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Use of Postweld Heat Treatments to Improve Ductility in Thin Sheets of Ti-6Al-4V

A postweld triplex heat treatment cycle produces a more ductile grain structure, resulting in elongation values up to 7%

BY K. BORGGREEN AND I. WILSON

ABSTRACT. Manual GTA welds made on 0.6 mm (0.024 in.) thick sheets of the 6Al-4V titanium alloy have only about 2% elongation in the as-welded condition. This is attributed to the presence of an acicular type of alpha structure mixed with a precipitated martensitic phase rather than to the large size of the prior beta grains. The formation of martensite is linked with a critical cooling rate of about 9°C/s (16°/F) which is nevertheless exceeded by these thin sheet metal welds. A more ductile internal grain structure was produced by the use of a postweld triplex heat treatment cycle which resulted in elongation values of up to 7%.

Introduction

The Intersecting Storage Ring (ISR) machine at CERN (European Centre for Nuclear Research) is a unique facility that enables physicists from all over the world to study high energy collisions between two intense proton beams circulating in opposite directions in an intersecting system of pipes. The requirement of good clean physics conditions for this machine means that these pipes must contain on the average a vacuum of 5×10^{-12} torr which is obtained by pumping the system for 24 hours (h) at 300°C (572°F).

The physics interest is centered around the eight intersection regions where the colliding proton beams produce particles which must be detected and analyzed. Both these procedures are hampered by the fact that all emerging particles and electro-magnetic radiation must pass through the vacuum retaining wall; there is, therefore, a fundamental need to make this as transparent as possible. As a general rule, this transparency to emerging particles is reflected in the strength-to-weight or stiffness-to-weight ratio of the metal from which the vacuum pipe is made.

The material requirements are in many ways similar to those required in the aeronautical and aerospace industries. For these reasons, the ISR has invested a certain development effort in the fabrication of titanium vacuum chambers for use in the experimental intersection regions. An example is shown in Fig. 1. By adopting the corrugated form shown for the central part of the chamber, a high ratio of stiffness to sheet metal thickness is obtained. In this particular case, the chamber is

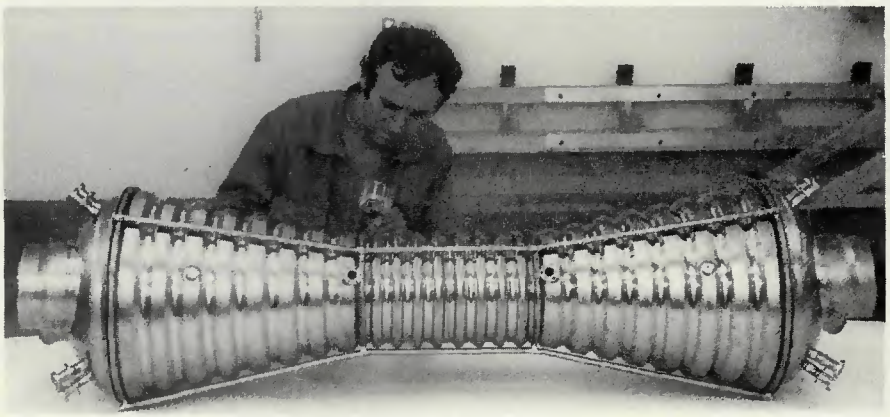

Fig. 1—A prototype titanium vacuum chamber for experimental intersection regions

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made from pure titanium, and the central section has a nominal thickness of 0.2 mm (0.008 in.).

For elliptical chambers such as shown in Fig. 2, the stiff corrugated form should also incorporate high strength. The titanium alloy 6Al-4V was, therefore, selected for these applications because of its high temperature strength properties, its reputation for weldability and because it can be superplastically formed (forming of the corrugated shapes often requires 40% elongation).

It should be noted that due to the complexity, size and uniqueness of these vacuum chambers, they have to be welded manually under a protective atmosphere in a glove-box and that the use of automatic or semi-automatic processes is in general excluded.

The various techniques associated with the actual welding of titanium are now well known[1] and with the availability of modern glove-box facilities, as long as the welds are cleaned correctly, contamination should no longer be a problem. It was found, however, that welds made on both the 0.7 mm (0.03 in.) superplastically formed chambers and the as-received sheet metal were somewhat brittle (<2% elongation) and were therefore unable to meet the ISR weld specification which requires >4% elongation. A similar result was obtained many years ago by Kohn *et al.* for 1 mm (0.04 in.) thick sheets.[2]

Recent investigations[3-5] have resulted in a better understanding of microstructure-mechanical property relationships in alpha-beta titanium alloys. These studies have shown that the as-welded mechanical properties of these high strength alloys are often poor but that the required characteristics can be obtained by the use of postweld heat treatments which control the decomposition and precipitation reactions responsible for the poor weld performance. This technique has been used in particular on the Ti-6Al-6V-2Sn alloy to improve ductility and fracture toughness.[6]

The internal structure of as-welded Ti-6Al-4V consists of fine acicular α needles dispersed in the prior β grains. Dispersions of this kind are generally associated with high strength and low ductility. The fine α needles can be changed to a more ductile plate structure by a high temperature anneal.

Ti-6Al-4V welds may contain a brittle martensitic phase, α', if cooled too quickly. This phase is not visible under an optical microscope but may be detected and differentiated from the α phase by X-ray diffraction techniques.[7] The critical cooling rate, below which no α' is formed, is around 9°C/s,[8] (16°F/s); this is well below the cooling rate of welds made on thin metal.

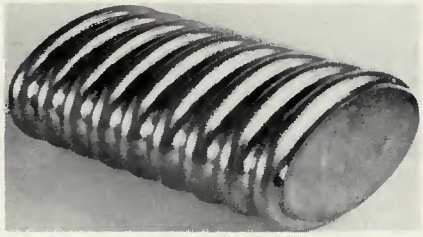


Fig. 2—A prototype piece of elliptical vacuum chamber

The formation of this unwanted phase can be avoided by preheating the component before welding to produce slower cool-down rates. This technique is dependent on part geometry and has in consequence received little attention. A more practical approach is to dissolve the α' back into solution after welding using a heat treatment cycle that incorporates a sufficiently slow cool-down rate.

The influence of grain size and shape on ductility is not well understood. Large reductions in ductility have been reported for large-grained alpha-beta alloys in sheet form having equiaxed structures although no reduction was observed when the structure was acicular.[9] It may be that the effect is small until the grain size becomes large compared with the sheet thickness. When welding thin sheets manually, it is not uncommon to find a single columnar prior beta grain across the thickness as shown in Fig. 3A. This provides a continuous grain boundary path for a crack to follow, but this concerns more structural reliability and fracture toughness than ductility.

