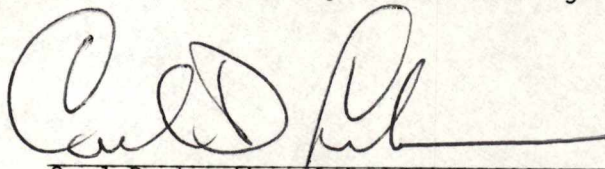


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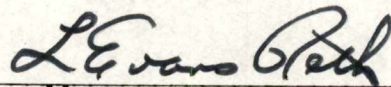
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HIGH TEMPERATURE PROPERTIES OF 2½Cr-1Mo
STEEL WELD METALS

A Thesis
Presented for the
Master of Science
Degree
The University of Tennessee, Knoxville

Michael R. Pendley

August 1980

3044924

DEDICATION

FOR LAURA ANN

In Loving Memory

"Secret gardens of the heart
Where the seasons change forever
I see you shining through the night
In the ice and snow of winter."

Collins

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ABSTRACT

$2\frac{1}{4}\text{Cr}-1\text{Mo}$ is a low-alloy ferritic steel used primarily for high temperature applications. While extensive data exists on $2\frac{1}{4}\text{Cr}-1\text{Mo}$ in wrought and cast forms, scant information is available on the high temperature properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal. Since welding is the primary fabrication method used with $2\frac{1}{4}\text{Cr}-1\text{Mo}$, the paucity of weld metal data is a matter of concern to the designers of pressure vessels constructed of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. The current study was initiated in response to the need for more weld metal data.

Stress-rupture tests were conducted at 850 F, 950 F and 1050 F on $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals deposited by the shielded metal arc (SMAW), submerged arc (SAW) and electroslag (ESW) welding processes. Room temperature and elevated temperature tension tests were also performed.

The stress-rupture strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal was shown to be a function of welding process characteristics, carbon content, and PWHT. The stress-rupture strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal was found to increase with increasing carbon content. Creep ductility in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal was shown to be independent of rupture time, but dependent on carbon content. Room temperature tension tests did not correlate directly with elevated temperature stress-rupture behavior, but a good correlation with stress-rupture behavior was observed with regard to the elevated temperature tensile properties.

The current study has shown the need for additional work in related areas, particularly the investigation of weldments with different carbon contents than those utilized in this study and of weldments produced by different welding processes, such as electron beam and gas tungsten arc welding.

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CHAPTER I

INTRODUCTION AND HISTORICAL REVIEW

$2\frac{1}{4}\text{Cr}-1\text{Mo}$ is a low-alloy ferritic steel used for pressure vessels, nuclear applications and pilot coal gasification plants. Since these applications call for service at temperatures significantly higher than 650 F, the creep behavior of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ is important to designers of such units. In addition, since welding is the primary fabrication method used with $2\frac{1}{4}\text{Cr}-1\text{Mo}$, the creep properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal are also of great importance.

The alloy with the nominal composition $2\frac{1}{4}\text{Cr}-1\text{Mo}$ is a high strength, low alloy ferritic steel used primarily for elevated temperature applications. This steel is noted for its excellent high temperature strength (imparted by the molybdenum content), oxidation resistance (imparted by the chromium content), and stability of microstructure during service.

As far back as 1940, $2\frac{1}{4}\text{Cr}-1\text{Mo}$ was singled out as one of the best low-alloy steels in terms of creep strength and microstructural stability (1). At the same time, new pressure vessel designs were calling for higher operating pressures and temperatures, necessitating the development of new materials with adequate strength at the higher temperatures. As a result, extensive research was done on the high temperature properties of Cr-Mo steels, including $2\frac{1}{4}\text{Cr}-1\text{Mo}$, leading to the publication of an extensive data review in 1953 (3). Specifications were prepared by ASTM for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ in a variety of forms. Normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ was given the specification SA 387 Grade D.

As pressure vessel designs began to call for materials with even higher strengths, industry investigated the use of $2\frac{1}{4}\text{Cr-1Mo}$ steel in the quenched and tempered condition. It was found that $2\frac{1}{4}\text{Cr-1Mo}$ could develop significantly higher tensile and yield strengths in the quenched and tempered condition, while maintaining adequate toughness. Subsequently, ASTM prepared specification A 542 to cover quenched and tempered $2\frac{1}{4}\text{Cr-1Mo}$.

The ASME Boiler and Pressure Vessel Code (85) has slowly allowed utilization of the higher strengths which can be produced in $2\frac{1}{4}\text{Cr-1Mo}$ by quenching and tempering. After several earlier code allowances, ASME prepared Division 2 of Section VIII which permitted the use of quenched and tempered materials with allowable stresses based on a safety factor of 3 on tensile strength (Division I of Section VIII calls for a safety factor of 4, but does not specify a stress analysis). However, Division 2 does not allow for the enhanced properties of quenched and tempered $2\frac{1}{4}\text{Cr-1Mo}$ above 650 F. Thus, while the ASME is preparing rules for elevated temperature service, utilization of the enhanced properties produced by the quenching and tempering of $2\frac{1}{4}\text{Cr-1Mo}$ is limited to non-creep conditions.

The creep behavior of $2\frac{1}{4}\text{Cr-1Mo}$ can be better understood by first considering the creep behavior of ferritic steels in general. The creep properties of ferritic steels are determined by the inherent strength of the matrix, the effect of carbide particles on the deformation on the matrix, and the stability of the microstructure. Superior creep properties are obtained with a fine dispersion of carbides. As carbide particles grow, they lose their ability to strengthen the matrix. Hence, a "perfect" creep resistant material would consist of a finely dispersed carbide phase which would not agglomerate at the service temperature.

One of the creep-resistance mechanisms mentioned above is that associated with the inherent strength of the matrix, or solid solution strengthening. Alloying elements which go into solid solution improve creep resistance by making cross slip and dislocation climb more difficult (100). The strengthening effect results from the distortion of the solvent lattice by the solute atoms. The magnitude of this effect will depend upon the amount of distortion produced by each solute atom and on the number of such atoms. Molybdenum is by far the most effective element in its ability to impart solid solution strengthening to ferritic steels. Unfortunately, it is not generally possible to produce very much matrix strengthening by this means since those atoms which produce great lattice strain (and thus high strength) have small solubility, and those which dissolve freely cause little lattice strain (17). This limitation may be overcome to some extent in the case of iron, since elements which are only slightly soluble in ferrite (i.e., carbon) are readily soluble in austenite. Hence, greatly supersaturated solutions can be obtained on quenching.

The creep of ferritic steels is primarily affected by the type and distribution of the carbide phase. Fine carbides are able to hinder the motion of dislocations and thus strengthen the matrix (99). Grain size has little influence on the creep properties of ferritic steels (17).

The carbide dispersions produced in ferritic steels are of two general types (100). The first type is a uniform dispersion of precipitates obtained by decomposing a highly supersaturated solution. The second type of dispersion consists of precipitates located mainly on dislocations, a type of dispersion obtained by decomposing a slightly supersaturated solution. The major limitation of carbide-dispersion strengthening is that

the carbide particles tend to grow under creep conditions, lessening their strengthening effects.

The creep resistance of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ will be influenced by the strengthening mechanisms mentioned above. The effectiveness of these mechanisms is in turn governed by the composition, heat treatment, and welding history of the material. Subsequent sections of this review will discuss these effects.

Compositional Effects on $2\frac{1}{4}\text{Cr}-1\text{Mo}$

Chemical composition plays a major role in determining the creep behavior of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel. Carbon acts with chromium, molybdenum, columbium, titanium and vanadium to form carbide precipitates which improve creep resistance. Other elements are deleterious to creep strength--these include aluminum, copper, nickel, tin and phosphorus. The effects of these and other elements will be discussed below.

Carbon content is an important factor in determining the microstructural characteristics of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Klueh and Canonico (66) studied the influence of carbon content on normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ base plate. They found that low carbon (0.003%) material consisted of proeutectoid ferrite and small spherical carbides. The microstructure of medium carbon (0.035%) material was similar, but with 5% tempered bainite. The high carbon (0.120%) material contained 55-65% tempered bainite with the balance being proeutectoid ferrite.

The carbon content also has a major influence on the mechanical properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel. Both tensile strength and stress-rupture strength have been observed to increase with increasing carbon content

(47,61,66). This effect appears to be temperature dependent. At 1050 F, the rupture strengths of the low and medium carbon steels mentioned above were similar, while that of the high carbon steel was much greater. However, at 850 F and 950 F, the stress-rupture strengths of the high and medium carbon steels were similar, while the low carbon steel was considerably weaker than both (67). The low strength of the low carbon material can be explained by the fact that it obtains little precipitate strengthening from the large, widely scattered carbides present in its microstructure.

To obtain optimum creep properties in ferritic steels such as $2\frac{1}{4}\text{Cr}-1\text{Mo}$, the carbon content should be between 0.1 and 0.2% (17). Such relatively high carbon contents allow more carbide particles to develop, thereby increasing the creep resistance of the material. However, lower carbon contents may be desirable in environments where decarburization is a potential problem. Also, low carbon materials will lose less molybdenum from solid solution (where it acts as a strengthening agent) due to carbide formation.

Molybdenum is an extremely potent agent in conferring creep resistance to $2\frac{1}{4}\text{Cr}-1\text{Mo}$ by means of solid solution strengthening (17). Molybdenum has long been recognized as the most effective solid solution strengthener in ferritic steels, especially at high temperatures. In addition, molybdenum contributes to the creep strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ by forming Mo_2C , a coherent M_2X -type precipitate which is widely recognized as the precipitate which contributes the most to creep strength (93).

The optimum level of molybdenum in ferritic steels appears to be around 1.0 weight percent. At this concentration, molybdenum saturates the

dislocation networks in ferritic steels (35). Greater concentrations of molybdenum lead to no further increase in creep strength.

Chromium acts to improve the oxidation resistance and microstructural stability of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ (1). Both of these properties are important for steels intended for elevated temperature service. Chromium also acts to resist decarburization of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ (38). However, chromium has a detrimental effect on creep strength. Thus, in determining optimum chromium levels for ferritic steels, a balance must be struck between the beneficial effects provided by chromium in terms of oxidation resistance and microstructural stability and detrimental effects in terms of reduced creep strength. A study of a number of different low-alloy steels found that an alloy with 2 $\frac{1}{4}$ % chromium had the highest creep strength while at the same time maintaining a stable microstructure and adequate oxidation resistance (1).

Titanium, vanadium and columbium are carbide formers. They increase the stress-rupture strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ by forming coherent precipitates, such as V_4C_3 and CbC_6 , in the ferrite matrix (71). These coherent precipitates contribute substantially to the creep strength of ferritic steels, apparently by precipitating on dislocation networks (35). Batte and Murphy (59) found that controlled additions of 0.03% titanium made to a basic $2\frac{1}{4}\text{Cr}-1\text{Mo}$ material increased stress-rupture strengths by 50% in short-time tests. Egnell (48) reported that additions of columbium to $2\frac{1}{4}\text{Cr}-1\text{Mo}$ produced a great response of the tensile and creep properties to heat treatment. In addition, titanium and columbium improve the resistance of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ to decarburization (38).

Unfortunately, the vanadium, titanium and columbium carbides coarsen with time and temperature and eventually lose their effectiveness in producing creep strength. Also, even though these elements are potentially of great effectiveness as solid solution strengtheners, in the presence of carbon they are almost entirely removed from solid solution to form carbides. Thus, the value of titanium, columbium and vanadium as solid solution strengtheners is greatly diminished (17).

Phosphorous, antimony, arsenic and tin promote temper embrittlement of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Temper embrittlement is defined as brittleness that results when certain alloy steels are held within, or cooled slowly through, a temperature range of 700 F to 1100 F (53). Temper embrittlement is evidenced by the shifting of the ductile to brittle transition temperature to higher temperatures. Temper embrittlement occurs only when certain alloy impurities are present, these being phosphorous, antimony, tin and arsenic. Since $2\frac{1}{4}\text{Cr}-1\text{Mo}$ is susceptible to temper embrittlement, the levels of these deleterious elements, especially phosphorous, should be kept as low as possible.

Manganese has several effects on $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Manganese is essential for good weldability in that it assists in the deoxidation of the weld metal. Manganese is known to prevent hot-cracking in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal. Bonta (111) found that manganese prevents the formation of iron sulfide inclusions which cause hot-cracking. North and Wallace (4) have stated that the hot-cracking susceptibility of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal can be controlled by having an Mn/S ratio of 40:1 in the weld. Manganese is also essential to ensure adequate toughness in the weld metal. However, Sachs et al. (40) found that excessively high manganese contents were detrimental to

creep strength. Also, Bruscatto (53) reported that $2\frac{1}{4}\text{Cr}-1\text{Mo}$ welds with low manganese levels were able to tolerate greater quantities of embrittling impurity elements without developing embrittlement. Bonta (111) has also stated that high manganese contents increase the susceptibility to temper embrittlement.

The optimum level of manganese must be determined by weighing its beneficial effects (control of hot-cracking, good weldability, good toughness) against its detrimental effects with regard to creep strength and temper embrittlement. Optimum manganese levels should be between 0.70 and 0.85%, in order to keep both high temperature strength and adequate toughness (71).

Silicon improves the weldability and high temperature strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Silicon increases the fluidity of the weld pool and deoxidizes the molten weld metal. Kanazawa et al. (124) found that silicon was very effective in increasing the high temperature strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Despite these beneficial effects, excessive silicon contents are known to promote temper embrittlement (53,111). Also, the notch ductility of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ welds decreases with increasing silicon content. In order to control temper embrittlement while maintaining good weldability, the optimum silicon content for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ should be between 0.20 and 0.40% (77).

Aluminum has been reported to reduce the creep strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel (98). Bonta (111) has reported that aluminum causes a severe degradation in the notch toughness of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal. Aluminum is an undesirable element in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ and should be kept at low levels.

Copper and nickel reduce the creep strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ (40,98). The copper content should be kept below 0.1% in $2\frac{1}{4}\text{Cr}-1\text{Mo}$, especially in weld metals, to avoid problems with embrittlement.

Boron has been reported to improve the creep strength of $2\frac{1}{4}\text{Cr-1Mo}$. Small additions of boron are thought to influence the grain boundary strength in $2\frac{1}{4}\text{Cr-1Mo}$ (40). Kanazawa et al. (124) have reported that boron additions increase the high temperature strength of $2\frac{1}{4}\text{Cr-1Mo}$.

It is evident that the creep properties of $2\frac{1}{4}\text{Cr-1Mo}$ steel are influenced by a number of elements. Enhanced creep properties can be obtained by establishing optimum levels for these elements.

Heat Treatment Effects on $2\frac{1}{4}\text{Cr-1Mo}$

The mechanical properties of $2\frac{1}{4}\text{Cr-1Mo}$ base metal and weld metal can be greatly enhanced by heat treatment. Because of this, $2\frac{1}{4}\text{Cr-1Mo}$ is available in a wide range of strength levels, making it a versatile design material. The common heat treatments used in connection with $2\frac{1}{4}\text{Cr-1Mo}$ include annealing, normalizing and tempering, and quenching and tempering.

Normalizing of $2\frac{1}{4}\text{Cr-1Mo}$, when performed according to ASTM specification A387-70a, involves austenitizing the material to a temperature within the range 1650-1850 F, followed by cooling in air. Common industrial practice involves austenitizing for 1 hour per inch of thickness ($\frac{1}{2}$ hour minimum and 4 hours maximum, regardless of thickness) (93). The material is then cooled in air.

The microstructure produced in normalized $2\frac{1}{4}\text{Cr-1Mo}$ is dependent on the cooling rate, which in turn is determined by the size and geometry of the material. Fast cooling rates generally produce microstructures consisting of bainite and martensite, while microstructures of slowly cooled materials are predominantly ferritic, although containing some bainite. As much as 8% retained austenite may be present in the as-normalized

structure. This carbon-enriched austenite precipitates as it transforms to ferrite during tempering (93).

Annealing of $2\frac{1}{4}\text{Cr-1Mo}$ is done with holding times and austenization temperatures similar to those used in normalizing. However, instead of air-cooling, the material is either cooled slowly in a furnace or held isothermally at a temperature close to 1300 F. Furnace annealing involves slow furnace cooling at a rate of 50 to 100 F/hr to a temperature below the ferrite-pearlite nose on the CCT diagram for $2\frac{1}{4}\text{Cr-1Mo}$, followed by cooling to room temperature at a slightly faster rate. Isothermal annealing, however, involves cooling the material to about 1300 F and holding for about 2 hours, to obtain nearly complete transformation to ferrite and pearlite (93). The material is then cooled to room temperature, avoiding the formation of lower transformation products.

The microstructures produced in $2\frac{1}{4}\text{Cr-1Mo}$ are similar for both methods of annealing. Cooling from 1700 F at 100 F/hr would produce a microstructure containing about 77% ferrite, 5% pearlite and 18% bainite (93). The amount of bainite will decrease with a decrease in the cooling rate. Any austenite which does not transform to ferrite and pearlite during the annealing treatment would turn to bainite upon cooling to room temperature. The structures produced by annealing are not determined by the size and geometry of the piece, as is the case in normalizing.

Another heat treatment in common use for $2\frac{1}{4}\text{Cr-1Mo}$ is quenching and tempering. This treatment involves the liquid quenching of $2\frac{1}{4}\text{Cr-1Mo}$ from austenitizing temperatures similar to those utilized for normalizing (i.e., 1650-1850 F). The cooling rate induced by quenching will depend upon the geometry and size of the piece and the quenching medium employed. The as-quenched

microstructure of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ is predominantly a bainite/martensite mixture (20) and a tempering operation is essential to ensure adequate toughness and ductility. Tempering will produce significant microstructural changes in both quenched and normalized materials.

The transformation of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ has been studied by at least 2 investigators (20,40). Transformation of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ occurs in 2 general ranges. Above 1250 F the transformation products are pearlite and pro-eutectoid ferrite, while below this temperature the products are bainite and martensite (20).

The continuous cooling transformation (CCT) diagram for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ is shown in Figure 1. The critical cooling time from an austenitizing temperature of 1795 F to the beginning of ferrite formation at 930 F is 125 seconds; however, 440 seconds are required for the beginning of pearlite formation (20). Rapid cooling rates will produce a bainitic/martensitic structure.

The isothermal transformation (IT) diagram, shown in Figure 2, also helps explain the transformation behavior of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. The important characteristics of this diagram include the ferrite and carbide transformation noses and the existence of an austenite bay over a limited range of temperature between the ferrite and bainite regions of the diagram. The isothermal transformation diagram also illustrates the short time (5 seconds) required for the start of the bainite transformation at temperatures below 900 F (40).

The composition of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ influences its transformation characteristics. Boron, nickel and vanadium retard the ferrite transformation, while manganese and nickel lower the transformation temperature of bainite (40).

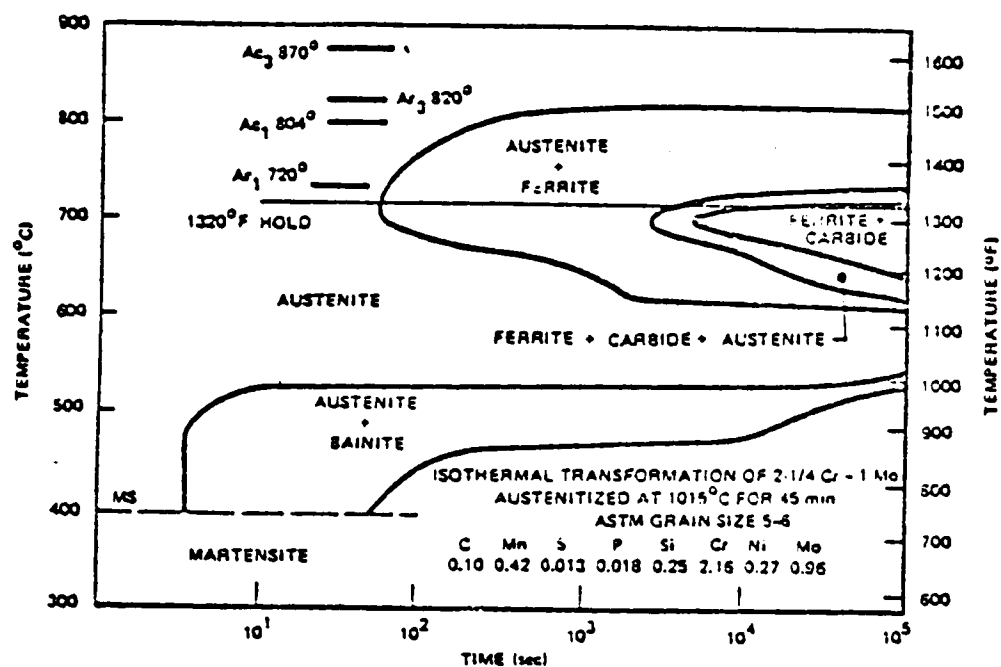


Figure 1. Continuous cooling transformation diagram for 2½Cr-1Mo (from Reference 105).

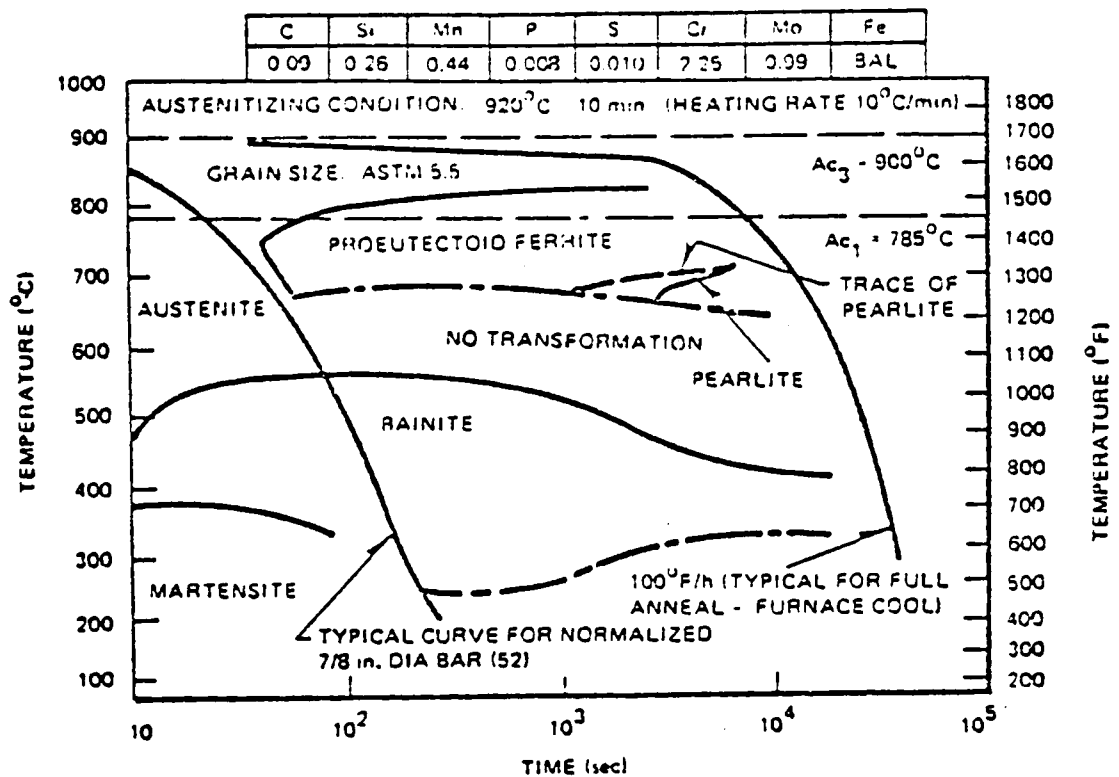


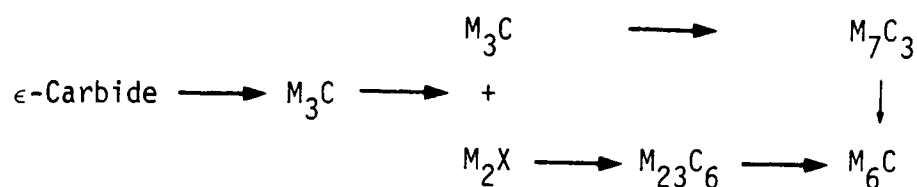
Figure 2. Isothermal transformation diagram for 2½Cr-1Mo (from Reference 106).

Two of the heat treatments mentioned earlier, normalizing and quenching, are usually followed by tempering. Tempering is often essential to improve the toughness and ductility of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Tempering produces significant changes in the carbide morphology of $2\frac{1}{4}\text{Cr}-1\text{Mo}$, as the carbides coarsen with time and temperature and lose their strengthening ability.

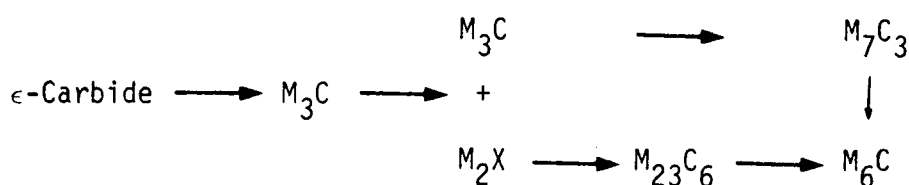
The following discussion of tempering will also include changes produced by aging at elevated temperatures (i.e., creep conditions). This is reasonable since alloy steels tempered for short times will not usually possess the equilibrium phase constitution and thus major changes in the distribution and structure of the carbides occur during service at elevated temperatures (35) (i.e., tempering and thermal aging are essentially the same process). For long-time extrapolation of creep data, a knowledge of tempering behavior is essential since the creep behavior will depend on the composition and location of the carbide phases existing at the service temperatures (62).

The tempering process will eventually produce an equilibrium microstructure. However, in tempered alloy steels this equilibrium state is not generally reached by simple precipitation and growth of the equilibrium carbides from the matrix. In the case of $2\frac{1}{4}\text{Cr}-1\text{Mo}$, several transitional carbides form before equilibrium is attained. Indeed, in complex alloy steels such as $2\frac{1}{4}\text{Cr}-1\text{Mo}$ several carbide transformations may proceed simultaneously.

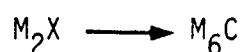
Baker and Nutting (9) reported on the sequence of carbide changes produced during tempering and subsequent thermal aging for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ in both the normalized and quenched conditions. The equations shown below summarize their findings:

1. Quenched $2\frac{1}{4}\text{Cr}-1\text{Mo}$:2. Normalized $2\frac{1}{4}\text{Cr}-1\text{Mo}$ (note that the bainitic area of the normalized material differed from the ferritic area in its response to tempering):

a. Bainite:



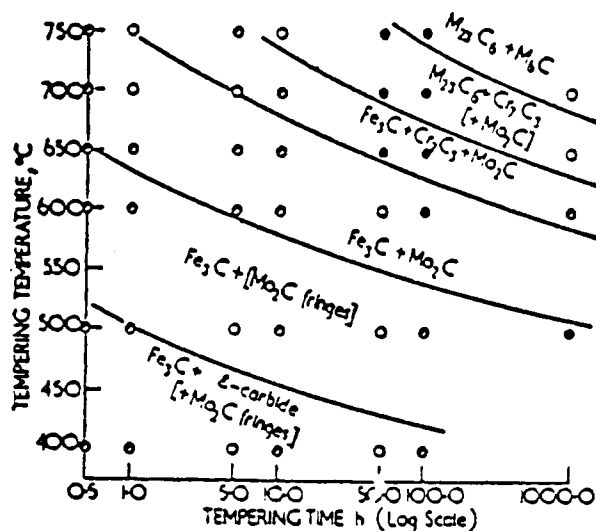
b. Ferrite:



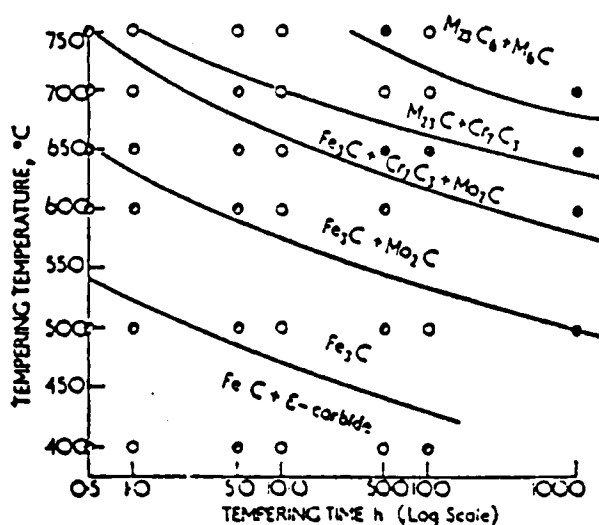
It may be seen from these equations that the M_2X precipitate forms much earlier in the ferritic regions of the normalized material and persists for much longer. This is an important effect on creep behavior, as will be discussed later.

Baker and Nutting (9) produced isothermal diagrams illustrating changes in carbide formation produced by tempering both normalized and quenched material. These are shown in Figure 3.

The carbide changes represented by the forementioned equations will now be discussed. The bainitic regions of the normalized steel will be considered with the quenched steels, since the sequence of carbide changes is almost identical for both.



Isothermal diagram showing the sequence of carbide formation on tempering the normalized steel



Isothermal diagram showing the sequence of carbide formation on tempering the quenched steel

Figure 3. Changes in carbide morphology produced by tempering of quenched and normalized $2\frac{1}{2}\text{Cr}-1\text{Mo}$ (from Reference 9).

After tempering, ϵ -carbide forms first in both the quenched $2\frac{1}{4}\text{Cr-1Mo}$ and the bainitic regions of the normalized material. The ϵ -carbide formed is more stable in $2\frac{1}{4}\text{Cr-1Mo}$ than in plain carbon steels. The ϵ -carbide does not completely transform to M_3C until after 5 hours at 930 F (9). This transformation takes place in situ, the carbide plates thickening and becoming sharper in outline. This M_3C carbide is iron-rich and is often termed Fe_3C or cementite, even though it may contain other elements.

The next change in carbide morphology is the precipitation by separate nucleation of a molybdenum-rich carbide M_2X . (This precipitate is also identified as Mo_2C and M_2C . However, since it may contain other elements the term M_2X is preferred.) Diffuse zones appear in the matrix and these are followed by the growth of well-defined needles. This fine M_2X precipitate has been identified by numerous investigators as the precipitate which contributes the most to the creep strength of $2\frac{1}{4}\text{Cr-1Mo}$. M_2X is coherent upon forming, although it loses coherency upon tempering. The primary strengthening effects are manifest just before the loss of coherency, at which point the precipitate becomes visible (89).

As the M_2X loses coherency, a chromium-rich carbide M_7C_3 (often termed Cr_7C_3) forms by nucleation within remaining M_3C particles or at the M_3C /ferrite interface (9). At the same time, a precipitate identified as M_{23}C_6 grows at the expense of the M_3C and M_2X precipitates. M_{23}C_6 is also believed to nucleate within M_3C precipitates or at M_3C /ferrite interfaces.

Finally, a precipitate identified as M_6C forms at the expense of the M_2X , M_{23}C_6 , and M_3C precipitates. Although this precipitate is most often termed M_6C , recent investigations suggest that it may actually be closer

to M_4C in $2\frac{1}{4}Cr-1Mo$ (62). The most important fact about the late forming carbides $M_{23}C_6$, M_7C_3 , and M_6C is that they coarsen with time, causing a loss in creep strength.

