# Optimized Postweld Heat Treatment Procedures for 17-4 PH Stainless Steels

*The effect of prior microstructure and room-temperature tensile properties on postweld heat treatment was investigated*

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ABSTRACT. The postweld heat treatment (PWHT) procedures for 17-4 PH stainless steel weldments of matching chemistry was optimized *vis-a-vis* its microstructure prior to welding based on microstructural studies and room-temperature mechanical properties. The 17-4 PH stainless steel was welded in two different prior microstructural conditions (condition A and condition H 1150) and then postweld heat treated to condition H900 or condition H1150, using different heat treatment procedures. Microstructural investigations and room-temperature tensile properties were determined to study the combined effects of prior microstructural and PWHT procedures.

## Introduction

The precipitation-hardening (PH) stainless steels provide both strength and corrosion resistance. Chromium imparts corrosion resistance, and strength comes from precipitation hardening by submicroscopic precipitates on aging at elevated temperatures. This combination of properties makes the PH stainless steels very popular for severe service conditions. The PH stainless steels are classified by structure: martensitic, semiaustenitic and austenitic. Of these the martensitic types are most popular, while the austenitic types are only used for special applications. Welding of PH stainless steels is much like welding conventional austenitic and martensitic stainless steels, and requires controlled procedures to keep the heat input low for developing best ductility and toughness (Ref. 1). Either matching composition or lower-strength filler metals are used, with the weldments of matching filler metals attaining strengths approximating the base metal PH type.

The martensitic PH stainless steel of the Type 17-4 PH can be obtained with a wide range of mechanical properties by suitable heat treatment in the temperature range 900°–1150°F (482°–621°C) (Refs. 2–5). This steel retains its useful strength up to about 900°F, and is used in different heat-treated conditions in nuclear, naval and aerospace applications, where high strength and toughness, good fabrication characteristics and moderate corrosion resistance are required. This material has a high strength-to-weight ratio, and some of its typical applications include aircraft and missile fittings, fasteners, gears, jet engine parts, valve parts, chemical process equipment, pump shafts, paper mill equipment and nuclear reactor components (Refs. 6–9).

The weldability of 17-4 PH stainless steel (SS) is reported to be excellent despite its similarity to AISI 400 series martensitic stainless steels (Refs. 10, 11), and it can be welded with any of the usual arc, resistance or high-energy-density welding processes. Preheating (Refs. 12–16) or PWHT is not required to prevent cracking or restore ductility (Refs. 10, 11). In this material, the heat-affected zone (HAZ), immediately adjacent to the fusion zone, is effectively annealed or softened by welding heating and cooling cycles (Refs. 12, 15, 17) because of the presence of retained austenite in the microstructure (Ref. 12). Hence, this material can be welded in the aged conditions without causing cracking (Refs. 11, 15), as the heat of welding causes local softening of the HAZ (Ref. 12). Further, welding in the solution-treated (ST) condition causes no appreciable precipitation hardening of the solution-treated structure as the heating time during welding is too short (Refs. 12, 14, 15). For welding 17-4 PH SS, filler metals and electrodes of either matching composition or low-strength high-ductility stainless steel are generally preferred (Refs. 1, 11, 15, 16). Weldments made with a matching filler metal can be aged to strength levels comparable to those of the base metal and are used for producing weldments with high strength. If, however, a lower strength level is permissible, austenitic stainless steel weld metals can be used.

In 17-4 PH SS, martensite which is stable at low temperatures, begins to transform to austenite at 1160°F (627°C) and transformation is completed at 1300°F (704°C). With further increase in temperature, the precipitates go into solution; this process being completed at 1900°F (1038°C). On cooling from 1900°F, transformation from austenite to martensite starts at 270°F (132°C) and the marten-

**KEY WORDS**

17-4 Stainless Welds
PWHT Procedure
17-4 PH Stainless
Aging
17-4 PH Weldability
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Martensite Matrix
Delta Ferrite
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Epsilon Copper