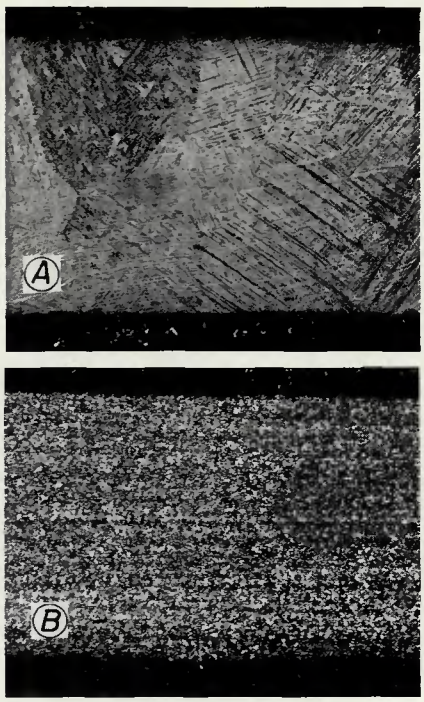


Fig. 3—Microstructure of: A—the as-welded metal; B—the as-received base metal. ×120 (reduced 50% on reproduction)

Electron beam welds have a finer grain size and are more ductile than manual welds. This would suggest that some gains in ductility may be possible by the use of grain refinement techniques. One technique that changes the pattern and rate of cooling with time and, therefore, acts in a positive way on both grain size and shape is the slow pulsed current arc welding technique.[10] Other techniques which favor nucleation by inducing vibrations in the weld pool exist and have given good results.[11-12] A grain refinement of up to 20 times has been obtained by inoculating Ti-6Al-6V-2Sn welds with yttrium.[14] This technique was very effective in increasing fracture toughness but produced a decrease in ductility.

Pure titanium as filler metal gives an improved as-welded ductility which is only slightly increased by subsequent heat treatment.[2] Ductility increases of this sort are, however, paid for by a reduction in strength.

It was concluded, therefore, that for these thin sheets of Ti-6Al-4V, the most promising way of obtaining an improved ductility at more or less constant strength would be to use post-weld heat treatments. The results of such an approach are given in this paper.

Experimental Details

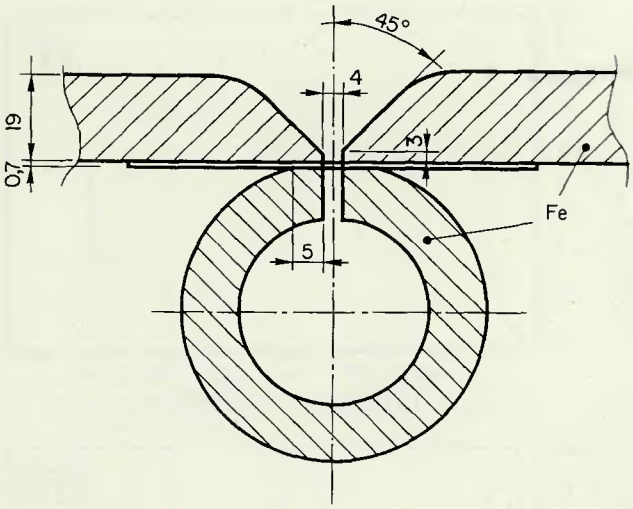
Material

The material used for the investigation was a commercially available Ti-6Al-4V alloy, 0.6 mm (0.024 in.) thickness, in the annealed, as-received condition. The exact chemical composition (wt-%) was: Fe—0.1; C—0.01; N—0.02; O—0.16; H—0.0125; Al—6.2; V—4.3; Ti—balance.

Weld Specimens

Test strips, each 130 × 40 mm (5.12 × 1.57 in.), were guillotined from the as-received sheets, milled on one edge, and degreased with perchlorethylene vapor. To assure maximum cleanliness, they were then washed with detergent in an ultrasonically agitated bath and water-rinsed before being dried with warm nitrogen. Immediately prior to welding, the surface oxide layer was removed from the vincinity of the weld with a clean stainless steel brush.

The machined edges of two of the strips were clamped end to end and butt-welded together in the soft-iron jig shown in Fig. 4 using a classical GTA welding process without filler



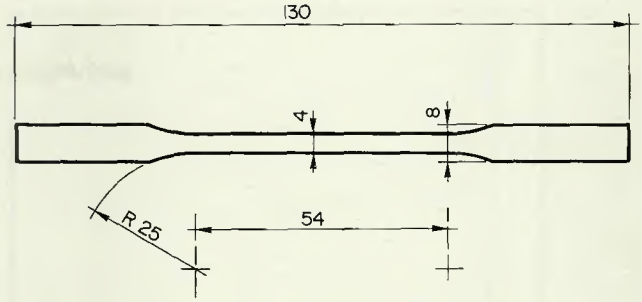


Fig. 5—Geometry of tensile test specimen used for longitudinal tests

Fig. 4 (left)—Soft iron jig used for manual GTA welding

metal. Argon gas was used for both the torch and the backing protection. Although the welds were done by hand, the welder was able to use the side of the jig to guide his torch and maintain a constant electrode distance of about 0.5 mm (3/16 in.).

With this jig, it was found that the welds done outside the glove-box were of the same quality as the welds done inside. Such a state of affairs is only possible when the welds are straight and simple, and use can be made of simple jigs for clamping and inert gas protection of both sides of the weld.

All test samples were, therefore, welded outside the glove-box. This resulted in a faster preparation rate as well as a better overall control of the welding process. All welds were subjected to both a visual examination to make sure they were a brilliant silver color and an X-ray examination to screen them against inclusions and porosity.

Heat Treatment

The as-welded strips were subjected to a triplex postweld heat treatment similar to that developed for the Ti-6Al-6V-2Sn alloy.[5] Two types of furnace were used, one giving a cooling rate of about 200°C/h (360°F/h) (FCA) and the other 2500°C/h (4500°F/h) (FCB). The level of vacuum in both furnaces was better than 2×10^{-5} torr. Between each cycle of the triplex treatment the specimens were allowed to cool to room temperature.

