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SURVEY OF WELDING PROCESSES FOR FIELD FABRICATION OF 2¼ Cr–1 Mo STEEL PRESSURE VESSELS

Topical Report, Subtask 1.2

By
G. E. Grotke

April 1980

Work Performed Under Contract No. AC01-78ET13511

Westinghouse R&D Center
Pittsburgh, Pennsylvania

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SURVEY OF WELDING PROCESSES FOR FIELD FABRICATION OF 2 1/4 Cr-1 Mo STEEL PRESSURE VESSELS

G. E. Grotke

April 1980

Topical Report, Subtask 1.2

Prepared for:
United States Department of Energy
Contract No. EF-78-C-01-2771

**Westinghouse R&D Center
1310 Beulah Road
Pittsburgh, Pennsylvania 15235**

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PREFACE

This report was prepared in performance of the Department of Energy Contract EF-78-C-01-2771 "Development of Automated Welding Processes for Field Fabrication of Thick Walled Pressure Vessels". This project is administered through Mr. S. J. Dapkunas, Office of Advanced Research of Technology with responsibility assigned to the Oak Ridge Operations Office, Mr. E. E. Hoffman and technical follow by the Oak Ridge National Laboratory, Mr. R. A. Bradley.

The Westinghouse Electric Company Program Manager is Mr. E. P. Loch of the Tampa Division with technical management provided by Mr. U. A. Schneider, Tampa Division, and Mr. G. G. Lessmann, Research and Development Center.

INTRODUCTION

This review has two purposes:

- A survey of the state of development of welding processes suitable for on-site fabrication of massive thick-wall coal-gasification pressure vessels requiring all-position welding.
- A review of the welding metallurgy of A387 Grade 22, Class 2 (2 1/4% Chromium-1% Molybdenum) steel.

These objectives are not separate; they entwine. The complications that arise in fabricating heavy-section weldments are well known. There is much less appreciation for the metallurgical/welding complications involved. Complications that originate with the production of thick plates or forgings, compounded, to a greater or lesser degree, by the welding process used, can result in unanticipated difficulties during post-weld heat treatment or in service:

1. In order to meet minimum mechanical properties in thick sections, the chemical compositions of steels are enriched. If the basic composition is susceptible to heat-affected-zone embrittlement, cold cracking, or stress-relief cracking, the severity will increase.
2. Heavy-section welds experience high restraint and develop high triaxial residual stresses. Design becomes more critical. This problem is complicated by the fact that when high-strength steels are employed, the mechanical property requirements for the steel and the weld are correspondingly higher.
3. If quenched and tempered steels are involved, there is an upper limit on preheat and on weld heat input.
4. Inspection and repair of welds becomes more difficult.

5. Stress relief of heavy-section weldments involves extremely slow heating and cooling. If a steel is susceptible to stress-relief cracking or to temper embrittlement, the likelihood of difficulty increases.

6. Limitations on both rolling reductions and cross-rolling ratios introduce anisotropic and through-thickness property variations.

The complications that some of the above factors introduce should always be borne in mind. Experiments are seldom conducted with materials, or under conditions, that are actually encountered with massive weldments.

Since the introduction of welding, fabrications have been undertaken with relatively good understanding of the properties of the base metal and with only fair to poor comprehension of the properties of weld deposits and heat-affected-zones. This situation is unfortunate, understandable, and unlikely to change. The 2 1/4% Chromium-1% Molybdenum steels in common with other low-alloy steels are susceptible to various, and somewhat related, forms of embrittlement when exposed to elevated temperatures during post-weld heat treatment or during service. Though information on embrittlement of weld metal and of weld heat-affected-zone microstructures is available, it is far from extensive. Specific information pertinent to thick-section weldments, either from direct tests or obtained under conditions that fully simulate thick-sections is scarce. Therefore, this review has, in considering embrittlement, included consideration of studies of plate material to draw attention to factors and microstructures known to be detrimental that are inherent, but to a degree controllable, in welding. The specialized fields of corrosion and hydrogen embrittlement during long-time service, since they are outside the scope of welding metallurgy, have not been included.

In preparation of this review, over one hundred papers and articles have been evaluated. Those that have been selected for discussion are considered to be comprehensive and best suited to aid in anticipation of the difficulties that may arise in thick-section welding of chromium-molybdenum steel. However, to extend the scope of pertinent information, the review has not been restricted solely to the A387 Gr. 22, Cl. 2 composition. The chromium-molybdenum steels include a variety of compositions having chromium in the range 1/2 to 9% and molybdenum between 1/2 and 1%. These steels were initially used in petroleum refineries to obtain improved resistance to corrosion caused by high-sulfur crude oils. Their oxidation resistance and excellent elevated-temperature strength have since led to further applications in heaters, large pressure vessels, and in steam power generation. Precautions during welding, dictated by composition and thickness, include avoiding hydrogen pickup, preheat, and postheat.

The 2 1/4 Cr-1 Mo composition is one of the most widely employed, with ASTM specifications covering plate, cast, forged, tube and pipe products. ASTM A387, Grade 22, Class 2 plate is usually furnished in the normalized and tempered condition to a minimum yield of 45 ksi and an ultimate strength in the range 75-100 ksi. The specification for the steel permits accelerated cooling to be employed with the permission of the purchaser. In order to obtain the minimum properties in thick plates, A387 Gr. 22, Cl. 2 steel must be quenched and tempered. ASTM A542 covers two classes of identical composition 2 1/4 Cr-1 Mo plate furnished only in the quenched and tempered condition to provide higher strengths than obtainable with A387 Gr. 22, Cl. 2.

The chemical composition determines the hardenability, hardness, toughness, and crack susceptibility of the heat-affected zone and governs the selection of acceptable welding procedures. Differences in strength, as far as welding characteristics are concerned are important only in their influence on weld-metal selection, residual-stress levels, and tolerable stress-relief temperatures. The information on A542 steel is more extensive than that available for A387 Gr. 22, Cl. 2.

In fact, a review of the welding metallurgy of the latter steel restricted to that material would yield little data and ignore much that is pertinent. Consequently, this review draws upon the general information on the 2 1/4 Cr-1 Mo composition.

SUMMARY AND CONCLUSIONS

Any evaluation of fabrication methods for massive pressure vessels must consider several welding processes with potential for heavy-section applications. These include submerged-arc and shielded metal-arc, narrow-joint modifications of inert-gas metal-arc and inert-gas tungsten-arc processes, electroslag, and electron beam. Obviously, these processes do not have equal potential. The advantage and disadvantages of each are discussed in this review.

Electroslag welding can be dropped from consideration for joining of 2 1/4 Cr-1 Mo steel because welds made with this method do not provide the required mechanical properties in the welded and stress relieved condition. The extension of electron-beam welding to sections as thick as 4 or 8 inches (100 or 200 mm) is too recent a development to permit full evaluation. The process has demonstrated promise but, to the present time, the investigations conducted have not been sufficient to prove reliability or to establish that autogenous single or two-pass weld deposits can meet mechanical-property requirements.

The manual shielded metal-arc and submerged-arc welding processes have both been employed, often together, for field fabrication of large vessels. They have the historical advantage of successful application but present other disadvantages that make them otherwise less attractive. The manual shielded metal-arc process can be used for all-position welding. It is however, a slow and expensive technique for joining heavy sections, requires large amounts of skilled labor that is in critically short supply, and introduces a high incidence of weld repairs. Automatic submerged-arc welding has been employed in many critical applications and for welding in the flat position is free of most of the criticism that can be leveled at the shielded metal-arc process. The horizontal and vertical position welding requirements for coal - gasification vessels, however, make it necessary to view the process in a different way. Specialized techniques have been developed that permit submerged-arc welds to be deposited in these positions. But, used in this manner, the applications are limited and the cost advantage of the process is lost.

Narrow-joint gas-shielded welding processes provide the greatest promise for present and future heavy-section fabrication and, primarily because of superior parametric control, low weld defect repair costs, and exceptional weld quality, the narrow-groove hot-wire gas-tungsten arc process provides the greatest advantage. The points supporting this conclusion are summarized in Tables I, II, and III, which present a comparison of the advantages and the disadvantages of the SAW and manual SMAW processes, and the GMAW and HW-GTAW processes in their narrow-joint modifications.

Submerged-Arc Welding

Because of the limitations on the size and weight of components that can be transported, coal-gasification pressure vessels will require on-site fabrication. We may assume that the usual practice of shop-fabrication of large but transportable sub-assemblies of coal-gasification vessels will not be possible since the preliminary vessel dimensions will not permit convenient over-land transportation of full courses. Field fabrication concepts are discussed in a Westinghouse Tampa Division report prepared for the Department of Energy[128]. The report presents concepts for receipt of 120° (2.1 rad) course sectors with pre-machined longitudinal seams to be joined in the vertical position, followed by machining and welding of the horizontal girth joints. If certain components, such as shells, are positioned in the flat position for seam welding, joining by the automatic submerged-arc process would seem the intuitive choice. However, the method proposed has the advantages of presenting six positions for simultaneous welding of the three vertical seams, offering good distortion control, avoiding the massive fixturing required to support a 123 ton (111,600 kg) course, and using a single welding process for all fabrication. Since distortion control on this heavy wall, large diameter sub-assembly is important for the fit-up of the out-of-position successive course welds,

TABLE I

EASE OF AUTOMATION

	Suitability For Automation	Can Be Visually Monitored	Potential For Remote Control	Automatic Control Rating	Equipment Cost	Field Welding Experience
SAW	Yes	No	No	3rd	Moderate	Some
SMAW (Manual)	No	Yes	No	NA	Low	Much
GMAW	Yes	Yes	Yes	2nd	High	Modest
HW-GTAW	Yes	Best	Best	Best	High	Least

TABLE II

PRODUCTIVITY FACTORS

	Can Be Visually Monitored	Repair With The Same Process	Suitability For Narrow Joints	Welder Skill Required	Repair Frequency	Out of Position Capability	
						Horizontal	Vertical
SAW	No	No	Limited	Little	Moderate	Some	Some
SMAW (Manual)	Yes	Yes	No	Much	Highest	Yes	Yes
GMAW	Yes	Yes	Yes	Little	Moderate	Yes	Yes
HW-GTAW	Best	Yes	Yes	Little	Least	Yes	Yes

	Productivity LBS/Hr	Cost Per Unit Deposited	Preheat Requirements	Cost of Weld Preparation & Backchipping	Field Welding Experience
SAW	High	Low	Moderate	High	Some
SMAW (Manual)	Low	High	High	High	Much
GMAW	Moderate	Low	Low	Low	Modest
HW-GTAW	Low	High	Low	Low	Least

TABLE III

QUALITY FACTORS

	Can Be Visually Monitored	Weld Quality	Repair Frequency	General Weld Quality	Possibility of Entrapped Flux	Possibility of Hydrogen Pickup	Possibility of Porosity
SAW	No	Good	Moderate	Good	Yes	Some	Some
SMAW (Manual)	Yes	Good	Highest	Fair	Yes	Yes	Most
GMAW	Yes	Good	Moderate	Good	No	No	Some
HW-GTAW	Best	Best	Least	Best	No	No	Least

	Typical Heat Input	Separate Arc/Filler Wire Control	Multi-Pass Grain Size Control	Flux Moisture Control	Possibility of Distortion	Field Welding Experience
SAW	Highest	No	Poor	Some	Most	Some
SMAW (Manual)	Low	No	Poor	Most	Some	Much
GMAW	Low	No	Good	None	Least	Modest
HW-GTAW	Low	Yes	Good	None	Least	Least

this would appear to eliminate flat-position submerged-arc as a preliminary process. This is especially true where "fairing" welding and resultant grind-blending is to be minimized and/or when a mechanized/ automated welding process is to follow. Submerged-arc welding can provide very high deposition rates. However, for pressure-vessel welding, this advantage is somewhat reduced by the need to use lower-deposition techniques to reduce heat input and obtain a degree of grain refinement in both the weld deposit and the heat-affected zone. To compensate, pressure vessel fabricators have developed welding techniques that employ remarkably narrow grooves [about 1 3/8-in. (35 mm) wide in 12-in. (305 mm) sections] so that the deposition efficiency for the process is good.

Submerged-arc welding has been amply demonstrated by extensive flat position use for welding nuclear and petroleum reactors. The equipment is simple and easily maintained, and highly skilled operators are not required. Though some other more sophisticated form of mechanized welding may eventually present a challenge, it is unlikely that a transfer in technology will occur in the near future since, for flat-position welding, the submerged-arc process has few disadvantages.

The major welding problem, and cost, in on-site assembly will occur in making horizontal and vertical welds. In recognition of the difficulties involved in making these junctures, their number will certainly be limited as much as possible within constraints imposed by cross-country transportation capability. Procedures for positioning and aligning cylindrical shell sections have been devised, and the lift capacity for handling appears to be adequate. Vertical welding with the submerged-arc process is restricted to the "Sub-Vert" technique in which the weld beads are deposited in the horizontal position by traversing the welding head through the thickness of the vessel wall. The procedure significantly lowers the deposition efficiency and increases cost since at the end of each short bead the welding process is terminated until the welding head has retracted and been repositioned for the next pass. A unique manual shielded metal-arc technique, called "Up-John", has been used for making vertical welds in the same manner. Special electrodes are employed to avoid arc stops and starts within the wall of the vessel.

Shielded Metal-Arc Welding

Shielded metal-arc welding is a low-deposition process not suited to narrow-groove configurations. Therefore, welding costs are relatively high. There is no doubt that good quality shielded metal-arc welds can be made by skilled welders, as long as the proper procedures are followed and controls are established to insure the integrity of low-hydrogen electrode coatings. There is also no doubt that all welders are not equally skilled and consistently attentive, or that electrode-moisture control procedures are always followed. Figures on the cost of weld repairs are difficult to obtain since they are usually factored into the general cost of welding. However, the cost of repairs is significant since the rate of repair for manual welds can run to as much as 30-40% compared to the 5-10% rates experienced for machine welds.

There is, furthermore, a shortage of welders who can qualify for pressure-vessel fabrication. The problem is a recurring one, influenced by upsurges in construction, but common to all industries where welding is used. Schedule slippages caused by a shortage of welders occurred in the late sixties and early seventies in nuclear plant construction. Testing and qualifying welders also posed serious problems. As many as 100 welders were tested for each welder who could meet the minimum standards required for qualification and be employed.

Electroslag Welding

The possibility of vertical seam welding by use of the electroslag process arises in any discussion of thick-walled assemblies since the extremely high deposition rates, and the potential for significant cost reductions are particularly attractive. However, the mechanical properties of the weld deposit and heat-affected zone are not acceptable in the welded and stress-relieved condition, and complete re-heat treatment of massive weldments presents problems that are staggering even when special facilities exist, and are insurmountable at site locations.

Electron-Beam Welding

Electron-beam welding, which after some twenty years of development has advanced to the point where it can be given consideration for thick-walled assemblies, is attractive from the standpoint of very high rates of welding and freedom from flux-moisture complications. However, further investigation and development is required. The electron-beam process provides all the conventional risks of weld-metal porosity and cracking problems plus a few unique to the process. Most cracking problems have been overcome by adjustments in welding parameters and by employing various types of beam oscillation. The "necklace effect", lack of fusion at the tip of non-penetrating welds, is a persistent difficulty that may not be entirely corrected. In that event, unless small mid-wall flaws are eventually considered to be tolerable, full-penetration electron-beam techniques would be required. Since welds in about 10-inch (254 mm) thick A387 Grade 22 Cl. 2 steel have been made, the problem is not insurmountable. Objections to near-term application of the electron-beam process are found in two areas. Though the process itself is not new, applications to thick-section welding are still at an early stage of development and it will be some time before the potential of the process can be adequately demonstrated and evaluated. Furthermore, aside from the complexities involved in adapting the process to field fabrication, the information on the properties of thick-plate electron-beam weldments is inadequate.

The techniques for heavy-section welding employ hard and soft vacuum. Demonstrations of girth welding on small vessels have used sealed and evacuated shell sections as the gun-containing chamber, with a circumferential chamber, also evacuated, around the outside of the vessel. Though the welding process itself is fast, the time to set up equipment, seal, and evacuate a huge vessel would be appreciable. Joint fit-up, alignment, and cleaning are also required, and the tolerance for joint fit-up does not appear to be well established. Recent results indicate that joint gaps of as much as 0.05 inch (1.3 mm) with an offset of 1/8 inch (3.2 mm) may be acceptable. Whether or not these can be attained in huge sections under field handling conditions remains unknown.

Pump-down difficulties in large chambers can be minimized by use of small gun-containment and back-side chambers that employ sliding shields or gaskets. However, the problem of the design of sliding shields, which would be vulnerable to damage in on-site operation, is only in the conceptual stage and remains a significant problem. Horizontal welds, because of the large volume of molten metal involved, develop severe centerline voids unless the beam and the weld-joint preparation is angled downward a few degrees and a supporting shelf which supports the molten pool is positioned just below the joint. Installation of the support and subsequent removal, imposes additional cost and simple shelves are not suited to vertical welding situations.

In the absence of hydrogen, cracking in the weld heat-affected zone does not appear to be a problem when electron-beam welds in A387 Gr. 22, Cl. 2 are prepared without preheat. The reported mechanical properties of weld deposits, though limited in number, are good. Since the base plate provides the weld metal and the composition range for the plate is wide, all heats of steel may not perform equally well. The inability to independently control weld metal properties by simply modifying the filler-metal chemistry is a disadvantage common to all autogenous welding processes.