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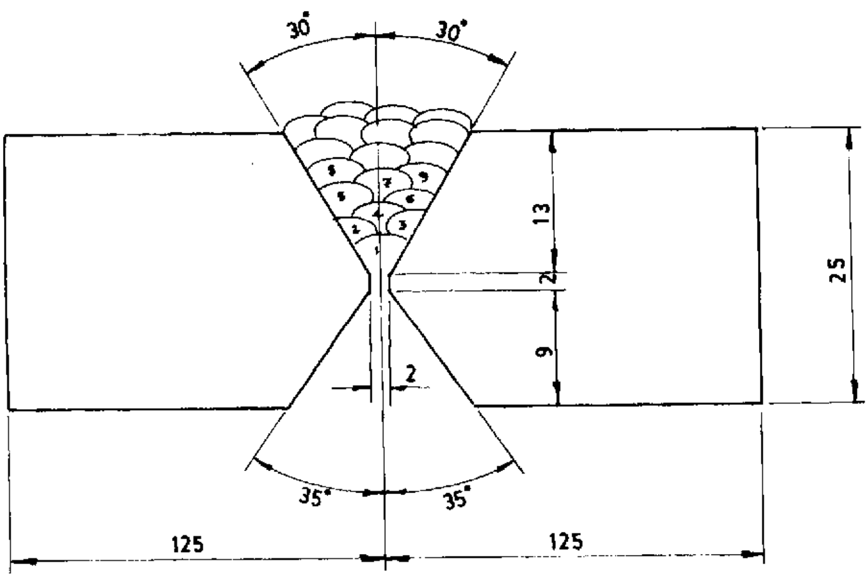



*Fig. 1 — Details of the double V-groove joint geometry used. All dimensions are in mm.*

**Table 1 — Chemical Composition of 17-4PH SS Base and Weld Metal Used (wt-%)**

| Metal | C | Mn | Si | S | P | Cr | Ni | Cu | Nb | Fe |
|---|---|---|---|---|---|---|---|---|---|---|
| Base | 0.064 | 0.55 | 0.33 | 0.001 | 0.021 | 16.8 | 4.70 | 3.8 | 0.20 | Balance |
| Weld | 0.060 | 0.52 | 0.38 | 0.007 | 0.020 | 16.5 | 4.5 | 3.7 | 0.25 | Balance |

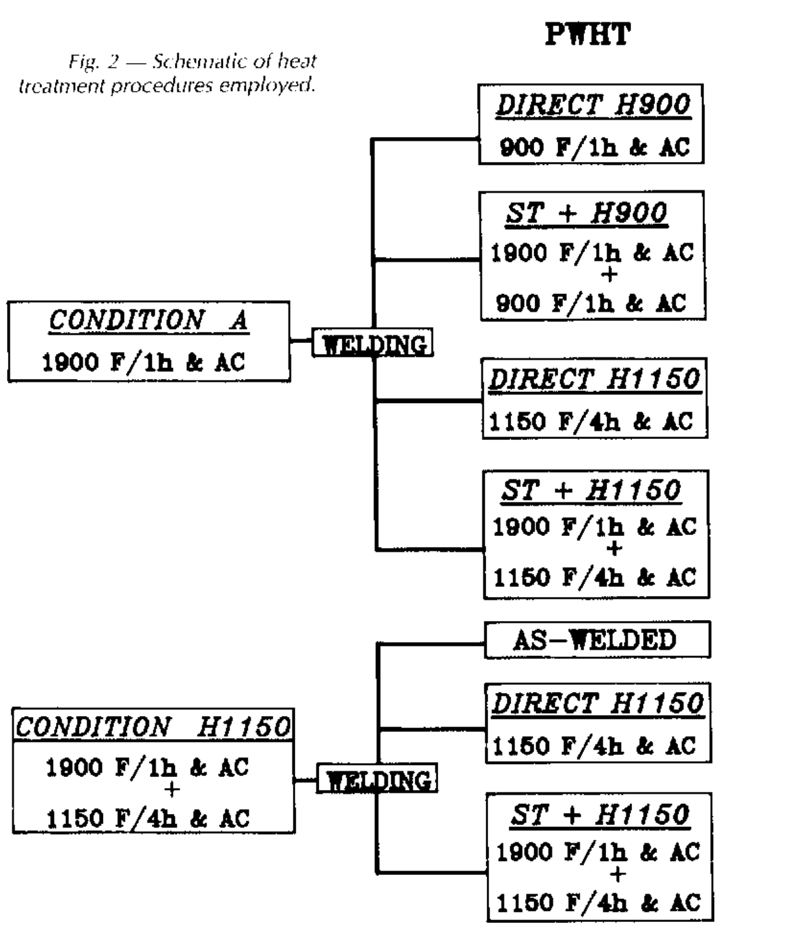



*Fig. 2 — Schematic of heat treatment procedures employed.*

**Table 2 — Welding Conditions Used for the Fill Passes**

| | |
|---|---|
| Welding process | SMAW |
| Welding consumable | 17-4PH SS |
| Consumable specification | E 630-16 |
| Consumable diameter (mm) | 3.15 |
| Welding position | 1G (Flat) |
| Polarity | DCEP (Reverse) |
| Welding current (Amps.) | 80–100 |
| Interpass temperature (°F) | 300 max. |