The final microstructure will consist of coarse and widely scattered M_6C particles in a carbon-saturated ferrite matrix. A small amount of $M_{23}C_6$ may also be present.

The ferritic regions of the normalized material undergo a sequence of carbide changes much different from that described above for quenched $2\frac{1}{4}Cr-1Mo$ and bainitic regions of the normalized material. In the ferritic areas, the fine M_2X precipitate forms rapidly. The M_2X needles quickly become coarser than those in the tempered bainite. These needles are quite stable, although some evidence of re-solution has been observed after 15,000-25,000 hours at 1050 F (54). The M_2X precipitate will eventually give way to M_6C .

It should be noted that the final microstructure of both areas will be similar, consisting, as mentioned earlier, of coarse M_6C particles in a ferritic matrix (54,63).

The mechanical properties of tempered $2\frac{1}{4}Cr-1Mo$ are dependent upon the severity of the tempering treatment (both time and temperature), decreasing with both increasing time and temperature. Figure 4 illustrates this, showing a decrease in tensile and yield strength with increasing temper parameter (a factor used to indicate the severity of the temper). This reduction in mechanical properties can be related to the carbide changes in tempering.

The hardness of $2\frac{1}{4}Cr-1Mo$ is observed to decrease during tempering. This hardness decrease can be attributed to the various microstructural

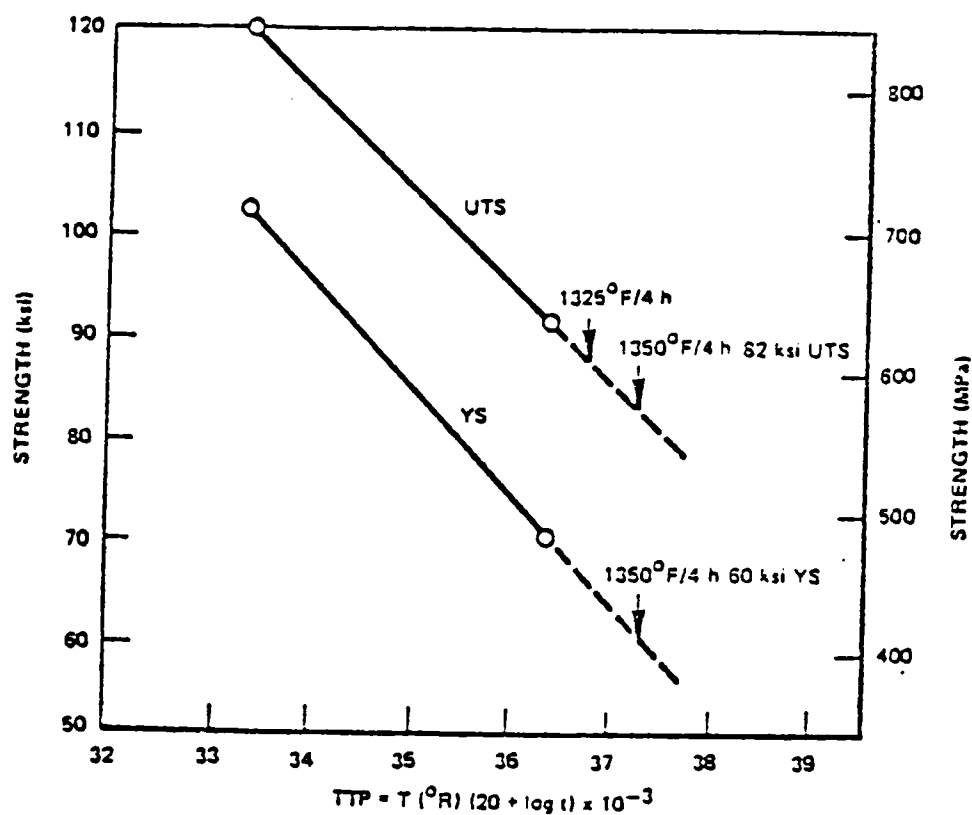


Figure 4. Effect of tempering on tensile and yield strength of 2½Cr-1Mo (from Reference 93).

constituents present. Murphy and Branch (54) observed that the bainite present in $2\frac{1}{4}\text{Cr}-1\text{Mo}$, while initially harder than the ferrite, softened with time, while the ferrite underwent a slight age-hardening. This is illustrated in Figure 5. Thus, the overall softening of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ during tempering is a function of bainite content, with those materials containing the most bainite showing the most softening.

A similar effect is observed on creep strength. High bainite material shows a strength advantage at short times; however, this strength is lost as the fine M_2X precipitate dissolves and the matrix recrystallizes. $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel with a high ferrite content starts out at a lower strength, but strengthens slightly under creep conditions. All important microstructural differences are gone after 50,000 hours at 1100 F (54). The creep behavior of both materials is then similar.

Because of the differing structures produced in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ by various heat treatments, it is of interest to compare the creep strength of annealed, normalized and tempered, and quenched and tempered material. The literature seems to be consistent in stating that quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ has a strength advantage over normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ up to a certain temperature. Above this temperature, spheroidization is responsible for a loss in strength of the quenched and tempered material. Hence, the normalized and tempered material has a strength advantage at higher temperatures. However, there is considerable disagreement in identifying the temperature at which the changeover takes place.

Steiner et al. (13) reported that quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ was superior to normalized and tempered material at temperatures up to 1050 F. Brown, Rege and Spaeder (26), however, found that the superiority

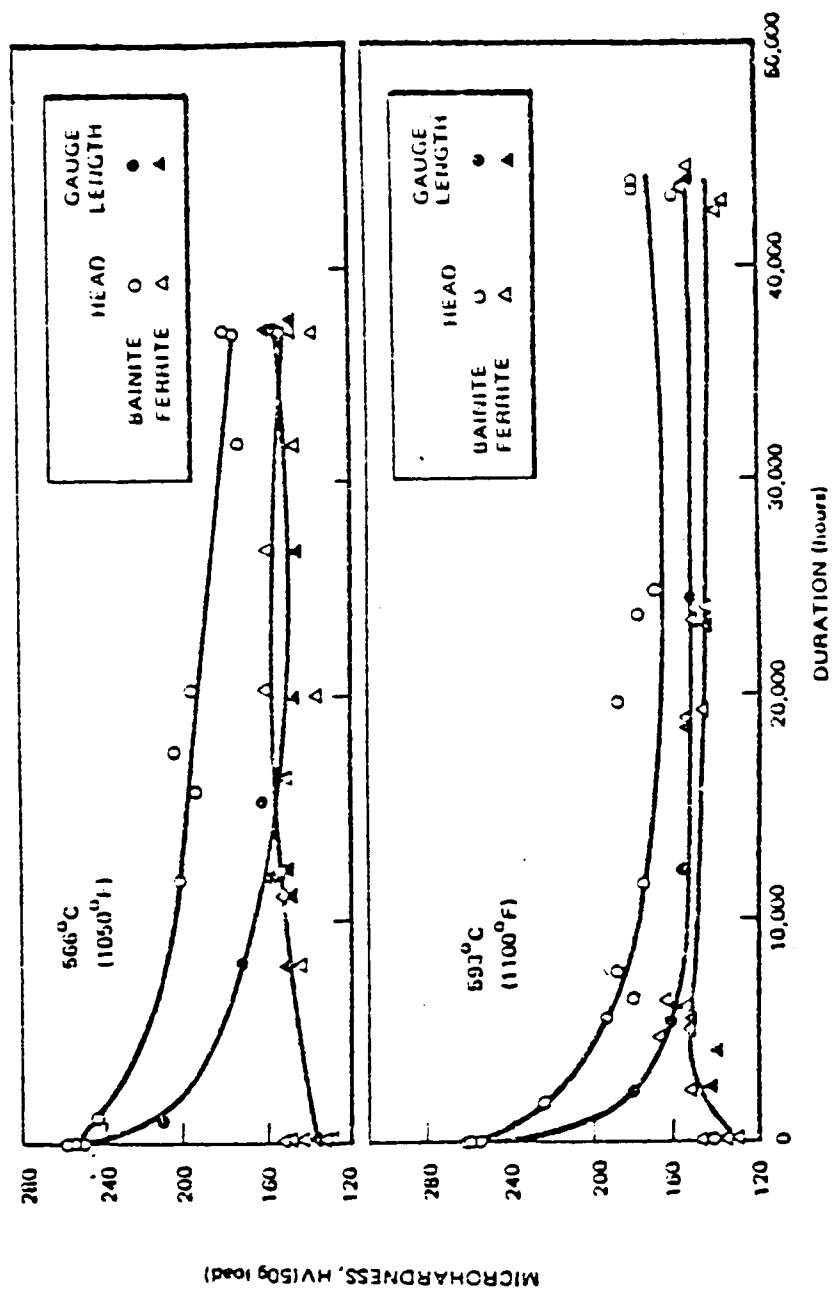


Figure 5. Change in hardness of individual constituents during creep testing of 2½Cr-1Mo (from Reference 54).

of the quenched and tempered material extended only up to 900 F. Pense and Stout (25) noted that quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ was prone to spheriodization in the stress-rupture range and suggested that its 10,000 hour stress-rupture strength would be superior to the normalized and tempered material at temperatures no higher than 950 F. Similarly, Robinson et al. (16) reported that spheriodization was responsible for a more rapid loss of creep strength in quenched and tempered material than in normalized material and suggested that the strength advantage produced by quenching and tempering is maintained only up to 1000 F. Finally, Baker and Nutting (9) found that normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ had superior creep resistance at 930 F because of the superior microstructural stability associated with the persistence of M_2X precipitates in the normalized steel.

The literature does not agree on the strength of the annealed material compared with either normalized and tempered or quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Inagaki (23) states that annealing produces stress-rupture strengths superior to those produced by other heat treatments. Emmanuel and Leyda (32), however, reported that annealing was superior to quenching and tempering at 800 F and 850 F, but not at 900 F. Miller and Benz (2) found that creep strength of normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ was greater than that of the annealed steel at 1000 F. The strength differences noted above may again be related to the microstructural changes induced by tempering and subsequent elevated temperature service. For test times up to 10,000 hours, normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ is reported to be superior to the annealed material. However, beyond 10,000 hours the annealed material is stronger than the normalized and tempered steel. This behavior can be related to the fact that M_2X carbides persist longer in ferrite than in

bainite. As mentioned earlier, M_2X carbides are very effective in enhancing creep strength. The normalized and tempered steel contains less ferrite and more bainite than the annealed steel. Hence, the M_2X carbides are not as stable as in the annealed material and transform more rapidly to coarser carbides. After this occurs, the annealed material, in which the M_2X carbides are still present, exhibits superior creep strength. After 50,000 hours at 1100 F, however, the creep properties of both materials would be similar due to the existence of an equilibrium microstructure containing M_6C carbides and no M_2X . Differences between the normalized and tempered and annealed materials decrease with increasing tempering temperature and decreasing carbon content (93).

In conclusion, it appears that the effect of heat treatment on the stress-rupture strength of $2\frac{1}{4}Cr-1Mo$ is uncertain in some areas. Quenched and tempered materials appear to have an advantage over normalized and tempered materials up to a temperature somewhere between 900 F and 1050 F. Above this temperature, the normalized and tempered materials seem to have superior creep strength. Annealed materials appear to have a strength advantage at temperatures below 900 F.

Welding Effects on $2\frac{1}{4}Cr-1Mo$

Welding is an important fabrication method used in the construction of pressure vessels, one of the primary applications for $2\frac{1}{4}Cr-1Mo$ steel. Hence, the effect of welding on the mechanical properties of $2\frac{1}{4}Cr-1Mo$ is of great interest. Since many of the vessels constructed of $2\frac{1}{4}Cr-1Mo$ operate at temperatures well in excess of 650 F, the effect of welding on the creep properties is especially important.

Due to the rapid cooling rates experienced by welds, the as-welded microstructure of most $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals will be predominantly bainitic. Multipass welding creates a complex weld microstructure due to variations in the thermal history of the individual passes. The microstructural constituents of a multipass $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld include tempered bainite, tempered martensite and proeutectoid ferrite. This complex microstructure leads to variations in hardness across the weld, as observed by Klueh and Canonico (66).

The weld metal microstructure developed in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ will be influenced by the carbon content. Low carbon (0.003%) and medium carbon (0.035%) $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal was found to consist primarily of proeutectoid ferrite, while high carbon (0.120%) weld metal was essentially 100% bainite with some martensite (66).

The heat-affected zone (HAZ) is another region of interest in the welding of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. Smith, Blanchard and Apps (38) studied the HAZ of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ and found that it consisted of 4 regions; a grain coarsened region, a grain refined region, a partially transformed region, and a subcritical region. The grain coarsened region had been heated to temperatures well above the A_{c3} , allowing considerable grain growth. The microstructure in this area consisted of upper bainite and tempered martensite, resulting in poor creep properties. The grain refined region had experienced temperatures just above the A_{c3} , allowing transformation from finer grained austenite on cooling. The microstructure consisted of a lath-like product similar to upper bainite. The partially transformed region had been heated to temperatures between the A_{c1} and A_{c3} . Microstructural changes in this region involved the dissolution of a fine Mo_2C (M_2X) dispersion and the

growth of coarse carbides in the ferrite matrix. The subcritical region was heated to just below the Ac_1 . Some coarsening of the precipitates was observed.

The microstructures produced in a weld, and hence the mechanical properties, depend upon a number of variables including the composition of the filler metal and base metal, welding process and conditions, preheat and post weld heat treatment (PWHT) temperatures, and flux characteristics. Each of these variables will influence the creep properties of $2\frac{1}{4}\text{Cr-1Mo}$ welds.

The carbon content of the weld deposit strongly influences stress-rupture strength. Several investigators, including Klueh and Canonico (67), have determined that stress-rupture strength increases with the carbon content of the weld deposit.

Certain carbide-forming elements such as columbium, vanadium and titanium also increase creep strength in $2\frac{1}{4}\text{Cr-1Mo}$ welds. Batte and Murphy (59) reported that an electrode containing columbium and titanium produced weld deposits with superior stress-rupture strengths. They postulated that this was due to the stabilization of the M_2X precipitate by solution of columbium and titanium, thus preventing the loss of strength which occurs when the fine M_2X precipitate is replaced by coarse carbides such as $M_{23}C_6$.

Differences between welding processes also affect the stress-rupture strength of $2\frac{1}{4}\text{Cr-1Mo}$ welds. Welding processes used with $2\frac{1}{4}\text{Cr-1Mo}$ include shielded metal arc (SMAW), submerged arc (SAW) and electroslag (ESW) welding. Automatic welding processes have also been used with $2\frac{1}{4}\text{Cr-1Mo}$.

Problems have been encountered in welding $2\frac{1}{4}\text{Cr-1Mo}$ with the electroslag process. Because of the high heat input inherent in this process, it is difficult to produce electroslag welded joints having adequate high temperature strength and toughness, despite extensive PWHT (71).

The submerged arc process also involves a high heat input. Rogers et al. (41) suggests that this high heat input produces a significant amount of tempering during welding and could explain the rapid tempering of $2\frac{1}{4}\text{Cr-1Mo}$ submerged arc welds during stress-relief.

The heat input level has also been looked at in conjunction with the reportedly inferior stress-rupture properties of $2\frac{1}{4}\text{Cr-1Mo}$ welds deposited by automatic processes. Sachs et al. (40) reports that tempering occurs during the automatic welding of $2\frac{1}{4}\text{Cr-1Mo}$, giving rise to the coarse carbides which normally occur during the later stages of tempering and which are associated with inferior creep properties.

The preheat temperature often has a significant influence on the mechanical properties of a weld deposit; however, this does not appear to be the case with $2\frac{1}{4}\text{Cr-1Mo}$. Bland (6) reported that preheat temperatures as high as 500 F had little or no effect on the mechanical properties of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal or HAZ. Klueh and Canonico (67) stated that the stress-rupture properties of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal specimens which were not given a preheat were little different from those given a 300-500 F preheat. Similarly, Apblett et al. (5) reported that preheats as high as 600 F resulted in only a slight decrease in the maximum HAZ hardness. Thus, it appears that preheating $2\frac{1}{4}\text{Cr-1Mo}$ is ineffective in producing softer transformation products. However, preheating may be necessary for other reasons.

Post weld heat treatment of $2\frac{1}{4}\text{Cr-1Mo}$ weldments is necessary to insure adequate ductility of the weld, relieve stresses induced by welding, and to equalize differences in microstructure, hardness, and mechanical properties produced by multipass welding. The ASME Boiler and Pressure Vessel Code (85) requires a minimum PWHT of 1250 F for $2\frac{1}{4}\text{Cr-1Mo}$. Interestingly, Leyda (83) reported that a PWHT of 1250 F produced the greatest stress-rupture strength in $2\frac{1}{4}\text{Cr-1Mo}$ weld metal. PWHT at temperatures less than 1050 F appear to produce no significant changes in $2\frac{1}{4}\text{Cr-1Mo}$ welds, while PWHT within the range of 1050 to 1150 F result in an undesirable increase in the hardness and a decrease in the ductility of $2\frac{1}{4}\text{Cr-1Mo}$ welds (6).

A number of articles have commented on the comparative stress-rupture strengths of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal and base metal. However, there is little agreement in the literature on this topic. Klueh and Canonico (67) found that, for a given carbon level, $2\frac{1}{4}\text{Cr-1Mo}$ weld metal was stronger than either the base metal or HAZ. Similarly, Pense and Stout (25) reported that for $2\frac{1}{4}\text{Cr-1Mo}$ welds (A542 plate), the weld composite was equal to or slightly greater in strength than the base plate. However, Inagaki et al. (23) stated that the stress-rupture strength of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal was inferior to base metal. Leyda et al. (83) also reported weld metal rupture strengths lower than the base metal. Finally, some authors report no difference between weld metal and base metal creep strength. Emmanuel and Leyda (32) found that $2\frac{1}{4}\text{Cr-1Mo}$ weld metals, deposited by either submerged arc or shielded metal arc processes, had long-time high temperature properties comparable to those of the base metal when heat treated to the same strength level. Robinson, Pense and Stout (16) stated that the

mechanical properties of A387 Grade D $2\frac{1}{4}\text{Cr}-1\text{Mo}$ were unaffected by welding. Hoyt (108) reported that the stress-rupture strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ welded joints was similar to that of unwelded base material. Davis and Harris (130) found that the rupture strength of shielded metal arc deposited weld metal was almost identical to that of wrought or cast $2\frac{1}{4}\text{Cr}-1\text{Mo}$.

This lack of agreement is also manifest when examining the literature on the comparative strengths produced by various welding processes. Bynum (94) and Batte and Murphy (59) reported that the welding process appears to have little or no effect on the stress-rupture properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$. However, Inagaki et al. (23) states that the stress-rupture strengths of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ submerged arc welds are inferior to shielded metal arc welds. Rogers (42) refutes this and states that differences often thought to exist between submerged arc and shielded metal arc $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals are actually differences in electrode composition rather than welding process type and that differences between different types of shielded metal arc weld metal can be as great as those between shielded metal arc and submerged arc weld metals. One reason for this lack of agreement on the effect of welding process type is the large number of variables which influence any given weld.

Conclusion

This review has discussed the effects of composition, heat treatment and welding on the creep properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel. It is apparent that further research is needed on these effects, especially in the area of welding processes, in order to take full advantage of the excellent high temperature properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel.

CHAPTER II

REVIEW OF STRESS-RUPTURE DATA ON

$2\frac{1}{4}\text{Cr}-1\text{Mo}$ WELD METALS

The elevated temperature properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel have been a subject of metallurgical interest for the last 40 years. Much of this interest has developed because of the popularity of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ as a pressure vessel steel. As a result, the high temperature properties of wrought and cast $2\frac{1}{4}\text{Cr}-1\text{Mo}$ have been well documented for material in a variety of forms and heat treatments. Unfortunately, little information has been made available on the elevated temperature properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal. Since welding is the primary fabrication method used in connection with $2\frac{1}{4}\text{Cr}-1\text{Mo}$, the paucity of elevated temperature weld metal data is a matter of concern for the designer.

As pointed out in the preceding literature review, much confusion exists concerning the relative stress-rupture strengths of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal and base metal. Also, little is known concerning the relative stress-rupture strengths of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals deposited by different welding processes.

The meager amount of weld metal data is further illustrated by the fact that an extensive literature survey yielded fewer than a dozen previous studies on the stress-rupture properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal. These studies are summarized in Table 1, with respect to the temperature range, welding processes and heat treatments involved. Unfortunately, most of the previous work has been conducted over narrow temperature ranges

TABLE 1
PREVIOUS STUDIES ON 2½Cr-1Mo WELD METAL

Authors	Ref.	Date	Temperature Range (°F)	Welding Process	Heat Treatment
Simmons & Cross	3	1953	1000-1100	SMAW	PWHT
Campbell	18	1965	1050-1200	SMAW	PWHT
Hopkin & Murray	19	1965	1022-1100	SMAW	PWHT
Voorhees & Freeman	22	1966	1000-1200	SMAW	PWHT
Carpenter & Emmanuel	29	1966	800-900	SAW,ESW	Q&T, PWHT
Emmanuel & Leyda	32	1968	800-900	SAW,ESW	Q&T, PWHT
Guyla	86	1968	850-950	ESW	Q&T
Williamson & Branch	39	1970	1067	SMAW	PWHT
Emmanuel & Leyda	116	1970	750-900	SAW	PWHT
Sangdahl & Voorhees	119	1972	850-900	ESW	Q&T
Klueh & Canonico	67	1976	850-1050	GTAW	PWHT
Leyda et al.	83	1978	1050-1100	SMAW,SAW	PWHT

and has involved only one welding process or heat treatment. Comparison of data from the various studies is made difficult by differences in base metal composition and heat treatment parameters.

Stress-rupture Data

Figures 6-17 present existing literature data on the stress-rupture properties of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal in the post weld heat treated condition at temperatures ranging from 750 to 1250 F. Each of these plots will be discussed separately below.

The stress-rupture data for $2\frac{1}{4}\text{Cr-1Mo}$ weld metal in the post weld heat treated condition at 750 F is shown in Figure 6. The only data available at this temperature is represented by 3 data points from a study by Emmanuel, Leyda and Rozic (116). This data is for submerged arc weld metal with a carbon content of 0.140%. The dashed line in Figure 6 represents minimum base metal properties from Code Case N-47 (Section III) of the ASME Boiler and Pressure Vessel Code (85). The data at 750 F lies above the minimum base metal properties.

The data available at 800 F is shown in Figure 7, and consists of 5 data points from a study by Emmanuel and Leyda (32) on submerged arc weld metal with a carbon content of 0.080%. There is little scatter evident, even though test times range from 8 to nearly 20,000 hours. This data lies well above the minimum base metal properties from Code Case N-47.

More extensive data is available at 850 F, as evidenced by Figure 8. Test times range from 1 to 9000 hours. Submerged arc weld metal data from studies by Emmanuel and Leyda (32) and Emmanuel, Leyda and Rozic (116) is shown along with gas tungsten arc weld metal data from work by Klueh and Canonico (67). The scatter present in the data can be attributed

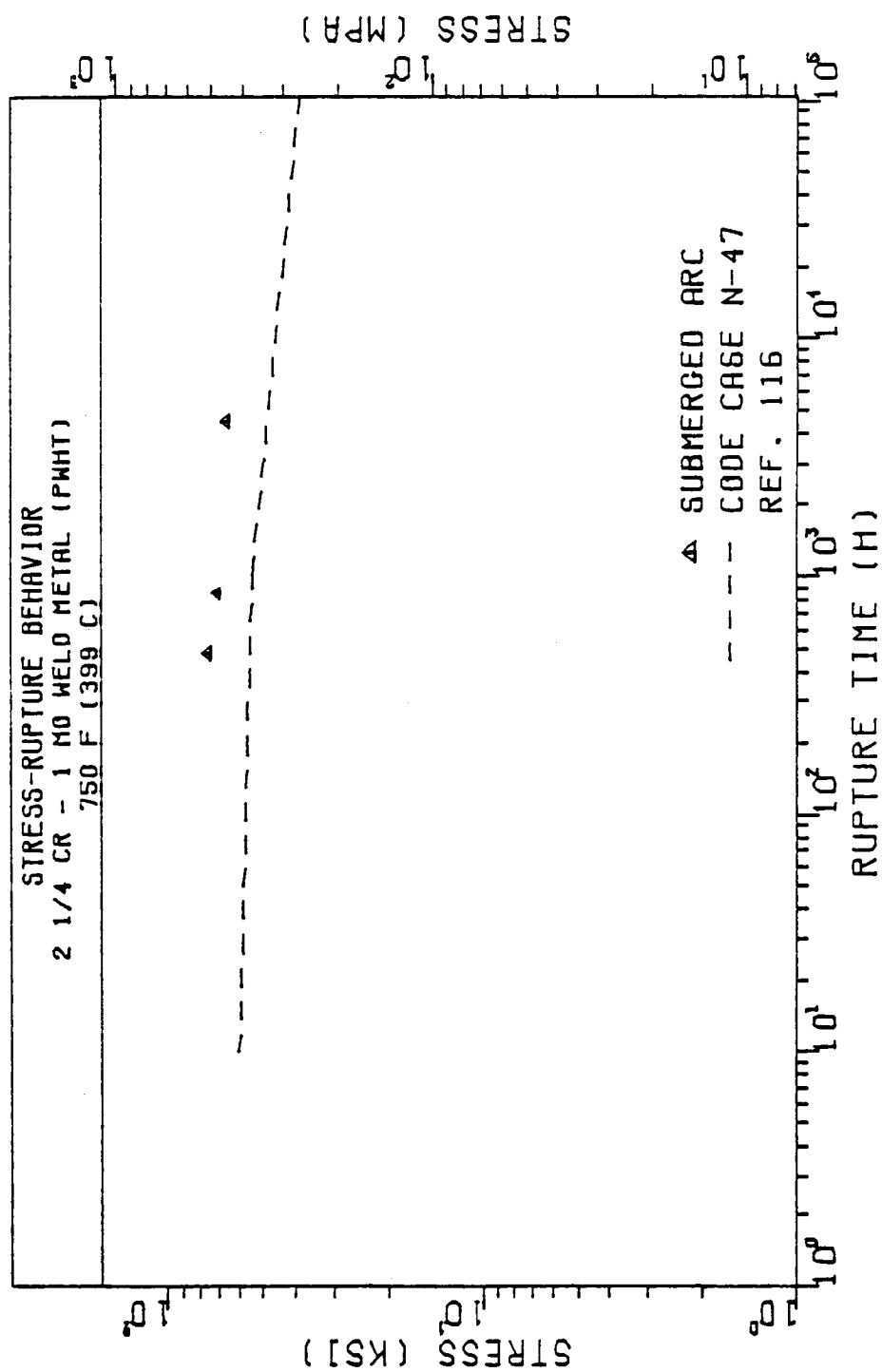


Figure 6. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 750 F.

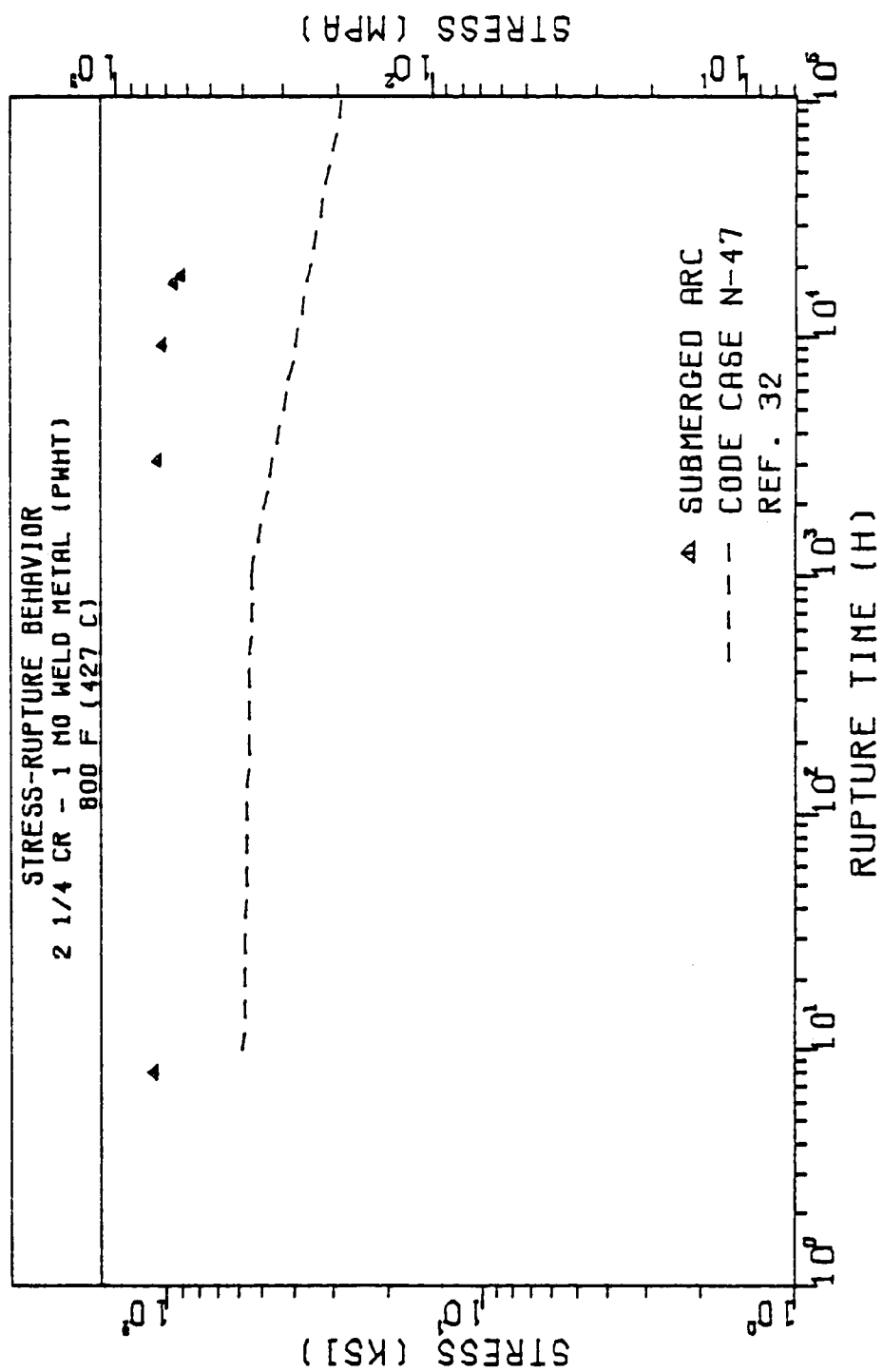


Figure 7. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 800 F.