The following variations of the triplex heat treatment were investigated:

Treatments with Variable Temperature: 730°C (1346°F)/6h/FCA +

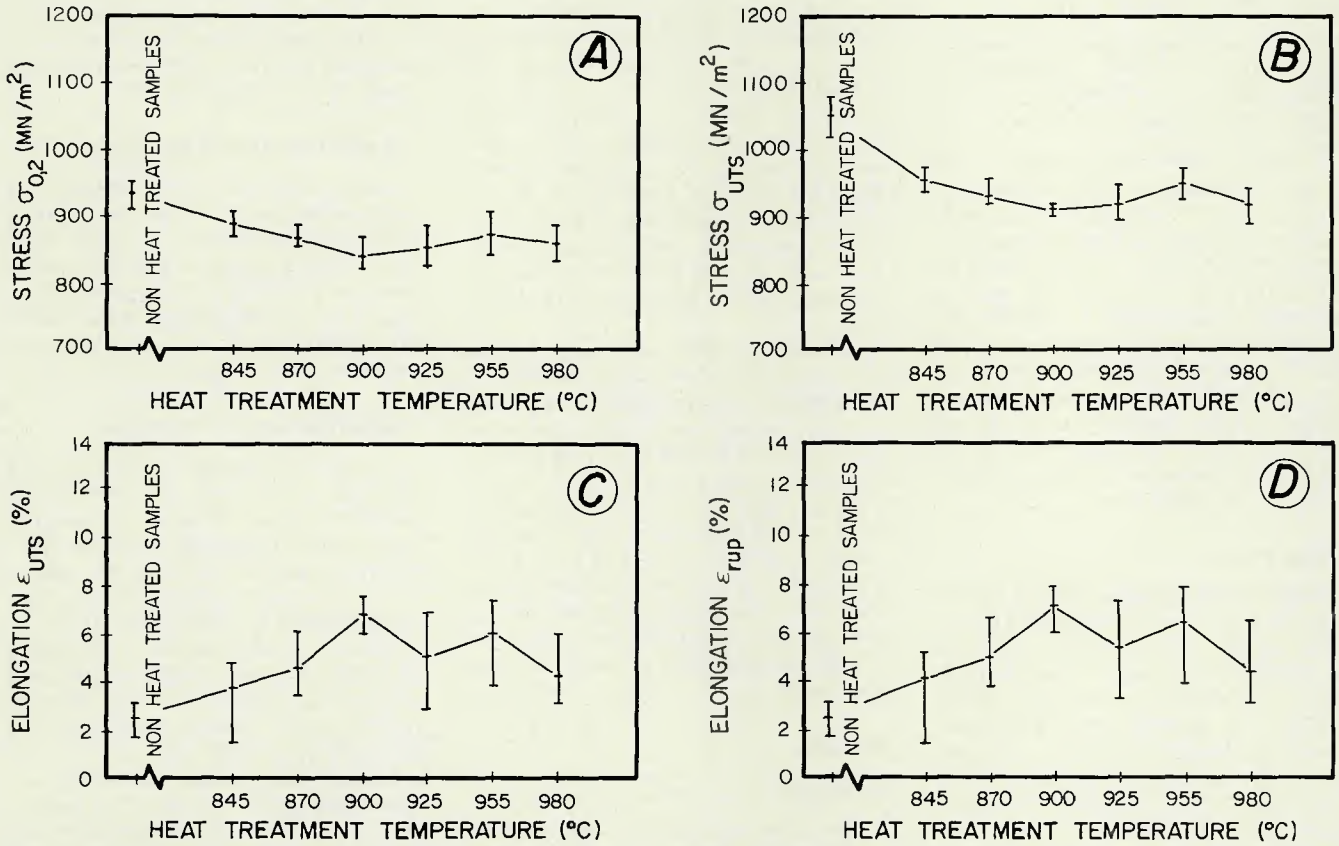


Fig. 6—Mechanical properties of the welded zone for different second cycle heat treatment temperatures: A—yield stress; B—ultimate tensile stress; C—elongation at maximum load; D—elongation at rupture

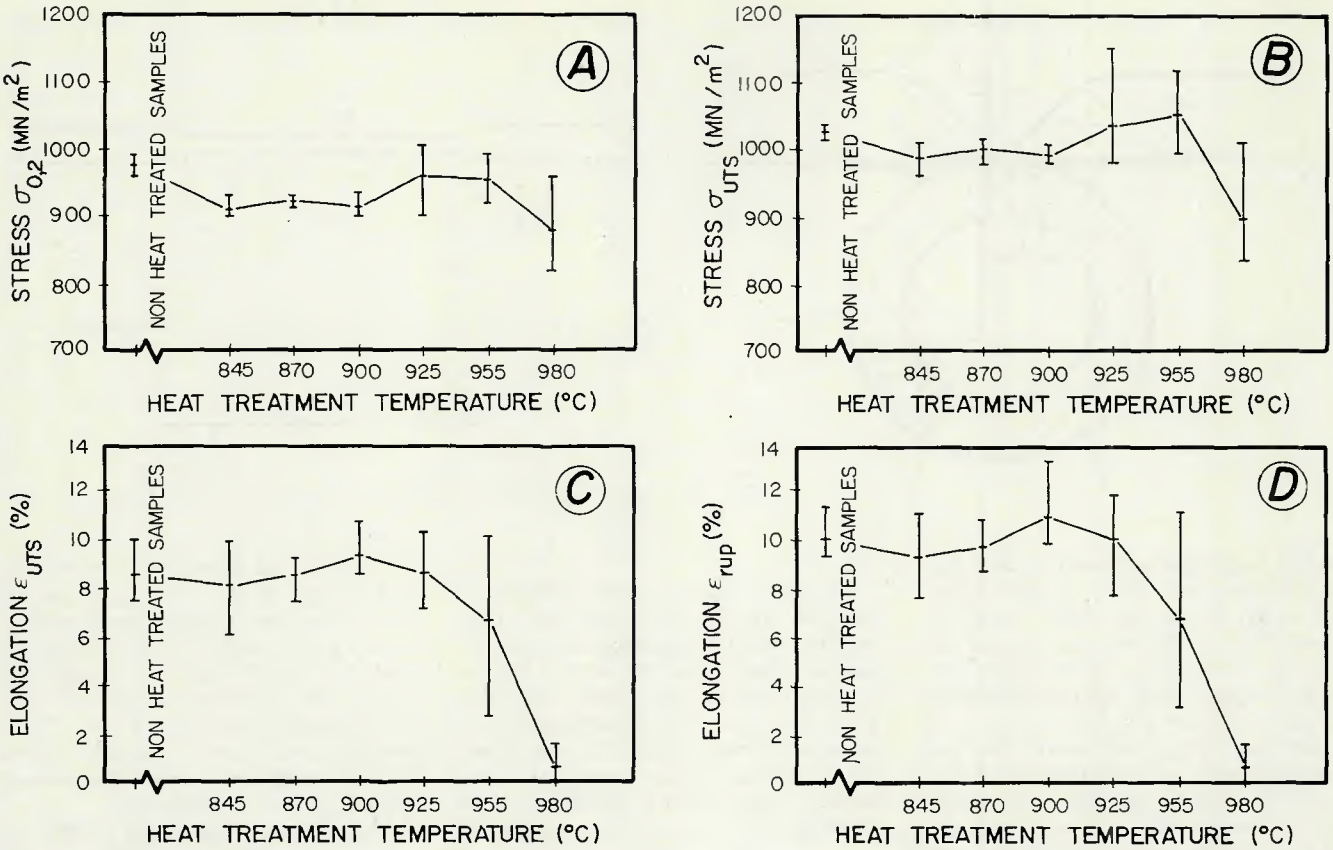


Fig. 7—Mechanical properties of the base metal for different second cycle heat temperatures: A—yield stress; B—ultimate tensile stress; C—elongation at maximum load; D—elongation at rupture

845-980°C (1553-1796°F)/3h/FCA or FCB + 730°C (1346°F)/3h/FCA.

