

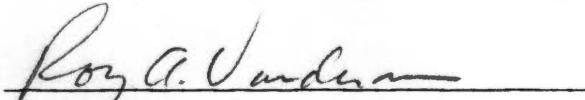
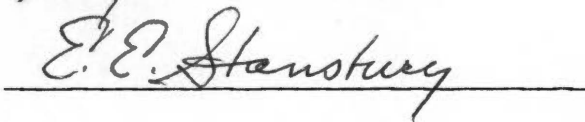
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Carl D. Lundin, Major Professor

We have read this thesis
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The Graduate School

THE WELDABILITY OF Cr-Mo STEELS FOR
FOSSIL ENERGY APPLICATIONS

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

Mark W. Richey

June 1984

DEDICATION

This thesis is dedicated to my family

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ABSTRACT

This research program was undertaken to provide fundamental and basic metallurgical information on the behavior of the heat-affect zone (HAZ) in Cr-Mo steel welds, as well as practical information on their relative weldability. The principal work was the evaluation of the stress-relief cracking (SRC) and hydrogen-assisted cracking (HAC) susceptibility of Cr-Mo steels ranging in Cr content from 2 1/4%-12%. Differences in observed cracking behavior were contrasted with differences in composition, on-cooling transformation behavior and weld microstructure.

The test methods were varied to investigate the effect of preheat, post-weld heat treatment (PWHT), metallurgical structure and restraint (externally applied) on the weld cracking susceptibility. Stress-rupture testing of the Gleeble simulated HAZ was employed in the SRC evaluation while the UT-Modified Hydrogen Sensitivity Test (UT-Mod HST) was employed in the HAC determination. In addition, the Gleeble apparatus was employed to determine the continuous-cooling transformation (CCT) behavior of the coarse grained heat-affected zone. (This region of the weld is known to be the region most susceptible to SRC and HAC.) The samples generated during testing provided test data as well as metallographic specimens for examination by optical and electron optic techniques (SEM). Evaluation of the metallographic specimens defined the effect of microstructure on the weld cracking susceptibility.

The basic metallurgical information resulting from this investigation made clear the reasons for the relative cracking behavior of different Cr-Mo materials. The practical data from this investigation provides a basis for the determination of suitable welding conditions as well as a guide for the selection of materials from the standpoint of fabricability.

The literature review provides a concise historical review and basis of theories for SRC and HAC in Cr-Mo steels which were employed to explain the weld cracking susceptibility of various Cr-Mo steels. It is apparent that this weldability investigation has answered many questions neglected by prior investigations.

The results of the CCT determination for the coarse grained HAZ in 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni, 12Cr-2Mo-1Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo show that martensite or martensite plus delta-ferrite is the predominate microstructure in the coarse grained HAZ. The microstructure of these materials is independent of cooling rates normally attendant upon welding. The M_s and M_f temperatures in the 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo are higher than the M_s and M_f temperatures determined for the 12Cr-2Mo-Ni materials. The lack of a detectable M_f temperature in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni suggests that austenite may be present in the HAZ of these two materials.

Results of the CCT behavior determination for the coarse grained HAZ of 3Cr-1.5Mo and 2 1/4Cr-1Mo show that the on-cooling microstructure is dependent on the cooling rate. Over the range of cooling

rates associated with commonly employed energy inputs the following types of HAZ microstructures may be observed: (1) for a fast cooling rate the microstructure is acicular martensite and lenticular martensite; (2) the microstructure at a slower cooling rate consists of lenticular martensite and bainite; (3) a further reduction in the cooling rate produces a microstructure which contains bainite or bainite plus ferrite. The 3Cr-1.5Mo HAZ is more likely to contain martensite than the 2 1/4Cr-1Mo HAZ because of its higher alloy content.

The SRC susceptibility of the Cr-Mo steels employed in this investigation is related to the extent of grain matrix strengthening which occurs during PWHT. An increase in grain matrix strength increases the susceptibility to SRC. During PWHT the grain matrix of 2 1/4Cr-1Mo and 3Cr-1.5Mo is strengthened more than that of 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo and as a result, the 3Cr-1.5Mo and 2 1/4Cr-1Mo is more susceptible to SRC. During PWHT at 1150°F (620°C) the grain matrix of the 3Cr-1.5Mo and 2 1/4Cr-1Mo is strengthened more than at 1250°F (680°C) PWHT. This increase in grain matrix strength at the lower PWHT temperature results in an increased SRC susceptibility. Apparently the amount of Cr present in the 3Cr-1.5Mo imparts improved ductility to the precipitate strengthened matrix, and as a result, this material is less susceptible to SRC than 2 1/4Cr-1Mo. An increase in the C or Si content in the nominal 2 1/4Cr-1Mo increases the likelihood of strengthening the matrix, thus increasing the SRC susceptibility.

The HAC susceptibility of the Cr-Mo steels is dependent on the microstructure, carbon content, on-cooling transformation behavior and to a certain degree on the level of restraint. The 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni is more susceptible to HAC than the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo because the austenite transforms to martensite at a lower temperature. The diffusivity of hydrogen in martensite is reduced at the lower temperature of transformation, thus hydrogen is less able to diffuse out of the HAZ, thereby increasing the cracking susceptibility. Among the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials, the HAC susceptibility is dependent on carbon content and delta-ferrite content. In general, an increase in carbon content results in an increase in HAC susceptibility. The 9Cr-2Mo is more susceptible to HAC than the 9Cr-1Mo material of a comparable carbon content because the increased amount of delta-ferrite present in the 9Cr-2Mo partitions carbon to the martensite, increasing both hardness and HAC susceptibility. For high levels of simulated restraint (2% and 4% augmented strain) the HAC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo is approximately the same as 9Cr-2Mo and 9Cr-1Mo, while at a lower simulated restraint level (1% augmented strain), the situation is reversed and the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials are more susceptible to HAC. For an equivalent carbon content (0.11% C), the bainitic 2 1/4 Cr-1Mo is less susceptible to HAC than the martensitic 9Cr-1Mo. The 3Cr-1.5Mo is more susceptible to HAC than the 2 1/4Cr-1Mo because martensite is more prevalent in the 3Cr-1.5Mo HAZ.

In general, the relative SRC and HAC susceptibility of Cr-Mo steels can be categorized in terms of subclasses of Cr-Mo steels. With respect to SRC, the 12Cr-2Mo-Ni, 9Cr-2Mo and 9Cr-1Mo materials are not as susceptible to SRC as the 3Cr-1.5Mo and 2 1/4Cr-1Mo. With respect to HAC, the 12Cr-2Mo-Ni alloys are more susceptible to HAC than the 9Cr-2Mo and 9Cr-1Mo. For an equivalent carbon level the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials are less susceptible to HAC than the 9Cr-2Mo and 9Cr-1Mo.

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NOMENCLATURE

Cr-Mo	chromium-molybdenum
HAZ	heat-affected zone
SRC	stress-relief cracking
HAC	hydrogen-assisted cracking
PWHT	post-weld heat treatment
UT-Mod HST	UT-Modified Hydrogen Sensitivity Test
CCT	continuous-cooling transformation
SEM	scanning electron microscope
M_s	martensite start
M_f	martensite finish
ASTM	American Society for Testing Materials
GTA	gas-tungsten arc
Cr.E.	chromium equivalent
MA	martensite/austenite
AR&TD	Advanced Research and Technology Development
ORNL	Oak Ridge National Laboratory
ESW	electroslag weld
AE	acoustic emission
Ni.E.	nickel equivalent
OLM	optical light microscope
DPH	diamond pyramid hardness
B_s	bainite start
B_f	bainite finish

8/5 t	time to cool from 800°C to 500°C
SRP	stress rupture parameter
RA	reduction in area
SRI	stress rupture index
EDAX	energy dispersive x-ray analysis
RPI	Rensselaer Polytechnic Institute

CHAPTER I

INTRODUCTION

Chromium-molybdenum (Cr-Mo) steels are widely used in the petroleum, power and process industries combining oxidation resistances with high elevated temperature strength and adequate toughness (1). The most commonly employed steels vary in chromium content from 0.5% to 12% and all contain some molybdenum with 1% being the normal nominal composition. ASTM specifications for Cr-Mo plate are shown in Table 1. The corrosion resistance and high temperature strength of these steels is a function of their chromium and molybdenum contents. The lower chromium materials (2 1/4Cr-1Mo) exhibit the best elevated temperature strength while the higher chromium materials (12Cr-1Mo) exhibit the best oxidation resistance. The hardenability of the chromium-molybdenum steels ranges from fully air hardenable to intermediate hardenability levels and thus, the gamut of metallurgical reactions must be considered depending on the alloy content (2). These materials may be supplied in the annealed, normalized and tempered or quenched and tempered conditions. The mechanical properties depend upon the particular heat treatment.

From a designer's point of view the selection of a chromium-molybdenum steel for a particular elevated temperature application is based on the following requirements (3).

1. Good fabricability resulting in a degree of soundness which easily meets code requirements;

Table 1. ASTM Specifications for Cr-Mo Steel Products.

Type	Forgings	Tubes	Pipe	Castings	Plate
1/2Cr-1/2Mo	A182-F2	A213-T2	A335-P2 A369-FP2 A426-CP2		A387-Gr2
1Cr-1/2Mo	A182-F12 A336-F12	A213-T12	A335-P12 A369-FP12 A426-CP12		A387-Gr12
1-1/4Cr-1/2Mo	A182-F11 A336-F11/F11A A541-C15	A199-T11 A200-T11 A213-T11	A335-P11 A369-FP11 A426-CP11	A217-WC6 A356-Gr6 A389-C23	A387-GR11
2Cr-1/2Mo		A199-T3b A200-T3b A213-T3b	A369-FP3b		
2-1/4Cr-1Mo	A182-F22/F22a A336-F22/F22A A541-C16/6A	A199-T22 A200-T22 A213-T22	A335-P22 A369-FP22 A426-CP22	A217-WC9 A356-Gr10 A643-GrC	A387-Gr22 A542
3Cr-1Mo	A182-F21 A336-F21/F21A	A199-T21 A200-T21 A213-T21	A335-P21 A369-FP21 A426-CP21		A387-Gr21
5Cr-1/2Mo	A182-F5/F5a A336-F5/F5A A473-501/502	A199-T5 A200-T5 A213-T5	A335-P5 A369-FP5 A426-CP5	A217-C5	A387-Gr5
5Cr-1/2MoSi		A213-T5b	A335-P5b A426-CP5b		
5Cr-1/2MoTi		A213-T5c	A335-P5c		
7Cr-1/2Mo	A182-F7 A473-501A	A199-T7 A200-T7 A213-T7	A335-P7 A369-FP7 A426-CP7		A387-Gr7
9Cr-1Mo	A182-F9 A336-F9 A473-501B	A199-T9 A200-T9 A213-T9	A335-P9 A369-FP9 A426-CP9	A217-C12	A387-Gr9

Source: "Chromium-Molybdenum Steels," Welding Handbook, 7th Edition, Vol. 4, Chap. 1, American Welding Society, (1983).

2. Adequate mechanical properties for the operating environment under design conditions;
3. Retention of desired properties for service under the design conditions and environment;
4. Freedom from property changes which inhibit operational transients and shut down conditions.

All too often structures are designed to meet the requirements based on function, steel availability and economic considerations, without regard to metallurgical characteristics. The main problem with this design approach is that the weld considerations (fabricability) are an afterthought or are taken for granted as trouble-free operations which do not require any particular attention. Recently, the lack of emphasis on the fabrication of the component and total structure has been recognized and is now being seriously treated.

In the past few years a new generation of Cr-Mo steels has been developed to (1) enhance resistance to hydrogen attack (3Cr-1.5Mo), (2) produce a concurrent increase in both creep strength and weldability (9Cr-2Mo and 12Cr-2Mo-Ni), and (3) improve creep strength to a level comparable to type 304 stainless steel (9Cr-1Mo-V-Nb). The evaluation of the fabricability of these new types of Cr-Mo steels would logically include a comparison with traditional types of materials of similar compositions, however review of the currently available literature on the fabricability (in particular weldability) of traditional types of Cr-Mo steels reveals that little work exists

on the weldability of these steels. It is apparent that if a new generation of Cr-Mo materials is to be employed for weld fabrication, additional investigations into the weldability of both the new generation and traditional type materials are required.

The weldability evaluation was initiated because of the need for an additional study on the weldability of Cr-Mo steels. The object of this work is to determine the relative weldability of both traditional types and newer generation Cr-Mo steels with respect to stress-relief cracking and hydrogen-assisted cracking. Particular attention has been placed on the effect of welding parameters (especially preheat), composition, weld microstructure and PWHT temperature.

CHAPTER II

LITERATURE REVIEW

Weldability

The ability of a material or combination of materials to be welded under fabrication conditions into a specific suitably designed structure and to perform satisfactorily in intended service is defined as its weldability (4). The concept of weldability requires the careful consideration of a large number of variables including the composition and mechanical properties of the base metal, welding consumables, energy input, welding sequence, joint design and structure constraint (5). One of the most common considerations assessing the weldability of a material is whether or not it will be free from cracking under imposed design and fabrication conditions. Coe (6) has drawn a clear distinction between cracking which occurs during service and that which takes place during or shortly after the welding operations. The former is not dealt with in this review but the latter, which may be divided into five categories, is the prime focus in this work:

1. Solidification cracking (weld metal hot cracking);
2. Liquation cracking (HAZ hot tearing);
3. Lamellar tearing;
4. Hydrogen induced cracking (HAZ delayed cold cracking);
5. Reheat cracking (heat treatment or stress-relief cracking).

Solidification cracking occurs upon solidification when solute redistribution occurs and both impurities and intentional alloy elements are segregated in concert with the development of solidification substructure. The segregation of these elements lowers the effective liquidus temperature of the material in the substructure and grain boundaries and may result in the formation of an intergranular or interdendritic film of liquid. These films may not be capable of supporting contractional stresses induced by thermal contractions and a crack may form.

Liquation cracking occurs when partial melting (liquation) takes place near the fusion boundary of a weld. This liquated region is unable to accommodate contractural stresses associated with the weld thermal cycle and a crack develops.

Lamellar tearing occurs in the base metal of a weldment and is the result of low base plate ductility combined with high localized stress. This form of cracking is associated with restrained corner or "tee" joints, particularly in thick section plate where the fusion boundary of the weld is more or less parallel to the plate surface.

The postulated mechanisms responsible for the first three types of cracking are well accepted and have been confirmed by direct observation. In the case of hydrogen induced cracking (hereto referred to as hydrogen-assisted cracking (HAC))* and stress-relief

*The term "hydrogen-assisted cracking" is intended to apply to all forms of cold cracking, including underbead cracking, toe cracking, delayed cracking, etc.

cracking, no single unifying theory has been postulated to explain the presence of cracks under various conditions. Because hydrogen-assisted cracking and stress-relief cracking is currently not or may well never be understood in a fundamental manner, this review is aimed towards examining these two types of cracking on a practical welding basis with emphasis on governing factors and necessary precautions.

Hydrogen-Assisted Cracking (HAC)

Hydrogen-assisted cracking is not confined to welding but may occur in steel during manufacturing, fabrication or in service. When cracking occurs as the result of welding, the cracks are situated either in the HAZ of the parent metal or in the weld metal itself. Cracks in the HAZ are usually situated at the weld toe, the weld root or in the underbead position as shown schematically in Figure 1 for fillet welds and butt welds. Usually in the fillet welds the HAZ cracks are oriented along the weld length, but in butt welds subsurface cracks can be transverse to the weld. Intergranular, quasi-cleavage or microvoid coalescence fracture modes have all been observed in hydrogen cracks. Beachem (7) has shown that the operating fracture mode depends upon microstructure, the crack-tip stress intensity and the concentration of hydrogen.

As outlined by Coe (8), HAC occurs when the following four conditions occur simultaneously:

1. Hydrogen is present;
2. Tensile stress acting on the weld;

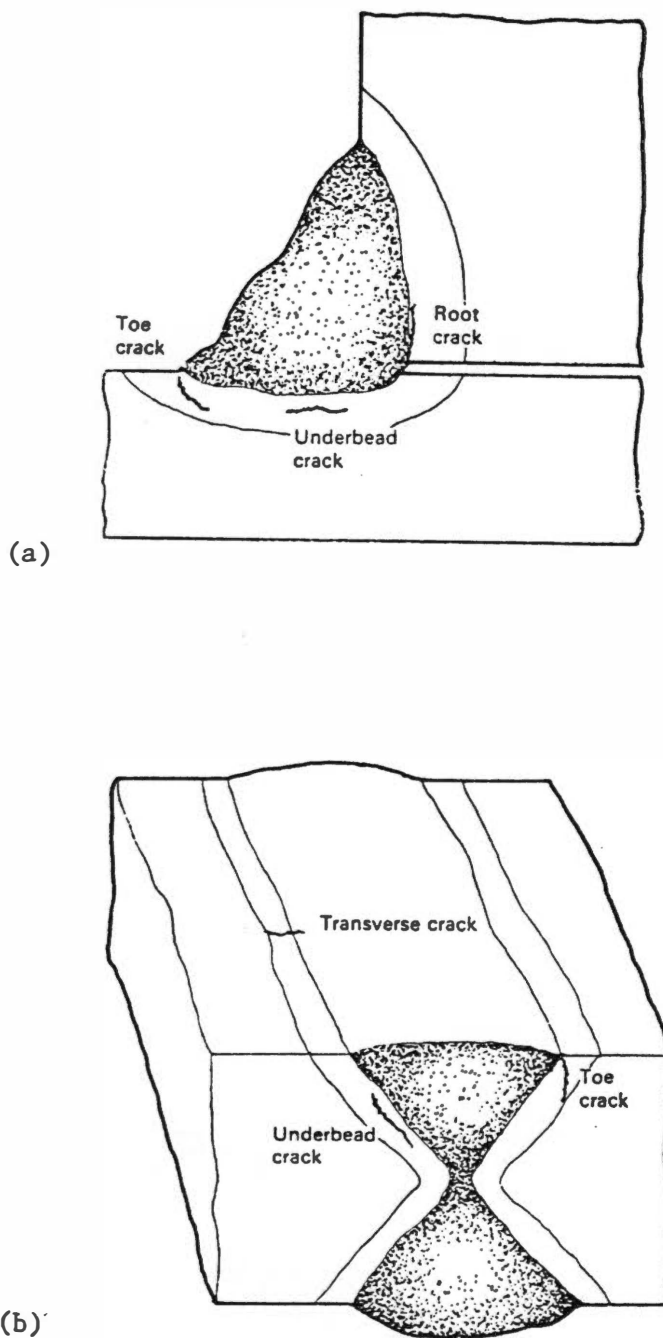


Figure 1. Heat-affected zone cracks. (a) Fillet weld, (b) Butt weld.

Source: Coe, F.R., Welding Steels Without Cracking, The Welding Institute, (1973).

3. A susceptible HAZ microstructure is present;
4. A low temperature is reached.

Hydrogen is absorbed by the weld metal from the arc during welding. Although much of the hydrogen may escape during cooling, some may diffuse into the HAZ and the parent metal. The amount diffused depends on the initial amount absorbed, size of the weld and decreasing solubility (Figure 2). Hydrogen is dissociated into atomic hydrogen (H^+) in the weld arc and is dissolved in the metal as an interstitial constituent or as recombined hydrogen gas located in traps. The interstitial solubility of hydrogen in molten steel is governed by Sievert's Law (9):

$$\%[H] = K\sqrt{P_{H_2}} \quad (2-1)$$

where $\%[H]$ is the percent of the hydrogen dissolved in the molten steel, K is a constant and P_{H_2} is the partial pressure of hydrogen in the atmosphere above the molten steel. In steel, hydrogen may also be trapped at grain boundaries, carbides, non-metallic inclusions, dislocations and microvoids. According to Lancaster (10), the total concentration of hydrogen in the traps may be up to 10^4 times the interstitial value. The portion of hydrogen which diffuses out of the weld at room temperature is known as diffusible hydrogen and is expressed as ml/100g of deposited metal. Typical weld hydrogen levels range from 30ml/100g of deposited metal in shielded metal-arc covered electrode welds to 2ml/100g of deposited metal in gas shielded metal-arc welds.

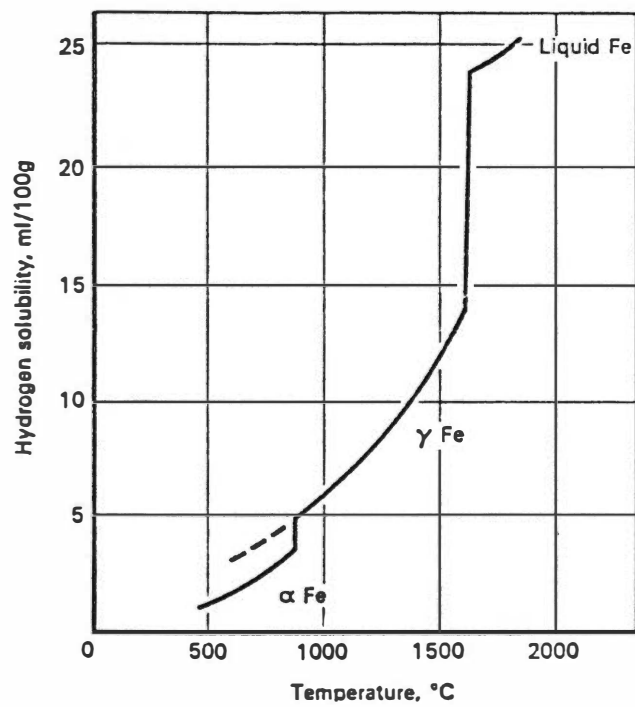


Figure 2. Solubility of hydrogen in weld metal.

Source: Coe, F. R., Welding Steels Without Cracking, The Welding Institute (1973).

Hydrogen is introduced into the arc atmosphere by hydrogen present in the consumables or hydrogen present on the surface of materials to be welded. Coe (8) described the principal sources of hydrogen in welding consumables as:

1. Moisture in electrode coatings or in the fluxes;
2. Any hydrogenous compound in the coating or flux;
3. Oil, dirt and grease on the surface or trapped in welding wires and electrode wires.

Coe also described the principal sources of hydrogen from the material to be welded as:

1. Oil, grease, dirt and paint on the weld surface;
2. Degreasing liquids employed to clean welding surfaces.

As the weld cools, stresses are developed as a result of thermal contraction and these stresses must be accommodated as strain in the weld. The stresses tend to be concentrated at the toe and root of the weld and at inclusions or any other defect which acts as a notch. Hydrogen appears to lower the stress level at which cracking at the notch occurs.

The region of the HAZ adjacent to the fusion zone experiences a temperature sufficient to produce a coarsened grain size which is more hardenable and less ductile than regions further away from the fusion boundary. The greatest risk of cracking is therefore in the coarsened region.

As the temperature of the weld decreases after welding the rate of diffusion of hydrogen decreases, thus reducing the amount of

hydrogen removed from the weld. As shown in Figure 3, this effect is particularly marked in the range of 68 to 300°F (20 to 150°C). Pre-heat slows down the cooling rate and allows more time for the escape of hydrogen.

Theory of Hydrogen-Assisted Cracking

Although much research, time and effort has been devoted to determine the exact cause and nature of hydrogen-assisted cracking in both welding and other metallurgical fields, there is presently no one unifying theory which explains precisely how the presence of hydrogen within metal structure leads to the formation of cracks under some circumstances and not in others. According to Bernstein (11) all the proposed theories of the role of hydrogen fall into one or more of four general categories:

1. Pressure;
2. Surface adsorption;
3. Binding energy;
4. Dislocation mobility modification.

The oldest, and until recently, the most popular of these theories, has been the pressure theory which was first proposed by Zapffe and Sims (12). According to this theory molecular hydrogen precipitates into pre-existing mosaic rifts in the lattice resulting in high internal pressures and subsequent failure along mosaic boundaries.

In the surface adsorption theory first proposed by Petch and Sables (13), hydrogen upon precipitation from the lattice is adsorbed

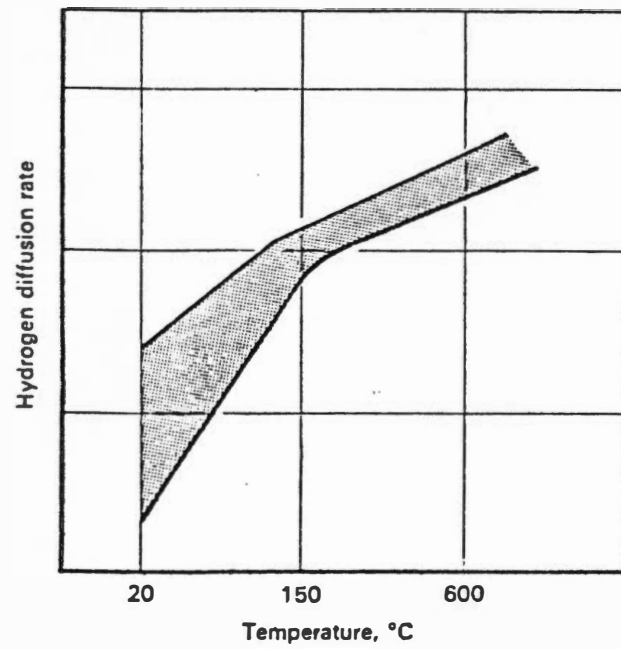


Figure 3. Diffusion rate of hydrogen through ferritic steel.

Source: Coe, F.R., Welding Steels Without Cracking, The Welding Institute, (1973).

on the surface of the internal lattice imperfections and microcracks. According to this theory the surface energy at the crack nucleus would be reduced and a lower stress could extend the submicroscopic rift into a major catastrophic crack.

The pressure theory and the surface adsorption theory were advanced before the exact nature of hydrogen cracking was recognized. Both theories were inadequate in explaining many observed details of delayed brittle fracture. For example, in some cases the effects of hydrogen in steel can be completely removed by baking (14). If the embrittlement of the steel by hydrogen was the result of hydrogen trapped in the microscopic voids, it is difficult to explain how at low baking temperatures this mechanism could operate.

Troiano (15) proposed a cracking theory based on the stress induced diffusion of hydrogen in an effort to more fully explain observed cracking behavior. In this theory, hydrogen distributed in a matrix will diffuse to the area of highest triaxial stress just ahead of a crack tip. When sufficient hydrogen is concentrated in this region a crack will initiate. The hydrogen is thought to reduce the cohesive strength of the lattice by interacting with the electron structure in a way as to lower the binding energy.

The modification of dislocation mobility model of hydrogen-assisted cracking proposed by Beachem (7), unites parts of the previous cracking models to explain the observed decrease in microscopic plasticity and changes of fracture mode associated with some forms of hydrogen-assisted cracking. This model suggests that the presence of

sufficiently concentrated dissolved hydrogen in the lattice just ahead of the crack tip unlocks dislocations, allowing them to multiply or move at reduced stresses. This process aids whatever deformation process the microstructure will permit (intergranular, quasicleavage or microvoid coalescence) and thus explains various fracture features observed in HAC.

Prevention of Hydrogen-Assisted Cracking

Crack avoidance in Cr-Mo steels must be accomplished by hydrogen and temperature control. Low hydrogen welding processes and procedures must be used in fabrication of Cr-Mo steels. Temperature control includes both preheat and post weld heating methods of preventing HAC. The general effects of preheating have been summarized by Coe (6) as follows:

1. Reducing cooling rate in the HAZ to produce softer and less susceptible microstructures;
2. Allows "self" tempering of martensite;
3. Promotes diffusion of hydrogen away from the HAZ after welding.

Because of the high hardenability of Cr-Mo steels, preheat has little effect on their HAZ microstructure and hardness. Apblett et al. (16) showed that in the 2 1/4Cr-1Mo material, softer transformation products could not be produced by increasing preheat temperature (within practical limits). Preheats as high as 600°F (315°C) resulted in only a minor decrease in the maximum HAZ hardness. Edmonds and Chew (17) found that in the higher hardenability 9Cr-1Mo material, energy input and preheat had no effect on the measured percentage

of ferrite present in the HAZ. It was also shown that the energy input did not significantly affect hardness in the HAZ of the 9Cr-1Mo material.

In Cr-Mo and other steels, preheat increases the rate of diffusion of hydrogen and thus accelerates its removal from the weld. Preheat requirements are related to chemical composition and thickness (Table 2). As thickness and alloy content increase, the preheat temperatures to guard against cracking also increase. Carbon content is the most important factor in establishing preheat temperatures. The presence or absence of free ferrite in the structure is a secondary consideration. To keep welding problems to a minimum, Cr-Mo steels are usually manufactured with a maximum carbon content of 0.15%. In addition to alloy content and material thickness, the mass of the joint and degree of restraint must also be considered when determining preheat temperatures.

When preheat cannot remove sufficient hydrogen to prevent cracking, a weld interpass temperature or post heating temperature high enough to avoid cracking must be held for a sufficient length of time to allow hydrogen to diffuse out of the HAZ before the weld is allowed to cool to ambient temperature. The decision as to post heat temperature and time must be based on consideration of the hydrogen concentration at the end of welding and a critical concentration below which cracking will not occur when the weld cools to ambient temperature. There is presently little information on critical hydrogen concentrations for Cr-Mo steels and post-heating procedures are generally based on practical experience (18).

Table 2. Recommended Minimum Preheat Temperatures for Welding Chromium-Molybdenum Steels.

Steel ^a	Thickness, in.		
	Up to 0.5	0.5 to 1.0	Over 1.0
1/2Cr-1/2Mo	100°F	200°F	300°F
1Cr-1/2Mo			
1-1/4Cr-1/2Mo			
2Cr-1/2Mo	150	200	300
2-1/4Cr-1Mo			
3Cr-1Mo	250	300	400
5Cr-1/2Mo	400	400	500
7Cr-1/2Mo			
9Cr-1Mo			

^aMaximum carbon content of 0.15%. For higher carbon content, the preheat temperature should be increased 100° to 200°F. Lower preheat temperatures may be used with gas tungsten arc welding.

Source: "Chromium-Molybdenum Steels," Welding Handbook, 7th Edition, Vol. 4, Chap. 1, American Welding Society, (1983).

Sometimes, for various reasons, it is impossible to preheat or post heat Cr-Mo steels which are to be welded. Under these conditions the only alternative is to use a combination of welding process and consumable which produces a weld metal that is insensitive to hydrogen and has a high solubility for hydrogen, thus preventing the introduction of hydrogen into the crack sensitive HAZ. This can be best achieved by employing austenitic weld metal. At room temperature the austenitic weld metal has a higher solubility for hydrogen and a lower diffusivity than the Cr-Mo HAZ. Thus, any hydrogen entering the HAZ from the weld can diffuse away faster than it is introduced and there is less chance of the hydrogen concentration reaching a critical level.

Compositional and Microstructural Effects on Hydrogen-Assisted Cracking

The HAZ microstructure is one of the most important factors in determining the HAC susceptibility of any steel. Linnert (14) has listed, in order of increasing propensity to HAC, six microstructures commonly found in the HAZ of steel welds:

1. Ferrite and carbides (e.g., pearlite);
2. Bainite;
3. Bainite mixed with martensite;
4. Martensite;
5. Martensite with some ferrite;
6. Martensite with internal twinning.

The microstructure produced in any steel weld is dependent upon the cooling rate through the transformation temperature range of the

austenite, the austenite grain size and composition (hardenability). The cooling rate is governed by energy input, plate thickness, joint geometry and preheat temperature.

Chromium-molybdenum steels are hardenable in the HAZ under normal welding conditions. The weld HAZ microstructure of a Cr-Mo steel is generally dependent on the alloy content. The weld HAZ of the more common types of Cr-Mo steels can be classified as follows:

1. Low alloy—bainitic and/or martensitic microstructure (0.5% to 5% Cr);
2. Martensitic—predominantly martensitic microstructure with some delta-ferrite present (5% to 12% Cr);
3. Duplex—martensitic-ferritic microstructure (9% to 12% Cr and 2% Mo).

Smith and his associates (19) have shown that the HAZ structure of 2 1/4Cr-1Mo steel can be classified as follows:

1. Grain coarsened region;
2. Grain refined region;
3. Region of partial transformation;
4. Subcritical region.

The grain coarsened region was found to consist of bainite and/or autotempered martensite.

Additions of over 5% chromium result in a martensitic structure on cooling through the transformation temperature. One of the more common martensitic Cr-Mo steels is 9Cr-1Mo. The chemical composition of the 9Cr-1Mo steel is such that it is close to the austenite-ferrite

phase boundary of the iron-chromium equilibrium diagram (Figure 4), and delta-ferrite may be present on cooling to ambient temperatures depending on the composition and austenitizing temperature (20).

The influence of various alloy modifications on the amount of delta-ferrite present in the martensitic materials after normalizing can be estimated by the chromium equivalent (Cr.E.) equation (21):

$$\begin{aligned} \text{Cr.E.} = & \text{Cr} + 6\text{Si} + 4\text{Mo} + 1.5\text{W} + 11\text{V} \\ & + 5\text{Nb} + 8\text{Ti} + 12\text{Al} - 40\text{C} - 30\text{N} - 4\text{Ni} - 2\text{Mn} - \text{Cu} . \quad (2-2) \end{aligned}$$

When the Cr.E. is less than ten, delta-ferrite is usually absent.

If the Cr.E. ranges from ten to twelve some delta-ferrite may be present. For Cr.E. greater than twelve, delta-ferrite may be present in large quantities.

During welding, the delta-ferrite in 9Cr-1Mo steel may be sufficiently stabilized so that it persists in the weld metal and in the HAZ close to the fusion boundary (23). Edmonds and Chew (17) have shown that the amount of delta-ferrite present in GTA (gas-tungsten arc) welds is a function of Cr.E. and is largely independent of welding conditions.

The HAZ in martensitic 12Cr-1Mo steel has been shown by researchers at Sandia National Laboratories (24) to consist of four distinct regions (Figure 5):

1. A two-phase region adjacent to the fusion line, containing martensite and some delta-ferrite;

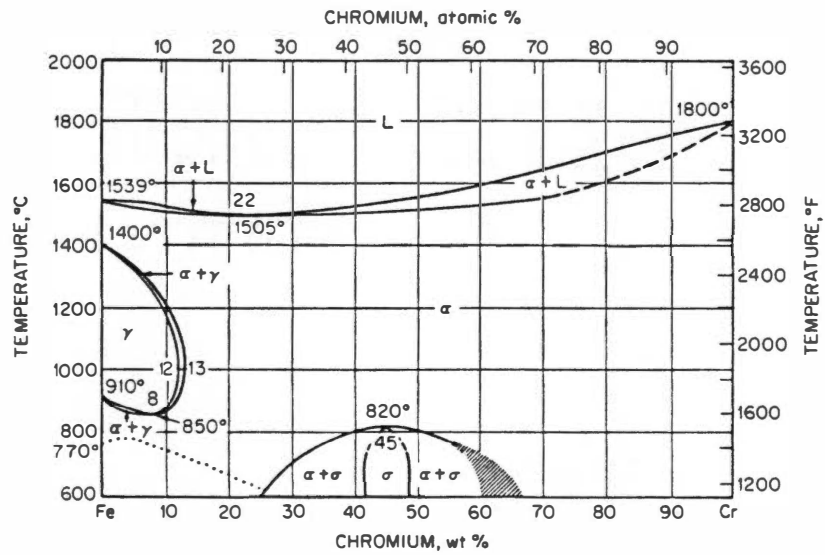


Figure 4. Iron-chromium phase diagram.

Source: Demo, J.J., "Structure and Constitution of Wrought Ferritic Stainless Steels," Handbook of Stainless Steels, Chap. 5, McGraw-Hill, (1977).

FOUR HAZ REGIONS

- ① $\gamma + \text{FERRITE} \rightarrow \text{MARTENSITE} + \text{FERRITE}$
- ② COARSE-GRAINED $\gamma \rightarrow \text{MARTENSITE}$
- ③ FINE GRAINED $\gamma \rightarrow \text{MARTENSITE}$
- ④ OVERTEMPERED B.M.

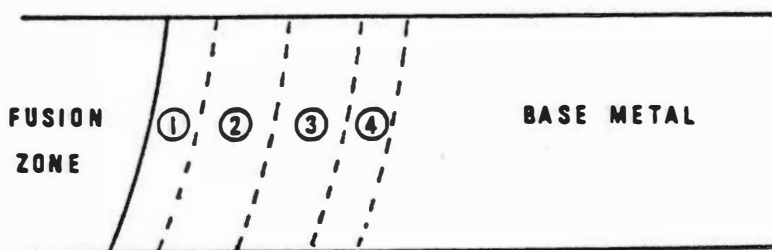


Figure 5. Four HAZ regions of HT-9 (12Cr-1Mo).

Source: Lippold, J.C., "Weldability of HT9: The Autogenous GTA Process," ADIP Performance Quarterly for the Period Ending July, 1980, DOE-ER-0045-2, 98-108 (1980).

2. A fully martensitic region exhibiting a relatively coarse prior austenite grain size;
3. A martensitic region exhibiting a finer prior austenitic grain size;
4. An overtempered base metal microstructure.

The HAZ structure of the 12Cr-1Mo material is explained by the observations of Fidler and Gouch (23). In the HAZ delta-ferrite forms around grain boundaries restricting grain boundary migration. The region of maximum grain growth is not immediately adjacent to the weld metal but is displaced further away in the HAZ. Due to the presence of delta-ferrite, only a portion of the HAZ structure is 100% martensite. The presence of the soft ferrite in the otherwise martensitic structure decreases the overall hardness of the steel.

Originally it was thought that the formation of delta-ferrite in martensitic Cr-Mo steels should be avoided because delta-ferrite was considered to reduce the maximum obtainable strength of the material (25). To insure that delta-ferrite would not be present, modest levels of austenite formers (particularly carbon) were added to the martensitic chromium-molybdenum steels. The increased level of carbon in these materials resulted in decreased weldability.

The recent development of lower carbon duplex (martensitic-ferritic) Cr-Mo steels shows that delta-ferrite is not necessarily detrimental to the physical properties of 9Cr-2Mo and 12Cr-2Mo materials (26,27). The lower carbon composition tends to close the gamma loop of the iron-chromium phase diagram, thus increasing the likelihood

that delta-ferrite will remain in the base metal on cooling to ambient temperatures from austenitizing temperatures.

Researchers at Climax Molybdenum Company (28) have identified three HAZ regions in the 12Cr-2Mo duplex material. The first region located in the HAZ near the unaffected base metal experiences a peak temperature into the (austenite plus ferrite) phase field which results in a partial transformation from martensite to austenite. The second region, which is closer to the fusion zone, experiences a higher peak temperature in the (austenite plus ferrite) phase range which results in the complete dissolution of carbides into the austenite. The third HAZ region is adjacent to the fusion zone. The peak temperature in this region is such that considerable amounts of delta-ferrite form. This ferrite is retained on cooling. The high peak temperature and relatively high cooling rate in the third region results in the formation of acicular austenite islands (from delta-ferrite), which transforms to martensite during cooling.

As the chromium content of Cr-Mo steels increases, the HAZ microstructures change from a bainitic/bainitic-martensitic microstructure to a martensitic-ferritic or fully martensitic microstructure. When the microstructure becomes more martensitic, the HAC susceptibility also increases. The increase in cracking susceptibility of the higher alloy Cr-Mo steels is reflected in Table 2 (p. 17).

Another important factor in determining the HAC sensitivity of Cr-Mo steels is the HAZ hardness. As the hardness of the HAZ increases the susceptibility to HAC increases. In Cr-Mo steels the

hardness of the HAZ is primarily dependent on the carbon content (29). The reduction of carbon content in martensitic Cr-Mo steels in an effort to improve weldability results in an increase in the amount of delta-ferrite present in the normalized and tempered microstructure. The resulting ferritic/martensitic microstructure is characteristic of the duplex Cr-Mo materials.

In addition to affecting the microstructural characteristics of Cr-Mo steels, both chromium and carbon, as well as molybdenum and nickel, lower the temperature at which the transformation of austenite to martensite starts and is completed (M_s and M_f temperatures) (14). From the standpoint of HAC sensitivity it is undesirable to decrease martensitic transformation temperatures because as the martensite transforms the solubility of hydrogen in the matrix is decreased, and at the lower transformation temperatures there is little chance for the hydrogen to diffuse out of the HAZ. Higher transformation temperatures also permit some tempering of martensite to occur which reduces hardness and lowers residual stresses.

Richly alloyed Cr-Mo steels may, on cooling, under welding conditions retain austenite. The addition of 6% nickel to 13Cr-1.5Mo steel has been shown by Arwidson et al. (30) to produce a high stability austenite which during tempering will withstand the rapid heating to high temperatures generated in the HAZ during welding. The retained austenite is able to absorb hydrogen and act as a sink to reduce the amount of hydrogen that can be absorbed by martensite and thus prevent hydrogen embrittlement of the martensite.

Weldability of Chromium-Molybdenum Steels

Although most of the applications of Cr-Mo steels require welding, there is little published work on the weldability of these steels prior to 1969 (31). Most of the literature since 1969 is the result of investigations into the weldability of Cr-Mo materials for use in liquid metal fast breeder (LMFB) reactors. In this application, Cr-Mo steels are stabilized with additions of vanadium and/or niobium to prevent decarburization of materials in contact with liquid sodium, or to enhance creep properties. These stabilized materials are the subject of most of the currently available information on weldability.

Smith et al. (19) were some of the first researchers to investigate the HAZ of both stabilized and standard composition 2 1/4Cr-1Mo. The grain coarsened region of samples experiencing a welding thermal cycle were metallographically examined and it was found that the grain coarsened region of the standard material consisted of martensite, while the niobium stabilized material exhibited upper bainite in the HAZ and was less crack sensitive. Thus, the results of this investigation suggested that the stabilized 2 1/4Cr-1Mo material had improved weldability over the unstabilized material.