Electron-beam welding, since it was first employed to join relatively thin materials at very high rates of travel, has a reputation for advantages as a low heat-input process. A change in this impression is required when high beam-power equipment is used to weld sections as thick as 4 or 8 inches (100-200 mm). To avoid weld-metal defects in thick sections, the rate of travel must be greatly reduced. In this instance, a wide heat-affected region is developed, in some cases extending as much as 1/2 inch (13 mm) on either side of the fusion line. The existence of a significant volume of both coarse-grained heat-affected-zone microstructure and unrefined fused metal, are cause for concern. Increased susceptibility to temper embrittlement and stress relief cracking have been associated with these microstructures. Stress-relief cracking occurs along prior austenite grain boundaries in columnar weld deposits or in the coarsened heat-affected base metal.

Studies involving weld metal have indicated that weld deposits are less susceptible to stress-relief cracking than the coarsened heat-affected zone. Whether this is also true when the weld metal and the base plate are of identical composition has not been established. The question may be unimportant, however, since in an electron beam weldment both structures will extend through the wall thickness of a vessel.

When the existence of stress-relief cracks in coarse-grained heat-affected zone microstructures adjacent to submerged-arc butt welds in some nuclear pressure vessels were first suspected, and later confirmed, there were serious questions about the possibility of crack propagation during anticipated service life. In this instance, the isolated coarse grained regions that contained microcracks were small and separated from each other by extremely tough fine-grained structures developed by a multipass welding technique. Fracture-mechanics evaluations showed that the flaws would not propagate, and link, during service. Since the presence of an embrittlement susceptible microstructure that provides a continuous through-wall path for stress-relief crack propagation exists in an electron-beam weldment, the potential for electron-beam welding of pressure vessels cannot be assessed until this risk has been evaluated. If the risk is unacceptable, electron-beam will join electroslag in the group of processes that cannot be employed for as-welded fabrication.

Narrow-Gap Gas-Metal-Arc Welding

The gas metal-arc process initially was developed to compete with submerged-arc for high deposition-rate flat-position or horizontal welding using argon or helium shielding in a spray-transfer mode. However, the major application for the process has not occurred in high deposition-rate applications. The development of short-circuit (dip-transfer) GMA techniques provided capability for wider application in all-position welding with deposition rates of 6 to 8 pounds (2.7 to 3.6 kg) per hour. The process commonly employs CO_2 for shielding though argon-CO_2 and argon-oxygen mixtures are also employed. The process is prone to lack-of-fusion defects; a problem that has not been entirely or consistently resolved.

Pulsed-arc GMA welding has been investigated in an attempt to retain the all-position capability of the short-circuiting process while avoiding its problems. Inert gas is employed for shielding. At low current the arc is maintained without metal transfer while transfer is in the spray mode during the high current pulse.

Flux-cored and self-shielded core wire versions of the GMA process were developed to compete with submerged-arc welding. Both employ tubular wires containing flux for arc control, shielding, and alloy addition. The flux-core process employs CO_2 shielding gas and with the introduction of small-diameter wires has replaced solid wire GMA welding in many applications. In the self-shielded process, a supplemental shielding gas is not used, but the flux contains inorganic materials that decompose and shield the arc. With both processes, particularly in the case of the latter, the quality of the weld deposits are considered unsuited to pressure-vessel applications. All successful mechanized GMAW equipment for piping and other specialized high-integrity applications have employed solid wires.

The problems involved in adapting either the GMAW process or GTAW hot-wire process in their mechanized narrow-groove modifications, to horizontal and vertical on-site applications are essentially the same. The equipment is similar in weight and size, demands equal protection from the elements, and both processes require competent, carefully trained, though not highly skilled, operators. A choice between them rests, primarily, on weld-metal quality and their ability to make defect-free welds with the consequent savings in time and money that results from a reduction in repairs. On that basis, the narrow-groove hot-wire GTA process has the advantage.

Westinghouse Narrow-Groove Hot-Wire Gas Tungsten-Arc Welding

The Westinghouse narrow-groove hot-wire gas-tungsten arc process is an innovative approach to thick-section welding. However, it is not a new or unproven technique. The hot-wire GTA process has been employed by Westinghouse Divisions for over four years for fabricating stainless and low-alloy steel nuclear vessels in sections to 6 inches (152 mm). During this time, more than 40 tons (36,300 kg) of filler wire has been deposited. The quality of the welds has been excellent and repairs have been few; less than one repair for each 100 miles (160 km) of weld deposited. Roughly half of the welds have been made in the horizontal position. In view of the commercial success of the process, it is difficult to envision a heavy-section welding application for which HW-GTA welding would not receive first consideration. Advantages of the process include:

- Use of a narrow groove that results in a 60 to 80% reduction in weld metal compared to conventional practice, and reduced cost for weld groove preparation.
- Flexibility in programmed manipulation including arc oscillation, dwell, and pulsing.
- Freedom from flux entrappment and the possibility of hydrogen cracking caused by moist flux. Since hydrogen cannot be introduced, preheat temperatures can be lowered and in some instances, preheat is not required.
- Resistance heating of the filler wire removes surface contamination and reduces the potential for porosity.
- An independent heat source provided by the tungsten electrode. This characteristic allows flexible control of heat input, weld puddle size, and weld-metal solidification. The weld-metal deposition rate is adjustable independent of arc energy, yielding precise control over weld puddle shape and penetration. Parameters may be tailored to produce excellent root pass welds and subsequent passes at higher rates of deposition.

- The weld is not obscured by flux and visual monitoring of the process is unusually good, compared to GMA welding. Vision is not impaired by fumes or smoke and the weld pool is placid.

- Use of a narrow groove, an inherently low heat-input process, and precise bead positioning in multipass deposits permits grain refining and tempering of the weld deposit and heat affected zones. Reductions in notch toughness, cracking, and increased severity of temper and creep embrittlement have been associated with coarse-grained microstructures.

- In some instances, the process can be used for root fusion eliminating weld back-ups, back gouging, and the cost of manual operations.

- Repairs can be made with the same process using fixturing employed for the basic welding operation.

- This process is characterized by the best EASE OF AUTOMATION (Table I), yields better than average PRODUCTIVITY FACTORS (Table II), and provides outstanding QUALITY FACTORS (Table III).

EMBRITTLEMENT

Three types of embrittlement can occur in 2 1/4 Cr-1 Mo steel during exposure to elevated temperatures during service or during thermal stress-relief treatment:

- Temper embrittlement
- Creep embrittlement
- Stress-relief embrittlement

The characteristics of all three types are similar. Embrittlement occurs at temperatures up to about 1100°F (590°C). Fractures in embrittled material are intergranular. The susceptibility is influenced by microstructure. Ferrite-pearlite and bainite-pearlite structures are less susceptible than quenched and tempered or weld microstructures.

Temper Embrittlement

This embrittlement can only occur in alloy steels during slow cooling or exposure in the temperature range 700-1100°F (370-590°C). Within the range there is a characteristic C-curve time-temperature relationship. Temper embrittlement occurs only in commercial steels that contain trace quantities of Sb, P, Sn, and As; and/or larger amounts of Si and Mn. Temper embrittlement, in Ni-Cr steels, has also been associated with Bi, Se, Ge, and Te. High purity alloy steels of the same composition are not susceptible[37]. The amount of embrittlement caused by a specific impurity depends on the specific alloying elements present. Embrittlement is delayed by small amounts of Mo and W; enhanced by Cr and Mn. Large amounts of Mo and W also may increase severity.

Temper embrittlement is indicated by elevation of the ductile to brittle energy and fracture appearance transition temperatures and by a change from transcrystalline to intercrystalline fracture as embrittlement proceeds. The fracture path follows prior austenite grain boundaries. A rapid quench through the embrittling temperature range will avoid the difficulty. The phenomenon is also reversible. Heat treatment at a temperature above the embrittlement range results in restoration of toughness.

- Emmer, Clauser, and Low[54] have reviewed the embrittlement characteristics of 2 1/4 Cr-1 Mo steel (A387/A542). The following is a summary of their more pertinent observations.

    1. Considerable evidence that A387/A542 steels are susceptible to temper embrittlement exists.
    2. With reference to work reported by Kerr[73] for A542 steel:
        (a) Both base plate and weld metal are susceptible. The FATT may be increased to a temperature as great as 300°F (150°C).

        (b) Significant embrittlement can occur at temperature as low as 800°F (430°C). The lower range, estimated to be about 700°F (370°F), was not determined.

        (c) Appreciable heat-to-heat variation has been observed; particularly in weld metals. Lower Mn and Si contents appear to reduce susceptibility.

        (d) For plate material, strength level and microstructure influence the embrittlement susceptibility:
            (1) Plate normalized and tempered to an ultimate strength less than 85 ksi (590 MPa) shows little embrittlement.
            (2) Quenched and tempered plate with an ultimate above 120 ksi (830 MPa) also showed little embrittlement.
            (3) Plate quenched and tempered to a strength between 85 and 100 ksi (590 and 690 MPa) shows appreciable embrittlement.

(e) Both quenched and tempered material and normalized and tempered material, at the same ultimate strength, will exhibit similar embrittled transition temperature. A step-cool* through the embrittling temperature range will, in about two weeks, result in at least as much embrittlement as six weeks at 900°F (480°C).

3. Work by Kerr was generally confirmed in a later study of A542 steel by Swift[74] who reported that step cooling resulted in increase in the FATT of at most 60°F (33°C) to levels of about 100°F (210°C). However, the ultimate strengths of the plates employed (with one exception, 86.9 ksi [600 MPa]) were either below 85 ksi (590 MPa) or about 120 ksi (830 MPa), strength levels at which Kerr found little embrittlement.

4. Bruscato[75] in a study of embrittlement of weld metal reported increased susceptibility with increased Mn, Si and impurity content. The maximum increase in transition temperature was 170°F (95°C) to a level of about 200°F (93°C). Tests were conducted after step-cooling treatments. Long time isothermal embrittling was not employed.

5. Clauser[76] and Emmer[77] in a two-part study of embrittlement of A542 steel found:

   (a) All commercial heats of A542 may be expected to undergo some temper embrittlement. After 5000 hr at 900°F the rise in the 60 ft-lb transition temperature may range from 50° to at least 130°F (28-72°C).

   (b) Water quenching after stress relief results in a lower transition temperature than furnace cooling from the stress relief temperature.

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*1100°F (590°C)/1 hr; 1000°F (540°C)/15 hr; 975°F (520°C)/24 hr; 925°F (490°C)/48 hr; 875°F (470°C)/75 hr; furnace cool to 600°F (320°C).

(c) The step-cooling treatment* was shown to be comparable to 500 hr at 900°F (480°C) and not a true indication long-time service embrittlement.

(d) A creep stress of 36 ksi (250 MPa) during embrittlement had little effect on the embrittlement of thin, air-cool material for times to 1000 hr. It should be noted that an "air cooled thin-section plate" was considered comparable to a water quenched heavy section plate.

(e) The toughness of the heat-affected zone of heavy section plate remained superior to the toughness of the base material after embrittlement.

- The Kawasaki Steel Corporation[96] has reported the properties of normalized and tempered 2 1/4 Cr-1 Mo steel with the composition modified to improve resistance to temper embrittlement. The material had low P (0.009%) and Si (0.08%) content with small additions of Ni (0.12%) and Cu (0.12%)**to compensate for the decrease in strength resulting from the low Si content. Tests were conducted with 0.8 inch (20 mm) and 4.3 inch (110 mm) thick plates. The 4.3 inch (110 mm) plate was water quenched from 1700°F (930°C). The thinner plate was quenched under two conditions to simulate mid-thickness water quenching of 1.7 inch (50 mm) and 9.8 inch (250 mm) thick plates.

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*1100°F (590°C)/1 hr/AC; 1000°F (540°C)/15 hr/AC; 957°F (520°C)/24 hr/AC; 925°F (490°C)/48 hr/AC; 875°F (470°C) 75 hr/AC.

**$\bar{X} = (10\ P + 5\ Sb + 4\ Sn + As) \times 10^{-2} = 9.25$ ppm

After quenching, the materials were subjected to various tempering and stress-relief treatments corresponding to tempering parameters* between 19.45×10^3 and 21.48×10^3. Following a step-cooling treatment**, Charpy tests revealed low susceptibility to temper embrittlement.

- Ueda, Ishikawa, and Sato[95] have investigated additions of aluminum and boron to 1 1/4%Cr-1/2%Mo steel to counter the deterioration in notch toughness and strength resulting from tempering and stress-relief treatments involving high temperature or very long times at temperature.

For normalized [1700°F (930°C)] material containing 0.06% Al and 7 to 18 ppm B, their results show:

- A considerable increase in toughness and room temperature strength even when stress-relieved using a tempering parameter* of more than 20×10^3. The Al-B treated steel will satisfy the ASTM specification for A387 Gr. 11 in thicknesses to 5.9 inches (150 mm) when tempered to a parameter of 21×10^3.
- The treatment is effective in increasing the elevated temperature yield and tensile strengths at 210 to 1110°F (100-600°C) with appreciable increase in yield between 570 and 750°F (300-400°C).

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*Tempering Parameter = T(log t + 20); T = °K, t = hr.

**1100°F (593°C) 1 hr, cool to
1000°F (538°C) 15 hr, cool to
977°F (525°C) 24 hr, cool to
925°F (496°C) 60 hr, cool to
876°F (469°C) 125 hr.

- The creep-rupture strength of Al-B treated material is similar to untreated steel in the region of low Larson-Miller indices and greater in the region of high indices.
- The treatment causes no deterioration in the resistance to temper embrittlement in the base plate or in simulated weld heat-affected-zones corresponding to the fusion line of welds made with a heat input of 71 KJ/in. (28 KJ/cm).

Addition of Al and B reduces the amount of intergranular and grain boundary ferrite in the predominantly bainitic microstructure. The authors see no reason that similar improvement in the properties of stress relieved 2 1/4 Cr-1 Mo steel could not be obtained by Al-B addition and also speculate on the possibility that improved weldability might be obtained by reducing the carbon content of treated steel by about 0.03%.

- Shaw[103] has investigated the temper embrittling characteristics of A387 Grade 22 steel in a study initiated in 1975, by the Task Group on Temper Embrittlement of the Division of Refining of the American Petroleum Institute. Though the work is not complete and future efforts may resolve some of the difficulties, the results introduce questions about the interpretation and reliability of embrittlement data.

The first phase of the program was a comparison of the performance of 65 different samples (plates, forgings, and weldments) in the as-received condition with performance after step-cooling embrittlement. This phase is complete. The second phase consists primarily of a study of 25 isothermally embrittled samples selected on the basis of the results in Phase I. These materials were subjected to five temperatures, 650, 725, 800, 875 and 950°F (343, 385, 427, 468, and 510°C) for 1000 and 10,000 hours. Tests after 20,000 hours will be completed in the first half of 1980. Minor tasks include determination of the effect of composition, strength and microstructure and of de-re-embrittlement.

Reasonably well established correlations between chemical composition (primarily Mn, Si, P and Sn) and the severity of embrittlement were not clearly evident in the data. Comparison of the test procedure employed with that used at other laboratories revealed three sources for this inconsistency:

- Fracture Appearance Transition Temperatures (FATT) are not at all reliable and confuse interpretation of data from laboratory to laboratory. A correlation between ΔFATT and ΔT_{40} is not always apparent. The data sometimes showed extreme differences between the two embrittlement criteria.

- Differences also result from variation in the chemical analytical techniques and the standards employed. Perhaps the most important factor is lack of a solid standard for spectrographic analysis of 2 1/4 Cr-1 Mo steel. Consequently, in conjunction with the API, the National Bureau of Standards is currently preparing a solid standard for the steel.

- Inconsistencies in Charpy test data also appeared in results obtained by two laboratories testing the same steel. A detailed comparative study showed the inconsistencies were, to a degree, the result of testing an insufficient number of specimens. Nevertheless, an unresolved bias in the data still appeared to exist.

Shaw also noted that a generally accepted step-cooling treatment intended to indicate the amount of embrittlement to be expected during long-term isothermal exposure was not always reliable, though reasonable for steels that do not embrittle greatly. Since the maximum embrittling temperature (MET) varies from steel to steel, the step-cooling treatment consisting of short-time isothermal steps ending with 925°F (496°C)-60 hour and 865°F (462°C)-100 hour treatments, a steel with an MET of 900°F (482°C) will embrittle more than at steel with an MET of 800°F (427°C).

A further complication was found with two isothermally embrittled steels in which substantial embrittlement took place between 1000 hours (41.7 days) and 10,000 hours (417 days), demonstrating that the shorter period of exposure does not give a reliable indication of the total embrittlement.

Creep Embrittlement

Creep embrittlement manifests itself by a decrease in ductility (usually measured by reduction in area) occurring under certain creep-rupture conditions. The ductility will generally reduce to a minimum and then increase with increasing time to rupture. The embrittlement is influenced by both the temperature and stress level, and by the strength and microstructure of the steel. Susceptibility is lower for low strength or normalized steels; greater for higher strength and quenched-and-tempered materials. Tensile ductility values can deteriorate to low levels as creep-embrittlement proceeds, with a change from a transgranular to intergranular rupture mode.

- The observations of Emmer, Clauser, and Low[54] on the general characteristics of creep embrittlement are:

1. The first indication of creep embrittlement of heat resisting steels was failure of flange bolts after several years of service in the temperature range 850-900°F (450-480°C). Failures were unexpected since the design stresses incorporated large safety factors based on laboratory tests.

2. Failure times ranged from 3000 to 12,000 hr depending on the steel and the load.

3. Characteristics such as damage irreversibility and stress/time dependence indicated that creep embrittlement differed from temper embrittlement, though the two are possibly related.