sitic transformation is completed at 90°F (32°C). A (ST + aging) PWHT is generally carried out on 17-4 PH SS weldments to prevent severe corrosion of the HAZ (Ref. 18), and to impart optimum mechanical properties (Refs. 15, 18). However, when thicknesses less than 12.5-mm (0.5-in.) thick are welded in the solution-treated (ST) condition (*i.e.*, with all precipitants in solution in the untempered martensitic structure), an aging PWHT at 900° to 1150°F can achieve properties approaching those obtained after a (ST + aging) PWHT (Ref. 18). Further, for service in hot concentrated chloride media, an aging PWHT at 1025°–1150°F (551°–621°C) is generally carried out (Refs. 1, 15). While a single-pass weld can be hardened by an aging PWHT, multipass welds show less uniformity in response to an aging PWHT as the successive application of welding heat results in variation in the microstructure of the weld metal (Refs. 10–12,16). Hence, for multipass welds, a ST-PWHT is required to restore the entire weld microstructure to a homogeneous condition to permit uniform hardening. Where a solution-treatment is not feasible for weldments thicker than 12.5 mm (0.5 in.), a direct aging PWHT at 1025°F is performed.

The effect of prior microstructure and PWHT on the microstructure and room-temperature tensile properties of 17-4 PH SS weldments, welded using matching AWS E630 consumable, was studied to optimize the PWHT procedure for the weldment *vis-a-vis* its microstructure prior to welding. This paper reports and discusses the results of this experimental investigation.

## Experimental Procedure

Plates 25 mm (1 in.) thick of 17-4PH SS were heat treated to two different microstructural conditions prior to welding, *viz:* 1) condition A (*i.e.*, solution-treatment (ST) at 1900°F for 1 h followed by air cooling) and 2) condition H1150 (*i.e.*, overaging at 1150°F for 4 h followed by air cooling). These plates were then welded using a double V-groove joint geometry as shown in Fig. 1. The root

passes were made by the gas tungsten arc welding (GTAW) process using AWS E308 filler metal, and the subsequent passes were made by the shielded metal arc welding (SMAW) process using 17-4PH SS electrodes conforming to AWS Specification E630-16. However, during subsequent preparation test specimen blanks, the root region was completely machined off to ensure that the weld metal consisted of only 17-4PH SS (*i.e.*, only the SMAW deposit). The chemical composition of the 17-4 PH SS base and weld metal are given in Table 1, and the welding conditions employed for the fill passes are listed in Table 2.

The plates welded in condition A (subsequently referred to as "condition A weldments") were postweld heat treated either to condition H900 (*i.e.*, aging at 900°F for 1 h followed by air cooling) using two different PWHT procedures, *viz*: 1) direct H900 and 2) ST + H900, or to condition H1150 using two different PWHT procedures, *viz*: 1) direct H1150 and 2) ST + H1150. The plates welded in condition H1150 (subsequently referred to as "condition H1150 weldments") were either retained in the as-welded condition, or postweld heat treated to condition H1150 using two different PWHT procedures, *viz*: 1) direct H1150 and 2) ST + H1150. A schematic of the heat treatment procedures employed is given in Fig. 2.

Transverse weld specimen blanks with the weld located at the center were obtained as shown in Fig. 3. These were metallographically polished and etched with Fry's reagent for optical microscopic examination and determination of microhardness profiles across the weld interface and HAZ. The microhardness measurements (in VPN) were made at intervals of 0.25 mm (0.01 in.) using a load of 500 g. Transverse weld tensile specimens, as shown in Fig. 4, were machined from the specimen blanks, and three specimens per condition were tested at room temperature using a nominal strain rate of 3.2 x $10^{-4}$S-1. The tensile properties, as well as the stress-strain behavior of these weldments, were analyzed as a function of prior microstructure and PWHT.

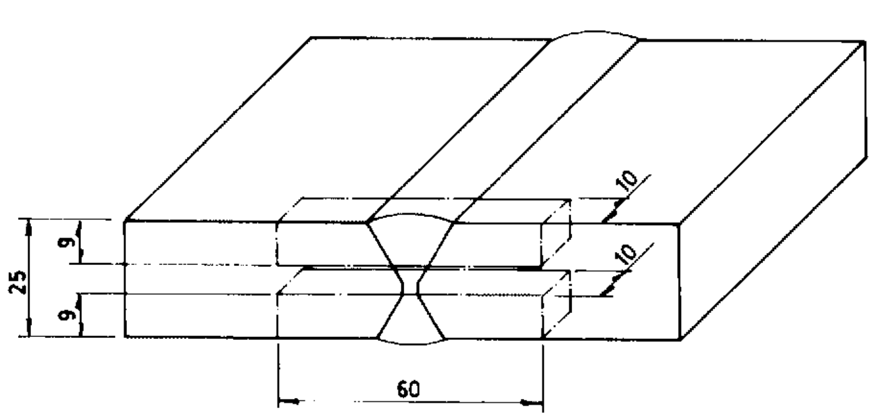



*Fig. 3 — Location of transverse weld specimen blanks. All dimensions are in mm.*