Colombe (32) and his associates employed the implant test to compare the relative cracking susceptibilities of stabilized and standard composition 2 1/4Cr-1Mo. The implant test (33) is a weld cracking test that applies a controlled stress on the weld HAZ and evaluates cracking susceptibility by means of a mechanical test

parameter. In this test, an actual weld is made over a notched cylindrical test specimen that is inserted into a hole in a test plate (Figure 6). Hydrogen is added to the weld by controlling the arc atmosphere. After the sample cools to a designated temperature (usually 300°F (150°C)), it is stressed to simulate the actual restraint that occurs in a weld during welding. If the specimen does not rupture within 24 hours the test is discontinued. The highest stress level for which no fracture occurs in 24 hours is called the critical stress level. The critical stress level is regarded as a relative measure of the hydrogen-assisted cracking susceptibility of a given material. Materials with a low cracking susceptibility have higher critical stress levels than materials with a high cracking susceptibility. It was the conclusion of Colombe et al. that the risk of HAC in the stabilized 2 1/4Cr-1Mo steel was less than that in the standard composition material.

Teraï et al. (35) investigated the HAC sensitivity of standard composition 2 1/4Cr-1Mo and stabilized composition 2 1/4Cr-1Mo-Nb employing both the Y-groove and U-groove restraint cracking tests. The Y-groove test and U-groove test are self restraint cracking tests used to determine the HAC susceptibility of welds. Both plate assemblies are shown in Figure 7. In these techniques, test welds are made at various preheats. The welds are allowed to cool to room temperature and sectioned to observe the location and percentage of

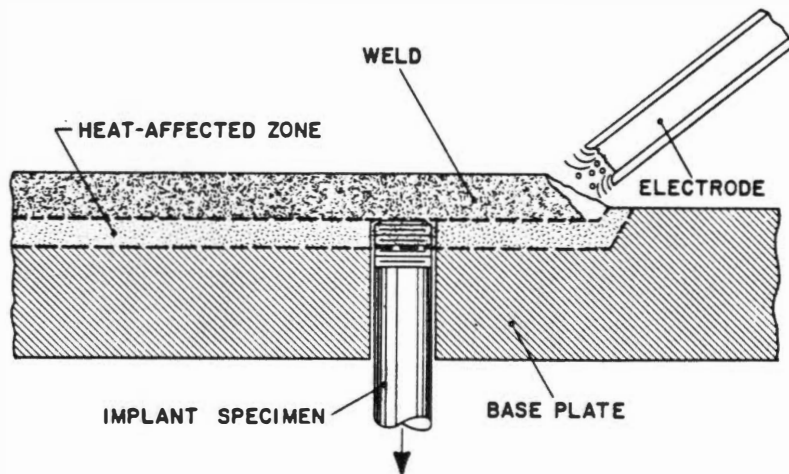


Figure 6. Implant test setup.

Source: Dickenson, D.W., and Ries, G.D., "Implant Testing of Medium to High Strength Steel--A Model for Predicting Delayed Cracking Susceptibility," Welding J., 58(7), 205s-211s (1979).

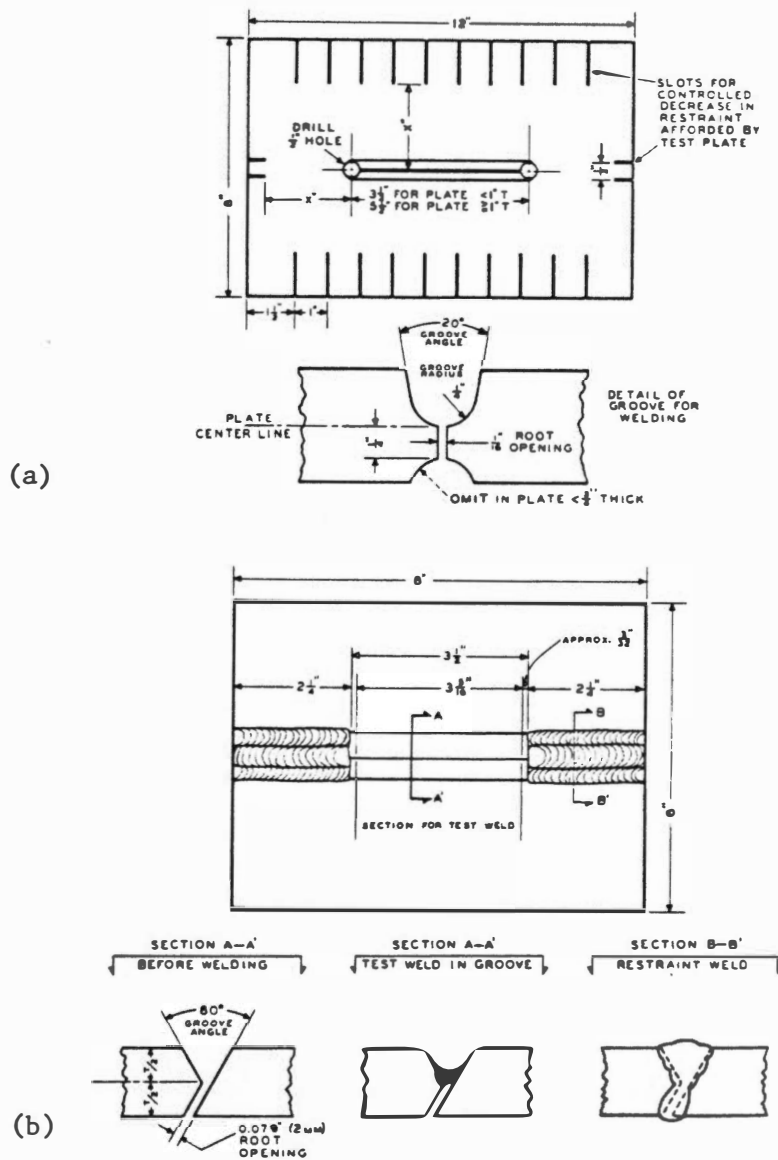


Figure 7. Hydrogen-assisted cracking restraint tests.
(a) U-groove, (b) Y-groove.

Source: Linnert, G.E., Welding Metallurgy, Vol. 2, American Welding Society, (1967).

cracking.* The Y-groove test is better suited to evaluate the cracking susceptibility of the base metal while the U-groove is better suited to evaluate the cracking sensitivity of weld metals. The Y-groove test results of Terai et al. showed that the preheat temperature to prevent cracking in the stabilized 2 1/4Cr-1Mo material was 440°F (225°C), 45°F (25°C) higher than that of the standard composition 2 1/4Cr-1Mo material. In contrast the results of the U-groove test showed that the preheat temperature to prevent cracking in the standard composition material was 620°F (325°C), 45°F (25°C) higher than that for the stabilized material. The preheat temperature differences between the two test methods were considered to be small and it was concluded that the weldment crack sensitivity for the stabilized steel was equivalent to that of the standard composition material.

Daikoku (36) and his associates employed the Y-groove restraint test to determine the relative weldability of duplex 9Cr-2Mo and standard composition 9Cr-1Mo. No cracks were observed in the duplex 9Cr-2Mo material at 212°F (100°C) preheat while the standard composition 9Cr-1Mo exhibited cracking at 390°F (200°C) preheat. Based on preheat levels to prevent cracking, the duplex 9Cr-2Mo material exhibited improved weldability over the standard composition 9Cr-1Mo material.

* Percentage cracking = $\frac{\text{sum of crack height}}{\text{sum of weld metal height}} \times 100\%$.

Wada (27,37) and his associates have employed the Y-groove restraint test to establish the relative level of weldability of 9Cr-1Mo-V-Nb, 9Cr-2Mo, 12Cr-1Mo-W-V, 12Cr-2Mo-0.1C-V and 12Cr-2Mo-0.05C-V. The duplex materials (9Cr-2Mo and 12Cr-2Mo) exhibited no cracks when welded at room temperature (75°F (24°C)). A preheat level of 300°F (150°C) was required to prevent cracking in the 9Cr-1Mo-V-Nb material while the 12Cr-1Mo-V material exhibited cracking at the 300°F (150°C) preheat level. Based on these two investigations, Wada and his associates concluded that the duplex materials are more readily weldable than the straight martensitic types. However, the metallurgical basis for this improvement was not investigated.

Yukitoshi et al. (38) investigated the HAC sensitivity of standard composition 9Cr-2Mo, and (vanadium/niobium) modified 9Cr-2Mo. Gas-tungsten arc Y-groove welds were made on all materials at ambient temperature and it was observed that no cracks occurred in the unmodified 9Cr-2Mo material, while cracking occurred over the entire length of the weld bead when the niobium content in the modified material exceeded 0.6%. The cracking was considered to be hydrogen related however.

In summary, review of the current literature reveals that a surprisingly limited amount of information exists on the relative HAC sensitivity of Cr-Mo steels, although these materials have been fabricated by welding for a number of years. Most of the studies to date have concentrated on establishing the relative cracking sensitivity of only a few selected materials and little effort has

been directed towards explaining the observed cracking behavior on the basis of compositional or microstructural differences. It is therefore apparent that a more broad based investigation of the relative HAC susceptibility of Cr-Mo materials is needed to more adequately address the effects of microstructure and composition on cracking sensitivity.

Stress-Relief Cracking

Stress-relief cracking (SRC) or reheat cracking may be defined as intergranular cracking in the HAZ that occurs during the exposure of welded assemblies to the elevated temperatures produced by either post-weld heat treatment, or high temperature service (39). McPherson (40) has summarized the main features of stress-relief cracking as:

1. Cracking that occurs in the coarse grained HAZ close to the fusion boundary.
2. Intergranular cracking along prior austenite grain boundaries.
3. Fracture morphologies which may be smooth and apparently featureless, or may show evidence of cavitation.
4. Cracking that occurs in the temperature range of 1020 to 1200°F (550 to 650°C) and which may show evidence of a "C-curve" time-temperature dependence.
5. Susceptibility to cracking which is related to certain types of alloying elements present. Steels containing elements of the group Cr, Mo and V are particularly sensitive.

6. Susceptibility to cracking of a particular steel type shows evidence of heat to heat variations, which suggests that minor elements are important.
7. Cracking favored by welding conditions and geometries that produce high residual stresses.

Two different theories have been proposed to explain the experimental observations of SRC. It was the conclusion of several SRC studies (39,41,42) that carbide precipitation within the grains during stress relieving created precipitate denuded zones (PDZ) at prior austenite boundaries. It was proposed that most of the strains resulting from stress relaxation were concentrated in the denuded zones causing a rapid increase in dislocation density and eventual intergranular cracking. Stone and Murray (41) observed that a decrease in austenitic grain size in Cr-Mo-V steels produced a decrease in SRC susceptibility. According to the precipitate-denuded region theory, the decrease in austenitic grain size would distribute the volume of the denuded region over more boundary surface. The increased area of the grain boundary was then able to absorb more dislocation movement, thus reducing the extent of strain accumulation. The denuded region was considered to result from segregation of Al and Si. According to Stone and Murray, both aluminum and silicon form a gamma loop phase boundary region with iron. The boundary of the loop occurs at 1.0 wt % and 2100°F (1150°C) for aluminum and 2.15 wt % and 2140°F (1170°C) for silicon. If the localized level of aluminum or silicon in the grain boundary region exceeds the

boundary of the gamma loop at elevated temperatures, ferrite may form. Carbon would then diffuse away from the grain boundary region into the austenite because of the lower solubility of carbon in ferrite. The grain boundary regions would be denuded of precipitate because the precipitation of carbides would be inhibited by the low carbon concentration.

Commonly observed heat-to-heat variations in SRC susceptibility prompted Boniszewski and Eaton (43) to suggest that SRC might be related to the presence of certain residual elements. Aluminum has been reported to be particularly detrimental (44,45). Recent studies by Hipsley and his coworkers (46,47) suggest that low ductility intergranular SRC in 2 1/4Cr-1Mo is promoted by increasing concentrations of impurity elements (particularly phosphorus). Swift (48) has proposed a mechanism of stress-relief cracking in 2 1/4Cr-1Mo steel based on dislocation locking and the formation of coherent precipitates both of which result in cracking along prior austenitic grain boundaries in the grain coarsened HAZ.

There is no doubt that various alloying elements contribute to the stress-relief cracking susceptibility of a material. Nakamura (49) and Ito (50) have derived equations which relate chemical compositions of a steel to its susceptibility to stress-relief cracking:

$$\Delta G = Cr + 3.3 Mo + 8.1V - 2 \quad (2-3)$$

$$P_{SR} = Cr + Cu + 2 Mo + 10V + 7Nb + 5Ti - 2 \quad (2-4)$$

When ΔG or P_{SR} are positive the steel is said to be susceptible to

reheat cracking. However, these formulas are not applicable to most chromium-molybdenum steels because they are limited to materials with a chromium level less than 1.5%.

Many of the SRC investigations of Cr-Mo steels to date have been limited to 2 1/4Cr-1Mo weld metal and base metal. In general, increasing the chromium level from 0.5 to 3% improves the creep ductility and thus reduces the susceptibility to stress-relief cracking (51). One possible explanation for this behavior is that the increased chromium level decreases the likelihood of the formation of harmful precipitates such as V_4C_3 because Cr_7C_3 or $M_{23}C_6$ carbides are formed instead.

Bland (52) studied the effect of PWHT on the ductility of 2 1/4Cr-1Mo and 1 1/4Cr-1/2Mo weld metal. Stress relief of the 2 1/4Cr-1Mo material at 1050°F (565°C) resulted in severe embrittlement while stress relieving at temperatures ranging from 1250° to 1350°F (680 to 730°C) caused no embrittlement. Stress-relief cracking was not observed at 1150°F (620°C). In all cases the embrittlement was accompanied by secondary hardening.

Swift and Rodgers (53) examined the stress-relief embrittlement of 2 1/4Cr-1Mo weld metal subjected to post weld heat treatments ranging from 1/2 hr at 500°F (260°C) to 64 hrs at 1350°F (730°C). This work indicated that SRC of 2 1/4Cr-1Mo weld metal was the result of precipitates inherent in the alloy system and was not the result of residual elements. Batte and Murphy (54) have also examined stress-relief cracking in 2 1/4Cr-1Mo weld metal and have determined that

the SRC susceptibility was only slightly dependent on the residual element content.

Murray (55) examined stress-relief cracking in carbon and low alloy steels. In their study, 1/2CrMoV, 1/2MoB and 2 1/4CrMo steels were shown to be susceptible to SRC while C-Mn, 1CrMo, 5CrMo and 9CrMo steels showed no evidence of cracking tendency under the same test conditions. It was determined that those steels which contained significant amounts of molybdenum or molybdenum plus vanadium exhibited increased susceptibility to SRC.

Pierce and Siverns (56) investigated the SRC of 1/2Cr-1/2Mo-1/2V and 2 1/4Cr-1Mo. Oil quenched laboratory specimens of both materials were preloaded and heated at 90°F/hr (50°C/hr) to 930 to 1290°F (500 to 700°C). Fracture occurred at stress levels lower than that for an isothermal hot tensile test at the same temperature. It was apparent from this investigation that some metallurgical change had occurred during reheating which was responsible for the observed fracture at lower stress levels.

Miller and Batte (57) examined SRC in 1/2CrMoV and 2 1/2Cr-1Mo. Detailed examination of the effects of residual elements on cracking susceptibility of both weld metal and parent metal suggested that the cracking susceptibility of both weld metal and parent metal was not entirely related to chemical composition. Preliminary results indicate that it is possible to improve crack susceptibility by refinement of the coarse grained HAZ. This observation suggests

that residual elements are not necessarily a major factor in determining the overall cracking susceptibility.

Recent studies by Hippsley and his associates (46,47) on the effect of phosphorus on stress-relief cracking in 2 1/4Cr-1Mo have shown that two distinct stress-relief fracture morphologies occur in different temperature ranges during the heating period of post-weld heat treatment. At low temperatures (800 to 900°F (425 to 480°C)), a smooth intergranular fracture mode was noted, but at higher temperatures (980 to 1115°F (525 to 600°C)) an intergranular microvoid coalescence mode was observed. There was no evidence of a precipitate-free zone or alloying element depletion observed when microvoid coalescence fracture occurred along the prior austenite grain boundaries. It was suggested that a fine grain boundary dispersion of MnS precipitates is associated with the intergranular microvoid coalescence. A comparison of the microstructure of the HAZ in four alloys by Hippsley (47) showed little difference between them, thus microstructural variations were not responsible for the variations in stress-relief cracking behavior observed. However, the variations may result from differences in the type and concentration of impurity elements present.

In summary, examination of the currently available literature shows that for the higher alloy materials (e.g., 2 1/4Cr-1Mo, 9Cr-1Mo, 9Cr-2Mo, 12Cr-1Mo and 12Cr-2Mo), very little information exists on SRC susceptibility. One investigation has suggested that 1/2Cr-1/2Mo-V and 2 1/4Cr-1Mo may be susceptible to SRC while 1Cr-1/2Mo, 5Cr-1/2Mo and 9Cr-1Mo may not be susceptible to cracking.

It is evident that additional information is needed on the SRC potential of the higher alloy chromium-molybdenum materials with particular emphasis on any compositional and microstructural variations which might account for observed differences in SRC susceptibility. There also exists a need to evaluate the effect of various post-weld heat treatment temperatures on SRC susceptibility.

Transformation Behavior

In this review it has been shown that the coarse grained HAZ adjacent to the fusion zone (peak temperature in excess of 2100°F (1150°C)) is the region of a weld in Cr-Mo steel which exhibits the highest potential for both HAC and SRC. It has also been pointed out that the transformed microstructure of the coarse grained HAZ is the most important factor in determining its susceptibility to HAC and SRC. For instance, a fully martensitic microstructure may be more susceptible to HAC than a bainitic structure (14), while their relative susceptibility to SRC may be the reverse (58). Furthermore, it is important to know the transformation temperature ranges in the HAZ, especially the M_s and M_f range since martensite which transforms at low temperatures (below 500°F (260°C)) may exhibit an increased susceptibility to HAC (14).

The transformation behavior of the HAZ in Cr-Mo steels is governed by austenite composition, peak temperature and to some extent cooling rate. Continuous-cooling transformation (CCT) diagrams may be developed to predict the microstructure and transformation temperatures in the coarse grained HAZ for conditions appropriate to

welding. CCT diagrams developed for welding conditions take into account the effects of peak temperature and incomplete homogenization on CCT behavior.

Review of current literature reveals that only a limited amount of information exists on the CCT behavior of the low alloy 2 1/4Cr-1Mo material under welding type conditions. No readily available information was found on the CCT behavior of martensitic or duplex Cr-Mo steels under welding conditions, however it is sometimes possible to estimate the microstructures which may be found in the HAZ of Cr-Mo steels based on conventional CCT diagrams which are readily available (29). Conventional CCT diagrams assume that the austenite is completely homogenized with a grain size of ASTM 7-8 prior to transformation. An actual weld HAZ exhibits various grain sizes (the coarse grained HAZ may exceed ASTM 1) as well as various degrees of incomplete homogenization and the CCT behavior may thus depart significantly from the conventionally determined behavior.

The on-cooling transformation of the HAZ of duplex Cr-Mo steels may be estimated from a CCT diagram such as that shown in Figure 8 for 12Cr-2Mo-Ni. Because of the duplex nature of this material, upon cooling from the austenitizing temperature a fully martensitic microstructure is never obtained. The weld HAZ will therefore always contain some untransformed ferrite. As depicted in the Cr-Fe phase diagram in Figure 9, a peak temperature between 2000°F and 2800°F (1095°C to 1540°C) in the HAZ results in the transformation of delta-ferrite from austenite. The cooling rate in the HAZ adjacent to the

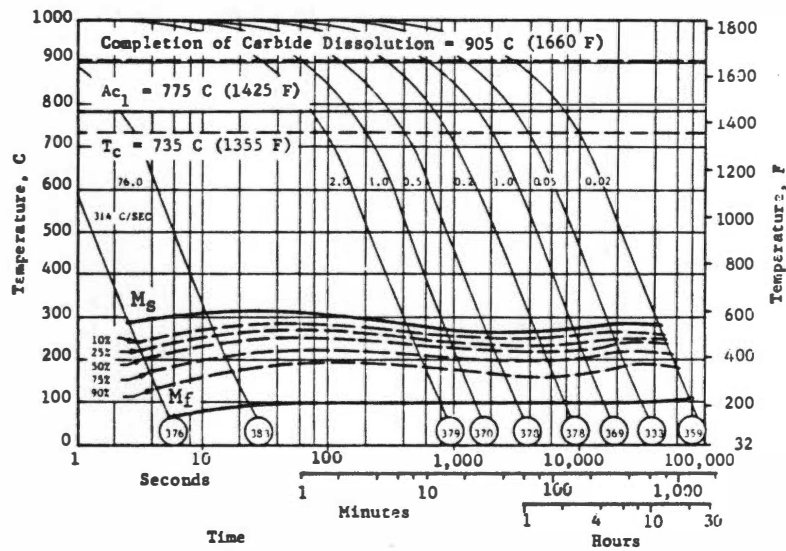


Figure 8. CCT diagram of 12Cr-2Mo-1Ni-0.08C-0.25V.

Source: Wada, T., Grobner, P.J., and Vineberg, E.J., "Duplex Structure 12Cr-2Mo Steels with Improved Weldability," Proc. Int. Conf. on Production, Fabrication, Properties and Applications of Ferritic Steels for High Temperature Applications, American Society for Metals, Warren, PA, (1981).

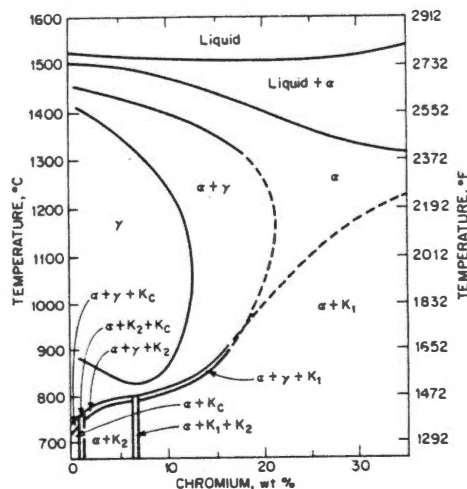


Figure 9. Constant section at 0.05% carbon through the Fe-Cr-C phase diagram.

Source: Lovejoy, P.T., "Structure and Constitution of Wrought Martensitic Stainless Steels," Handbook of Stainless Steels, McGraw-Hill, (1977).

fusion zone is sufficient to retain delta-ferrite in the final ambient temperature microstructure.

The CCT diagram of a martensitic Cr-Mo steel (Figure 10), like that of the duplex steel, is relatively simple, in that martensite is the primary on-cooling transformation product present at ambient temperature. However, on-heating, austenite may also transform to delta-ferrite (Figure 11) in the HAZ and be retained in the final structure. However, less delta-ferrite may form in the HAZ of the martensitic type Cr-Mo steel than in the duplex Cr-Mo steel because it is higher in carbon and lower in molybdenum (Cr.E. for the martensitic materials approximately 9 to 12).

Apblett and his associates (16) examined the simulated weld CCT behavior of 2 1/4Cr-1Mo in an effort to find preheat conditions that would result in softer transformation products in the HAZ. Bainite was found to be the primary transformation product in regions of the simulated HAZ experiencing a peak temperature greater than 2000°F (1095°C). On-cooling any untransformed austenite existing below 650°F (345°C) was found to transform to martensite. Practical preheat levels, up to 600°F (315°C), were found to have no appreciable effect on transformation products in the HAZ.

The CCT behavior of 2 1/4Cr-1Mo is more complex than that of the predominately martensitic higher alloy Cr-Mo steels. The bainitic/bainitic-martensitic HAZ microstructure of the 2 1/4Cr-1Mo material is not independent of cooling rate as is the higher alloy Cr-Mo materials. For any given preheat temperature, an increase in energy input causes

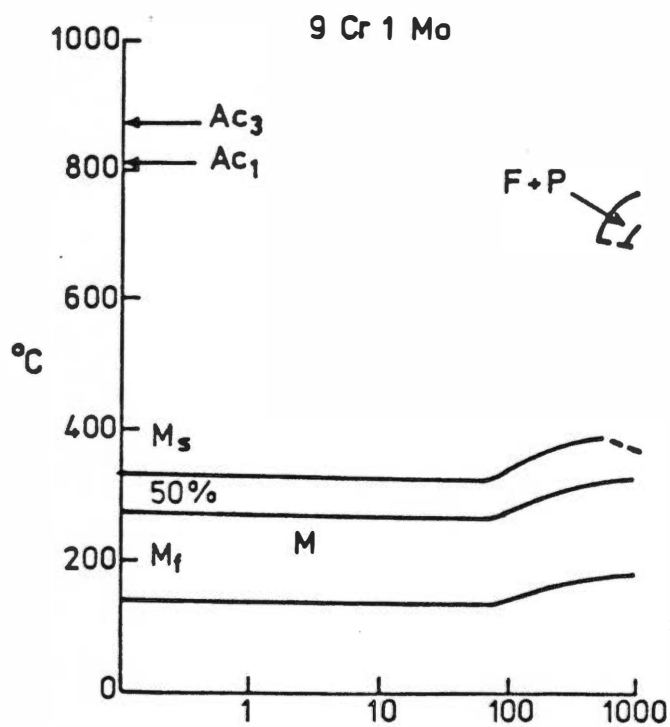


Figure 10. CCT for air cooled 9Cr-1Mo steel.

Source: Orr, J., Beckitt, F.R., and Fawkes, G.A., "The Physical Metallurgy of Chromium-Molybdenum Steels for Fast Reactor Boilers," Proc. Int. Conf. on Ferritic Steels for Fast Reactor Steam Generators, British Nuclear Energy Society, London, England, 91-109 (1978).

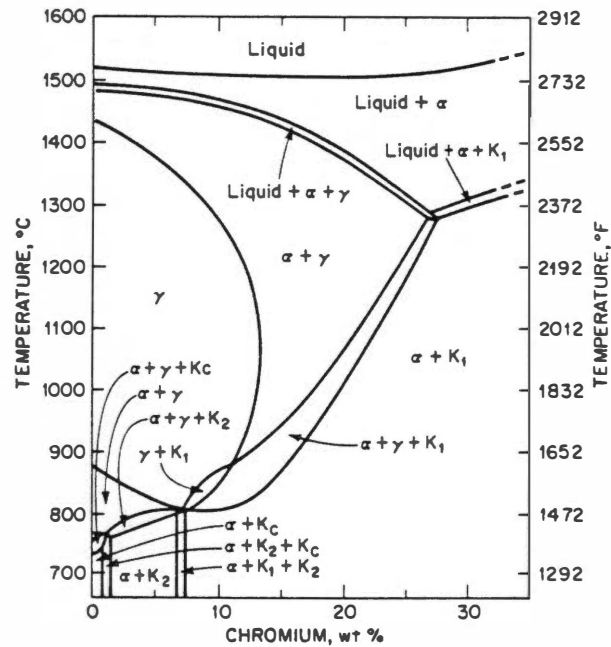


Figure 11. Constant section at 0.10 carbon through the Fe-Cr-C phase diagram.

Source: Lovejoy, P.T., "Structure and Constitution of Wrought Stainless Steels," Handbook of Stainless Steels, McGraw-Hill, (1977).

a decrease in the cooling rate. The resulting influence of reduced cooling rates on CCT behavior of the HAZ in 2 1/4Cr-1Mo can be estimated by examining the conventional CCT diagram (Figure 12). According to this diagram as the cooling rate decreases, the microstructure proceeds from martensite and/or fine bainite to a coarse bainite (60). Carpenter and Emmanuel (61) have also noted that coarser bainite is produced at slower continuous-cooling rates.

It has already been mentioned that at sufficiently high cooling rates, martensite is present in 2 1/4Cr-1Mo. Habraken and Economopoulos (62) have shown that for 2 1/4Cr-1Mo cooled at 6.6°F/sec (3.7°C/sec) the martensite is present in the form of martensite/austenite (MA) "islands" within the bainite. Wada and Eldis (63) have also observed MA "islands" in 2 1/4Cr-1Mo steel cooled at 0.33°F/sec (0.18°C/min). In the discussion of his work, Wada suggests that during continuous-cooling, austenite islands form instead of carbides during the bainite transformation. During continued cooling a large fraction of this austenite then transforms to martensite.

In summary, it has been shown that the transformation behavior of the HAZ adjacent to the fusion zone has considerable influence on HAC and SRC susceptibility of welds in Cr-Mo steels. CCT diagrams are essential to appraise the transformation behavior of the adjacent HAZ region. The literature review reveals that little information is available on the transformation of Cr-Mo steels under welding conditions. Thus, it is apparent that additional CCT information applicable to welding of Cr-Mo steels is needed. This CCT

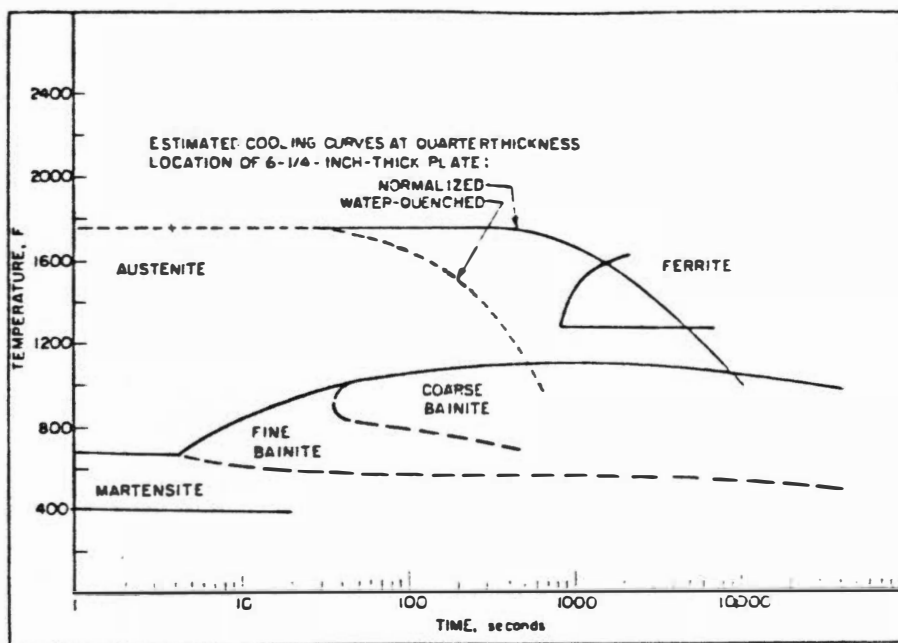


Figure 12. CCT diagram for 2 1/4Cr-1Mo.

Source: Brown, R.M., Rege, R.A., and Spaeder, C.E., "The Properties of Normalized and Tempered and of Quenched and Tempered 6 1/4-Inch-Thick Plates of 2 1/4Cr-1Mo Steel," Symposium of Heat treated Steels for Elevated Temperature Service, ASME, New Orleans, LA, (1966).

information is essential to the basic understanding of HAC and SRC in Cr-Mo steels.

CHAPTER III

EXPERIMENTAL PROCEDURES

The experimental procedures employed during the course of this investigation are discussed in this chapter.

Materials

Materials for this investigation were received from the Advanced Research and Technology Development (AR&TD) Fossil Energy Materials Program, at Oak Ridge National Laboratory (ORNL). Fifteen Cr-Mo materials were employed in this investigation. These materials can be separated into three general types based on their welded microstructure:

1. Duplex structure—12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni, 12Cr-2Mo-1Ni, 9Cr-2Mo-V-Nb, and 9Cr-2Mo;
2. Martensitic structure—9Cr-1Mo-0.11C, 9Cr-1Mo-0.06 and 9Cr-1Mo-V-Nb-0.08C (four heats designated as 394, 383, 176 and 148);
3. Bainitic structure—Lukens 3Cr-1.5Mo and 2 1/4Cr-1Mo (three commercial heats designated as Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C).

The chemical compositions of the investigation materials are listed in Table 3. Note that the three 12Cr-2Mo materials vary in Ni and Mo content (6%Ni-0.7%Mn, 3%Ni-2.3%Mn and 1%Ni-0.8%Mn) and contain additions of Nb and/or V. One of the 9Cr-2Mo materials has been modified with additions of V and Nb (0.10%V and 0.06%Nb). Four of

Table 3. Chemical Analysis.

Designation	Supplier	Composition (Weight %)																						
		C	Mn	P	S	Si	Ni	Cr	Mo	V	Nb	Ti	Co	Cu	Al	B	W	As	Sn	Zr	N	O	Cr.E.	Ni.E.
12Cr-2Mo-6Ni	Climax	0.066	0.74	0.012	0.014	0.29	5.92	11.96	2.09	0.21	0.09	<0.01	0.016	0.03	0.073	<0.001	0.05	<0.001	0.001	<0.001	0.039	0.002	-	9.27
12Cr-2Mo-3Ni	Climax	0.060	2.27	0.011	0.014	0.30	3.03	11.91	2.08	0.22	0.10	<0.01	0.014	0.01	0.054	<0.001	0.04	<0.001	0.001	<0.001	0.013	0.002	-	6.31
12Cr-2Mo-1Ni	Climax	0.068	0.82	0.011	0.009	0.31	1.03	11.86	2.04	0.20	<0.01	<0.01	0.011	0.02	0.014	<0.001	0.03	<0.001	0.002	<0.001	0.028	0.010	-	4.20
9Cr-2Mo-V-Nb	ORNL	0.034	0.61	0.009	0.014	0.05	0.05	9.62	2.07	0.10	0.06	<0.01	0.019	0.05	0.010	0.001	0.03	0.005	0.002	<0.001	0.027	0.005	16.12	-
9Cr-2Mo	Sumitomo	0.061	0.66	0.015	0.004	0.25	0.06	8.60	1.93	0.03	<0.01	<0.01	0.017	0.04	0.029	<0.001	0.02	0.003	0.002	<0.001	0.012	0.004	14.16	-
9Cr-1Mo-V-Nb(148)	ORNL	0.089	0.47	0.021	0.006	0.28	0.16	9.24	0.96	0.21	0.05	<0.01	0.019	0.08	0.002	0.001	0.01	0.002	0.004	0.001	0.035	0.008	11.12	-
9Cr-1Mo-V-Nb(176)	ORNL	0.081	0.37	0.010	0.003	0.11	0.09	8.61	0.89	0.21	0.07	<0.01	0.100	0.04	0.007	0.001	0.01	0.001	0.001	0.001	0.055	-	9.59	-
9Cr-1Mo-V-Nb(383)	ORNL	0.083	0.46	0.010	0.004	0.41	0.09	8.46	1.02	0.20	0.07	<0.01	0.055	0.03	0.002	0.001	0.05	0.001	0.001	0.001	0.051	-	11.52	-
9Cr-1Mo-V-Nb(394)	ORNL	0.084	0.46	0.010	0.003	0.40	0.09	8.57	1.02	0.20	0.07	<0.01	0.055	0.04	0.014	0.001	0.05	0.001	0.001	0.001	0.053	-	11.61	-
9Cr-1Mo-0.11C	ORNL	0.110	0.59	0.017	0.008	0.63	0.07	9.38	0.99	0.05	0.01	<0.01	0.081	0.03	0.001	0.001	0.01	0.004	0.009	<0.001	0.021	0.011	11.29	-
9Cr-1Mo-0.06C	ORNL	0.062	0.46	0.018	0.005	0.52	0.20	9.28	0.98	0.01	<0.01	<0.01	0.021	0.08	0.005	0.001	0.01	0.004	0.003	0.001	0.019	-	11.90	-
Climax 3Cr-1.5Mo	Climax	0.082	0.86	0.006	0.009	0.16	0.21	3.02	1.43	0.12	<0.01	<0.01	0.005	0.02	0.005	<0.001	<0.01	<0.001	0.001	0.001	0.010	-	-	-
Kawasaki 2 1/4Cr-1Mo-0.12C	Kawasaki	0.12	0.53	0.006	0.003	0.10	0.17	2.36	0.98	0.01	<0.01	<0.01	0.010	0.16	0.028	<0.001	<0.01	0.001	0.001	0.001	0.011	0.003	-	-
Lukens 2 1/4Cr-1Mo-0.10C	Lukens	0.10	0.44	0.006	0.026	0.19	0.16	2.18	1.02	0.01	<0.01	<0.01	0.014	0.15	0.003	<0.001	<0.01	0.006	0.011	<0.001	0.007	0.007	-	-
Lukens 2 1/4Cr-1Mo-0.12C	Lukens	0.12	0.41	0.006	0.003	0.22	0.02	2.27	1.05	0.01	<0.01	<0.01	0.004	0.08	0.008	<0.001	<0.01	0.002	0.003	<0.001	0.015	0.004	-	-

the 9Cr-1Mo materials also contain V and Nb (0.2%V and 0.07%Nb). The carbon content of the modified 9Cr-1Mo material is nominally 0.08%. The remaining two 9Cr-1Mo materials are standard composition and contain no additions of V and Nb. These two 9Cr-1Mo materials differ mainly in carbon content (0.11 and 0.06%C).

Among the bainitic materials the Climax 3Cr-1.5Mo has the highest Cr, Mo and Mn content (3.02, 1.43 and 0.86%), and the lowest carbon content (0.08%C). The Climax 3Cr-1.5Mo material also contains 0.12%V. Other than nominal differences in minor elements, the bainitic 2 1/4Cr-1Mo materials differ in C, S, Si and Ni content. The Lukens 2 1/4Cr-1Mo-0.10C material has the lowest carbon content. The carbon content of the Lukens 2 1/4Cr-1Mo-0.12C and Kawasaki 2 1/4Cr-1Mo-0.12C is the same. The Kawasaki 2 1/4Cr-1Mo-0.12C material has the highest Ni content (0.17%) and the lowest Si content (0.10%). The Lukens 2 1/4Cr-1Mo-0.10C material contains the highest amount of residual sulfur (0.026%).

A description of the as-received form of the investigation materials can be found in Table 4. The Kawasaki 2 1/4Cr-1Mo-0.12C ring forging was the only material not initially in plate form. Test specimens were extracted from the 1/4 thickness location in materials thicker than 1 in (25.4mm) and from the 1/2 thickness location in materials less than 1 in (25.4mm). The longitudinal direction of all test specimens was located in the rolling or forging direction of the as-received material.

Table 4. As-Received Form and Heat Treatment Condition of Investigation Materials.

Type	Material	Form	Heat Treatment
Duplex	12Cr-2Mo-6Ni	5/8 in (16mm) Plate	Normalized 1920°F (1050°C) Tempered 1290°F (700°C)
	12Cr-2Mo-3Ni	5/8 in (16mm) Plate	Normalized 1920°F (1050°C) Tempered 1290°F (700°C)
	12Cr-2Mo-1Ni	5/8 in (16mm) Plate	Normalized 1920°F (1050°C) Tempered 1350°F (730°C)
	9Cr-2Mo-V-Nb	1/2 in (12.7mm) Plate	Normalized 1900°F (1040°C) Tempered 1400°F (760°C)
	9Cr-2Mo	1 in (25.4mm) Plate	Normalized 1740°F (950°C) Tempered 1430°F (780°C)
Martensitic	9Cr-1Mo-V-Nb	3/8 in (9.5mm) Plate	Normalized 1900°F (1040°C) Tempered 1400°F (760°C)
	9Cr-1Mo-0.11C	3/4 in (19mm) Plate	Normalized 1900°F (1040°C) Tempered 1400°F (760°C)
	9Cr-1Mo-0.05C	3/4 in (19mm) Plate	Normalized 1900°F (1040°C) Tempered 1400°F (760°C)
Bainitic	Climax 3Cr-1.5Mo	1/2 in (12.7mm) Plate	Normalized 1750°F (955°C) Tempered 1295°F (700°C)
	Lukens 2 1/4Cr-1Mo-0.10C	6 in (330mm) Plate	Austenitized 1750°F (955°C) Water Quenched Tempered 1270°F (690°C)
	Lukens 2 1/4Cr-1Mo-0.12C	13 in (150mm) Plate	Austenitized 1750°F (955°C) Water Quenched Tempered 1225°F (665°C)
	Kawasaki 2 1/4Cr-1Mo-0.12C	16 in (400mm) Ring Forging	Austenitized 1960°F (1070°C) Water Quenched Tempered 1200°F (650°C)

A description of the prior heat treatment of the investigation materials is also shown in Table 4. The duplex and martensitic materials were normalized at the temperature indicated. The 2 1/4Cr-1Mo materials were austenitized at the indicated temperature followed by water quenching. All materials were tempered at the indicated temperature and allowed to air cool. The duration of all austenitizing and tempering operations was one hour per inch (25.4mm) of thickness.

Continuous-Cooling Transformation Behavior

The CCT behavior of the 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni, 12Cr-2Mo-1Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-0.11C, Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials was determined by the dilatometric technique employing the Gleeble apparatus. This method of evaluation may be directly related to actual weld HAZ transformations.

Test specimens 0.250in diameter by 4in in length (6.35mm by 102mm) were subjected to a simulated weld thermal cycle which duplicates that experienced in the coarse grained HAZ. A dilatometer attachment to the Gleeble was used to determine the on-cooling transformation temperatures. Thermal cycles appropriate to arc welding processes were utilized to construct CCT diagrams which are applicable to actual welding conditions.

The Gleeble

The Gleeble is a device developed at Rensselaer Polytechnic Institute (RPI) which is capable of reproducing the thermal cycle

found at a particular point of interest in the HAZ. The point of interest in this study was the coarse grained HAZ experiencing a peak temperature of 2400°F (1315°C). The coarse grained HAZ of an actual weld is so narrow that it would be almost impossible to determine its transformation behavior. One of the unique features of the Gleeble is that it is capable of reproducing this narrow region in a size suitable for testing.

The Gleeble consists of a high speed, time-temperature resistance-heating control device coupled with a high-temperature tensile testing apparatus. The essential features of the Gleeble employed in this investigation are shown in Figure 13. As shown in the figure, the test specimen was clamped between the two water-cooled copper alloy wedge grips. The grips were contained in platens, one fixed and one movable, which in turn were connected to the secondary power transformer. The specimen was heated by its own resistance to the passage of an electric current supplied by the power transformer, or was cooled by the longitudinal extraction of heat to the water cooled jaws. In a 0.25in (6.34mm) diameter steel specimen, the Gleeble can reproduce any weld HAZ cycle requiring a heating rate of less than 3000°F/sec (1670°C/sec) and a cooling rate at 1000°F (540°C) of less than 350°F/sec (195°C/sec). The control maintains an accuracy to within $\pm 15^{\circ}\text{F}$ ($\pm 8^{\circ}\text{C}$) of the desired temperature.