4. Creep embrittlement is more severe under conditions that permit small amounts of strain over long periods.

5. Creep embrittlement occurs in the range 800-1100°F (425-590°C).

6. Embrittlement is influenced by grain size, microstructure and operating temperature. Coarse-grain material is more susceptible than fine-grained material. A morphology leading to an ultimate strength greater than 110 ksi (760 MPa) will lead to severe embrittlement, as will lower service temperature. Embrittlement is more severe if a material is quenched from high austenitizing temperatures.

7. Creep damage is unaffected by subsequent heat treatments, and creep-induced void formation is irreversible.

8. Failure occurs along prior austenitic grain boundaries and may be revealed by increasing tendency to form grain-boundary cavities or cracks prior to failure.

9. The creep-embrittlement mechanism appears to be associated with:

   (a) Fine, intragranular precipitates which develop during the creep process and strengthen the grain interior.

   (b) Subsequent formation of denuded zones adjacent to prior austenite grain boundaries which lowers the boundary flow stress.

   (c) The formation of elongated grain boundary carbides which reduces the cohesive strength of the boundaries.

10. Temper-embrittling elements (P, Sn, Sb, and As) together with Cr and Mo apparently cause creep embrittlement. No specific alloy addition or special heat treatment that will avoid or minimize creep embrittlement has been found.

• Bruscato[37] as part of a study of temper embrittlement of 30 different 2 1/4 Cr-1 Mo shielded metal arc electrodes (see elsewhere) also investigated the creep embrittlement characteristics of two deposits. The weld metals selected had high and low levels of temper embrittlement impurity elements. Creep-rupture tests were conducted (at 900 and 1050°F [480 and 565°C]) at stresses to provide rupture in times to 2000 hours. The deposits were tested in two heat treated conditions:

1. Quenched in water from the 20 hr stress-relief treatment at 1250°F (675°C). This treatment avoided any temper embrittlement that might occur during a conventional slow cool.

2. Furnace cooled to 600°F (320°C) after stress relief and step aged to induce a high degree of temper embrittlement.

The results of these comparisons (considered "preliminary") are summarized as follows:

1. In 900°F (480°C) tests the deposit with high impurity was inferior (lesser RIA at fracture) to the low impurity deposit.

2. In 1050°F (565°C) tests the difference was more pronounced. The ductility of the deposit with high level of temper embrittling impurity deteriorated to low levels at rupture times of a few hundred hours. The low impurity deposit did not exhibit this deterioration in 2000 hours.

3. A temper embrittling treatment prior to testing delayed creep embrittlement. The relationship by which the temper embrittlement mechanism influences the creep embrittlement mechanism is not understood.

4. The two mechanisms have the same underlying cause -- impurity constituents.

● Steiner, de Barbadillo, Penn, and Stout[104] in a study conducted for PVRC compared the creep-rupture properties of ten welded, quenched and tempered pressure vessel steels. The steels included A517, A533, A212B, A543, A387B and A542 Class 2. The steels were tested in the 800 to 1100°F (426 to 593°C) range for rupture lives to 10,000 hours. Fabrication discontinuities, such as mechanical and metallurgical notches were incorporated in the study. In order to determine the effect of metallurgical notches on creep-rupture strength, standard 0.252-inch (6.4 mm) diameter creep specimens were adapted to include weld metal and weld heat-affected-zone structures, as well as base plate material in the test section. Mechanical notches were superimposed on both welded and unaffected base plates to determine the influence of these combined effects on rupture properties.

Two regions of interest were included in the temperature ranges employed. The lower temperature region, between 800 and 1000°F (426 and 537°C), was primarily related to service behavior with specific interest in long-time data. The higher temperature region, between 1000 and 1200°F (537 and 648°C) was of interest in relation to thermal treatment during fabrication where shorter-time data are applicable. To evaluate data obtained from the higher temperature portion of the studies, welded restraint tests thermally treated in the range 800 to 1200°F (426 to 537°C) were included in the investigation. Welding was performed by the shielded metal-arc process (E11018 electrodes) using 30 KJ/inch (1180 KJ/m) heat input for both Lehigh restraint specimens. The Lehigh test specimens were examined both as-welded and after stress relief at 1150°F (620°C) for one hour. Welded rupture test specimens were prepared with heat inputs of 70 and 30 KJ/inch (2755 and 1180 KJ/m). Preheat or post heat was not employed.

The creep rupture characteristics were found to depend primarily on the presence of Cr, Mo, and V. The carbon steel had generally low rupture strength at 1000°F (537°C), but good rupture ductility. The Mn-Mo and Ni-Cr Mo had intermediate strengths but generally low rupture ductilities, while the Cr-Mo steels with less than 1% Cr had intermediate rupture strengths but much higher rupture ductilities. The Cr-Mo steels with over 1% Cr had the highest rupture strengths at 1000°F (537°C) and good rupture ductilities. In the probable service temperature range, quenched and tempered steels retain a useful advantage in creep-rupture strength over normalized grades. The range increases as the chromium and molybdenum content increases. At higher temperatures, normalized microstructures would be superior.

The steels showed a wide range of ductility before rupture. The carbon and lower alloy showed high values of reduction of area, while some of the deeper-hardening steels were embrittled. Only A517F (T-1) and A543 (HY-80) were especially sensitive.

Welding lowered the creep-rupture strength of some steels and had no effect on others. A517E (SSS-100) and A517F (T-1) were reduced in strength more than 20%, A517A (T-1A) and A542 lost 10%. The remainder showed no effect.

The presence of a notch strengthened some steels and weakened others, but no correction to composition, microstructure, or smooth-bar properties was apparent. A387B showed notch strengthening, while A542 was weakened by notching.

Stress-relief cracking was produced in the A542 steel and in A517 steels (A387B not tested) by subjecting Lehigh specimens to a 1050°F (565°C) one hour treatment. The cracks formed in the coarse grained heat-affected-zone structure and appeared to be closely related to cracks formed in welded stress-rupture specimens which exhibited very low ductility (less than 15%) in the 900-1100°F (482-593°C) range.

- Lundin[115] in a review of the development and physical metallurgy of 2 1/4 Cr-1 Mo weld metal, emphasizing creep-rupture properties, notes that further research and understanding of the effects of welding processes is needed:

  - The microstructures produced in a weld, and therefore the mechanical properties, depend on a number of variables including the composition of the filler metal, the welding process and parameters, post-weld heat treatment temperatures, and flux characteristics. Each of the variables influence creep-rupture properties.

  - Difficulties with obtaining adequate high temperature strength and toughness in electroslag welded joints, despite extensive post-weld heat treat, are not restricted to that process. The relatively high heat-input typical of most submerged-arc welding practices may result in significant tempering during welding and explain the rapid tempering of 2 1/4 Cr-1 Mo submerged arc welds during stress relief. Tempering during automatic welding has also been reported to give rise to coarse carbides that normally occur during the later stages of tempering and are associated with inferior creep properties.

  - A number of articles have commented on the relative creep-rupture strengths of 2 1/4 Cr-1 Mo weld metal and base metal, with little agreement. The creep-rupture strength of weld metal has been reported as superior to base plate, equal to base plate, and inferior to base plate.

  - Lack of agreement also appears in comparison of creep-rupture strength as influenced by various welding processes. A statement in one report[116] that creep-rupture strengths of submerged-arc welds are inferior to shielded metal-arc welds is refuted in another report[117] that states that the differences result from composition variables and that differences between SMAW deposits can be as great as those between SMAW and SAW weld metals.

• Lundin[118,119] provided some creep and tensile data for three 2 1/4 Cr-1 Mo weldments made by submerged-arc, electroslag, and shielded metal-arc welding processes. Though the fabricators of the weldments had all indicated they expected the welds to meet Class 2 strength requirements, the submerged-arc weld was the only one that exceeded 75 ksi (518 MPa) minimum room-temperature ultimate after a 25 hour stress relief at 1300°F (704°C). The electroslag and shielded metal-arc welds had ultimates of 61.5 and 70.3 ksi (424 and 485 MPa), respectively. The short-time creep-rupture performance (less than 1000 hours to failure) of the electroslag weld was only slightly inferior to that of the submerged-arc weld and both exceeded the minimum values of Code Case 1592. The performance of the shielded metal-arc weld (E9015-B3L Class) was below minimum levels at 850, 950, and 1050°F (454, 510, and 656°C). This relatively poor performance was attributed to a low carbon content.

The room-temperature tensile data and creep-rupture behavior did not correlate. The electroslag weld, for example, had lower tensile strength than the shielded metal-arc weld but exhibited higher creep strength. Performance in hot tensile tests at 850, 950, and 1050°F (454, 510, and 565°C) correlated better. At 850°F (454°C), for short times, the shielded metal-arc deposit had a higher creep strength. However, at 950 and 1050°F (510 and 565°C) the electroslag weld was stronger. Apparently, hot tensile strength is a useful indicator of creep-rupture behavior for 2 1/4 Cr-1 Mo welds.

Further objectives for this program, conducted for the Weld Metal and Weld Procedures Subcommittee of the Fabrication Division of PVRC, include:

1. Long-time creep-rupture tests including tests of the electroslag weldment in the quenched and tempered condition.
2. Fabrication and testing of a higher-carbon shielded metal-arc weldment.
3. Computer analysis of existing creep data on 2 1/4 Cr-1 Mo base metal and weld metal to reflect the influence of composition, heat treatment, welding process, and other parameters.
4. Metallographic examination of welds and base metals.

Embrittlement and Cracking of Heat-Affected-Zone and Weld Metal

Post-weld stress relief is employed to reduce residual stresses and to temper the weld joint. Stresses present after welding are reduced by a creep process in which the elastic strain is converted to plastic strain at elevated temperature. At temperatures where creep can occur, both the weld heat-affected zone and weld metal may undergo changes that result in a decrease in notch toughness and cracks may develop in the coarse-grained region of the heat-affected zone. Cracking in the weld metal is not usually observed. The loss in toughness is stress-relief embrittlement. The cracks, which occur along prior austenite boundaries, are called stress-relief or reheat cracks.

During welding, the crack-susceptible portion of the heat-affected zone experiences brief exposure to temperatures sufficient to coarsen the microstructure and dissolve carbides. The cooling rates typical of welding processes are rapid enough to retain alloy elements in solution and to result in formation of bainitic or martensitic microstructures that, in the quenched condition, possess limited ductility. The inability of the susceptible microstructure to accommodate to creep strain during stress relief arises from a mechanism similar to that for creep embrittlement:

- Strengthening of the grain matrix by precipitation,
- Formation of elongated grain-boundary carbides that reduce the cohesive strength of the prior austenite boundaries,
- Development of an alloy denuded region adjacent to the boundaries resulting in a local reduction in flow stress.

Stress-relief cracking was rarely encountered with plain carbon steels. The phenomenon became important with the introduction of alloy steels to meet demands for increased strength, toughness, and heat resistance. The formation of stress-relief cracks is promoted by high residual stress and is often associated with reinforcements or notches that act as stress concentrators. In their small size, location, and intergranular characteristic, stress-relief cracks and cold or underbead cracks are similar. In instances when the cause of cracking is unknown, the possibility of undetected cold cracking, or cold cracking aggravated by stress-relief cracking, must be explored. Cold cracking can usually be avoided by control of flux moisture content or by increase in preheat or weld heat input. Techniques for avoiding stress-relief cracking are not equally simple nor equally effective.

- Pertinent features of stress-relief cracking summarized by Emmer, Clauser, and Low[54] include:

1. Stress-relief cracking can be expected to be most severe in steels which exhibit low creep ductility and in steels that are capable of precipitation.
2. Stress-relief cracking is dependent on the strength levels of the parent matrix and the welding electrode.
3. Stress-relief cracking is dependent on grain size, morphology, and stress-relieving temperature. Coarse-grained material is more susceptible than fine-grained material. At lower stress-relieving temperatures cracking becomes more severe.
4. Cracking is dependent upon the heating rate to the stress-relief temperature. Depending on the prior thermal treatment the weldment has received, the size of the part, and other factors, both high and low heating rates may lead to cracking.
5. The cracks are intergranular and exhibit low ductility with little or no evidence of deformation.
6. Chromium, molybdenum and vanadium enhance susceptibility to stress-relief cracking. Molybdenum additions tend to retard temper embrittlement.
7. Stress-relief cracking has been shown to occur in steels having ferrite plus carbide microstructures, austenitic microstructures and martensitic (quenched and tempered) microstructures. It occurs in the temperature range 200-1800°F (90-980°C) depending on the alloy considered.

• Pense, Stout and Kottcamp[28] have provided some information on stress-relief embrittlement of A387 steel resulting from relatively long exposure. The study included nine materials used or considered for use in pressure vessels. Among these was A387 austenitized at 1700°F (925°C), in both the normalized and the spray-quenched conditions. The normalizing cooling rate matched that of a heavy section. The cooling rate for the spray quench was equivalent to that near the surface of a heavy plate. After cooling, the materials were stress-relieved 1 hr at the recommended temperature; A387 at 1350°F (735°C). Since in the course of fabrication, large pressure vessels may receive repeated stress-relief treatments and experience long periods at temperature, the steels were again stress relieved for 100 hr before tensile and impact testing.

Almost all the steels compared suffered some loss in ultimate strength after the 100 hr stress relief, usually only 5-10%; for T-1, HY-80, and A387, the reduction was about 25%. The ultimate strength of the normalized A387 was reduced from 90 to 72 ksi (620-495 MPa); spray-quenched from 102 to 82 ksi (703-565 MPa). The reduction in strength was almost always accompanied by a decrease in notch toughness; generally small. For T-1 and the spray-quenched A387 steel, however, the decrease was appreciable. The normalized A387 steel showed an increase in the 15 mil (0.38 mm) transition temperature of about 25°F (14°C) to slightly above 0°F (-18°C). The increase in transition for the spray-quenched material was 110°F (60°C) to about -40°F (-40°C). The spray-quenched material exhibited the greatest decrease in toughness of all the steels compared. However, stress relief at 1350°F (730°C) is a high temperature and would be expected to cause a more marked effect. Change in toughness appeared to be associated with carbide agglomeration on ferrite boundaries. A condition that would not be corrected except by reheat treatment.

● Almost all papers dealing with stress-relief cracking refer to formulae, similar to the carbon-equivalent, for predicting the sensitivity to cracking:

Nakamura, et.al.[105]

$$\Delta G = \%Cr + 3.3(\%Mo) + 8.1(\%V) - 2 \lesssim\gtrsim 0$$

Ito and Nakanishi[106]

$$P_{SR} = \%Cr + 2(\%Mo) + 10(\%V) + 7(\%Nb) + \%Ti - 2 \lesssim\gtrsim 0$$

Both formula indicate freedom from cracking at values less than zero, and increasing susceptibility at values greater than zero. They were derived from extensive series of welding tests and have proven useful, though not infallible, in preselecting steels for welding. These equations are of doubtful value for 2 1/4 Cr-1 Mo steels since they indicate a higher level of sensitivity than experience confirms. Other steels, with lower ΔG or P_{SR} values are known to be much more prone to develop cracking during stress relief. However, the susceptibility to stress relief cracking is not controlled by the chemical composition of the steel alone, but is also strongly influenced by the heat input employed during welding.

● This point has been demonstrated by Ujiie, Sato, and Matsumato[107] who obtained data showing that steels that are normally considered insensitive to cracking can be made crack sensitive when high heat-input welding conditions are employed. Their results show that stress-relief cracking can occur with high heat input even if ΔG is negative, but will not necessarily occur under low input conditions even when the ΔG value is positive.

● Similar results were obtained in a study by Grotke, Bush, Manjoine, and Slepian[108] using a quenched and tempered SA-508 Class 2 steel that had exhibited underclad stress-relief cracks when it was stainless clad using a high heat input 6-wire, submerged-arc technique. Extensive metallographic examinations of 9 3/4-inch (248 mm)-thick single-wire submerged-arc butt welds made with a low, 65 KJ/in (2560 KJ/m), heat input and a high, 130 KJ/in. (5120 KJ/m), heat input revealed cracks only in the high heat-input weldment. Prior examination of a similar production weld in the same steel made with an intermediate input of 95 KJ/in. (3740 KJ/m) also revealed stress-relief microcracking, but the cracks were smaller and occurred less frequently.

- Stress-relief cracking in steam plant pipe joints involving 1/2 CrMoV steel pipe welded with 2 1/4 CrMo weld metal has been the subject of considerable attention in the United Kingdom. A survey of butt-welded joints in service showed 2% of the welds examined were cracked in the HAZ, 2% had circumferential weld-metal cracks, and 5.5% contained transverse weld-metal cracks. Most of the cracking was attributed to stress-relief cracking that occurred during manufacture or early in service at about 1020°F (550°C). Miller and Batte[40] employed a notch-bend relaxation test using Charpy-size specimens to evaluate the susceptibility of deposits made with 18 different 2 1/4 CrMo manual electrodes and submerged-arc/flux combinations. There was no apparent correlation of stress-relief cracking with the welding process, bulk residual-element composition (Cu,P,Al,Sn,Sb,Pb,Ni), or deposit hardness even though the range of crack susceptibility was wide. These investigators concluded that the influence of residual elements is too complex to be quantifiable when the residual-element composition is within the range generally experienced in production materials and that the apparent relationship reported by others was probably a consequence of investigations based on particular batches of weld or base metals.