## Results and Discussion

### Microstructure

The microstructure of the unaffected base metal in weldments after the condition H900 PWHT consists of a matrix of equiaxed aged martensite, while that in weldments after the condition H1150 PWHTs consists of heavily overaged martensite. Further, in all cases, delta ferrite was observed as stringers along the prior working direction in the unaffected base metal. The martensite obtained after the solution treatment is supersaturated with copper and, on aging at 900°F copper-rich precipitation occur. On overaging these coherent precipitates transform from bcc structure to incoherent fcc epsilon-copper precipitates. During welding, in the HAZ immediately adjacent to the weld metal, the martensite transforms to austenite during the heating cycle; during the subsequent cooling cycle, some of this austenite transforms back to martensite (called retransformed-martensite) and some of it is retained as austenite (retained-austenite). In an earlier investigation (Ref. 19), it was found that the HAZ in the as-welded condition A bead-on-plate weld is comprised of three different microstructural zones, namely, zones of 1) retransformed martensite and retained austenite, 2) overaged martensite, and 3) underaged martensite. In contrast, the as-welded overaged-condition bead-on-plate weld is composed of almost entirely heavily overaged martensite, *i.e.*, Zone 2 of Ref. 19.

Microscopic examination showed that near the weld interface, the HAZ microstructure for all the conditions contains retained austenite (lightly etched regions in Figs. 5–7) and retransformed martensite, with the martensite (darkly etched regions in Figs. 5–7) being aged/overaged to varying degrees depending on the PWHT. Also, the delta-ferrite stringers present in the base metal become discontinuous in the HAZ in the regions near the weld interface, because of transformation of a part of the delta ferrite in this region at the high temperatures experienced during welding.

Figure 5A–B shows the microstructure of the HAZ near the weld interface for condition A weldments subjected to condition H900 PWHT. The HAZ microstructure after the direct H900 PWHT shows nonuniformly distributed large grains (Fig. 5A), while after the ST + H900 PWHT shows uniformly distributed smaller grains (Fig. 5B) caused by transformation during the ST step.

Overaging at 1075°F (635°C) and above results in retransformation of a significant amount of the martensite to austenite, which forms predominantly along the martensite lath boundaries (Ref. 20); on subsequent air cooling, much of this austenite is retained in the

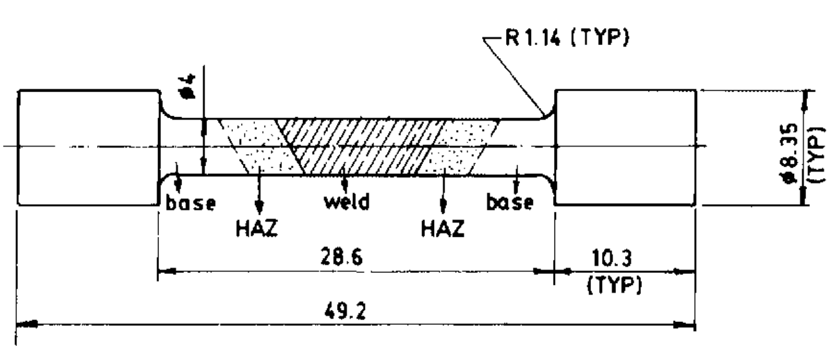



*Fig. 4 — Geometry of transverse weld round tensile specimens. Location of weld and HAZ are shown schematically. All dimensions are in mm.*

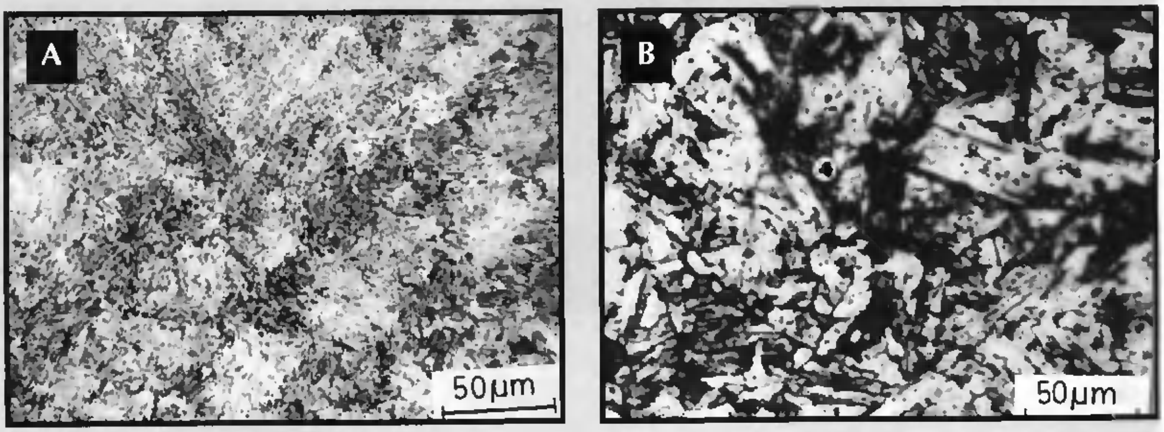


Fig. 5 — Microstructure of HAZ in condition A weldments. A — After direct H900 PWHT; B — after ST + H900 PWHT.