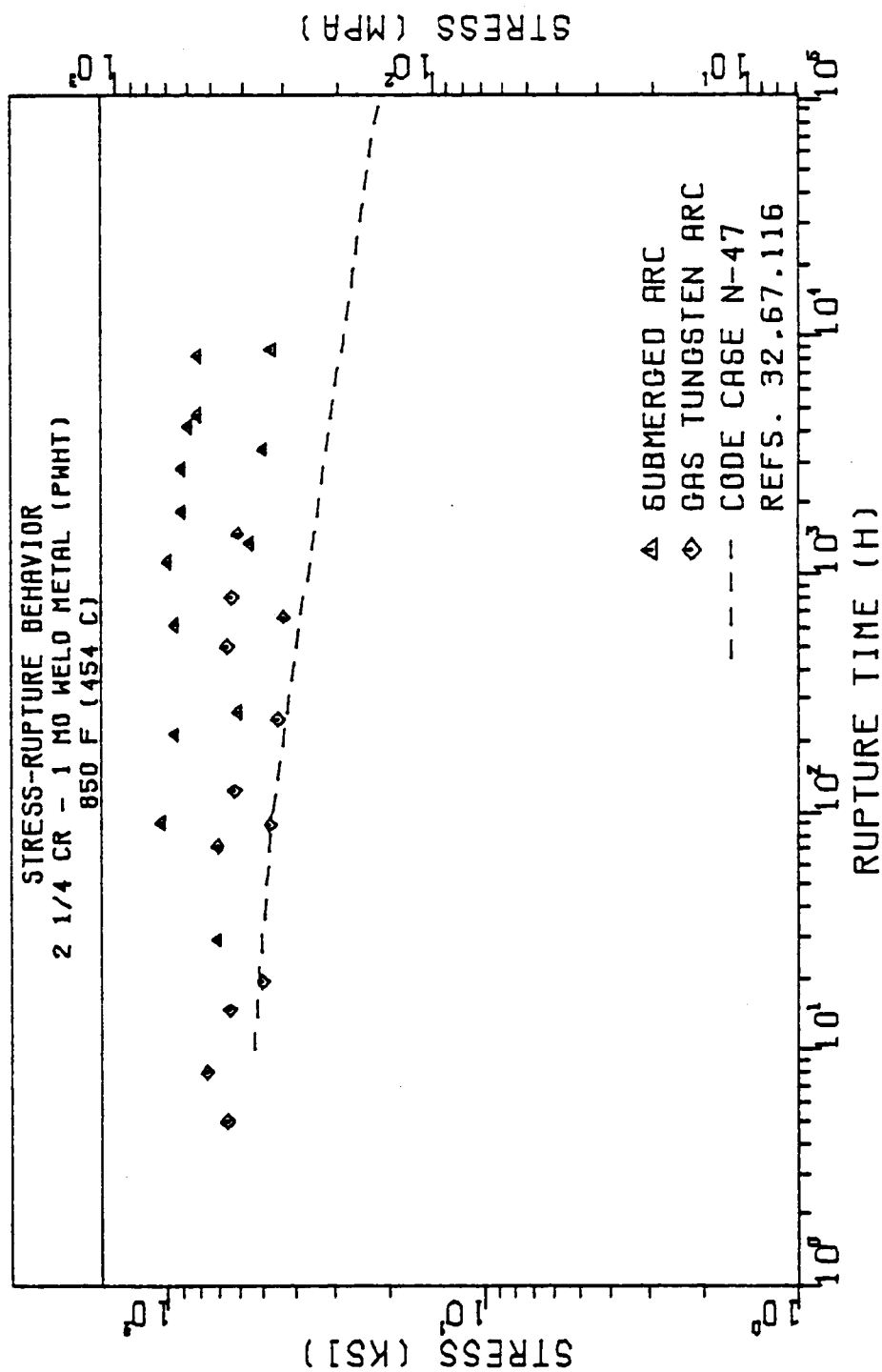


Figure 8. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 850 F.

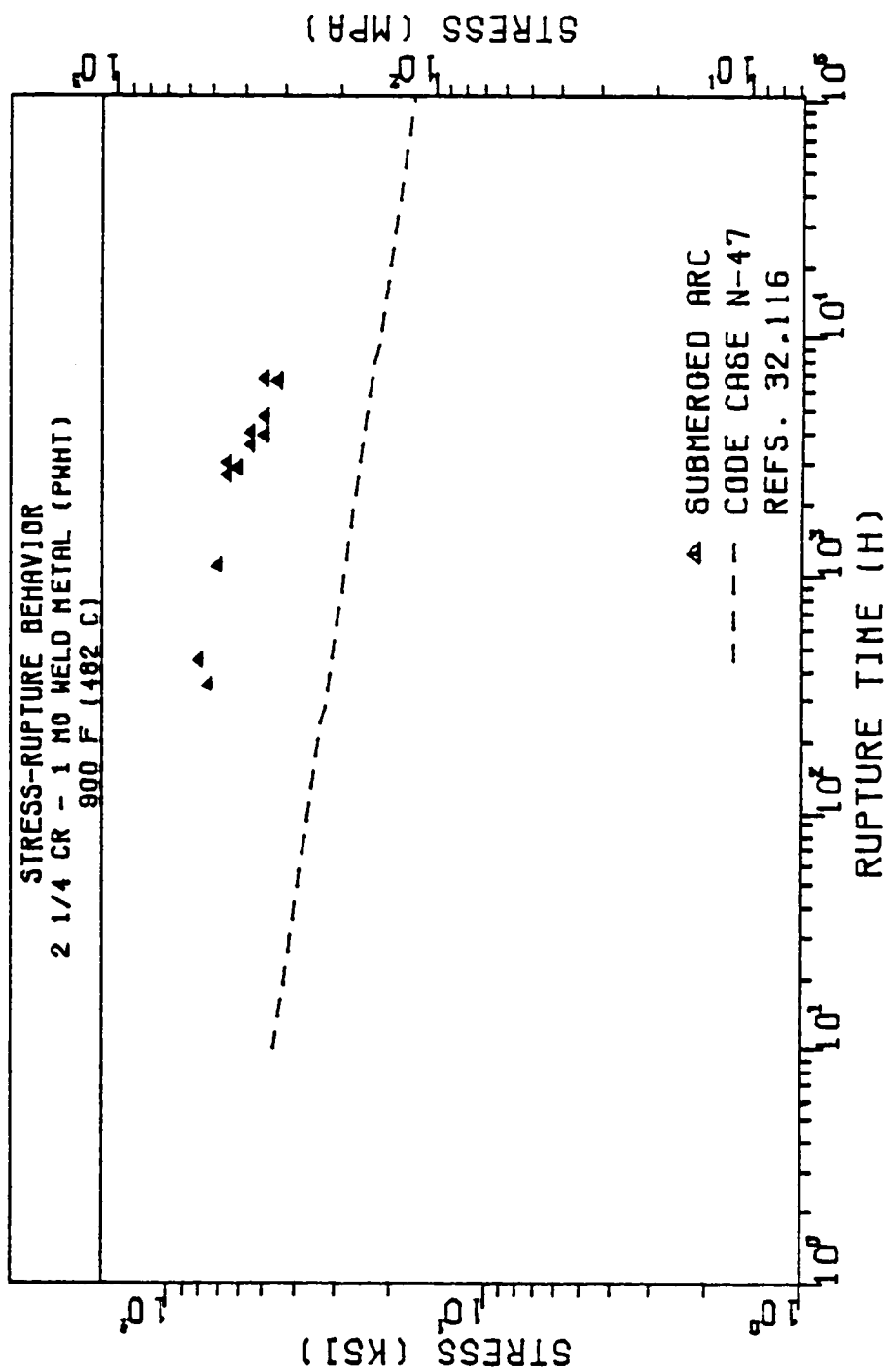


Figure 9. Stress-rupture data for postweld heat treated 2 1/4Cr-1Mo weld metal at 900 F.

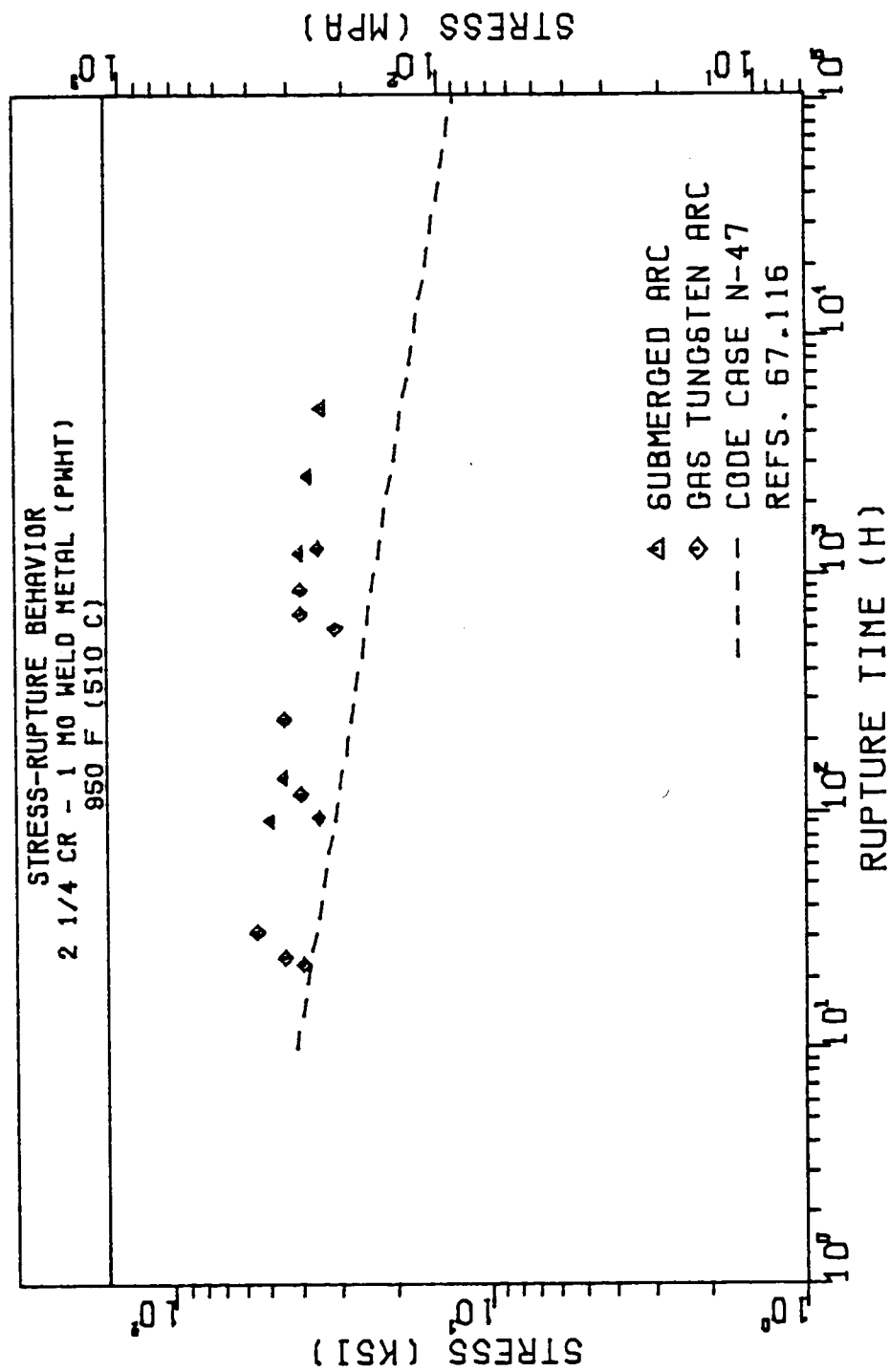


Figure 10. Stress-rupture data for postweld heat treated 2 1/4Cr-1Mo weld metal at 950 F.

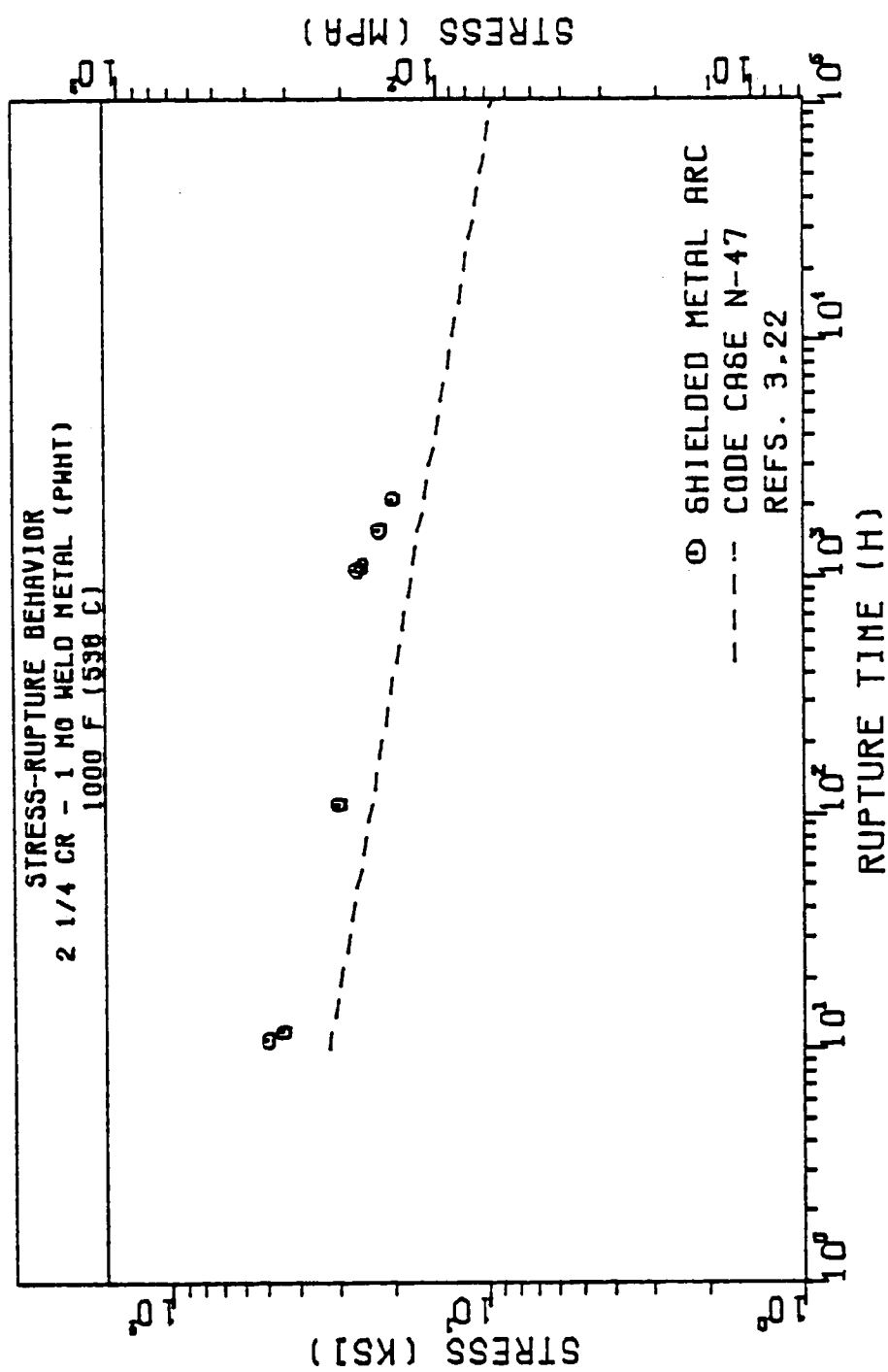


Figure 11. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 1000 F.

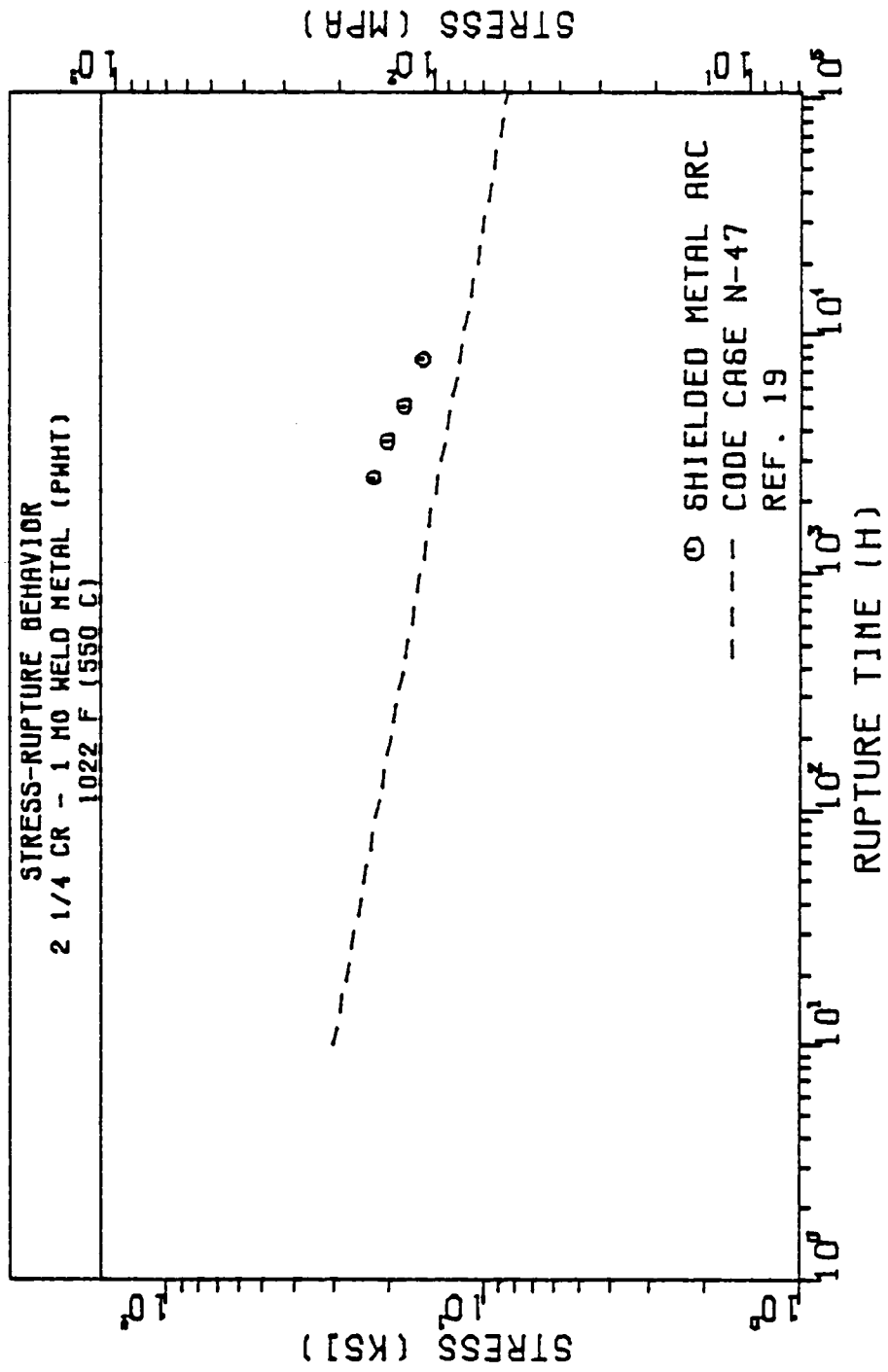


Figure 12. Stress-rupture data for postweld heat treated 2 1/4Cr-1Mo weld metal at 1022 F.

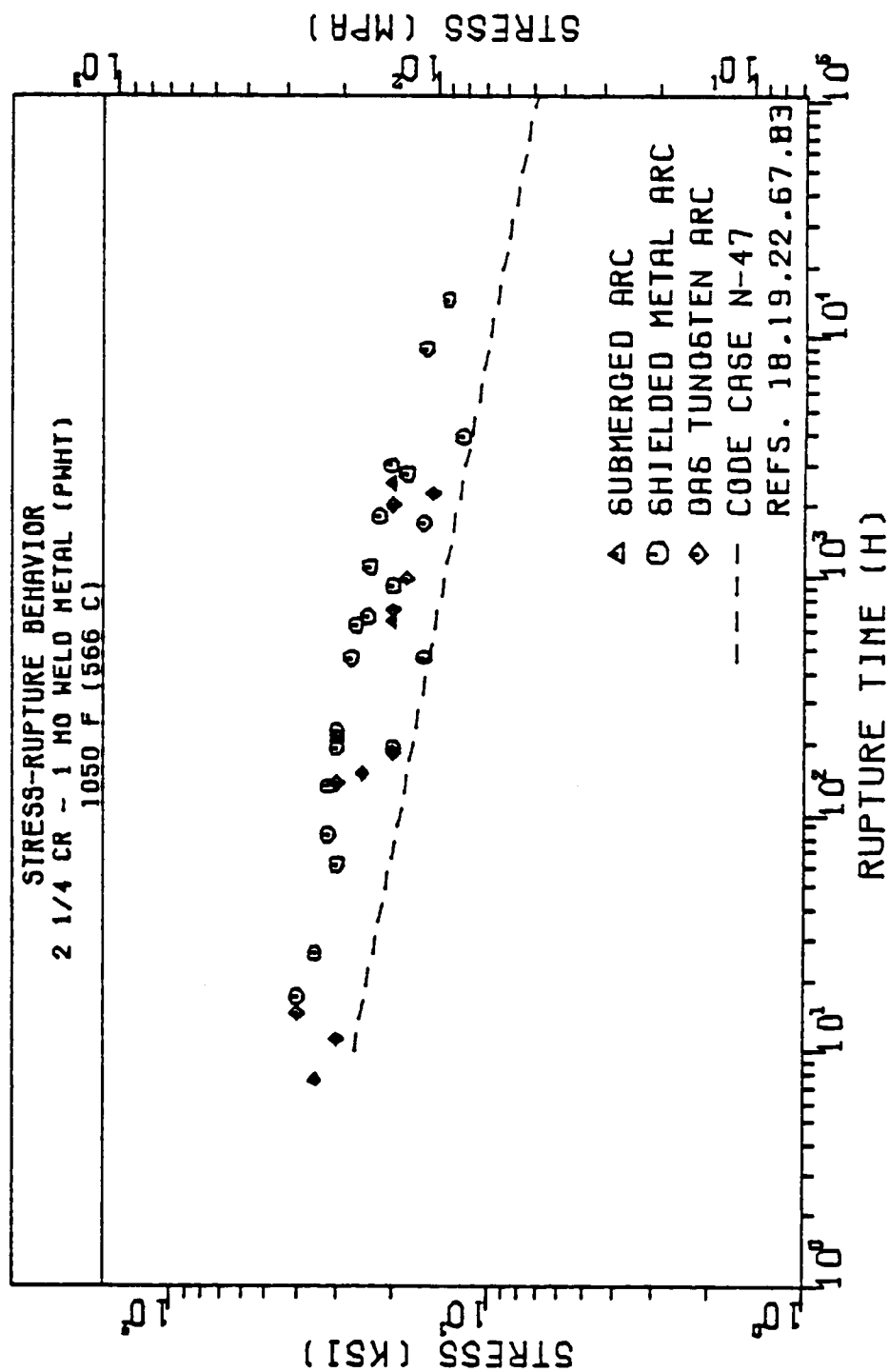


Figure 13. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 1050 F.

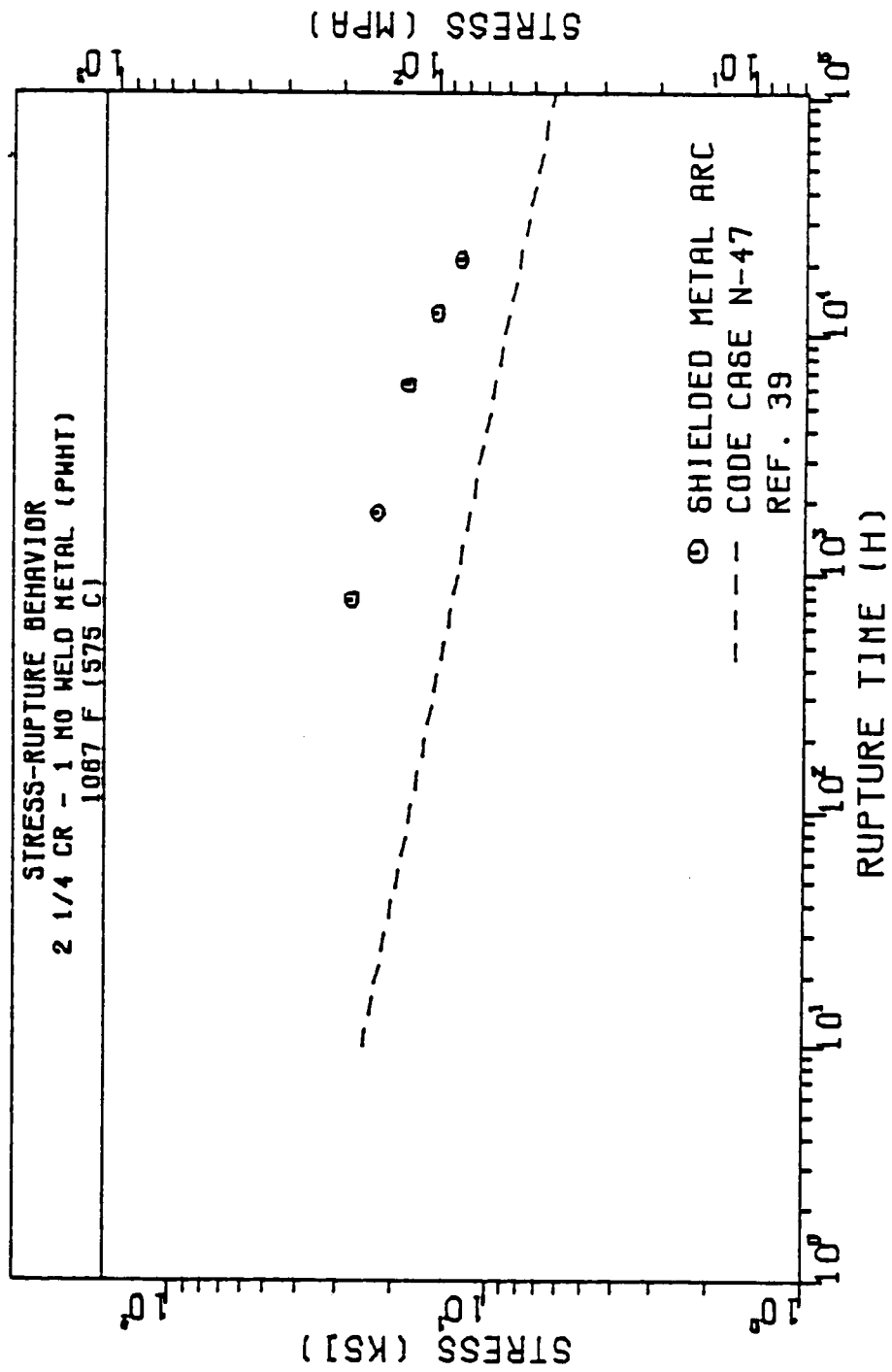


Figure 14. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 1067 F.

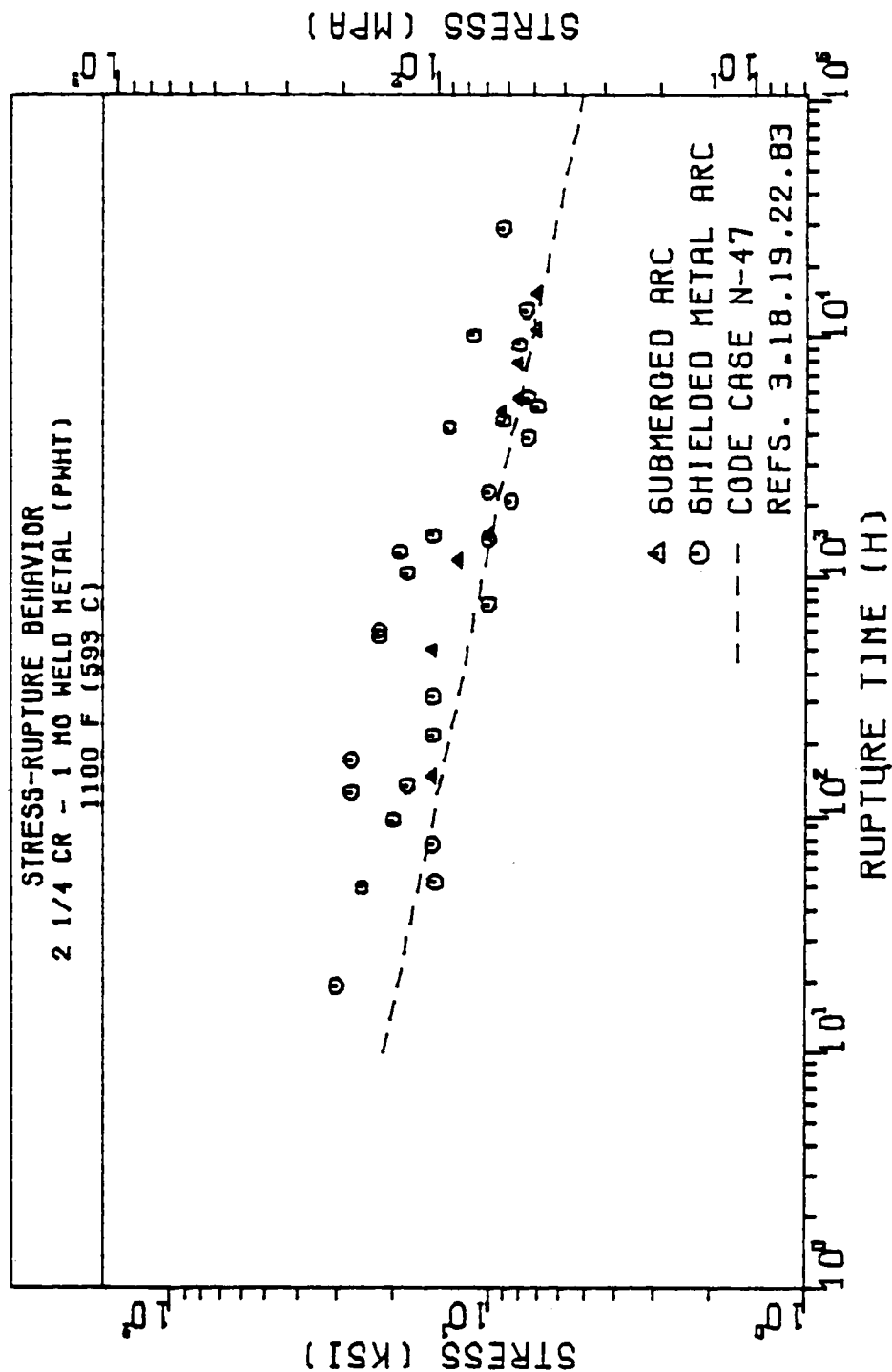


Figure 15. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 1100 F.