Miller and Batte also noted that sensitivity to stress-relief cracking can be reduced by an increase in the amount of grain refinement obtained in multipass welds. SMAW experiments with electrodes of different diameters using various weave patterns indicated that deposition techniques that refined about 90% of the initial coarse bainite avoided cavitation and cracking even in a susceptible material under mechanical restraints similar to those in production welds.

● Jones[62] investigated the stress-relief cracking susceptibility of four 2Cr-1Mo submerged-arc deposits to determine whether changes in wire or flux composition would reduce cracking difficulties. Stress-relief cracking occurred more frequently in submerged-arc deposits than in shielded-metal-arc deposits. Combinations of two different filler wires and three fluxes were employed to make welds under constant conditions. The preheat and maximum interpass temperatures were 390°F (200°C) and 660°F (350°C), respectively. Comparisons were based on short-time stress-rupture tests employing heating at a rate of 150°F/hr (85°C/hr) to a test temperature of 1275°F (690°C) to simulate thick-plate stress-relief conditions. Jones found that small variations in chemical composition had little effect on the rupture strength of the weld metal. However, wide variations in ductility (RIA) were observed, though the minimum exceeded 20%. In all specimens, cavitation and cracking were concentrated along prior austenite grain boundaries in the columnar regions of the deposits in a manner typical of stress-relief cracking in weld metal. The weld-metal ductility decreased initially with increasing time to rupture to a minimum where it remained constant. The minimum ductility correlated with Mn content. Increased Mn refined the bainitic structure and increased the ductility. Maximum rupture ductility was associated with a high Mn content and use of a basic flux.

● Swift and Rogers[66] have provided a general review of the embrittlement of a variety of weld metals. Though their principal interest was the 2 1/4 Cr-1 Mo composition, little specific information was presented. Whenever weld-deposit data were lacking, pertinent information for plate material were used. They note that weld preheat can have either a beneficial or detrimental influence on the toughness of shielded-metal-arc deposits. Welds in 2 1/4 Cr-1 Mo pipe showed some increase in embrittlement when a 300°F preheat was employed. However, no trend in the effect of preheat was evident when the weld deposits were subsequently stress relieved at 1300°F (700°C). No information was found on the influence of other welding parameters, but tempering of quenched 2 1/4 Cr-1 Mo plate in the range 400-1100°F (200-590°C) was reported to cause embrittlement.

In a discussion of electroslag welding of 2 1/4 Cr-1 Mo, it was noted that a post weld heat treatment such as normalizing or austenitizing and quenching was required to refine the coarse-grained structure. In one instance, the as-welded Charpy energy of 5 ft lbs (7 Joules) at 75°F (24°C) increased to 108 ft-lbs (146 Joules) following a reaustenitize and temper treatment. No specific information on gas-metal-arc, gas-tungsten-arc, or submerged-arc weld 2 1/4 Cr-1 Mo deposits was presented. It was noted, however, that the toughness of Cr-Mo welds apparently decreases with increased Cr content, probably merely a consequence of the strengthening effect of Cr.

Quench-aging embrittlement of carbon and alloy steels rapidly cooled from a stress relieving temperature sufficiently high to cause solution of iron carbides or nitrides was also discussed. The phenomena can effect plate, heat-affected zones, and weld metal. In the case of the 2 1/4 Cr-1 Mo composition, slow cooling from a temperature above 1200°F (650°C) reduces or eliminates the effect. Therefore, the effect would be insignificant in heavy-gage weldments which are slowly cooled from the stress-relief temperature.

In a discussion of heat-affected zone embrittlement, the authors noted that when 2 1/4 Cr-1 Mo plate was tempered at 700°F (371°C) after water quenching from 1750°F (955°C) or 2300°F (1260°C) the impact energy transition temperature was higher than that of the as-quenched plate. Comparative values were not presented. In summarizing the available data the authors noted:

1. The degree of embrittlement (temper, creep, stress-relief) increases with increasing grain size for both weld metal and plate.

2. Stress-relief and creep embrittlement are more severe at higher strength levels; temper embrittlement appears more pronounced at lower strength levels.

3. Higher impurity content appears to increase the severity of all three types of embrittlement. In the case of stress-relief embrittlement, however, definitive data are lacking.

4. The influence of welding parameters on each type of embrittlement has not been sufficiently examined. No correlation between welding processes and embrittlement appears in the literature. Based on available information, lower heat input processes generally reduce embrittlement.

5. Despite evidence that all three types of embrittlement may be related, well established correlations have not been generated and mechanistic evaluation of the factors affecting plate and weld metal have not been conducted.

- Bland[78] investigated shielded-metal-arc welding of 2 1/4 Cr-1 Mo pipe with a wall thickness of 5/8 inch (16 mm). He considered the results were applicable to plate of at least 1 inch (25 mm) in thickness. The comparison included four weld metals that, deposited without preheat, had ultimate strengths in the range 123-133 ksi (848-917 MPa). The pipe employed had an ultimate strength of 67 ksi (462 MPa). Tests included a correlation of strength and hardness using tensile specimens, longitudinal and side-bend tests, impact tests, and an investigation of preheat requirements. Bland concluded:

1. The use of preheat up to 500°F (260°C) has little or no effect on the strength, ductility or impact properties of weld metal deposited by low hydrogen electrodes, or of the heat-affected-zone.

2. Postheating at temperatures below 1050°F (565°C) produces no significant effect on weld metal properties. Postheating in the range 1050°F (565°C), for 16 hr to 1150°F (620°C) for 2 hr produces an undesirable increase in hardness and decrease in ductility. The heat-affected zone, however, is relatively unaffected by these treatments.

3. Unrestrained welds made without preheat or postheat were sound. Under restrained conditions, a preheat of 300°F (150°C) was recommended to avoid root cracking.

4. The ductility of welds in the no preheat/no postheat condition was entirely satisfactory based on the results of longitudinal and side-bend tests.

5. Post weld treatments of at least 2 hr at 1150°F (620°C) or 1/2 hr at 1250°F (675°C) are required to produce significant reduction in the hardness of weld deposits of heat-affected-zones.

6. Though the 2 1/4 Cr-1 Mo steel is air hardening, test results indicate no adverse affect from increased hardness. The notch-bend properties of the heat-affected zone in the no preheat/no postheat condition was superior to the weld deposits in any thermally treated condition.

7. The weld deposits exhibited impact properties substantially lower than the base metal; but not unacceptably low.

- Kluek and Canonico[64] investigated the microstructure and tensile properties of 2 1/4 Cr-1 Mo TIG-welded steel with carbon contents of 0.003, 0.035 and 0.11%. The 0.11% C steel, prepared by vacuum remelting stock from a commercial heat, is typical of commercial compositions. Tests were conducted with 1/2-inch (13 mm) plate, normalized at 1700°F (927°C) and tempered at 1300°F (704°C). Material from the respective plates, processed to 1/8-inch (3.2 mm)-diameter filler rod, was used to join the respective butt-welded assemblies.

The maximum hardnesses obtained in the heat-affected-zone and weld metal in the 0.11% carbon steel assembly were 364 DPH and 390 DPH, respectively. After stress relief at 1300°F, the peak hardness in both regions reduced to 228 DPH; only 28 DPH higher than the base metal. The base-metal microstructure was 55-65% tempered bainite, balance proeutectoid ferrite. The weld deposit was essentially bainite with some martensite present.

In this study, welds were made both without preheat and preheated to 300°F (150°C) with the plates welded to a 2-inch-thick strongback. When preheat was not employed the interpass temperature was stated to be less than 200°F (95°C). The interpass temperature for welds preheated to 300°F (150°C) ranged from 300°F to 500°F (260°C). Microcracking was not observed.

The minimum preheat/interpass temperature recommendations for A387 Gr. 22 and all classes of A542 steel found in WRC Bulletin 191 are 150°F (65°C) for thicknesses to 1/2 inch, 250°F (120°C) for over 1 to 2 inches, and 300°F (150°C) for thicknesses above 2 inches.

- Kluek and Canonico[65] also studied the creep-rupture properties of weld metal, transverse weldments, and normalized and tempered base metal for three heats of 2 1/4 Cr-1 Mo steel with 0.003, 0.035, and 0.11% carbon at 850, 950, and 1050°F (454, 510, and 565°C). Material preparation and welding procedures were discussed in a previous paper[64] that reported the microstructures and tensile properties. The high-carbon analysis is typical of commercial compositions.

For a fixed temperature and carbon content, the creep rupture properties of the weld deposits were invariably greater than those of transverse weldments or base metals, which had similar strengths for a given carbon content. The rupture properties of weld metal, transverse weldments, and base metals exhibited an effect of carbon content that was related to temperature. At the highest test temperature 1050°F (565°C) the 0.11% carbon steel had the highest strength, while the strengths of the other two steels were similar. At 850 and 950°F (454 and 510°C), the strength of the 0.035% carbon steel approached that of the 0.11% steel at long rupture times (2000 hr). Both were considerably stronger than the low-carbon materials. Failures in transversely welded specimens occurred in the base metal, with a single exception. The exception was in the high-carbon steel tested at 1050°F (565°C) at 15 ksi (102 MPa), the lowest stress employed. In this instance, failure initiated in the heat-affected zone. Similar failures would be anticipated at lower stresses.

● Batte and Murphy[24] have reviewed the creep-rupture properties at 1020-1110°F (550-600°C) of 2 1/4 Cr-1 Mo deposits to assess the general level of weld properties, the effect of welding processes and parameters, and to determine reasons for the reported superior creep strength of a particular SMAW electrode. Since the data reviewed was obtained from various forms of test specimens, the authors restricted their consideration to the results of tests conducted with material from all-weld metal pads or transversely butt-welded specimens that failed in the weld deposit. The deposits were made predominately by the SMAW and SAW processes, but a few were made by the gas-metal arc process (GMAW). To obtain an overall picture from weld metals tested to various temperatures, they used the mean data (to 20,000 hr) derived by the International Standards Organization[79] (ISO) for 2 1/4 Cr-1 Mo wrought materials. The data used for the ISO lines include both normalized and normalized and tempered material. Conventional 20% scatterbands about the ISO mean line were accepted as a reasonable basis for comparison with weld metal results.

Less than 5% of the reviewed results were below the ISO $\pm$ 20% limits about the mean for wrought material. Since a similar proportion of wrought material results were also below this limit, these data refute suggestions that welds are up to 20% weaker than wrought material.

Discounting isolated contradictory results, the data as a whole suggest that the welding process has no significant effect on creep-rupture strength. The data were insufficient to reveal the effect of flux composition though basic fluxes that result in cleaner deposits may improve ductility. Increasing the time or temperature of stress relief appeared to decrease creep-rupture strength. The electrode that exhibited superior creep-rupture performance contained slightly higher levels of titanium and nio-titanium additions, which increase strength and slowed the response to tempering. The effect of such additions on other weld metal properties was not established.

• Bruscato[37] compared the embrittling characteristics of thirty different shielded metal-arc welding electrodes (AWS-ASTM E9015-B3L, E9016-B3, and E9018-B3) in the stress-relieved condition and following a step-cool embrittling treatment. Deposits were made in V-grooved, 1-inch thick plates of A387 Grade D steel with essentially constant conditions. Preheat and interpass temperatures were maintained at 200-300°F (90-150°C). Stress relief was performed at 1275°F (690°C) for 20 hours followed by furnace cooling to 600°F (320°C). Since specifications generally stipulate a minimum impact energy at 50°F (10°C) of 40 ft-lbs, evaluation was based on the average performance of three specimens at that temperature.

Of the thirty electrodes, 27 were standard type 2 1/4 Cr-1 Mo compositions with a low-carbon rimmed-steel core. The three others, that provided the "cleanest" deposits, were fabricated with vacuum-melted, low-carbon steel cores. Of these, one had moderate levels of Mn plus Si (0.70 and 0.50%, respectively), and provided excellent toughness in the embrittled condition. The other two had high Mn + Si content and performed less well. The 27 standard electrodes exhibited wide variation in toughness at 50°F (10°C) in the stress-relieved condition even before embrittlement. The range was 12 to 110 ft-lbs (16 to 149 Joules). However, in all but three instances, the 40 ft-lbs (54 Joule) aim was either met or exceeded. After embrittlement the range was 5 to 92 ft-lbs (7 to 125 Joules) with 16 electrodes still meeting the 40 ft-lb (54 Joule) level. The high-purity, moderate Mn + Si weld deposit was best of all tested; 122 ft-lbs (165 Joules) stress relieved, 101 ft-lbs (140 Joules) embrittled.

Bruscato obtained the chemical analyses of all deposits and developed an embrittling factor for four potent embrittling elements:

$$\bar{X} = \text{Embrittling Factor} = \frac{10P + 5Sb + 4Sn + As}{100}$$

By plotting $\bar{X}$ versus the Mn + Si content, he obtained good correlation with the toughness of the embrittled materials. Bruscato also noted that for the electrodes tested, the phosphorus and tin content constituted over 90% of the Embrittlement Factor $\bar{X}$

Though welds with low manganese and silicon can tolerate greater quantities of embrittling impurity elements, a reduction in Mn + Si must be weighed against the beneficial effects of these elements in weld deposits. Manganese and Si deoxidize the molten pool and improve the notch toughness of the deposit. Bruscato unsuccessfully attempted reductions in Mn and Si. Below 0.50% Mn the deposits would not meet minimum toughness requirements even before the embrittling treatment. There is little freedom for reduction. Brucato's data shows that for the 24 "standard" electrodes that met the 40 ft-lb (54 Joules) aim in the stress-relieved condition, the average Mn content is only 0.68%.

In a few instances where a Charpy curve was obtained, the highest temperature for 40 ft-lbs (54 Joules) was 170°F (75°C). The maximum shift in transition temperature at 40 ft-lbs (54 Joules) was 120°F (67°C). Bruscato notes that 2 1/4 Cr-1 Mo pressure vessels normally operated between 700 and 900°F (370/480°C), temperatures at which the welds will perform satisfactorily. The only concern would occur during shut-down and start-up. Problems during these periods could be avoided by heating to about 200°F (90°C) prior to applying pressure to the vessel.

Since composition and embrittlement are related, it is interesting to note the range of pertinent elements reported by Bruscato for the "standard" 2 1/4 Cr-1 Mo deposits and for the base metal:

	Mn %	Si %	P ppm	Sn ppm	Sb ppm	As ppm
Welds	.54/.78	.32/.98	60/180	70/300	0.9/22	20/130
Base	.45	.23	65	150	27.7	284

REVIEW OF WELDING PROCESSES

Electron Beam Welding

- Terai and Nagai[81,82] have provided a general review of EB welding equipment and application to large and heavy section components, noting the three basic systems that have received consideration in recent years:

Local Chamber Welding

With this system the gun is moved over the workpiece and only the small volume between the gun and the work is evacuated. The major problem is maintenance of an adequate vacuum seal. This approach overcomes problems of size limitations associated with large structures.

Full Chamber Welding

Chamber welding with an auxiliary chamber to enclose the workpiece. This is an extension of conventional EB welding which with only small amounts of further development could offer a practical solution. Cost increases rapidly with increased component size and would be prohibitive for huge assemblies.

Non-Vacuum Welding

With this method, the electron beam, produced under high vacuum, is led through a nozzle with gradual reductions in vacuum down to atmospheric pressure. The technique eliminates problems with workpiece size but is limited in both working distance and penetration depth. Extensive short-range commercial application is unlikely.

Local-Chamber Equipment

The authors describe four types of local chamber equipment under evaluation in 1978 basically suitable for pressure-vessel applications. Gun sizes range from 12 to 60 KW at vacuums from 5 x 10^{-4} to 2 x 10^{-4} torr, with capability for flat, horizontal, and vertical welding. The thickness ranges for steel is 1.2 in. (30 mm) to 4.7 in. (120 mm).

Full-Chamber Equipment

This group also consists of four basic types. One unit developed by Sciaky (France) and the General Electric Co. comprises a 42 ft. (13 m) high by 36 ft. (11 m) diameter sphere fitted with jigs and a 30 kV, 30 KW gun. The unit will accept a wide variety of shapes and sizes of structures. Penetration in steel is limited to 2.8 inches (70 mm). Sciaky (France) also has a 100KW system for horizontal girth welding of cylindrical shells from the inside using a rotating electron gun suspended from a plate, at the top of the vertically positioned barrels which also support the pumping equipment. The barrels form the evacuated chamber, sealed with ring gaskets at the top and bottom, and with a cylindrical, U-section seal over the joint being welded. The unit is capable of welding 7.8 inch (200 mm) thick steel but is limited to short gun-to-workpiece distances in cylindrical objects.

Kawasaki Heavy Industries is using a 75-100 KW unit developed by the Welding Institute (UK) capable of welding 9.8 inch (250 mm) steel. The gun is mounted on a large vacuum chamber and produces a beam allowing long working distances under low vacuum (1×10^{-1} to 1×10^{-2} torr). This gun has been designed to overcome problems with gases or vaporized material from the material being welded causing discharge within the gun, which occurs at ratings of over 50 KW. Kawasaki is also working on another approach, suitable for flat, horizontal or vertical single-pass welding of steel to 12.6 inches (320 mm) thick. The 180 kV, 120 KW gun permits large variations in gun-to-workpiece distance. At a thickness of 12.6 inches (320 mm), the welding speed is 3.9 inches (100 mm) per minute.

Electron-beam welding has been used by Kawasaki Heavy Industries for nuclear reactor pipe work for which an accuracy of straightness better than 1 in 2000 is required. The process has also been used for steel boiler drums with results that suggest that nuclear heat exchangers could be fabricated by the same technique. The drums, 18 ft (5600 mm) long, 5 ft (1500 mm) in diameter, with 2 inch (50 mm) walls are welded with local shielding for the gun and an internal backup shield. A comparison shows a 2 inch (50 mm) thick joint made from both sides by submerged-arc welding would require 26 hours to complete. Single side EB welding, with a backup plate, reduces the time to 1/2 hour. Tensile and impact tests of welds in HY 90 steel show excellent performance. Comparison of SMAW, GMAW, and EB welds showed that the toughness of the EB welds was superior and similar to the base plate.