The temperature of the sample was monitored by a fine wire thermocouple percussion welded to the sample. The electronic

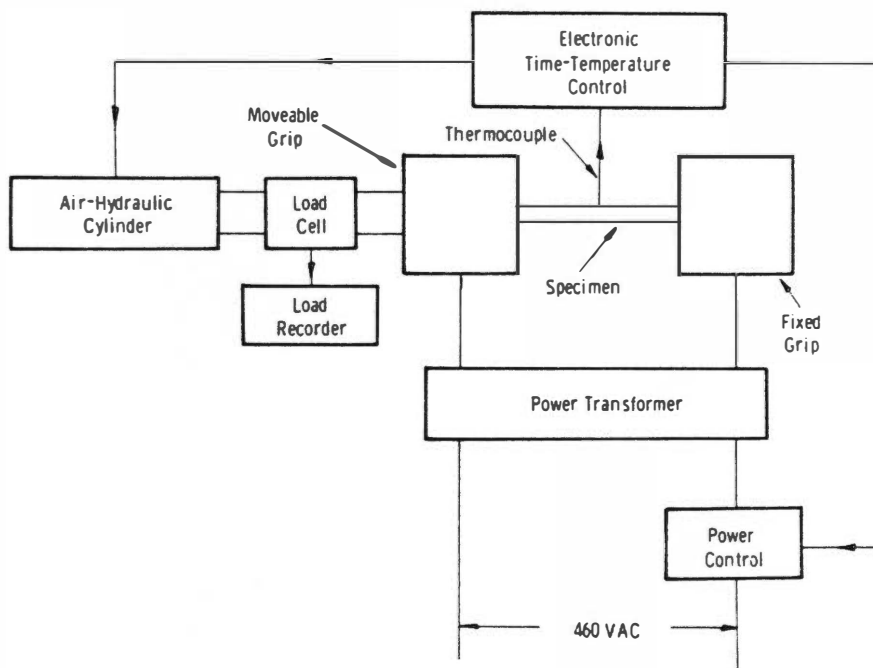


Figure 13. Schematic diagram of the Gleeble apparatus.

Source: Stout, R.D., and Doty, W.D., Weldability of Steels, Welding Research Council, (1978).

controller compares the thermal couple output 120 times per second with the voltage corresponding to the instantaneous temperature of the desired thermal cycle and increases, reduces or interrupts the heating current as required. The thermocouple output was recorded on a direct-developing recording oscillograph.

A schematic representation of the thermal cycles employed in the CCT investigation is shown in Figure 14. These nine thermal cycles are based on a peak temperature of 2400°F (1315°C). The simulated weld microstructure produced at this peak temperature is analogous to the actual microstructure of a weld which is displaced a few grains away from the fusion zone. The first eight thermal cycles which range up to 150 KJ/in (5.9 KJ/mm) produce cooling rates which are similar to those encountered in the last pass of a metal arc weld in a 2in (51mm) Cr-Mo steel plate at 400°F (205°C) preheat. The ninth thermal cycle simulates an electroslog weld (ESW) in a 4in (102mm) 2 1/4Cr-1Mo plate (48 volts, 750 amps, wire speed 6.57in/sec (167mm/sec)) (65). The time required to cool between 800°C and 500°C (8/5t) is listed at the bottom of each cooling curve. The cooling rates employed in this investigation range from the relatively fast cooling rate encountered in a shielded-metal arc (SMA) weld arc strike (8/5t—0.53sec) to the relatively slow cooling rate encountered in an electroslog weld (8/5t—2050).

The Dilatometer

The dilatometer is an attachment to the Gleeble apparatus which measures specimen dilation (contraction or expansion) during rapid

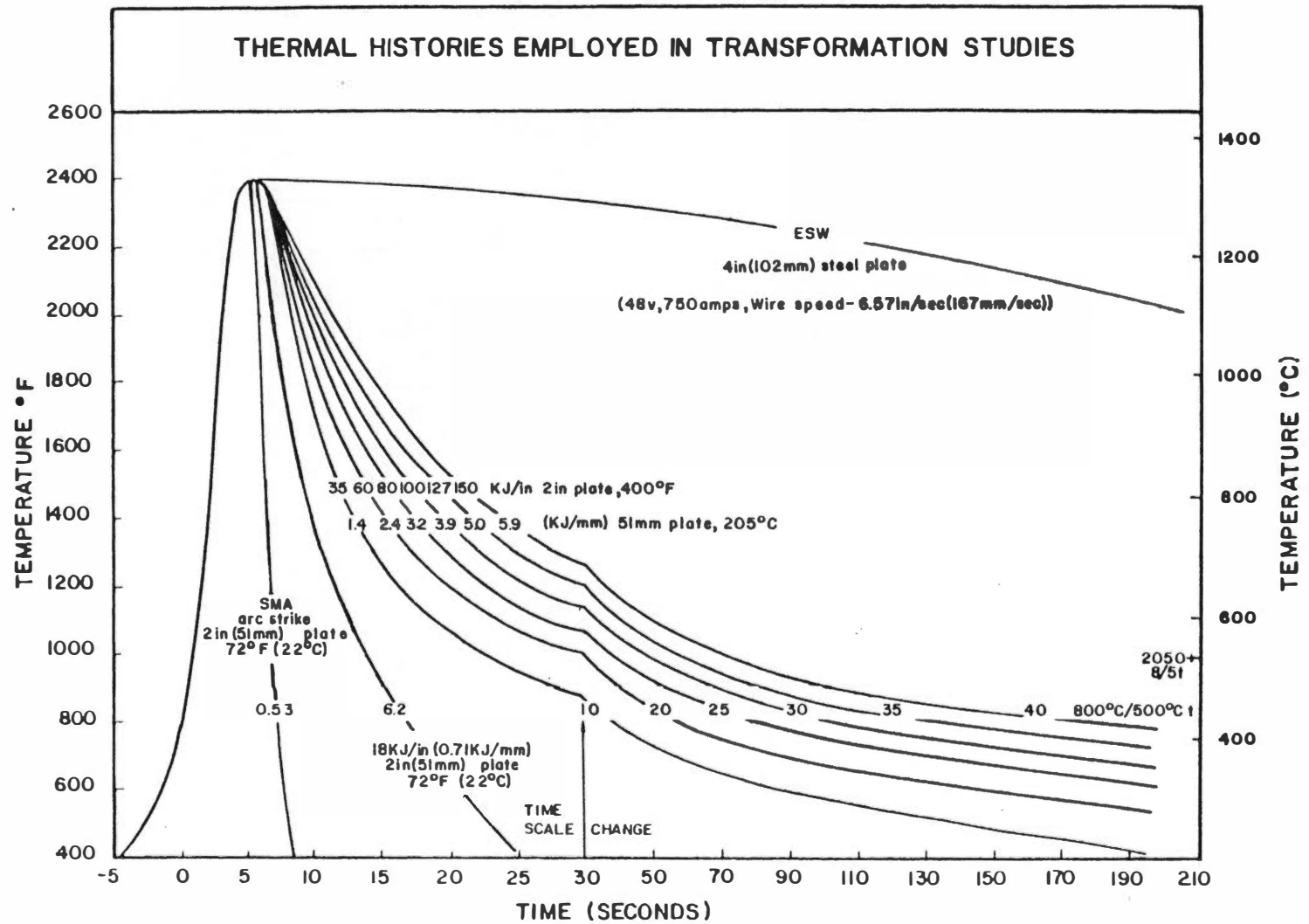


Figure 14. Gleeble thermal cycles employed in transformation determinations.

heating and cooling. As shown in Figure 15, the dilatometer consisted of quartz contact rods connected to a precision rectilinear transducer by a solid-hinge mechanical linkage. The change in volume accompanying an on-cooling phase change was transferred through the hinge to the rectilinear potentiometer which was in turn connected to a precision bridge circuit. The output of the bridge was connected to a multiple channel recording oscillograph.

Both dilation vs. time and temperature vs. time plots were obtained from the multiple channel recording oscillograph. An example of dilation vs. temperature plots constructed from the Gleeble data is shown in Figure 16. The transformation start and stop temperatures were selected as the point where reasonable departure from smooth thermal contraction initiated. The transformation temperatures determined in this manner for the entire series of energy inputs were plotted as temperature vs. log-time from peak temperature, thereby generating the conventional CCT diagram.

Confirming tests were made for each weld thermal cycle investigated. To insure that the same initial microstructure was present in each run, a new specimen was employed. Specimens from different thermal cycles were prepared for metallographic evaluation. The chemical composition of the etchants employed during the course of this investigation are listed in Table 5.

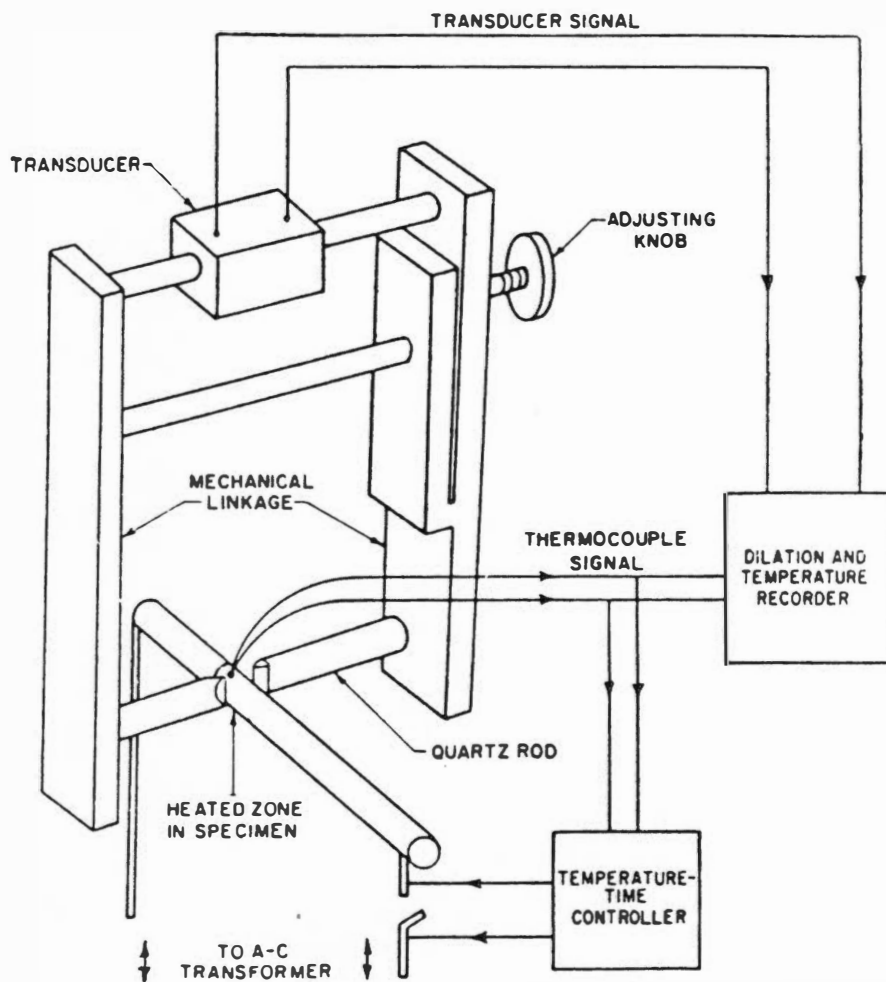


Figure 15. Schematic representation of the Gleeble dilatometer.

Source: Rathbone, A.M., Connor, L.P., and Gross, J.H., "Weldability of a High Strength Alloy Plate Steel With a Minimum Yield Strength of 140 Ksi," Welding J., 43(12), 551s-563s (1964).

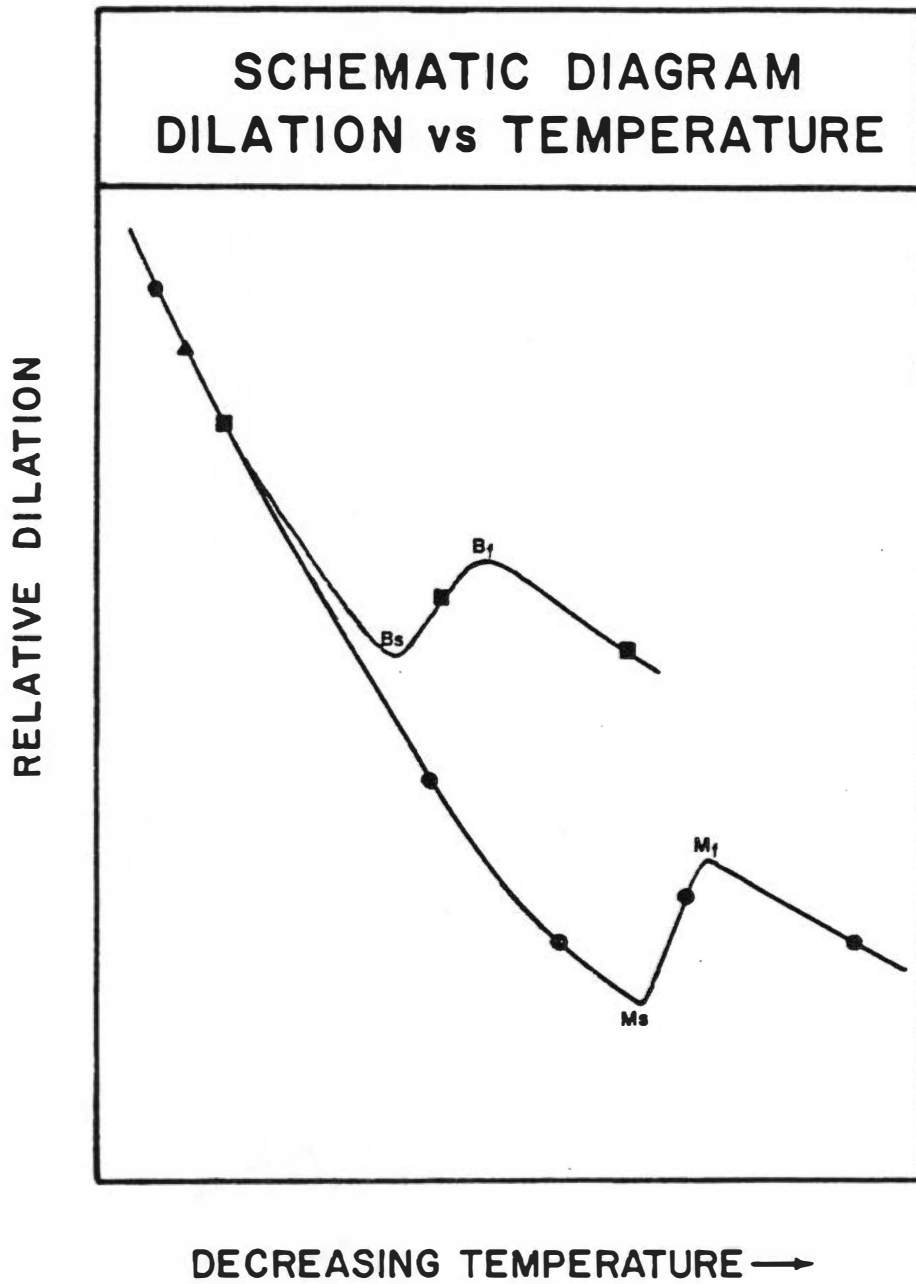


Figure 16. Example of dilation vs. time plot for on-cooling transformation.

Table 5. Etchants for Cr-Mo Steels.

Etchant	Use	Composition	Comments
Modified HCl-Picral	12Cr-2Mo-Ni 9Cr-2Mo, 9Cr-1Mo-V-Nb, 9Cr-1Mo	900ml saturated Picric acid in Methanol 100ml HCl 20ml H ₂ SO ₄ 20g FeCl ₃ 20g CuCl ₂	Dilute Final mixture with 50% Methanol when ready to use.
Modified Wazau's	9Cr-2Mo 9Cr-1Mo-V-Nb 9Cr-1Mo	300ml Methanol 180ml HCl 100ml H ₂ O	Dilute Final mixture with 50% Methanol when ready to use.
Modified 5% Nital	9Cr-2Mo	900ml Methanol 50ml HNO ₃ 100ml H ₂ O 20ml H ₂ SO ₄ 20g Fe ₃ Cl ₃ 20g CuCl ₂ 20ml HCl	Make two solutions, mix when ready to use. Dilute final mixture with 50% Methanol when ready to use.
Picral	2 1/4Cr-1Mo	Methanol saturated with Picric acid	
Nital-Picral	2 1/4Cr-1Mo 3Cr-1.5Mo	925ml Methanol 50ml Picral 25ml HNO ₃	

Stress-Relief Cracking (SRC) Determination

Stress-relief or reheat cracking is intergranular cracking that occurs in the HAZ during exposure of welded assemblies to elevated temperatures produced either by post-weld heat treatment (PWHT) or high temperature service. Cracking can be attributed to the combined effects of embrittlement and strain.

The potential for SRC in the 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni, 12Cr-2Mo-1Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-0.11C, Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials was examined employing the Gleeble. The rapid tensile-test capability of the Gleeble enables it to impose simultaneous stress and temperature conditions similar to those experienced during welding. A schematic representation of the thermal treatment-stress cycle applied in the SRC determination is shown in Figure 17.

In this investigation a thermal cycle simulating an energy input of 60 KJ/in (2.36 KJ/mm) in a 2in (51mm) plate at 400°F (205°C) with a peak temperature of 2400°F (1315°C) is imposed on the test specimen (Figure 18) to produce a microstructure which is representative of the grain coarsened region of the HAZ (this region has been shown to be sensitive to SRC). After the weld thermal cycle the sample is reheated to the PWHT temperature and a stress applied to simulate restraint under PWHT conditions. The PWHT temperatures selected in this investigation reflect temperatures

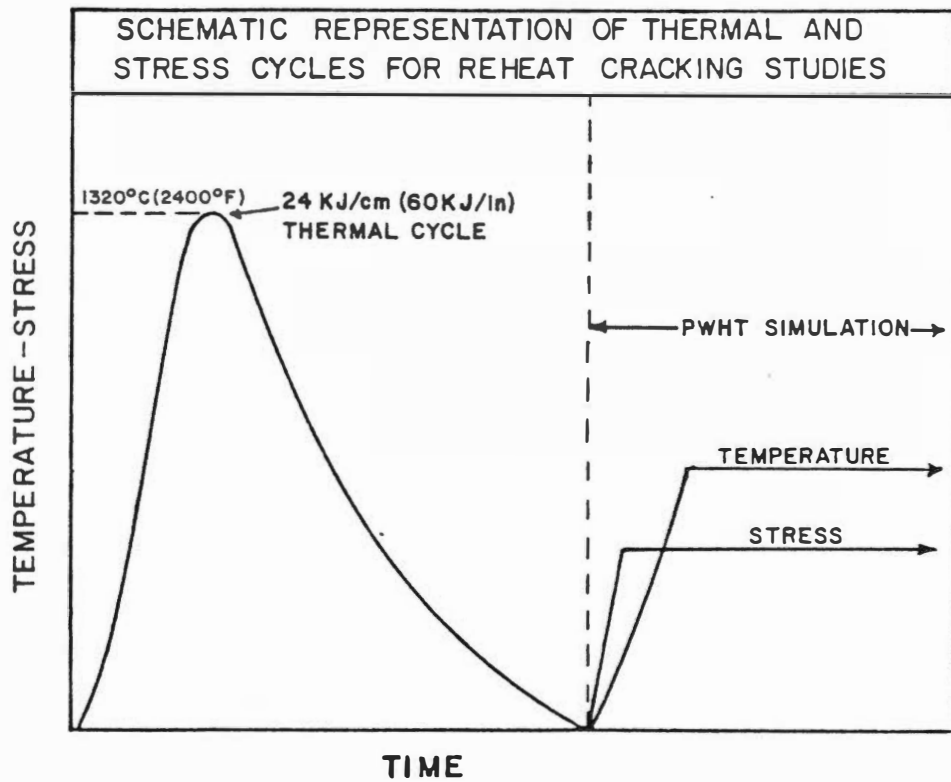


Figure 17. Schematic representation of the thermal treatment-stress cycle employed in SRC determination.

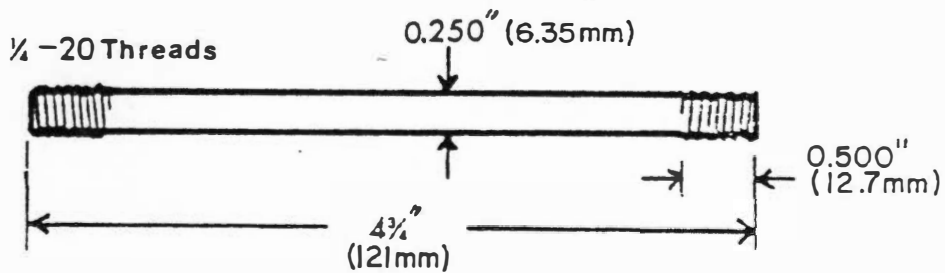


Figure 18. Gleeble sample employed in SRC evaluation.

employed in actual service. The various PWHT temperatures utilized for the various Cr-Mo materials are listed in Table 6.

The stress levels employed in this investigation were selected to cause rupture within 120 minutes. If the specimen is not ruptured in 120 minutes or less the test is discontinued. For comparison purposes one stress level is selected to cause rupture in approximately ten minutes. Both the ductility (reduction in area) and fracture features of the ruptured specimen reflect the SRC sensitivity of the material. The ruptured specimens were retained for further metallographic and fractographic evaluations. This test method has been successfully employed at The University of Tennessee to evaluate the SRC susceptibility of SA508 and SA533.(67).

Hydrogen-Assisted Cracking Evaluation

Hydrogen-assisted cracking (HAC) is low temperature, (below 600°F (315°C) delayed cracking which occurs in the fusion zone or HAZ of a weld due to a tensile stress acting on a susceptible microstructure in which hydrogen is present. The HAC susceptibility of the 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni, 12Cr-2Mo-1Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb (148, 176, 383 and 394), 9Cr-1Mo-0.11C, 9Cr-1Mo-0.06C, Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials was evaluated by the UT-Modified Hydrogen Sensitivity Test (UT-Mod HST). This test is a variation of the RPI augmented strain test (68,69,70) and includes the following modifications:

Table 6. Post-Weld Heat Treatment Temperatures Employed in SRC Evaluation.

Material	PWHT Temperature	
	°F	(°C)
12Cr-2Mo-6Ni	1290	700
12Cr-2Mo-3Ni	1290	700
12Cr-2Mo-1Ni	1290	700
9Cr-2Mo-V-Nb	1350	730
9Cr-2Mo	1320	715
9Cr-1Mo-0.11C	1350	730
Climax 3Cr-1.5Mo	1250	680
	1150	620
Lukens 2 1/4Cr-1Mo-0.12C	1250	680
	1150	620
Lukens 2 1/4Cr-1Mo-0.10C	1250	680
	1150	620
Kawasaki 2 1/4Cr-1Mo-0.12C	1250	680
	1150	620

1. Reduced specimen size.
2. Autogenous GTA weld.
3. Hydrogen introduced directly into weld pool.
4. Materials evaluated on a pass-fail basis.
5. Effect of preheat, hold time and straining temperature determined.
6. Acoustic emission determination of crack initiation.
7. No cutting or polishing steps required for crack observation.

The specimen shown in Figure 19 is used in the UT-Mod HST. The specimen is held in a copper fixture which is preheated to the desired temperature. An autogenous gas-tungsten arc (GTA) weld is made along the length of the sample employing the parameters shown in Table 7. Hydrogen is added directly to the shielding gas, saturating the fusion zone. After welding, the sample is allowed to cool to room temperature and is strained in the fixture shown in Figure 20. The nominal augmented strain on the surface is given by the approximation:

$$\epsilon \approx t/2R \quad (3-1)$$

where ϵ = nominal augmented strain, t = specimen thickness (0.125in (3mm)) and R = bending radius. Die blocks are available to produce strain levels of 1%, 2% and 4%. Susceptibility to cracking is determined by observing if cracks form on the specimen face strained in tension. Cracking is evaluated as a function of preheat, strain level and straining temperature. Two straining

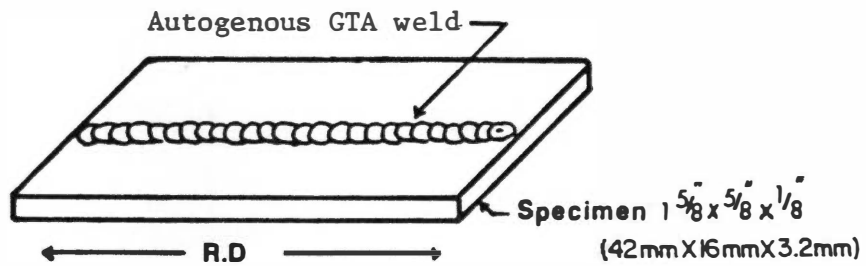


Figure 19. UT-Mod HST specimen.

Table 7. UT-Modified Hydrogen Sensitivity Test Welding Parameters.

Variable	Setting
Current	80-100 amps
Voltage	10 volts
Travel Speed	5ipm (12.7 cmpm)
Shielding Gas	5% H ₂ - 95% Ar
Energy Input	10.8 KJ/in (0.43 KJ/mm)

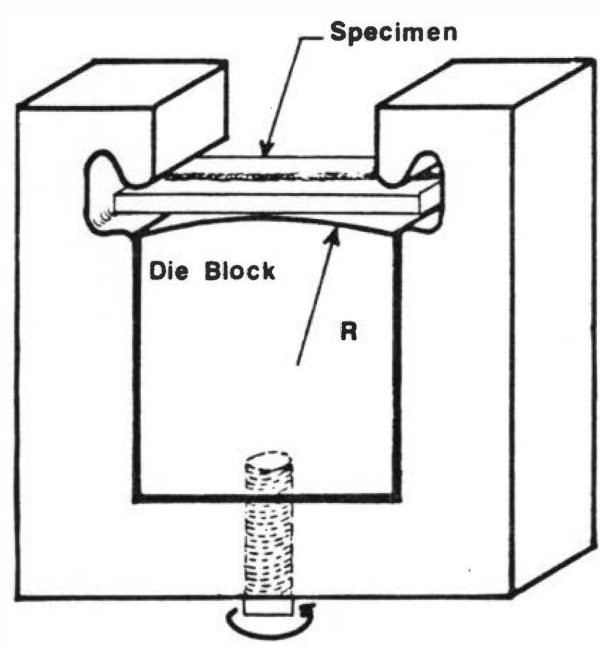


Figure 20. UT-Mod HST straining fixture.

temperatures were employed in this investigation (32°F (0°C) and 75°F (24°C)). The 32°F (0°C) straining temperature is considered to be below the apparent ductile to brittle transformation temperature (DBTT) of the HAZ in most Cr-Mo materials, and provides the most conservative assessment of HAC.

After straining the specimen and fixture are placed in a padded vice and an acoustic emission (AE) sensor from a TEC 912-10 AE instrument is attached to the die block. The output of the AE instrument is recorded on a strip chart recorder. The specimen is held in the fixture for 24 hours. The time for crack initiation is taken as the time at which sustained AE bursts above the background level were recorded. If the specimen does not crack in 24 hours the test is terminated. For any strain level the preheat level at which at least two tests remain crack free is designated as the preheat required to prevent cracking. After testing, the UT-Mod HST specimens are prepared for metallographic examination.

The UT-Modified Hydrogen Sensitivity Test is intended to provide a comparative relative ranking of the HAC sensitivity of Cr-Mo materials based on augmented strain, hydrogen content, microstructure, preheat and welding conditions. Although the augmented strain levels employed in the sensitivity test have no direct correlation with actual measurable restraint encountered during welding, the 4% strain level is analogous to the restraint encountered when welding large thickness multipass joints, while the lower strain levels (2% and 1%) are analogous to restraint produced during the welding of thinner joint sections.

CHAPTER IV

RESULTS AND DISCUSSION

In this weldability evaluation, the CCT, SRC and HAC behavior of the full range of Cr-Mo steels varying in Cr content from 2 1/4% to 12% was determined. In addition to materials which are currently employed in fossil energy applications (2 1/4Cr-1Mo and 9Cr-1Mo), the materials employed in the weldability investigation include a group of new generation Cr-Mo steels which have been formulated to:

(1) enhance resistance to hydrogen attack (3Cr-1.5Mo), (2) produce a concurrent increase in both creep strength and weldability (9Cr-2Mo and 12Cr-2Mo-Ni) and (3) improve creep strength to a level comparable to that of type 304 stainless steel (9Cr-1Mo-V-Nb). The primary objective of the weldability investigation was to provide a relative ranking of the cracking susceptibility of Cr-Mo steels which, from a fabrication viewpoint, facilitates proper alloy selection.

In this study it became apparent that the observed variations in the weld cracking behavior of Cr-Mo steels is related to both compositional and microstructural differences. The predominantly bainitic microstructure which forms in the coarse grained HAZ of the low alloy materials (2 1/4Cr-1Mo and 3Cr-1.5Mo) is more susceptible to SRC than the martensitic or martensitic/ferritic microstructure which forms in the coarse grained HAZ of the higher alloy materials (9Cr-1Mo, 9Cr-2Mo and 12Cr-2Mo-Ni). In contrast, at an equivalent

carbon content the HAC susceptibility of the predominantly bainitic materials is lower than that of the martensitic or martensitic/ferritic materials.

Continuous-Cooling Transformation Behavior of Cr-Mo Steels

The heart of the cracking behavior of Cr-Mo welds is the type and nature of the HAZ CCT transformation which take place over the full scope of welding conditions employed during fabrication. A thorough understanding of the transformation behavior of the HAZ helps to predict the weldability and assists in selecting methods to control certain aspects of SRC and HAC. The transformation behavior of the weld HAZ is determined under the constraints of fast heating rates, short austenitizing times and rapid cooling rates attendant upon welding. During welding the HAZ experiences an entire spectrum of peak temperatures which in turn, yield an entire range of HAZ microstructures. While information on the transformation behavior of Cr-Mo steels is readily available in the form of conventional CCT diagrams, this information is not generally applicable to welding conditions because it is based on slow heating rates, low austenitizing temperatures of substantial duration and relatively slow cooling rates. Most importantly, the conventional CCT diagrams do not address the transformation behavior of the coarse grained HAZ, and it is this region that has been shown to be the weld region most susceptible to SRC and HAC.

This portion of the weldability investigation was initiated because of the lack of available and applicable CCT diagrams which adequately describe the behavior of the coarse grained HAZ. The principal method of the CCT determinations was on-cooling dilatometric measurements using simulated HAZ specimens (Gleeble). The Gleeble method effectively duplicates the austenitizing conditions and cooling rates characteristic of the HAZ in a specimen of a size suitable for direct measurement. The Gleeble method is described in greater detail in Chapter III.

Continuous-Cooling Transformation Behavior of Duplex and Martensitic Cr-Mo Steels (12Cr-2Mo-Ni, 9Cr-2Mo and 9Cr-1Mo)

Evaluation of available conventional CCT diagrams determined for 12Cr-2Mo-1Ni and 9Cr-1Mo indicated that under the most commonly encountered arc welding conditions, martensite would be the primary transformation product encountered (1,28). The use of two simulated HAZ thermal cycles was deemed sufficient to predict the range of CCT behavior because the transformation behavior of the duplex and martensitic materials should be largely independent of the cooling rate.

A summary of the on-heating and on-cooling transformation temperatures for both the duplex and martensitic materials is presented in Tables 8 and 9. The on-heating transformation temperatures (Ac_1 and Ac_3) shown in Table 8 are characteristic of the full range of thermal cycles employed in the CCT evaluation. The Ac_1 and Ac_3 temperatures of the 12Cr-2Mo-Ni materials were found to be a function

Table 8. Summary of On-Heating Transformation Temperatures for 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo.

Material	On-Heating Transformation Temperatures			
	Ac_1		Ac_3	
	°F	(°C)	°F	(°C)
12Cr-2Mo-6Ni	1230	665	1360	740
12Cr-2Mo-3Ni	1340	730	1430	780
12Cr-2Mo-1Ni	1560	850	1780	970
9Cr-2Mo-V-Nb	1620	880	1810	990
9Cr-2Mo	1570	855	1730	945
9Cr-1Mo-0.11C	1680	915	1800	980

Table 9. Summary of On-Cooling Transformation Temperatures and Final Hardness Values for 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo. Simulated Weld Thermal Cycles Based on a Preheat Temperature of 400°F (205°C) in a 2in (51mm) Plate.

Material	On-Cooling Transformation Temperatures														
	35 KJ/in (1.38 KJ/mm)					60 KJ/in (2.36 KJ/mm)					150 KJ/in (5.91 KJ/mm)				
	M _s		M _f		DPH	M _s		M _f		DPH	M _s		M _f		DPH
	°F	(°C)	°F	(°C)		°F	(°C)	°F	(°C)		°F	(°C)	°F	(°C)	
12Cr-2Mo-6Ni	330	165	ND	ND	405	310	155	ND	ND	409	-	-	-	-	-
12Cr-2Mo-3Ni	380	195	ND	ND	390	380	195	ND	ND	398	-	-	-	-	-
12Cr-2Mo-1Ni	520	270	320	160	331	520	270	340	170	336	-	-	-	-	-
9Cr-2Mo-V-Nb	680	360	560	295	227	-	-	-	-	-	670	355	540	280	222
9Cr-2Mo	760	405	630	330	278	-	-	-	-	-	740	395	610	320	280
9Cr-1Mo-0.11C	760	405	620	325	352	-	-	-	-	-	730	390	600	315	360

DPH - Diamond Pyramid Hardness.

ND - No definite M_f temperature observed on-cooling to 75°F (24°C).

of the Ni content. The higher the Ni content, the lower the Ac_1 and Ac_3 temperatures. The Ac_1 and Ac_3 temperatures observed in the 9Cr-2Mo and 9Cr-1Mo materials were higher than those observed in the 12Cr-2Mo-Ni materials. The 9Cr-1Mo-0.11C material displayed Ac_1 and Ac_3 temperatures which were higher than those determined for the 9Cr-2Mo and 9Cr-2Mo-V-Mb materials.

The on-cooling transformation temperatures and final HAZ hardnesses of the 12Cr-2Mo-Ni, 9Cr-2Mo and 9Cr-1Mo materials are tabulated in Table 9. For welding conditions typical of those encountered in fabrication, martensite was the only transformation product revealed during continuous cooling. Continuous cooling to 75°F (24°C) was required to determine the complete on-cooling transformation in the 12Cr-2Mo-Ni materials. A thermal cycle which cools below the preheat temperature (400°F (205°C) to 75°F (24°C)) is analogous to the thermal cycle experienced by the last pass of a weld followed by cooling to ambient temperature from the preheat temperature. For cooling rates corresponding to energy inputs ranging from 35 KJ/in (1.38 KJ/mm) to 150 KJ/in (5.91 KJ/mm) no differences in M_s and M_f temperatures in the 12Cr-2Mo-Ni, 9Cr-2Mo and 9Cr-1Mo materials were observed. This is because the ferrite/bainite "nose" of the CCT diagram is at significantly longer times than those encountered under typical welding conditions and as a result, only martensite forms on-cooling. In the 12Cr-2Mo-3Ni, 12Cr-2Mo-1Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C ferrite forming elements (Cr, Si and Mo) are present in sufficient quantity to

stabilize delta-ferrite and the final HAZ microstructure in martensite with varying amounts of delta-ferrite.

Continuous-cooling transformation behavior of 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni. The most striking feature of the CCT results shown in Table 9 is the differences in the M_s and M_f temperatures for the three different 12Cr-2Mo-Ni materials. The 12Cr-2Mo-Ni materials differ mainly with respect to their Ni and Mn content. The austenite forming influence of Ni, Mn as well as other alloying elements can be mathematically expressed as:

$$Ni.E. = (Ni) + (Co) + 0.5(Mn) + 0.3(Cu) + 30(C) + 25(N) \quad (4-1)$$

where Ni.E. is the Ni equivalent and each element is expressed in weight percent. In Figure 21 the average M_s and M_f temperatures for the three 12Cr-2Mo-Ni materials are shown as a function of their Ni.E. Based on the M_f temperature of the 12Cr-2Mo-1Ni material the anticipated M_f temperatures of the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni materials are displayed in Figure 21.

It is evident that an increase in the Mn and/or Ni content effectively depresses both the M_s and M_f temperatures. The M_f temperature of the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni materials is depressed below the anticipated M_f temperature. In the experimental investigation, no definite transformation finish was detected on cooling to 75°F (24°C). The lack of a transformation finish in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni materials suggests the retention of austenite at 75°F (24°C).

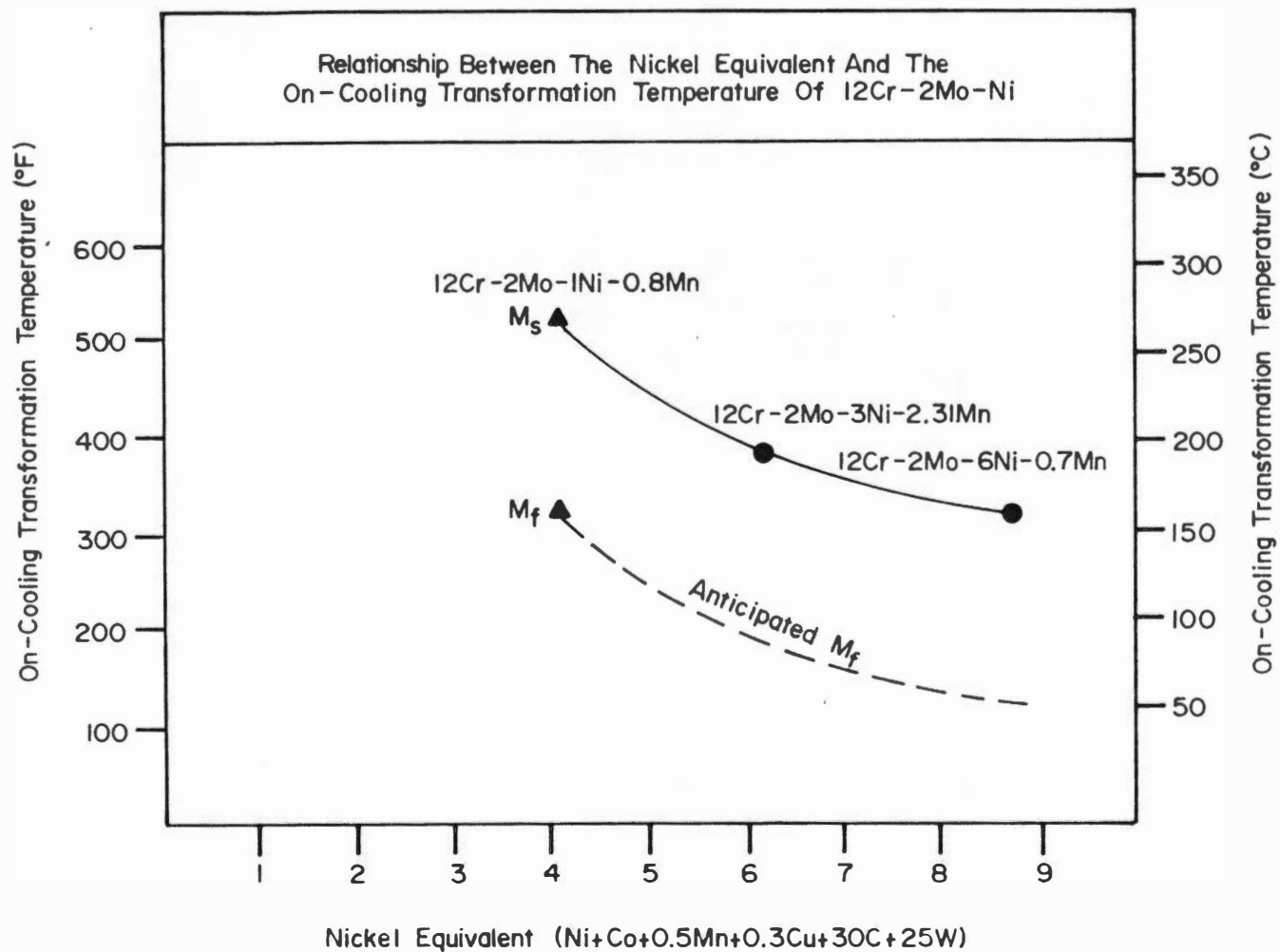


Figure 21. Effect of the nickel equivalent on the on-cooling transformation temperature of 12Cr-2Mo-Ni alloys.

The presence of austenite in the weld HAZ below 300°F (150°C) would increase the HAC susceptibility because the hydrogen present in austenite below this temperature is less able to diffuse out of the HAZ. Austenite which may be present during PWHT may: (1) isothermally transform to ferrite plus carbide, or, (2) upon cooling to ambient temperature transform to "fresh" untempered martensite. The presence of ferrite plus carbide or untempered martensite in the HAZ microstructure would decrease the HAZ toughness.

Microstructure produced on continuous-cooling of 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni. For a specific HAZ peak temperature, the effect of Ni on the microstructure of the 12Cr-2Mo-6Ni material is depicted in Figure 22. Nickel tends to stabilize the gamma loop region of the Fe-Cr phase diagram. Thus, at a peak temperature of 2400°F (1315°C) only austenite is present. Upon cooling, most of the austenite transforms to martensite producing the microstructure shown in Figure 23. The lack of an observable transformation finish suggests that some austenite may be retained in the 12Cr-2Mo-6Ni material. Arwidson and his co-workers (30) employed x-ray diffraction techniques to show that austenite was in fact retained in a continuously-cooled material of a composition similar to the 12Cr-2Mo-6Ni material employed in this study.