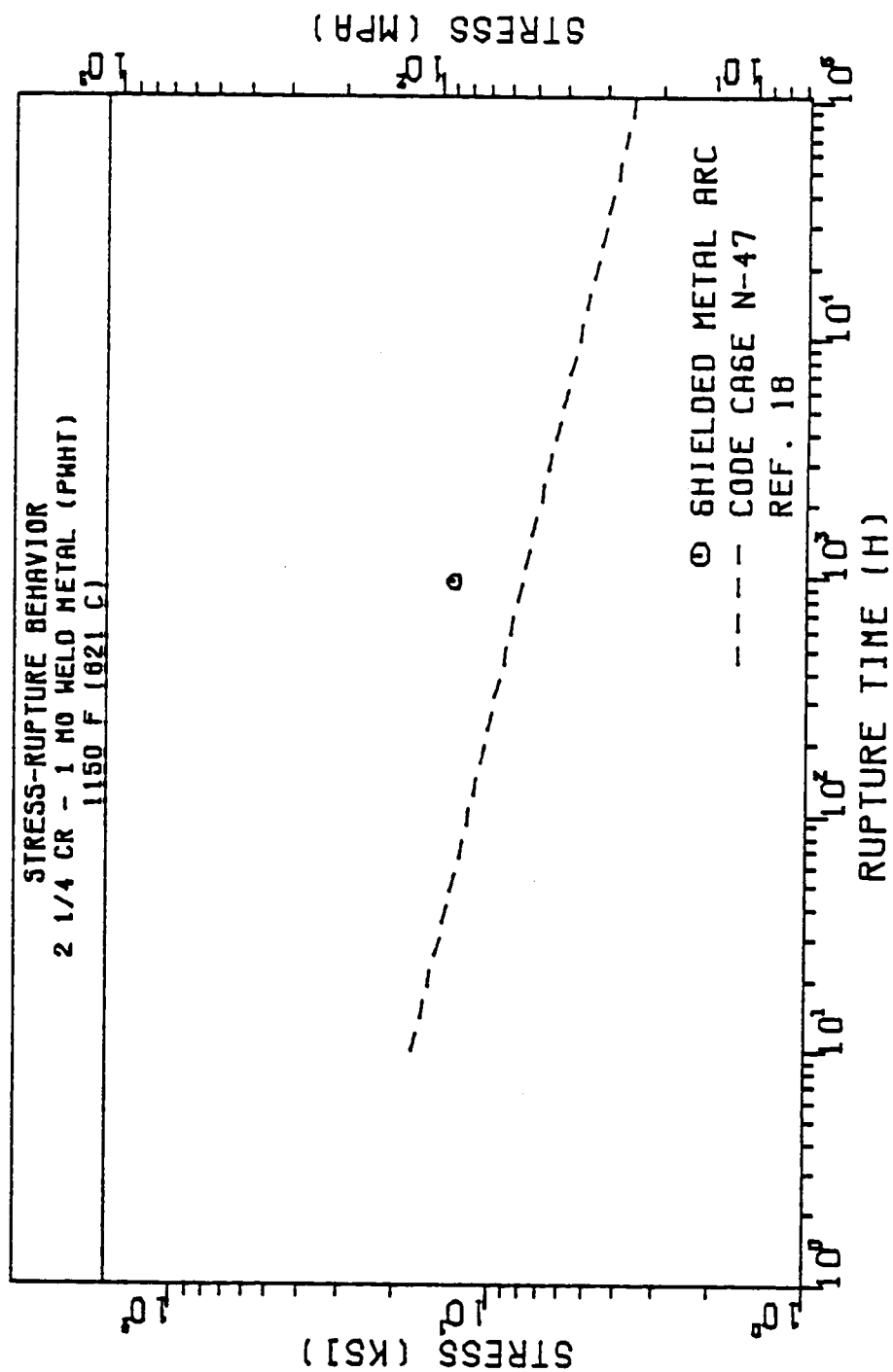


Figure 16. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 1150 F.

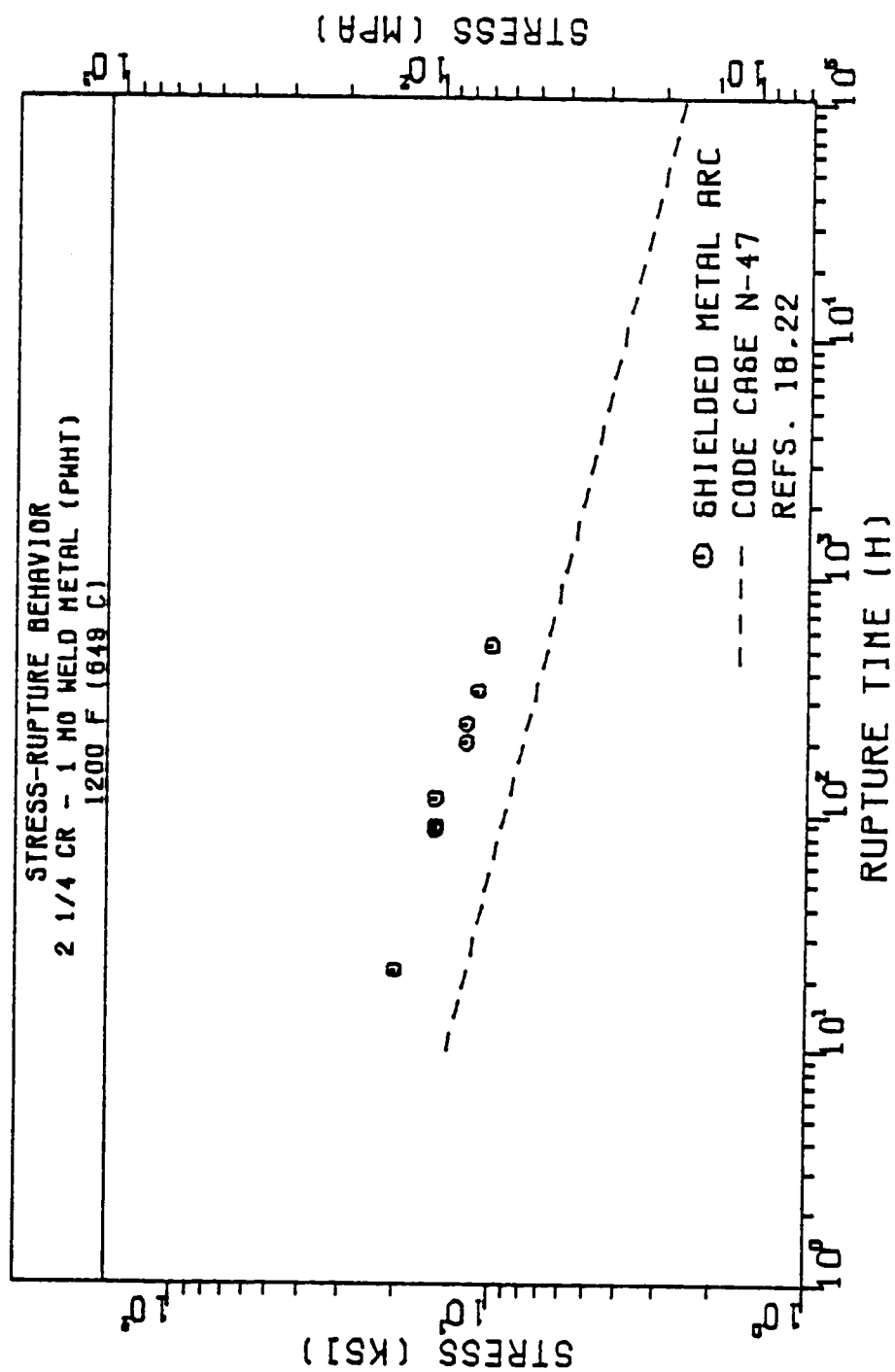


Figure 17. Stress-rupture data for postweld heat treated 2 $\frac{1}{4}$ Cr-1Mo weld metal at 1200 F.

to differences in carbon content and post weld heat treatment parameters. The data lies above the minimum base metal properties, except for several data points representing gas tungsten arc (GTA) weld metal of very low carbon content (0.003%).

Figure 9 (page 35) shows the data available at 900 F. All of the data is for submerged arc weld metal and is obtained from studies by Emmanuel and Leyda (32) and Emmanuel, Leyda and Rozic (116). Test times range from 350 to 7000 hours. There is little scatter in the data due to the fact that both studies involved weld metal with similar post weld heat treatments and carbon contents (0.073 and 0.080%). The data lies generally above the minimum base metal reference line.

At 950 F, data is available for both submerged arc and gas tungsten arc weld metals, as shown in Figure 10 (page 36). Test times range from 20 to 5000 hours. The data illustrates the influence of carbon content. The gas tungsten arc data reported by Klueh and Canonico (67) represents 3 different weld metals of varying carbon content. The low carbon (0.003%) displays the lowest stress-rupture strength, while the medium carbon (0.035%) weld metal exhibits somewhat higher strength. The high carbon (0.120%) weld metal has the highest stress-rupture strength. The submerged arc data, reported by Emmanuel, Leyda and Rozic (116), is also for weld metal with a high carbon content (0.140%) and is in close agreement with high carbon gas tungsten arc data. Again, the weld metal data lies above the minimum base metal values.

At 1000 F, data is available only for shielded metal arc weld metal, as shown in Figure 11 (page 37). This data comes from studies by Simmons and Cross (3) and Voorhees and Freeman (22). The carbon content ranges

from 0.080 to 0.120%. No data is available at times greater than 2000 hours. The data shows little scatter and lies above the base metal reference line.

Only 4 data points have been obtained at 1022 F. These points, shown in Figure 12 (page 38), are from a study by Hopkin et al. (19) on shielded metal arc weld metal. The points range in rupture times from 2500 to 8000 hours; no short-time data is available. The data shows marked linearity, and lies above the minimum base metal properties.

Figure 13 (page 39) shows the literature data at 1050 F. This data represents shielded metal arc, submerged arc, and gas tungsten arc weld metal from studies by Campbell (18), Hopkin et al. (19), Voorhees and Freeman (22), Klueh and Canonico (67) and Leyda et al. (83). Test times range from 1 to 15,000 hours. The data shows some scatter due to variations in carbon content (0.003 to 0.120%) and post weld heat treatment. The data exceeds minimum base metal properties.

Only 5 points have been obtained at 1067 F. This data, shown in Figure 14 (page 40), is from work by Williamson and Branch (39) on shielded metal arc weld metal. These points represent long-time data, ranging from 800 to over 20,000 hours. The data shows little scatter and lies well above the minimum base metal properties.

More weld metal data is available at 1100 F than at any other temperature. This data, for submerged arc and shielded metal arc weld metal, is shown in Figure 15. Test times range from 10 to 30,000 hours. The data at 1100 F is from studies by Campbell (18), Hopkin et al. (19), Leyda et al. (83), Simmons and Cross (3) and Voorhees and Freeman (22). The data shows some scatter, due to differences in carbon content (0.038

to 0.120%) and post weld heat treatment. Most of the data lies above the minimum base metal reference line.

At 1150 F, one 1 data point has been reported, as shown in Figure 16. This data point was reported by Campbell (18) in a study on shielded metal arc weld metal. This point exceeds the minimum base metal properties.

At 1200 F, data has been reported only for shielded metal arc weld metal. This data, shown in Figure 17, is from studies by Campbell (18) and Voorhees and Freeman (22). Only short-time data (less than 500 hours) has been reported. The data shows almost no scatter and lies well above the base metal reference line.

$2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal has also been tested in the quenched and tempered condition. Figures 18-21 present stress-rupture data for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal in the quenched and tempered condition at temperatures ranging from 800 to 950 F.

At 800 F, data has been reported by quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal deposited by the submerged arc and electroslag processes. This data is shown in Figure 18 and is from studies by Carpenter and Emmanuel (29), Emmanuel and Leyda (32) and Emmanuel, Leyda and Rozic (116). Test times range from 1 to nearly 20,000 hours. This data is for relatively high carbon weld metal, ranging from 0.110 to 0.140%. A striking feature of the data is the small amount of scatter present. The data exceeds minimum base metal properties.

Figure 19 presents data on quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal at 850 F obtained from work by Emmanuel and Leyda (32), Gulya (86), Emmanuel, Leyda and Rozic (116) and Sangdahl and Voorhees (119) on submerged arc and electroslag weld metal. The data covers a wide range of rupture times,

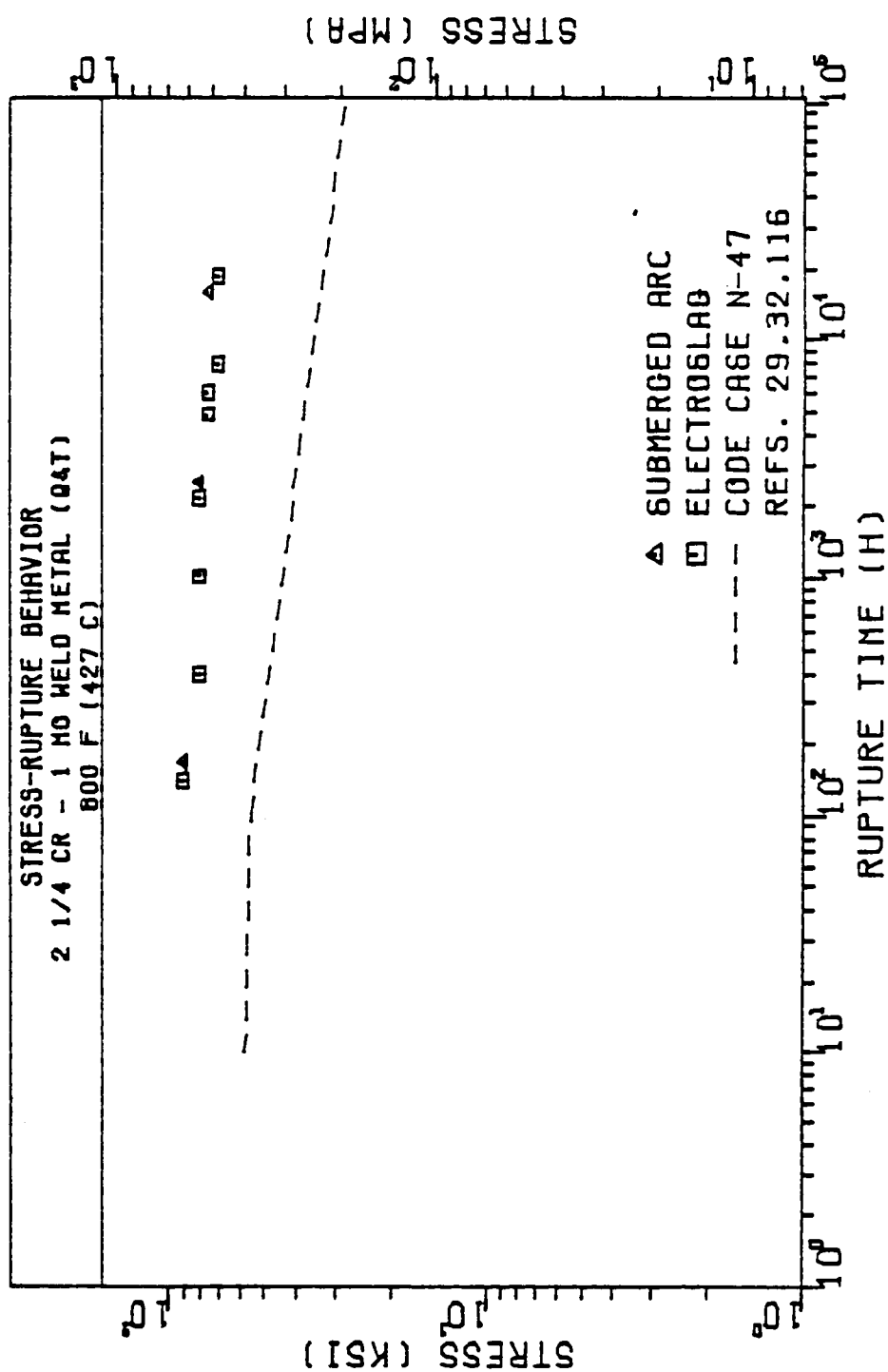


Figure 18. Stress-rupture data for quenched and tempered 2 $\frac{1}{4}$ Cr-1Mo weld metal at 800 F.

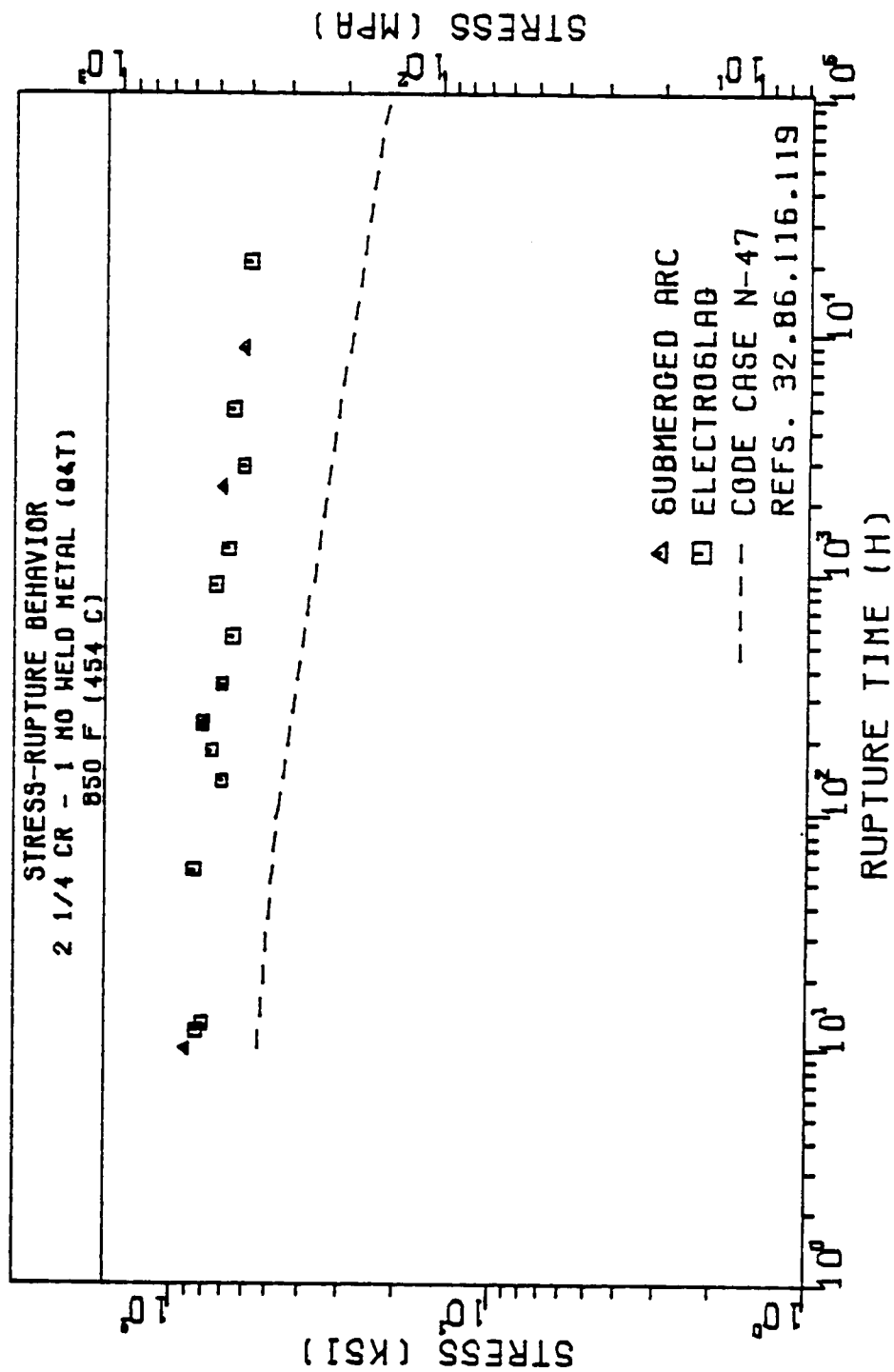


Figure 19. Stress-rupture data for quenched and tempered 2 $\frac{1}{4}$ Cr-1Mo weld metal at 850 F.

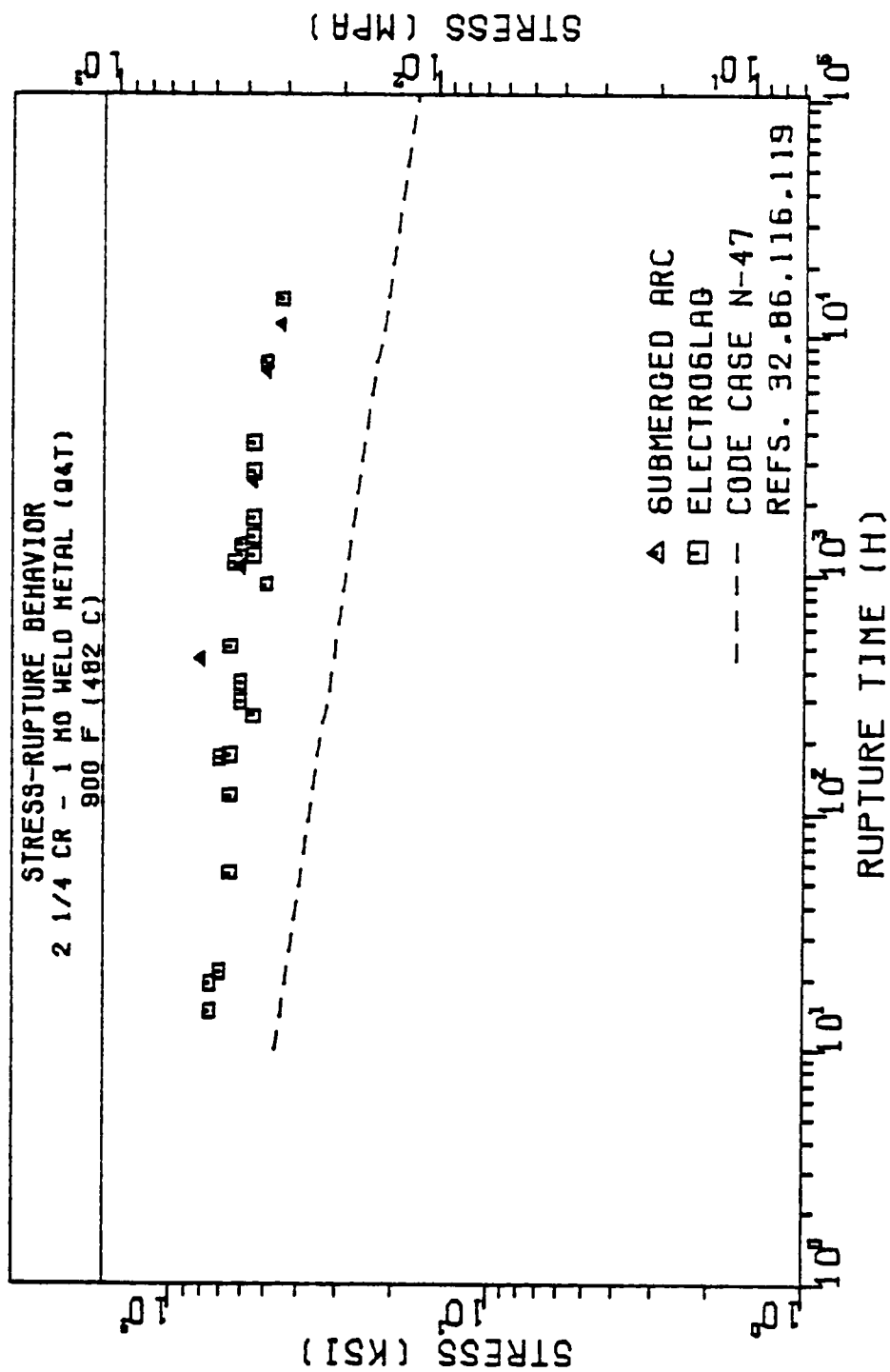


Figure 20. Stress-rupture data for quenched and tempered 2 1/4 Cr-1 Mo weld metal at 900 F.

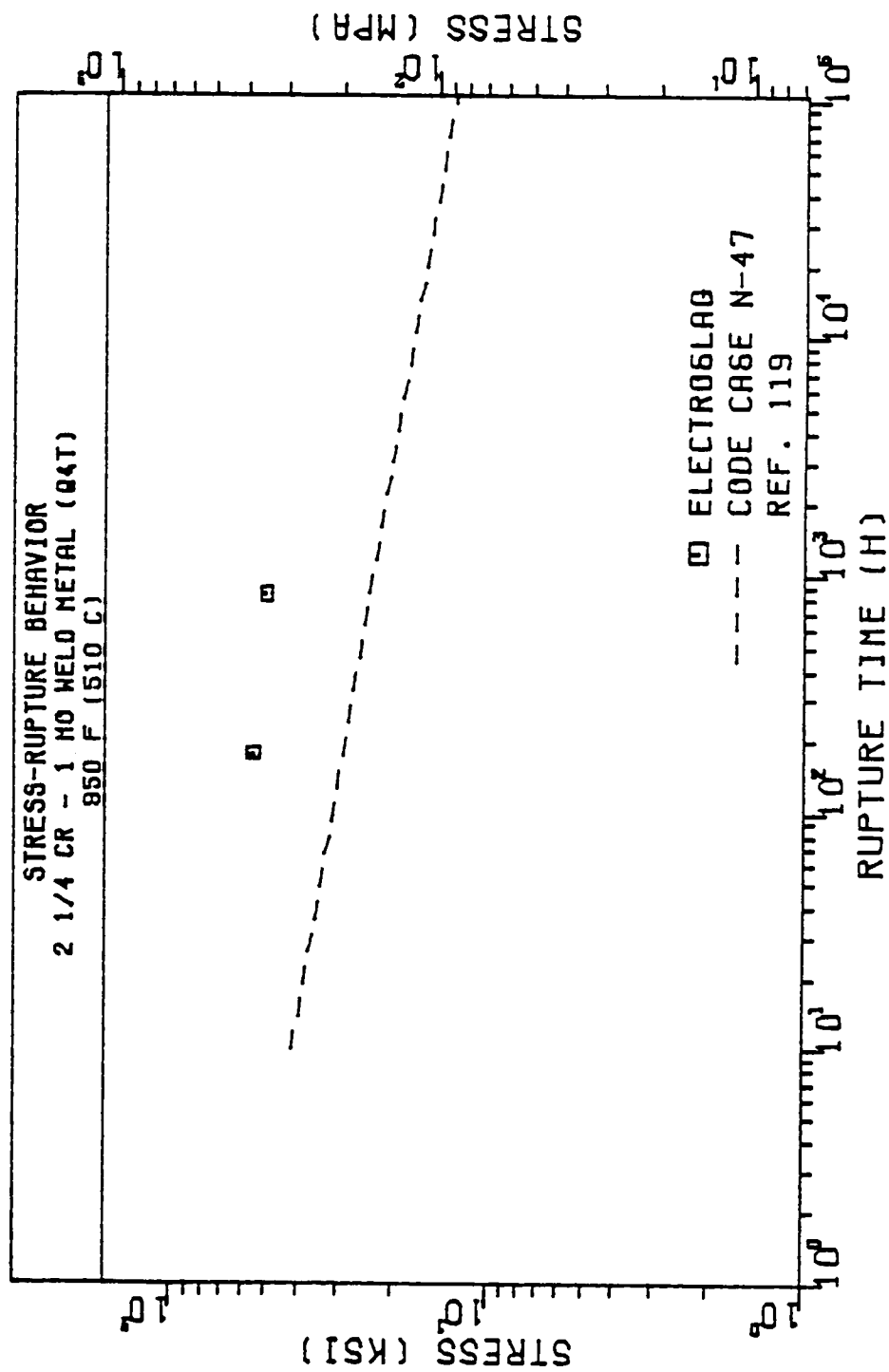


Figure 21. Stress-rupture data for quenched and tempered 2 $\frac{1}{4}$ Cr-1Mo weld metal at 950 F.

from 1 to over 20,000 hours. The carbon contents of the weld metals were relatively high, ranging from 0.090 to 0.140%. Very little scatter is seen in the data. The data lies well above the minimum base metal reference line.

An extensive amount of data has been reported for quenched and tempered weld metal at 900 F, as seen in Figure 20. This data, for submerged arc and electroslag weld metal, was reported by Emmanuel and Leyda (32), Gulya (86), Emmanuel, Leyda and Rozic (116) and Sangdahl and Voorhees (119). This data is also for relatively high carbon weld metal (0.090 to 0.150%) and exceeds minimum base metal properties. As was the case at 800 and 850 F, very little scatter is present in the data. The data covers a wide range of rupture times, from 10 to 15,000 hours.

Two data points have been reported for Sangdahl and Voorhees (119) for quenched and tempered 2½Cr-1Mo at 950 F, as shown in Figure 21. This data, for high carbon (0.140%) electroslag weld metal, lies well above the minimum base metal reference line. No data is available at times greater than 1000 hours.

Parametric Analysis

Stress-rupture parameters are mathematical relations in which rupture life is a function of both stress and temperature. Through the use of stress-rupture parameters, data obtained at different temperatures can be combined to characterize stress-rupture behavior. The first stress-rupture parameter to be developed was the Larson-Miller parameter:

$$P = \text{function of stress} = T(C + \log t_r) \times 10^{-3}$$

where T is the absolute temperature and t_r is time to rupture. C is a

constant which depends on the data. For $2\frac{1}{4}\text{Cr}-1\text{Mo}$, C values ranging from 18 to 23 have been used (15,16,19,31,117). Numerous other parametric evaluations have been developed with each having certain advantages for specific materials within a given temperature range. A number of these have been discussed by Conway (136).

Wattier (131) has used the Larson-Miller parameter to characterize $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal. Figures 22-23 are stress vs. Larson-Miller parameter plots for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal within given carbon ranges. The data shown in these plots were obtained from studies by Voorhees and Freeman (22), Leyda et al. (83) and Smith (107). Figure 22 is a stress-parameter plot for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal with carbon contents below 0.05%. The data is markedly linear and is best characterized by a constant of $C = 18.9$. Figure 23 presents stress parameter data on weld metal with carbon contents between 0.05 and 0.10%. This data also shows little scatter and is best characterized by a C constant of 15.8.