The authors feel that further development is required to solve problems with jigs, positioning equipment, and sliding shields. One problem still to be overcome is acceptance and approval of the technique for fabrication of pressure vessels, a topic mentioned only briefly in cases 1461 and 1471 of the ASME Boiler and Pressure Vessel Code. Even these cases imply that EB welding is intended for small parts rather than structures such as drums or vessels.

- Russel, Brown, and Adams(39) have examined the feasibility of fabricating a large cylindrical pressure vessel using electron-beam welds and reported results of tests conducted with a low-alloy steel to establish the tolerance for joint gaps. The vessels envisioned were 20 ft (6 m) in diameter with walls to 3 inches (75 mm) thick. Though very large structures have been welded by the electron-beam process, they have been quite light and their application has justified the use of the highest standard of machining to achieve the fit-up necessary for successful welding. Fabrication of large pressure vessels will necessitate manipulation and machining to close tolerances extremely heavy pieces which may distort under their own weight. Some means of insuring increased fit-up tolerance will be necessary if reproducible welds are to be made.

Maintenance of equilibrium in the weld zone is particularly important for any mechanized welding process. Anything that tends to upset this equilibrium, including variation in the joint gap or mismatch of plates, cannot be accommodated for as easily as in manual welding which employs wider beads and is subject to operator control and changes in technique. Since the diameter of the electron beam is small, it can pass through relatively narrow gaps without heating the metal on either side. It is difficult to machine the required extremely tight butt joints in huge structures. Therefore, techniques must be developed to obtain the maximum tolerance for gaps between faces. This approach will increase the width of the fusion zone and possibly require use of filler metal.

Tests to establish gap tolerance were conducted with a Hamilton Standard 150 kV, 6 KW machine with a fixed gun in the downhand position, using a hard vacuum of 5×10^{-4} mm Hg. To obtain more flexibility in welding parameters the thickness of the steel was limited to 1 inch (25 mm). A beam spinner that rotated the beam in a small circle (0.060 in. dia. [1.2 mm]) was employed to widen the fusion zone and reduce the incidence of root defects in nonpenetrating welds. Porosity was not a problem since the steel used was either vacuum degassed or vacuum-arc remelted. Preliminary melt run and close butt experiments showed that full penetration welds were sound but that wide blind welds exhibited some solidification cracks. Such cracking was observed only in welds deliberately made wider than normal by beam oscillation or by defocusing and only in vacuum-degassed material though the two steels used had essentially identical compositions. The reason for the difference in behavior was not established though the vacuum-degassed steel was observed to contain more inclusions. Another defect associated with a majority of the blind welds was a series of cold shuts (necklace defect) at the weld root. These necklace defects had a greater tendency to form in the roots of sharply tapering welds and sometimes were completely absent when the roots were round. Unfortunately, welds with rounded roots are normally wider and prone to solidification cracking. Defocusing, beam oscillation and spinning over a wide range of frequencies and amplitudes failed to eliminate the problem when high-voltage equipment was used. However, a low-voltage 30 kV, 9 KW machine was found capable of making defect-free welds with rounded roots. The 30 kV machine had an electron beam with hollow as opposed to the normal gaussian, energy distribution.

Difficulty was encountered in producing good top and bottom bead contours at the same time. One contour could be improved only at the sacrifice of the other. Therefore, the use of a blind weld into a backing material was considered a valuable procedure to avoid drop-through and to insure all root defects would occur in a region subsequently removed from the fabrication. Drop'through problems occurred on melt runs and with close butt joints and was aggravated by the presence of any joint gap. Substantial gaps would, therefore, require some form of underbead support.

As mentioned above, a top-surface depression was produced when the underbead profile was good. To correct this condition, filler metal addition was investigated using a two pass technique using a non-keyhole second pass to avoid the necklace defect. Conventional single-pass welding into a backup or arc weld deposit, with the necklace defect in the backup material, was also employed. Satisfactory welds could be made in close butt joints. However, when the joint gap was extended to investigate spacings to 0.04 inch (1 mm), top bead depression occurred with both techniques. The weld sink could be eliminated by adding filler metal during the first pass if the gap was small. When substantial gaps were involved, (0.02 in. [0.5 mm]) or more, the results were inconsistent. Feeding wire into the gap is the preferable procedure. However, this can be done only if the gap is consistent. In instances of periodic gap narrowing, the first pass could be made without filler to produce a uniform gap that could be closed with filler metal during the second pass. Either cold or hot-wire feed may be employed.

An alternative procedure was developed using gaps of 0.015-0.035 in. (0.4-0.9 mm) and powder mixture of parent-metal composition. The technique is of greatest value for gaps wide enough to permit dense compacting of the pre-placed powder. Some weld sink still occurred as a consequence of the low density of the powder, requiring a shallow sealing pass using wire or powder. The powder mixture technique is a possible source of trouble because of incorrect or inhomogeneous mixtures.

The following ranges of gap tolerance and welding conditions were proposed:

- 0-0.010 in. (0-0.25 mm) maximum.
  A two-pass technique, the second pass made with filler wire, beam deflection advantageous. Alternatively, a single-pass weld with filler addition.

- 0.020 in. (0.5 mm) maximum.
  Transverse oscillation or beam spinning required to widen the fusion zone and a two pass technique with the second pass made with filler metal. Metal powder could be preplaced in the original joint to reduce the amount required subsequently.

- 0.035 in. (0.9 mm) maximum.
  The technique is similar to the above except that preplacing of powder was necessary to reduce the weld sink. Wider gaps permit better compacting of powder. Adequate filling of wide gaps is facilitated by increased beam deflection and slow travel speeds.

Though feeding of powder from a hopper was not investigated, the authors believed that a system could be developed that would permit joint filling in a single pass. A powder-paste system was also considered to be possible for positional welding.

During these experiments it was noted that gaps tended to narrow ahead of the beam though shims and tack welds were placed approximately 4 in. (100 mm) apart along the test piece. Typically, an original cold gap of 0.020 in. (0.25 mm) might be reduced by half after a partially welded test piece was cooled. The possibility of devising a thermal technique for gap closing was not investigated.

A few horizontal and vertical welds were also made. Gap space and welding details were not provided. These welds had top and bottom contours that were superior to those made in the downhand position. Radiographic and metallographic examination showed they were free of flaws.

The authors also provided a summary of the state-of-the-art and anticipated problem areas influencing the potential for EB welding of large vessels:

Seam Tracking

The narrowness of the electron beam and the high depth to width ratio of the fusion zone make it imperative that the joint faces are perpendicular to the work surface or at some constant angle to it. Once aligned, the beam must remain aligned with the joint throughout welding. It is unlikely that joints in large components will be straight. Therefore, some method of seam following must be employed. Mechanical tracking systems, using a stylus in the groove, and scanning electron beam systems have been developed. If seam following by beam deflection is employed, best results are obtained with a high voltage gun which has a long working distance.

Inspection

Radiography will reveal areas of porosity and lack of fusion. Ultrasonic examination using a 45° shear wave probe is an effective means of detecting lack of fusion and estimating the size of defects. Visual observation of the underside, together with dye-penetrant will detect either lack of fusion or penetration defects only if careful weld-bead dressing is carried out.

Vacuum Environment

Workpieces would either be fully enclosed or protected locally by chambers fitted with static or sliding seals. Sealants are available which may be painted on chamber surfaces to seal pinholes or porous regions and reduce outgassing. Sealing against leakage from the outside presents the major problem since efficient functioning of seals must be maintained over a reasonable period of time without the need for frequent maintenance or replacement. Seal properties include resistance to heat, radiation, and chemical attack. Demountable, re-usable seals for chamber or chamber-to-workpiece closures will be most vulnerable to damage. Seals for low voltage power lead ins and angular/linear shaft movements are available. High voltage kinetic cable lead-ins for high power guns are not fully developed and may impose limits on some moving gun concepts. For many large chambers it is considered worthwhile to provide an evacuated volume between an inner and outer seal to reduce the leak rate and provide a means for testing flange sealing by quickly evacuating the enclosed pocket. It is unlikely that demountable mating flange faces will retain original close limits in heavy duty use. Therefore, solid gasket materials, only capable of about 25% portion compression, are not the best choice. Inflatable seals are attractive for gaps with a varying of tolerances. Sealing techniques for localized sliding seals present the greatest design problems.

Joint Fit-Up

Meticulous joint-gap cleaning is a vital requirement. All contaminants and turnings must be removed. Solvent cleaning of newly machined surfaces will suffice if carried out before surface deterioration has occurred and the joint is temporarily protected until welding. Components should be first rough machined to dimensions sufficiently oversize to permit parallel-gap remachining after mating surfaces have been positioned and faired. Single-point tools, milling cutters, slitting saw/grinding wheels or abrasive disks could be employed, depending on the amount of material to be removed.

Development Costs

Excluding personnel training, protection against health hazard, and the cost of more accurate workmanship standards, the authors estimate that establishing a large scale electron-beam welding facility would require five years from determination of requirements to production of structures at a cost between $500,000 and $1,300,000 depending on the vessel size and the procedure selected.

● Russell, Rodgers, and Stearn[49] surveyed the electron beam weldability of a variety of structural steels to assess joint properties and susceptibility to defects. The materials, all 0.87-inch thick and welded with a 6 KW 150 kV machine, included plain carbon, carbon-manganese, Mn-Cr-Mo-V, and 2 1/4 Cr-1 Mo. The latter steel was vacuum degassed. The others were either semi-killed, silicon-killed, or silicon-killed and aluminum treated.

Porosity and cracking are the two main problems encountered in electron beam welding of steels with high gas content and inclusion levels. Preliminary tests were conducted to optimize focus distance for parallel-sided welds. Thereafter, beam deflection was employed to minimize porosity. Circular patch melt-run specimens, examined radiographically and by sectioning, were used to evaluate crack susceptibility. From the results of these tests, conditions were selected and employed to prepare defect-free butt welds for determination of mechanical properties. Sufficient specimens were prepared for tests in the as-welded and the stress-relieved conditions. The tests included transverse tensile, side-bend, Charpy Vee-notch and crack opening displacement.

All of the steel except the vacuum-degassed 2 1/4 Cr-1 Mo showed some tendency to form porosity. In all cases, however, porosity could be eliminated by techniques that promote vacuum degassing during welding. Steels with high oxygen content were most prone to porosity. Hydrogen and nitrogen levels had no detrimental effect, particularly in the case of nitrogen. The best steels had the highest nitrogen content. High inclusion content did not indicate a steel was prone to porosity. The best steel, 2 1/4 Cr-1 Mo, had an intermediate inclusion level. However, the silicate level was very low. This material could be welded with a wide range of process parameters before porosity became a problem.

Neither weld metal or HAZ cracking appeared to be a major problem. Radiography and ultrasonic tests failed to reveal cracks in any steel. Some steels showed small cracks in polished sections. The tendency to form cracks was reduced at travel speeds below 8 inches per minute. The resistance to cracking exhibited by these steels was anticipated because their S and P contents were generally low. Steels with highest levels exhibited more and larger flaws. It is probable that increased susceptibility would be encountered with material at the high side of the range for S or P, and under conditions of greater restraint.

• Weber is investigating electron beam welding of 8-inch (200 mm) thick A387 Gr. 22 Cl. 2 steel at the Research and Development Division of the Babcock and Wilcox Co.[80] The objective is demonstration of a procedure for welding in both the horizontal and vertical positions meeting the mechanical property requirements of Section VII, Division 2 of the ASME Boiler and Pressure Vessel Code. The program includes:

- Development of procedures for partial penetration welding [about 4 1/2-inch (115 mm) deep] at vacuums of 0.06, 40 and 100 microns.
- Development of a technique for welding the full 8-inch (200 mm) thickness, from both sides, in the horizontal and vertical positions.
- Establishing a start/stop technique and a means for repairing a weldment in the event of equipment malfunction.
- Determining the limits of joint mismatch and joint fitup, and their combinations, that can be welded without modification of parameters.
- Determining the influence of magnetic fields on the electron beam and development of a technique to avoid excessive beam deflection.
- Evaluation of mechanical properties in both the as-welded and the stress-relieved conditions.

The work reported to date deals with the development of a technique for making partial penetration welds [4 1/2 inch (200 mm) deep] in 6-inch (150 mm) thick steel, and a demagnetization procedure.

The principal difficulty in making horizontal welds was the formation of relatively large internal voids when molten metal drained from the groove. The situation was somewhat improved by tack welding a shelf to the test pieces just below the beam location to support the liquid metal until it could solidify. Though the size of the voids was also influenced by beam oscillation variables, changes in amplitude, frequency, and pattern were not effective in eliminating the flaws. Use of a support shelf together with a beam angled downward 3° (0.052 rad) from the horizontal has been successful. Since the width of the fusion zone is only about 1/4 inch (6.4 mm), the angled beam technique will apparently require a 176° (3.1 rad) included angle bevel-and vee-groove joint preparation rather than the square butt originally envisioned.

Variations in parameters have also eliminated problems with root porosity, centerline solidification voids, and transverse fusion-zone cracks. However, the presence of a relatively small lack-of-fusion defect at the point of deepest penetration (called a necklace defect) is reported. Further modification in current, travel speed gun standoff, focal location, and oscillation pattern are expected to eliminate the necklace defect. Despite formation of a broad heat-affected-zone [about 1/2-inch (13 mm) wide], no HAZ flaws were observed in a total of 95 welds.

Soft-vacuum conditions were abandoned when it was found that they provided no benefit and introduced problems with soot and erratic equipment operation. Latter work was performed with hard vacuum (10^{-4} Torr or less).

Calculations showed that a magnetic field of 4 gauss was sufficient to deflect an electron beam by 0.1 inch (2.5 mm) after 4 inches (100 mm) of travel in a vacuum. Since this amount of deflection was unacceptable, it appeared that demagnetizing of plates might be necessary. However, experiments showed that within the metal, at fields to 300 gauss, no deflection occurred and reduction in residual field would not be required. Above the surface of the joint the presence of a magnetic field was important. A small (1 gauss) field was sufficient for improper weld placement. A Mu metal shield cone was developed that will prevent magnetic interaction with the electron beam in fields to 40 gauss.

At the Heavy Section Welding Symposium on April 20, 1979, at Oak Ridge, TN, Weber reviewed his program and presented further information showing two macrographs of defect-free sections from 8 inch (200 mm) thick welds in A387 steel. The plate was welded from both sides without preheat, using a beam power of about 30 KW. The travel speed for the joint, including both passes, was 1 inch (25 mm) per minute.

In a more recent summary, at the Triple Engineering Show in Chicago on November 14, 1979, Weber showed excellent welding results with all process techniques successfully resolved. However, agglomeration of base-metal inclusions in the autogenous weld metal produced detectable defects. Hence, special base-metal acceptance criteria may be required for electron beam welding.

- At the Heavy Section Welding Symposium held April 20, 1979, at Oak Ridge, TN, Farrell briefly reviewed the operation of a portable electron beam system being developed in France through a cooperative effort involving Sciaky and six international concerns. The welder rated at 100 KW has a 12 inch (305 mm) diameter gun that is 30 inches (760 mm) long which operates at a pressure of 10^{-5}. The gun is contained in a partial-pressure chamber at 5×10^{-3}. Though the attempt has not been made, the aim is a single-pass weld in 12 inch (305 mm) thick nuclear pressure vessel steel. The unit has been employed to weld a 4 inch (100 mm) wall, 10 foot (3 m) diameter vessel of 304L. The joint is automatically tracked, and no mismatch correction was employed for the initial attempt. A gun to work distance control will be required. Two TV cameras are employed to monitor the progress of welding. The gun was mounted inside the vessel which was fitted with a temporary cover. External back-side shielding was provided using silicone-rubber gaskets made in the shop. The circumferential seam was welded in the horizontal position at 4 inches (100 mm) per minute and completed in 1 hour. A total of five such joints have been made.

Slides were shown of sections from test welds in 5.2 inch (130 mm) thick A533 Gr. B and 9.8 inch (250 mm) thick A387 Gr. 22 Cl. 2 steels, welded in the horizontal position. The A387 Steel was welded at 6 inches (510 mm) per minute with a 0.8 inch (20 mm) beam. Mechanical properties, or the presence or absence of defects were not discussed.

Electroslag Welding

- The results of electroslag welding tests conducted with a 6 1/4 inch (159 mm) thick plate of 2 1/4 Cr-1 Mo steel are furnished by Brown, Rege, and Spaeder[101].

The composition of the filler wire used was similar to that of the base metal though slightly lower in sulfur and silicon and higher in manganese. Comparison of the average width of the weld [about 3 inches (76 mm)] with the original 1 3/8 inch (35 mm) gap used in welding shows that the weld deposit was approximately 45% filler wire and 55% base metal. No defects were found in the weldment either by sectioning or by radiographic examination.

The weldment was made without preheat using normalized and tempered plate with gas-cut edges. Two electrodes were employed, each with an alternating current of 1600 amperes and 50 volts. After welding the weldment was cut in half and one portion was quenched from a temperature of 1700°F (927°C) and tempered at 1275°F (690°C).