Figure 22 also describes the effect of Ni on the extent of the austenite field for the 12Cr-2Mo-3Ni material. A reduction in the Ni content from 6% to 3% reduces the effective size of the gamma loop. For a 2400°F (1315°C) peak temperature two phases (austenite plus

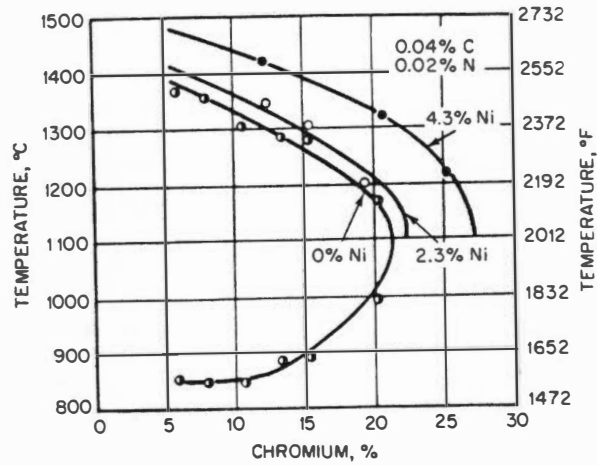


Figure 22. Shift in gamma loop with Ni.

Source: Lovejoy, P.T., "Structure and Constitution of Wrought Martensitic Stainless Steels," Handbook of Stainless Steels, McGraw-Hill, (1977).

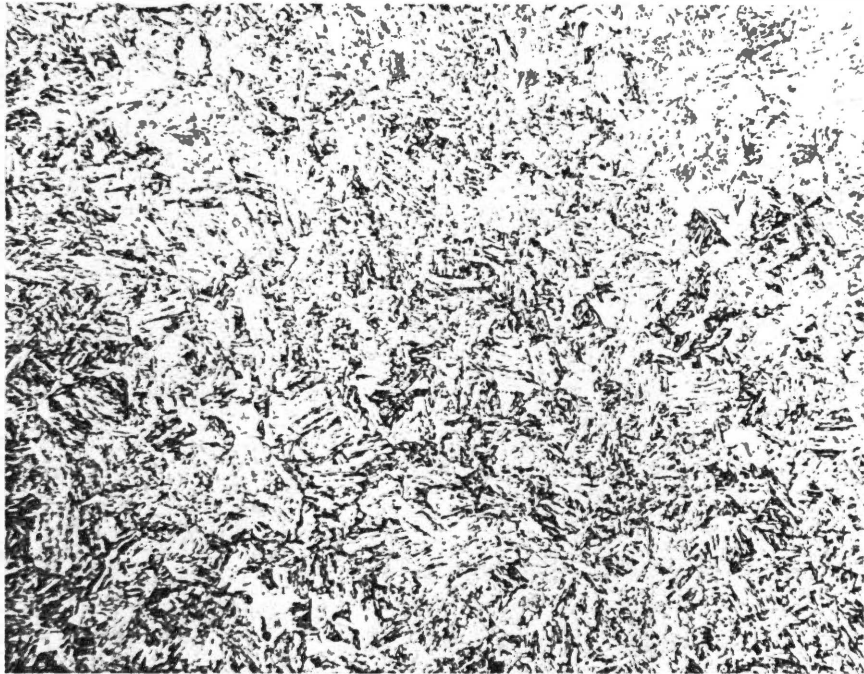


Figure 23. Photomicrograph of 12Cr-2Mo-6Ni Gleeble specimen, 60 KJ/in (2.36 KJ/mm). OLM. 200X. HCl-Picral.

delta-ferrite) are present and during subsequent cooling, most of the austenite transforms to martensite, while the delta-ferrite is retained in the microstructure (Figure 24). Because of the retention of delta-ferrite in the microstructure, the 12Cr-2Mo-3Ni material is softer than the 12Cr-2Mo-6Ni material (376 DPH vs 409 DPH). The lack of an observable transformation finish in the 12Cr-2Mo-3Ni material suggests the presence of retained austenite in this material.

Although the transformation results indicate that austenite is retained in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni HAZ, the presence of austenite in either material was not detected by either optical microscopy or scanning electron microscopy. The lack of metallographic evidence of retained austenite in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni HAZ does not dispute the dilatometric results because in most steels, austenite is only observable by quantitative microscopy when the amount present exceeds 15% (71). Arwidson et al. (30) showed that x-ray diffraction techniques were required to detect retained austenite in a material of a composition similar to the 12Cr-2Mo-6Ni material employed in this investigation. It is apparent that more powerful methods of austenite detection such as direct x-ray comparison or dark field electron microscopy are required to disclose the existence of austenite in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni HAZ.

A decrease in the Ni content from 3% to 1% results in a further reduction in the size of the gamma loop region in the Fe-Cr diagram shown in Figure 22, and as a result, at a peak temperature of 2400°F



Figure 24. Photomicrograph of 12Cr-2Mo-3Ni Gleeble specimen, 60 KJ/in (2.36 KJ/mm). OLM. 200X. HCl-Picral.

(1315°C) there is more delta-ferrite present in the 12Cr-2Mo-1Ni than there is in the 12Cr-2Mo-3Ni. On cooling, the delta-ferrite is retained in the microstructure (Figure 25), and thus results in a lower bulk hardness (336 DPH). The balance of the microstructure in the 12Cr-2Mo-1Ni is fully martensite as evidenced by a clearly detectable M_s and M_f temperature.

Summary of the continuous-cooling transformation behavior of 12Cr-2Mo-Ni. The 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni materials are but a subclass of the Cr-Mo family. The coarse grained HAZ of 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni contains martensite, delta-ferrite and possibly austenite. Simulated HAZ transformation studies of 12Cr-2Mo-Ni materials show that the on-cooling transformation temperatures in the coarse grained region are primarily a function of the Ni content. An increase in the Ni content decreases both the M_s and M_f temperatures. The absence of a detectable M_f temperature in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni materials suggests that austenite is retained in the HAZ of both 12Cr-2Mo-Ni materials. Additional studies utilizing direct x-ray comparison or dark field electron microscopy detection methods are required to verify the presence of austenite in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni HAZ.

In the literature review on HAC, it was shown that the presence of austenite in the HAZ below 300°F (150°C) is undesirable because if austenite transforms to martensite below 300°F (150°C), hydrogen is unable to readily diffuse out of the newly formed martensite, thus

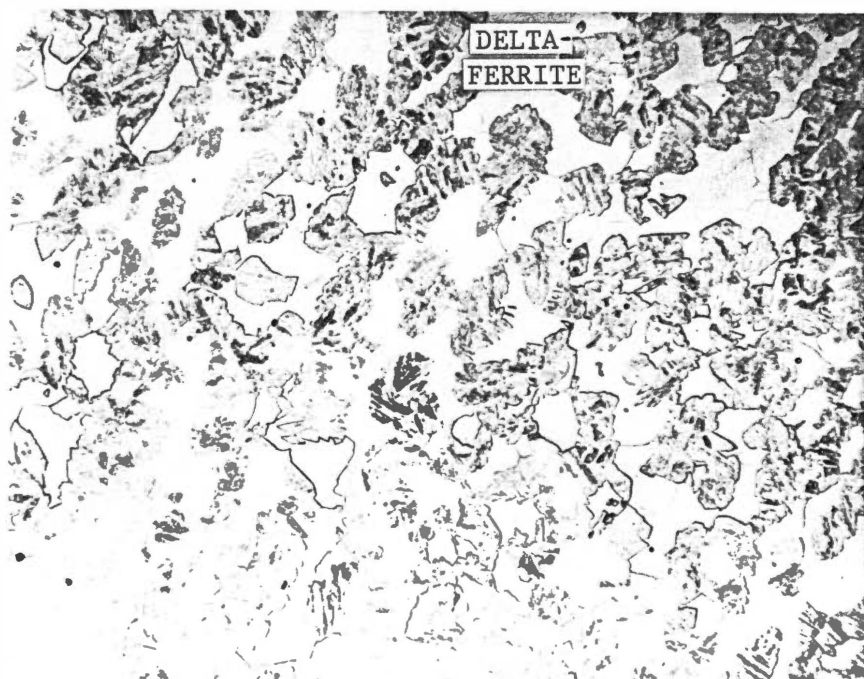


Figure 25. Photomicrograph of 12Cr-2Mo-1Ni Gleeble specimen, 60 KJ/in (2.36 KJ/mm). OLM. 200X. HCl-Picral.

enhancing the tendency for HAC. The possible influence of retained austenite on the HAC susceptibility and subsequently the weldability of 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni is considered in the following discussion on HAC.

Continuous-cooling transformation behavior of 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo. There is no doubt that one of the most important factors in determining the M_s and M_f temperatures in 9Cr-2Mo and 9Cr-1Mo type materials is the carbon content of the austenite. An increase in austenite carbon content results in a decrease in the M_s and M_f temperature. During the weld cycle, the austenite carbon content is a function of peak temperature, time at peak temperature, prior microstructure and bulk (total) carbon content. For the welding thermal cycles employed in this investigation it was found that the microstructure in the coarse grained HAZ is primarily a function of peak temperature (delta-ferrite formation) and austenite carbon content.

It is possible to relate the composition and coarse grained HAZ microstructure of 9Cr-2Mo and 9Cr-1Mo in terms of the Cr.E. equation (Eq. 2-2, p. 20). The Cr.E. is a mathematical expression which relates the ferrite and austenite forming tendencies of various elements in a nominal 9Cr material. If the Cr.E. is less than 10, delta-ferrite would not be expected. If the Cr.E. ranges from 10 to 12, some delta-ferrite would be expected to form at elevated temperature and be sufficiently stable to be retained to room temperature.

If the Cr.E. is greater than 12, large quantities of delta-ferrite would be expected.

A plot of the average M_s and M_f temperatures vs Cr.E. for the 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb(176), 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C materials is shown in Figure 26. The 9Cr-1Mo-V-Nb(176) and 9Cr-1Mo-0.06C materials included in this analysis were part of an earlier CCT investigation at The University of Tennessee (72). The chromium equivalents of the 9Cr-1Mo materials (9Cr-1Mo-V-Nb(176), 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C) range from 9 to 12. The M_s and M_f temperatures of the three 9Cr-1Mo materials increased as the bulk carbon content decreased. For example, the M_s and M_f temperature for the 9Cr-1Mo-0.11C material was 750°F (400°C) and 610°F (320°C), respectively. A decrease in carbon level to 0.06% results in an increase in the M_s and M_f temperature to 800°F (425°C) and 630°F (330°C), respectively.

The chromium equivalents of the two 9Cr-2Mo materials (9Cr-2Mo-V-Nb and 9Cr-2Mo) range from 12 to 17. The M_s and M_f temperature of the 9Cr-2Mo-0.06C material is approximately the same as that of the higher carbon content 9Cr-1Mo-0.11C material. The M_s and M_f temperature for 9Cr-2Mo-V-Nb is approximately 120°F (67°C) lower than the M_s and M_f temperature of 9Cr-2Mo-0.06C. The massive amounts of delta-ferrite in the continuously-cooled microstructure, offset the lower bulk carbon content and in contrast to the 9Cr-1Mo materials, the M_s and M_f temperatures of the 9Cr-2Mo-V-Nb and 9Cr-2Mo materials decrease as the bulk carbon content decrease. The

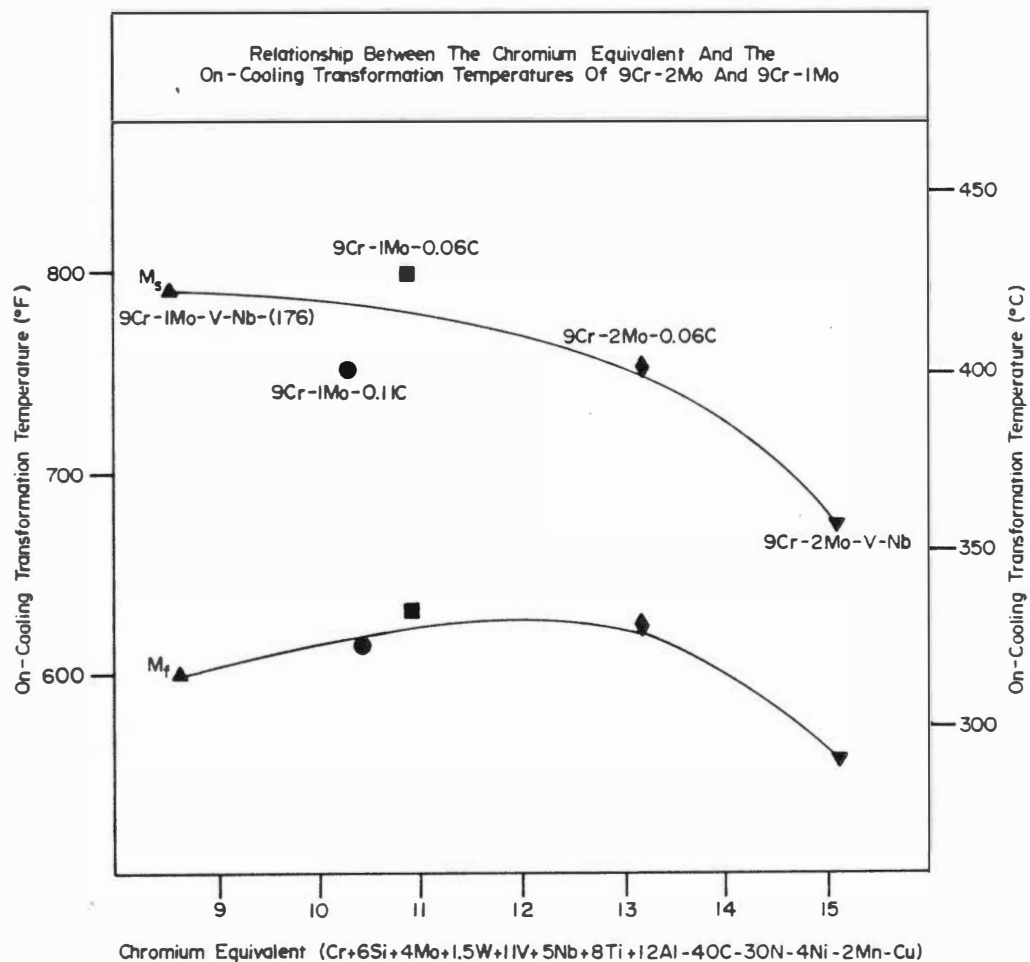


Figure 26. Effect of Cr.E. on M_s and M_f temperatures of 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo.

delta-ferrite present in the coarse grained HAZ acts to partition the carbon to the austenite. The carbon enrichment of the austenite results in the lower M_s and M_f temperatures observed in the 9Cr-2Mo-V-Nb and 9Cr-2Mo.

Microstructures produced on continuous-cooling of 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo. The HAZ microstructure of 9Cr-1Mo-V-Nb(176), 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C consists of martensite and islands of delta-ferrite (Figures 27-29) and as expected the M_s and M_f temperatures of the three 9Cr-1Mo materials are related to the carbon content; for example, an increase in bulk carbon content results in a decrease in on-cooling transformation temperatures.

The delta-ferrite observed in the 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C materials (Figures 28 and 29) formed in the same manner as the delta-ferrite observed in the 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni materials (Figures 24 and 25, pp. 80 and 82). For the 2400°F (1315°C) peak temperature both austenite and delta-ferrite are present. On cooling the austenite transforms to martensite, while the delta-ferrite is retained in the final microstructure. The 9Cr-1Mo-V-Nb(176) material is richer in austenite stabilizing elements than ferrite stabilizing elements, thus this material is fully austenitic at the peak temperature (2400°F (1315°C)). On cooling, the austenite completely transforms to martensite producing the microstructure shown in Figure 27.

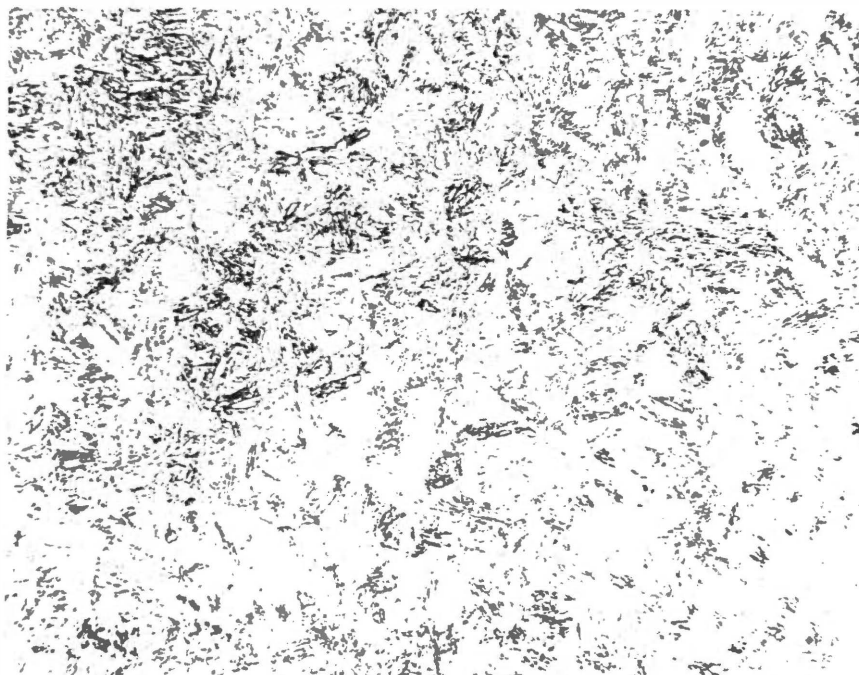


Figure 27. Photomicrograph of 9Cr-1Mo-V-Nb (176) Gleeble specimen, 18 KJ/in (0.71 KJ/mm). OLM. 200X. Modified Wazau's.

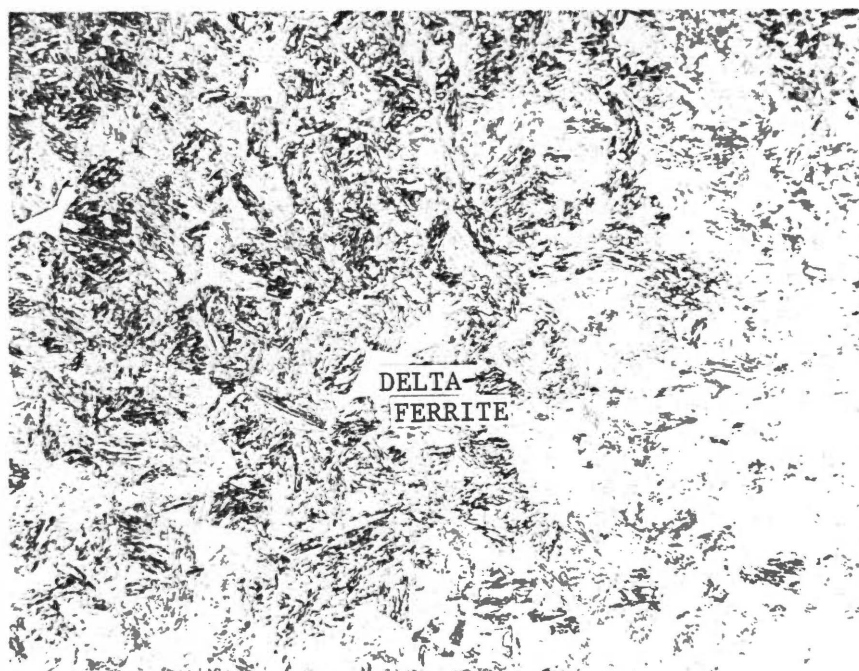


Figure 28. Photomicrograph of 9Cr-1Mo-0.11C Gleeble specimen, 150 KJ/in (5.91 KJ/mm). OLM. 200X. HCl-Picral.



Figure 29. Photomicrograph of 9Cr-1Mo-0.06C Gleeble specimen, 18 KJ/in (0.71 KJ/mm). OLM. 200X. Modified Wazau's.

The coarse grained HAZ microstructure of the 9Cr-2Mo-V-Nb and 9Cr-2Mo materials consists of massive amounts of delta-ferrite and martensite (Figures 30 and 31). It was shown earlier that the massive amounts of delta-ferrite influence the M_s and M_f temperature because the solubility of carbon in delta-ferrite is considerably lower than the solubility of carbon in austenite and as a result, carbon is rejected from the delta-ferrite into the austenite, increasing the carbon concentration in the austenite. The temperature at which the austenite transforms to martensite is effectively lowered by carbon-enrichment. This explains why the 9Cr-2Mo-V-Nb and 9Cr-2Mo materials have M_s and M_f temperatures which are lower than those predicted by the bulk carbon content.

The HAZ microstructure of the 9Cr-2Mo material like that of the 12Cr-2Mo-Ni, 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C materials is the result of a peak temperature excursion into the delta-ferrite plus austenite two phase region. The 9Cr-2Mo is richer with respect to ferrite forming elements than the 12Cr-2Mo-Ni, 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C materials; thus more delta-ferrite is retained in the martensite plus delta-ferrite HAZ microstructure (approximately 30%). The 9Cr-2Mo-V-Nb is richer with respect to ferrite forming elements than the 9Cr-2Mo. For the peak temperature (2400°F (1315°C)), the microstructure is fully ferritic. Although not detected by the Greeble, upon cooling, austenite apparently forms along the ferrite grain boundaries and extends in a Widmanstätten fashion into the ferrite grains. During further cooling, the austenite which formed



Figure 30. Photomicrograph of 9Cr-2Mo-V-Nb Gleeble specimen, 150 KJ/in (5.91 KJ/mm). OLM. 200X. HCl-Picral.



Figure 31. Photomicrograph of 9Cr-2Mo Gleeble specimen,
150 KJ/in (5.91 KJ/mm). OLM. 200X. HCl-Picral.

from the delta-ferrite transforms to acicular martensite thus, the martensite forms an almost continuous network along the ferrite grain boundaries (Figure 30). The martensite plus ferrite structure as observed in the 9Cr-2Mo-V-Nb simulated coarse grained HAZ is similar to the 12Cr-2Mo HAZ microstructure described by Wada (28).

Martensite hardness determination for 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo-V-Nb(176). The hardness of the martensite in Cr-Mo steels for Cr contents ranging between 9-12% Cr is primarily a function of carbon content (73). Partitioning of carbon to austenite and subsequent enrichment of the martensite in the 9Cr-2Mo-V-Nb and 9Cr-2Mo materials should be accompanied by an increase in martensite hardness. The martensite hardness levels of the 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo-V-Nb(176) materials are shown in Figure 32. The martensitic 9Cr-1Mo-V-Nb(176) material is included in this analysis as a base line reference because the HAZ microstructure for a peak temperature of 2400°F (1315°C) contains no delta-ferrite. Although the nominal carbon content of the 9Cr-2Mo material is lower than that of the 9Cr-1Mo-V-Nb(176) material (0.06%C vs 0.08%C), the martensite hardness of these two materials is approximately the same (420 DPH). The similarity in the martensite hardness of the 9Cr-2Mo and 9Cr-1Mo-V-Nb(176) materials suggests that the 30% delta-ferrite present in the 9Cr-2Mo-V-Nb HAZ enriches the carbon content of the martensite to approximately the same level as the martensite found in the 9Cr-1Mo-V-Nb(176) material. Even though the carbon content of 9Cr-2Mo and 9Cr-1Mo-V-Nb(176) is twice that of 9Cr-2Mo-V-Nb

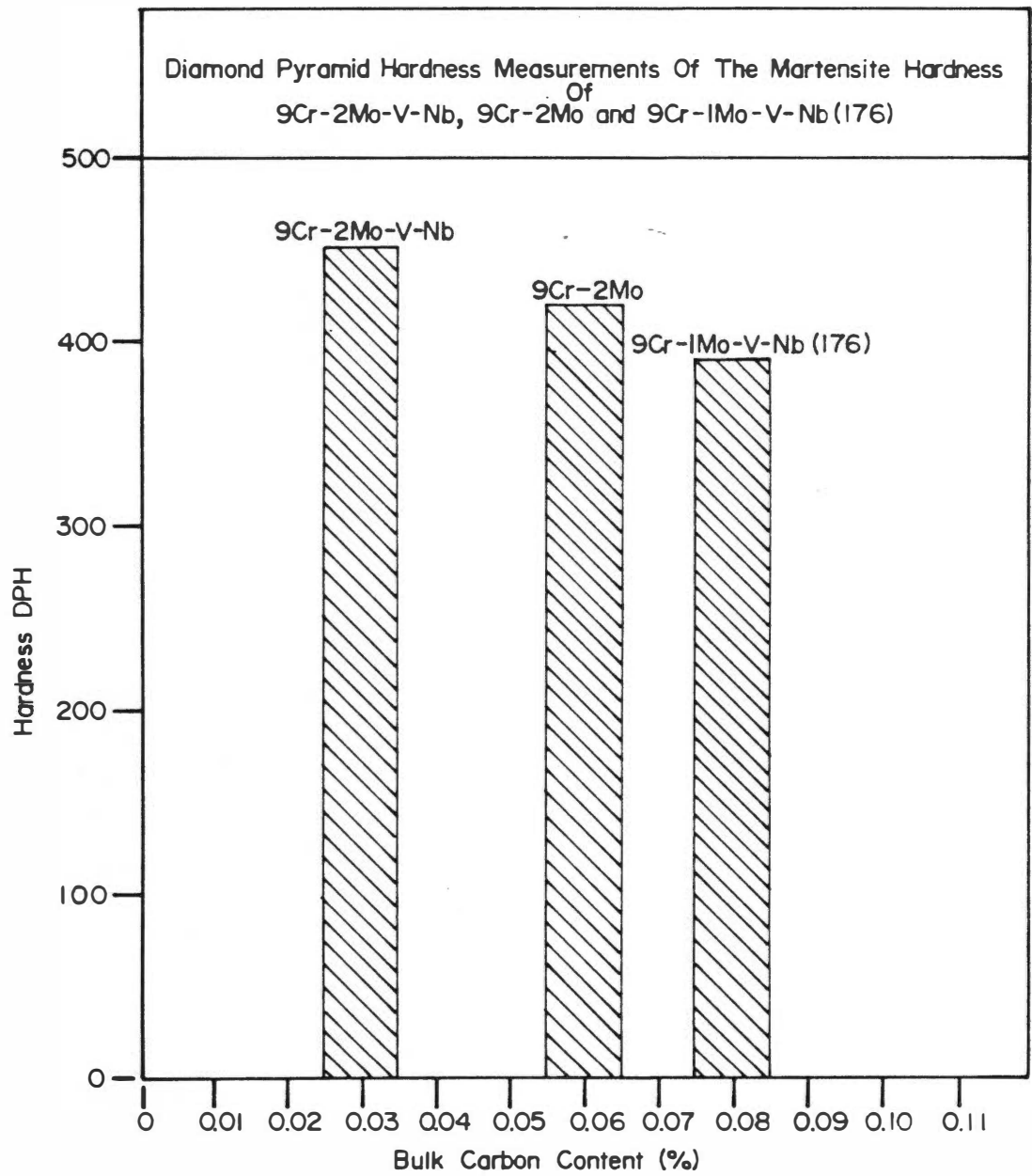


Figure 32. Histogram of martensite hardness in 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo-V-Nb(176).

(0.06%C and 0.08 vs 0.03%C), the martensite found in the 9Cr-2Mo-Nb is harder than that found in the 9Cr-2Mo and 9Cr-1Mo-V-Nb(176) materials (440 DPH vs 420 DPH). These data show that there is undoubtedly extensive carbon enrichment of the martensite in the 9Cr-2Mo-V-Nb material which follows from the presence of approximately 80% delta-ferrite.

Summary of the continuous-cooling transformation behavior of 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo. Like the 12Cr-2Mo-Ni materials, the 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo materials are a subclass of the Cr-Mo steel family. Similarly, the coarse grained HAZ of 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo contains both martensite and delta-ferrite. The amount of delta-ferrite present in the HAZ of 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo can be expressed in terms of the Cr.E. equation. The Cr.E. equation is a mathematical expression of the ferrite forming tendencies of various alloying elements. An increase in the Cr.E. reflects an increase in the amount of delta-ferrite retained in the HAZ microstructure.

The small amounts of delta-ferrite present in the HAZ of the 9Cr-1Mo type materials has little effect on the on-cooling transformation behavior. The on-cooling transformation temperatures in the 9Cr-1Mo type materials are primarily dependent on carbon content. An increase in the bulk carbon content reduces the M_s and M_f temperature. In contrast, the on-cooling transformation temperatures of the 9Cr-2Mo-V-Nb and 9Cr-2Mo materials decrease in spite of a decrease in the bulk carbon content. Hardness measurements of the

two 9Cr-2Mo materials show a level of hardness which is higher than that expected based on the respective bulk carbon contents. The increase in hardness suggests an increase in the martensite carbon content. The increase in the martensite carbon content results from the massive amounts of delta-ferrite (30 to 80%) present in the two 9Cr-2Mo materials at elevated temperature. Carbon is rejected into the surrounding austenite during the HAZ thermal cycle because of the low solubility of carbon in delta-ferrite. On cooling, the carbon enriched austenite transforms to martensite at a lower temperature than expected based on the bulk carbon content.

In the literature review of HAC, it was shown that a localized increase in the carbon content of martensite may increase the HAC susceptibility of the martensite. The presence of a continuous network of carbon-enriched martensite along prior delta-ferrite grain boundaries can result in an increase in HAC susceptibility as well as a decrease in HAZ toughness. The microstructure resulting from the weld HAZ exposure helps explain the HAC behavior described in later sections.

Continuous-Cooling Transformation Behavior Determination of 3Cr-1.5Mo and 2 1/4Cr-1Mo

CCT diagrams appropriate to the continuous cooling conditions encountered during arc welding of Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C and Kawasaki 2 1/4Cr-1Mo-0.12C are presented in Figures 33-35. HAZ thermal cycles ranging from 18 KJ/in (0.71 KJ/mm) to 150 KJ/in (5.91 KJ/mm) were utilized in the CCT determination. A

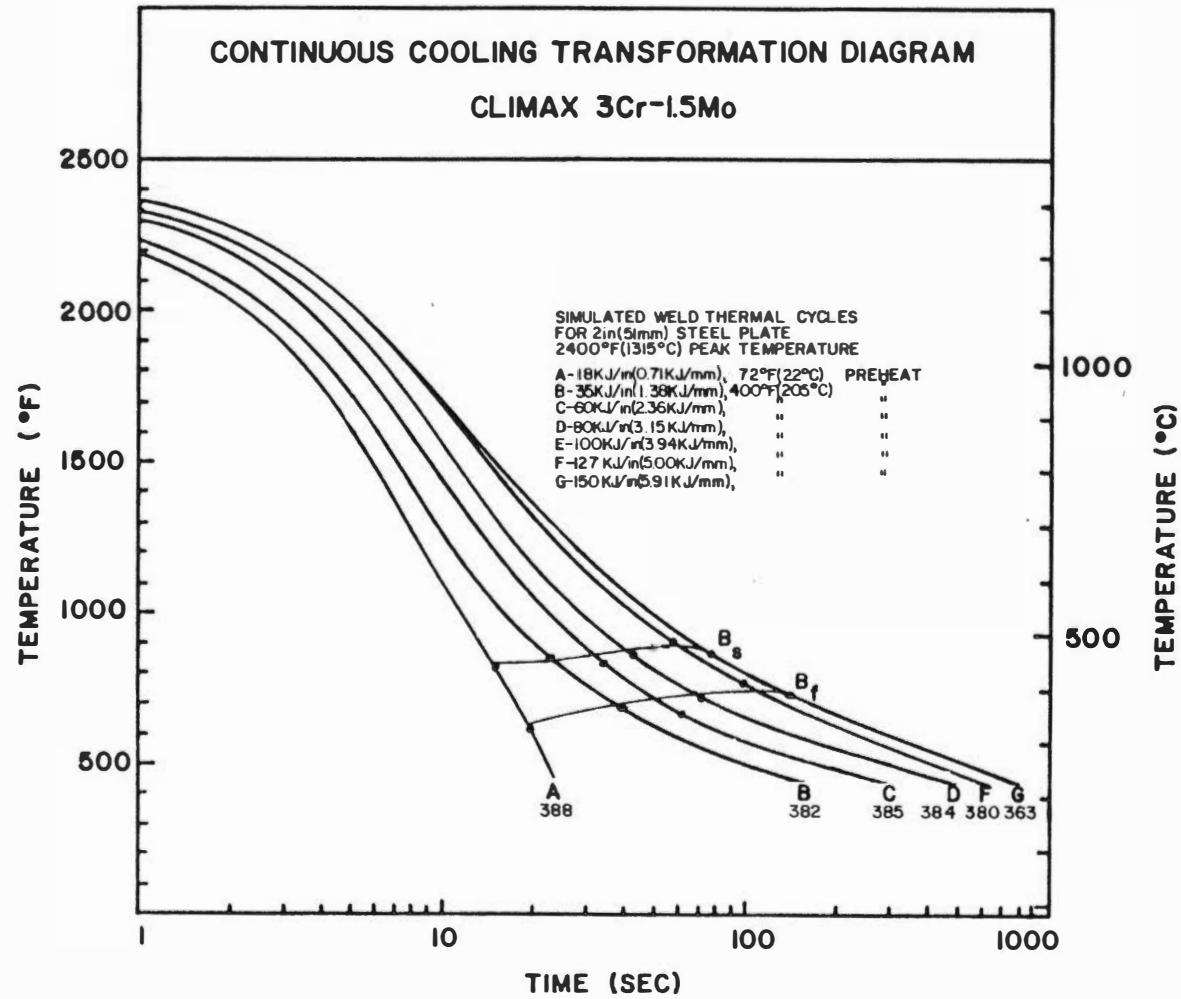


Figure 33. Continuous-cooling transformation diagram for Climax 3Cr-1.5Mo.

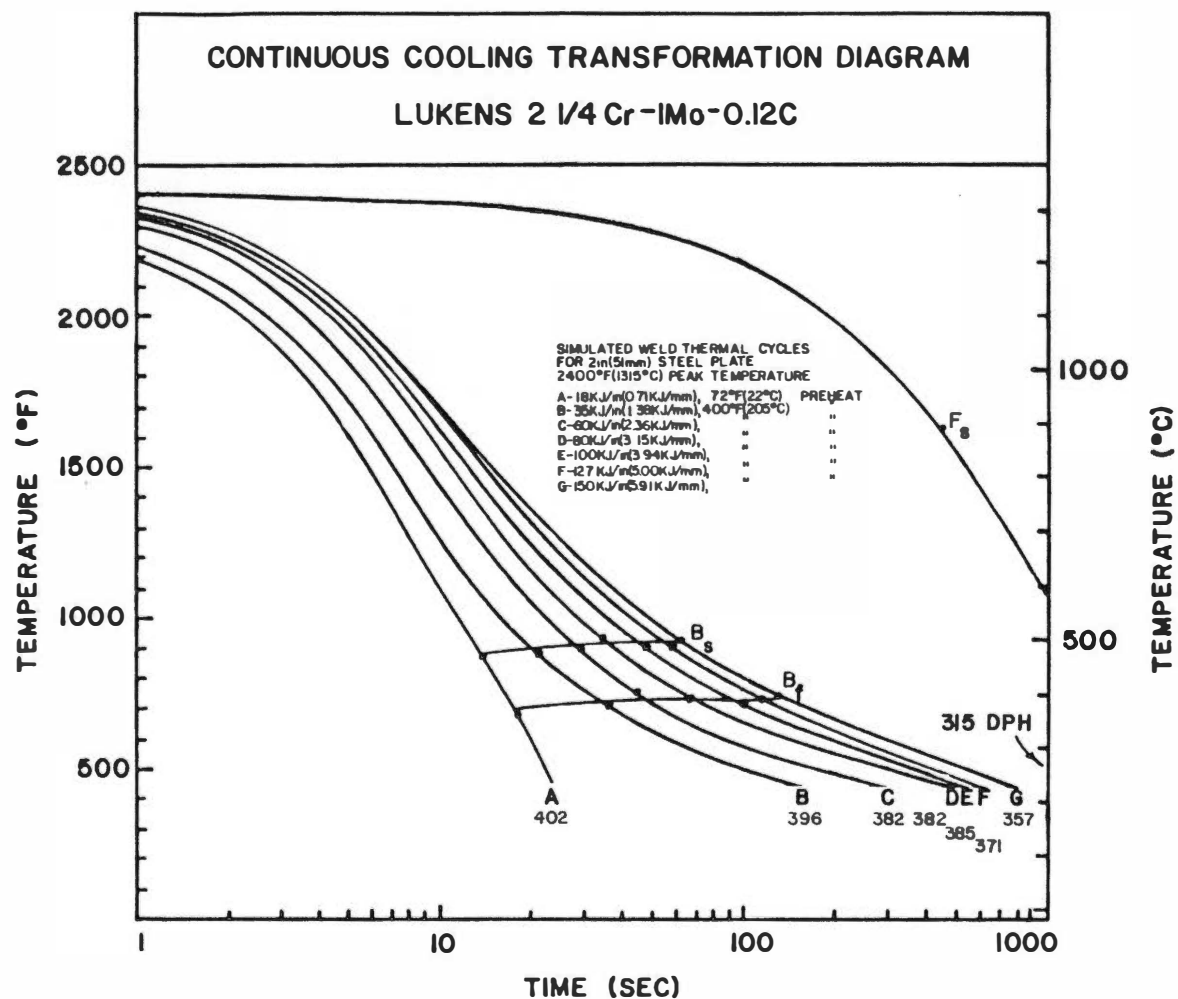


Figure 34. Continuous-cooling transformation diagram for Lukens 2 1/4Cr-1Mo-0.12C.

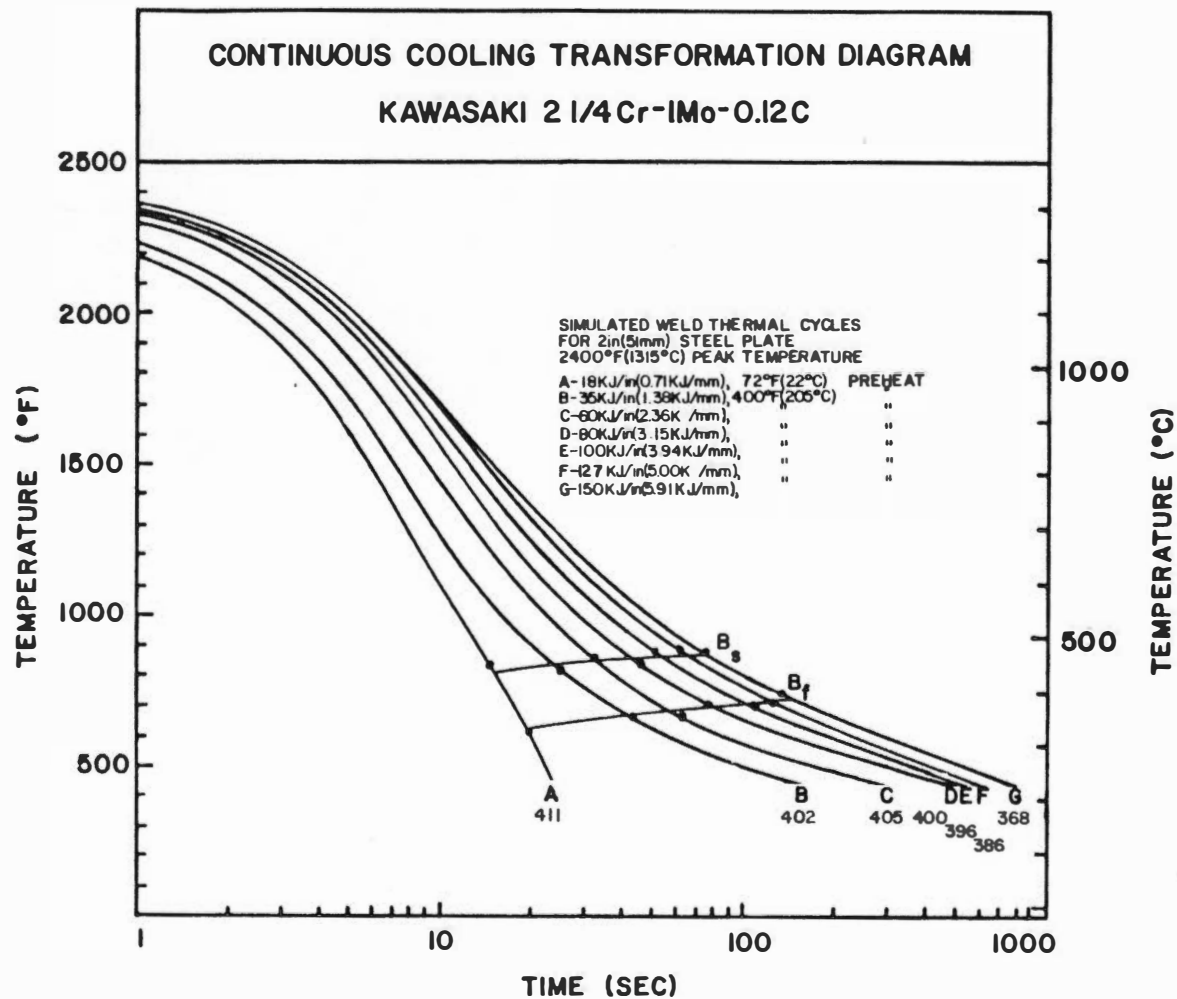


Figure 35. Continuous-cooling transformation diagram for Kawasaki 2 1/4Cr-1Mo-0.12C.

thermal cycle which simulates an electrosag weld (ESW) in a 4in (102mm) plate (48 volts, 750 amps wire speed 6.57 in/sec (167 in/mm) was employed to locate the position of the ferrite "nose" in the Lukens 2 1/4Cr-1Mo-0.12C material. Because the Lukens 2 1/4Cr-1Mo-0.12C, Kawasaki 2 1/4Cr-1Mo-0.12C and Climax 3Cr-1.5Mo materials are similar in composition, the position of the ferrite "nose" in each material should be approximately the same.

