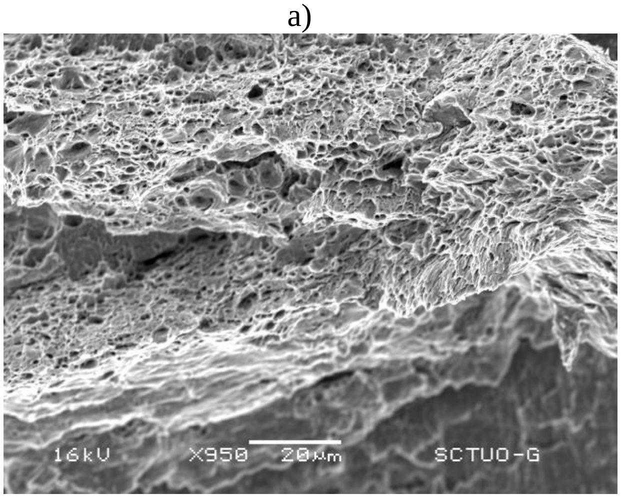

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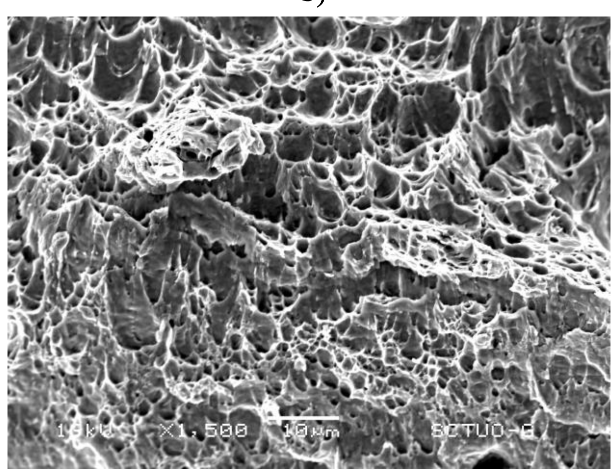

d)

Figure 6 a) As-welded condition, b) Low temperature intermediate treatment, LTIT-Brittle c) Low temperature intermediate treatment, LTIT-Tough, d) PWHT condition.

Table 4. Fracture parameters

2.25Cr1Mo0.25V Weld Metal	K_Q ($MPa\sqrt{m}$)	K_{max} ($MPa\sqrt{m}$)	J_Q (kJ/m^2)
AW	54	82	---
LTIT-Brittle	85	142	---
LTIT-Tough	230*	---	236
PWHT	427*	---	800

*Values obtained after applying Eq. 1

Discussion and application

The main aim of the present study was to analyze the possibility of using a low temperature intermediate heat treatment (350ºC, 4 hours) as a substitute for the conventional intermediate heat treatment (680ºC, 4 hours) in order to reduce costs and manufacturing time. This treatment produces a soft tempering of the bainitic microstructures generated during the weld thermal cycle that gives rise to a significant reduction in yield strength and an increase in fracture toughness with respect to the as-welded material.

We shall now attempt to demonstrate that the low temperature intermediate heat treatment applied to high thickness CrMoV welds upgrades their toughness to a level that ensures completion of the reactor manufacturing process without the risk of cold cracking. It should be noted that the ASME construction code now allows the avoidance of intermediate heat treatments only in longitudinal and circumferential welds in shell-to-shell and shell-to-head cases. We shall now proceed to compare the critical crack size which will give rise to the fracture of the plate (the reactor may have diameters of around 5 m or even larger and, as already mentioned, thicknesses greater than 100 mm) if residual stresses as high as the material yield strength are considered to exist as a consequence of welding operations in very restrained welds [7], or with a magnitude of half the material yield strength in a more compliant construction detail. It is likewise assumed that welding can produce semi-elliptical surface defects with a depth to length ratio, a/2c=0.1 (a typical arc strike defect) [8]. The stress intensity factor, K_I, corresponding to the considered geometry (very large diameter and with defect depths much lower than the thickness of the plate) is [9]:

$$K_I = [1.13 - 0.09(a/c)]\sigma\sqrt{\pi a/Q} \quad (3)$$

$$Q = 1 + 1.464(a/c)^{1.65} \quad (4)$$

The critical defects shown in Table 5 were then calculated applying the fracture criterion for the onset of instable crack growth in the weld metal, K_I=K_Q, in the as-welded condition and also after the low temperature intermediate heat treatment (in this case, the lower toughness of the two tests was used), assuming both exposed situations: the presence of a maximum residual stress equal to the yield strength [9] and also with a magnitude of half this value. The critical defects after the post-weld heat treatment were also similarly determined for the sake of comparison alone (they are larger than the plate thickness).

The results thus obtained show that, as a result of the application of the low temperature intermediate heat treatment, the critical defect depth increases by a factor of 4.6 with respect to the critical defect in the as-welded state. Moreover, even in the most unfavorable situation ($\sigma_R = \sigma_{ys}$), the critical defect is quite large (a=3.7 mm, 2c=37 mm). Furthermore, this defect would not grow in an unstable manner. Unstable growth would only occur under a stress intensity factor of 142 $MPa\sqrt{m}$ (Table 4). Moreover, assuming a residual stress as great as the yield strength, the critical defect would have a depth a=10.3 mm and a length 2a=103 mm. As this critical defect seems to be large enough so as to be rarely produced in the workshop, the low temperature intermediate heat treatment thus seems to ensure the safe manufacture of the pressure vessel. However, the weld metal in the as-welded condition under residual stresses similar to the yield strength gives rise to a critical defect with a depth of only 0.8 mm, which is unacceptable, as it seriously jeopardizes the subsequent steps in the process of manufacturing the reactor.

Table 5. Critical crack depths (a/2c=0.1) under residual stresses equal to the yield strength and half the yield strength

Weld metal	a [mm] $\sigma_R = \sigma_{ys}$	a [mm] $\sigma_R = \sigma_{ys} / 2$
AW	0.8	3.2
LTIT	3.7	14.8
PWHT	182	729

It is also worth noting that the size of the critical defect after the low temperature intermediate treatment in the most restrained condition is even larger than that corresponding to the as-welded condition under a more compliant construction detail.

Conclusions

The mechanical properties of the weld metal obtained by submerged arc welding of high thickness CrMoV plates in the as-welded condition, after a low temperature intermediate heat treatment (350ºC, 4 hours) and after a post-weld heat treatment (705ºC, 10 hours) have been determined. The low temperature intermediate heat treatment produces a soft tempering of the as-welded bainitic microstructure which gives rise to a significant decrease in yield strength and tensile strength, along with an increase in elongation, reduction of area and fracture toughness.

It was likewise demonstrated that the applied low temperature intermediate heat treatment seems to ensure the safe manufacture of the pressure vessel and could therefore be used in some circumstances to replace the conventional intermediate heat treatment, thereby reducing both the final cost and the manufacturing time of this type of equipment.

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