Treatments with Variable Time at Temperature: 730°C (1346°F)/6h/FCA + 900°C (1652°F)/1-24 h/FCB + 730°C (1346°F)/3h/FCA.

It was decided to investigate this range of temperatures because the related alloy Ti-6Al-6V-2Sn responds well to a similar triplex treatment at 927°C (1701°F), 25°C (45°F) below its β-transus. For the Ti-6Al-4V with a β-transus of about 980°C (1796°F), the corresponding temperature would be 955°C (1751°F). This, however, is its superplastic temperature where deformation under self-weight might be important and therefore there is an interest in finding if possible a somewhat lower temperature.

Tensile Testing

In the absence of a standard tensile specimen for thin sheet welded metal, a tensile test specimen with a width of 4 mm (0.16 in.) was chosen—Fig. 5. Such a specimen is composed almost entirely of heat-affected and fused metal and will, therefore, give results which are uninfluenced by the base metal properties. The effect of specimen width on measured characteristics is discussed in the Appendix.

This form of specimen was conveniently machined from the heat-treated strips by spark erosion. The rolling direction (RD) unless otherwise specified was taken parallel to the weld (IIRD).

Measurements of yield strength ($\sigma_{0.2}$) ultimate strength (σ_{uts}) and elongations (ϵ_{uts}, $\epsilon_{rupture}$) were made for a fixed rate of straining of 10^{-3}/s. Reference to weld ductility in this work means uniform uniaxial elongation over 50 mm (1.97 in.) of welds running parallel to the testing direction.

All results unless otherwise stated are based on a statistical sample of nine specimens. Arithmetic mean values are given together with a scatter bar indicating the maximum and minimum measurements.

Joint efficiency was determined from simple transverse welded specimens (20 × 6 mm, i.e., 0.79 × 0.24 in.) having the rolling direction parallel to the weld (IIRD) and perpendicular to the tensile direction. In this case results are based on a sample of four specimens.

Results

As-Welded

By optimizing the welding parameters, the elongation of nonheat-treated manual welds was increased from an initial value of less than 2% to about 2.5%. Electron beam welds made on similar samples gave 5%. These values are to be compared with a base metal elongation of about 9%.

Postweld Heat-Treated Specimens

Initial attempts at improving the mechanical properties of the welds by the triplex type of heat treatment were very encouraging, giving an immediate increase of 100% in elongation for a 10% drop in strength. An optimization of the heat treatment procedure was, therefore, undertaken.

Longitudinal Welded Specimens

First investigations showed that no marked difference in mechanical properties was obtained by varying the furnace cooling rate in the second part of the triplex treatment. All subsequent treatments, therefore, only included the fast cooling rate (FCB).

Figure 6 compares the mechanical properties of heat-treated samples and as-welded samples for six different heat treatment temperatures between 845 and 980°C (1553 and 1796°F). The corresponding results for the base metal are given in Fig. 7. Figure 7 shows that the heat treatment has no detrimental effect on the base metal properties for temperatures well be-

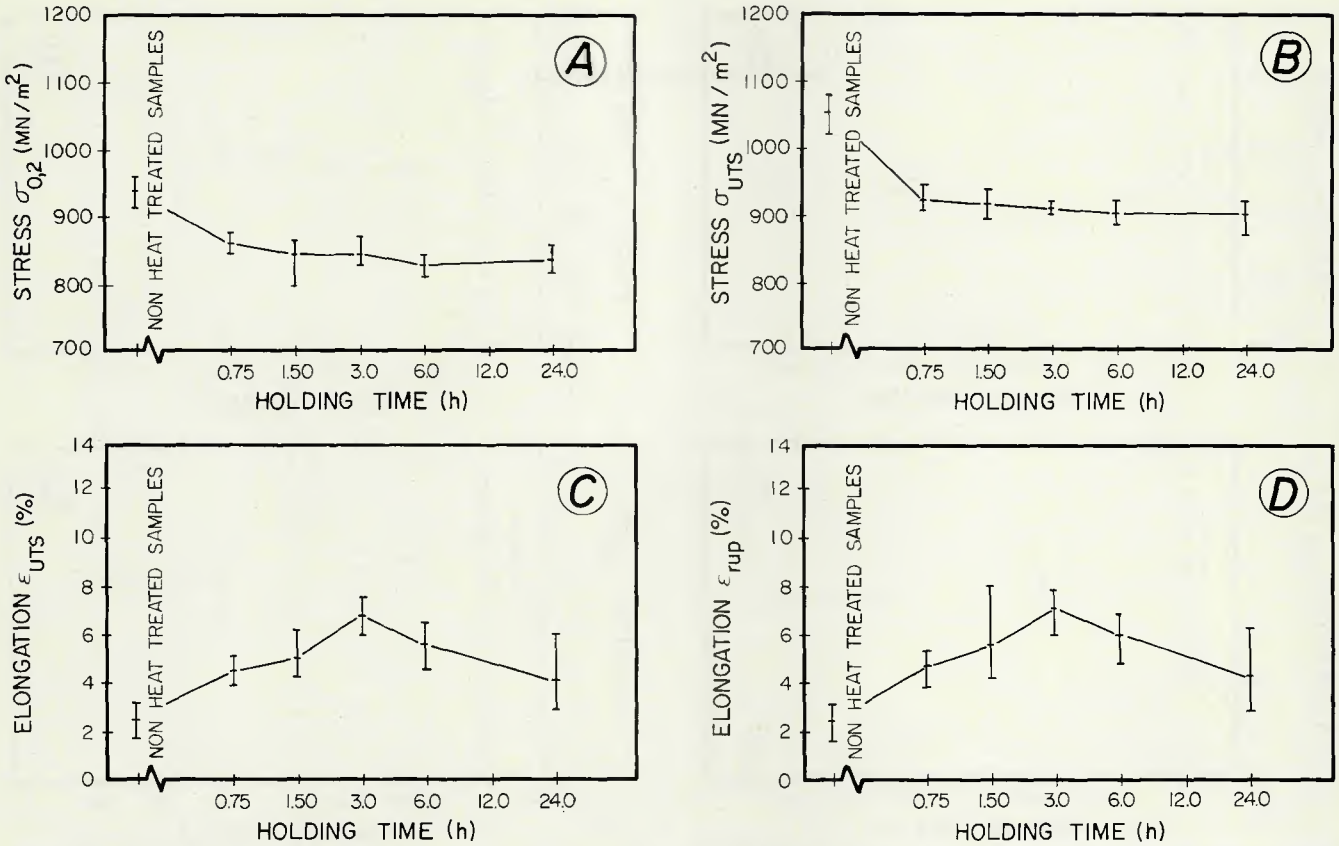


Fig. 8—Mechanical properties of the welded zone for different second cycle holding times at 900°C: A—yield stress; B—ultimate tensile stress; C—elongation at maximum load; D—elongation at rupture

low the β-transus. Higher temperatures, however, produce a small drop in strength and a significant reduction in elongation; this effect is not observed in the welded specimens.