In the as-deposited condition the strength of the weld metal was high; 125 ksi (862 MPa) and 158 ksi (1089 MPa), respectively, for the yield and ultimate. However, the 75°F (24°C) energy was only 5 ft-lbs. (6.8 J) with 2% shear. The normalized and tempered base plate had a yield strength of 74 ksi (510 MPa), an ultimate of 95 ksi (510 MPa), and an absorbed energy of 83 ft-lbs (572 J) at 50°F (10°C).

When the weldment was reaustenitized at 1700°F (927°C) and tempered at 1275°F (690°C) the yield and ultimate strength of the weld metal and parent plate were identical; 78 and 97 ksi (538 and 669 MPa), respectively. The 75°F (24°C) Charpy energy values for the weld deposit, grain-coarsened HAZ, and base plate, respectively were 108, 147, and 125 ft-lbs (146, 199, and 196 J) with 100% shear failures.

- Mandich, Fagleman, and Gulya[102] have reported properties of an electroslag weld in 7 13/16 inch (198 mm) thick plates of quenched and tempered 2 1/4 Cr-1 Mo steel. The weldment was made with two-wire AC equipment using Arcos V flux. Current and voltage ranges for each wire were 600-680 amperes and 42-50 volts, respectively. The chemical composition of the filler wire was not given. However, the carbon content of the weld deposit was 0.09%, 0.05% less than the parent plates, indicating that a low-carbon wire was employed. The weldment was austenitized at 1750°F, water quenched, and tempered at 1125°F.

The room temperature yield and ultimate strengths of the 7 13/16 inch (198 mm) thick plate was 95.7 and 113 ksi (660 and 780 MPa), respectively. The weld metal exhibited yield and ultimate strength values of 91.4 and 108.3 ksi (631 and 747 MPa).

Tensile tests of the weld metal at 850°F (454°C) showed a yield of 76.5 ksi (528 MPa) and an ultimate of 85 ksi (587 MPa). The corresponding values for tests at 900°F (482°C) were 75.3 ksi (520 MPa) and 81.5 ksi (562 MPa).

The Charpy energy values at 10°F (-12°C) for the plate, grain-coarsened HAZ and the weld deposit were, respectively, 25, 48, and 40 ft-lbs (34, 65 and 54 J).

Narrow-Gap GMA and GTA Cold-Wire Welding

- Ducrot, Koffel, and Sayegh[93] have reviewed the development and progress of narrow-gap techniques and describe multiple-torch equipment capable of providing increased deposition or positioned welding. The authors note that early interest in all-position capability for narrow-gap welding resulted in two difficulties:

  - A tight restraint on the maximum gap width leading to the use of small, fragile welding heads (torch, sensor, gas nozzle).

  - Problems with weld repairs because of the difficulty of introducing tools into such narrow grooves.

Closer examination of general requirements led to appreciation of the fact that positioned welding is seldom required and extremely small gaps [0.3 inch (8 mm)] impose serious constraints on the process. Subsequent designs for narrow gap welding in thicknesses in excess of 4 inches (100 mm) expanded gaps to 0.6 to 0.8 inches (16 to 20 mm). The new direction had the following advantages:

- Relaxation of tight gap tolerances.
- Sturdier torch and gas supply equipment.
- Introduced the possibility of bending the filler wire, to direct the arc, near the point of deposition rather than outside the workpiece.
- Ease of repair.

Since the larger gap required more metal to fill it, the newer developments were led in several directions.

- The use of a welding head with three torches, which because of small bead size, could be operated in all positions.

- Use of a welding head with one or two oscillating torches for applications where welding in all positions was not required.

- Oscillating the arc and use of CO_2 gas shielding to obtain better operating conditions.

The three-torch equipment is similar to the first narrow-gap machines, with an additional axial torch that deposits the feed wire between the lateral beads. Sturdy torches with curved tips enable the feed wire to be directed to the desired deposit points to fill the gap with a three-bead layer.

Increasing the groove width has also permitted development of a single torch system that uses arc oscillation and wire bending near the point of deposit. The deflection can be as much as 20° (0.35 rad) insuring good fusion between the parent metal and the deposit. The addition of a motorized device which controls the rotation of the torch about the axis of the straining system permits arc oscillation[94]. The welding cycle can comprise either a constant welding speed with an adjustable dwell time for the wire which is a function of the arc position on each lateral surface, or a variable welding speed. In the latter instance, incremental forward movement of the torch occurs only when the arc is directed toward the groove sidewall. The welding speed is zero when the torch is rotating. The combination provides a very flexible oscillating method for general purpose applications. Joint tolerances, however, are so tight they are difficult to obtain in heavy industry and the welding speed is slow.

To speed the process, a second torch can be added and, if they are each adjusted to cover two-thirds to three-quarters of the gap, the gap tolerance problem is also overcome. The shielding gas is normally 80% argon and 20% CO_2. This proportion reduces splatter which is a function of the impurities in the gas and the welding parameters. If the parameters are optimized, two torch welding can be done with CO_2 shielding with excellent fusion between beads and the parent metal. Under certain conditions, depending on the metallurgical quality demanded of the joint, it is possible to weld pieces that have been flame cut.

Practically, increases in gap width does not mean that the welding time must increase correspondingly. The addition of a supplementary torch to a machine is not expensive because the main element of the machine remains unchanged and the diameter of the filler wire can usually be increased so that deposition rates of the same order as those for submerged-arc welding may be obtained without encountering the porosity, bead-to-bead fusion problems and high-heat input difficulties associated with the latter process.

A choice of technique must be guided by the specific applications envisioned. However, the newer trends in design for welding of wider gaps make narrow-gap equipment more reliable, sturdier, and industrially justifiable.

- Narrow-gap welding with twisted filler wires, under investigation at Kobe Steel, has been reported by Kimura, Nagai, and Kashimura[91]. The technique was developed to overcome difficulties with lack-of-fusion defects at groove sidewalls and to extend the range of welding conditions within which sound welds could be obtained. Only flat-position welding using Ar + 20% CO_2 shielding is discussed.

High-speed motion pictures were used to study the rotational movement of the arc and melting characteristics of the wires. When the twisted wire is composed of two wires of different diameters, the arc deflects in the axial direction of the larger wire. The smaller wire supports the arc less frequently and the arc from it is absorbed by that from the larger wire. Consequently, as the wires are melted the arc rotates in a continuous manner with the droplets from the smaller wire integrating with those from the larger diameter wire. In this instance, the small diameter wire acts only as a filler.

When equal diameter wires are intertwined, the arc is generated alternately from the tips of both so that rotation of the arc is discontinuous. Rotation is observed at low welding currents where globular transfer occurs, but the rotation is less smooth than that observed in the spray transfer range.

The radius of gyration increases with a decrease in pitch of the twist. For the same pitch the speed increases with increase in the welding current, and the radius of gyration is greater with increase in arc voltage. Arc rotation is not clearly observed when three wires of the same diameter are included in the twisted electrode.

Twisted wire gives about 10% greater deposition than a single wire of the same cross section. The increase is attributed to the greater over-all stickout distance and increased resistance heating.

The weld bead produced by a twisted wire has less total penetration than that deposited by a single wire but has increased penetration at the groove sidewalls. The technique eliminates sidewall fusion defects in full-width welding without mechanical weaving or modulation of current or voltage. Optimum conditions employ a machined straight-wall groove 0.55 to 0.70 inch (14 to 18 mm) wide, and a twist electrode consisting of two 0.078-inch (2 mm)-diameter wires. Travel speed is 12-14 inches (300-350 mm) per minute. Test pieces from welds to 4.7 inches (120 mm) thick, in A387 Gr. 11, A387 Gr. 22, and A516 Gr. 70 have shown excellent performance.

The narrow-gap system at Kobe Steel is video monitored and gated to detect the width of the weld groove at the arc location and to automatically adjust the travel speed when the groove width varys. If the welding current and voltage are maintained constant, and the product of the groove width and welding speed are constant, the height of weld metal deposited is uniform even when the groove width changes. Experiments show that the variation in thickness of the deposit is less than ± 0.008 inch (± 0.2 mm) for changes in groove width from 0.5 to 0.70 inch (12 to 18 mm).

Electrode centering within the gap is also monitored and automatically controlled with an accuracy of 0.02 inch (0.5 mm).

● In a recent paper[89], Sawada, Hori, Kawahara, Takao, and Asano discuss the development of both metal-inert gas and tungsten-inert gas narrow-gap welding and the commercial applications of these techniques by Backcock-Hitachi. Narrow gap welding, introduced in the Kure Works in 1976, now accounts for 80% of butt joints in products over 3/4 inch (20 mm) in thickness. The principal process is a single pass/one layer GMA technique using a 11/32-inch (9 mm) wide nearly-square groove, 3/64-inch (1.2 mm) diameter filler wire, and a pulsed power source. Fusion to sidewalls is facilitated by mechanical arc oscillation and by controlling the shape of the molten metal surface by streams of shielding gas consisting of an 80% argon-20% carbon dioxide mixture. The electrode is fed through an insulated guide tube by rollers. Prior to entering the rolls, the wire is deformed into a wave shape by a "flapping plate". Primary shielding is supplied at the tip of the torch which is less than 7/32-inch (6 mm) wide; secondary shielding gas from the top of the groove.

Development of the flapping-plate electrode-bending mechanism led to the first practical applications. Following this, higher deposition dual-electrode equipment and portable single-wire equipment for out-of-position welding were devised The latter two types use a pair of bending rolls to deform the electrode rather than the flapping plate. The waved electrode is almost straightened as it passes through the contact tube but springs back as it emerges. Consumption of the electrode causes the arc to oscillate from one side of the groove to the other. Oscillation amplitude is controlled by the amount of wave introduced into the wire and by the stand-off distance. In production welding, the oscillation frequency is adjusted to between 60 and 80 cycles per minute. Adequate fusion to sidewalls can be maintained even if the groove gap changes from 9/32 inch (7 mm) to 19/32 inch (15 mm). The roll-bending device used on the dual-wire and portable equipment is more compact and convenient to install. The flapping-plate design provides the best oscillation control.

Additional control to avoid undercutting and improved sidewall fusion is obtained by altering the shape of the molten pool using the stream pressure of the primary shielding gas. Gas directed toward the rear of the pool produces a concave shape but tends to push the puddle forward, in the direction of welding. Consequently, a second stream at the front of the puddle is required to produce a stable configuration and an ideal contact angle at the sidewalls of the groove. Formation of a suitably concave bead surface prevents undercutting and reduces the need for repair. When unsatisfactory contours are observed during welding, the regions are dressed with air grinders or rotary files. Defects found after completion of welding are removed by arc or gas gouging and repairs are made with shielded metal-arc deposits. Fewer flaws are encountered than are typical with conventional submerged-arc techniques.

Tungsten-arc welding equipment for welding in narrow gaps apparently was developed only for applications, such as pipe welding, where it can be used for the initial passes to obtain superior control of penetration and root fusion.

Problems with "arc-climb" at groove sidewalls were overcome by using a low-voltage pulsed-current power source. Splatter and interruption of welding made the use of a short-circuiting arc technique impractical. The spray-transfer, pulsed arc is stable and the typical wine-glass shaped penetration problem was overcome by arc oscillation.

The thickest section joined with the narrow-gap method is apparently 12 inches (300 mm). The joint required 69 passes with each pass deposited at a maximum speed of 10 inches (250 mm) per minute. Excluding interruptions, the full section could, therefore, be welded at a rate of 1.4 hours per foot (300 mm). Another example, a comparison of 4-inch (100 mm) thick welds made with narrow-gap and submerged-arc processes gives elapsed times of 3.5 hours and 8 hours, respectively. The process is considered to be superior to submerged-arc welding from standpoints of lower heat input (about half), reduction in the width of the heat-affected-zone, reduced dilution of the weld metal by base metal addition, and lowered diffusable hydrogen content. The hydrogen level is below the critical value for delayed cracking in 2 1/4 Cr-1 Mo steel. Therefore, it is permissible to lower the preheat temperature and intermediate post-weld heat treatments may not be necessary. Studies to establish the need for intermediate PWHT and more reasonable preheat and interpass temperatures are underway.

The narrow-gap process has been employed with a variety of steels including SA-516 Gr. 70, SA-299, SA-387 Gr. 22 Cl. 1, and SA-533 Gr. B Cl. 1. Mechanical tests indicate that the properties of the joints are equal to or superior to those of submerged arc welds and that the narrow-gap GMA process is suitable for nuclear pressure vessels.

Groove preparations employ either a backing bar or a double U-groove at the extreme bottom of the joint. When a backup is employed, chipping is required. Back chipping is not required with the U-groove geometry. Because of the small size and location of the lands, the root face can be completely penetrated by a subsequent GMA or SA bead. Standard practice is to maintain the weld gap between 5/16 and 25/64 inch (8-10 mm) at the position being welded. To maintain these limits, the groove has an initial bevel of 0.6 to 1.5 degrees (0.011 to 0.026 rad).

Continuous operation (in circular girth welds) of about 4 hours duration is possible before tip cleaning is required. A welder controls the operation by visual observation either directly or by means of an arc monitor.

For welding outside the shop, where turning rolls or positioners are not available and out-of-position welding is often required, a portable unit capable of both horizontal and vertical welding is employed. Application includes vertical-up welds in 6 1/2-inch (162 mm) thick steel. To make on-site welding easier, a seam-tracking sensor was developed. For shop welding, visual observation is preferred.

The economic advantage of GMA narrow-gap welding increases with increasing plate thickness. The advantage differs with different shops but the break-even point for comparison between narrow-gap and submerged-arc welding occurs at a thickness of about 1 inch (25 mm).

Electroslag welding is more economical than narrow gap welding of extremely thick plates. However, since ESW welds require normalizing after welding to improve notch toughness, the cost, including stress-relief, is lower for the narrow gap process.

- Use of pulsed narrow gap orbital pipe welding is described by Hill and Graham[92]. The all-position capability required for pipe welding eliminated all processes except metal inert-gas, metal active-gas, and tungsten inert-gas. Of these, the TIG cold-wire process was considered best suited for obtaining the weld quality required for joining 1/2 CrMoV piping with 2 Cr-1 Mo filler wire.

The welding system is capable of welding fixed pipe in wall thickness to 6 inches (150 mm) and from 3 to 25 inches (180 mm to 650 mm) in diameter. The groove employed has nearly parallel sides and is 0.43 inch (11 mm) wide. This equipment is approved by the CEGB for welding steam turbine high-pressure pipe.

Metallurgical examinations and mechanical tests show the properties of the welds are equal to and in most instances superior to welds made using manual metal-arc or submerged-arc processes. Under standard conditions the heat-affected zones show only a small percentage of unrefined structure, except in the region under the capping pass. This region can be refined by TIG remelting using a 30% reduction in heat input and the same pulsing conditions.

Residual stress measurements of pipe welds revealed that the as-welded surface stresses were compressive in both the axial and circumferential directions whereas they were tensile in similar manual metal-arc joints. After a 12 hour stress relief at 1290°F (700°C) the stress levels were essentially the same.

This process is now used on production welding of both plain-carbon and low-alloy steel pipe work and is being marketed.

Submerged-Arc Welding

● Bunn and Berger[35] describe a modification of the usual technique for submerged-arc welding of the cylindrical shells of thick-wall nuclear reactor vessels developed by Combustion Engineering. With this procedure, called "Sub-Vert", the joint is a square-butt type positioned with its axis vertical. The groove opening, only about 1-inch (25 mm) wide is sufficient to accept a submerged-arc welding wire guide which extends into the groove to deposit successive passes in the flat position through the thickness of the plate. The width of the gap is bridged with two passes, directed against the sides of the groove, that together constitute a weld layer about 5/32-inch (4 mm) thick. The operator initially aligns the welding nozzle and starts the weld. At the end of the bead, deposited away from the operator, the nozzle retracts and automatically repositions upward for deposition of the next bead. The major application has been welding of the longitudinal seams in reactor vessels in thicknesses of 9 to 12 inches (228-305 mm). The largest assembly[90] was a cylindrical 22-inch (560 mm) thick closure head for a liquid-metal reactor made by joining two large plates.

The process has the following advantages:

- The square-butt preparation is easier to make than the double-U groove usually employed for reactor shell seams. If machining capacity is not available, plates can be prepared by oxygen cutting and grinding.
- Fit-up and preparation for welding is simplified since the shells are not moved from the vertical position following fit-up.
- All seams (usually two to four per shell) can be welded simultaneously. This technique eliminates the need for rotating shells to permit inside and outside welding to control radial distortion.
- Preheating of stationary shells is simplified as is maintaining temperature within the desired range.

- The investment for equipment is approximately 50% of the investment for conventional equipment of equivalent production capacity.
- The technique uses small beads which by successive refinements improve the properties of both the heat-affected zone and the weld deposit.

A manual shielded metal-arc welding technique, called "Up-John", similar in most respects to the Sub-Vert technique is also used by Combustion Engineering for welds to 12 inches (305 mm) in thickness. The principal application is the welding of torus heads in thicknesses to 5 inches (127 mm). Special electrodes 22 to 25 inches (560 to 635 mm) long and usually 1/4 inch (6.4 mm) in diameter have been developed for this application to avoid the necessity of stops and starts within the narrow groove and the hazard of porosity. Arc strikes occur only on a runoff tab. For some applications electrodes have been made with excentric coating to produce a preferred arc direction and insure good sidewall fusion.

- The cost of a six-man Up-John station is 1/50 that of a comparable submerged-arc station.
- Four or five times as much set-up and welding time is required for conventional submerged-arc welding.