Following Wattier's example, Figures 24-26 are stress vs. Larson-Miller parameter plots for $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal in the post weld heat treated condition. Figure 24 is for weld metal with carbon contents less than 0.05%. This data was obtained from studies by Campbell (18), Voorhees and Freeman (22), Klueh and Canonico (67) and Leyda et al. (83). The data shows some scatter at higher stresses, but is generally linear. Figure 25 presents stress-parameter data for weld metal with intermediate carbon contents, between 0.05 and 0.10%. This data is from work by Cross and Simmons (3), Hopkin et al. (19), Voorhees and Freeman (22), Emmanuel and Leyda (32), Williamson and Branch (39), Leyda et al. (83) and Emmanuel Leyda and Rozic (116). There is a fair amount of scatter present in the

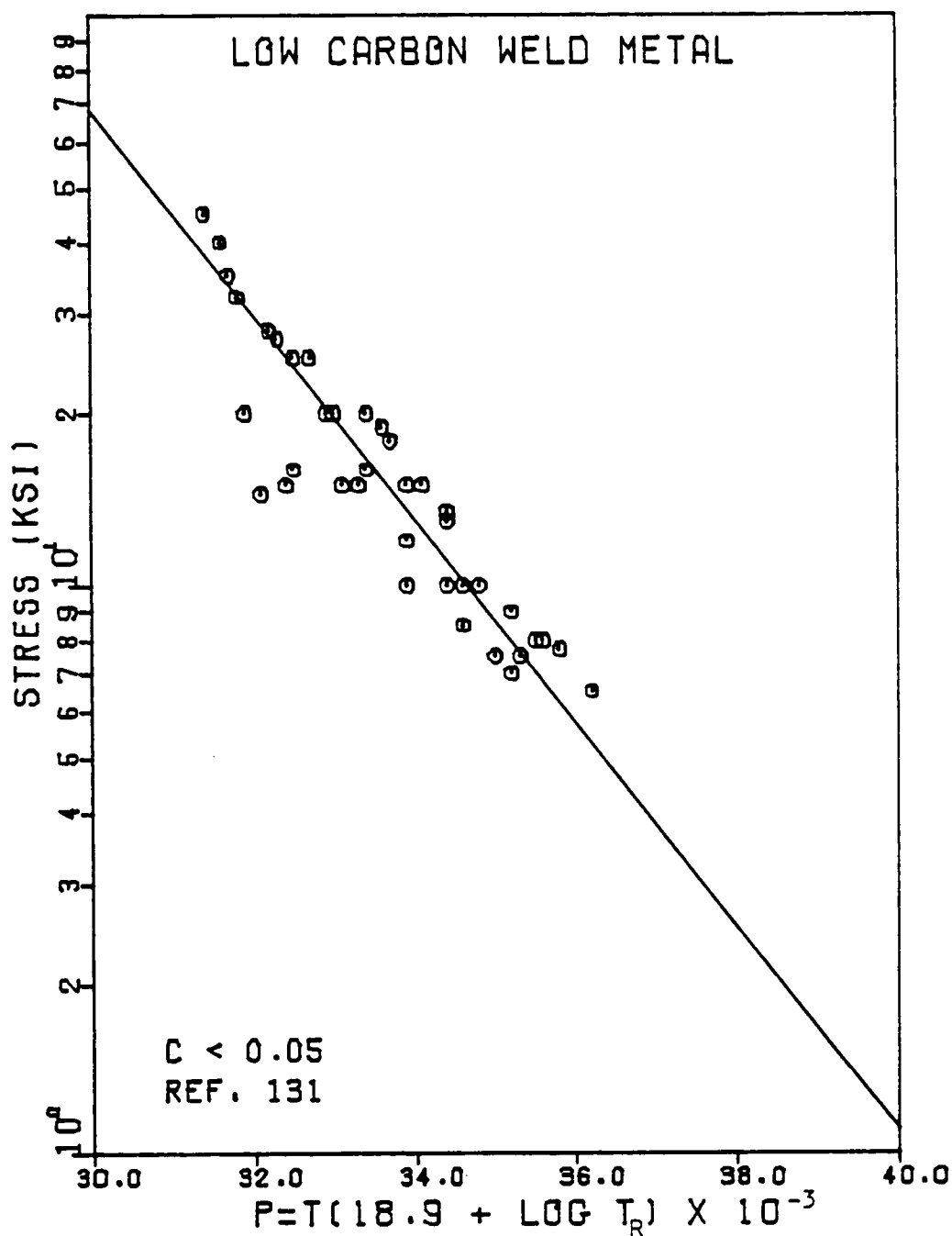


Figure 22. Stress-parameter data for low carbon 2½Cr-1Mo weld metal (after Reference 131).

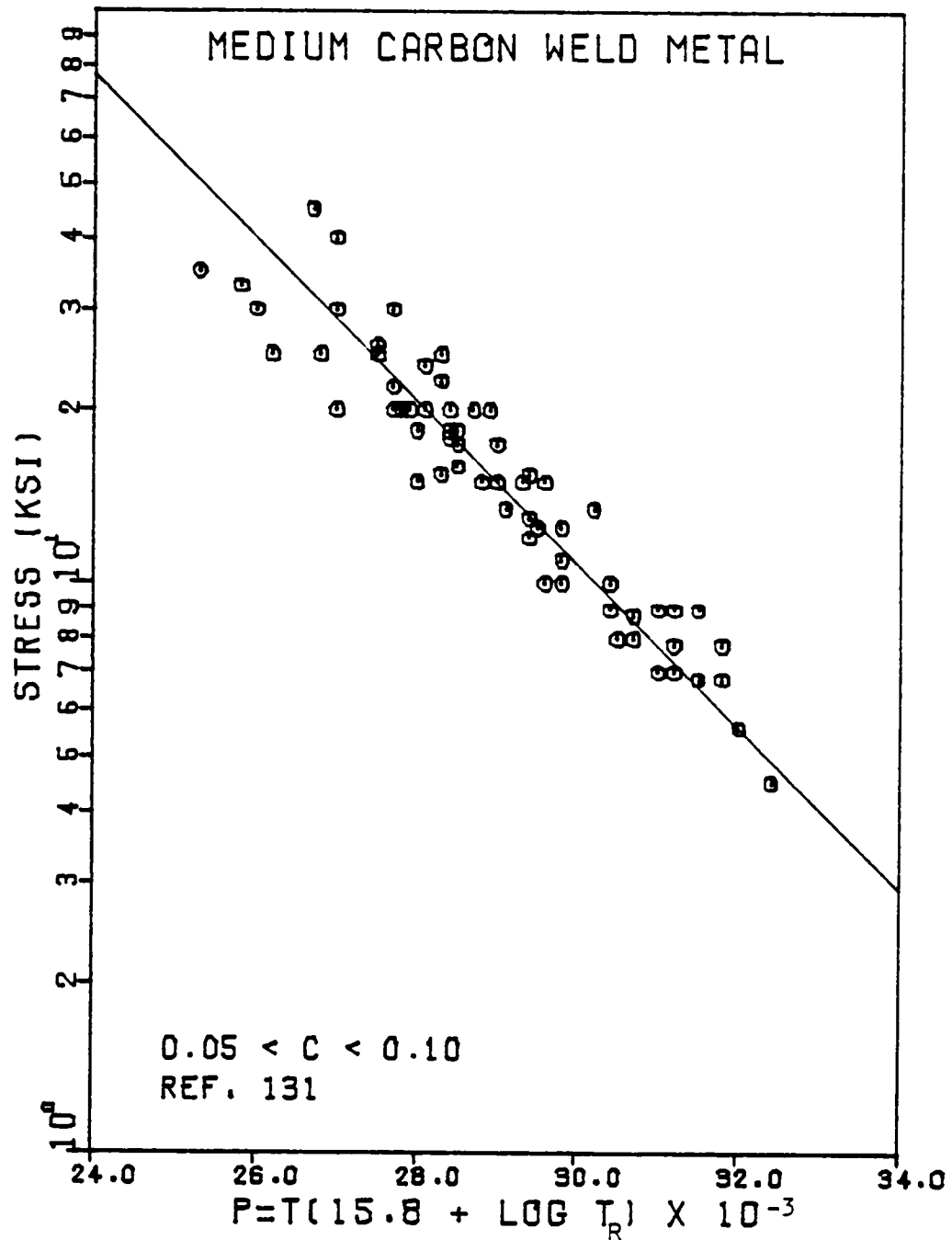


Figure 23. Stress-parameter data for medium carbon $2\frac{1}{2}\text{Cr}-1\text{Mo}$ weld metal (after Reference 131).

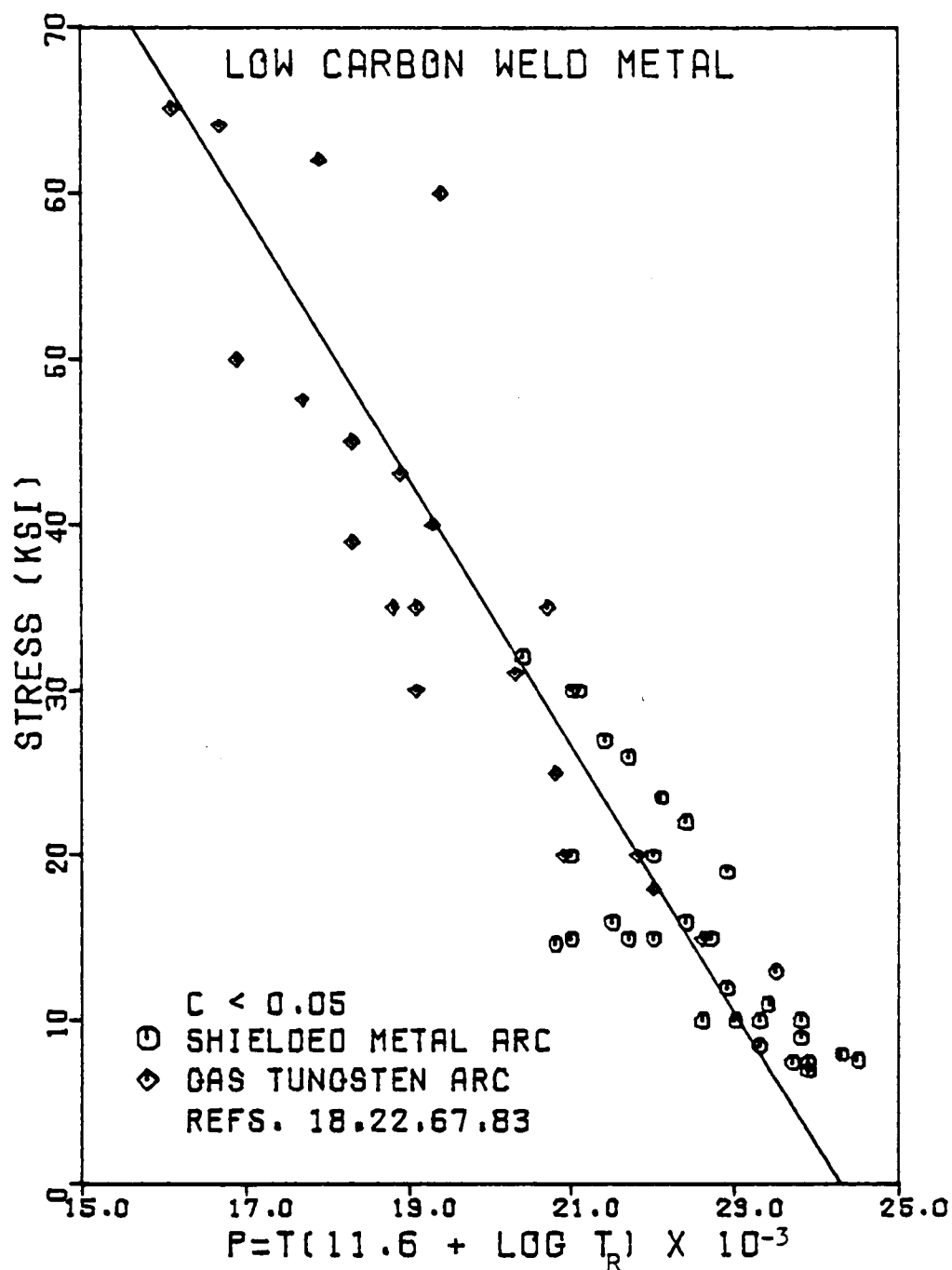


Figure 24. Stress-parameter data for low carbon $2\frac{1}{2}\text{Cr}-1\text{Mo}$ weld metal.

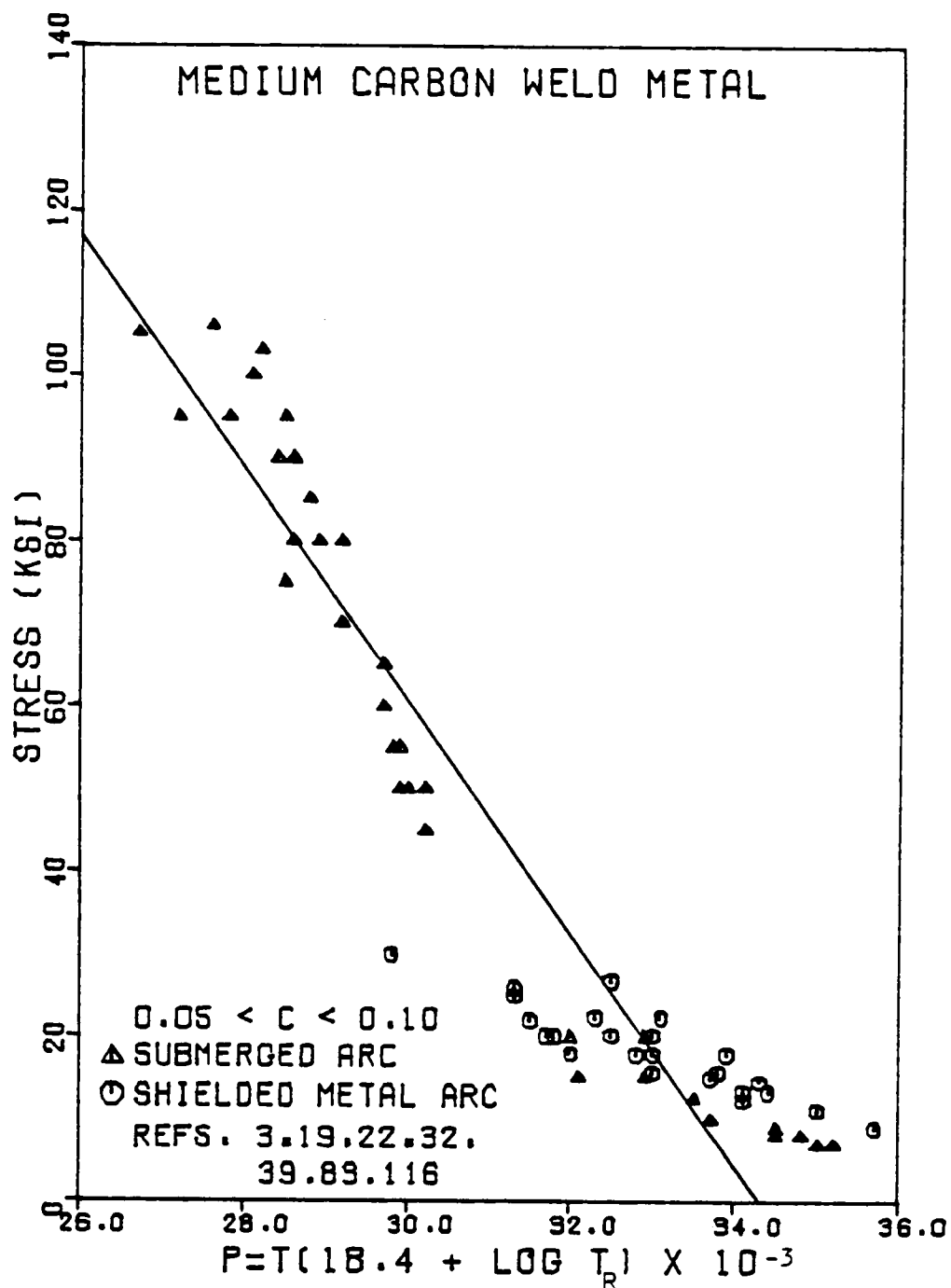


Figure 25. Stress-parameter data for medium carbon 2½Cr-1Mo weld metal.

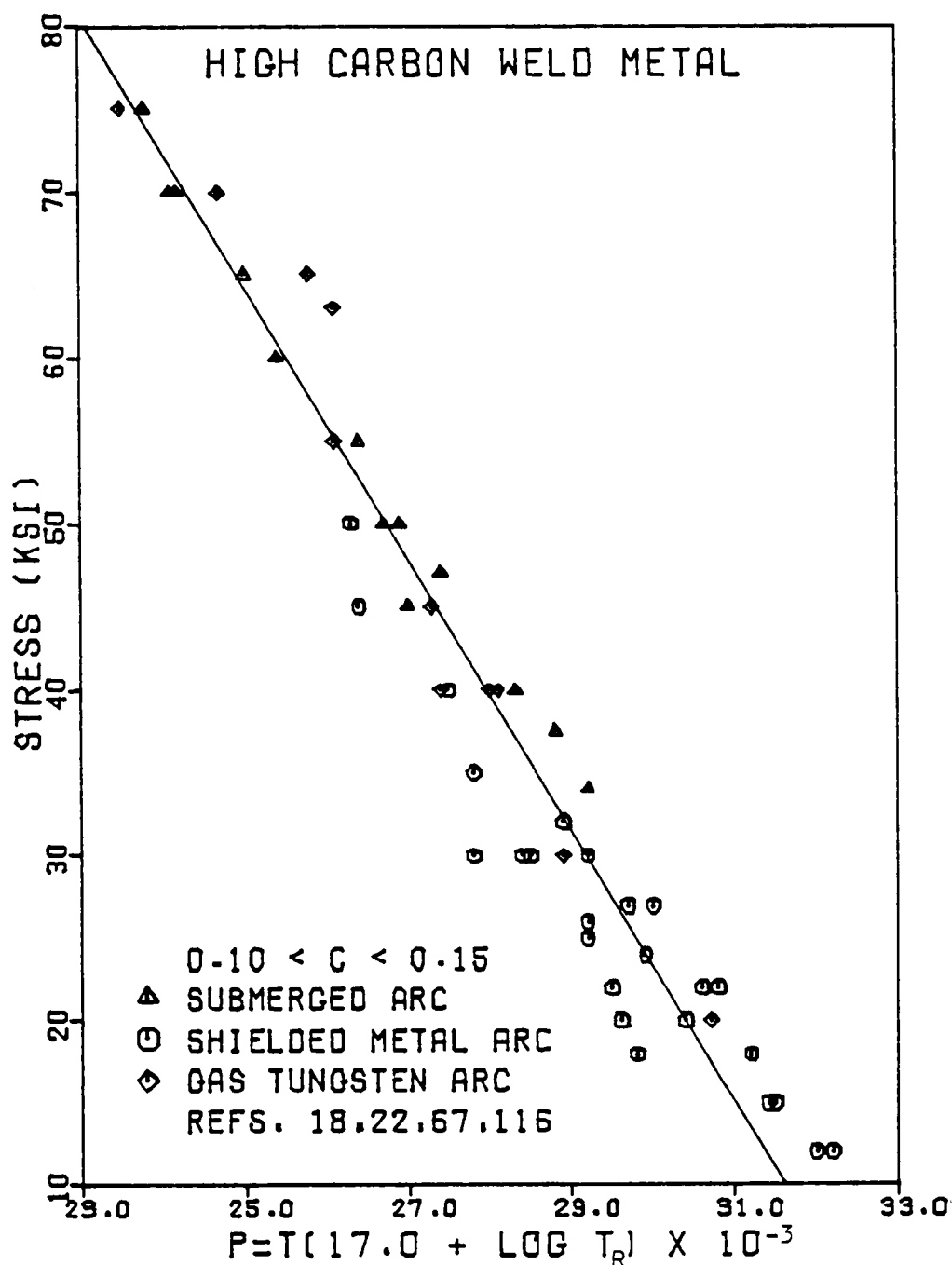


Figure 26. Stress-parameter data for high carbon $2\frac{1}{2}\text{Cr}-1\text{Mo}$ weld metal.

data. Figure 26 is a stress-parameter plot for high carbon (greater than 0.10%) weld metal from studies by Campbell (18), Voorhees and Freeman (22), Klueh and Canonico (67) and Emmanuel, Leyda and Rozic (116). This data is markedly linear and is characterized by a constant $C = 17.0$.

Figure 27 is a stress-parameter plot for quenched and tempered weld metal. The data was obtained from studies by Carpenter and Emmanuel (29), Emmanuel and Leyda (32), Gulya (86), Emmanuel, Leyda and Rozic (116) and Sangdahl and Voorhees (119). Very little scatter is observed in the data.

Conclusion

The preceding discussion of the available data on 2½Cr-1Mo weld metal reveals an almost total lack of data in the following areas:

1. Post weld heat treated shielded metal arc weld metal below 1000 F.
2. Post weld heat treated electroslag weld metal.
3. Quenched and tempered electroslag weld metal above 900 F.

In addition, there is no data for which it is possible to compare the relative strengths of shielded metal arc, submerged arc and electroslag weld metals at the same temperature. At most a comparison can be made between 2 of these processes, and even then difficulties arise due to differences in base metal composition and heat treatment.

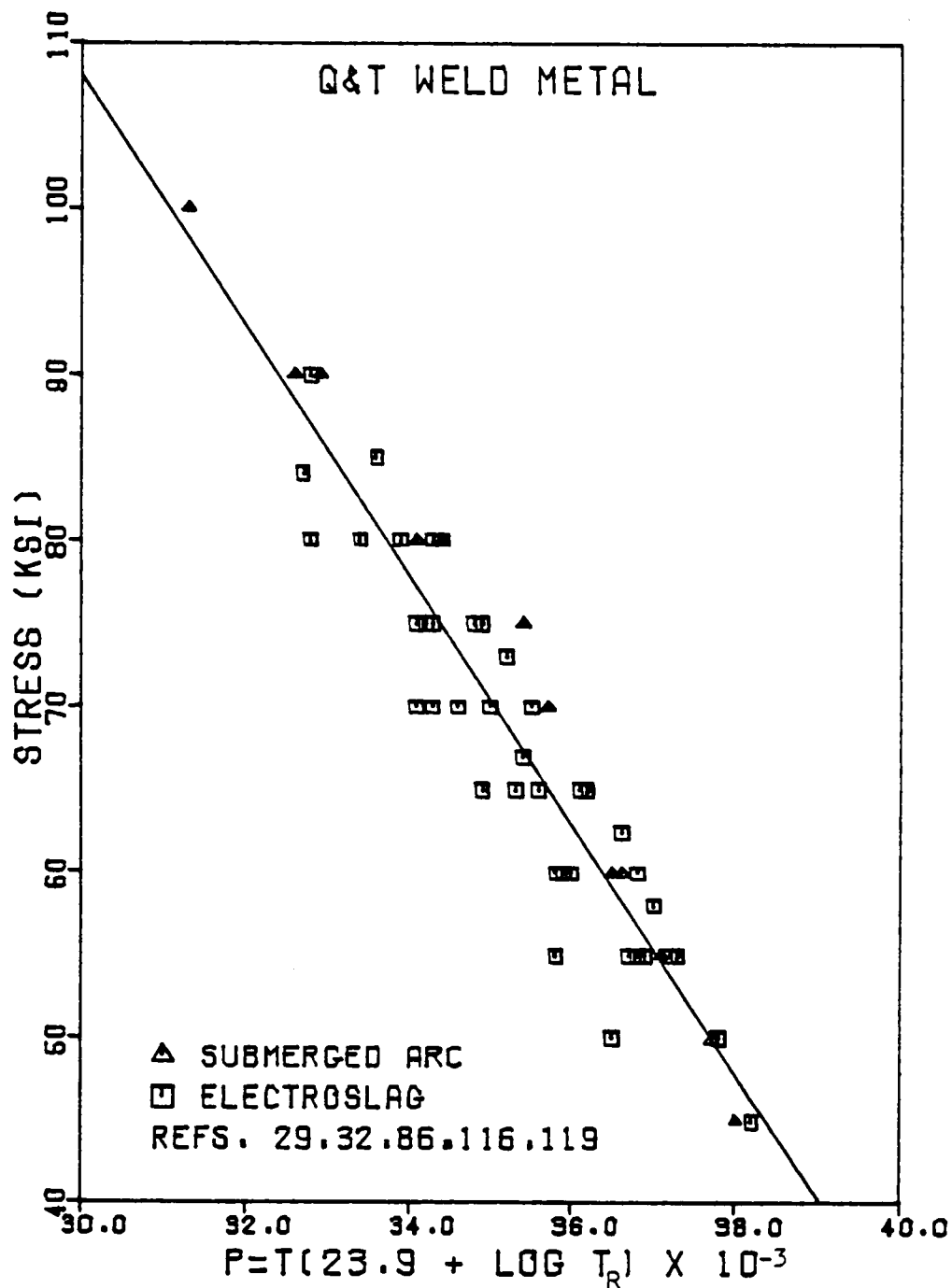


Figure 27. Stress-parameter data for quenched and tempered 2½Cr-1Mo weld metal.

CHAPTER III

TEST MATERIALS AND EXPERIMENTAL PROCEDURE

As pointed out in the preceding data review, little information is available on the stress-rupture properties of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal. Due to the lack of weld metal data, it is difficult to study the effects of welding process type, post weld heat treatment, and other variables on stress-rupture strength. In a response to the paucity of weld metal data, the Pressure Vessel Research Committee of the Welding Research Council sponsored a program at The University of Tennessee, Knoxville to investigate the stress-rupture properties of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal.

Test Materials

Most heavy wall $2\frac{1}{4}\text{Cr-1Mo}$ pressure vessels are built of SA387 Grade 22 material treated to class 2 properties (75 ksi ultimate tensile strength). This material is also the prime candidate for future coal gasification vessels. The chemical composition ranges for SA387 Grade 22 steel are listed in Table 2. Thus, due to the widespread current and expected future use of this material, SA387 Grade 22 steel was selected as the base material for the weldments to be fabricated for the program. The material was in the form of $2\frac{1}{2}$ inch thick plate. In order to avoid variations in base metal chemistry between the different weldments, the base metal for all 3 weldments came from the same heat. The composition of the base metal in this study is given in Table 3.

A principal aim of the program was to compare the stress-rupture properties of $2\frac{1}{4}\text{Cr-1Mo}$ weld metals deposited by different welding processes.

TABLE 2
CHEMICAL COMPOSITION REQUIREMENTS FOR SA387 GRADE 22 STEEL
(WEIGHT PERCENT)

Carbon	0.150 max
Manganese	0.25-0.66
Phosphorous	0.035 max
Sulfur	0.035 max
Silicon	0.50 max
Chromium	1.88-2.62
Molybdenum	0.85-1.15

TABLE 3
CHEMICAL COMPOSITION OF BASE METAL
(SA387 GRADE 22) USED IN FABRICATION OF WELDMENTS
(WEIGHT PERCENT)

Carbon	0.104
Manganese	0.40
Phosphorous	0.008
Sulfur	0.022
Silicon	0.22
Chromium	2.25
Molybdenum	0.94

Three welding processes were selected for study because of their widespread application in industry: shielded metal arc, submerged arc and electroslag welding. Three weldments, each utilizing one of the fore-mentioned processes, were manufactured by major fabricators using standard welding procedures. The weldments were to be fabricated so as to meet SA387 Grade 22, class 2 properties (75 ksi ultimate tensile strength, 40 ksi yield strength).

After welding, each weldment was post weld heat treated at 1300 F for 25 hours. This PWHT is similar to that used in the fabrication of heavy wall pressure vessels.

After PWHT, the welds were radiographed to reveal any defects or discontinuities. Defects were to be reported, but not repaired.

The following paragraphs will discuss the fabrication of the 3 weldments. Welding procedures and process variables will be detailed for the 3 weldments.

The shielded metal arc weldment was manufactured by Babcock and Wilcox. Over 97 individual passes were required to complete the weld. The electrodes employed were of the E9015 B3L type. The chemical composition of weld pads from the electrodes and the actual weld metal is given in Table 4. For the first 17 passes, 1/8 inch electrodes were used. The weld was completed with 5/32 inch electrodes. The welding process parameters are listed in Table 5. Figure 28 shows a macrograph of the SMA weld.

A preheat temperature of 300 F was specified. This temperature was maintained until welding was completed. The maximum interpass temperature was 500 F. After welding, the entire weldment was post weld heat treated

TABLE 4
CHEMICAL COMPOSITION OF SMAW WELD METAL
AND WELD PADS FROM E9015-B3L ELECTRODES
(WEIGHT PERCENT)

	1/8" Electrode	5/32" Electrode	Weld Metal
C	0.033	0.034	0.028
Mn	0.76	0.38	0.36
P	0.011	0.011	0.008
S	0.018	0.018	0.021
Si	0.63	0.72	0.82
Cr	2.13	2.20	2.19
Mo	0.90	0.94	0.91

TABLE 5
WELDING PROCESS VARIABLES FOR SMAW WELDMENT

Zone Number	Passes	Electrode Size	Travel Speed (in/min)	Current (A)	Voltage (V)
1	1-17	1/8 inch	6	145	110-150
2	18-97	5/32 inch	8	195	150-200

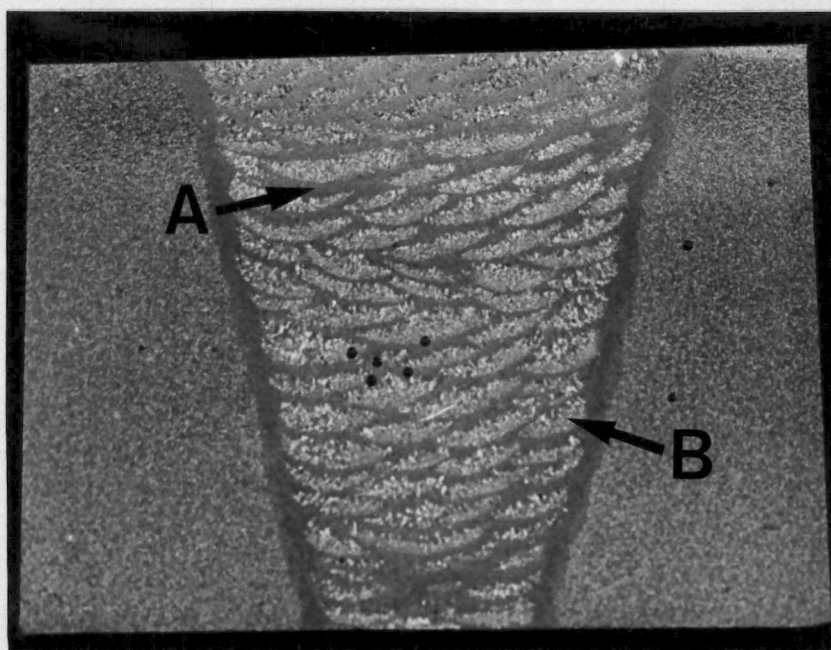


Figure 28. Macrograph of SMAW weld. Areas labeled "A" and "B" refer to fine grained ferritic regions and coarse grained ferritic regions, respectively.

at 1300 F for 25 hours. Following PWHT, the weldment was furnace cooled at 50 F/hour to 600 F, and then cooled in air. After PWHT, the SMAW weldment received a radiographic examination. No defects were reported.

The submerged arc weld, fabricated by Chicago Bridge and Iron, was composed of 27 individual passes. The electrode wire employed was Arcos Type 521 MC. The flux used was Linde Type 0091. The compositions of the electrode wire and deposited weld metal are given in Table 6. The welding process parameters are listed in Table 7. Figure 29 presents a macrograph of the weld.

As with the SMAW weld, a preheat of 300 F was maintained until the weld was completed. The maximum interpass temperature was 500 F. The weldment was post weld heat treated at 1300 F for 25 hours. The weld was radiographed after the PWHT. No defects were reported.

The electroslag weld was fabricated by the Foster-Wheeler Energy Corporation. The weld consists of a single pass, as shown in Figure 30. Raco Type 521 electrode wire was employed with Arcosite BV flux. The compositions of the electrode wire and weld metal are given in Table 8. The welding process parameters are listed in Table 9.

Due to the high heat input inherent in the electroslag welding process, no preheat was required. After welding, the weldment was normalized by heating to 1650 F, followed by cooling in still air. The weld was then post weld heat treated at 1300 F for 25 hours. The weld was radiographed after the PWHT. No defects were reported.