Having investigated various second cycle heat treatment temperatures and finding little difference between the resulting mechanical properties, it was decided to fix the temperature at 900°C (1652°F) for the final series of heat treatments and investigate the effect of holding time. Thus, 900°C was chosen because there appears to be a tendency towards maximum elongation at this temperature; it is also well below the potentially troublesome superplastic and β-transus temperatures.

Five different holding times between 1 and 24 h were investigated. Figure 8 compares the resulting mechanical properties of heat-treated and nonheat-treated welded samples. Increasing time at temperature produces a slight drop in strength with a corresponding increase in elongation which reaches a maximum after 3 h. The behavior of the base metal to these same heat treatment cycles is given in Fig. 9 and shows a slight increase in strength.

Transverse Welded Specimens

Transverse weld tests were made for all the above mentioned heat treatments. All the samples failed in the base metal.

Heat-Treated EB Welded Specimens

Electron beam welded specimens were subject to the following heat treatment cycle to see if any further gains in ductility could be expected from small grained samples: 730°C (1346°F)/6h/FCA + 900°C (1652°F) /3h/FCB + 730°C (1346°F)/3h/FCA.

In the as-welded condition, the mean values for $\sigma_{0.2}$, σ_{uts}, ϵ_{uts} were found to be 960 MN/m², 1080 MN/m² and 5.0% respectively. After heat treatment these values become 870 MN/m², 940 MN/m² and 9.2%. This weld elongation is equal to that of the heat-treated base metal. Transverse weld test specimens for both as-welded and heat-treated samples all failed in the base metal.

Effect of Rolling Direction

Although the measured strength of heat-treated base metal samples was found to be greater than that of the heat-treated welded samples, all transverse weld test specimens failed in the base metal. This apparent discrepancy is explained by the anisotropic behavior of these thin sheets.

Table 1 gives the result of a series of tests made on specimens with the welds perpendicular to the rolling direction ($\perp$ RD). It can be seen that the strength of the heat treated base metal falls below that of the heat-treated weld when tested in a direction perpendicular to the rolling direction and that a failure in the fused zone is only found in heat-treated transverse welded specimens when the tensile direction is parallel to the rolling direction.

Microstructures

As-Welded

The two-phase structure of the as-received sheet metal consists of small equiaxed α grains of about 5 μ diameter within a β matrix. This is shown in Fig. 3B. Manual welds made on thin sheets of Ti-6Al-4V require high heat inputs which produce large grained structures with single prior β-grains traversing the thickness as shown in Fig. 3A. With respect to the base metal this represents a grain growth of more than 100. In comparison, low heat input electron beam welded specimens have typically more than 4 grains across the thickness.

Heat Treatment

An intuitive explanation of the role

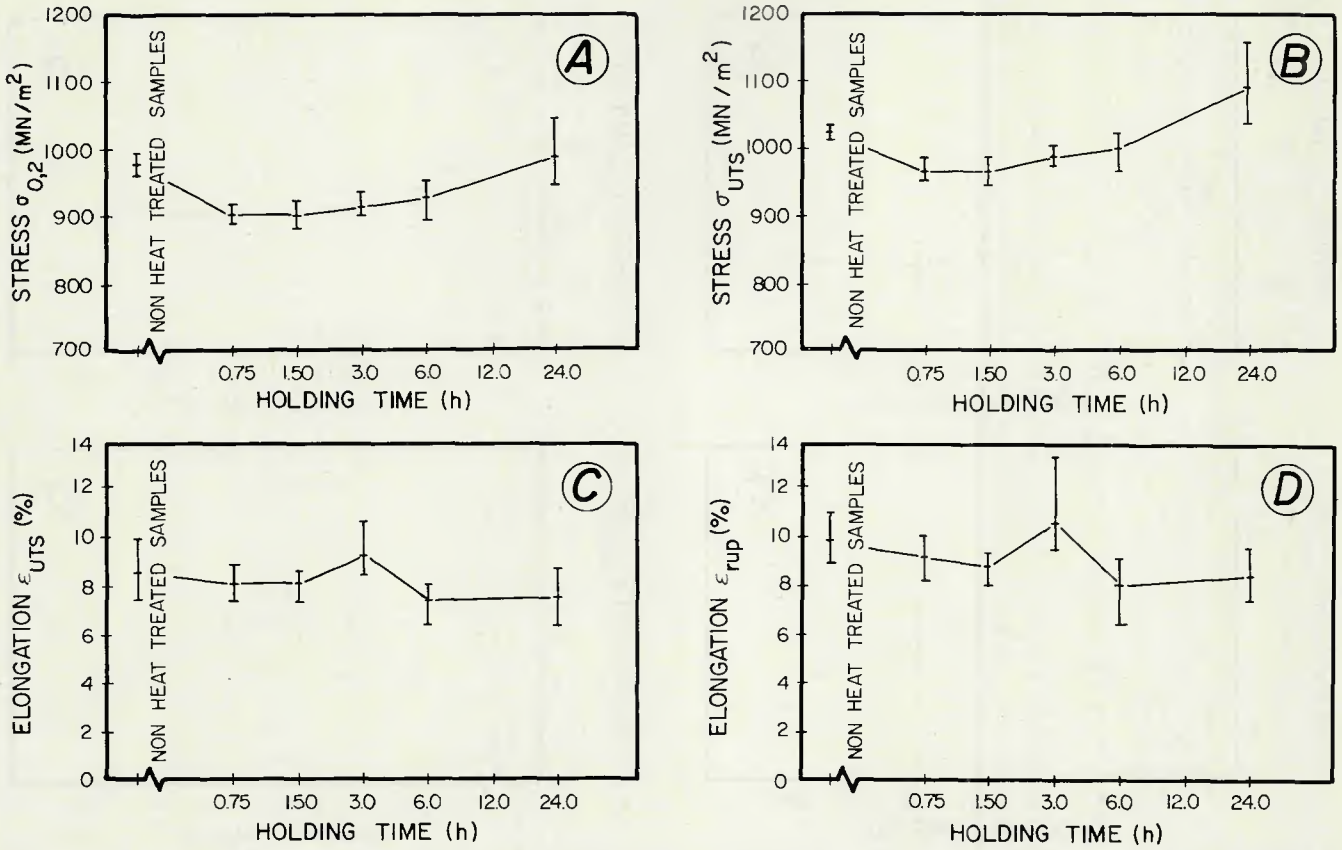


Fig. 9—Mechanical properties of the base metal for different second cycle holding times at 900°C: A—yield stress; B—ultimate tensile stress; C—elongation at maximum load; D—elongation at rupture

played by each part of the triplex heat treatment cycle adopted for this study has been given in detail elsewhere.[6] This is summarized as follows:

The first part allows the α' and retained β to decompose into $\alpha + \beta$ at a temperature which favors α nucleation rather than growth; the second part uses the previously formed nucleation sites to encourage globularization of the α phase at a temperature just below the β-transus; the β-rich structure on cooling produces fine α platelets which are subsequently coarsened during part three to obtain the mechanical properties desired. It is worth noting that the prior β-grain size in the fused zone is not changed by this heat treatment.