Continuous-cooling transformation diagrams of 3Cr-1.5Mo and 2 1/4Cr-1Mo. Bainite was the primary austenite transformation product observed upon cooling of the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials under welding conditions. From Figures 33-35, it is evident that the temperatures at which the bainite transformation both begins and ends increases with decreasing HAZ cooling rates. The B_s and B_f temperatures determined for the Kawasaki 2 1/4Cr-1Mo-0.12C material were 20 to 60°F (11 to 33°C) lower than the B_s and B_f temperatures determined for the Lukens 2 1/4Cr-1Mo-0.12C material. The Kawasaki 2 1/4Cr-1Mo-0.12C material is slightly richer in Cr, Mo and Mn than the Lukens 2 1/4Cr-1Mo-0.12C material. The alloying elements C, Cr, Mo and Mn act to depress the B_s and B_f temperatures, carbon being the most potent. The lower B_s and B_f temperatures observed for the Kawasaki 2 1/4Cr-1Mo-0.12C material may result from the richer alloy content.

The B_s and B_f temperatures observed for the Kawasaki 2 1/4Cr-1Mo-0.12C material are similar to the B_s and B_f temperatures determined for the Climax 3Cr-1.5Mo material. Comparing the alloy

contents of these two materials, the Climax 3Cr-1.5Mo material is richer in Cr, Mo and Mn, while the Kawasaki 2 1/4Cr-1Mo-0.12C material is richer in carbon (0.08% vs 0.12%). Apparently, the net effect of alloy content on the transformation behavior of the Climax 3Cr-1.5Mo and Kawasaki 2 1/4Cr-1Mo-0.12C materials is approximately the same because there is little difference in the B_s and B_f temperatures.

For comparable cooling rates, the B_s and B_f temperatures determined for the Lukens 2 1/4Cr-1Mo-0.12C material are within $\pm 50^\circ\text{F}$ (28°C) of the B_s and B_f temperatures determined for 2 1/4Cr-1Mo (Ishiguro et al. (74)). For comparable cooling rates, the B_s and B_f temperatures determined for Climax 3Cr-1.5Mo are within $\pm 40^\circ\text{F}$ (22°C) of the B_s and B_f temperatures determined for 3Cr-1.5Mo (Wada et al. (75)).

Microstructure produced on continuous cooling of 3Cr-1.5Mo and 2 1/4Cr-1Mo. The effect of cooling rate on the HAZ microstructure of the Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C and Kawasaki 2 1/4Cr-1Mo-0.12C materials can be observed in Figures 36-44. An increase in the energy input decreases the cooling rate. Upon transformation, slower cooling rates allow additional time for the growth of ferrite laths resulting in an increase in lath spacing. The increase in lath spacing results in a coarser appearance of the bainitic microstructure. The coarsening of bainite at slower cooling rates has also been observed in prior transformation investigations of 2 1/4Cr-1Mo (60,61).



Figure 36. Photomicrograph of Climax 3Cr-1.5Mo Gleeble specimen, 35 KJ/in (1.38 KJ/mm). OLM. 200X. Nital-Picral.

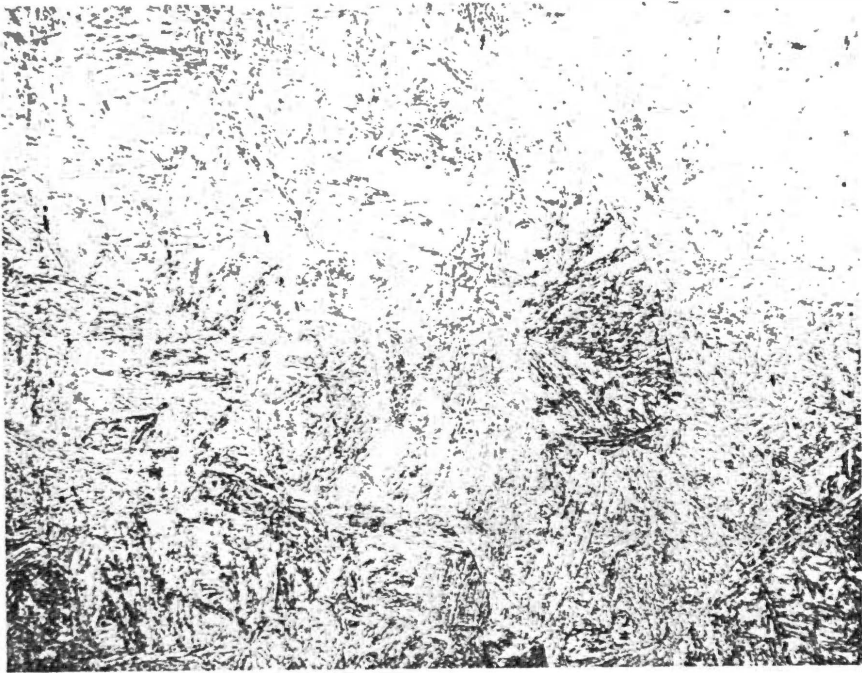


Figure 37. Photomicrograph of Climax 3Cr-1.5Mo Gleeble specimen, 150 KJ/in (5.91 KJ/mm). OLM. 200X. Nital-Picral.



Figure 38. Photomicrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, 35 KJ/in (1.38 KJ/mm). OLM. 200X. Nital-Picral.



Figure 39. Photomicrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, 150 KJ/in (5.91 KJ/mm). OLM. 200X. Nital-Picral.

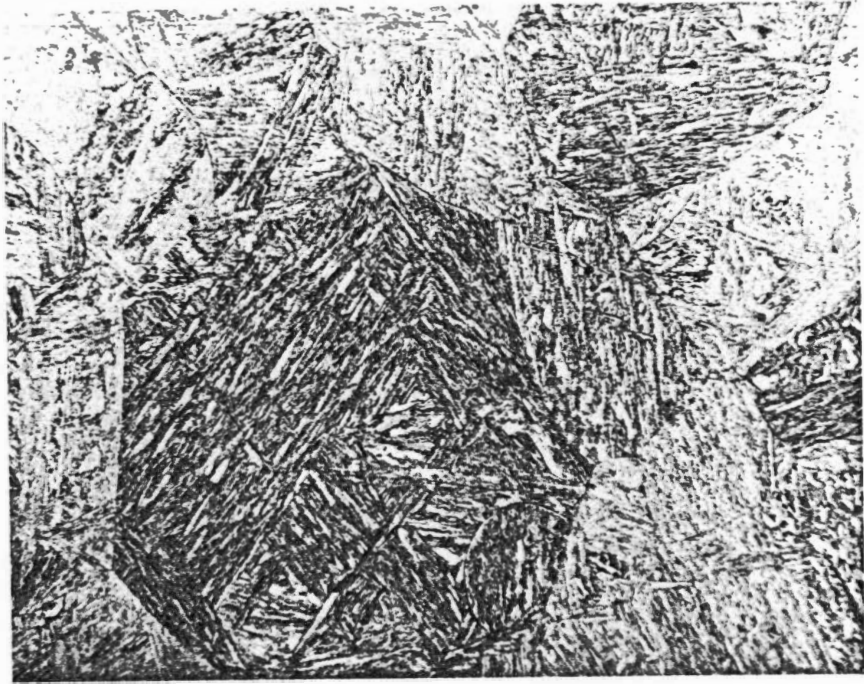


Figure 40. Photomicrograph of Kawasaki 2 1/4Cr-1Mo-0.12C Gleeble specimen, 35 KJ/in (1.38 KJ/mm). OLM. 200X. Nital-Picral.

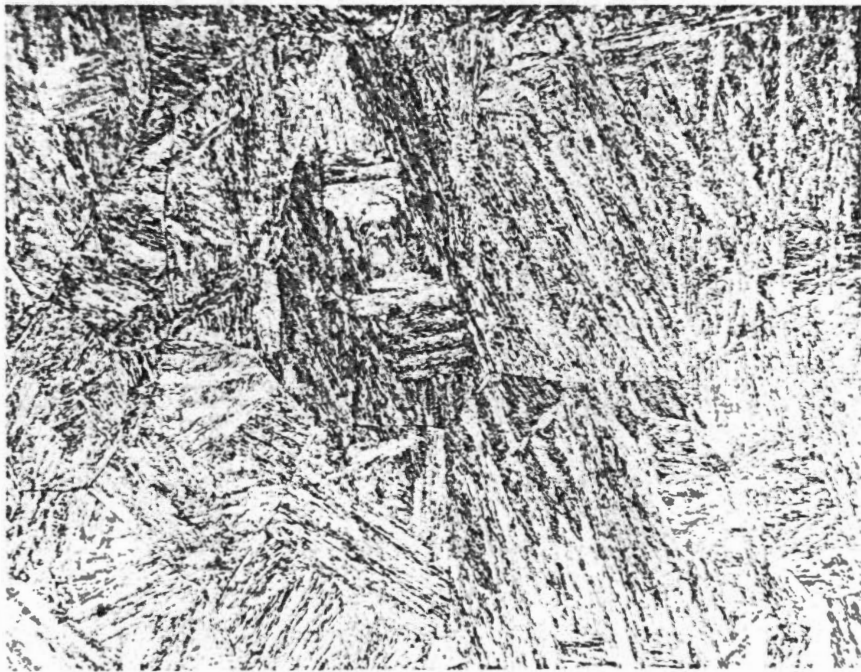


Figure 41. Photomicrograph of Kawasaki 2 1/4Cr-1Mo-0.12C Gleeble specimen, 150 KJ/in (5.91 KJ/mm). OLM. 200X. Nital-Picral.

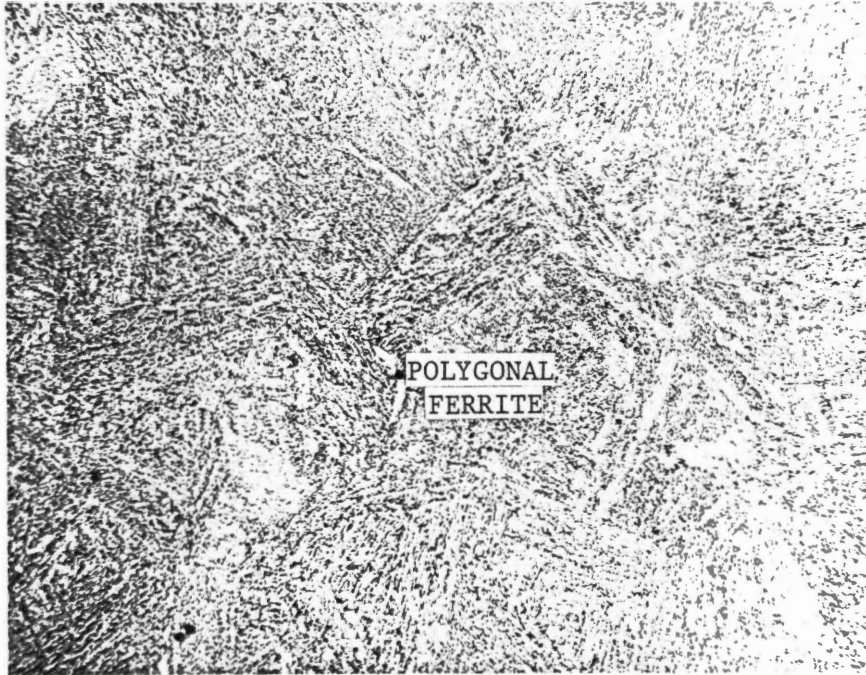


Figure 42. Photomicrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, ESW. OLM. 200X. Nital-Picral.

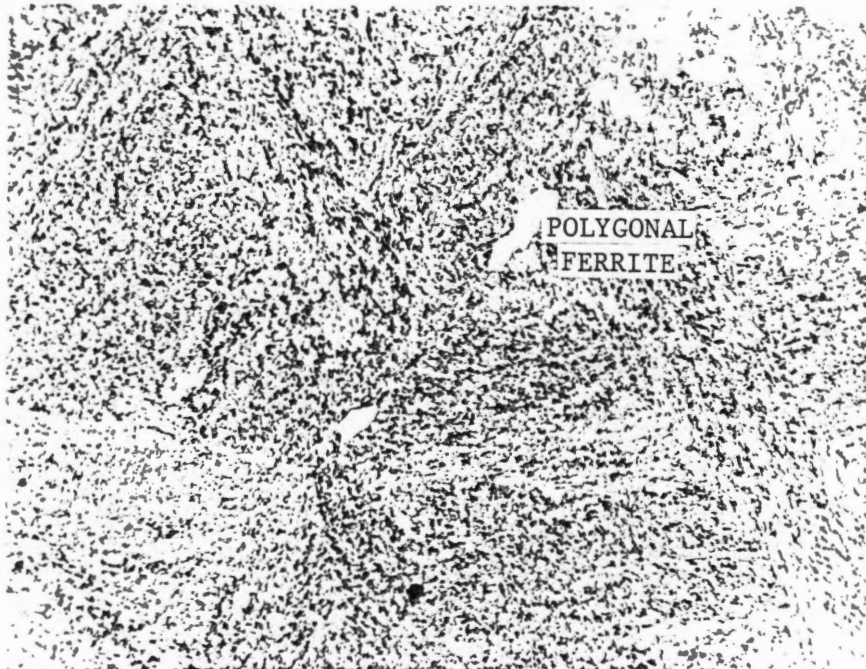


Figure 43. Photomicrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen showing polygonal ferrite, ESW. OLM. 500X. Nital-Picral.

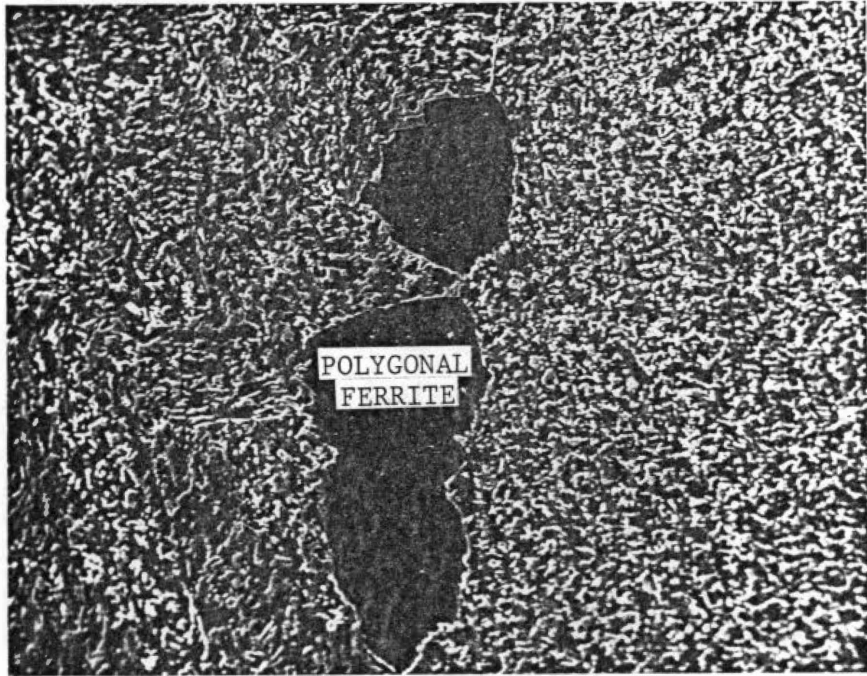


Figure 44. Scanning electron micrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen showing polygonal ferrite, ESW. SEM. 1000X.

For the ESW thermal cycle, the cooling rate is slow enough to permit the formation of polygonal ferrite along the prior austenite grain boundaries (Figures 42-44). The balance of the ESW HAZ structure is coarse bainite. For a cooling rate corresponding to an 8/5 time of 2050 seconds, the polygonal ferrite forms at 1640°F (895°C). The ferrite transformation finish was not detected. The resulting small amounts of polygonal ferrite in the ESW microstructure (Figure 42), indicates that this thermal cycle is just within the ferrite "nose" of the 2 1/4Cr-1Mo-0.12C CCT diagram. The presence of polygonal ferrite in the microstructure contributes to the reduction in hardness (315 DPH) observed in the ESW continuous-cooling transformation evaluation.

In the literature, the weld HAZ microstructure of 3Cr-1.5Mo and 2 1/4Cr-1Mo is generally described as predominantly bainitic or bainitic-martensitic. For weld energy input/cooling conditions which correspond to those employed in actual practice, it is hard to discern the differences between martensite and bainite (Figures 36-43). To provide a base line for the possible delineation of martensite in 3Cr-1.5Mo and 2 1/4Cr-1Mo, simulated HAZ specimens were subjected to a 2400°F (1315°C) peak temperature followed by helium quenching. Photomicrographs of the helium quenched HAZ specimens are shown in Figures 45-47. In general, the microstructure after helium quenching is finer than the microstructure corresponding to an energy input of 35 KJ/in (1.38 KJ/mm) or 150 KJ/in (5.91 KJ/mm). Using light microscopy it was not possible to distinguish any prominent

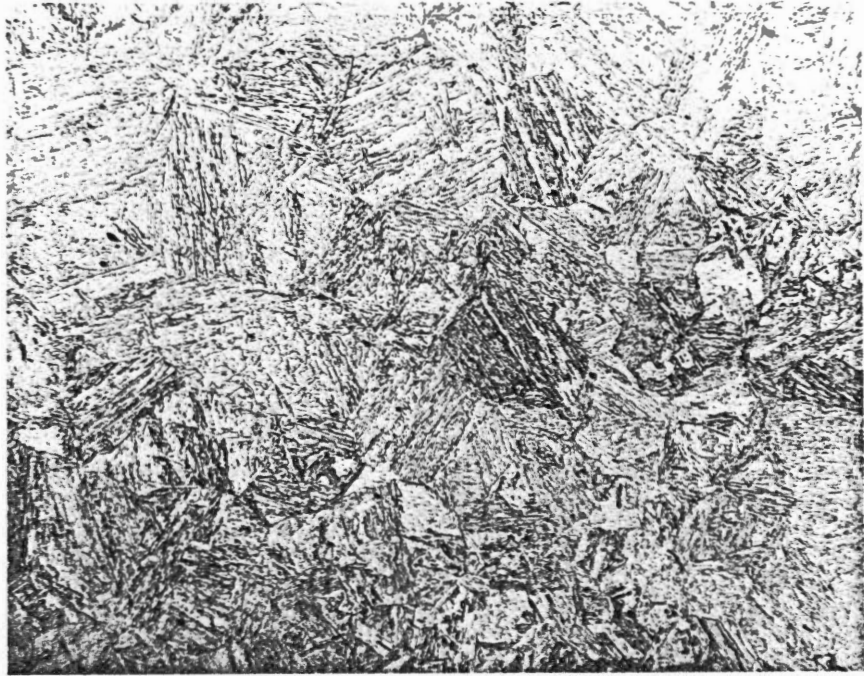


Figure 45. Photomicrograph of Climax 3Cr-1.5Mo Gleeble specimen, helium quenched. OLM. 200X. Nital-Picral.

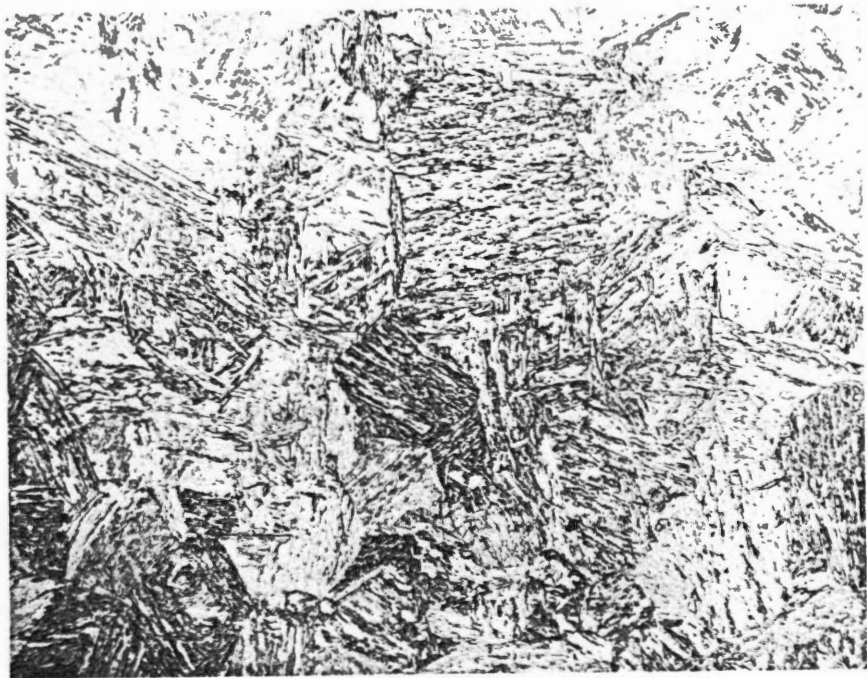


Figure 46. Photomicrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, helium quenched. OLM. 200X. Nital-Picral.

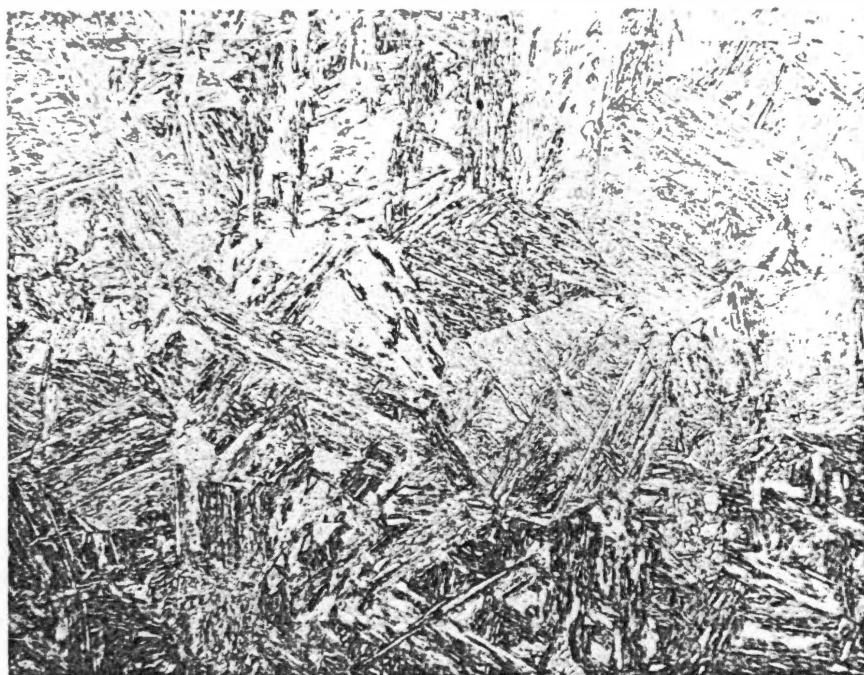


Figure 47. Photomicrograph of Kawasaki 2 1/4Cr-1Mo-0.12C Gleeble specimen, helium quenched. OLM. 200X. Nital-Picral.

microstructural features after helium quenching which could be characterized as martensite. Metallographic evaluation in the scanning electron microscope (SEM) was required to discern the subtle changes in the on-cooling microstructure which were associated with a change in cooling rate.

Using the SEM, four distinct HAZ microstructures were observed in the helium quenched, 35 KJ/in (1.38 KJ/mm), 150 KJ/in (5.91 KJ/mm) and ESW thermal cycle specimens (Figures 48-58). These four microstructures can be described as:

1. Acicular-lenticular (autotempered martensite);
2. Bainitic-lenticular;
3. Intermediate granular-bainitic;
4. True granular-bainitic.

Upon helium quenching, an acicular microstructure is produced (Figures 48, 52 and 56). Large plates of lenticular cross-section are located between some of the acicular structures. The presence of the lenticular plates in the acicular microstructure produce a "roughening" effect on the HAZ microstructure. This same type of "roughening" has been observed in water quenched 2 1/4Cr-1Mo and has been attributed to autotempering of the first large martensite plates to form (76). The fine structures within the lenticular plates have been attributed to the precipitation of cementite during autotempering (76,77).

The slower cooling rate of the 35 KJ/in (1.38 KJ/mm) thermal cycle produces a conventional bainitic microstructure in which the

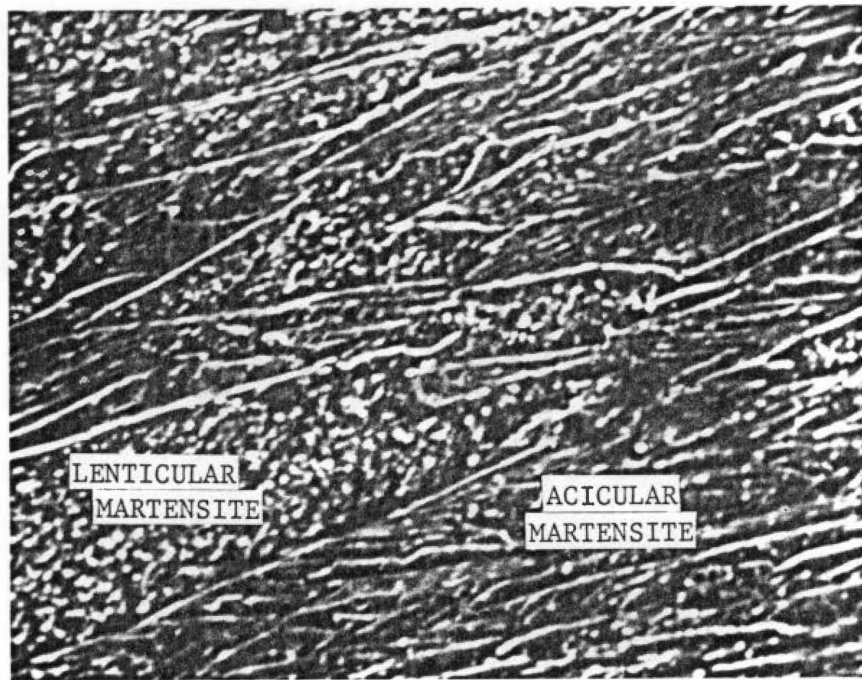


Figure 48. Scanning electron micrograph of Climax 3Cr-1.5Mo Gleeble specimen, helium quenched. SEM. 10,000X.

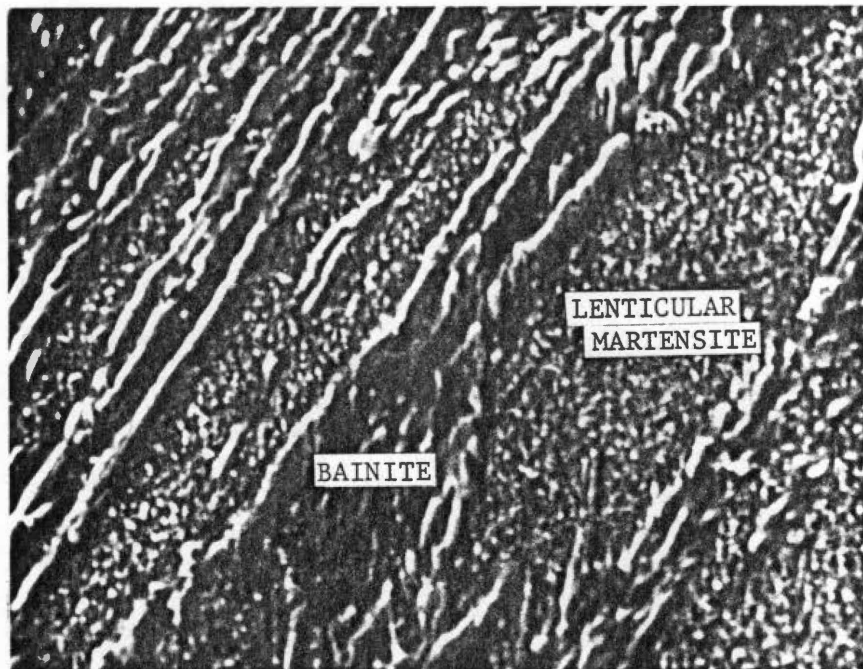


Figure 49. Scanning electron micrograph of Climax 3Cr-1.5Mo Gleeble specimen, 35 KJ/in (1.38 KJ/mm). SEM. 10,000X.

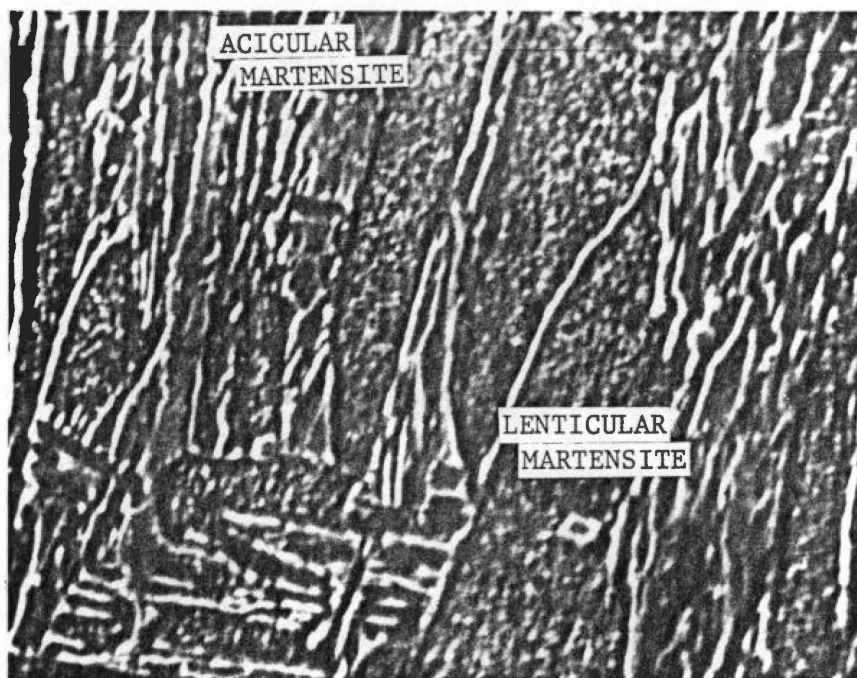


Figure 50. Scanning electron micrograph of Climax 3Cr-1.5Mo Gleeble specimen showing acicular structure, 35 KJ/in (1.38 KJ/mm). SEM. 10,000X.

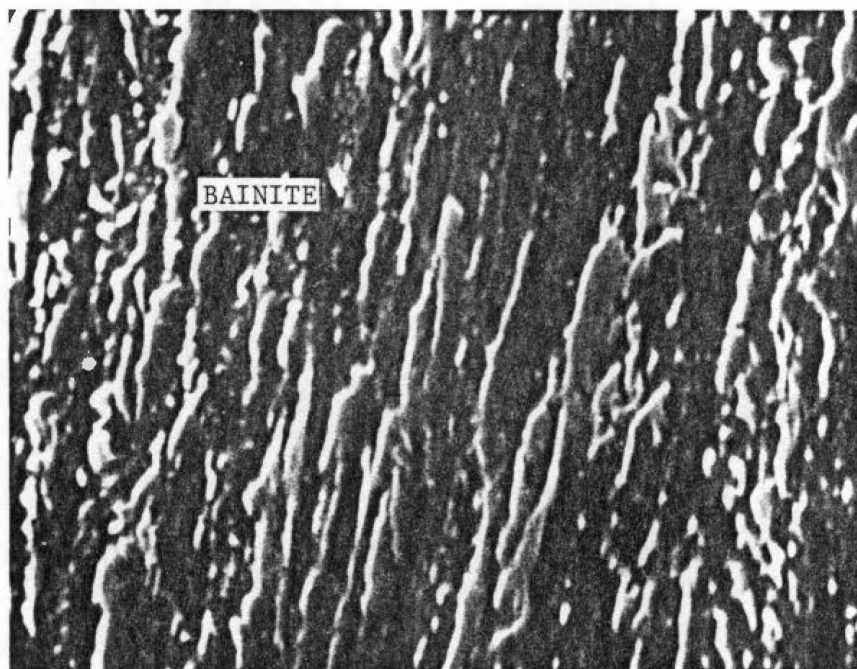


Figure 51. Scanning electron micrograph of Climax 3Cr-1.5Mo Gleeble specimen, 150 KJ/in (5.91 KJ/mm). SEM. 10,000X.

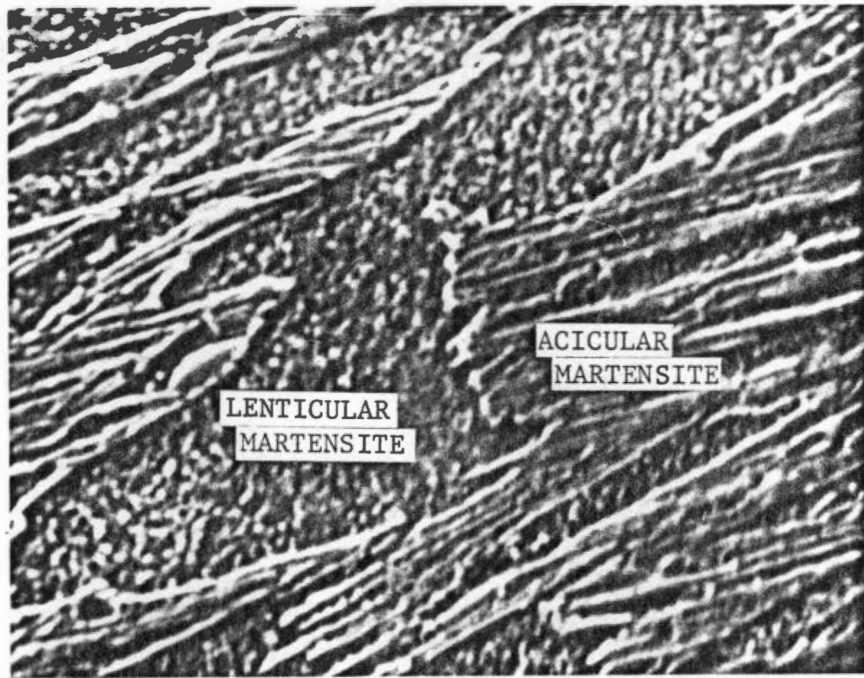


Figure 52. Scanning electron micrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, helium quench. SEM. 10,000X.

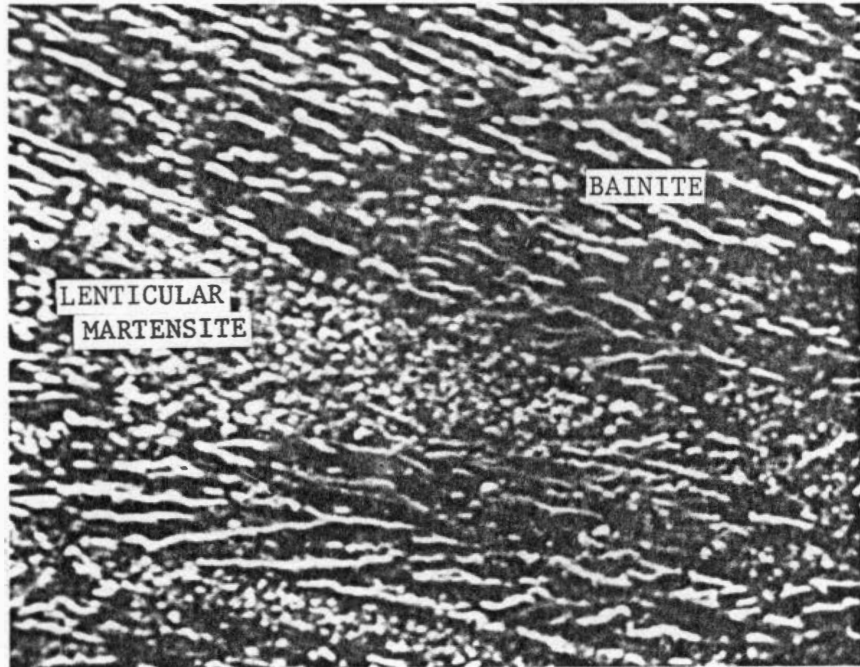


Figure 53. Scanning electron micrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, 35 KJ/in (1.38 KJ/mm). SEM. 10,000X.

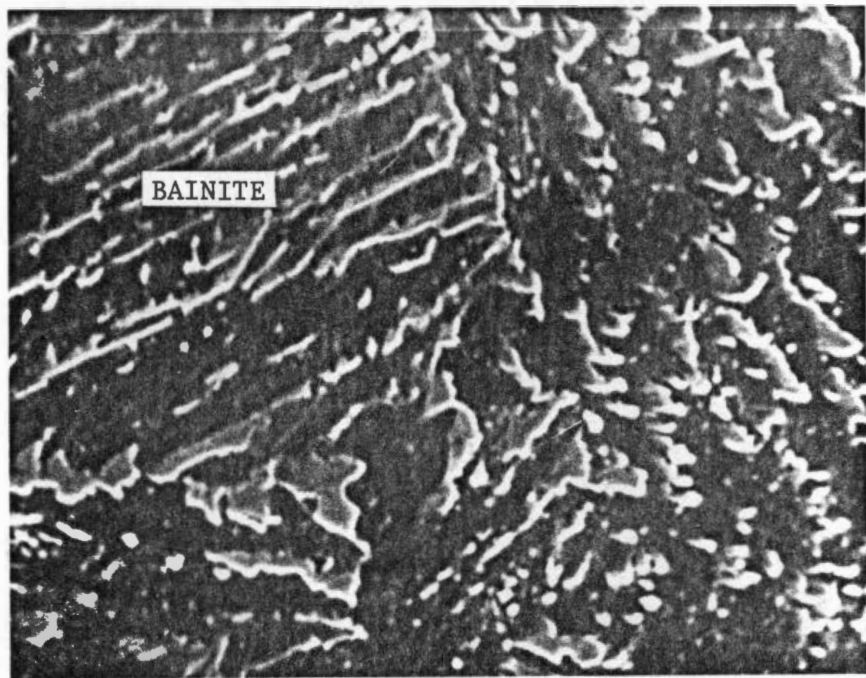


Figure 54. Scanning electron micrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, 150 KJ/in (5.91 KJ/mm). SEM. 10,000X.

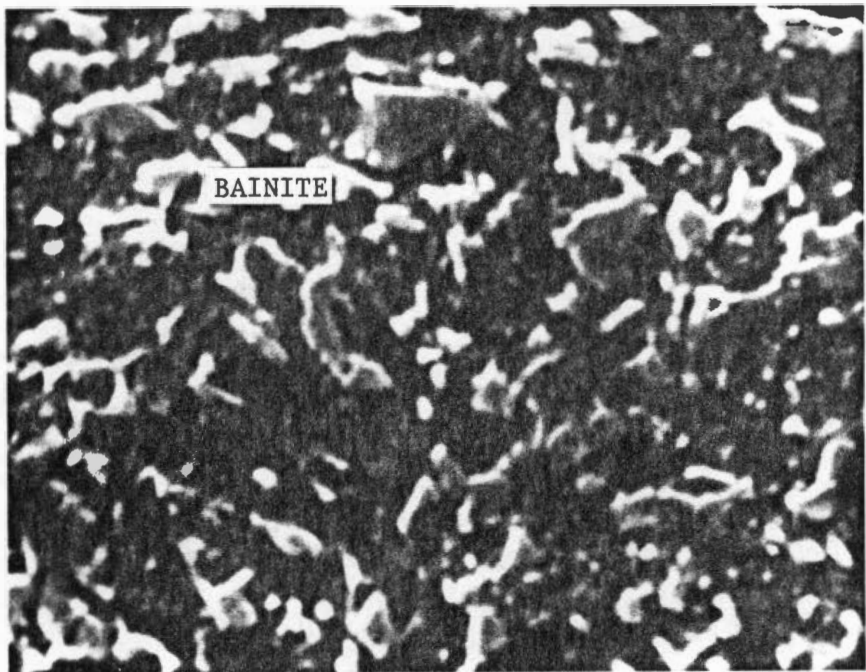


Figure 55. Scanning electron micrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, ESW. SEM. 10,000X.

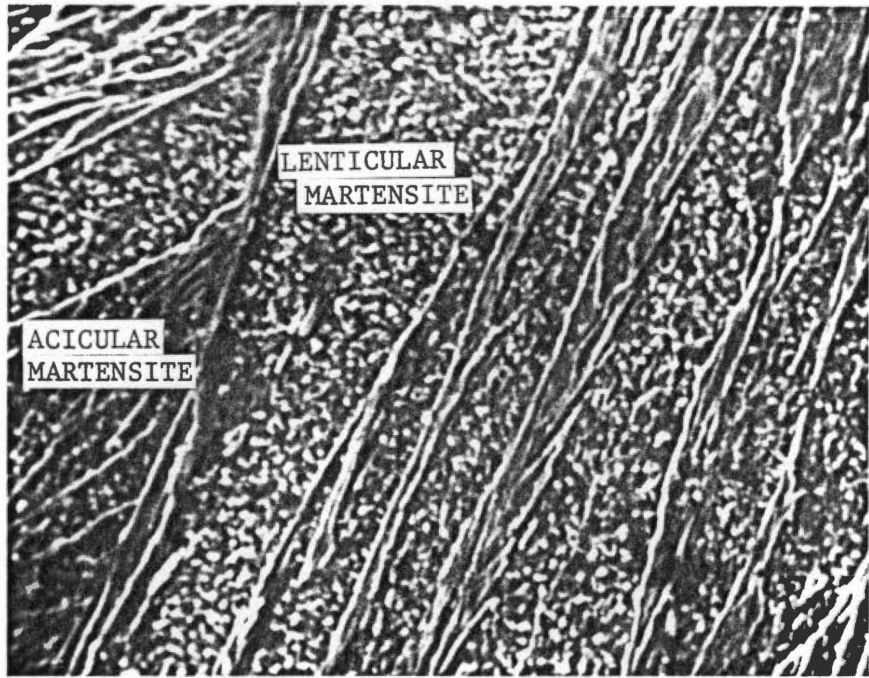


Figure 56. Scanning electron micrograph of Lukens 2 1/4Cr-1Mo-0.12C Gleeble specimen, helium quench. SEM. 10,000X.