A portion of the electroslag weld was removed and returned to Foster Wheeler for a quenching and tempering treatment to increase the strength. This material was normalized at 1650 F for 2¼ hours and then

TABLE 6
CHEMICAL COMPOSITIONS OF SAW WELD METAL AND ELECTRODE WIRE
(WEIGHT PERCENT)

	Electrode Wire	Weld Metal
C	0.136	0.124
Mn	0.79	0.38
P	0.012	0.010
S	0.019	0.019
Si	0.195	0.034
Cr	2.44	2.14
Mo	0.997	1.00

TABLE 7
WELDING PROCESS VARIABLES FOR SAW WELDMENT

Zone Number	Passes	Electrode Size	Travel Speed (in/min)	Current (A)	Voltage (V)
1	1-23	3/16 inch	20-24	600-675	30-34
2	24-27	3/16 inch	20-24	600-650	32-34

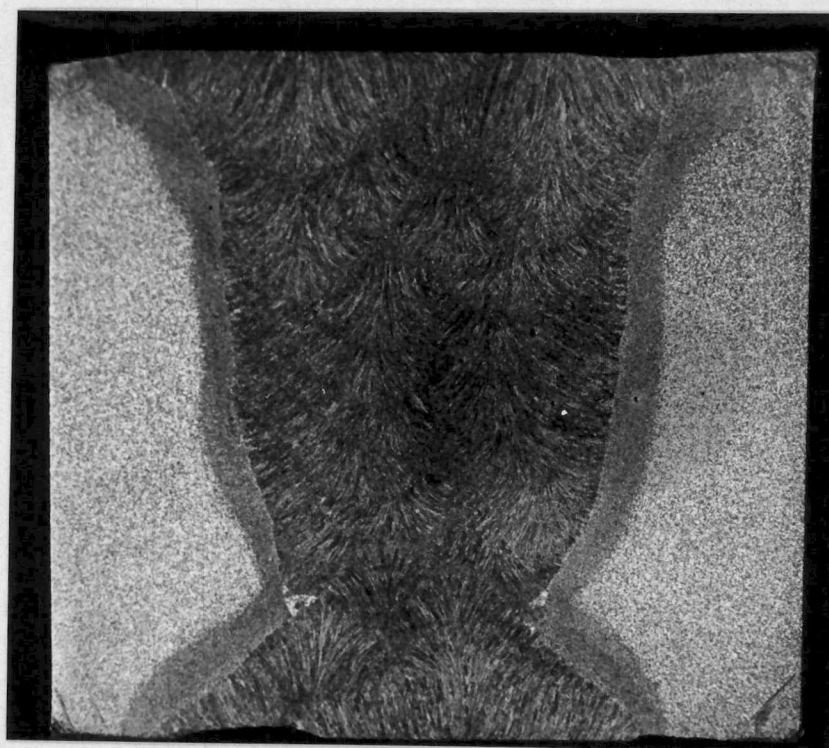


Figure 29. Macrograph of SAW weld.

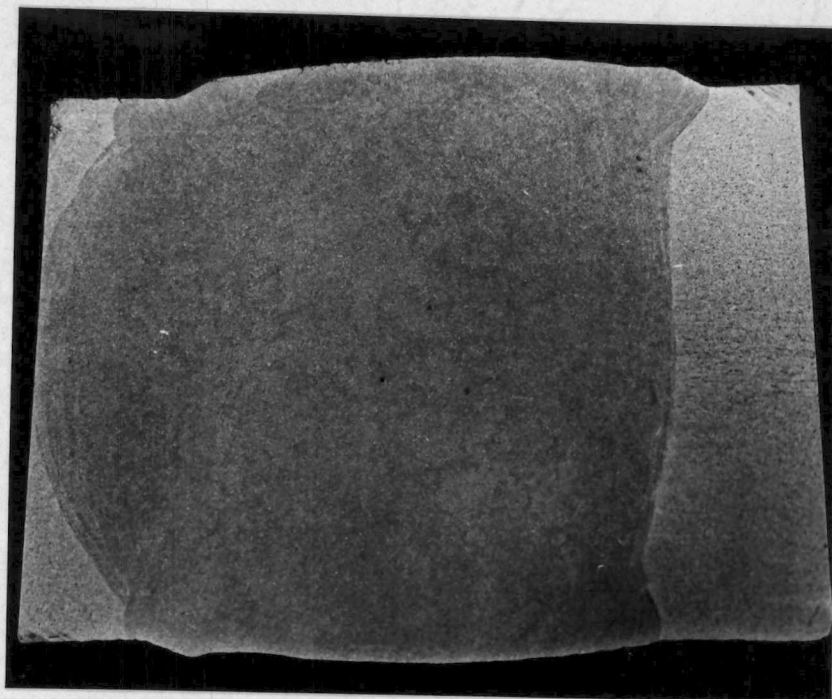


Figure 30. Macrograph of ESW weld.

TABLE 8
CHEMICAL COMPOSITIONS OF ESW WELD METAL AND ELECTRODE WIRE
(WEIGHT PERCENT)

	Electrode Wire	Weld Metal
C	0.096	0.107
Mn	0.61	0.44
P	0.017	0.010
S	0.010	0.016
Si	0.09	0.06
Cr	2.56	2.40
Mo	1.10	0.94

TABLE 9
WELDING PROCESS VARIABLES FOR ESW WELD

Passes	1
Wire Size	1/8 inch
Current	530-600 A
Voltage	44 V

water spray quenched. The material was then tempered at 1300 F for 25 hours, a temper identical to that received by the other weldments.

Test Specimens

The 3 weldments and base material were sectioned to provide specimens for stress-rupture testing, room temperature and elevated temperature tension testing, and metallographic examination.

All-weld metal stress-rupture specimens were removed along the longitudinal axis of each weld, as shown in Figure 31. The stress-rupture specimens had a gage length of 1.125 inches and a gage diameter of 0.125 inches, as shown in Figure 32. This specimen design was chosen to allow the extraction of specimens from weldments of varying size. This specimen design also allows for the extraction of transverse specimens containing weld metal, heat affected zone (HAZ), and base metal within the same gage length.

In addition to the stress-rupture specimens, all-weld metal specimens for room temperature and elevated temperature tension tests were extracted along the longitudinal axis of the 3 weldments. Base metal tension specimens with longitudinal orientations were employed for comparative purposes. The room temperature and elevated temperature tension test specimens conform to ASTM E8-79.

Testing Procedure

The following paragraphs will discuss the equipment and operating procedure used in the conduct of the stress-rupture tests. Testing procedure for the room temperature and elevated temperature tension tests will also be discussed.

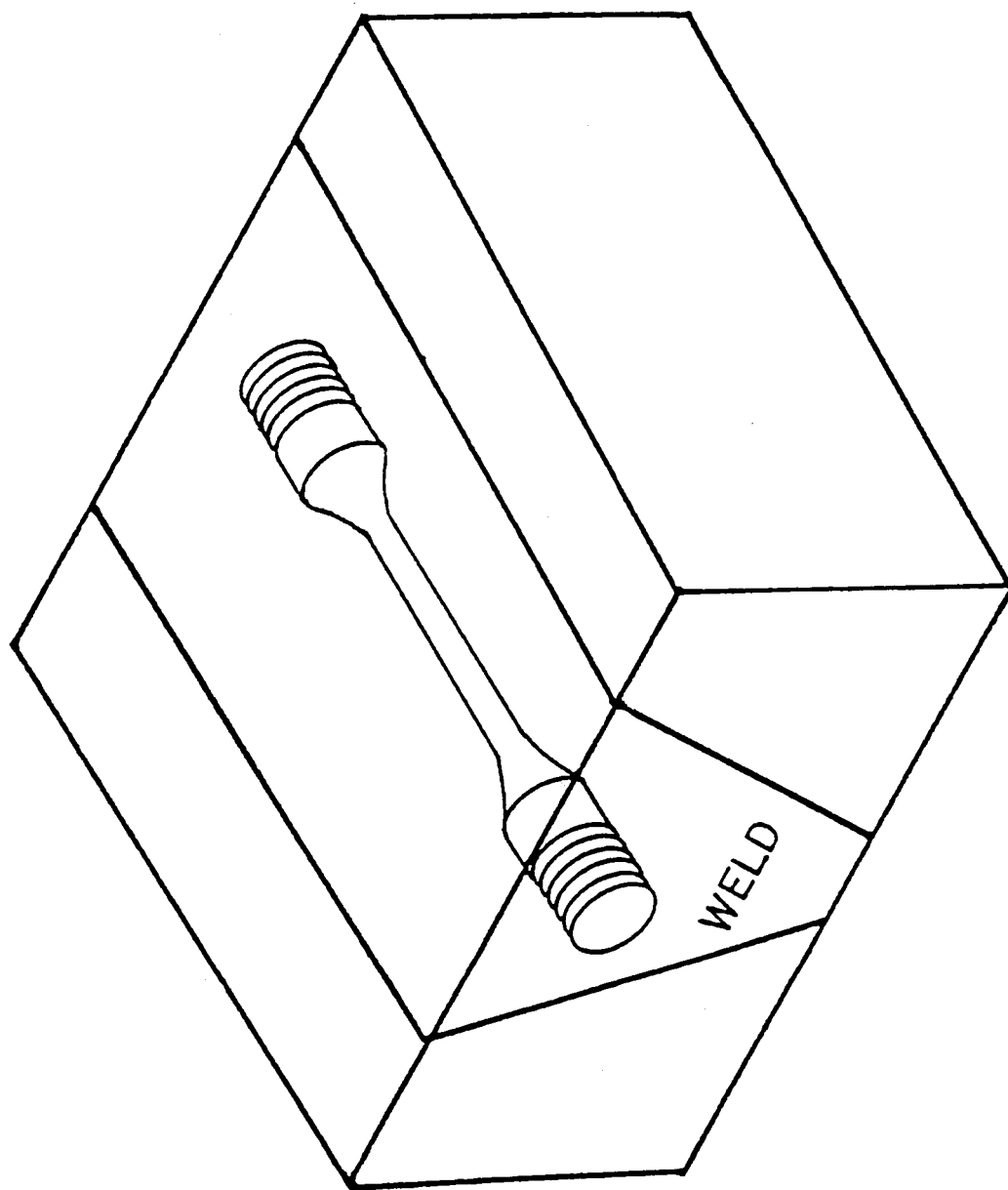


Figure 31. Schematic of all-weld metal stress-rupture specimen.

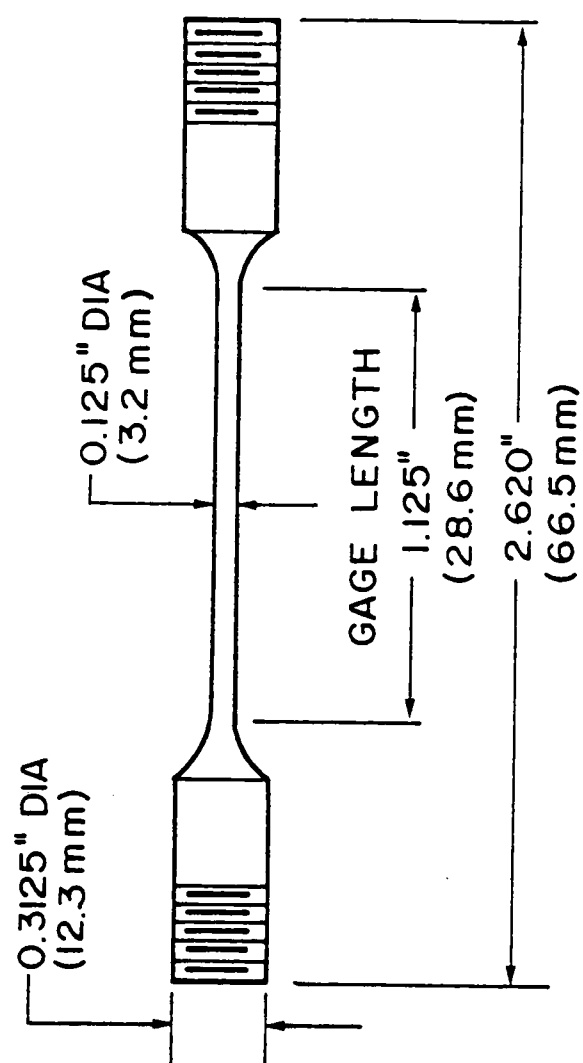


Figure 32. Stress-rupture test specimen.

Stress-rupture testing comprised a major part of the experimental work involved in the program. Over 100 tests were conducted on specimens extracted from the 3 weldments. Stress-rupture tests were conducted at 850 F, 950 F and 1050 F. These 3 temperatures were selected because they cover the current and anticipated operating range of pressure vessels constructed of 2½Cr-1Mo. Test times ranged from 1 to nearly 10,000 hours. Duplicate tests were conducted at each stress level, in order to determine the reproducibility of test results.

The stress-rupture tests were conducted on 8 constant-load dead-weight type creep frames. The design of the creep frames is shown in Figure 33. In the conduct of the test, a stress-rupture specimen was placed in the specimen holder and loaded in the creep frame. When loaded, the specimen was located in the center of a 3 zone furnace. The zones were independently controlled, allowing for the establishment of a constant temperature profile (± 1 F) over the specimen gage length. A thermocouple wired to the sample provided accurate temperature readings. The specimen was stressed by means of a deadweight load. Changes in specimen length were measured by an extensometer mounted on the specimen holder. The duration of each test was measured by a clock which shut off when the test terminated.

Experimental observations made during the conduct of each test included periodic strain readings, temperature measurements, and time to rupture. Other information was also recorded including stress level, specimen identification number, and specimen dimensions prior to and after testing.

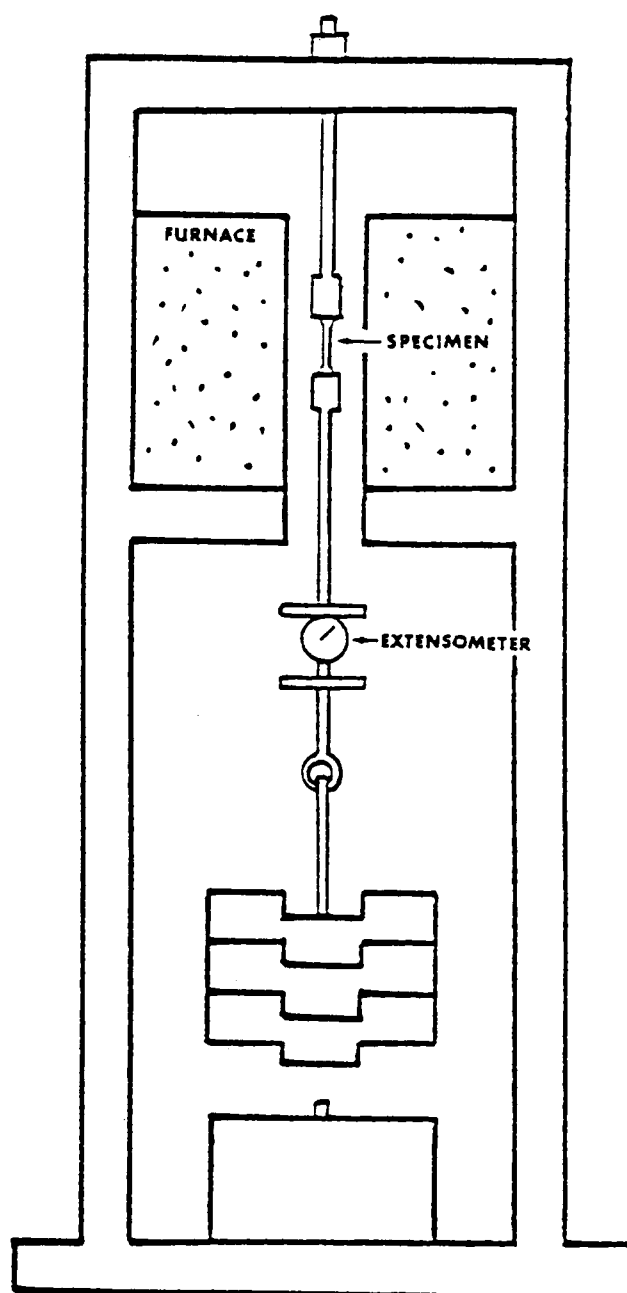


Figure 33. Schematic of creep frame.

Room temperature tension tests were conducted according to ASTM E8-79. Both all-weld metal and base metal specimens were tested.

Elevated temperature tension tests were performed at 850 F, 950 F and 1050 F (139).

CHAPTER IV

RESULTS AND DISCUSSION

The preceding literature and data review have emphasized the need for more information on the high temperature properties of 2½Cr-1Mo weld metal deposited by various welding processes. The current program was initiated at The University of Tennessee, Knoxville in direct response to the need for more data. This chapter discusses the results of the experimental work to date and compares these with the results of previous studies.

Stress-rupture Test Results

Stress vs. time to rupture curves were obtained at 850 F, 950 F and 1050 F for SAW, SMAW and ESW weld metal. Stress-rupture data was obtained for the ESW weldment in both the normalized and tempered and quenched and tempered conditions. Figures 34-36 present stress-rupture data for the SAW, SMAW and ESW (normalized and tempered) weld metals at 850 F, 950 F and 1050 F, respectively. The individual data points shown on these curves are listed in the Appendix. The solid lines drawn through each set of weld metal data represent least-squares line fits of the data. The dashed line shown on each curve represents minimum base metal properties from Code Case N-47 (Section III) of the ASME Boiler and Pressure Vessel Code (85).

Figure 34 shows the stress-rupture properties of the 3 weld metals at 1050 F. The regression coefficients, R^2 , obtained for the least-squares line fits are 98.9% for the SAW data, 98.7% for the SMAW data, and 99.4%

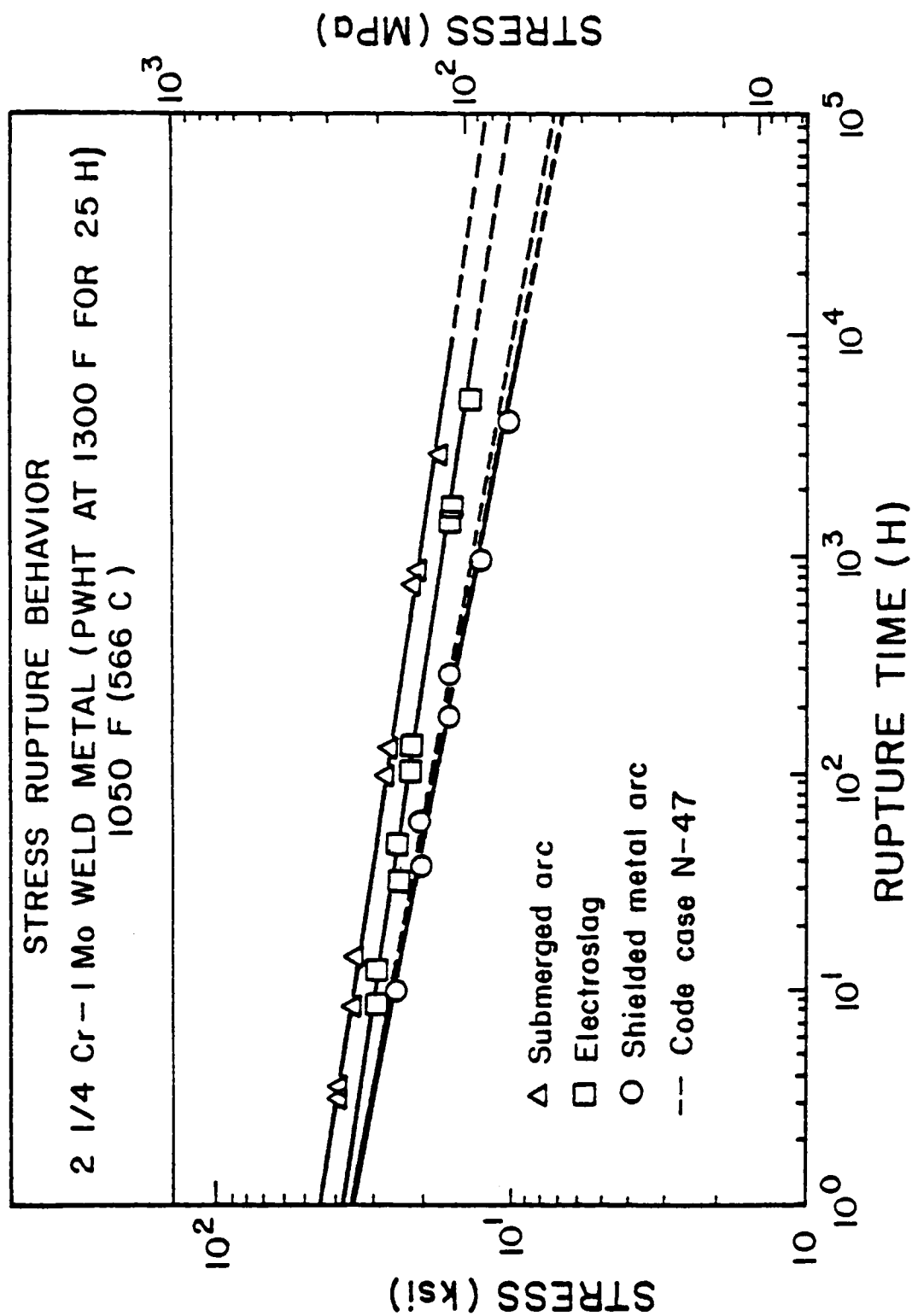


Figure 34. Stress-rupture curves for 2 $\frac{1}{4}$ Cr-1Mo weld metals at 1050 F.

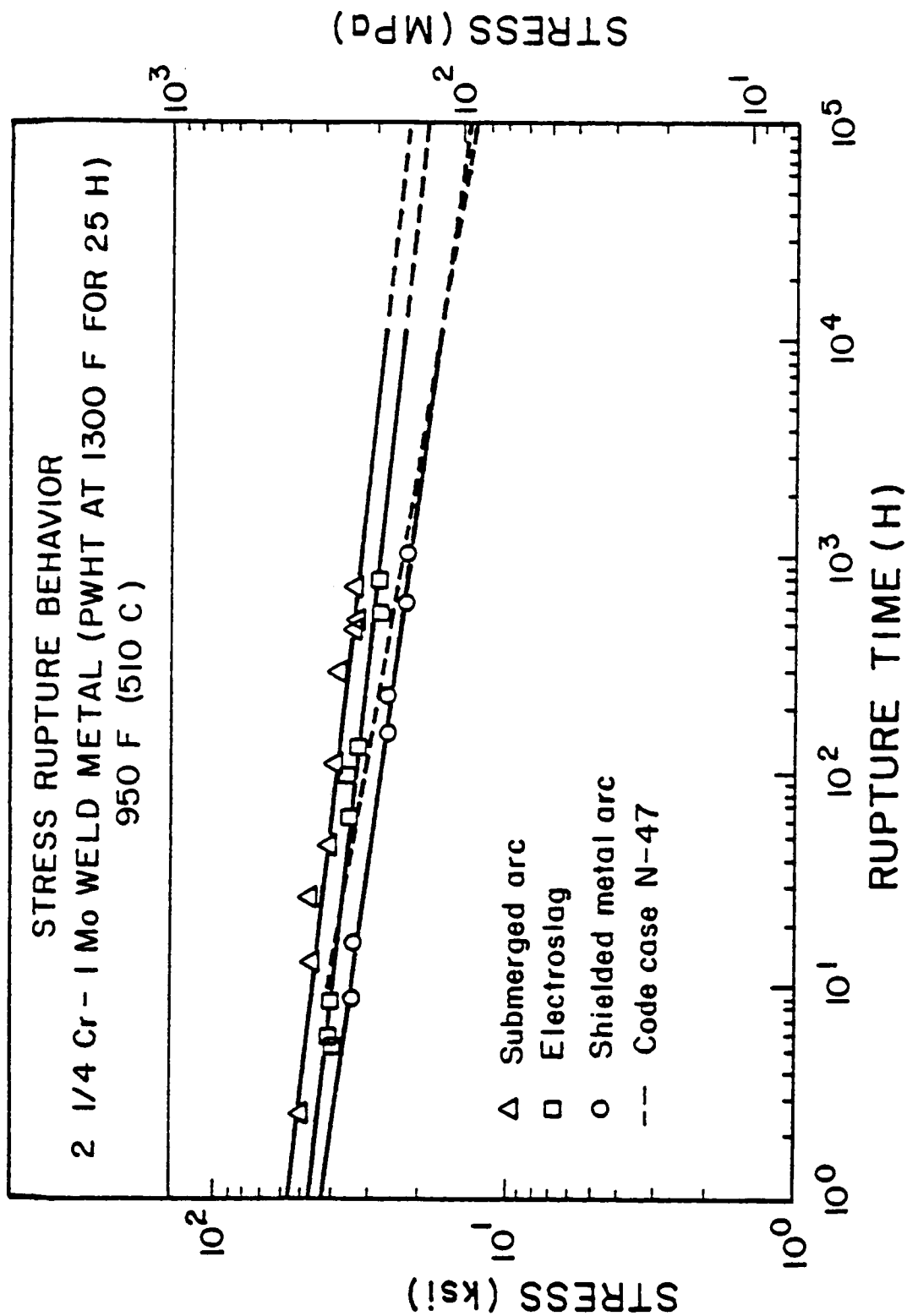


Figure 35. Stress-rupture curves for 2 1/4Cr-1Mo weld metals at 950 F.

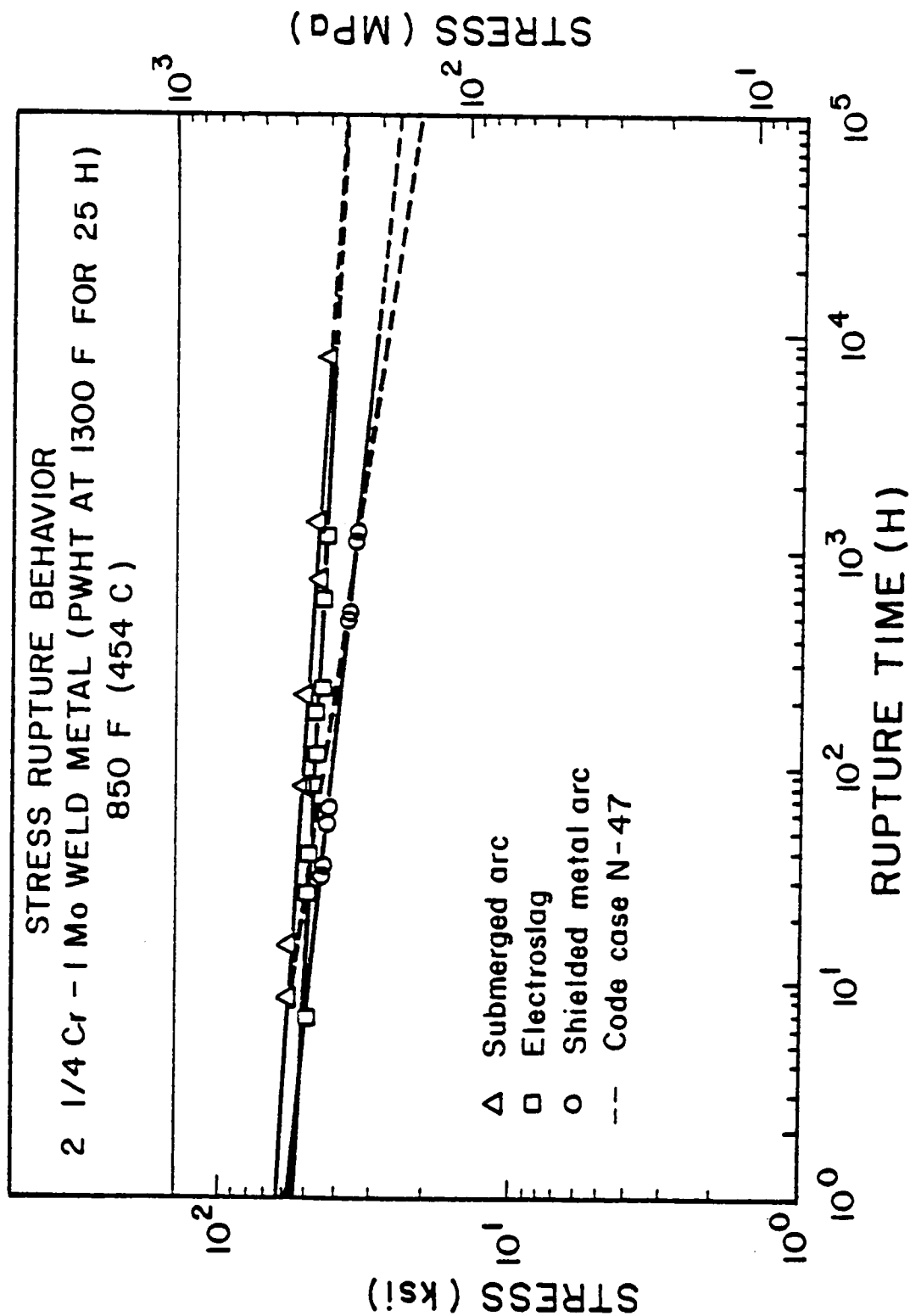


Figure 36. Stress-rupture curves for 2¼Cr-1Mo weld metals at 850 F.

for the ESW data. These coefficients represent excellent line fits, and hence good reproducibility of data. As can be observed from this figure, the SAW weld metal displays the highest stress-rupture strength, while the SMAW weld metal exhibits the lowest strengths. The ESW weld metal displays intermediate rupture strengths. The SAW and ESW rupture strengths lie above the minimum base metal properties represented by the Code Case N-47 reference line, while the SMAW rupture strength is slightly lower than the base metal values. An interesting observation is that the slopes of the Code Case N-47 reference line and the SMAW data are similar.

The stress-rupture behavior at 950 F, illustrated in Figure 35, is similar to that observed at 1050 F. Namely, the SAW weld metal again displays the highest stress-rupture strength, followed by the ESW material. The SMAW weld metal is again the weakest. As observed at 1050 F, the SAW and ESW data lie above the base metal reference line, while the SMAW data lies below it. The line fits are again very good, being characterized by regression coefficients of 98.0%, 93.3% and 95.3% for the SMAW, SAW and ESW data, respectively.

The results at 850 F, shown in Figure 36, are similar to those seen at 950 F and 1050 F. The SAW weld metal again displays the highest stress-rupture strength. The SMAW weld metal is again the weakest, except for short times where it shows slight superiority over the ESW material. The ESW weld metal again displays intermediate rupture strengths. Interestingly, the rupture strengths of the SAW and ESW weld metals appear to converge at very long times. The SAW and ESW data again lie above the Code Case N-47 base metal reference line, while the SMAW data lies below it, except for very long times. Once again, the least-squares line

fits are very good being characterized by regression coefficients of 99.4%, 97.3% and 95.3% for the SMAW, SAW and ESW data, respectively.

The longest stress-rupture test conducted was on the order of 10,000 hours. However, most designs are based on 100,000 hour properties. Since it is economically and practically unfeasible to conduct 100,000 hour stress-rupture tests, some means must be found to predict long-time rupture strengths from short-time data. Two such methods are parametric evaluations and straight-line extrapolation. The following paragraphs will discuss the application of these 2 methods to the data produced by this study.

As stated in Chapter II, parametric evaluations are mathematical relations in which rupture life is a function of temperature and stress. Since the parameter value is a function of stress, a smooth curve should result from a parameter vs. log stress plot. Such a curve can be extrapolated to predict stress-rupture behavior at times exceeding actual test times.