martensite matrix (Ref. 7), while the remaining is retransformed to unaged martensite (Refs. 7, 21). In fact, after 4 h overaging at 1100° and 1150°F, respectively, about 6 and 16 vol-% retained austenite is reported to be present (Ref. 21).

Figure 6A–B shows the microstructure of condition A weldments subjected to condition H1150 PWHT, the HAZ of which is comprised of regions of retransformed martensite and retained austenite and of overaged martensite. Both the PWHT procedures to condition H1150 result in the formation of retained austenite in the HAZ, with its amount after the direct H1150 PWHT (Fig. 6A) being higher than after the ST + H1150 PWHT — Fig. 6B. The reduced amount of retained austenite present after the ST + H1150 PWHT is due to transformation during the ST step.

Figure 7A–C shows the microstructure of condition H1150 weldments, the HAZ of which is also comprised of regions of retransformed martensite and retained austenite and of overaged martensite. In all the cases, retained austenite is present in the HAZ in varying degrees; its amount being lowest in the as-welded condition (Fig. 7A) and highest after the direct H1150 PWHT — Fig. 7B. The lower amount of retained austenite observed after the ST + H1150 PWHT (Fig. 7C) is due to the ST step, which transforms a greater amount of the retained austenite.

**Microhardness Profile**

Figure 8A shows the microhardness profile across the weld interface and HAZ for condition A weldments subjected to condition H900 PWHT. The formation of the hardness peak in the HAZ after the direct H900 PWHT is due to aging of the unaged region in the HAZ, *i.e.* ,Zone 3 of Ref. 19. The aging process results in an increase in the hardness due to precipitation of coherent copper-rich precipitates. On the other hand, the ST step in the ST + H900 PWHT results in dissolution of the copper-rich precipitates, thereby effectively evening out almost all the hardness variation of the HAZ.

Figure 8B shows the microhardness profile across the weld interface and HAZ for condition A weldments subjected to condition H1150 PWHT. Both the PWHT procedures overage the HAZ; the overaging process involving formation and coarsening of incoherent epsilon-copper precipitates. The hardness trough in the HAZ near the weld interface obtained after the direct H1150 PWHT can be attributed to the presence of an increased amount of the retained austenite and nondissolution of copper-rich precipitates in the absence of the ST step, which almost evens out the hardness variations in the HAZ as observed after the ST + H1150 PWHT.

Figure 8C shows the microhardness profile across the weld interface and HAZ for the condition H1150 weldments. The hardness trough in the as-welded HAZ is due to the presence of heavily overaged martensite containing coarse incoherent epsilon-copper precipitates, *i.e.*, Zone 2 of Ref. 19. The effect of welding induced overaging in the as-welded HAZ decreases (*i.e.*, the hardness increases) with increasing distance from the weld interface; the initial dip in hardness near the weld interface can be attributed to the retained-austenite formed and the complete/partial dissolution of coarse epsilon-copper precipitates in the HAZ during welding interfering with the overaging process. The direct H1150 PWHT results in partial recovery of hardness near the weld interface due to reprecipitation of incoherent