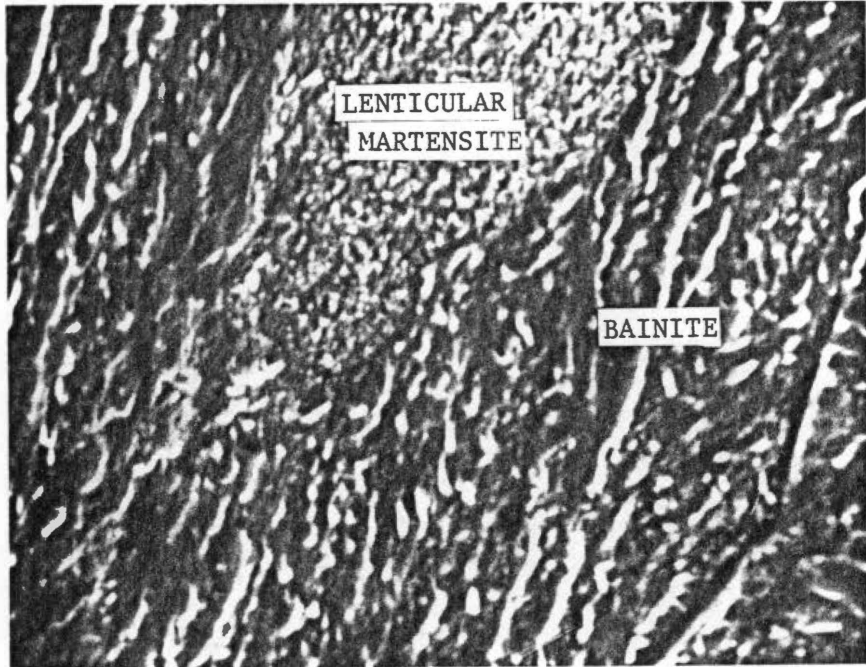


Figure 57. Scanning electron micrograph of Kawasaki 2 1/4Cr-1Mo-0.12C Gleeble specimen, 35 KJ/in (1.38 KJ/mm). SEM. 10,000X.

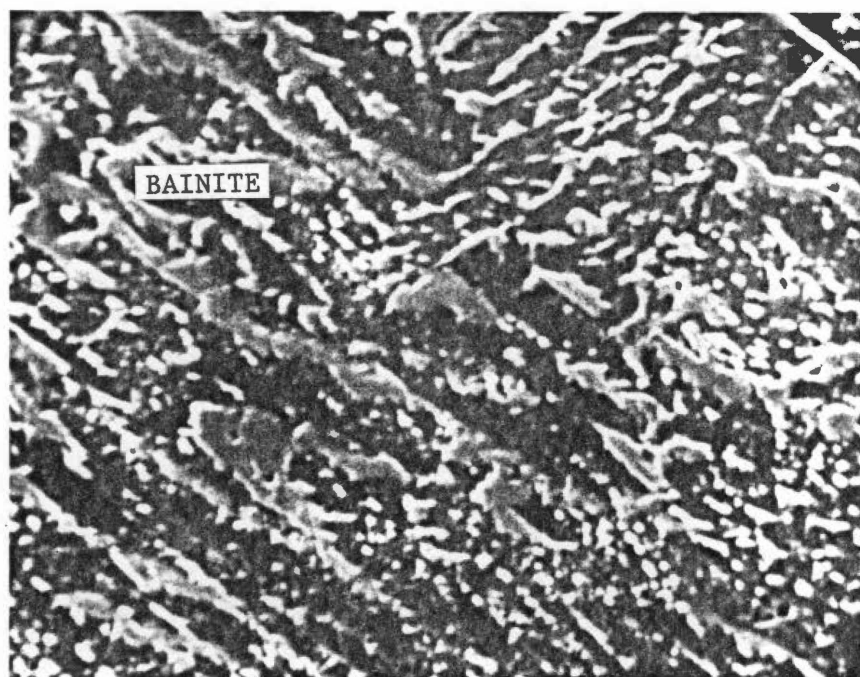


Figure 58. Scanning electron micrograph of Kawasaki 2 1/4Cr-1Mo-0.12C Gleeble specimen, 150 KJ/in (5.91 KJ/mm). SEM. 10,000X.

lenticular structures may still be observed (Figures 49, 53 and 57). In the Climax 3Cr-1.5Mo specimen corresponding to a 35 KJ/in (1.38 KJ/mm) energy input scattered regions which are acicular-lenticular in nature may be observed (Figure 50). The extent of the acicular structures in the Climax 3Cr-1.5Mo 35 KJ/in (1.38 KJ/mm) simulation is less than that observed in the helium quenched specimens.

The bainite formed during the 35 KJ/in (1.38 KJ/mm) thermal cycle consists of elongated discontinuous carbide particles precipitated along ferrite lath boundaries. The precipitates along the lath boundaries are most likely cementite, however it is possible that during continuous cooling some carbon-enriched austenite does not transform as bainite, but is retained. Sometimes this enriched austenite upon further cooling may transform to martensite.

The cooling rate for the 150 KJ/in (5.91 KJ/mm) thermal cycle is sufficiently slow to produce a completely bainitic microstructure (Figures 51, 54 and 58). The elongated precipitates in the fully bainitic microstructure are irregular and no longer follow the distinctive ferrite lath boundaries as in the case of the conventional bainitic structure. The precipitates for the 150 KJ/in (5.91 KJ/mm) thermal cycle take on a granular aspect. Because this transition microstructure has attributes of both the conventional and granular bainitic structures, it is best described as an intermediate granular structure.

The coarse bainite microstructure resulting from the relatively slow cooling rate of the ESW thermal cycle is shown in Figure 55. The precipitates formed during the ESW thermal cycle exhibit the blocky morphology of a true granular bainite. Habraken and Economopoulos (62) have shown that in 2 1/4Cr-1Mo these blocky transformation products contain both austenite and martensite. Wada (63) has described such transformation products as martensite-austenite islands.

Summary of the continuous-cooling transformation behavior of 3Cr-1.5Mo and 2 1/4Cr-1Mo. Unlike 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo, the 3Cr-1.5Mo and 2 1/4Cr-1Mo coarse grained HAZ transformation behavior is dependent on the HAZ cooling rate (energy input). For cooling rates corresponding to a weld arc strike ($8/5 \ t = 0.53 \text{ sec}$), the HAZ structure contains lenticular regions of tempered martensite and acicular regions of untempered martensite. For a slower cooling rate corresponding to an energy input of 35 KJ/in (1.38 KJ/mm, $8/5 \ t = 10 \text{ sec}$), a microstructure containing lenticular martensite and bainite is produced. A further decrease in the cooling rate, to a level corresponding to an energy input of 150 KJ/in (5.91 KJ/mm, $8/5 \ t = 40 \text{ sec}$), produces a completely bainitic microstructure. An extremely slow cooling rate, such as that encountered in an electroslog weld ($8/5 \ t = 2050 \text{ sec}$), produces a blocky bainitic microstructure which contains small regions of polygonal ferrite.

In terms of weldability, the variation in the 3Cr-1.5Mo and 2 1/4Cr-1Mo HAZ microstructure resulting from a change in cooling rate (energy input) is very important. In the literature review, it was explained that a bainitic HAZ microstructure is less susceptible to HAC than a martensitic HAZ. In the ensuing sections, it will be shown that at an equivalent carbon content the predominantly bainitic microstructure produced in the 3Cr-1.5Mo and 2 1/4Cr-1Mo HAZ is less susceptible to HAC than the martensite plus ferrite or martensite plus ferrite and austenite HAZ microstructure found in the 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-2Mo and 9Cr-1Mo materials. In the following discussion on HAC, it will also be noted that any increase in cooling rate which increases the likelihood of forming HAZ martensite in 3Cr-1.5Mo or 2 1/4Cr-1Mo increases the HAC susceptibility.

Stress-Relief Cracking (SRC) Behavior of Cr-Mo Steels

Stress-relief cracking (also called reheat cracking) is intergranular cracking in the HAZ of weldments which occurs during PWHT. Post-weld heat treatment involves reheating the weldment in order to: (1) temper the hardened weldment to a strength and toughness level comparable to that designed for the structure, and (2) reduce the residual stresses in the weld by plastic flow at elevated temperatures (creep). It is indeed unfortunate that certain steels when given a PWHT to produce beneficial effects on the weldment for service, develop SRC cracks which render the welded structure unacceptable.

Stress-relief cracking results directly from the inability of a welded HAZ microstructure to elastically/plastically accommodate the strains which result from the relief of residual stresses during PWHT. The ability of a particular HAZ microstructure to accommodate strain during PWHT depends on the level of stress which is to be relieved, the temperature and composition. The level of stress in a welded structure is determined by section thickness, restraint, and the presence of notches or other stress concentrators. In Cr-Mo steels, the inability of the microstructure to accommodate strain results from the carbide precipitation reactions that strengthen the grain matrix and the presence of residual elements (P, As, Sb, Sn and Al) which reduce the strength of the grain boundary regions. The precipitation reactions are controlled by: (1) Composition—the presence of alloying elements which increase the amount of carbon precipitates formed (C, Cr, Mo and V). (2) The weld thermal cycle—the fast cooling rates and high austenitizing temperatures associated with welding increase the amount of carbide forming elements remaining in solution in the transformed HAZ, which during PWHT precipitates as carbides. (3) Post-weld heat treatment time and temperature—carbide formation is a diffusional process and is therefore time and temperature dependent.

The object of the SRC investigation was to determine the effect of stress, composition, microstructure and PWHT temperature on the coarse grained HAZ SRC susceptibility of 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo, 3Cr-1.5Mo and 2 1/4Cr-1Mo. The coarse grained HAZ

was selected as the main focus of the SRC evaluation because the large grain size and extensive on-heating resolution of carbides in this region results in a maximum susceptibility to SRC. In this investigation, stress-rupture tests of the Gleeble simulated HAZ were employed extensively. The stress-rupture test has been successfully utilized to determine the SRC susceptibility of A517F, SA508 and SA533 (67,78).

In the stress-rupture test method, the Gleeble is used to simulate the coarse grained HAZ associated with a HAZ peak temperature of 2400°F (1315°C) and an energy input of 60 KJ/in (2.36 KJ/mm) in a 2in (51mm) plate at 400°F (205°C) preheat. After the grain coarsening thermal cycle, a preselected tensile stress comparable to the residual stress encountered in an actual weldment is applied. The temperature of the specimen is then increased to a PWHT temperature which is representative of that employed in actual fabrication and the time required to rupture the sample is recorded. After the test, the fracture surface is examined and the RA determined. In the stress-rupture test method, intergranular fracture accompanied by a low reduction in area (RA) is associated with an increased susceptibility to SRC. The stress-rupture test method for SRC susceptibility is described in greater detail in Chapter III.

The results of the stress-rupture tests presented in Table 10 show the influence of stress and PWHT temperature on the time to failure, RA and fracture morphology for the Cr-Mo materials studied. In order to directly compare the stress-rupture strength and

Table 10. Summary of Stress-Relief Cracking Test Results.

Material	PWHT °F	PWHT (°C)	Stress		Time to Fracture (min.)	% RA	Type Fracture
Lukens 2 1/4Cr-1Mo-0.12C	1250	680	22.4	154	49.5	13	Intergranular
"	"	"	28.1	194	21.5	13	"
"	"	"	34.6	239	12.7	17	"
"	"	"	38.7	267	2.7	36	"
"	1150	620	55.0	378	133.0	2.9	"
"	"	"	68.0	468	31.8	1.7	"
"	"	"	77.4	532	10.1	2.6	"
"	"	"	77.4	532	9.3	3.1	"
Lukens 2 1/4Cr-1Mo-0.10C	1250	680	31.4	216	39.0	19.6	"
"	"	"	31.7	218	12.4	23.2	"
"	"	"	35.5	245	8.4	29.9	"
"	"	"	39.5	272	2.5	29.5	"
"	1150	620	64.8	446	53.0	14.6	"
"	"	"	74.1	510	4.1	11.9	"
"	"	"	75.0	517	11.0	14.7	"
Kawasaki 2 1/4Cr-1Mo-0.12C	1250	680	33.4	230	42.0	21.5	"
"	"	"	38.9	267	11.4	22.1	"
"	"	"	47.0	323	4.3	21.3	"
"	1150	620	53.0	364	97.5	7.4	"
"	"	"	67.0	461	11.5	8.2	"
"	"	"	70.0	481	5.0	7.1	"
Climax 3Cr-1.5Mo	1250	680	39.5	272	34.0	36	"
"	"	"	53.4	368	17.6	29	"
"	"	"	59.0	407	8.3	27	"
"	1150	620	76.6	528	121.0	8.3	"
"	"	"	77.4	533	64.8	10.8	"
12Cr-2Mo-6Ni	1290	700	33.0	228	150.0	60	Dimple
"	"	"	35.8	247	37.3	69	"
"	"	"	38.7	267	16.4	67	"
12Cr-2Mo-3Ni	"	"	20.3	140	75.2	79	"
"	"	"	25.0	172	12.0	84	"
"	"	"	35.0	241	0.5	87	"
12Cr-2Mo-1Ni	"	"	22.4	154	71.6	72	"
"	"	"	27.0	186	8.0	80	"
"	"	"	29.3	202	4.6	81	"
"	"	"	31.0	214	3.1	85	"
9Cr-1Mo-0.11C	1350	730	13.4	92	4.1	75	"
"	"	"	11.4	79	43.0	79	"
"	"	"	8.1	56	184.0	78	"
9Cr-2Mo	1320	715	15.0	103	40.5	89	"
"	"	"	18.5	128	10.1	89	"
"	"	"	20.5	141	2.7	91	"
9Cr-2Mo-V-Nb	1350	730	20.2	139	67.0	77	"
"	"	"	22.5	155	13.9	82	"
"	"	"	24.4	168	9.3	81	"

PWHT - Post weld heat treatment

RA - Reduction in area

stress-rupture ductility of the Cr-Mo materials, the time to failure and reduction in area values are plotted as a function of applied stress (Figures 59, 60 and 61). For equivalent times-to-fracture, the duplex (12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo) and martensitic (9Cr-1Mo) materials exhibit higher rupture ductility and lower rupture strength than the bainitic materials (3Cr-1.5Mo and 2 1/4Cr-1Mo). The differences between the stress-rupture properties of the duplex, martensitic and bainitic materials are apparently related to a difference in the carbide precipitation reaction which occurs during PWHT. According to Baker and Nutting (77), the carbides formed in the Bainitic materials during PWHT should be Fe_3C and Mo_2C . The presence of a fine dispersion of coherent Mo_2C precipitates strengthens the bainitic matrix, resulting in the high rupture strength and the low rupture ductility observed.

In the 2 1/4Cr-1Mo and 3Cr-1.5Mo the strengthening effect of the Mo_2C coherent precipitate is more pronounced at 1150°F (620°C) than at 1250°F (680°C) as evidenced by the higher rupture strengths observed at the lower PWHT temperature. In the duplex and martensitic materials, the majority of the carbides formed during PWHT have been shown to be large spheroidal M_{23}C_6 with small amounts of M_2C and MC (1,73,79). The martensite matrix of the duplex and martensitic materials is not strengthened by the precipitation of coarse M_{23}C_6 precipitates, thus the rupture strength of the duplex and martensitic materials are lower than those observed for the bainitic materials.

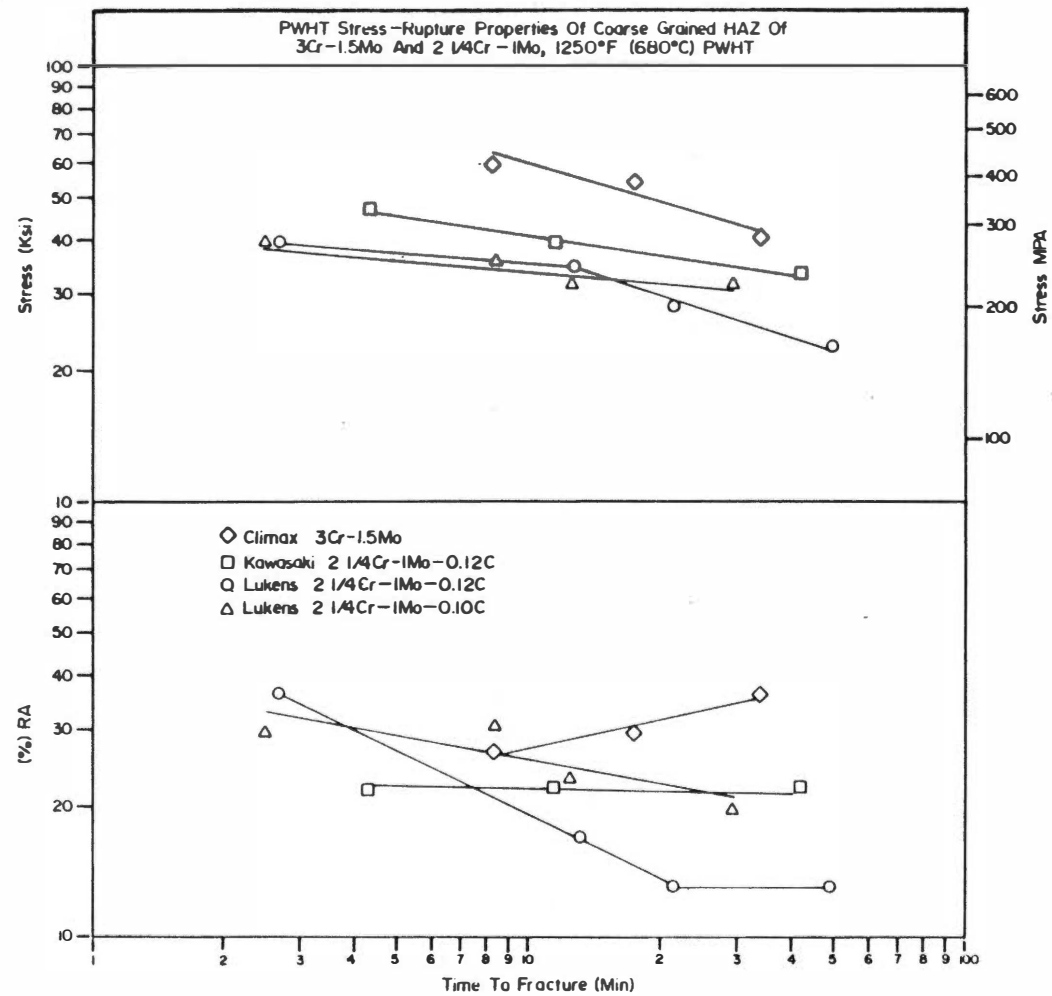


Figure 59. Results of SRC test for 3Cr-1.5Mo and 2 1/4Cr-1Mo (PWHT - 1250°F (680°C)).

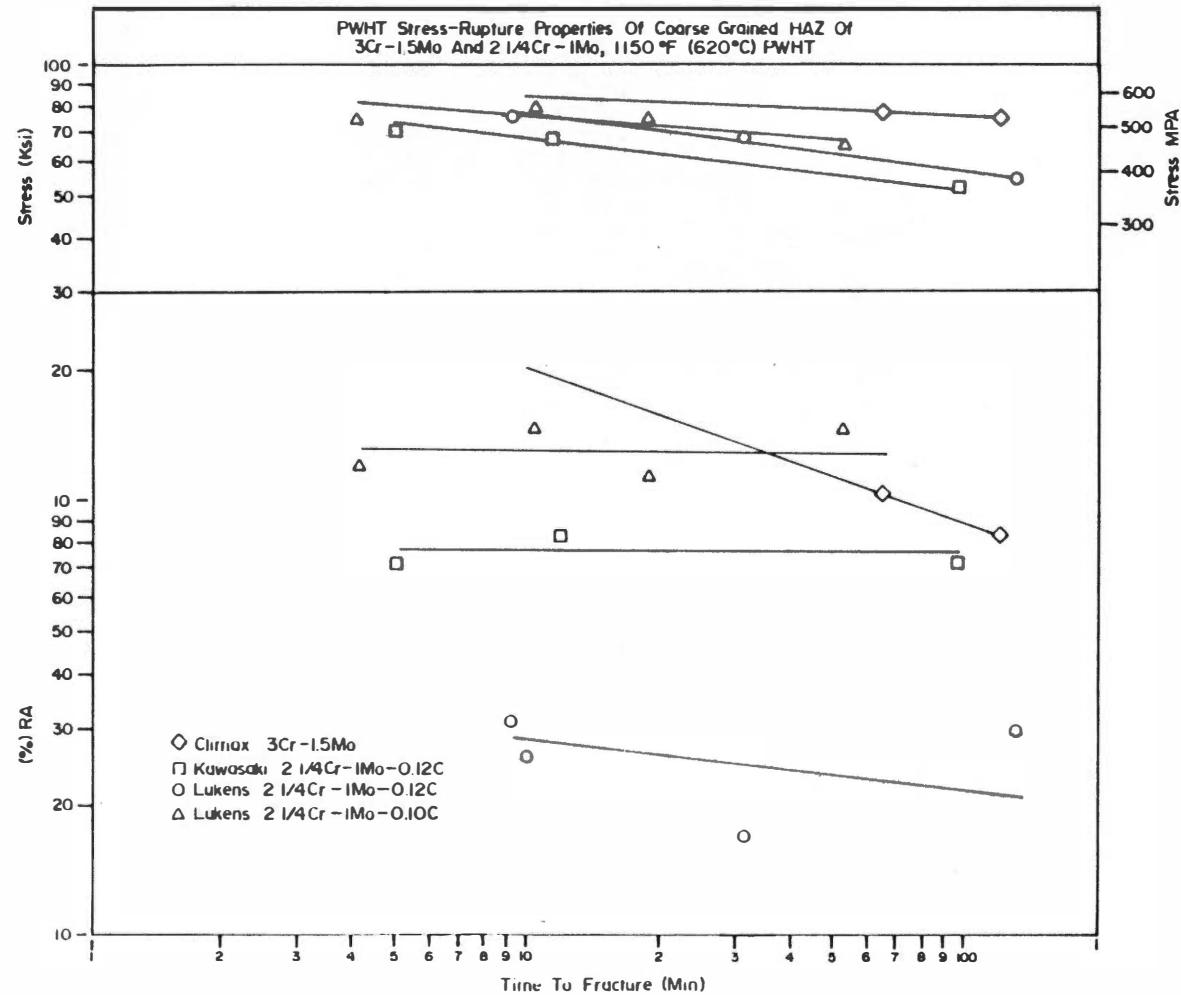


Figure 60. Results of SRC test for 3Cr-1.5Mo and 2 1/4Cr-1Mo (PWHT - 1150°F (620°C)).

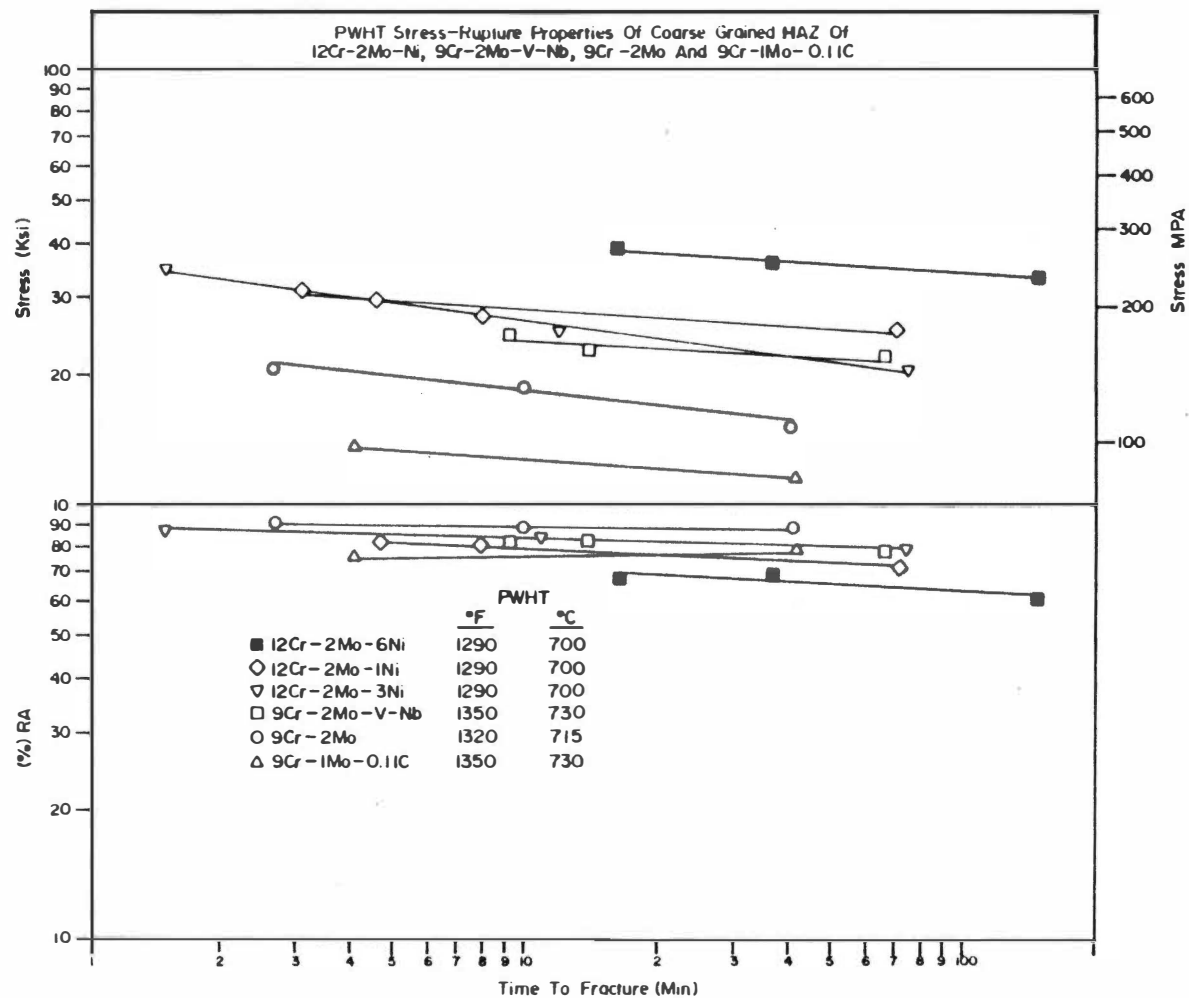


Figure 61. Results of SRC test for 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo.

Evaluation of the Relative SRC Susceptibility of Cr-Mo Steels

The SRC susceptibilities of the Cr-Mo steels examined in this investigation can be evaluated on the basis of rupture ductility or a rupture strength and rupture ductility parameter. Vinckier and Pense (80) have developed the following criteria for SRC based on the 1110°F (600°C) rupture ductility (%RA) of various steels previously subjected to a simulated HAZ cycle:

SRC Susceptibility	%RA	HAZ Peak Temperature	
		°F	°C
Extremely susceptible	5	2190	1200
Susceptible	10	2370	1300
Slightly susceptible	15	2460	1350
Not susceptible	20	All peak temperatures	

Researchers at United States Steel have developed a stress-rupture parameter (SRP) based on both stress-rupture strength and stress-rupture ductility (80). The SRP is the product of the stress for a rupture time of 10 minutes and the corresponding reduction in area. A high value of the SRP is indicative of good stress-rupture strength and ductility, consequently a low susceptibility to SRC would be expected. A low SRP value is indicative of poor stress-rupture properties and increased SRC susceptibility. High strength martensitic steels with a $SRP > 15 \times 10^5$ have been shown to exhibit good stress-rupture strength and ductility, and show a low susceptibility to SRC.

Relative SRC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo. The rupture ductilities of the four bainitic materials (Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C and

Kawasaki 2 1/4Cr-1Mo-0.12C) for the 1250°F (680°C) PWHT range from 13-36% RA (Figure 59). With the exception of the Climax 3Cr-1.5Mo material, the rupture ductilities of the bainitic materials decreased as the time to rupture increased. The increase in rupture ductility with increased time to fracture observed in the Climax 3Cr-1.5Mo may be attributed to the overaging of a coherent precipitate (81). Only the Lukens 2 1/4Cr-1Mo-0.12C material exhibited 1250°F (680°C) PWHT rupture ductilities below 20% RA which, according to the Vinckier and Pense criteria, indicates a "slight susceptibility" towards SRC.

When the PWHT temperature of the Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials was reduced from 1250°F (680°C) to 1150°F (620°C), the rupture strengths at equivalent times to rupture increased (Figure 60). The 1150°F (620°C) PWHT temperature corresponds to the upper realm of temperatures over which the strengthening effect of a coherent precipitate is maximized (82). The increase in rupture strength observed at 1150°F (620°C) PWHT is accompanied by a decline in rupture ductility. The decline in rupture ductility may be indicative of an increase in SRC susceptibility at 1150°F (620°C) PWHT.

The rupture ductilities of the Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials at 1150°F (620°C) PWHT range from 8 to 20% RA which, according to the Vinckier and Pense criteria, indicates a "slight susceptibility" to SRC. The loss in ductility observed in the Lukens 2 1/4Cr-1Mo-0.12C

material at 1150°F (620°C) (1.7 to 3% RA) is more substantial than the loss in ductility observed in the other three bainitic materials. According to the Vinckier and Pense criteria any material exhibiting a RA below 5% RA should be considered "extremely susceptible" to SRC.

Based on rupture-ductility considerations the Lukens 2 1/4Cr-1Mo-0.12C is the only bainitic material tested which shows an inclination towards SRC at 1250°F (680°C) PWHT. For this PWHT temperature, the Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials display virtually no SRC tendency. In contrast, all four bainitic materials demonstrate some degree of SRC susceptibility at 1150°F (620°C) PWHT. The Lukens 2 1/4Cr-1Mo-0.12C also exhibits the greatest susceptibility to SRC at 1150°F (620°C). The SRC susceptibility of the Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C and Kawasaki 2 1/4Cr-1Mo-0.12C at 1150°F (620°C) PWHT is approximately the same as the Lukens 2 1/4Cr-1Mo-0.12C material at 1250°F (680°C) PWHT.

The SRP was employed to more clearly define the relative SRC susceptibility of the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials. In the previous discussion, it was shown that the SRP takes into consideration both rupture-strength and rupture-ductility. The stress-rupture parameters of the four bainitic materials at 1250°F (680°C) and 1150°F (620°C) PWHT have been calculated and are tabulated in Table 11. The SRP values determined for the bainitic materials in this investigation range from 163×10^4 to 22.8×10^4 and are above the SRP value of 15×10^4 , below which high strength martensitic steels have been shown to be susceptible to SRC.

Table 11. Stress-Rupture Parameter and Index.

Material	PWHT		Stress (Ksi)	RA (%)	Stress-Rupture Parameter ($\times 10^4$) ¹	Stress-Rupture Index ²
	°F	(°C)				
Climax 3Cr-1.5Mo	1250	680	60	27	162	100
Kawasaki 2 1/4Cr-1Mo-0.12C	"	"	41	22	90.2	56
Lukens 2 1/4Cr-1Mo-0.10C	"	"	34	25	85.0	52
Lukens 2 1/4Cr-1Mo-0.12C	"	"	35	19	66.5	41
Climax 3Cr-1.5Mo	1150	620	86	20	172	100
Lukens 2 1/4Cr-1Mo-0.10C	"	"	76	13	98.8	57
Kawasaki 2 1/4Cr-1Mo-0.12C	"	"	68	8	54.4	32
Lukens 2 1/4Cr-1Mo-0.12C	"	"	76	3	22.8	13

¹Reduction in area (RA) x Stress (Ksi) for Fracture in 10 minutes.

²Percentage of Climax 3Cr-1.5Mo stress-rupture parameter (SRP).

By employing the SRP, it is possible to rank the cracking susceptibilities of the four bainitic materials in terms of a stress-rupture index (SRI). For a fixed PWHT temperature, the stress-rupture index is the stress-rupture parameter of each material expressed as a percentage of the greatest stress-rupture parameter. The stress-rupture index calculations are also included in Table 11. A histogram of the SRI for each bainitic material at 1250°F (680°C) and 1150°F (620°C) PWHT is shown in Figure 62. The Climax 3Cr-1.5Mo material has the greatest stress-rupture parameter at either PWHT temperature, and thus is considered the least susceptible to SRC. The SRP of the Climax 3Cr-1.5Mo was therefore employed as a base line for the SRI calculations at both PWHT temperatures (1250°F (680°C) and 1150°F (620°C)).

Based on the SRI, the Lukens 2 1/4Cr-1Mo-0.12C material is the most susceptible to SRC at both PWHT temperatures. For 1250°F (680°C) PWHT, there is little difference in the SRC susceptibility of the Lukens 2 1/4Cr-1Mo-0.12C and Kawasaki 2 1/4Cr-1Mo-0.12C materials, however, at 1150°F (620°C) PWHT, the Kawasaki 2 1/4Cr-1Mo-0.12C material is more susceptible to SRC than the Lukens 2 1/4Cr-1Mo-0.12C material.

In order to explain the relative difference in the SRC susceptibility of the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials, one must consider the two primary theories which, in recent years, have arisen to explain the varied experimental observations on SRC. The foremost theory is based on the internal strengthening of grains by the

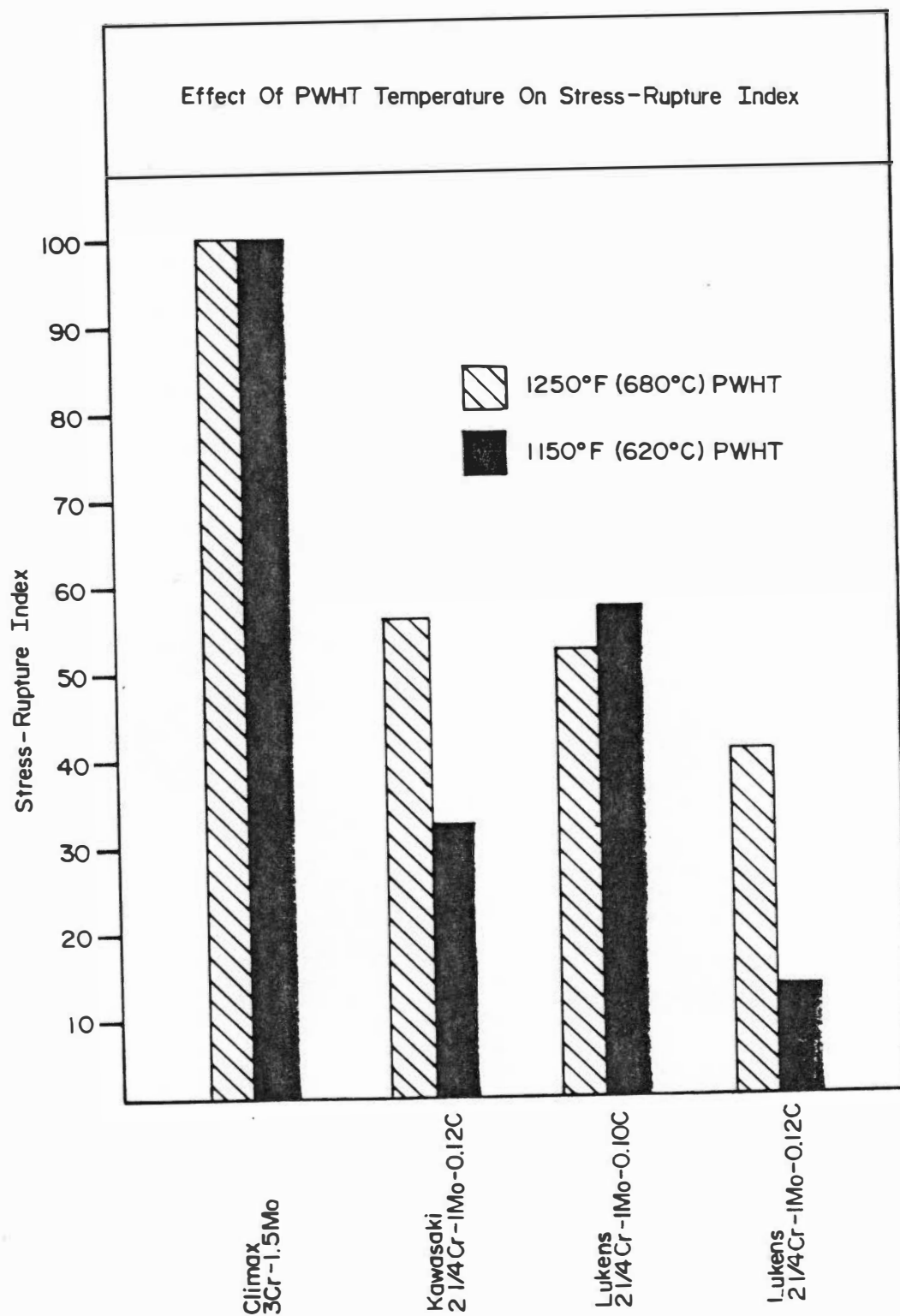


Figure 62. Histogram of stress-rupture index.

formation of a coherent precipitate and the subsequent accumulation of strain along weaker prior austenite grain boundaries. The second theory is based on the segregation of embrittling elements such as P, As, Sn and Sb along prior austenite grain boundaries which results in the degradation of grain boundary strength.

The SRC susceptibility of the bainitic materials in this investigation can be characterized in terms of grain matrix strengthening and/or grain boundary degradation. Among the bainitic materials, there is little difference in the bulk amount of embrittling elements present (Table 3, p. 48). The elements C, Mo, Cr, Si and Mn have been shown to alter the effectiveness of embrittling elements which may be present (47,83). The differences in SRC susceptibilities between the bainitic materials may be related to differences in these nominal alloying elements.

The relatively lower SRC susceptibility observed in the Climax 3Cr-1.5Mo can apparently be attributed to the combined effects of V (0.12%) and Cr (3.02%). Vanadium tends to encourage the formation of V_4C_3 carbides which like Mo_2C carbides are initially coherent and strengthen the grain matrix. If V_4C_3 carbides form instead of Mo_2C , Mo is free to scavenge embrittling elements such as P and prevent grain boundary embrittlement. While V may have detrimental effects on the SRC susceptibility of 0.2 to 1% chromium Cr-Mo-V steels (41, 84), no enhanced SRC susceptibility has been noted for Cr contents ranging from 2.25-2.9% (41,50). Apparently, the amount of Cr present in the Climax 3Cr-1.5Mo material imparts sufficient ductility

to the V strengthened matrix and results in an overall lower susceptibility to SRC.

The relative SRC susceptibilities of the Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials are approximately the same at 1250°F (680°C) PWHT. The increase in the SRC susceptibility of the Kawasaki 2 1/4Cr-1Mo-0.12C at 1150°F (620°C) PWHT may be related to the higher carbon content of this material. Carbon has a strong affinity for Mo and may enhance the formation of matrix strengthening Mo_2C precipitates during PWHT. For the 1150°F (620°C) PWHT, the grain matrix of the higher carbon content Kawasaki 2 1/4Cr-1Mo-0.12C may be strengthened more than the grain matrix of the Lukens 2 1/4Cr-1Mo-0.12C. An increase in grain matrix strength of the Lukens 2 1/4Cr-1Mo-0.12C results in an increase in SRC susceptibility.

Among the four bainitic materials tested, the Lukens 2 1/4Cr-1Mo-0.12C contained the greatest proportion of C and Si. The relatively high SRC susceptibility observed in the Lukens 2 1/4Cr-1Mo-0.12C material at both PWHT temperatures (1250°F (680°C) and 1150°F (620°C)) can be possibly attributed to the combined effects of the C and Si present in this material. In the previous discussion, it was shown that an increase in the carbon content encourages the formation of Mo_2C carbides which strengthen the grain matrix and intensify the SRC susceptibility. The formation of Mo_2C is enhanced by an increase in the Si content (83). Silicon may also act to form grain boundary regions which are denuded of precipitates and are therefore weaker than the bulk of the grain matrix (85). During PWHT, strain may be

concentrated in the precipitate denuded zone, eventually leading to low ductility fracture along prior austenite grain boundaries.

Relative SRC susceptibility of 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo. The rupture-ductilities of the duplex (12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo) and martensitic (9Cr-1Mo) materials range from 60 to 90% (Figure 61, p. 125) and are considerably greater than those observed in the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials (Figures 59 and 60, pp. 123 and 124). The increased stress-rupture ductility of the duplex and martensitic materials, reflects a decrease in relative SRC susceptibility. The rupture-ductility observed in the duplex and martensitic materials are well above the 20% RA level at which the Vinickier and Pense criteria predicts there should be no susceptibility to SRC.

The difference in the SRC susceptibility of the duplex, martensitic and bainitic materials is apparently related to the carbide precipitation reaction which takes place during PWHT. It was shown in the earlier discussion, that upon PWHT, the bainitic materials form M_2C (Mo_2C) or M_4C_3 (V_4C_3) type carbides which are initially coherent and strengthen the grain matrix resulting in an increase in SRC susceptibility. In contrast, upon PWHT the duplex and martensitic materials form carbides which are predominately $M_{23}C_6$. The $M_{23}C_6$ carbides found in the duplex and martensitic materials do not have the strengthening effect of the M_2C and M_4C_3 carbides found in the bainitic materials. Because of a reduction in the strength of the grain interiors, the duplex and martensitic

materials are better suited to accompany the strains associated with PWHT, and as a result, are less susceptible to SRC. The decreased grain matrix strength of the duplex and martensitic materials is reflected in Figures 59, 60 and 61 (pp. 123, 124 and 125). In order to produce equivalent times to rupture, lower stress levels are required in the duplex and martensitic materials than in the bainitic materials.

Metallography and Fractography of SRC Specimens

In the literature review, it was shown that intergranular fracture was a manifestation of SRC. Fractographic and metallographic examinations of the fracture surfaces of the SRC specimens were conducted to substantiate the SRC susceptibility of the 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo, 3Cr-1.5Mo and 2 1/4Cr-1Mo materials.