In this study, various stress-rupture parameters were considered in an effort to obtain the best parametric evaluation describing the stress-rupture behavior of the 3 weld metals. An Orr-Sherby-Dorn parameter with a linear stress dependence was found to best describe the data (132). The parametric evaluations obtained for the weld metals are shown below:

$$\text{SAW: } \log t_r = -17.091 - 0.02112 \sigma + 19380/T$$

$$\text{SMAW: } \log t_r = -18.158 - 0.02279 \sigma + 19320/T$$

$$\text{ESW: } \log t_r = -22.709 - 0.02692 \sigma + 24240/T$$

where t_r is time to rupture, σ is the stress in MPa, and T is the temperature in degrees K. The regression coefficients, R^2 , for these equations

are 95.2%, 96.5% and 95.2%, respectively, indicating an excellent data fit.

An alternate method of predicting long-time rupture strengths involves the straight-line extrapolation of short-time stress-rupture data to longer times. In a study on $2\frac{1}{4}\text{Cr-1Mo}$, Smith (112) found that straight-line extrapolation of short-time stress-rupture data yielded a better assessment of 100,000 hour rupture strength than did parametric techniques.

Table 10 lists the 100,000 hour rupture strengths for the 3 weld metals at the test temperatures. The SAW weld metal exhibits the highest long-time rupture strength at all 3 temperatures. The SMAW weld metal displays the lowest 100,000 hour rupture strengths, while the ESW material exhibits strengths which are intermediate as compared to the SAW and SMAW weld metals. Also shown in Table 10 are minimum base metal stress-rupture properties from Code Case N-47 (Section III) and Section VIII, Div. 1 of the ASME Boiler and Pressure Vessel Code (85). The 100,000 hour rupture strengths for Section VIII, Div. 1 were obtained by dividing the maximum allowable stresses in Table UCS-23 of Section VIII by 0.8, as defined in Appendix P. The SAW and ESW weld metals exceed the minimum base metal rupture strengths at all 3 temperatures. The SMAW rupture strength is higher than the minimum base metal values at 850 F, but is lower at 950 F and 1050 F.

The major criticism of the straight-line extrapolation method is that structural changes may occur within the material and change the slope of the log stress vs. log rupture time curve. Wilson (125) has stated that a decrease in steepness is often observed in log stress vs. log rupture time curves for $2\frac{1}{4}\text{Cr-1Mo}$ at 1000 F and 1050 F. However, Smith (112)

TABLE 10

ESTIMATED 100,000 HOUR RUPTURE STRENGTHS AND MINIMUM
BASE METAL VALUES

Weld Metals	<u>Stress (ksi)</u>		
	Test Temperature		
	1050 F	950 F	850 F
SAW	12.1	22.4	37.5
ESW	10.2	19.3	36.5
SMAW	6.6	13.7	24.9
<u>Minimum Base Metal Properties</u>			
Section VIII, Div. 1 ¹	7.25	13.8	20.4
Code Case N-47	7.0	12.6	21.0

¹100,000 hour rupture strength = maximum allowable stress
divided by 0.8
A387 Grade D - normalized and tempered

suggested that log stress vs. log rupture time curves for 2½Cr-1Mo are linear at temperatures up to 1100 F. In this study, it was decided that the straight-line extrapolation method could be used with confidence because of the stability of the as-PWHT microstructure. Post weld heat treatment conducted at temperatures significantly higher than the service temperature is known to stabilize the microstructure of the weld metal and HAZ (133).

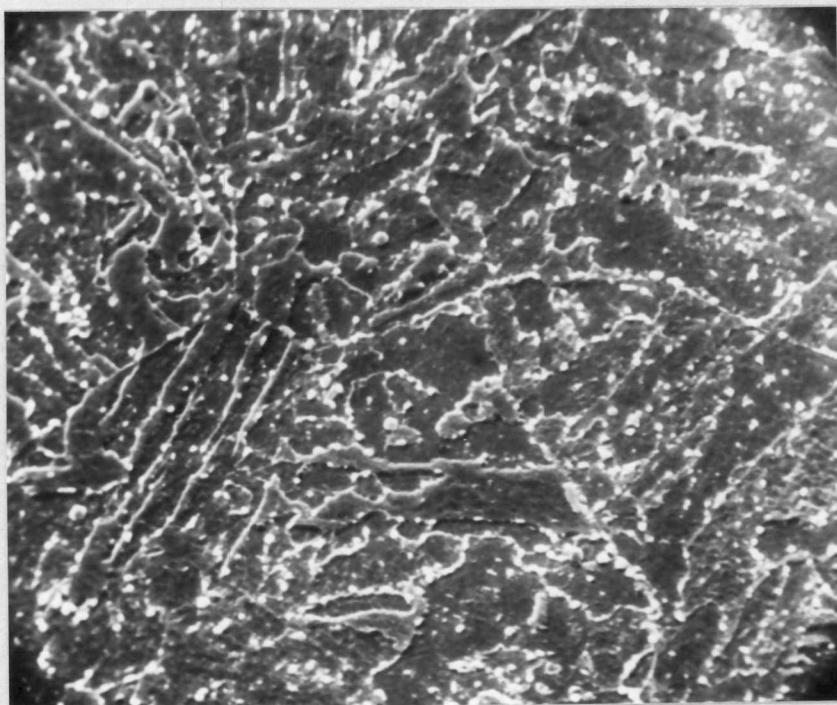
In this study, the stabilizing effect of the PWHT was confirmed by metallographic examination. The as-PWHT (after welding and PWHT but prior to exposure in the creep range) microstructure of the SAW weld consisted entirely of tempered bainite, as shown in Figure 37A. Figure 37B presents a scanning electron micrograph of the SAW weld. After almost 3000 hours at 1050 F, the microstructure of the SAW weld was essentially unchanged, as seen in Figure 38.

The ESW weld was also bainitic in the as-PWHT condition, as shown in Figure 39A. Figure 39B presents a scanning electron micrograph of the ESW weld. Figure 40 shows the microstructure after 1500 hours at 1050 F. Little change was noted in the microstructure.

The SMAW weld in the as-PWHT condition is seen in Figure 41. Because of the low carbon content of the SMAW weldment, it possessed a predominantly ferritic microstructure. The SMAW weld consists of alternating regions of fine grained ferrite and coarse grained ferrite (areas "A" and "B", respectively, in Figure 41A and 41B). These regions may also be observed in the macrograph of the SMAW weld, Figure 28, page 66. The fine grained regions result from multipass welding. When a weld pass is deposited over a previous pass, the prior pass experiences temperatures



A. Light micrograph, showing bainitic structure.
Polished and etched with 2% nital. 250X.



B. SEM micrograph. 2000X.

Figure 37. Micrographs of SAW weld in the as-PWHT condition.

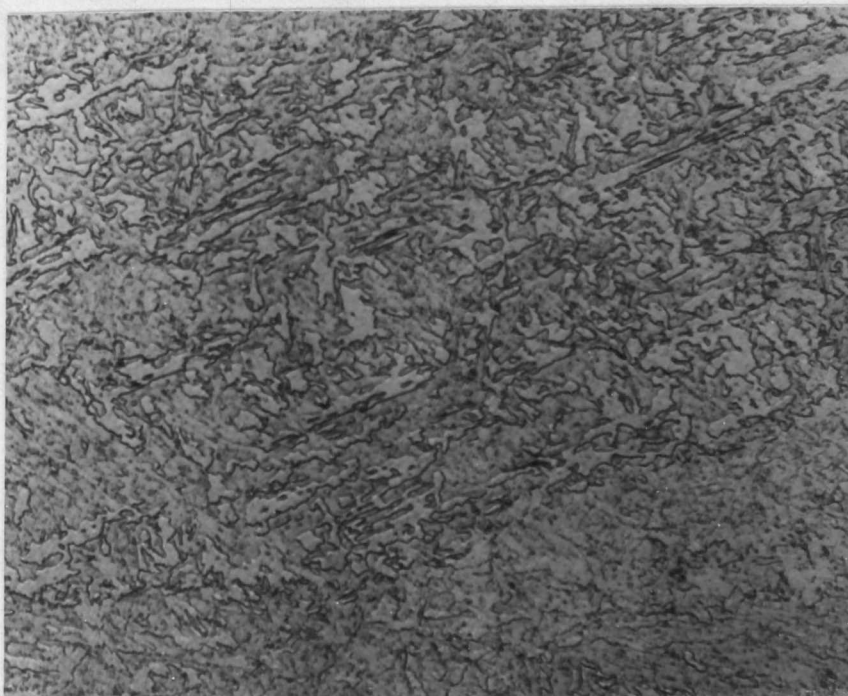
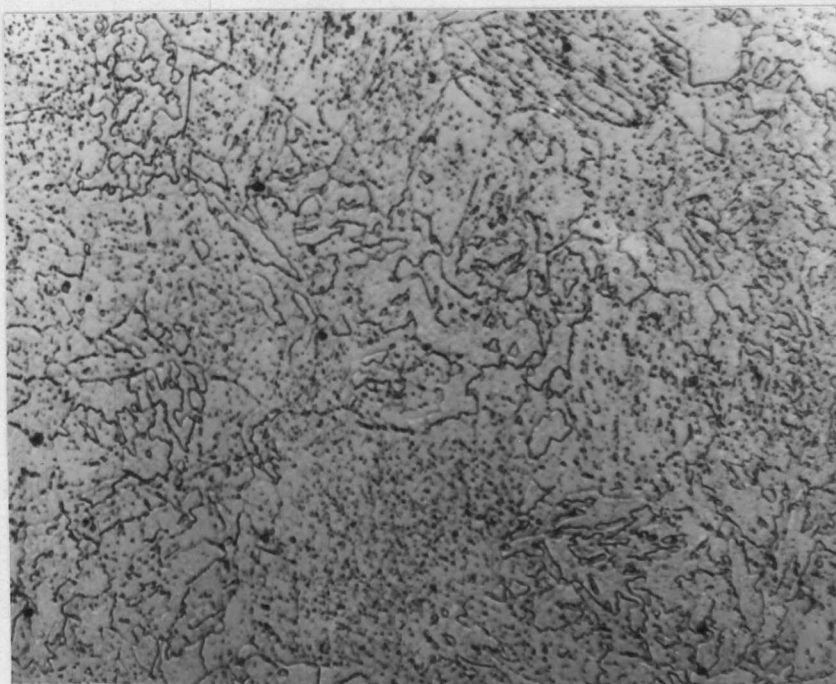
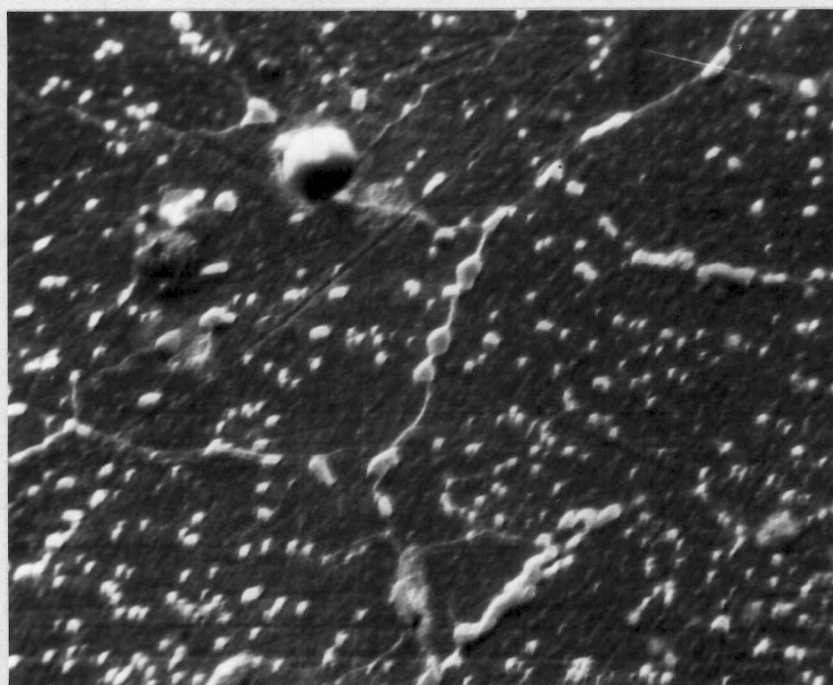


Figure 38. Light micrograph of SAW weld metal after 3000 hours at 1050 F showing bainitic structure. Polished and etched with 2% nital. 250X.



A. Light micrograph, showing bainitic structure.
Polished and etched with 2% nital. 250X.

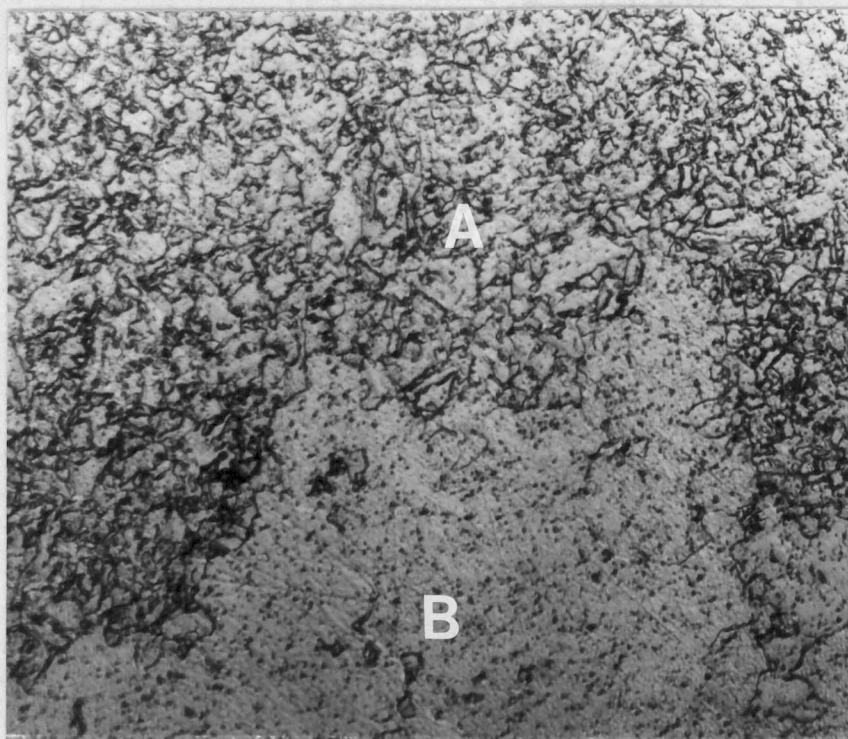


B. SEM micrograph. 2000X.

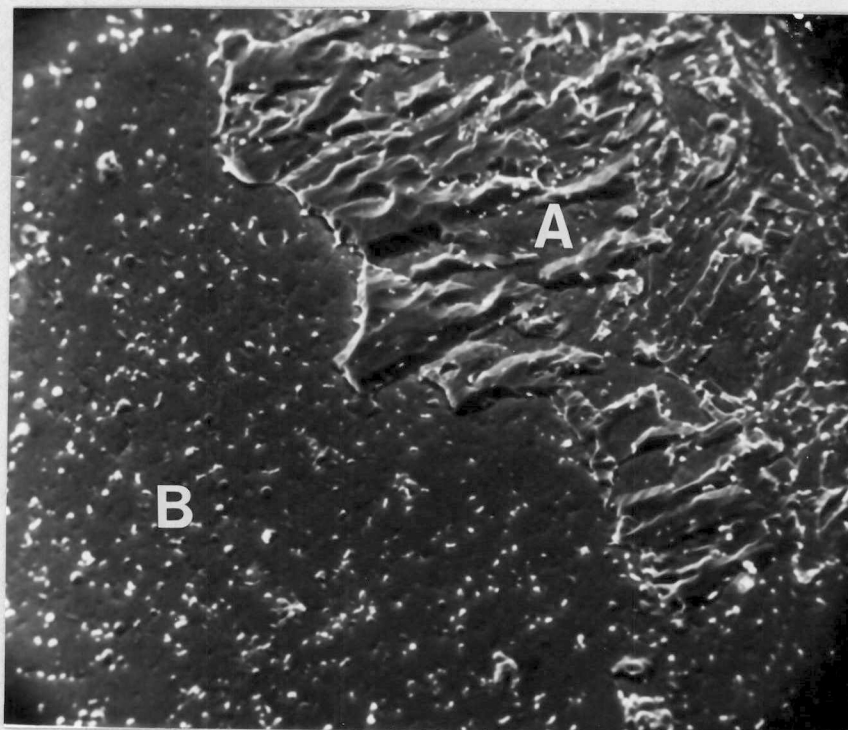
Figure 39. Micrographs of ESW weld in the as-PWHT condition.



Figure 40. Light micrograph of ESW weld metal after 1500 hours at 1050 F showing bainitic structure. Polished and etched with 2% nital. 250X.



A. Light micrograph. Polished and etched with 2% nital. 250X.



B. SEM micrograph. 1000X.

Figure 41. Micrographs of SMAW Weld in the as-PWHT Condition. Areas labeled "A" and "B" refer to fine grained ferritic regions and coarse grained ferritic regions, respectively.

ranging from the melting temperature to the preheat temperature. That portion of the pass heated below the A_1 temperature experiences tempering. The region heated between A_1 and A_3 is known as the partially transformed region. The portion of the weld which experiences temperatures from A_3 to about 2000 F completely transforms to fine grained austenite. Upon cooling, the austenite retransforms to a fine grained ferritic structure. Hence, this region is known as the fine grained region. Finally, the region which experiences temperatures above 2000 F up to the melting point forms a coarse grained ferritic microstructure upon cooling from coarse grained austenite. This area is therefore known as the coarse grained region. In a multipass weld, each of these regions will be adjacent to the individual passes. Since the SMAW weldment had a low carbon, essentially ferritic, microstructure, the grain refinement was more pronounced than in the higher carbon SAW and ESW welds. Thus, the microstructure of the SMAW weld consists of coarse grained ferrite regions and grain refined regions. The properties of the SMAW weld metal are a composite of the coarse grained and fine grained regions, and depend upon the relative volume of these regions. After 1000 hours at 1050 F, little change is noted in the microstructure of the SMAW weld, as shown in Figure 42.

As the preceding discussion pointed out, long-time extrapolations of short-time stress-rupture data depend upon the stability of the weld metal microstructure. The microstructure of $2\frac{1}{2}\text{Cr}-1\text{Mo}$ weld metals is essentially bainitic. During thermal aging, such as tempering and service exposure, the bainitic microstructure changes to a carbon saturated ferritic matrix with the precipitation of carbides (54). The carbides eventually coarsen and become stable through this diffusion controlled

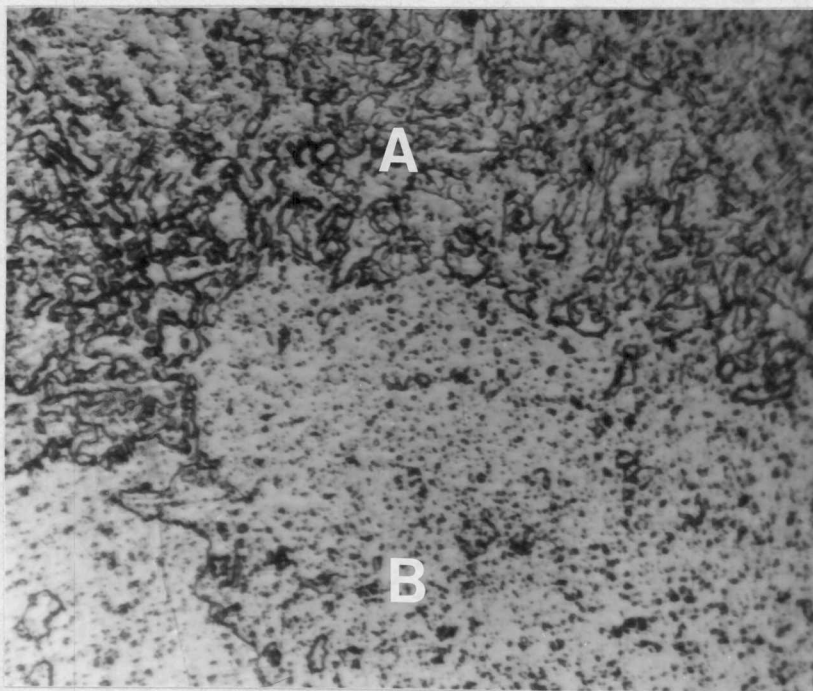


Figure 42. Light micrograph of SMAW Weld after 1000 hours at 1050 F. Areas labeled "A" and "B" refer to fine grained ferritic regions and coarse grained ferritic regions, respectively. Polished and etched with 2% nital. 250X.

temper process. The bainite, which is unstable at the service temperature, has higher stress-rupture strength than the ferrite/carbide microstructure. Long-time extrapolations based on an unstable microstructure such as bainite may not be valid and thus tempering for stability is a means by which the property predictions become more valid.

This point is emphasized in Table 11. In this table, the 100,000 hour rupture strength estimated for the SAW weld metal at 850 F is compared with data from Emmanuel, Leyda and Rozic (116) for SAW weld metal at 850 F. The 3 weldments listed in the table received different post weld heat treatments. The post weld heat treatments are ranked in severity by the Larson-Miller temper parameter:

$$T.P. = T(20 + \log t) \times 10^{-3}$$

where T is the tempering temperature in degrees R and t is the tempering time. The post weld heat treatments are further characterized by determining the equivalent exposure time at 850 F for each temper. These calculations are based on the Larson-Miller temper parameter. The 100,000 hour rupture strength of the SAW weld which was post weld heat treated at 1075 F for 6 hours is much higher than that for the other SAW welds which received much more severe tempers. Since the carbon contents are similar, it appears that the high rupture strength predicted for the weld which received a mild temper may be based on an unstable microstructure and, hence, may not be valid. This conclusion is confirmed by observing that the high strength weldment had an equivalent service exposure of only 22,229 hours, compared to much greater times for the lower-strength weldments.

TABLE 11

COMPARISON OF SAW WELD METAL AT 850 F WITH LITERATURE DATA

Data Source	Carbon Content	PWHT	Temper Parameter	Equivalent Time at 850 F (h)	100,000 Hour Rupture Strength (ksi)
Current Study	0.124	1300 F for 25 h	37.7	560,244,266	37.5
Emmanuel et al (116)	0.140	1125 F for 24 h	33.9	738,624	38.0
Emmanuel et al (116)	0.090	1075 F for 6 h	31.9	22,229	55.0

The influence of PWHT may also be illustrated by the data in Table 12 for SMAW weld metals of similar carbon content at 1050 F. The SMAW weld metal employed in this study received a PWHT at 1300 F for 25 hours. One of the SMAW welds investigated by Leyda et al. (83) received a PWHT at 1350 F for 12 hours. These 2 post weld heat treatments are characterized by similar temper parameters. These 2 welds have almost identical 100,000 hour rupture strengths. The other weld shown in Table 12 was post weld heat treated at a lower temperature resulting in a lower temper parameter and a higher stress-rupture strength. However, the likelihood again exists that this high rupture strength estimate is based on an unstable microstructure, especially when one considers the temper related equivalent service exposure (7432 hours at 1050 F). This equivalent service exposure is much less than 100,000 hours, for which the estimate was made. The other 2 welds have much longer equivalent service exposures (87,226 and 184,969 hours) and are much more likely to possess stable microstructures. Hence, it appears that long-time extrapolations of stress-rupture data from weldments receiving a mild temper must be viewed with suspicion.

The preceding paragraphs discussed the effects of PWHT on stress-rupture strength in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals. Another important factor is carbon content. Klueh and Canonico (67) have reported that the rupture strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals increases with increasing carbon content. The results of this study appear to confirm this conclusion. The SAW weld metal had the highest carbon content, 0.124%, and displayed the highest stress-rupture strength at all 3 test temperatures. Conversely, the SMAW weld metal which had the lowest carbon content (0.028%) exhibited

TABLE 12

COMPARISON OF SMAW WELD METAL AT 1050 F WITH LITERATURE DATA

Data Source	Carbon Content	PWHT	Temper Parameter	Equivalent Time at 850 F (h)	100,000 Hour Rupture Strength (ksi)
Current Study	0.028	1300 F for 25 h	37.7	87,226	6.6
Leyda et al (83)	0.036	1350 F for 12 h	38.2	184,969	6.57
Leyda et al (83)	0.036	1250 F for 12 h	36.0	7,432	7.31

the lowest stress-rupture strength. The ESW weld metal had an intermediate carbon content of 0.107% and correspondingly showed stress-rupture strengths which were intermediate as compared to the SAW and SMAW weld metals. It should be carefully noted, however, that the stress-rupture properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals do not depend solely on carbon content but are modified by welding process deposition characteristics and influenced very strongly by PWHT.

The low carbon content on the SMAW weldment can be attributed to the low carbon levels used in many SMAW electrodes of the $2\frac{1}{4}\text{Cr}-1\text{Mo}$ type. Low-alloy electrodes are characteristically kept low in carbon to avoid problems with weld metal cracking and poor notch toughness (137).

Creep Ductility

An important consideration in the high temperature behavior of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ is creep ductility. A loss of rupture ductility with increasing time to rupture is common to creep. Swift and Rogers (65) observed a decrease in ductility for long time tests on $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal. However, the weld metals involved in this study did not show a loss of ductility at long times, as evidenced by the ductility vs. rupture time plots presented in Figures 43-45. These plots show that rupture ductility is not a function of rupture time in the weld metals studied.

Table 13 lists the average rupture ductilities, in terms of elongation and reduction in area, for the weld metals at the 3 test temperatures. These ductility values are comparable to those obtained from other studies on $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal (32,67,116). In general, the weld metals exhibited a decrease in ductility with increasing carbon content. The high carbon (0.124%) SAW weld metal shows low ductility when compared to the SMAW and

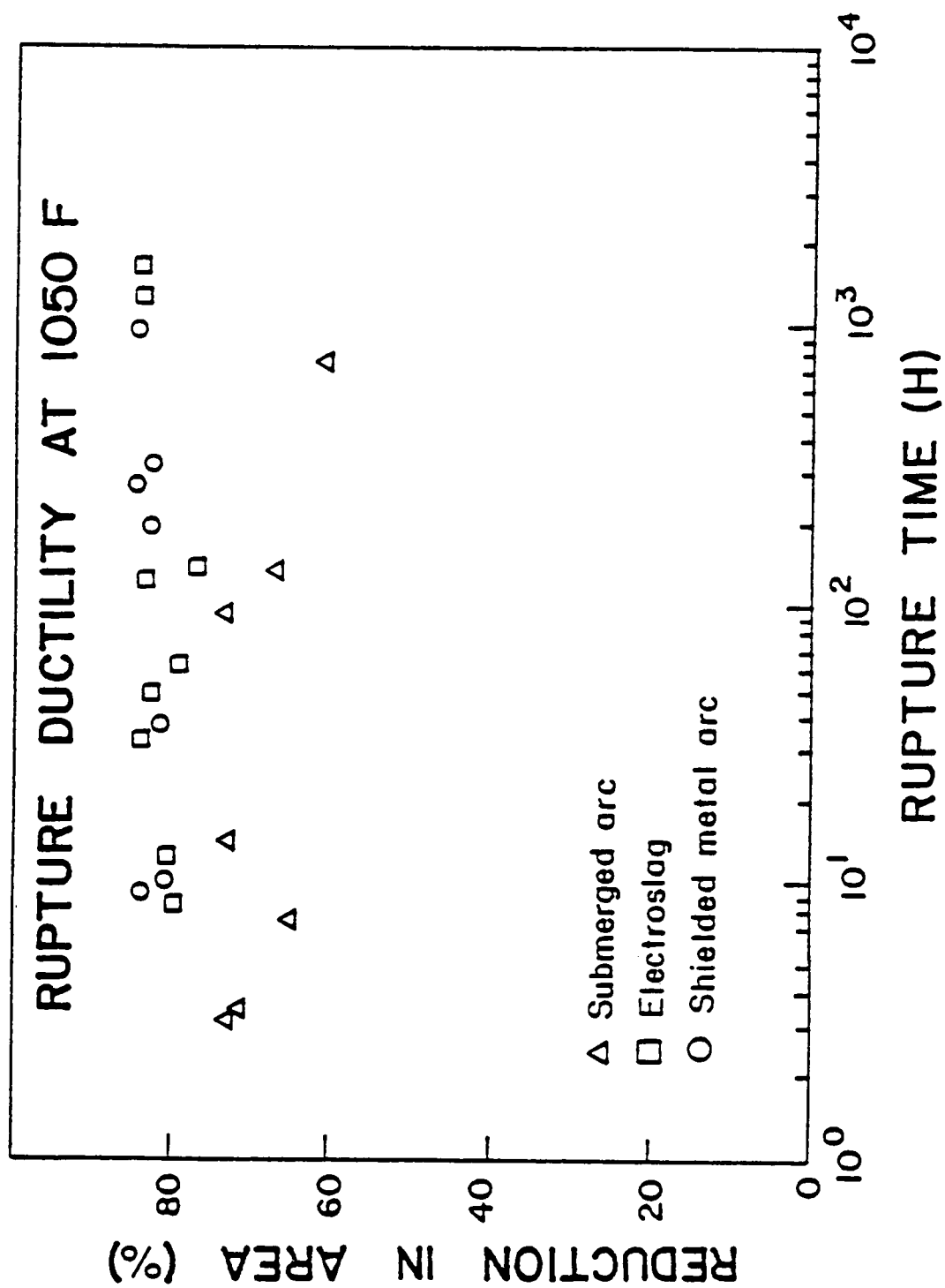


Figure 43. Rupture ductility of 2½Cr-1Mo weld metals at 1050 F.