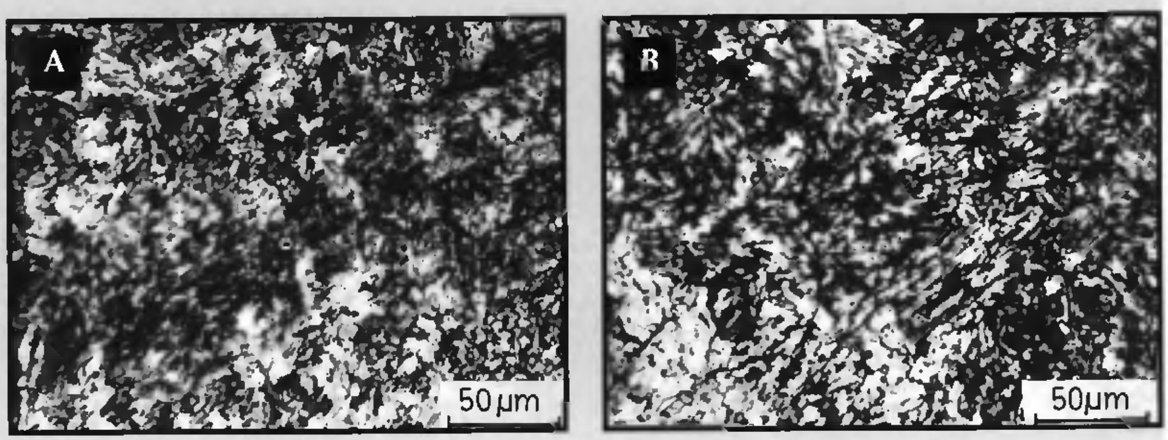
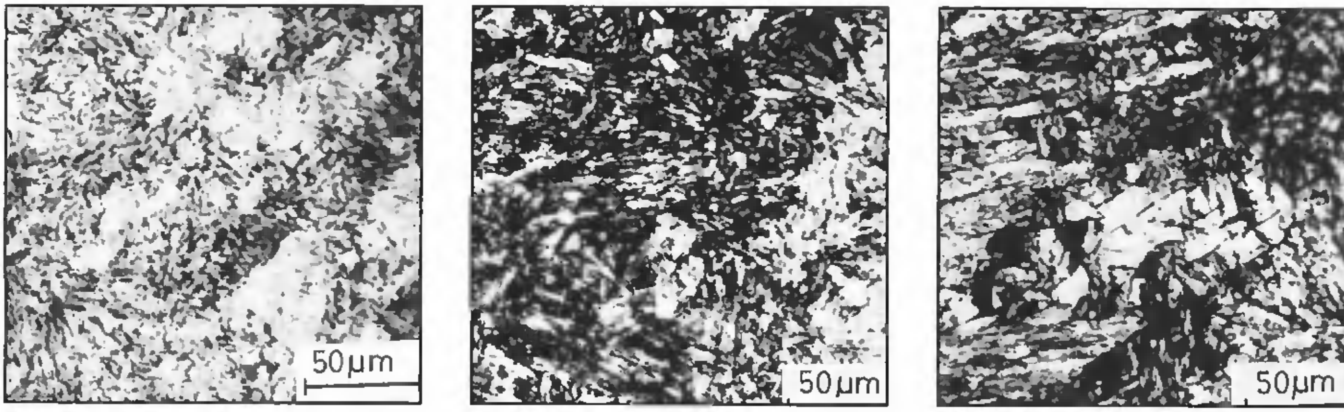


Fig. 6 — Microstructure of HAZ in condition A weldments. A — After direct H1150 PWHT; B — after ST + H1150 PWHT.




A B C

*Fig. 7 — Microstructure of HAZ in condition H1150 weldments. A — In as-welded condition; B — after direct H1150 PWHT; C — after ST + H1150 PWHT.*

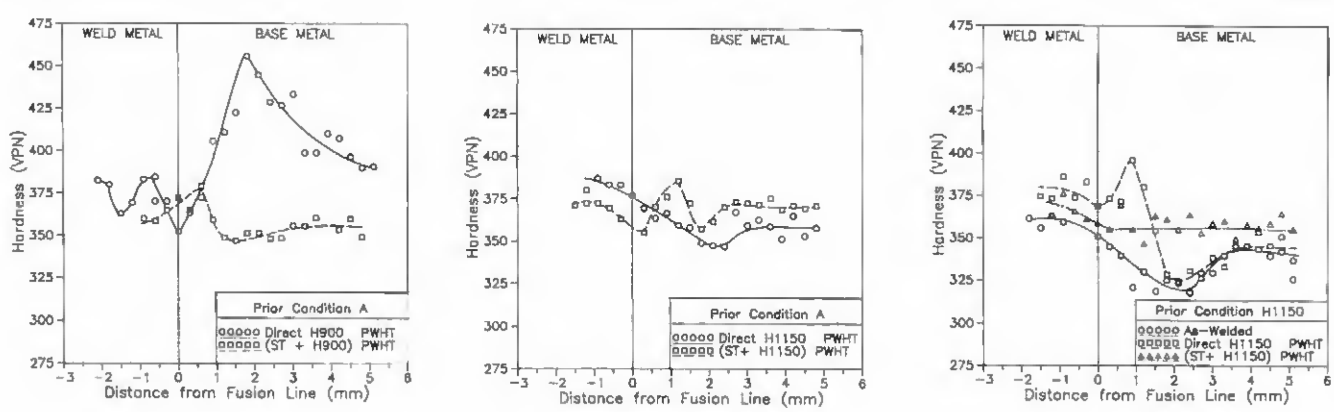
epsilon-copper precipitates in this region, where these were likely to have been completely dissolved due to more pronounced effect of heating during welding. However, the direct H1150 PWHT does not have any significant effect on the hardness of the as-welded HAZ farther away from the weld interface line as it only coarsens the partially dissolved coarse incoherent epsilon-copper precipitates. As expected, the hardness trough, in the as-welded HAZ is completely evened out by the ST step in the ST + H1150 PWHT; this PWHT procedure results in complete dissolution and reprecipitation of the incoherent epsilon-copper precipitates throughout the HAZ.