Effect of Varying the Temperature

The changes in microstructure in the fused zone obtained for three of the heat treatment temperatures investigated are shown in Fig. 10A. The structure in all cases consists of α plates densely packed in a β matrix with little or no evidence of globularization. With increasing heat treatment temperature, the resulting plate structure is seen to coarsen rapidly with a tendency at the highest temperatures for the larger plates to break up into shorter ones.

The corresponding microstructures for the base metal are given in Fig. 10B. The laminated structure of the as-received sheet (Fig. 3B) is gradually replaced by a coarsening equiaxed grain structure as the temperature increases. For the highest temperature, there is a marked tendency towards the formation of α plates which are created during the slow cool down of the β phase.

It is interesting to see in Fig. 11A that the β transformation in the outer surface layers has been suppressed at the high heat treatment temperatures by the presence of an α stabilizing element. Oxygen, nitrogen and carbon are α stabilizing but their contents after heat treatment were not found to be above the specification values of 0.20%, 0.05% and 0.08% respectively. However, it was found that this surface effect could be completely eliminated as shown in Fig. 11B by placing the samples during the heat treatment cycle between alumina plates which are known to act as a getter for oxygen.

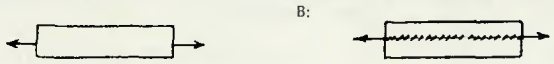
Effect of Varying Time at Temperature

The changes in microstructure in the fused zone obtained by varying the holding time during the second part of the triplex heat treatment are shown in

Table 1 —Results of Tests on Specimens with Welds Perpendicular to Rolling Direction

Specimen	Sketch[a]	RD	$\sigma_{0,2}$, MN/m²	σ_{uts}, MN/m²	ϵ_{uts}, %
			Untreated specimen:		
Base metal	A	↔	970	1020	8.5
Base metal	A	↕	920	960	7.1
GTA weld	B	↔	940	1050	2.5
GTA weld	B	↕	950	1080	2.3
			Heat treated specimen:		
Base metal	A	↔	910	990	9.2
Base metal	A	↕	780	860	8.6
GTA weld	B	↔	840	910	6.8
GTA weld	B	↕	820	880	5.4

[a] A: B:



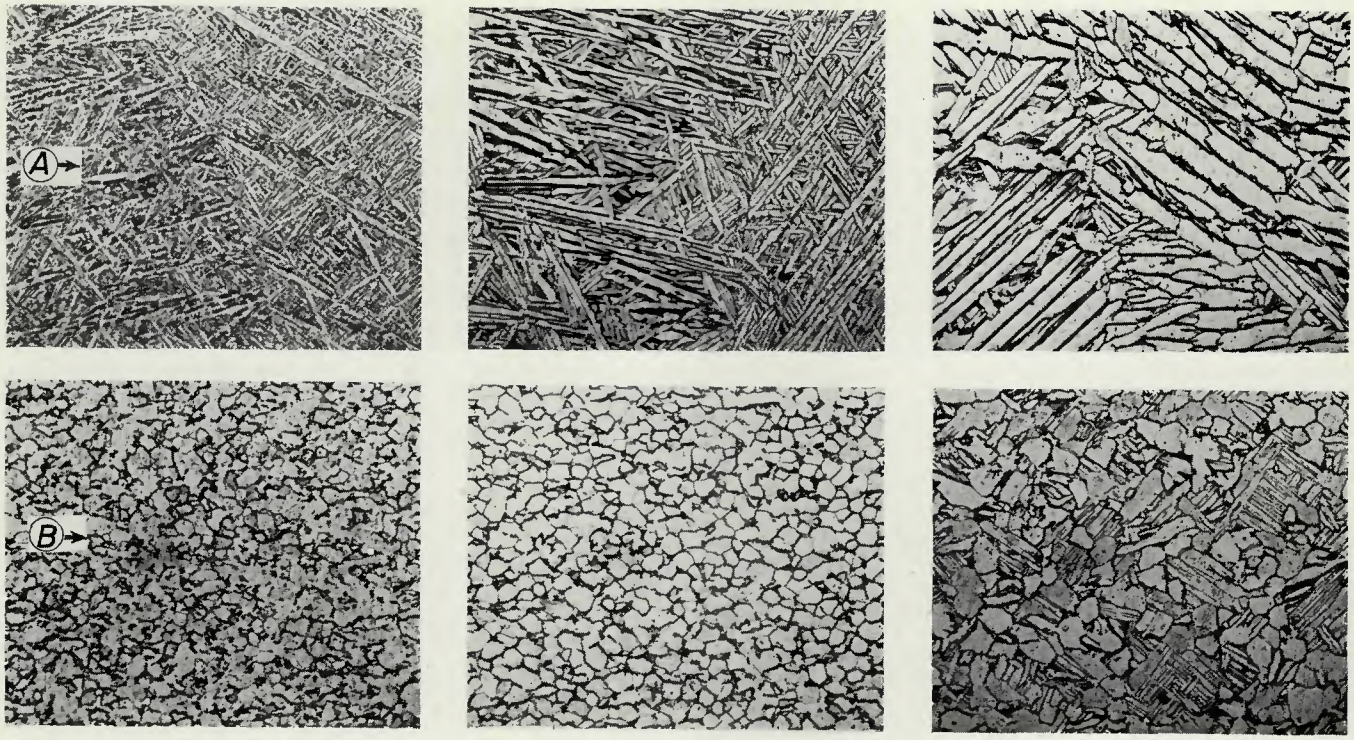


Fig. 10—Microstructures of for three different second cycle heat treatment temperatures (left to right—845°C, 900°C and 955°C): A—fused zone; B—base metal. ×500 (reduced 50% on reproduction)

Fig. 12A. Again, there is no evidence of globularization and the α-plate structure simply coarsens as the holding time increases.