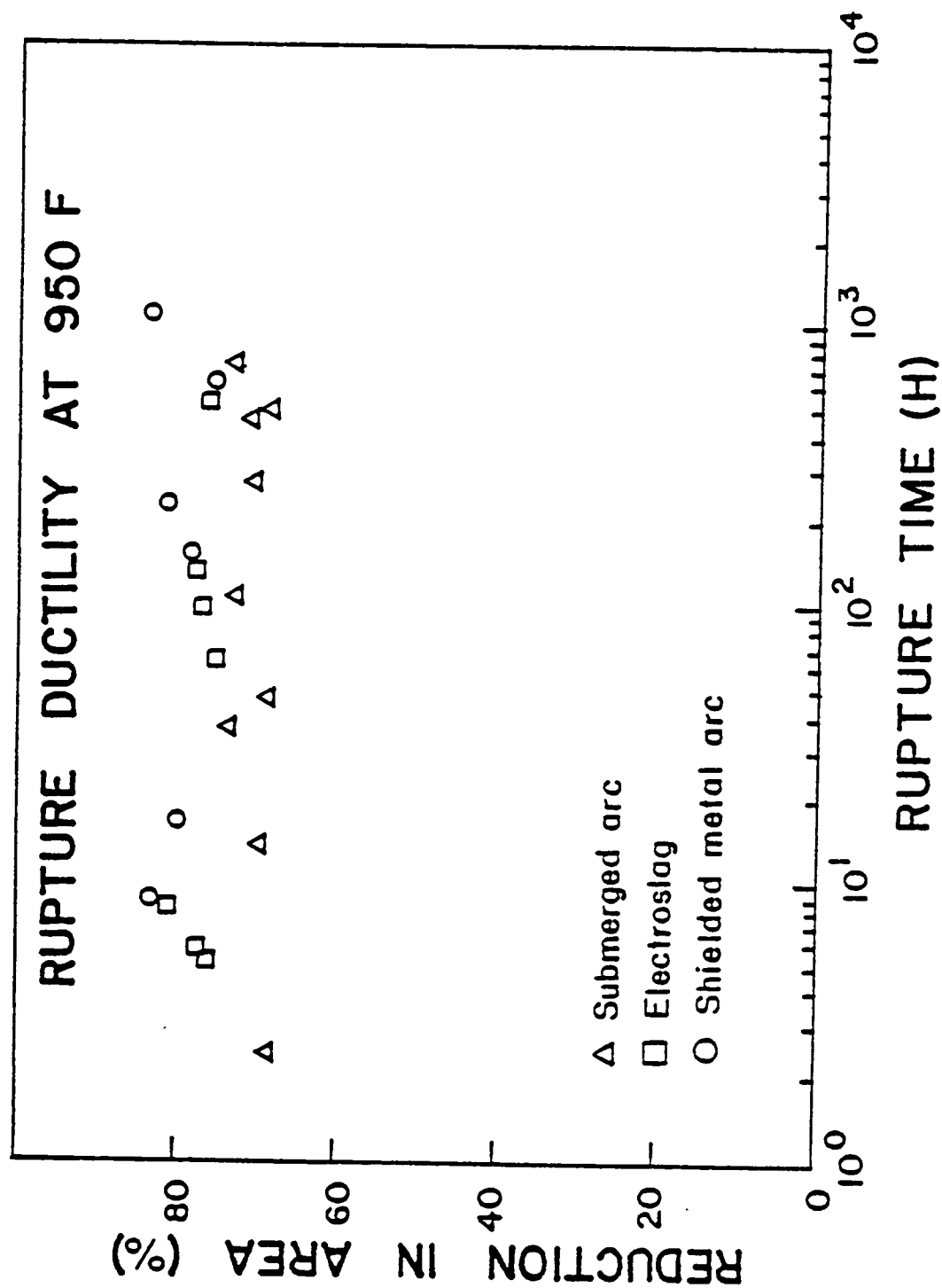


Figure 44. Rupture ductility of 2 $\frac{1}{4}$ Cr-1Mo weld metals at 950 F.

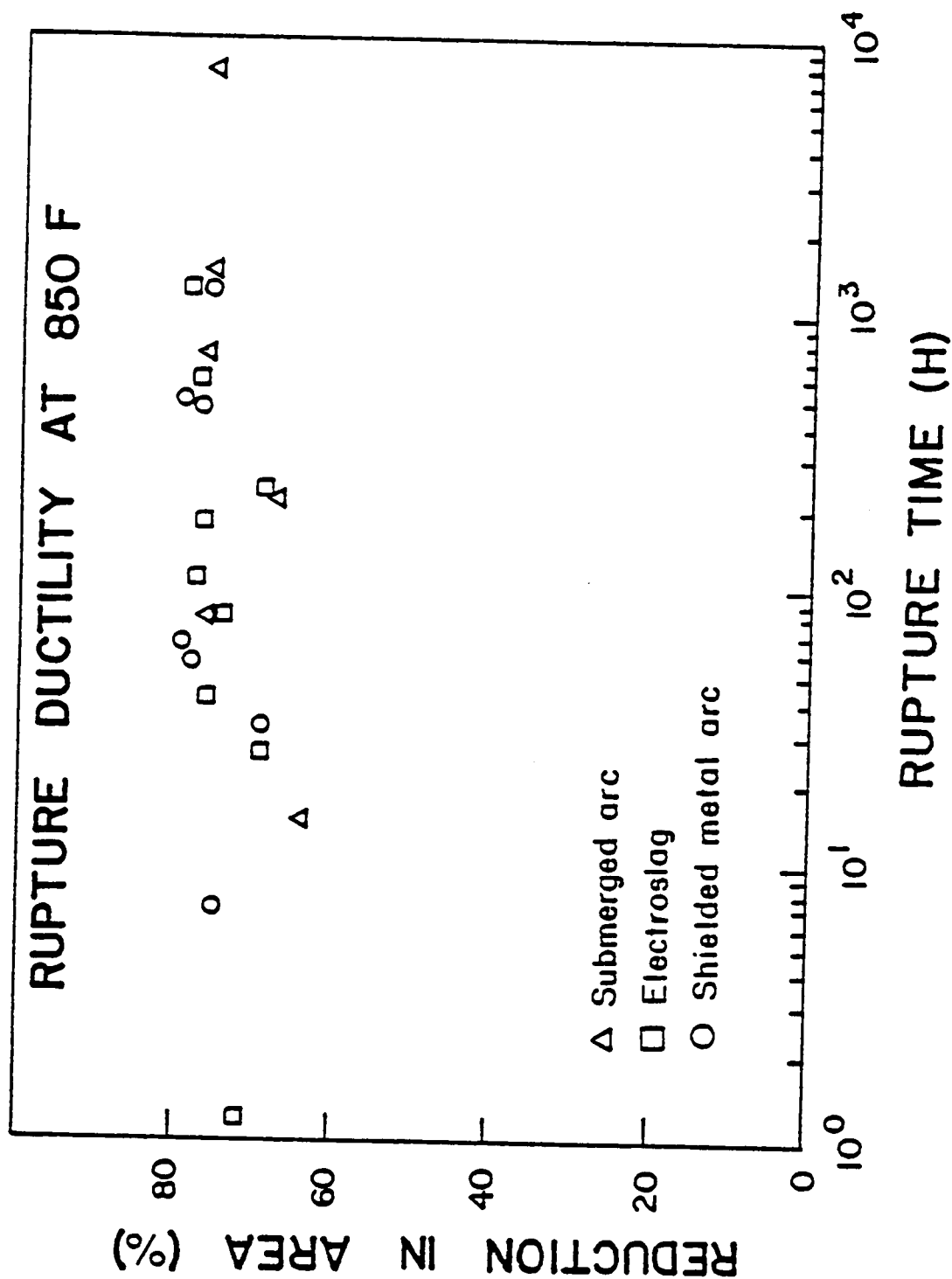


Figure 45. Rupture ductility of 2½Cr-1Mo weld metals at 850 F.

TABLE 13
AVERAGE RUPTURE DUCTILITY

	SMAW	ESW	SAW
% Elongation			
850 F	9.5	10.5	6.8
950 F	18.1	15.3	11.2
1050 F	18.0	17.0	11.8
% RIA			
850 F	76.1	75.3	72.5
950 F	79.6	77.2	70.8
1050 F	82.8	82.1	69.9

ESW weld metals, both of which have lower carbon contents. This low comparative ductility of the SAW weld metal becomes more pronounced with increasing temperature. Klueh and Canonico (67) stated similar findings in a study on $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals of varying carbon content. They found that high carbon (0.110%) $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal showed lower ductility than low carbon (0.003%) and medium carbon (0.035%) weld metal. The authors found that the difference in ductility between the high carbon weld metal and both the low and medium carbon weld metals increased with increasing temperature.

Results for Quenched and Tempered ESW Weld Metal

Because of current interest in the use of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal in the quenched and tempered condition, it was decided to test a portion of the ESW weldment after a quenching and tempering treatment. As described in Chapter III, the ESW material was reheated to 1650 F and then water spray quenched. The material was post weld heat treated at 1300 F for 25 hours. The quenched and tempered ESW weld metal was entirely bainitic, as seen in Figure 46.

Figures 47-49 present stress-rupture data for the ESW weldment in both the quenched and tempered and normalized and tempered conditions. At 850 F, the stress-rupture strength of both materials is nearly identical, as seen in Figure 47. Both materials lie above the minimum base metal reference curve from Code Case N-47. At 950 F, the quenched and tempered material has higher stress-rupture strength than the normalized and tempered weld metal, as seen in Figure 48. Both materials lie above the minimum base metal reference curve. Similarly, at 1050 F the quenched and tempered ESW weld metal has higher stress-rupture strength than the

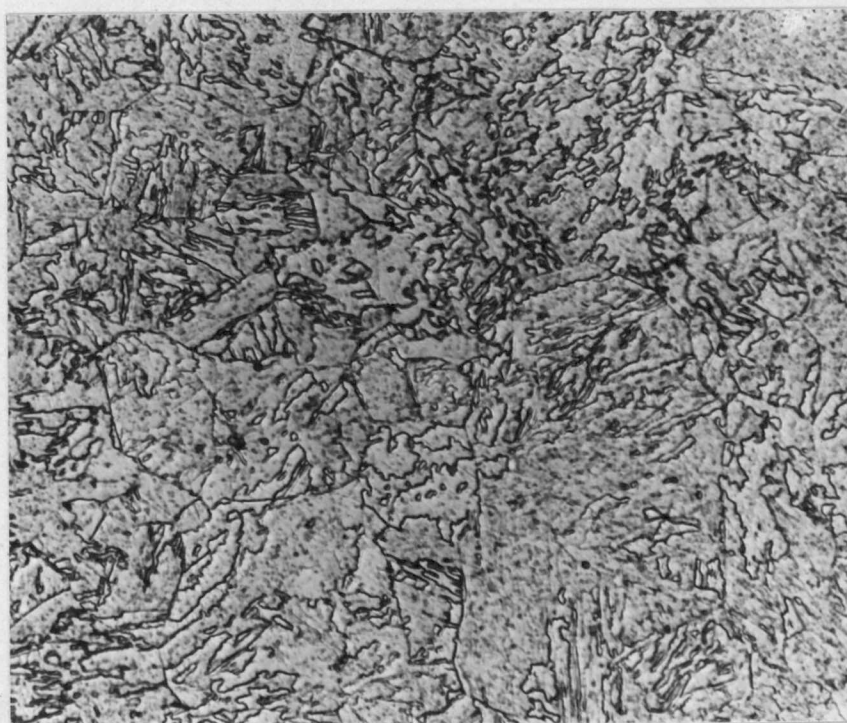


Figure 46. Light micrograph of quenched and tempered ESW weld metal in the as-PWHT condition showing bainitic structure. Polished and etched with 2% nital. 250X.

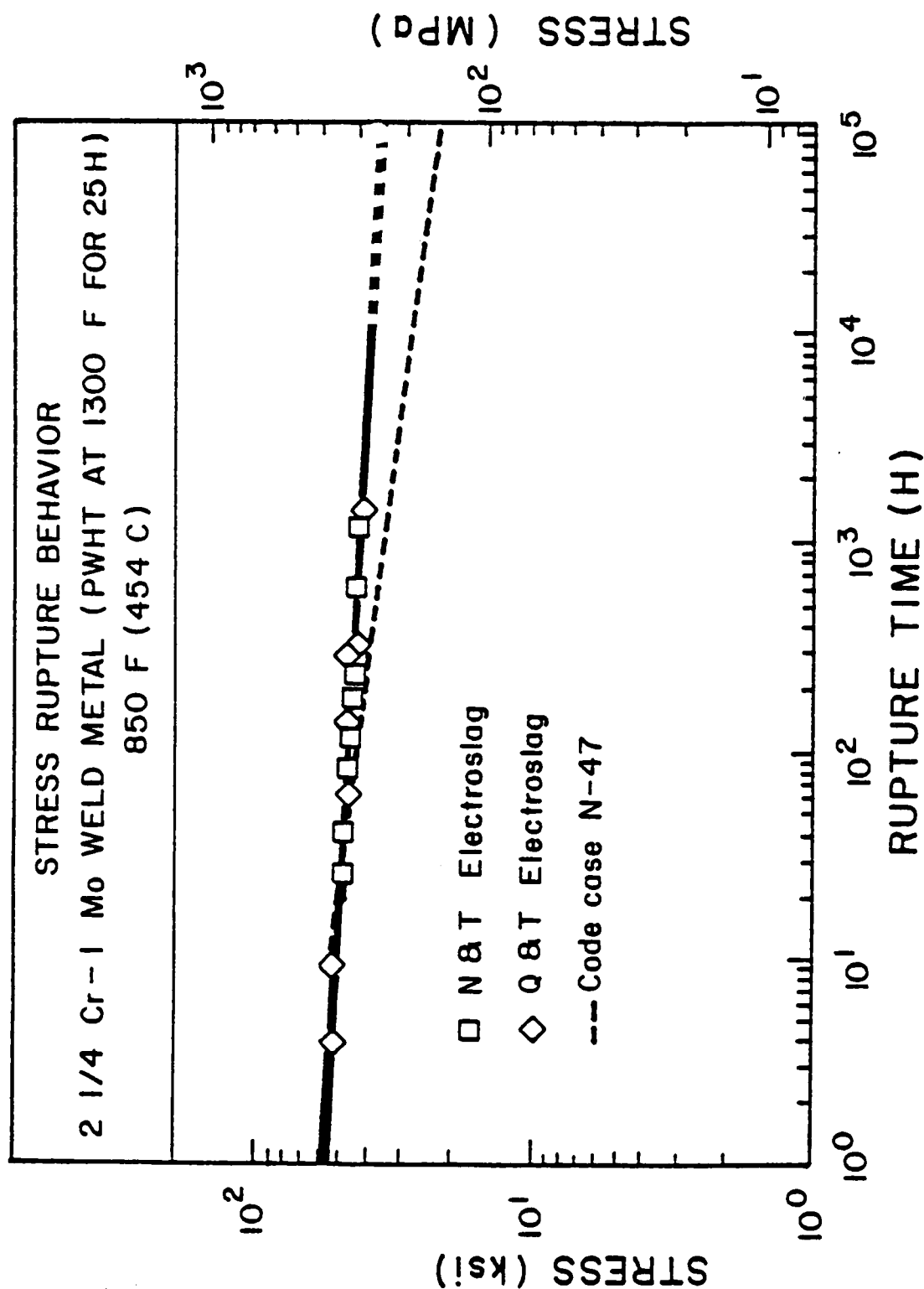


Figure 47. Stress-rupture curves for quenched and tempered and normalized and tempered 2 $\frac{1}{4}$ Cr-1Mo electroslag weld metals at 850 F.

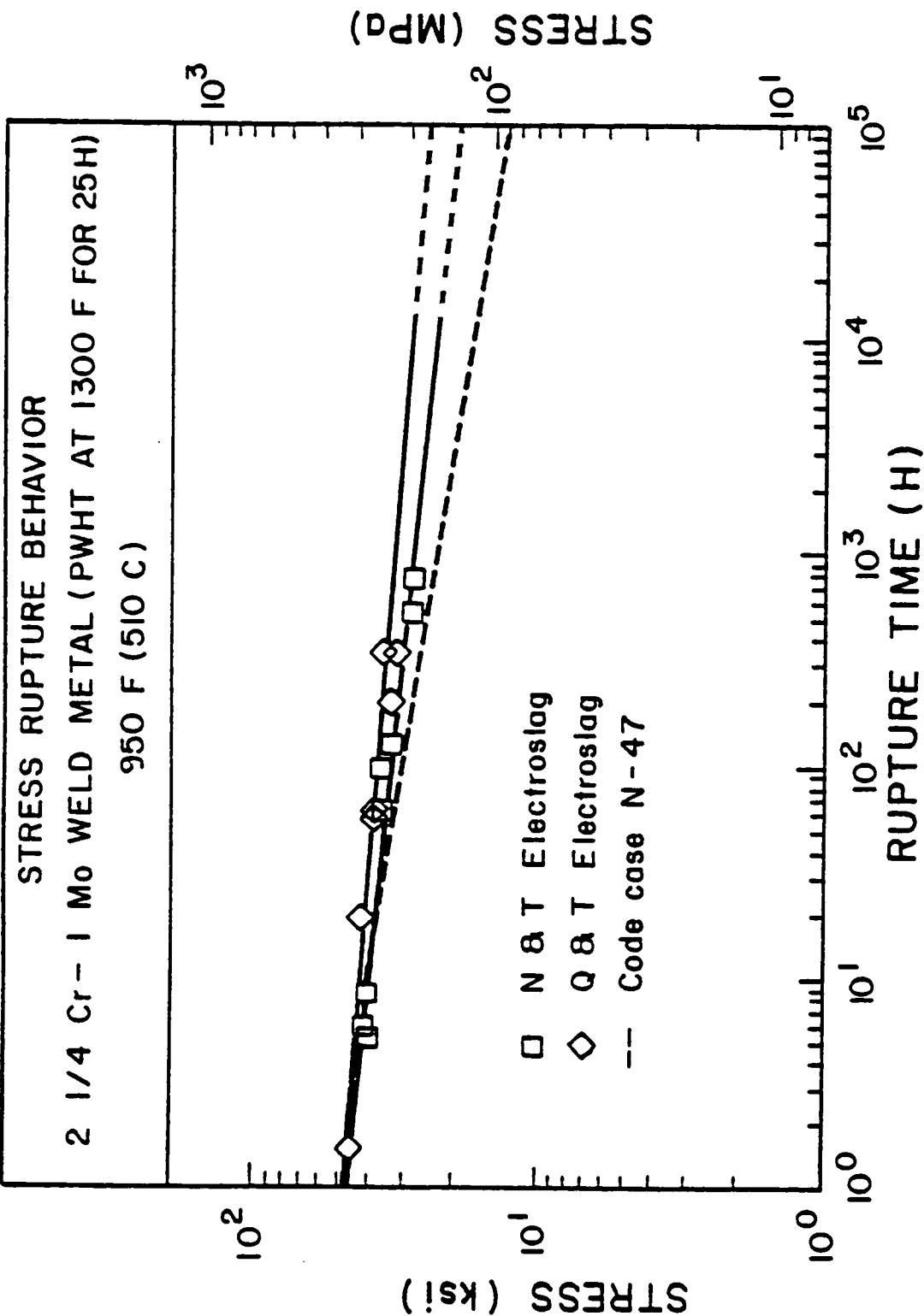


Figure 48. Stress-rupture curves for quenched and tempered and normalized and tempered 2¹/₄Cr-1Mo electroslag weld metals at 950 F.

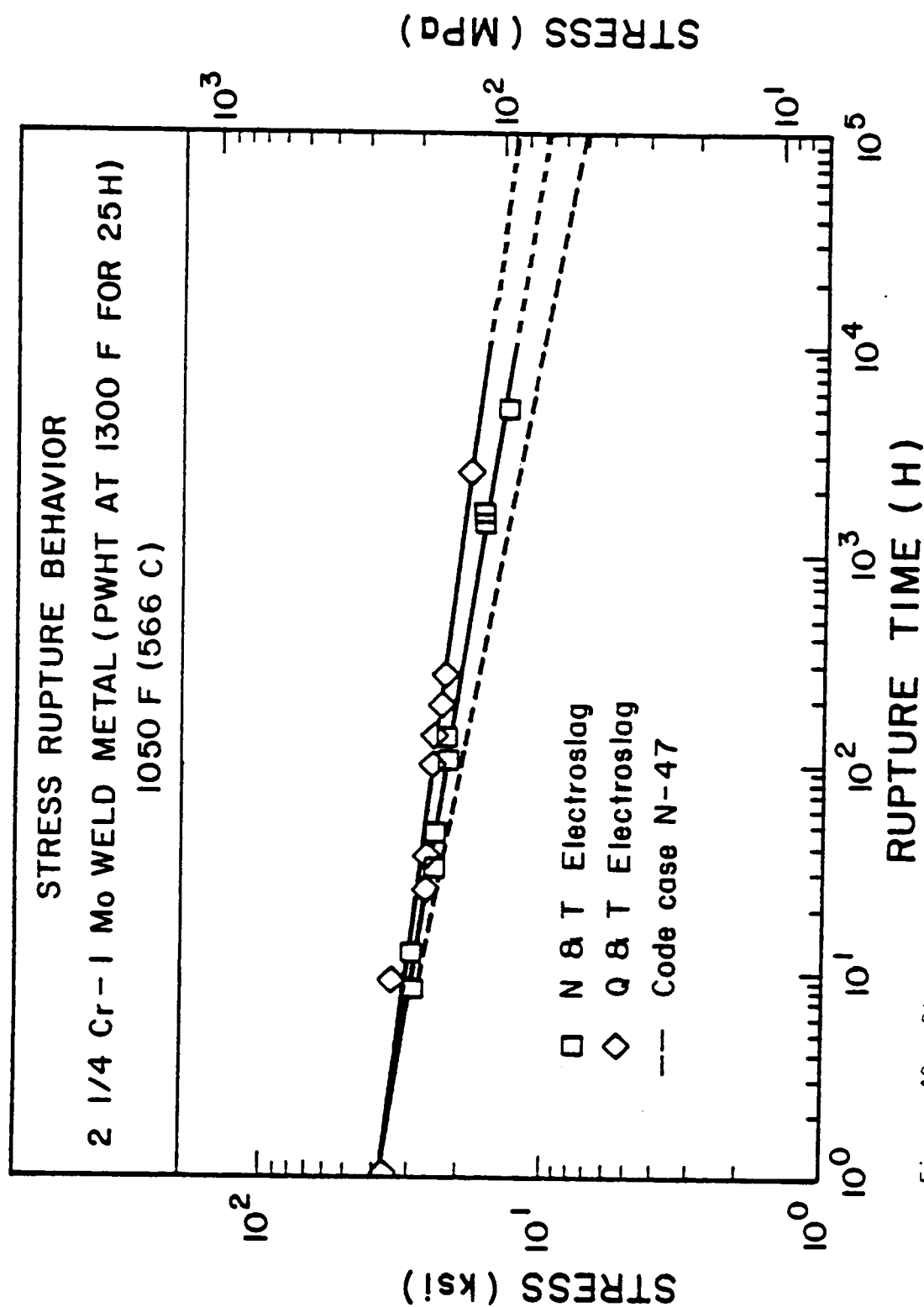


Figure 49. Stress-rupture curves for quenched and tempered and normalized and tempered 2 $\frac{1}{4}$ Cr-1Mo electroslag weld metals at 1050 F.

normalized and tempered weld metal, as shown in Figure 49. Again, both materials have rupture strengths which exceed minimum base metal values.

Table 14 lists estimated 100,000 hour rupture strengths for both the quenched and tempered and normalized and tempered ESW weld metal. The quenched and tempered material is superior to the normalized and tempered material at 1050 F and 950 F, but shows no advantage at 850 F. This finding is in contrast to some earlier studies. As mentioned in Chapter I, Brown, Rege and Spaeder (26) found that quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ was superior over normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ only at temperatures up to 900 F. Pense and Stout (25) reported that quenched and tempered material was superior to normalized and tempered material at temperatures not exceeding 950 F, while Robinson et al. (16) found that the strength advantage produced by quenching and tempering is maintained only up to 1000 F. Data from this study, however, shows that the quenched and tempered material has a distinct strength advantage over normalized and tempered material at 950 F and 1050 F.

Tension Test Results

Room temperature tension tests were performed on specimens taken from the weldments and base material. The results of these tests are shown in Table 15. The SAW weld metal shows the highest ultimate and yield strengths. Interestingly, the SMAW weld metal has higher room temperature strength than the ESW weld metal, even though the ESW weld metal has higher stress-rupture strength than the SMAW weld at all 3 test temperatures. Thus, it appears that room temperature mechanical properties do not correlate directly with elevated temperature stress-rupture behavior.

TABLE 14

ESTIMATED 100,000 HOUR RUPTURE STRENGTHS FOR
QUENCHED AND TEMPERED AND NORMALIZED AND TEMPERED
ELECTROSLAG WELD METAL

<u>100,000 HOUR RUPTURE STRENGTH (ksi)</u>		
	Q&T	N&T
1050 F	13.1	10.2
950 F	23.5	19.3
850 F	36.2	36.5

TABLE 15

ROOM TEMPERATURE TENSION TEST RESULTS FOR
ALL-WELD METAL AND BASE PLATE SPECIMENS

	SAW	ESW (N&T)	SMAW	BASE METAL
UTS	82.5 ksi	61.5 ksi	70.3 ksi	65.8 ksi
0.2% Yield Strength	65.4 ksi	33.8 ksi	41.8 ksi	26.9 ksi
% Elongation	23.8	40.3	33.5	31.8
% Reduction in Area	40.5	49.6	51.3	72.5

In light of this conclusion, elevated temperature tension tests were conducted at the 3 test temperatures. The results of these tests are presented in Table 16. The SAW weld metal has the highest ultimate and yield strengths at all three temperatures. The ESW weld metal is stronger than the SMAW weld metal at 1050 F and 950 F, but is weaker at 850 F. This corresponds directly to the observed stress-rupture behavior, especially when one notes that the SMAW material has higher stress-rupture strength than the ESW weld metal for short times at 850 F, but is weaker at 950 F and 1050 F. Thus, it appears that elevated temperature tension tests are useful in predicting elevated temperature behavior.

Closure

Over a hundred stress-rupture tests have been conducted on $2\frac{1}{4}\text{Cr-1Mo}$ weld metals deposited by various welding processes. These stress-rupture tests were supplemented by room temperature and elevated temperature tension tests, chemical analysis, and metallographic examination. The data produced by this study has provided a better understanding of the high temperature properties of $2\frac{1}{4}\text{Cr-1Mo}$ steel weld metal.

The rupture strength of $2\frac{1}{4}\text{Cr-1Mo}$ weld metal was shown to be a function of welding process characteristics, carbon content and PWHT. Creep ductility in $2\frac{1}{4}\text{Cr-1Mo}$ weld metal was found to be independent of rupture time, but dependent on carbon content. Specific conclusions are detailed in Chapter V.

The current study has shown the need for research in related areas. Future work in this program is discussed in Chapter VI.

TABLE 16

ELEVATED TEMPERATURE TENSION TEST RESULTS FOR
ALL-WELD METAL SPECIMENS

	UTS (ksi)	0.2% Yield Strength(ksi)	% Elongation	% RA
1050 F				
SAW	46.1	44.2	30.9	82.3
SMAW	32.7	25.3	53.0	88.2
ESW (N&T)	34.2	25.0	51.6	86.8
950 F				
SAW	55.8	50.9	22.6	71.3
SMAW	41.1	28.4	34.1	81.5
ESW (N&T)	43.6	25.8	35.0	77.9
850 F				
SAW	63.5	54.0	20.4	70.1
SMAW	53.0	35.5	28.3	74.9
ESW (N&T)	50.6	28.9	25.5	71.6

CHAPTER V

CONCLUSIONS

The study of the high temperature properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals has produced the first set of coherent data on $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals deposited by different welding processes. This study has yielded conclusions which are summarized as follows:

1. The stress-rupture strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals was shown to be a function of welding process characteristics.
2. The stress-rupture strength of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal was observed to increase with increasing carbon content, confirming an earlier study.
3. PWHT at temperatures significantly higher than the test temperature was found to have a stabilizing effect on the weld metal microstructure, confirming earlier work.
4. It is suggested that PWHT at low temperatures and short times does not produce a stable microstructure in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal and long-time extrapolations based on weld metals receiving such mild tempers may not be valid.
5. Quenched and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ electroslag weld metal was found to have superior stress-rupture strength at 1050 F and 950 F as compared to normalized and tempered electroslag weld metal of identical chemical composition.
6. The $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals studied did not show a loss in creep ductility with increasing time to rupture.

7. The ductility of 2½Cr-1Mo weld metal was observed to decrease with increasing carbon content.
8. Room temperature tension tests did not correlate directly with elevated temperature stress-rupture behavior; however, a good correlation with stress-rupture behavior was observed with regard to the elevated temperature tensile properties.

CHAPTER VI

FUTURE WORK

This initial study of the high temperature properties of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals has shown the need for expanded research in related areas.

Future efforts in this program will include:

1. Stress-rupture testing of transverse specimens extracted from the weldments employed in the current study. Some of these will be all-weld metal specimens, while others will be composite specimens containing weld metal, HAZ, and base metal in the gage length.
2. A study to determine the severity of PWHT (as defined by the Larson-Miller temper parameter) required to stabilize the microstructure in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals.
3. Fabrication and stress-rupture testing of higher carbon (0.06 and 0.10%) SMAW weldments, with variations in other alloying elements.
4. Fabrication and stress-rupture testing of a narrow-gap "hot wire" GTAW weldment.
5. Fabrication and stress-rupture testing of an electron beam (EB) weldment. EB welding is being considered for future utilization in heavy sections.
6. Extensive metallography of $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metals in the as-welded and as-PWHT conditions, and after exposure in the creep range.
7. Continued computer analysis of data on $2\frac{1}{4}\text{Cr}-1\text{Mo}$ weld metal, in order to place property predictions on a more firm footing.

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APPENDIX

TABLE 17
STRESS-RUPTURE TEST RESULTS AT 850 F

	Stress-(ksi)	Rupture Time (h)
SMAW Weld Metal		
	48.8	1.0
	48.8	6.8
	44.7	32.9
	44.7	36.0
	42.4	55.0
	42.4	64.8
	36.6	479.5
	36.6	505.1
	34.0	1104.8
	34.0	1240.0
SAW Weld Metal		
	56.5	8.5
	56.6	15.0
	50.2	82.1
	50.2	214.9
	45.5	739.5
	45.5	1351.2
	42.4	7832.5
ESW Weld Metal (N&T)		
	48.8	0.8
	48.0	26.2
	48.0	40.6
	46.7	83.4
	45.4	113.2
	43.9	253.5
	43.9	600.0
	42.4	1183.9
	42.4	1186.4
ESW Weld Metal (Q&T)		
	52.8	3.9
	52.8	9.1
	46.7	61.2
	46.7	136.8
	46.5	280.4
	43.6	326.7
	41.1	1415.4
	43.6	2027.3

TABLE 18
STRESS-RUPTURE TEST RESULTS AT 950 F

	Stress-(ksi)	Rupture Time (h)
SMAW Weld Metal		
	41.5	0.1
	33.7	8.8
	33.7	15.7
	25.6	151.7
	25.6	224.5
	22.0	614.9
	22.0	1029.4
SAW Weld Metal		
	49.6	2.5
	46.3	13.0
	46.3	26.4
	40.7	46.1
	37.8	108.7
	37.8	293.6
	33.3	462.1
	33.3	497.2
	32.5	728.9
ESW Weld Metal (N&T)		
	43.9	0.7
	40.7	5.8
	39.8	5.3
	39.8	8.4
	35.0	62.9
	35.0	99.3
	32.5	132.4
	27.6	556.4
	27.6	790.2
ESW Weld Metal (Q&T)		
	46.3	1.5
	42.3	5.8
	42.3	20.5
	37.9	60.8
	33.1	208.0
	35.0	345.0
	32.0	348.7

TABLE 19
STRESS-RUPTURE TEST RESULTS AT 1050 F

	Stress-(ksi)	Rupture Time (h)
SMAW Weld Metal		
	24.6	9.7
	24.6	10.0
	20.3	37.7
	20.3	61.4
	16.3	182.3
	16.3	289.5
	12.6	946.9
	10.2	4237.4
SAW Weld Metal		
	37.8	3.2
	37.8	3.5
	34.1	8.3
	34.1	14.2
	26.5	98.7
	26.5	132.9
	21.1	754.4
	21.1	867.2
	17.1	2940.7
ESW Weld Metal (N&T)		
	28.9	8.5
	28.9	12.4
	24.4	33.0
	24.4	47.7
	22.0	103.9
	22.0	136.4
	16.3	1462.3
	16.3	1627.1
	13.8	5184.9
ESW Weld Metal (Q&T)		
	36.2	1.1
	34.1	9.4
	34.1	9.4
	26.5	25.0
	26.5	36.5
	24.6	101.9
	24.6	136.7
	22.8	190.6
	22.8	270.2

VITA

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