### Tensile Properties

The average room-temperature tensile properties of transverse weld specimens in the various PWHT conditions are presented in Table 3, while their true stress/true strain plots are given in Fig. 9A–C. The room-temperature transverse tensile properties determined for the base metal in conditions H900 and H1150 are also presented in Table 3 for comparison.

For condition A weldments subjected to condition H900 PWHT, both the PWHT procedures result in adequate yield strength (YS) and ultimate tensile strength (UTS) compared to those for the base metal in condition H900. The ST + H900 PWHT results in slightly higher total elongation (TE) and reduction in area (RA) compared to the base metal in the same condition. In fact, the ST + H900 PWHT results in the highest YS, TE and RA values among all the PWHTs used, while the UTS values are slightly lower than that after the direct H900 PWHT. Further, tensile fracture occurs in the HAZ, which is known from earlier studies (Refs. 12, 17, 19, 22) to be a soft zone. The slightly higher ductility after ST + H900 is also attributed to the soft HAZ. The true stress/true strain plots (Fig. 9A) show that the strength and ductility of the weldments are not significantly affected by the PWHT procedure. However, from Table 3, it is observed that the YS, TE and RA values for the weldment subjected to the ST + H900 PWHT are higher. Hence, for better tensile properties of condition A weldments for use in condition H900, the ST + H900 PWHT procedure is preferable to the direct H900 PWHT procedure.

The tensile properties of condition A weldments after direct H1150 PWHT (Table 3) show that an overmatched weldment (*i.e.*, with failure occurring in the base metal) is obtained after the direct overaging PWHT, as this PWHT route results in YS and UTS values that are slightly higher than those for the base metal in condition H1150. This is associated with a TE value identical to that for the base metal in the same condition. The true stress/true strain plots (Fig. 9B) show that, compared to the direct H1150 PWHT, the ST + H1150 PWHT results in a markedly lower weldment strength and ductility. Further, Table 3 shows that the ST + H1150 PWHT results in considerably lower strength and ductility values in comparison to the weldment subjected to the direct H1150 PWHT, as well




A B C

*Fig. 8 — Microhardness (500-g load) profiles measured across the weld interface and HAZ. A — Condition A weldments after condition H900 PWHT; B — condition A weldments after condition H1150 PWHT; C — condition H1150 weldments.*

**Table 3 — Average Room-Temperature Tensile Properties of Transverse Weld Specimen of 17-4PH SS Weldments (transverse base metal properties for comparison)**

| Base Metal Condition | Postweld Heat Treatment (PWHT) Procedure | Yield Strength (MPa) | Ultimate Tensile Strength (MPa) | Uniform Elongation (%) | Total Elongtion (%) | Reduction in Area (%) | Failure Location |
|---|---|---|---|---|---|---|---|
| A | Direct H900 | 944 | 1189 | 11 | 14 | 34 | HAZ |
| A | ST + H900 | 1052 | 1159 | 12 | 17 | 43 | BM |
| A | Direct H1150 | 950 | 1122 | 12 | 17 | 43 | BM |
| A | ST + H1150 | 560 | 680 | 7 | 9 | 39 | HAZ |
| H1150 | — | 779 | 1022 | 11 | 16 | 59 | HAZ |
| H1150 | Direct H1150 | 836 | 1020 | 10 | 16 | 60 | HAZ |
| H1150 | ST + H1150 | 924 | 1097 | 12 | 17 | 46 | HAZ |
| H900 | Base Metal | 1156 | 1314 | — | 15 | 61 | BM |
| H1150 | Base Metal | 931 | 1103 | — | 17 | 65 | BM |

as the base metal in condition H1150. Hence, for condition A weldments for use in condition H1150, the direct H1150 PWHT procedure is to be used for superior tensile properties.