Both the sub-Vert and the Up-John methods employ somewhat lower heat input than conventional single-or dual-wire submerged arc processes. No difficulties have been encountered with inspection or maintenance of acceptable mechanical properties. The principal advantages are a shorter schedule, reduction in the handling of extremely heavy components, and reduced capital expenditure for a given production rate.

● Arnold[97] has described field welding of girth joints (horizontal position) in large storage tanks using the automatic submerged-arc process. The technique was first employed by Chicago Bridge and Iron Company in 1950, and, in essentials, remains unchanged. The equipment is fully automatic consisting of a variable speed carriage that supports two welding heats, positioned on both sides of the joint, and flux-holding shelves located just below the weld groove. The lower plate has a square preparation. The upper plate is double chamfered at about 45° (0.79 rad). The carriage which also supports guiding devices, reels, flux holders, and electrical controls, straddles the upper plate and uses the accurately machined upper edge as a track. The inner and outer welding heads may be positioned directly opposite each other or several inches apart. The head spacing and the welding current determine the depth of penetration of the weld deposit. Provision for preheating is provided by torches that precede the welding heads.

For tank walls in the thickness range 1/4 to 1 1/2 inches (6.4 to 38 mm) welding may be performed by depositing one to three beads from each side of the joint. Aside from the time advantage when compared to shielded metal-arc welding, the submerged-arc technique provides less risk of hydrogen pickup or underbead cracking.

<u>Westinghouse Narrow-Groove Welding</u>

● Development of narrow-groove welding techniques at the Westinghouse Electric Corp. have been directed to exploiting the superior weld quality and economic advantages affected using narrow joints for making thick-section welds on heavy equipment such as reactor vessels, steam generators, and large gate valves. These assemblies, in sections to as much as 12-inches (305 mm) thick normally require extensive preweld machining, and the deposition of large volumes of metal requiring hundreds of hours of welding time. Productivity improvements of four times the present rates have been demonstrated for specific heavy-section manufacturing applications. Production applications have been developed for both gas metal-arc welding and for gas tungsten-arc welding with a hot-wire feed. Further improvements and refinements in the processes are anticipated from studies directed toward modification in welding parameters for various metallurgical systems, improvements in shielding design, modification of filler-metal chemistry, arc pulsing and oscillation, wire positioning, and metal-transfer modes.

Weld quality and process reliability have been maintained by parametric optimization and precision process controls. Non-contact feedback systems have been developed for both proximity and seam tracking to automatically adjust the torch position in response to variations or changes in contour of the weld surface. The use of voltage-current slope variability and a pulsed current input system provide the flexibility of altering the weld pool size to accommodate various weld positions, joint geometries, and filler-metal base-metal combinations.

Studies have demonstrated that the processes can be used with gaps as narrow as 3/8 inch (9.5 mm) yielding a substantial savings over conventional submerged-arc practice in thick section welding. In one instance, comparison showed a 33% reduction in pre-weld machining time, a 70% reduction in welding time, and a 90% reduction in the total amount of weld metal deposited.

Technical advantages include:

- Low heat input
- Development of a narrow region of heat-affected microstructures
- Controlled dilution from the base metal
- Efficient transfer of elements from the filler metal to the pool, assuring better control of deposit composition
- Reduced distortion
- Refining and retempering of weld metal and heat-affected-zone microstructures.

Weld quality and reliability is enhanced by a fully automated process which provides a low defect rate and by the absence of fumes or dust which permits access for closed-circuit television monitoring of the welding operation. Weld repairs, when required, are no more difficult to perform than for other processes.

Seam tracking is obtained with a torch-mounted eddy current sensor that compares and compensates in accord with the intensity of signals received from parallel current-conducting tubes affixed to the surface of the weldment on either side of the narrow groove. The non-contact eddy-current tracking system was designed and adopted following evaluation of electro-mechanical systems which generally worked well in wider joints but were subject to failures in narrow grooves due to geometrical constraints. Arc proximity control is provided by an automatic voltage control with fully adjustable voltage/ amperage response.

Westinghouse Narrow Groove Hot-Wire GTA Welding

Narrow-groove hot-wire tungsten arc welding may be performed in the downhand, horizontal and vertical positions with or without arc oscillation. The equipment employs a high-reliability solid-state microprocessor, in an isolated chamber supplied with filtered, humidity and temperature controlled air. The unit provides a preprogrammed welding sequence with continual, in-process recording of welding variables

Extraordinary measures have been employed to assure the quality of the inert shielding gas. The supply tubing system is of stainless steel and employs a positive pressure design incorporating a mass-flow meter and oxygen and moisture monitors that detect changes in gas purity between the supply manifold and the torch itself. Metal-diaphragm pressure regulators have been installed to avoid problems with moisture absorption in rubber diaphragm designs.

The process has been investigated for joining A553 Gr. B Cl. 2, A387 Gr. 11, A387 Gr. 22 Cl. 2, ASTM 4140, and both 304 and 405 stainless steel in thicknesses ranging from 4 to 8 inches (100 to 200 mm).

- The hot-wire GTA process is generally regarded as a high-quality technique limited in applications to relatively small assemblies. Therefore, Westinghouse experience with extensive use of hot-wire GTA welding for massive, thick-section weldments is unusual. The process has been employed exclusively at the Pensacola Division since 1969 for all major welding of stainless steel nuclear reactor internals. In total, about 84,000 pounds (38,100 kg) of Types 308 and 308L filler wires have been deposited in approximately 8700 feet (2650 m) of weld seams. The joint thicknesses have been in the range 3/8 to 6 inches (9.5 to 152 mm), with the majority between 2 1/4 and 4 inches (57 and 101 mm). Roughly 45% of the welds have been made in the horizontal position. Weld quality has been excellent. Fewer than a dozen localized areas failed radiographic acceptance criteria and required repair.

In March of 1976, following development and refinement of narrow-groove welding techniques, this concept was introduced into reactor-internals fabrication. Presently, all welds, regardless of thickness, are made with the narrow-groove GTAW-HW process.

Since October of 1978, the narrow-groove HW-GTAW technique has also been employed to weld low-alloy, high-strength steels used to fabricate thick-walled nuclear steam-system pressurizers. Using a 1400 cubic foot (40 m^3) pressurizer as a typical example, sixteen major pressure retaining joints are made with the process, replacing a combination of submerged-arc and shielded metal-arc welds. Each pressurizer requires approximately 1200 pounds (544 kg) of deposited weld metal, or some 40 miles (64 km) of filler wire.

These examples clearly indicate the scope and success of HW-GTAW application within Westinghouse and a commitment to expanded use of the process for flat, horizontal, and vertical welding in all materials.

- Most filler wires in existing specifications intended to cover both GMAW and GTAW processes were specifically designed for GMA welding with CO_2 shielding gas and do not effectively meet the needs of those who employ them for more critical GTA welding. A filler wire for use with CO_2 shielding must be enriched in composition to compensate for reduced strength caused by losses in carbon, manganese, silicon and other deoxidizers. When such wires are used for GTA welding with inert-gas or inert-gas/low oxygen mixtures, the compositions of wires and deposits are essentially unaltered. These weld deposits can be much stronger and much less ductile than desired. Furthermore, the permissible maximum values for sulfur are generous. While the upper limit is seldom approached, heats with sulfur in excess of 0.015% are encountered much too often.

These problems are considered in a study conducted by Monley[110] at the Westinghouse Tampa Division, who evaluated the suitability of various filler wires for narrow-groove HW-GTA welding of SA-533 Gr. A Cl. 2 low alloy steel. The requirements for the weld deposits in the 1125°F (605°C)-24 hr. stress-relieved condition were 90-115 ksi (620-795 MPa) ultimate, 70 ksi (483 MPa) minimum yield and 50 ft-lbs (68 Joules) C_v minimum at 70°F (21°C). Among the filler wires examined, by deposition in 1-inch (25 mm)-thick plates of plain-carbon steel, were one sample of AWS 5.18 E70S-6 and three samples of AWS 5.18 E70S-1B. These wires are designed for GMA out-of-position applications and provide a minimum ultimate of 72 ksi (497 MPa) in the as-welded condition. The mild steel E70S-6 wire was judged to have marginal strength in the stress-relieved condition. All of the alloy steel E70S-1B wires exhibited properties that were acceptable, in this particular application, in the stress-relieved condition.

Comparison of the wire and the deposit chemistry for the various filler wires showed, as expected, high transfer efficiency for carbon, manganese, and silicon. The average values were 102, 91, and 101%, respectively. Though these wires must meet an as-welded minimum ultimate of only 72 ksi (497 MPa), the as-welded values for the three E705-1B deposits were appreciably stronger, ranging between 94.5 and 117.2 ksi (652 and 810 MPa). The 117.2 ksi value exceeds the minimum requirement by more than 45 ksi (310 MPa).

Monley also noted that analysis of the E70S-1B filler wires showed that a balanced composition of manganese and silicon was used for deoxidation without deliberate addition of special deoxidizers such as Ti, Zr, and Al. Though special deoxidizers are sometimes considered necessary for manual GTA welding, freedom from porosity showed these special deoxidizing elements are apparently not necessary with the HW-GTAW process.

A distinct feature evident in an etched cross-section of some of the HW-GTA weldments was the appearance of dark-etching spots, located near the center of individual weld beads, and found generally throughout the entire cross section. These spots were poorly resolved with the optical microscope. Using the scanning electron microscope at 2000X, however, examination revealed that each macroscopic spot consisted of a cluster of small inclusions. Microprobe traces and EDAX analyses of the regions showed Mn/S, Mg/S, and S alone, indicating that the spots were clusters of sulfide inclusions. No microstructural differences were observed, using the scanning electron microscope, between the regions containing the clusters and the adjacent cluster-free regions.

This observation was further studied by obtaining sulfur prints on sections of autogenous GTA weld beads on SA-533 Gr. A Cl. 2 steel plates. Longitudinal sections revealed intermittent alignment of sulfides in the welding direction. It is evident that in the absence of desulfurization by a flux, sulfur in GTAW deposits is rejected during solidification to the center of beads, to form precipitates along an intermittant linear path parallel to the length of the welds.

The fracture location in transverse tensile specimens and openings in transverse side-bend specimens were found to be related to the presence and location of sulfides when they were present. With a sulfur content of 0.020% in the weld metal, openings in guided side-bend tests caused by sulfides were only marginally within ASME Code limits. At 0.010% sulfur or less, the macroscopic sulfides were not evident and bend specimens performed well.

Since a filler wire specification for a maximum of 0.010% sulfur introduces complexities from the standpoint of both cost and availability, the compositions of all the filler wires used on the Tampa HW-GTAW program were reviewed to establish the likelihood of locating and procuring selected low sulfur heats. From these data it appears that about 40% of the heats melted to specifications that require a maximum of 0.035% S will contain less than 0.010% S and that procuring low-sulfur wires should not be a formidable problem.

In addition to a requirement for procurement of low-sulfur filler wires for narrow-groove HW-GTA welding, six special heats will also be prepared and evaluated in an attempt to obtain alteration in the shape and distribution of sulfides, and to provide some additional deoxidation. The heats will include a near-commercial composition with low sulfur and an addition of aluminum, a low-silicon material also fortified with aluminum, low-and high-sulfur heats for comparisons, a heat deoxidized with Ti and Zr, and a heat with rare-earth additions.

In summary, there is mounting evidence that separate filler-wire categories, or specifications, should be established for processes such as HW-GTAW that employ inert-gas shielding and have high transfer efficiency. Preliminary evaluation of 2 1/4 Cr-1 Mo HW-GTAW deposits at Westinghouse Tampa shows that ultimate strengths in excess of 100 ksi (690 MPa) can be obtained in the heat-treated condition. A leaner-alloy modification of AWS 5.28-79 ER90S-B3 composition including lower carbon and tighter control of sulfur and phosphorus is recommended[111].

● Use of narrow groove, hot-wire GTA welding in a demonstration at the Westinghouse Tampa Division of retubing of a full-size nuclear steam generator is described in a paper by Atkin and others[113]. In the final phase of this program, a model D-4 steam generator with over 9000 tube ends was dismantled and rebuilt under simulated radioactive control conditions in 62 days. The space limitations and obstructions within a typical reactor containment vessel were included in the mock-up area. The retubing process was structured as a systematic program encompassing design, development and testing, and demonstration phases. The necessary disciplines included guidelines for qualification of personnel, design and equipment control, specifications, handling, fabrication, field changes, and quality assurance to applicable portions of the ASME Code, Sections III, IX, and XI.

The shell of the 68-foot (20.7 m)-high, 350 ton (3.18×10^5 kg) vessel fabricated from SA-533 Gr. A Cl. 2 low-alloy steel was circumferentially parted 8 feet (2.4 m) below the upper elliptical-head weld seam and rewelded after the retubing operation. At this elevation the vessel drum had an outside diameter of 14.7 feet (4.5 m) and a 4-inch (101 mm)-thick wall. A circumferential track fitted to the drum, see Figure 1, carried two narrow groove machining and welding units which initially milled through the wall providing a groove preparation with a 7/16-inch (11 mm)-radius root and an included angle of 6 to 8 degrees (0.10-0.14 rad.). Special stabilization and alignment hardware installed within the drum held the severed section in place during cutting and enabled it to be accurately repositioned for welding.

The horizontal-position girth weld was made with AWS 5.18 E70S-1B (low S, low Si) filler wire using a preheat and interpass temperature of 225-250°F (107-121°C) maintained with resistance heaters. The welding operation is shown in Figure 2. Six days were required to deposit somewhat more than 70 passes to fill the joint. This is roughly one-tenth the time required for manual welding. Instrumentation for monitoring and controlling the welding operation was located in a welder control-station trailer, Figure 3, that provided an air conditioned

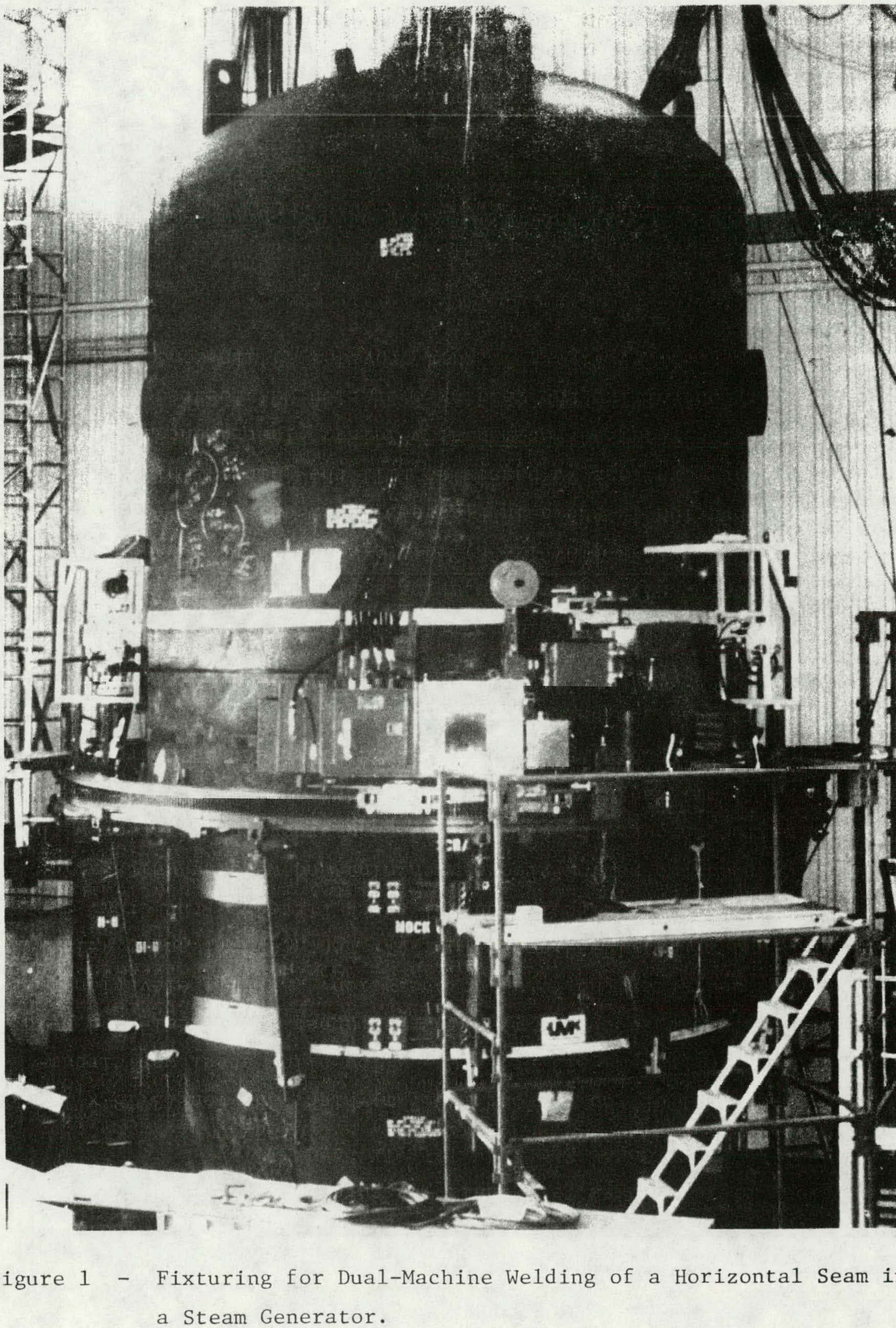

Figure 1 - Fixturing for Dual-Machine Welding of a Horizontal Seam in a Steam Generator.