The corresponding microstructure for the base metal, given in Fig. 12B shows essentially the same evolution with time as obtained with temperature. Grain boundary α, produced by a similar heat treatment in the Ti-6Al-6V-2Sn alloy, is not observed in any of the Ti-6Al-4V samples.

Hardness Measurements

The microhardness profiles shown in Fig. 13 were determined for one specimen in the as-welded condition and after the first and final parts of the following heat treatment cycle: 730°C (1346°F)/6h/FCA + 900°C (1652°F) /3h/FCB + 730°C (1346°F)/3h/FCA.

In the as-welded condition, the fused and heat-affected zone has a hardness of 410 HV compared with 360 HV for the base metal. This zone is about 4 mm (0.16 in.) wide; it therefore accounts for all the material contained in the tensile test specimens used in this work.

After the first heat treatment there is a noticeable drop in hardness in the fused and heat-affected zone due to the disappearance of the α′ phase. At the end of the complete heat treatment cycle this same zone has a hardness of about 340 HV. This further drop in hardness is associated with the creation of a coarser α-plate structure. The base metal hardness increases slightly after heat treatment attaining a value of about 370 HV.

Discussion and Conclusion

The triplex type of postweld heat treatment used in this work was originally developed to increase the fracture toughness of the Ti-6Al-6V-2Sn alloy.[6] It was reported that the fracture toughness parameter K_Q, the crack propagation energy W/A and the elongation for the welded zone increased by factors of 2, 10 and 10 respectively. For the Ti-6Al-4V alloy which has a similar heat-treated structure, it would be expected that the increase in ductility produced by this same sort of triplex treatment would be accompanied by a corresponding increase in fracture toughness. For our particular application this is of little interest and has, therefore, not been investigated because already in the as-welded condition the Ti-6Al-4V alloy is known to have good fracture toughness characteristics.

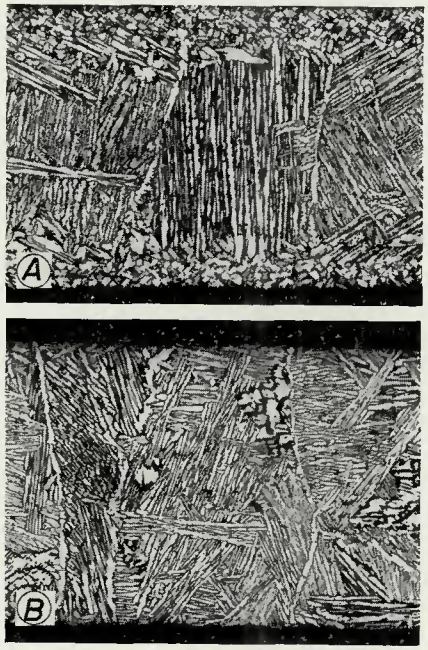


Fig. 11—Crosssection of a heat treated base metal sample; 980°C second cycle heat treatment temperature: A—with an α-stabilized surface layer; B—without the α-stabilized surface layer. ×120 (reduced 50% on reproduction)

Although these triplex treatments have proved very effective in increasing ductility, it is not excluded that a similar result could not be obtained more simply. As an example, a single cycle treatment was tried, i.e., 870°C (1598°F)/2 h/FCA. It was found that the resulting mechanical properties were similar to those obtained for the 870°C triplex treatment. Whether this result is typical or not has to be investigated. For Ti-6Al-6V-2Sn welds, this same single cycle was investigated and found to be less effective than the triplex treatment.[6]

After the 980°C (1796°F) triplex treatment, the base metal elongation fell drastically from 10% to 1%. The same treatment, however, only produced a slight decrease in weld elongation. This was linked to the fact that, although the resulting internal struc-

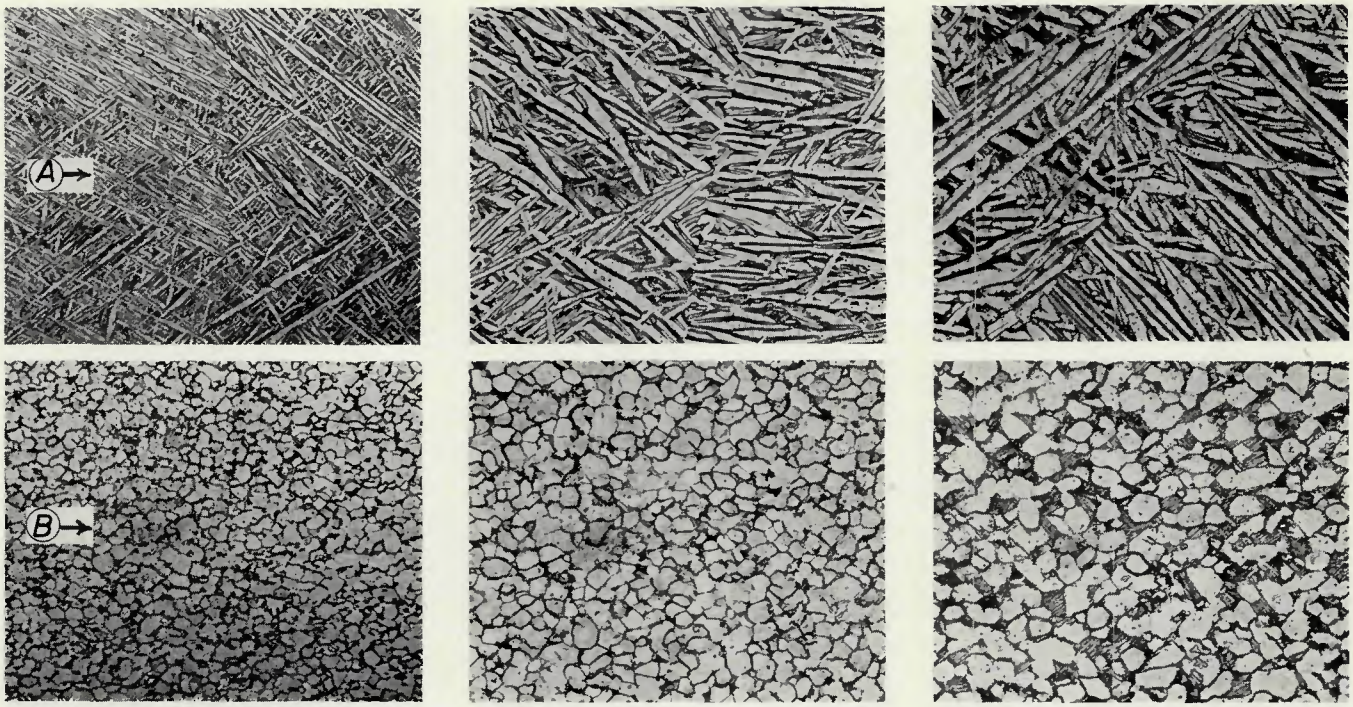


Fig. 12—Microstructures of for three different second cycle holding times (left to right—1, 6 and 24 h) at 900°C; A—fuse zone; B—base metal. ×500 (reduced 50% on reproduction)

tures were similar, the base metal had undergone excessive grain growth resulting in a grain size which was large compared to the thickness. This suggests the presence of some kind of grain size effect.