The tensile properties for the condition H1150 weldments (Table 3), show that welding in the overaged condition results in an undermatched weldment (with failure occurring in the HAZ), as the YS and UTS values for the weldments are lower than those for the base metal in condition H1150. It is also observed that the TE values are not significantly affected by any of the PWHTs, while the different PWHTs have only a marginal influence on the RA values. Further, the ST + H1150 PWHT results in higher YS, UTS and TE values but lower RA values than in the as-welded condition and after the direct H1150 PWHT. Compared to the base metal in condition H1150, the RA value of the weldment with/without PWHT is always lower. The true stress/true strain plots (Fig. 9C) show that the strength and ductility of as-welded weldments and weldments subjected to PWHT by the two procedures are almost similar. However, the ST + H1150 PWHT results in marginally superior strength and ductility. However, Table 3 shows that the ST + H1150 PWHT results in a lower RA value compared to that in the as-welded condition and after the direct H1150 PWHT, while all their other tensile properties are almost similar. Also, in the as-welded condition, the YS is lower than after the direct H1150 PWHT, while all their other tensile properties are almost similar. However, except for YS and RA values, there is only a marginal variation in the other tensile properties of condition H1150 weldments irrespective of the PWHT given. Hence, it would be possible to avoid PWHT of 17-4PH SS weldments when they are in the overaged condition, as their strength and ductility in the as-welded condition is satisfactory.

## Practical Implications

From the preceding discussion, it follows that for use in condition H900, condition A weldments are to be heat treated using the ST + H900 PWHT procedure. Also, a better combination of properties is obtained by direct H1150 PWHT of condition A weldments, and by the ST + H1150 PWHT of condition H1150 weldments. Further, the optimal combination of tensile properties (*i.e.*, higher strength and ductility) for using 17-4PH SS weldments in condition H1150 is achieved by welding in condition A and then subjecting it to the direct H1150 PWHT procedure. It is, however, reported that 17-4 PH SS weldments are more prone to severe HAZ corrosion in the as-welded condition (Ref. 18). Thus, the corrosion resistance required for a particular service condition should determine the necessity for PWHT of this weldment and not its mechanical properties alone, and for corrosion resistance, the ST + H1150 PWHT route is to be adopted.

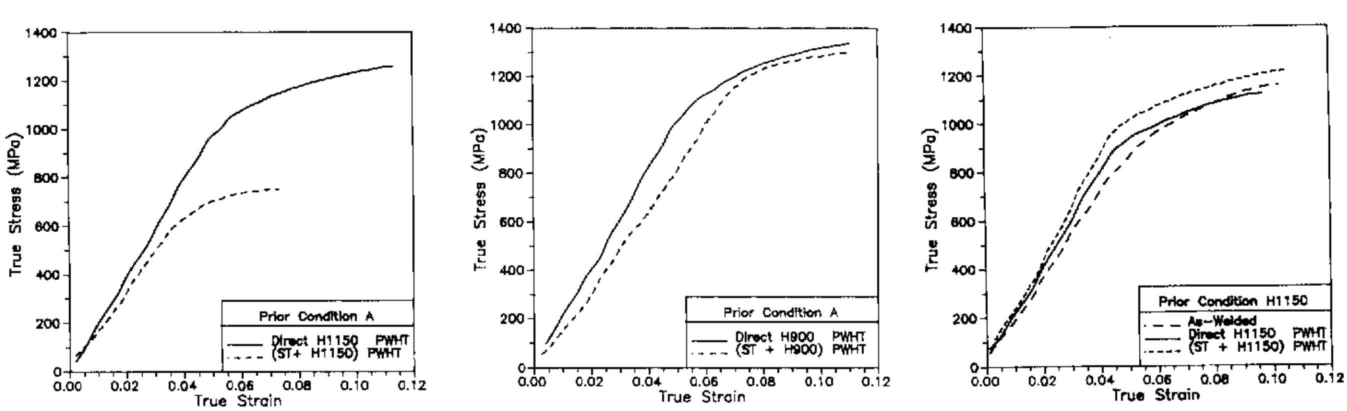
## Conclusions

1) The HAZ adjacent to the weld interface contains retained austenite and retransformed martensite. The retransformed martensite is aged or overaged to varying degrees depending on the PWHT given.

2) The PWHT involving the ST step prior to aging/overaging reduces the amount of retained austenite in the HAZ. However, the condition H1150 PWHTs result in an increase in the amount of retained austenite in the HAZ.

3) Compared to the direct PWHT procedures, the ST step prior to H900 PWHT effectively reduces the hardness variation across the HAZ, while that prior to the H1150 PWHT eliminates the hardness trough in the HAZ adjacent to the weld interface.

4) Optimum room-temperature tensile properties for use in condition H900 is obtained on using the ST + H900 PWHT procedure after welding in condition A. To obtain optimum tensile properties for use in condition H1150, the direct H1150 PWHT route is to be used if




A B C

*Fig. 9 — True stress/strain curves at room temperature. A — Condition A weldments after condition H900 PWHT; B — condition A weldments after condition H1150 PWHT; C — condition H1150 weldments.*

welded in condition A, while no PWHT needs to be given if welded in condition H1150 (note, however, that corrosion resistance is not optimum in the H1150 as-welded condition). The optimal combination of tensile properties for use in condition H1150 is achieved by welding in condition A and using the direct H1150 PWHT procedure.

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