Metallography and fractography of 3Cr-1.5Mo and 2 1/4Cr-1Mo SRC specimens. Scanning electron fractographs and optical photomicrographs of the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials are shown in Figures 63-74. The fracture morphology of the Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.12C and Kawasaki 2 1/4Cr-1Mo-0.12C bainitic materials exhibits the classic intergranular brittle appearance associated with SRC. In Figures 63-74, two fracture modes can be identified on the basis of fracture appearance. In the Climax 3Cr-1.5Mo and Lukens 2 1/4Cr-1Mo-0.12C specimens subjected to 1250°F (680°C) PWHT, ductile areas can be observed along the grain boundary regions (Figures 63 and 66). This

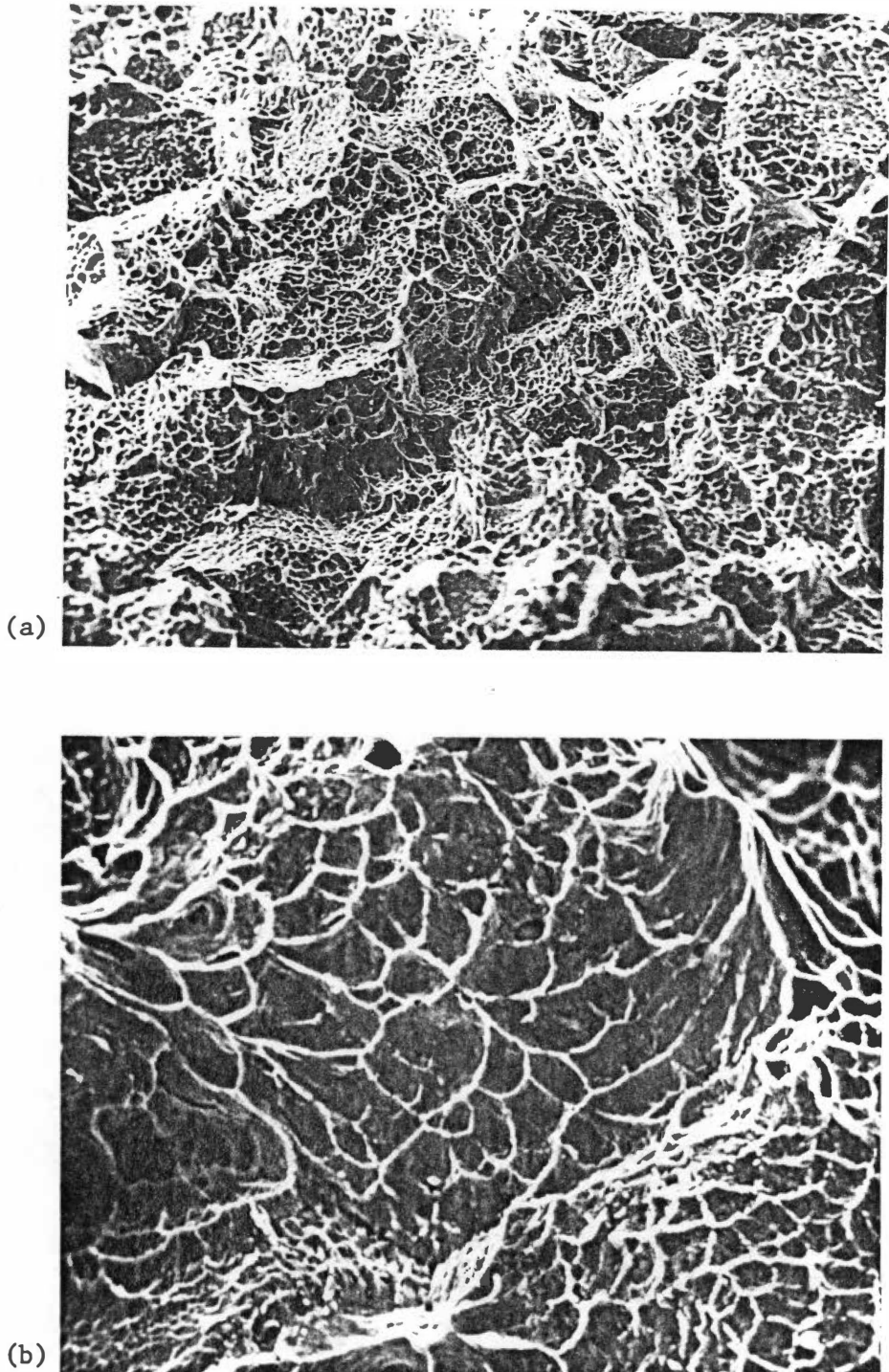


Figure 63. Fractographs of Climax 3Cr-1.5Mo SRC specimen, 1250°F (680°C). 59.0 Ksi (407 Mpa), time to fracture 8.3 min, 27% RA. SEM, (a) 200X, (b) 1000X.

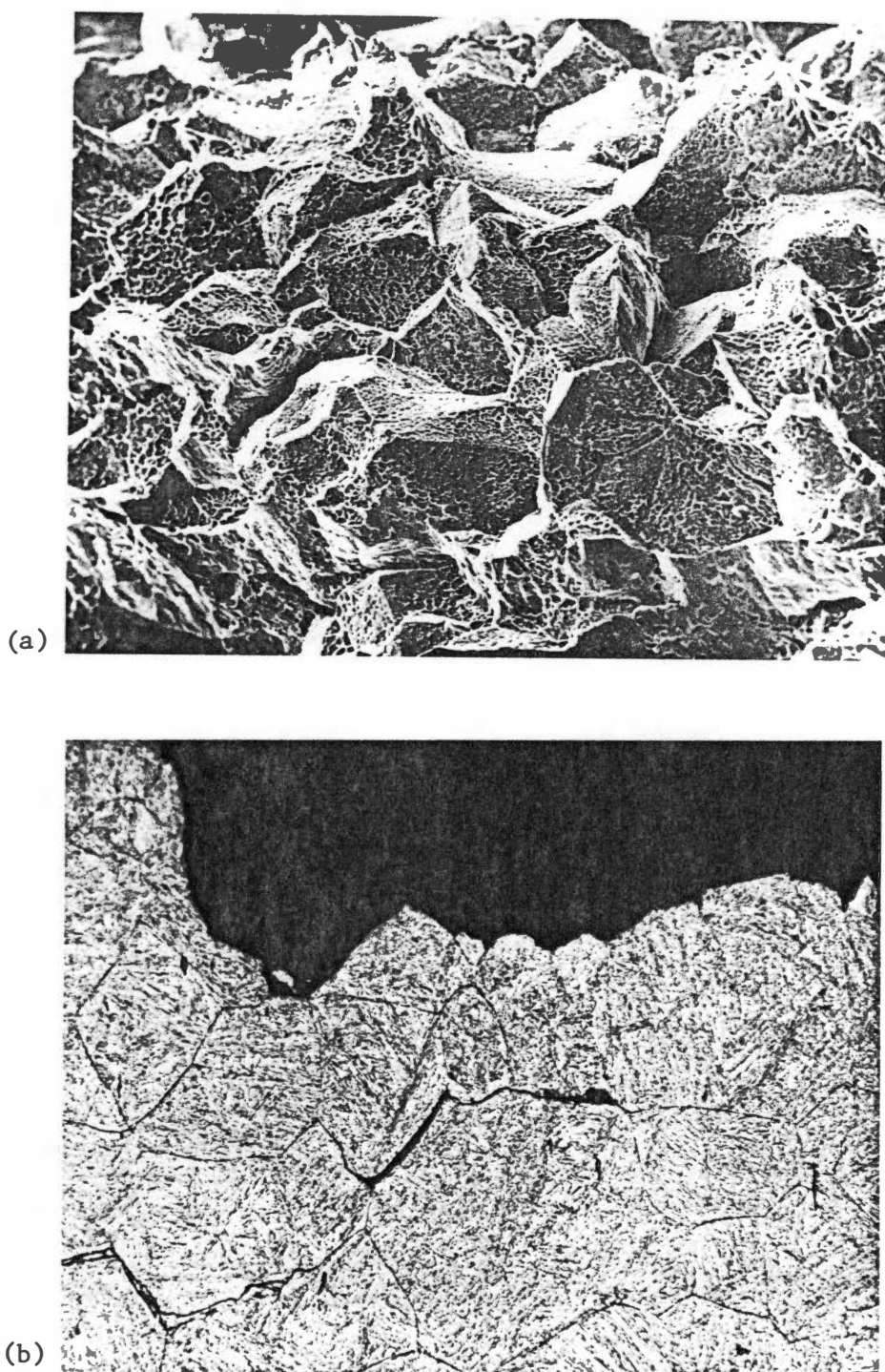


Figure 64. Fractographs of Climax 3Cr-1.5Mo SRC specimen, 1150°F (620°C). 77.4 Ksi (533 Mpa), time to fracture 64.8 min, 10.8% RA. (a) SEM, 200X, (b) Metallographic cross section of (a) showing secondary cracking. OLM. 200X. Picral-Nital.

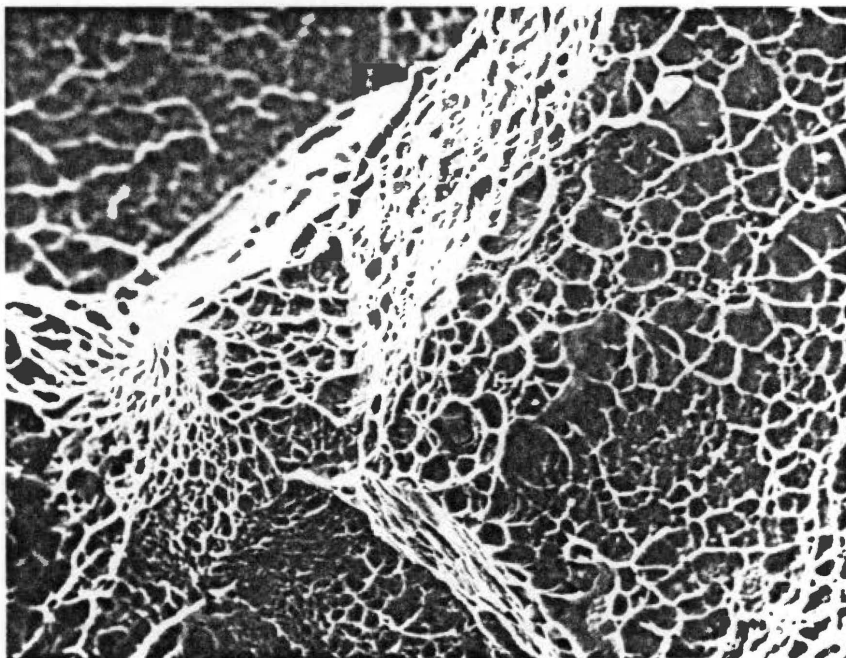


Figure 65. Higher magnification view of Climax 3Cr-1.5Mo SRC specimen, 1150°F (620°C). SEM. 1000X.

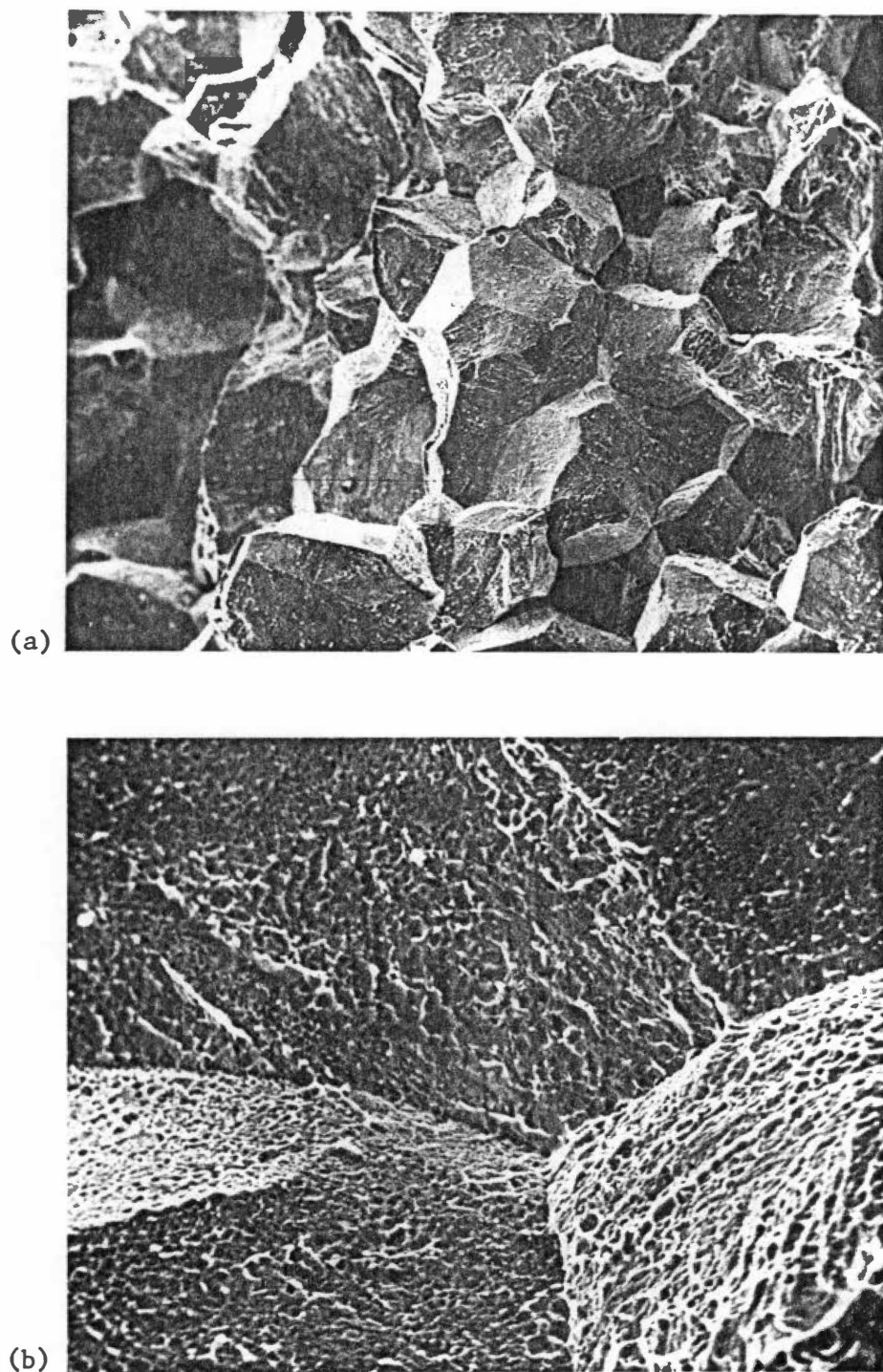


Figure 66. Fractographs of Lukens 2 1/4Cr-1Mo-0.12C SRC specimen, 1250°F (680°C). 22.4 Ksi (154 Mpa), time to fracture 49.5 min, 13% RA. SEM, (a) 200X, (b) 1000X.

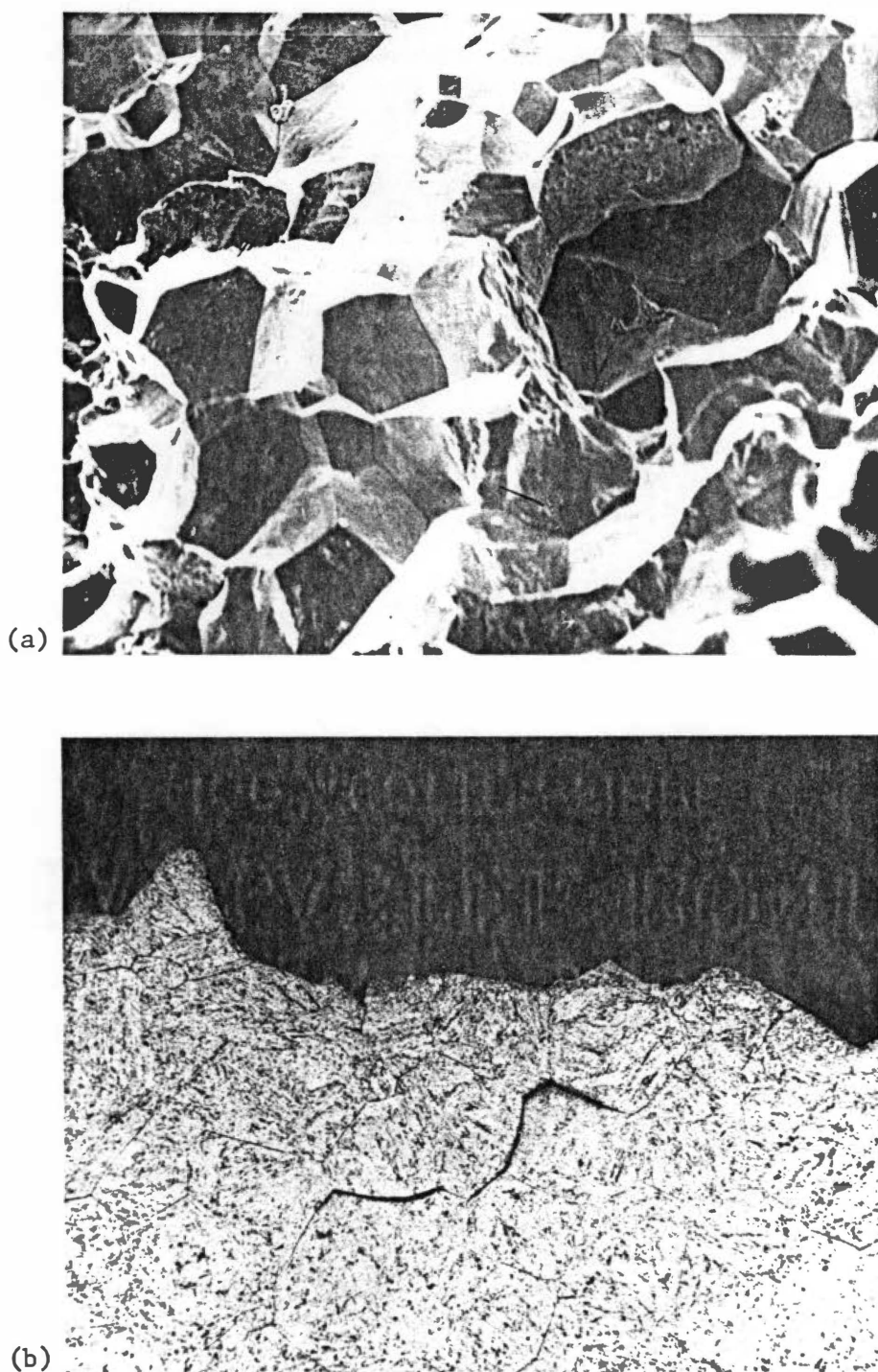


Figure 67. Fractographs of Lukens 2 1/4Cr-1Mo-0.12C SRC specimen, 1150°F (620°C). 77.4 Ksi (533 Mpa), time to fracture 10.1 min, 2.6% RA. (a) SEM, 200X, (b) Metallographic cross section of (a) showing secondary cracking. OLM. 200X. Picral.

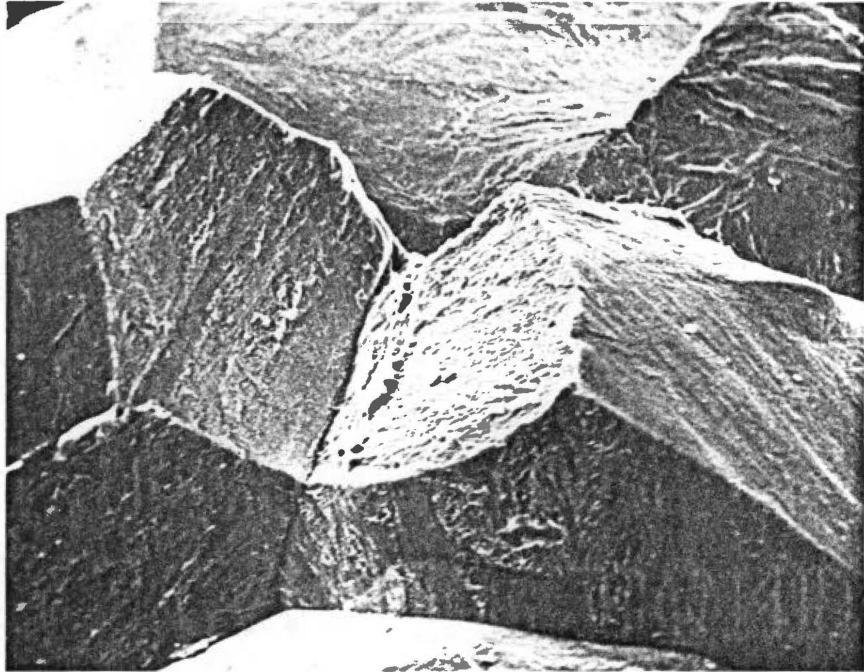


Figure 68. Higher magnification view of Lukens 2 1/4Cr-1Mo-0.12C SRC specimen, 1150°F (620°C). SEM. 1000X.

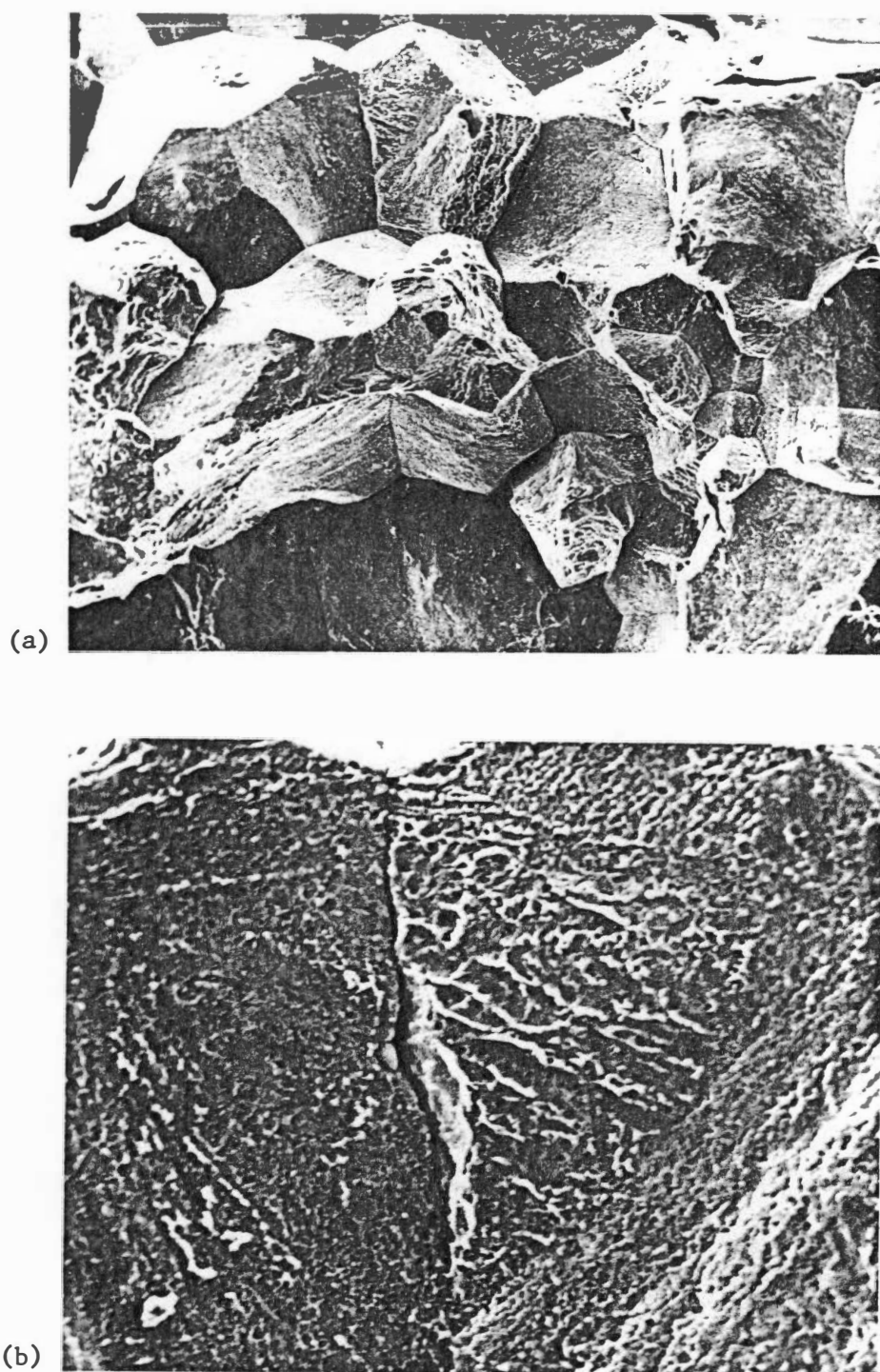


Figure 69. Fractographs of Lukens 2 1/4Cr-1Mo-0.10C SRC specimen, 1250°F (680°C), 31.7 Ksi (218 Mpa), time to fracture 12.4 min, 23.2% RA. SEM. (a) 200X, (b) 1000X.

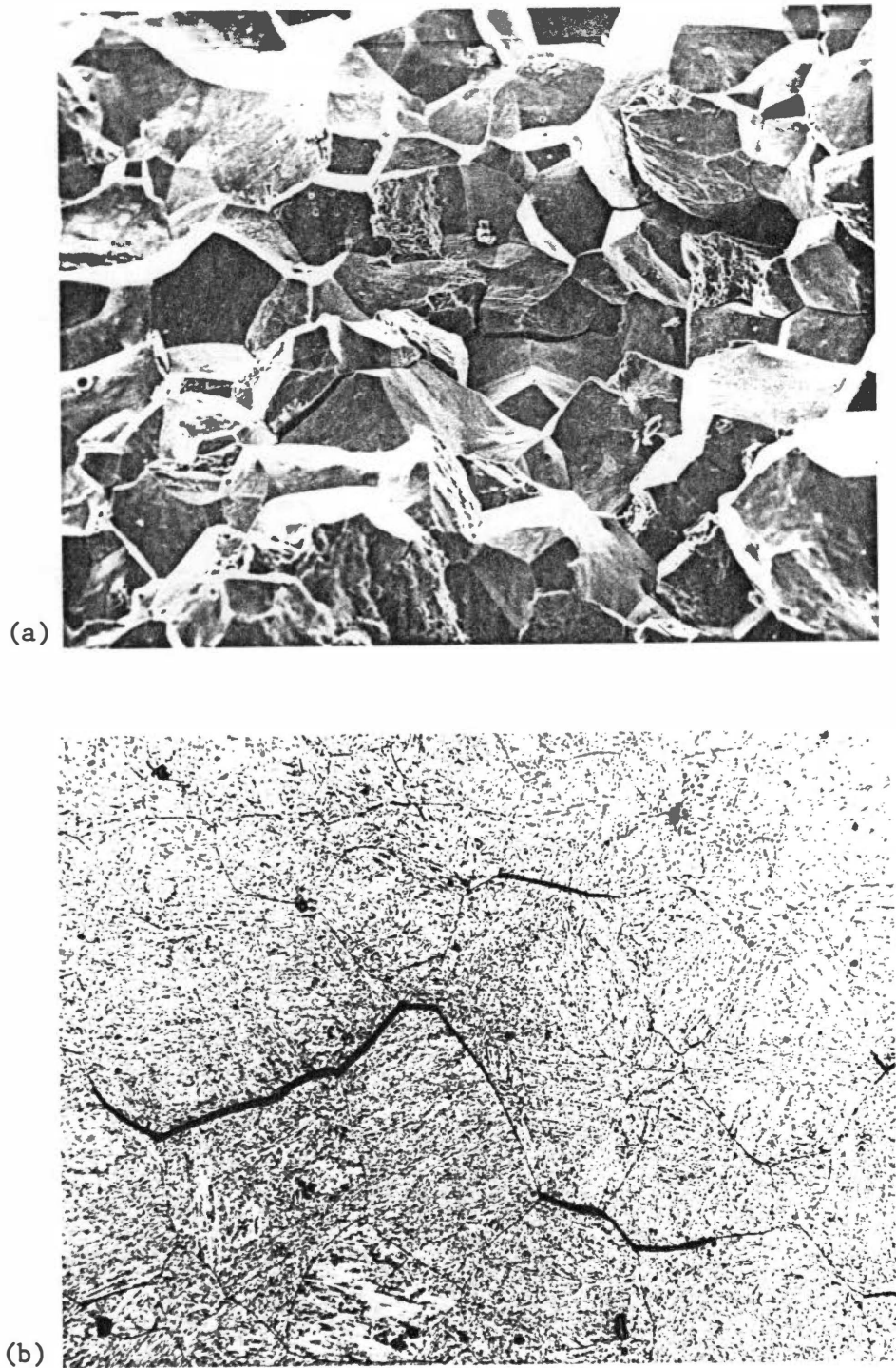


Figure 70. Fractographs of Lukens 2 1/4Cr-1Mo-0.10C SRC specimen, 1150°F (620°C). 75.0 Ksi (517 Mpa), time to fracture 11.0 min, 14.7% RA. (a) SEM, 200X, (b) Metallographic cross section of (a) showing secondary cracking. OLM. 200X. Picral.

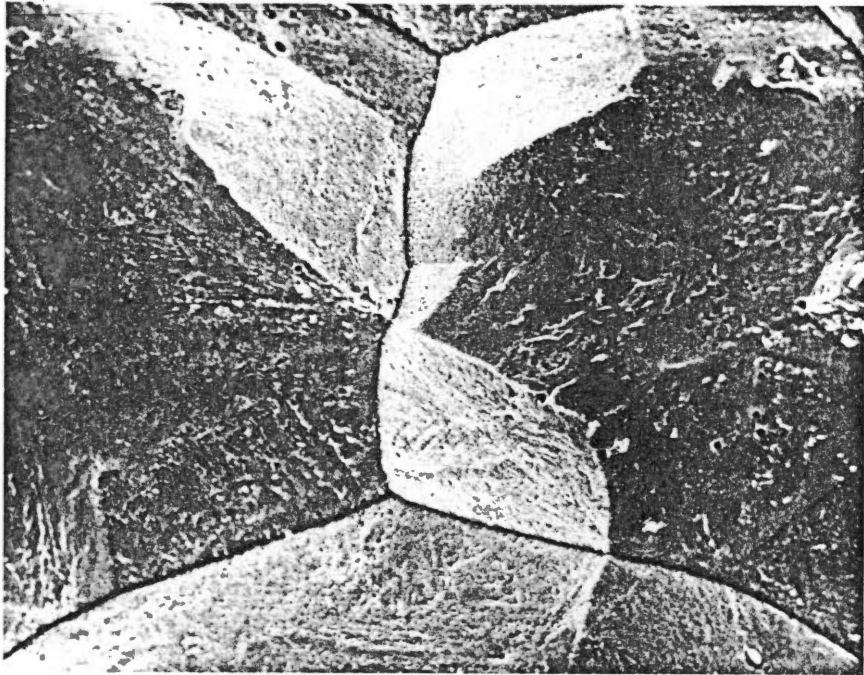


Figure 71. Higher magnification view of Lukens 2 1/4Cr-1Mo-0.10C SRC specimen, 1150°F (620°C). SEM. 1000X.

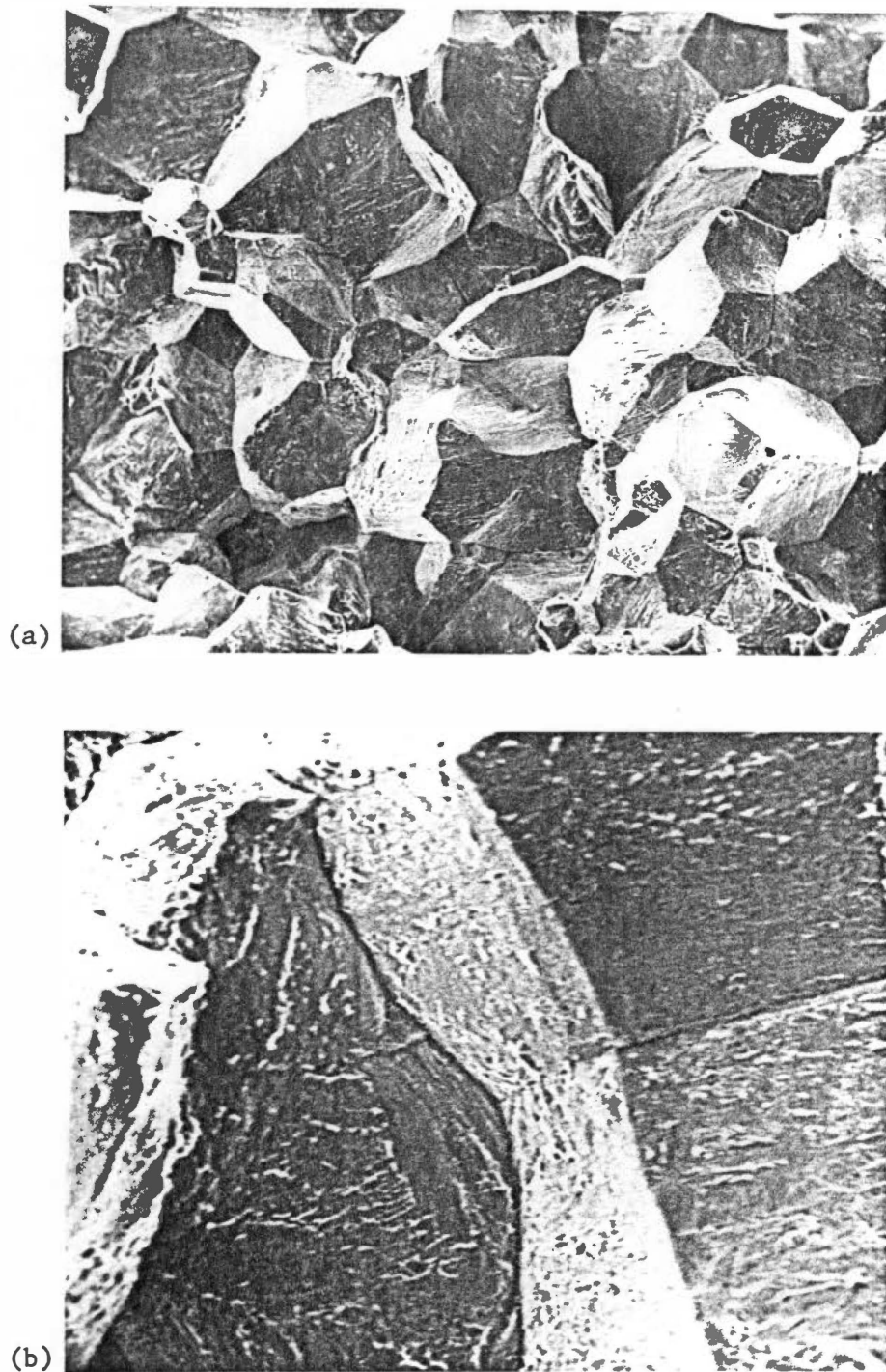


Figure 72. Fractographs of Kawasaki 2 1/4Cr-1Mo SRC specimen, 1250°F (680°C). 38.9 Ksi (268 Mpa), time to fracture 11.4 min, 22.1% RA. SEM, (a) 200X, (b) 1000X.

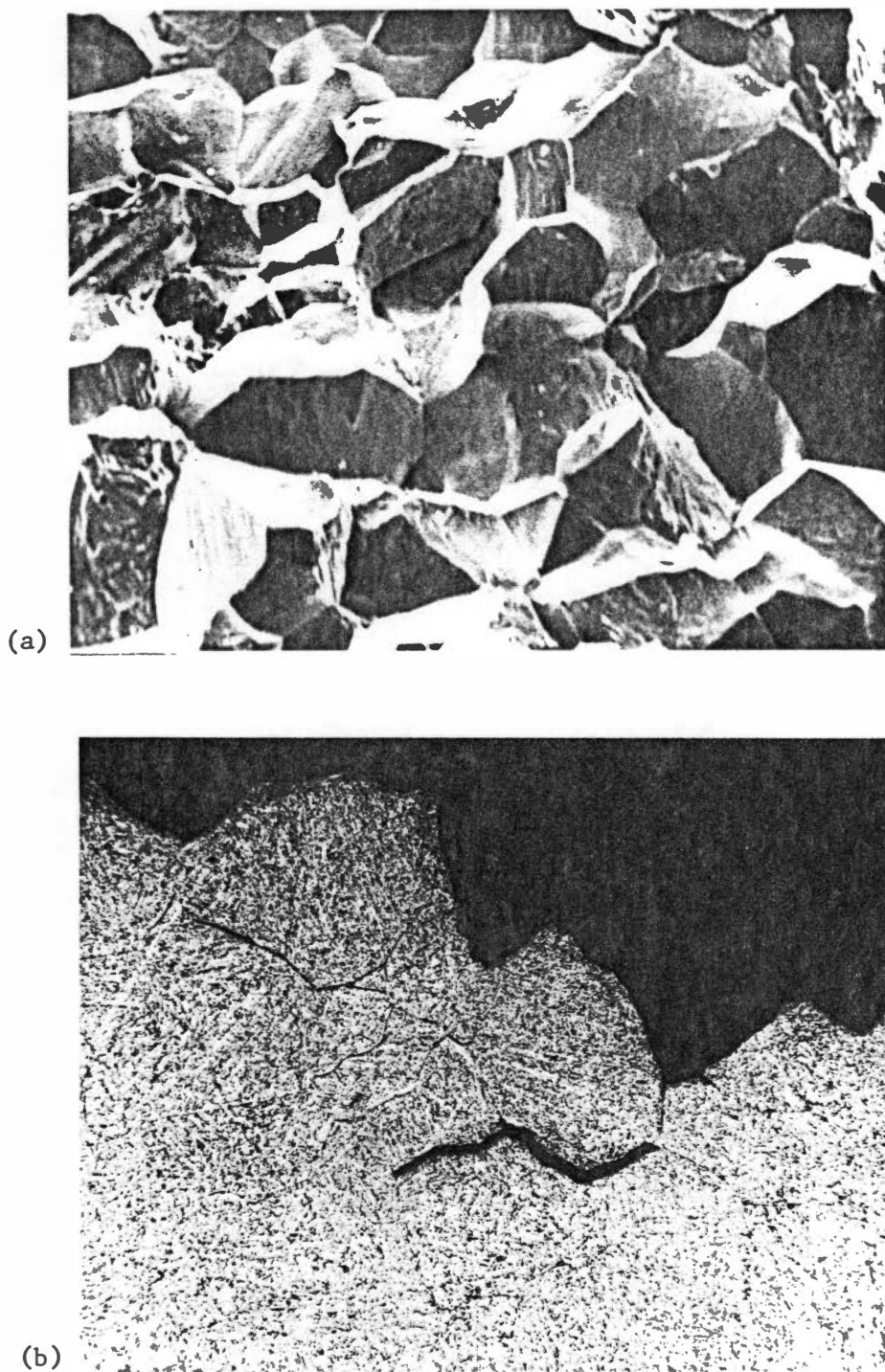


Figure 73. Fractographs of Kawasaki 2 1/4Cr-1Mo-0.12C SRC specimen, 1150°F (620°C). 70.0 Ksi (482 Mpa), time to fracture 5.0 min, 7.1% RA. (a) SEM, 200X. (b) Metallographic cross section of (a) showing secondary cracking. OLM. 200X. Picral.

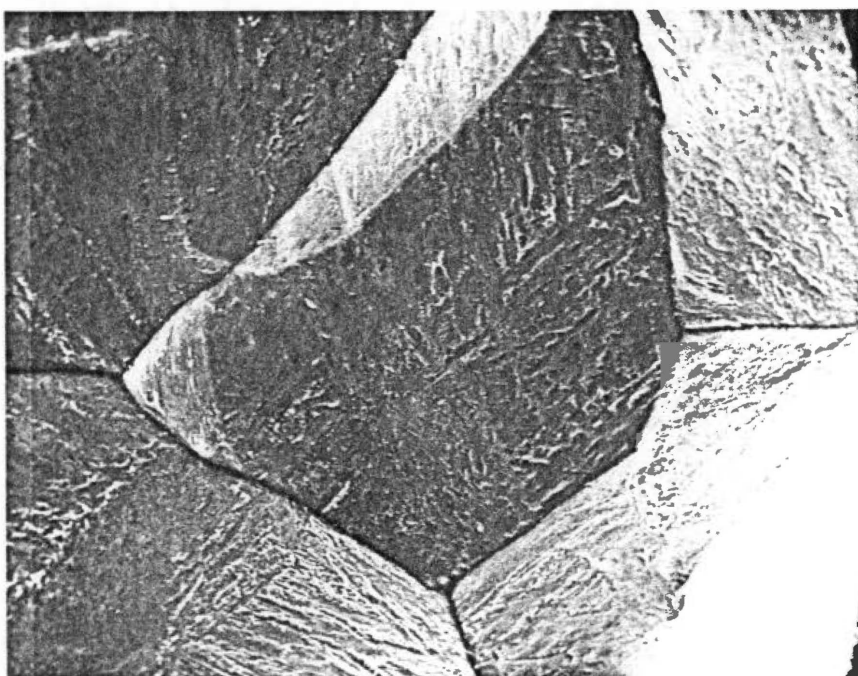


Figure 74. Higher magnification view of Kawasaki 2 1/4Cr-1Mo-0.12C SRC specimen, 1150°F (620°C). SEM. 1000X.

fracture mode is described as intergranular microvoid coalescence. Comparing the Climax 3Cr-1.5Mo (Figure 63) with the Lukens 2 1/4Cr-1Mo-0.12C (Figure 66), the Climax 3Cr-1.5Mo material exhibits the greatest proportion of grain boundary ductility. For the lower PWHT temperature (1150°F (620°C)), only the Climax 3Cr-1.5Mo material exhibits the microvoid coalescence fracture mode (Figures 64 and 65), where as the Lukens 2 1/4Cr-1Mo-0.12C exhibits a low-ductility intergranular fracture mode which is dominated by "smooth" intergranular facets (Figures 67 and 68). The Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C specimens exhibit the low ductility fracture mode at both PWHT temperatures (Figures 69-74).