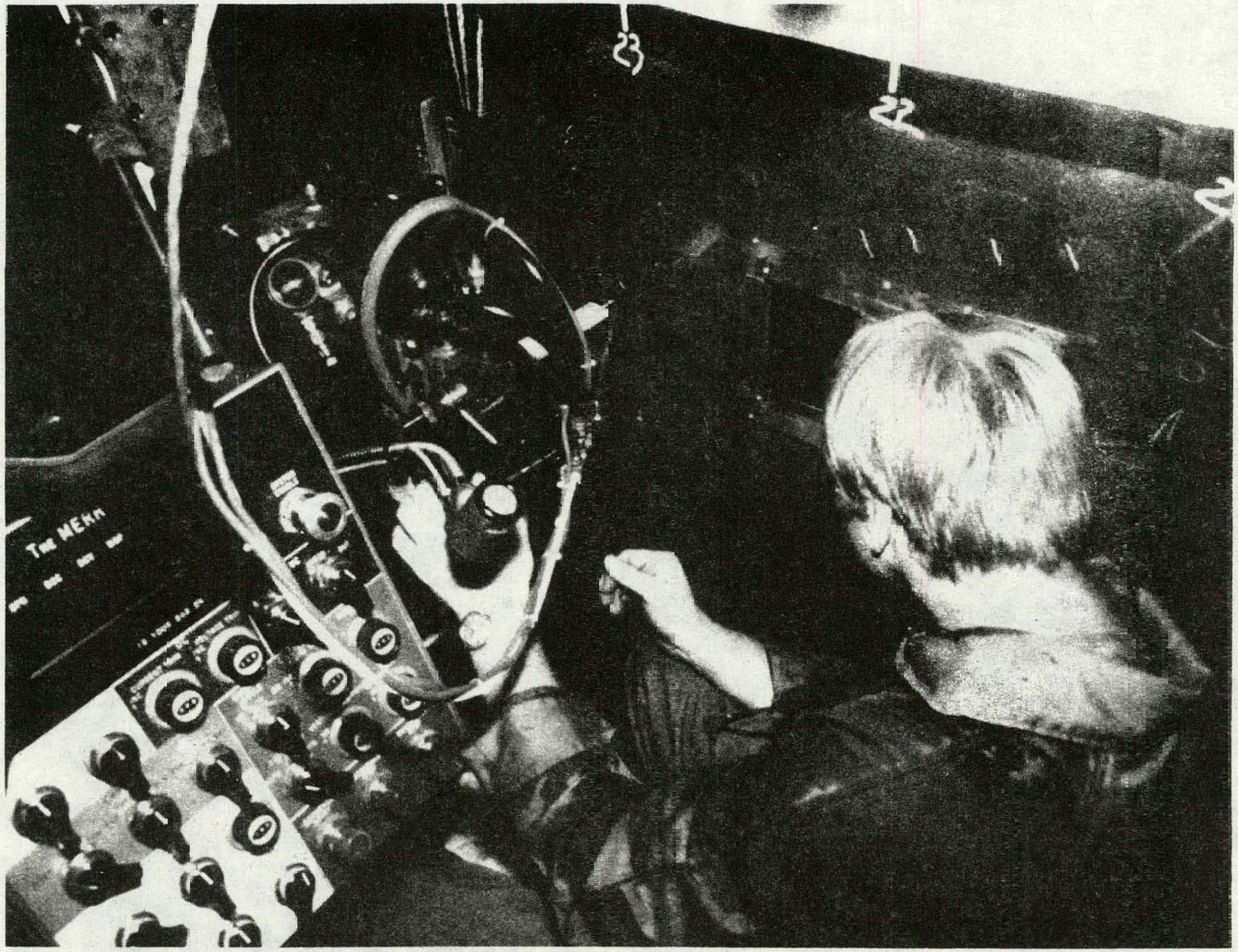

Figure 2 - Setup for Narrow-Groove Hot-Wire GTA Welding in the Horizontal Position.

Figure 3 - Trailer Interior Showing Controls, Television Monitor and Microprocessor.

environment with absolute atmospheric filtration. From the welder's station, complete remote operation of the welding system was provided from distances up to 250 feet (76 m). No repairs were required during the deposition process. Ultrasonic and x-ray inspections showed the completed weld was free of defects. The joint was stress relieved at 1125°F ± 25°F (607°C) for 2.5 hours.

- As part of a study of the welding of heavy-section Type 405 stainless steel, Friedman[112] investigated narrow-groove HW-GTA welding of 2-inch (51 mm)-thick plates and a 5.6-inch (142 mm)-thick forging using Type 410 Ni-Mo flux-cored filler wire.

A logical choice in filler metals for joining Type 405 stainless steel is Type 410 (0.12% C max., 11.5-13.5% Cr) since it provides a weld deposit that is similar in composition to the base metal. However, the 410 filler has a high hardenability, is prone to cracking, and requires high preheat and a post-weld heat treatment. The 410 Ni-Mo modification typically contains only 0.03% carbon and has additions of 4 to 4 1/2% nickel and 1/2% molybdenum. The nickel imparts substantially improved toughness while molybdenum serves as a solid-solution strengthener. The deposit is fully martensitic at room temperature, but compared to Type 410, is resistant to cracking even when employed for welding of heavy sections without preheat.

Three narrow-groove HW-GTA welds were prepared, two in 2-inch (51 mm) plate and one in a 5.6-inch (142 mm)-thick forged bar. All joints were double-U preparations with a 3/8 inch (10 mm) root radius and a 6° (0.10 rad.) included angle. The filler wire was 0.045 inch (1.14 mm) diameter. One of the 2-inch (51 mm) assemblies and the 5.6 inch (142 mm)-thick assembly were welded without preheat on one side and with a 400°F (204°C) preheat for the other side. The thicker weldment required a total of 50 passes. The remaining 2-inch (51 mm) weldment was prepared without preheat.

The weldments were inspected by a combination of longitudinal and shear-wave ultrasonics, x-ray, and both macro-and micro examinations. No weld metal or heat-affected-zone cracks, or general porosity, were detected. Fusion tie-in problems at the weld sidewalls were minimal though a few spots were found; largest approximately 0.15 inch (3.8 mm) in length. Macrographs showed the difficulty was a result of improper bead placement. Worm-hole porosity occurred locally at the ends of specimens at stop and start regions. The occurrence is not unusual since the rough and irregular contours in these areas were not removed by grinding.

Mechanical tests included weld-metal tensile, weld-metal Charpy impacts, and side-bends, conducted in both the as-welded condition and after a 2 hour stress relief at 1150°F (620°C). The stress relief treatment provided an improvement in ductility and excellent mechanical properties. Ultimate strengths ranged from 110 to 120 ksi (760 to 828 MPa), 0.2% yield strengths ranged from 80 to 90 ksi (552 to 621 MPa), C_v toughness values were uniform and averaged about 75 ft-lbs (102 Joules), and 6 of the 8 bend specimens bent 180° (3.14 rad). The remaining two bend specimens failed by tearing in the heat-affected zone of the Type 405 base metal which was appreciably less ductile than the weld deposit.

- Monley(114) investigated in-process monitoring of horizontal-position narrow-groove HW-GTA welds using remote video, radiographic, and ultraonsic techniques. He concluded that only the ultrasonic method will presently provide a high degree of confidence.

The use of television viewing was considered primarily for use in radioactive environments since remote monitoring may be the only practical means for visual control where radiation exposure must be limited. Both black and white and color video systems were developed and constructed for prototype use. The black and white system consisted of a thallium-iodide light source, a lens, a narrow-band filter, and a TV camera and display. This optical arrangement uses a narrow-band light source focused toward the arc in the weld groove. Reflected light is viewed from the opposite side through the narrow-band filter. The camera sees both the filtered light spectrum from the arc and the auxiliary illumination. Good visibility is restricted to the weld pool and the region immediately adjacent to the pool.

The color video system uses only light emitted by the welding arc. Light is transmitted through a fiberoptic bundle to a color camera and monitor. Attenuation of the output from the arc was obtained by use of a gradient-density filter which required accurate superposition of the darker portion of the filter over the brighter portion of the image. The inherent problem is the change in the shape of the arc during welding, causing misalignment with the dark area of the filter.

Transmission of the image to the monitor is a problem common to all systems. Access to the weld groove is restricted and installation of video equipment cannot be permitted to obstruct direct viewing. The equipment is somewhat delicate and must be protected from damage and exposure to preheat temperatures. Improved fiberoptics are helpful in this respect and close observation of the arc can be obtained using a close-up (zoom) lens. In addition to the technical difficulties associated with obtaining a clear and detailed image, the TV monitor does not supply sufficient 3-dimensional information under dynamic conditions. Zoom lenses and multiple images from different angles are, to date, the best methods for enhancing perspective. Actual remote operation for field use through a monitoring system has not been developed and will require further testing and evaluation to establish its full capability.

In-process radiography was attempted during welding of a large circumferential weld seam using an Iridium source outside the shell to examine selected areas. Subsequent comparison of final radiographs and the in-process radiographs revealed an inherent difficulty with the in-process technique. The difference in the thickness of the base metal and metal in the partially filled groove results in masking of defects at the weld-base metal interface. In a narrow-groove joint with near parallel sides, the use of in-process radiography is severely limited because the width of the weld deposit is the only region that can be examined. Since the majority of narrow-groove defects would be expected to occur at the weld-base metal interface, the in-process technique is inherently ineffective.

An in-process ultrasonic technique was developed to detect two specific defects, lack-of-fusion and porosity, that occur under certain conditions along the bottom wall of horizontal narrow-groove HW-GTA welds. The bottom bead in a (2G) narrow groove presents tie-in problems that require changes in torch angle to assure good penetration. Porosity in beads at the same location is indicative of a shielding problem controlled or influenced by the torch angle, depth of the groove, and the convective forces acting on the shielding gas flowing along the preheated wall. A 45° (0.79 rad) shear wave technique from the upper side of the weld was found to be the best method to detect these flaws with maximum signal response. A high-temperature couplant was employed on the 250°F (121°C) preheated surfaces. During in-process UT tests on a 4-inch (101 mm) thick mock-up weld, conducted after each 3/4-inch (19 mm) of weld metal was deposited, three flawed regions were detected and repaired. Radiography, after the weld was completed and stress relieved, showed no defects.

ON-SITE FABRICATION OF THICK WALL-VESSELS

● Though Wakefield[87] in a review of field fabrication of large vessels refers to field-rated automatic MIG and TIG welding (Graver Tank and Manufacturing Co.), automatic submerged-arc and manual shielded metal-arc welding are the processes that have been commonly employed. Submerged-arc welding is used for site-factory assembly. Manual metal-arc welding is employed, in most instances, for joining components during vessel erection where tolerance for fit up difficulties and positioned welding is required. The fact that both processes have a history of successful application, and are well understood, explains why the literature dealing with on-site fabrication provides little detailed information on the welding operation, while dwelling on the complexities of transportation, protection of personnel and equipment, heavy component erection, and machining.

Manual metal-arc welding is slow, and the field assembly of a large vessel demands many highly skilled welders; presently in short supply, with no improvement in the labor pool projected through 1985[19] Use of manual welding, other than the ability to handle poor joint fit up has been promoted by the lack of mechanized welding equipment considered suitable for use under adverse field-welding conditions, and the excessive cost of unreliable field-welding performance. Manual metal-arc welding equipment is inexpensive, readily portable and easy to maintain or replace.

It has already been established that mechanized welding has a high duty cycle and produces higher quality welds than are obtainable with manual welding. Reproducibility, and a significant reduction in the number of repairs needed, results in a significant reduction in total welding time. The incentive for the large-scale introduction of mechanized welding processes to on-site fabrication has existed for some time. Whether or not the potential for economic advantage can be exploited hinges more on developing confidence in equipment reliability than on any other item.

A partial list of site-assembled vessels in service since 1964, found in reference 100, for use in the petroleum, alumina, aerospace, paper, and chemical industries shows 19 heavy-wall assemblies including six fabricated with A387 steel. The largest of the 2 1/4 Cr-1 Mo vessels is a 13-foot (4 m) diameter x 70 foot (21 m) long refinery reactor having a shell thickness of 7 1/4 inches (108 mm). The design pressure and temperature are, respectively, 1600 psi (11 MPa) and 800°F (426°C). The 575-ton (521,600 kg) vessel was shipped in ten complete shell rings and two head sections. Defect-free welding (shielded metal-arc) of all eleven girth seams was demonstrated by gamma radiography and ultrasonic inspections. All weld seams were preheated and post-weld heat treated at 1150°F (620°C) in position.

- Bagni and Somigli[85] have described submerged-arc fabrication of a large petrochemical reactor made from A387 Gr. D steel that was shipped in two halves, assembled, and joined with a single shielded metal-arc girth weld in the field. The 12-foot (3.66 m) diameter, 62-foot (18.9 m) high vessel, requiring five shell courses of 4.4-inch (112 mm) thick plate was designed to operate at 610°F (320°C) at a pressure of 1820 psi (12.5 MPa). Design requirements, to reduce the wall thickness, called for a room temperature ultimate strength of 82.3 ksi (567 MPa) and a yield at 610°F (320°C) of 45.4 ksi (313 MPa) with 100% joint efficiency. These values are intermediate to those for normalized and tempered 387 Gr. D and quenched and tempered A542.

To obtain the desired properties, some portions of the bottom half of the reactor, including the longitudinally welded shell courses, were quenched and tempered after welding. For fabricating the courses, the procedure was:

- Hot rolling
- Seam welding at 445-480°F (230-250°C) preheat
- Normalizing. The normalized courses were placed in plate-bending rolls to be sized and cooled.

- Non-destructive examination
- Austenitizing and spray quenching
- Tempering
- Non-destructive examination

The technique avoided ovalization. The increase in diameter was never more than 0.2 inch (5 mm).

The upper and lower sections were delivered to the site and positioned vertically with supports. The bearing face between the two supports was fitted with a steel-aluminum-steel pad, about 1.20 inch (30 mm) thick to allow for contraction during girth welding. The juncture was made by a shielded metal-arc weld. Induction heating coils, directly above and below the joint, were used for preheating and for post-weld heat treatment.

- Details of the on-site assembly, by shield-arc welding, of a liner, diagrids, and gas baffle for an advanced gas-cooled nuclear reactor have been provided by Greener[86].

The liner which serves primarily as a leak-tight membrane to prevent seepage of reactor gas through the concrete containment vessel is a 62 foot (18.9 m) diameter, 63 1/2 foot (19.4 m) high vertical cylinder with a flat roof and floor is made from plate in the thickness range 1/2 to 1 1/4 inch (13 to 32 mm). The membrane liner was produced using shop prefabricated units comprising 16 full-height wall panels, 16 floor segments, and 30 roof segments.

The moderator graphite core of the reactor is supported by a diagrid, fabricated from 1 3/8 inch (35 mm) plate which experiences temperatures as high as 660°F (350°C) and is subject to both irradiation from the core and vibrations induced by the turbulance of the inlet gas. In plane view, the diagrid is a 16 sided polygon 39 ft (11.9 m) across the flats with structural members about 6 feet (1.8 m) deep. The size required much of the fabrication to be carried out on site. The unit was successfully welded and stress relieved using a specially designed electric furnace constructed around the assembly.

The gas baffle was in the form of a 1-inch (25 mm) wall circular shell, 45 feet (13.7 m) in diameter, fitted with a 2 5/8 inch (66 mm) thick torespherical dome. The baffle was assembled and stress relieved in a large covered tank.

- The field-fabrication experience of the Chicago Bridge and Iron Company in site-erection of nuclear power generation plants and other large facilities is reviewed in references 98, 99, and 100.

A 545 MWe boiling water reactor (BWR) is 17 feet (5.2 m) in diameter, 64 feet (19.5 m) high, has an average shell thickness of more than 5 inches (127 mm), and weighs 500 tons (4.5×10^5 kg). Site assembly is the only practical way to erect vessels of this size and weight at locations inaccessible by navigable waterways or where overland transport is either not economical or impossible.

BWR vessels are shipped to the site as 13 shop-fabricated major sections: The top head, four shell rings of two sections each, the bottom head knuckle section and a bottom head dollar plate. Each shell ring is formed from two plates, with the half-ring segments temporarily joined to permit stainless steel overlay welding and to insure correct dimensions and alignment prior to shipment. All nozzles are installed and clad in the shop. Subassembly at the site, by shielded metal-arc welding, of two or more components of sizes that can be handled by a 150 ton (1.36×10^5 kg) derrick reduces the number of sections to six. The bottom head is welded into a single assembly, each shell ring is completed, the upper flange is joined to the top shell ring, and the top head is welded to the head flange. These six pieces are then assembled within the containment vessel. About three-quarters of all assembly occurs at the site. The construction sites are equipped with several cylindrical steel igloos which provided protection from the weather and double as stress-relieving furnaces. The igloos furnish dust-free enclosures, which can be air-conditioned for temperature and humidity control as required. These enclosures are used not only in the assembly yards, but inside the containment vessels where some heat treating must be performed in the final location.

All welding procedures used on site were developed at the CB&I welding laboratory to insure structural integrity and dimensional tolerances. When welding is underway, each inch (25 mm) of deposit is magnetic-particle inspected and, upon completion of welding, the joints are examined ultrasonically and by radiography using gamma sources.

Preheating and intermediate stress relieving is obtained by utilizing either electric resistance heaters or gas burners, or combinations of both. Final stress relief of the girth welds between subassemblies in the containment vessel is performed at 1100/1175°F (590/635°C). Temporary baffles are installed below and above the joint on the inside of the vessel and the external surface is covered with insulation. CB&I has used similar stress-relieving techniques on hundreds of components and vessels where conventional furnace methods were impractical or not available.

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