Part of the benefit of the high temperature heat treatment under vacuum may be due to the elimination or redistribution of small amounts of contaminating elements in the welded zone. Hydrogen is, for instance, easily pumped out of a contaminated weld at these high temperatures. This is interesting, because this means that the microplasma technique of welding can be used even for titanium. Some tests were successfully done with this valuable technique using an argon–7% hydrogen shielding gas mixture. The elongation was found to increase from an as-welded value of 1.3% to a post-weld heat-treated value of 5.2%.

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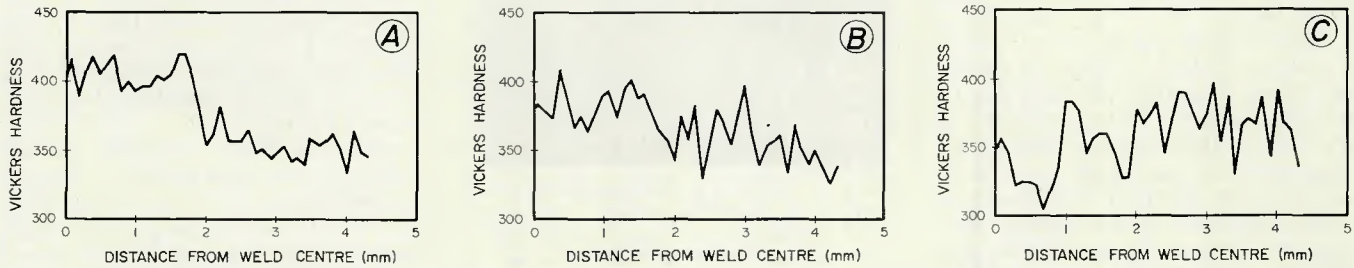

Fig. 13—Hardness profiles at the welded zone in: A—the untreated condition; B—after the first part of the triplex treatment; C—after the complete triplex treatment. 100 g load

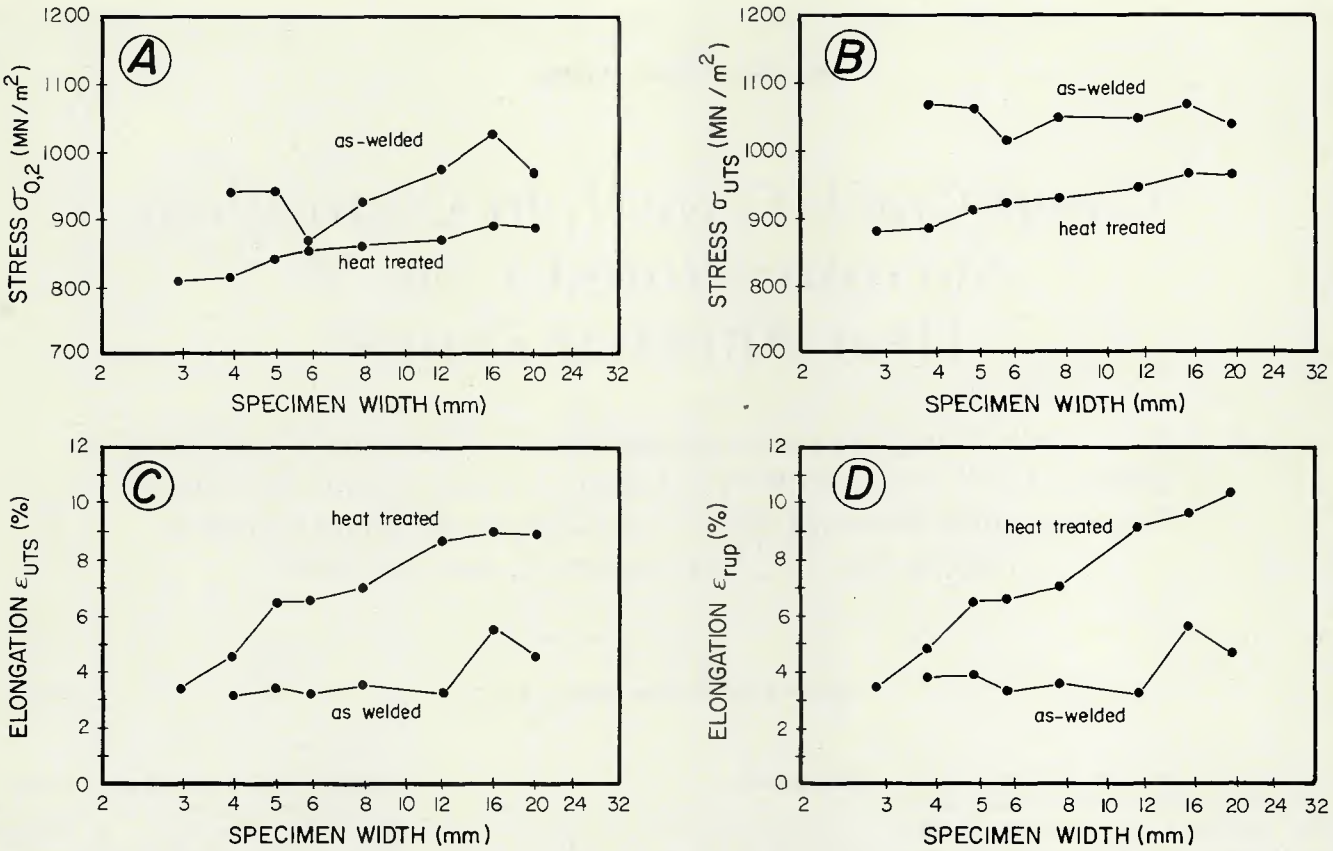

Fig. 14—Mechanical properties of the welded zone for different specimen widths: A—yield stress; B—ultimate tensile stress; C—elongation at maximum load; D—elongation at rupture

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Appendix

A few specimens taken from a different batch of material to that used for the main part of this work were used to investigate the effect of specimen width on the measured characteristics of longitudinally welded samples. In this case, the as-welded samples had an elongation of around 3.5%.

The value of $\sigma_{0.2}$, σ_{uts} and ϵ_{uts} measured for the different specimen widths are given in Fig. 14 for as-welded samples and postweld heat-treated samples (triplex treatment with a second cycle of 900°C, i.e., 1652°F, for 3 h). Although these results are based on a single measurement, the tendency is for all properties except as-welded elongation to approach those of the base metal for specimen widths of 20 mm (0.79 in.).

This dependency of measured elongation on percentage specimen base metal content may explain in part the higher values obtained for electron beam welded specimens of the same width.

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