The two distinct SRC modes observed in this investigation have been observed in 2 1/4Cr-1Mo by other investigators (46,47). Evidence from these two investigations suggest that the low-ductility intergranular fracture mode is associated with grain boundary enrichment by P. Although energy dispersive x-ray analysis (EDAX) showed no evidence of P, Cu, Sn, As, Sb or Al segregation, the possibility of grain boundary embrittlement by residual elements cannot be ruled out because the depth of the segregate region may be less than the beam excitation depth of the EDAX apparatus. Auger spectroscopy is a technique potentially capable of surface analysis for grain boundary embrittling elements. Clearly, an additional investigation is required to adequately assess the effect of residual grain boundary elements on the SRC behavior of the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials employed in this study.

Hippesley and his co-workers (46,47) have shown that the cavities observed in the intergranular microvoid coalescence mode are nucleated on small manganese sulphide inclusions. It was possible to distinguish small particles within some of the cavities of the 3Cr-1.5Mo fracture surfaces (Figure 75a). The particles are approximately 0.5μ in diameter and have been identified by EDAX as manganese sulphide inclusions (Figures 75b and 75c). While Hippesley has suggested that the very fine facets observed in Figure 66b (p. 139) are nucleated around grain boundary carbides (47), the SEM revealed no particles within the facets. Additional work employing transmission electron microscopy of particles extracted from the fracture surface is necessary to verify the presence of carbides.

Fractography of 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo SRC specimens. Scanning electron microscope fractographs of the duplex (12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo) and martensitic (9Cr-1Mo) materials are shown in Figures 76-81. The fracture features of the duplex and martensitic materials are very ductile in nature and do not exhibit the brittle SRC morphology observed in 3Cr-1.5Mo and 2 1/4Cr-1Mo. The primary mode of fracture in the duplex and martensitic materials is microvoid coalescence. The ductile fracture mode in the 9Cr-2Mo-V-Nb, 9Cr-2Mo and 9Cr-1Mo (Figures 79-81) show considerably larger dimples than those observed in the 12Cr-2Mo-Ni material (Figures 76-78). Both equiaxed and irregular dimples are present in the 12Cr-2Mo-6Ni (Figure 76). Similar fracture features

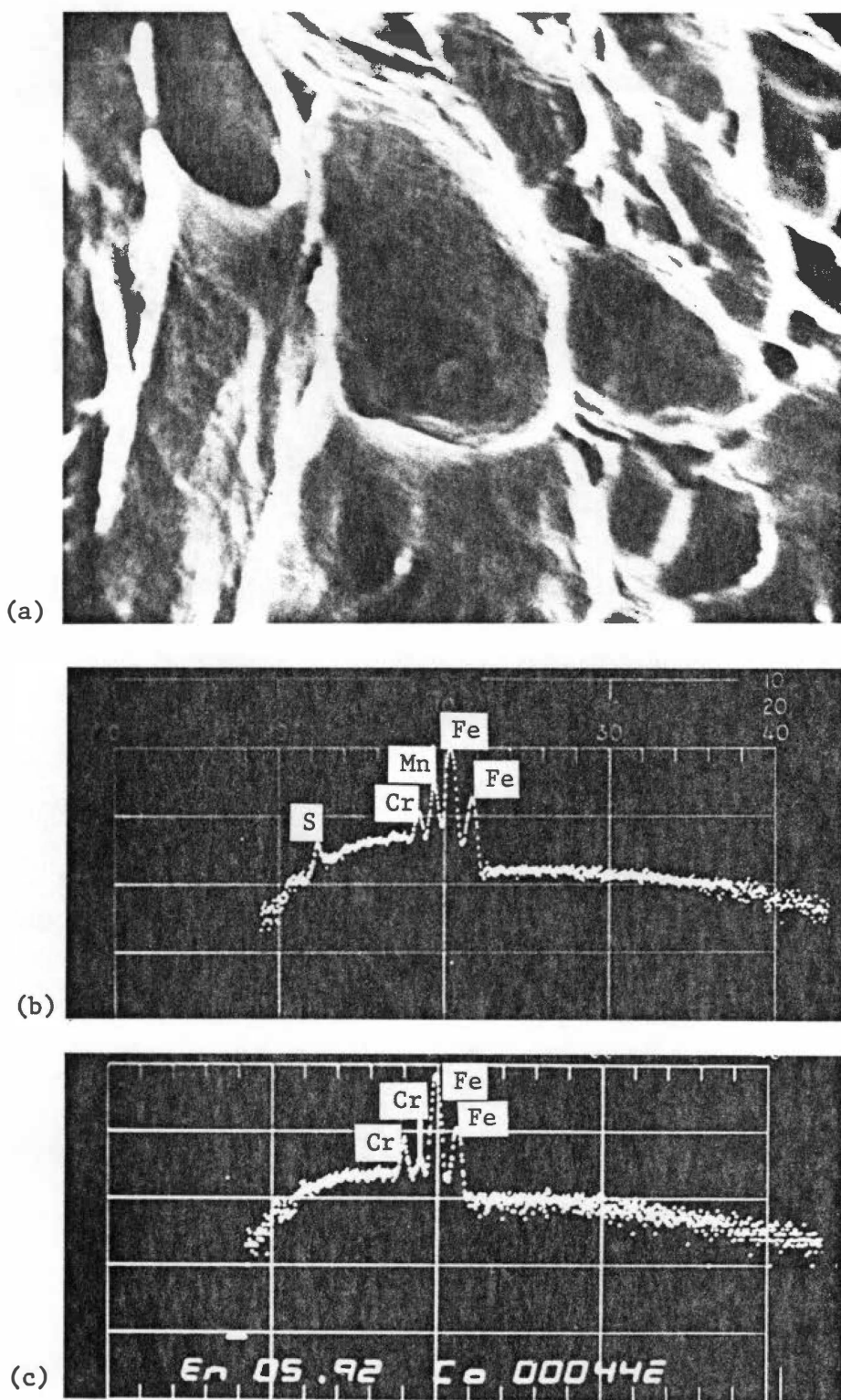


Figure 75. Inclusion located within the dimple cavities of Climax 3Cr-1.5Mo SRC specimen. (a) SEM, 10,000X, (b) EDAX spectrum of inclusion, (c) EDAX spectrum of region adjacent to inclusion (as shown in (a)).

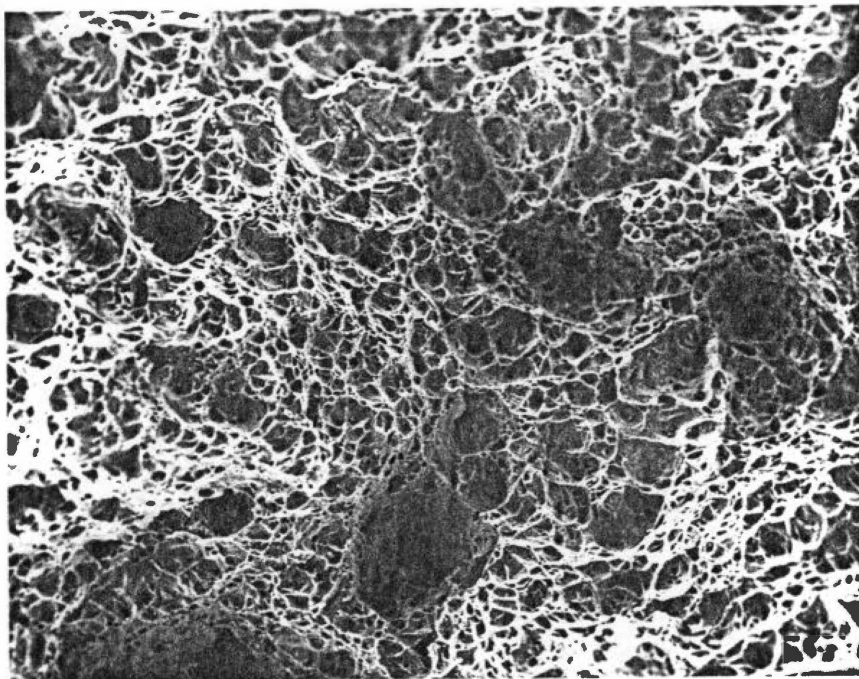


Figure 76. Fractograph of 12Cr-2Mo-6Ni SRC specimen, 1290°F (700°C). 35.8 Ksi (247 Mpa), time to fracture 37.3 min, 69% RA. SEM. 200X.

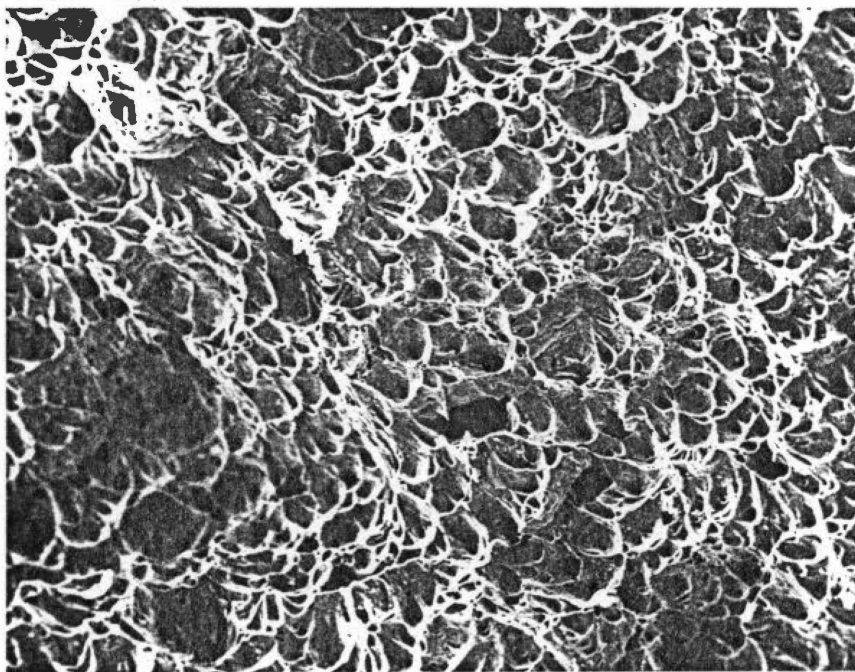


Figure 77. Fractograph of 12Cr-2Mo-3Ni SRC specimen, 1290°F (700°C). 25.0 Ksi (172 Mpa), time to fracture 12.0 min, 84% RA. SEM. 200X.

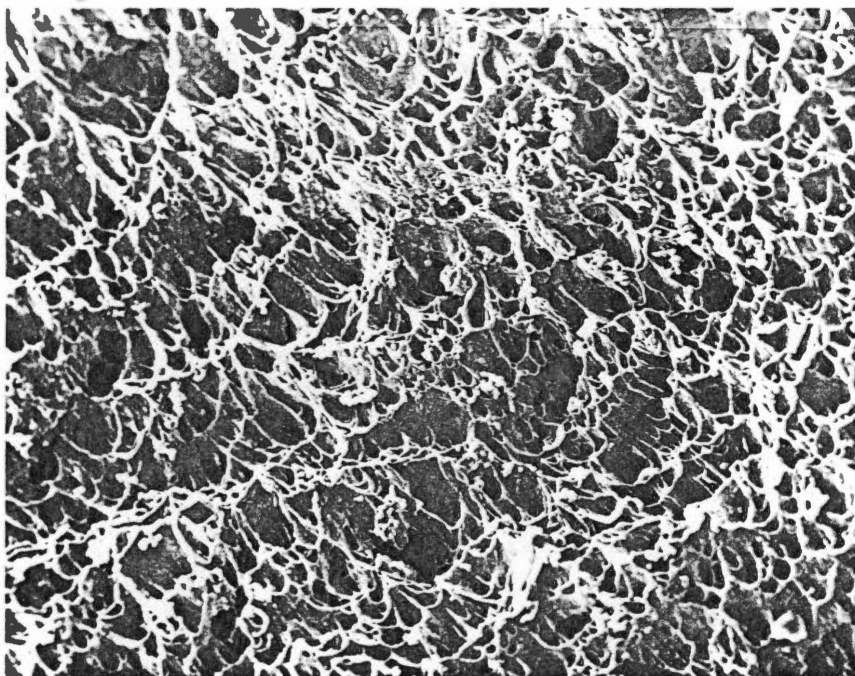


Figure 78. Fractograph of 12Cr-2Mo-1Ni SRC specimen, 1290°F (700°C). 27.0 Ksi (186 Mpa), time to fracture 8.0 min, 80% RA. SEM. 200X.

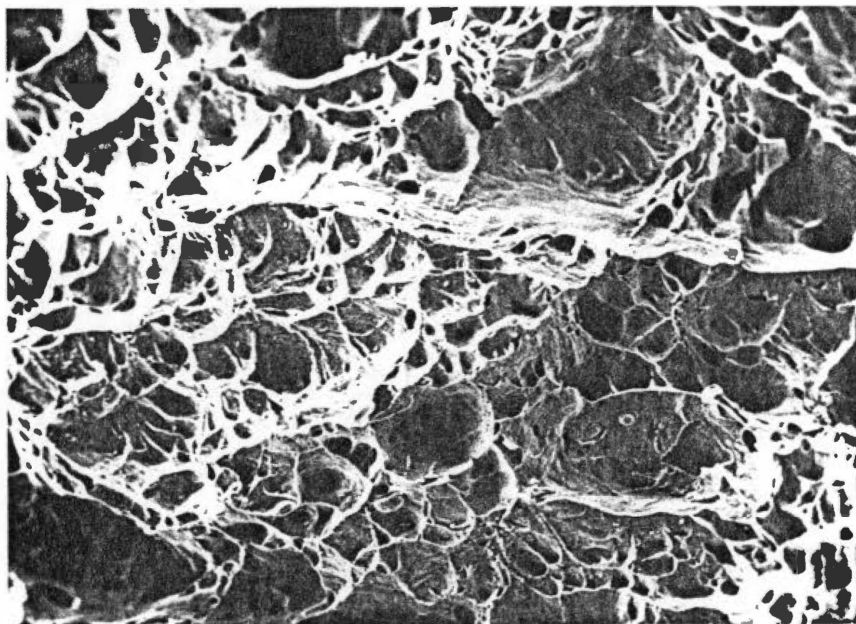


Figure 79. Fractograph of 9Cr-2Mo-V-Nb SRC specimen, 1350°F (730°C). 24.4 Ksi (168 Mpa), time to fracture 9.3 min, 81% RA. SEM. 200X.

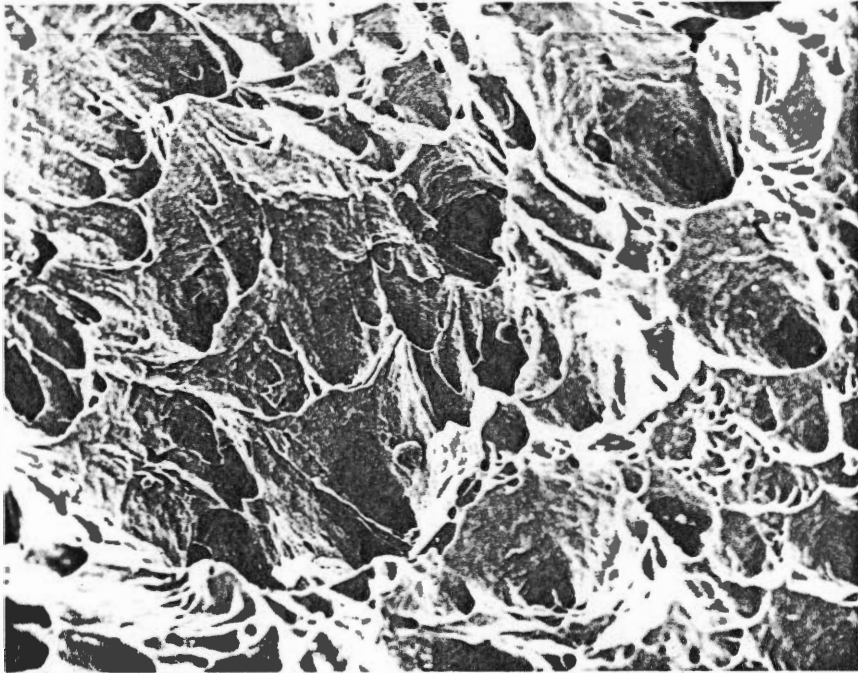


Figure 80. Fractograph of 9Cr-2Mo SRC specimen, 1350°F, (730°C). 15.0 Ksi (103Mpa), time to fracture 40.5 min, 89% RA. SEM. 200X.

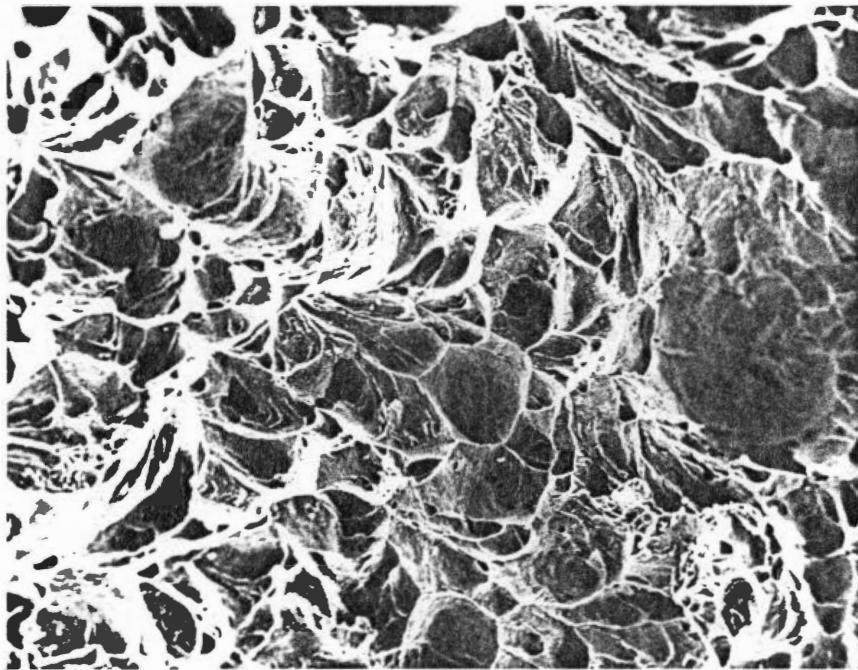


Figure 81. Fractograph of 9Cr-1Mo-0.11C SRC Specimen, 1350°F (730°C). 11.4 Ksi (56 Mpa), time to fracture 43.0 min, 79% RA. SEM. 200X.

and similar rupture ductilities were observed in 9Cr-1Mo-V-Nb(176) and 9Cr-1Mo-0.06C when they were subjected to the stress-rupture test method (72).

Summary of SRC Behavior of Cr-Mo Steels

In this investigation, the relative SRC susceptibility of the coarse grained HAZ of the duplex (12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo), martensitic (9Cr-1Mo) and bainitic (3Cr-1Mo and 2 1/4Cr-1Mo) subclasses of Cr-Mo steels was determined. The relative SRC susceptibility of the three subclasses of Cr-Mo steels was based on the rupture-strength, rupture ductility and fracture morphology of stress-rupture Gleeble simulated HAZ SRC specimens. When subjected to the stress-rupture Gleeble simulation, materials susceptible to SRC exhibit rupture ductilities below 10% RA and display an intergranular fracture mode.

The SRC susceptibility of Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C was evaluated at 1250°F (680°C) and 1150°F (620°C) PWHT. For 1250°F (680°C), only the Lukens 2 1/4Cr-1Mo-0.12C material exhibited fracture features and rupture ductilities associated with SRC. All four bainitic materials showed a sharp decrease in rupture-ductility when the PWHT was reduced to 1150°F (620°C), indicating an increase in SRC susceptibility. For 1150°F (620°C) PWHT, all four bainitic materials exhibited the fracture mode and rupture-ductilities associated with SRC. The increase in SRC susceptibility at 1150°F

(620°C) is probably the result of an increase in the amount of grain matrix strengthening (formation of coherent M_2C (Mo_2C) type precipitates) which occurs at 1150°F (620°C) PWHT.

The stress-rupture parameter (SRP) was employed to establish the relative SRC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo on a material to material basis. The SRP is the product of the stress for a rupture time of 10 minutes and the corresponding reduction in area. A high value of the SRP is indicative of good stress-rupture strength and ductility, consequently a low susceptibility to SRC is expected. The SRP reveals that at 1250°F (680°C) and 1150°F (620°C) PWHT the Climax 3Cr-1.5Mo is the material least susceptible to SRC, and the Lukens 2 1/4Cr-1Mo-0.12C is the material most susceptible to SRC. For both PWHT temperatures, the relative SRC susceptibility of the Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C is in the intermediate range of that determined for the Climax 3Cr-1.5Mo and Lukens 2 1/4Cr-1Mo-0.12C materials.

It is possible to attribute the material to material difference in the relative SRC susceptibility of the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials to the interaction of elemental species which alter carbide precipitation (C, Cr, Mo, Si and V) and those which encourage grain boundary degradation (P, S, As, Sb, Al and Si). Any combination of alloying elements which increase grain matrix strength and/or reduce grain boundary strength, enhances the SRC susceptibility.

Apparently, the Climax 3Cr-1.5Mo is the bainitic material least susceptible to SRC because the Cr level present in this material

imparts increased ductility to the grain matrix. It is possible that the Lukens 2 1/4Cr-1Mo-0.12C exhibits the greatest SRC susceptibility because this material contains the greatest amount of C (0.12%) and Si (0.22%). Carbon and Si both encourage the formation of Mo_2C carbides which strengthen the grain matrix. The formation of Mo_2C carbides release embrittling elements from the grain matrix so that they might embrittle the grain boundaries. In addition to enhancing carbide formation, Si acts to reduce the strength of the grain boundaries by forming regions which are denuded of precipitates. The total C plus Si content of the Lukens 2 1/4Cr-1Mo-0.10C (0.10C, 0.19Si) and Kawasaki 2 1/4Cr-1Mo-0.12C (0.12C, 0.10Si) is less than the total C plus Si content of the Lukens 2 1/4Cr-1Mo-0.12C (0.12C, 0.22Si), thus the Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C materials are less susceptible to SRC.

The SRC susceptibility of the duplex (12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb and 9Cr-2Mo) and martensitic (9Cr-1Mo) subclasses of Cr-Mo steels was evaluated at commonly employed PWHT temperatures. The rupture ductilities of the duplex and martensitic materials were considerably greater than those observed in the bainitic materials (60 to 90% RA vs 1.7 to 36% RA). The difference in the rupture ductilities of these three classes of Cr-Mo materials is apparently related to the difference in the tempering response of the martensite plus delta-ferrite microstructure and the bainite microstructure. During PWHT, the bainitic materials are strengthened by the formation of coherent M_4C_3 (V_4C_3) and M_2C (Mo_2C) precipitates, resulting in a decrease in

rupture ductility. Only small amounts of matrix strengthening precipitates form in the duplex and martensitic materials resulting in an increase in the rupture ductility. The greater stress-rupture ductility observed in the duplex and martensitic materials represents a lower SRC susceptibility. Fractographic examinations of the duplex and martensitic materials revealed no mode of fracture which would characterize the duplex and martensitic materials as being susceptible to SRC under the welding and PWHT conditions simulated in this investigation.

In the prior discussion of CCT behavior, it was shown that the transformation behavior of Cr-Mo steels could be categorized in terms of various subclasses of Cr-Mo steels. From the prior discussion of SRC, it is evident that the SRC susceptibility can also be categorized in terms of different subclasses of Cr-Mo materials. For the welding and PWHT conditions examined in this examination, the duplex and martensitic classes of Cr-Mo steels exhibited no indication of SRC susceptibility. In contrast, the bainitic Cr-Mo steels displayed fractographic and rupture-ductility characteristics which describe them as being susceptible to SRC. The degree of SRC susceptibility in the bainitic class of Cr-Mo steels was shown to be strongly influenced by the PWHT temperature.

During the course of this investigation, an attempt was made to explain how differences in composition and PWHT temperature affect the SRC susceptibility of Cr-Mo steels. The results of this investigation suggest that composition and PWHT temperature influence

the SRC susceptibility by: (1) controlling the precipitation process, which during PWHT, strengthens the grain matrix and (2) controlling the degradation process which, during PWHT, weakens the grain boundary regions. The results of this investigation indicate that a detailed study of the precipitation and degradation processes employing transmission microscopy and Auger surface analysis is needed to adequately assess the affect of PWHT temperature and composition on the SRC susceptibility of Cr-Mo steels.

Perhaps it would also be wise to include a series of direct weldment tests (e.g. Y-groove) in any subsequent SRC studies. A direct weldment test is a test for SRC in which the specimen to be examined is produced by an actual welding process. After being subjected to the normal PWHT, the welded specimen is examined for cracking. The direct weldment evaluation for SRC duplicates the complex interactions of stress and strain during PWHT more closely than the stress-rupture Gleeble simulated HAZ test employed in this investigation. While the results of the stress-rupture Gleeble simulated HAZ test are useful in comparing the relative SRC susceptibility of Cr-Mo steels, data from actual weldment tests are more directly applicable to actual fabrication conditions.

Evaluation of the Hydrogen-Assisted Cracking (HAC)

Susceptibility of Cr-Mo Steels

Hydrogen-assisted cracking (sometimes called cold cracking) is low temperature (below 600°F (315°C)) cracking which results from the

embrittlement of Cr-Mo steel HAZ's by hydrogen and is a function of microstructure and stress state. Hydrogen-assisted cracking (HAC) occurs in the fusion zone and/or HAZ and like SRC, renders a welded structure unacceptable, necessitating repair. Hydrogen-assisted cracking is often delayed in nature, occurring minutes, hours or even days after welding is complete.

The degree of HAC susceptibility in a Cr-Mo steel depends on hydrogen content, composition, microstructure, restraint and preheat. In Cr-Mo steels, the bainitic and/or martensitic microstructure produced during welding exhibits a definite susceptibility to HAC and as a result, HAC is more prevalent over the entire range of Cr-Mo steels than any other form of weld cracking (hot cracking, SRC, lamellar tearing and temper embrittlement). In contrast to hot cracking, SRC and creep embrittlement, which are very sensitive to the residual amounts of elemental species (S, P, Sn, As and Al), HAC is very sensitive to the amount of bulk alloying elements (in particular C, Cr, Ni, Mn and Si).

The object of the HAC evaluation was to determine the effect of hydrogen, composition, microstructure, restraint and preheat on the weldability of Cr-Mo steels ranging in Cr content from 2 1/4-12%. In this investigation, the UT-Modified Hydrogen Sensitivity Test (UT-Mod HST) was employed to rank the relative HAC susceptibility of 12Cr-2Mo-Ni, 9Cr-2Mo, 9Cr-2Mo-V-Nb, 9Cr-1Mo-V-Nb, 3Cr-1.5Mo and 2 1/4Cr-1Mo. The UT-Mod HST test is a variation of the RPI augmented strain test (68,69,70) and has shown extreme promise in evaluating

the HAC susceptibility of SA508, SA533 and C-Mn-Si cast steels (67,86).

In the UT-Mod HST, an autogenous GTA weld is made along the length of a small test specimen (1 5/8in x 5/8in x 1/8in (42mm x 16mm x 3.2mm)). Hydrogen is added directly to the shielding gas, saturating the fusion zone. After welding, the sample is cooled to room temperature and strained in a fixture for which die blocks are available to produce nominal augmented strain levels of 1%, 2%, and 4%. Cracking susceptibility is determined by observing if cracks form on the specimen face strained in tension. Cracking is evaluated as a function of preheat, strain and straining temperature. For any particular strain level, the preheat level for which at least two tests remain crack free for 24 hours is designated as the preheat required to prevent cracking. The UT-Mod HST test and its associated procedures are described in greater detail in Chapter III.

It is important to note that the object of the UT-Mod HST is to produce a relative ranking of HAC based on strain, hydrogen content, microstructure and preheat. The preheat determined by the UT-Mod HST is based on relatively high levels of hydrogen (the weld fusion zone is saturated) and restraint, and as a result, the preheat determined by the UT-Mod HST is conservative in nature and is generally higher than that commonly employed in welded fabrication.

Evaluation of the HAC Susceptibility of the 12Cr-2Mo-Ni Materials

The UT-Mod HST results for the 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni materials are summarized in Table 12. The three

Table 12. Summary of UT-Mod HST Results for 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni.

Material	Preheat °F (°C)	Nominal Strain	Result
12Cr-2Mo-6Ni	75 (24)	4%	No H ₂ , no cracking
12Cr-2Mo-6Ni	75 (24)	4%	Cracked on straining
12Cr-2Mo-6Ni	750 (400)	1%	Cracked 8 minutes
12Cr-2Mo-3Ni	75 (24)	4%	No H ₂ , no cracking
12Cr-2Mo-3Ni	750 (400)	1%	Cracked 6 minutes
12Cr-2Mo-1Ni	75 (24)	4%	No H ₂ , no cracking
12Cr-2Mo-1Ni	75 (24)	1%	Cracked 38 minutes
12Cr-2Mo-1Ni	750 (400)	1%	Cracked 14 hours

12Cr-2Mo-Ni materials exhibited delayed cracking when welded at 750°F (400°C) preheat and strained at 75°F (24°C) to 1%. No cracks were observed in the specimens welded with 100% argon (0% hydrogen) shielding gas and strained to 4% at 75°F (24°C). The fact that the specimens cracked in a delayed manner only in the presence of hydrogen demonstrates the role that hydrogen plays in the cracking behavior of the three 12Cr-2Mo-Ni materials.

Metallographic evaluation of 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni UT-Mod HST specimens. A metallographic evaluation of the 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni UT-Mod HST specimens was undertaken to determine the effect of microstructure on the observed cracking behavior. Photomicrographs of the UT-Mod HST specimens which are transverse and longitudinal to the welding direction are shown in Figures 82-87. The weld fusion zone microstructure of the 12Cr-2Mo-6Ni material (Figures 82 and 83) contains martensite, ferrite and possibly austenite, while the fusion zone of the 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni materials (Figures 84-87) is predominately martensite and ferrite.

Three distinct ferrite morphologies can be observed in the fusion zone of the three 12Cr-2Mo-Ni materials. In the 12Cr-2Mo-6Ni material (Figures 82 and 83), the ferrite is located along prior solidification subboundaries. The ferrite observable in the 12Cr-2Mo-1Ni material (Figures 86 and 87) is in the form of large ferrite islands surrounded by an almost continuous band of martensite.

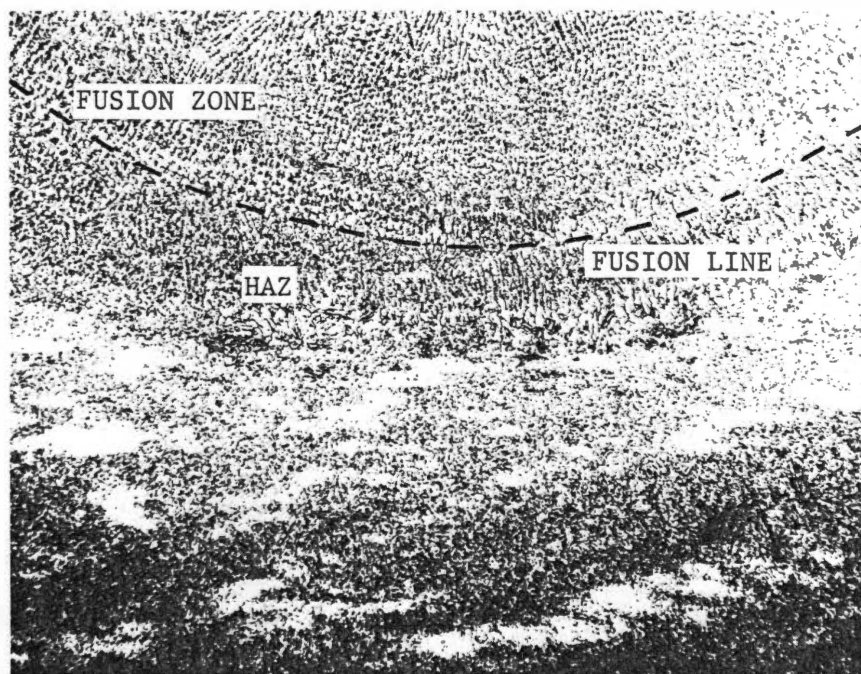


Figure 82. Transverse view of 12Cr-2Mo-6Ni UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified HCl-Picral.

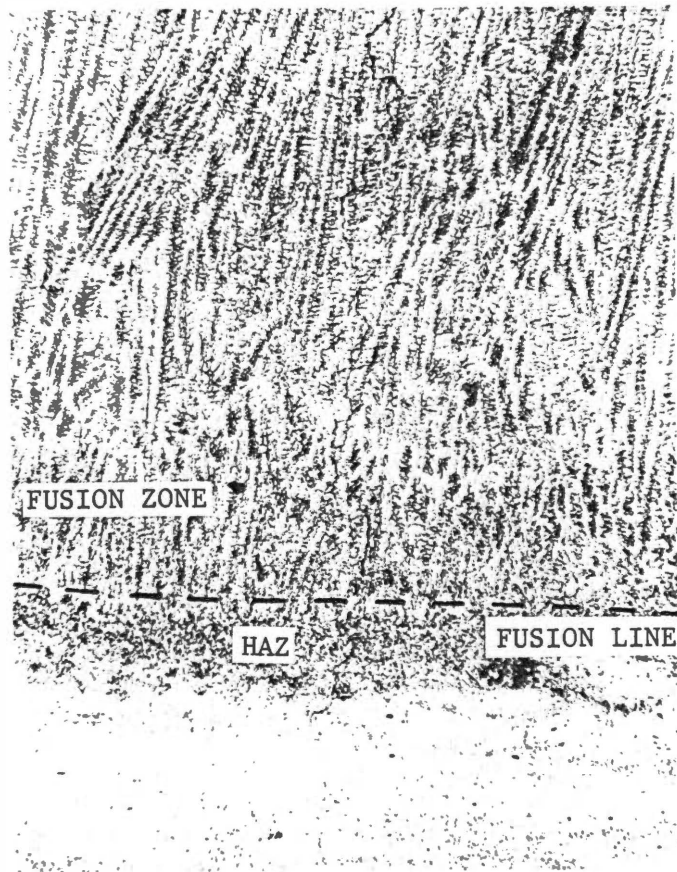


Figure 83. Longitudinal view of 12Cr-2Mo-6Ni UT-Mod HST specimen. 750°F (400°C) preheat, 1% nominal strain at 75°F (24°C). OLM. 100X. Modified HCl-Picral.

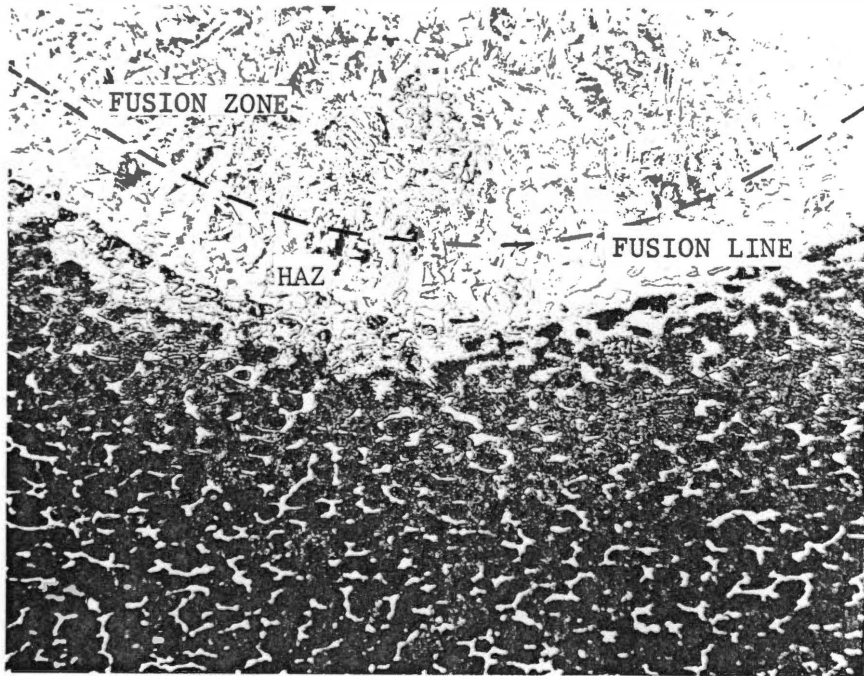


Figure 84. Transverse view of 12Cr-2Mo-3Ni UT-Mod HST specimen.
75°F (24°C) preheat. OLM. 100X. Modified HCl-Picral.



Figure 85. Longitudinal view of 12Cr-2Mo-3Ni UT-Mod HST specimen. 750°F (400°C) preheat, 1% nominal strain at 75°F (24°C). OLM. 100X. Modified HCl-Picral.

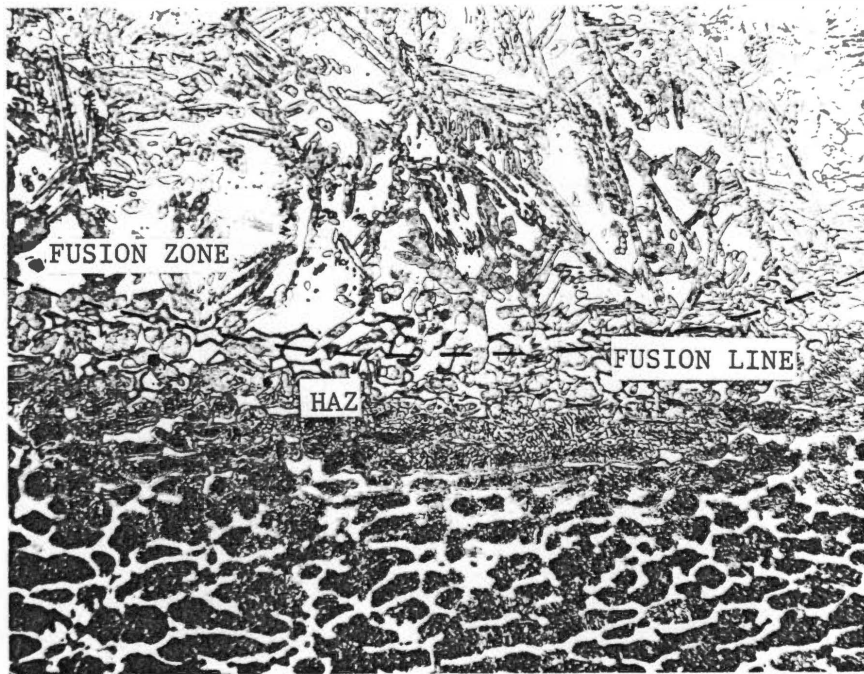


Figure 86. Transverse view of 12Cr-2Mo-1Ni UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified Wazau's.

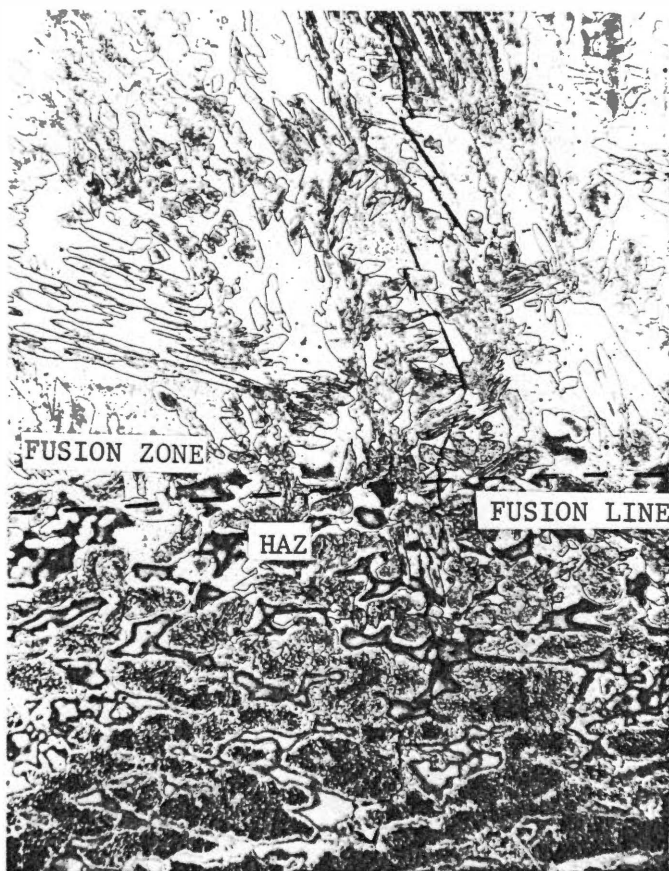


Figure 87. Longitudinal view of 12Cr-2Mo-1Ni UT-Mod HST specimen. 750°F (400°C) preheat, 1% nominal strain at 75°F (24°C). OLM. 100X. Modified Wazau's.

The ferrite in the 12Cr-2Mo-3Ni (Figures 84 and 85) exhibits an angular morphology.

The HAZ region immediately adjacent to the fusion zone in the 12Cr-2Mo-6Ni material is predominately martensitic, while this same HAZ region in the 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni materials is martensitic/ferritic. The HAZ in the 12Cr-2Mo-1Ni contains only martensite and delta-ferrite as was evidenced by the detectable on-cooling transformation finish observed in the HAZ CCT evaluation. The HAZ CCT evaluation of the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni materials suggests that austenite may be retained in the HAZ, however, no metallographic evidence of austenite was observed in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni UT-Mod HST specimens. It was pointed out in the discussion on transformation behavior that austenite may not be metallographically detectable in the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni HAZ because it is present in quantities of less than 15%. It is apparent that an additional investigation employing direct x-ray analysis or dark field electron microscopy is needed to resolve the presence of austenite in the weld fusion zone and HAZ of 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni.

Cracks which are representative of those observed in the three 12Cr-2Mo-Ni materials are shown in Figures 83, 85 and 87. In general, the cracks are perpendicular to the direction of the maximum tensile stress on the surface of the specimen. The cracks traverse both the HAZ and fusion zone cutting across both ferrite and martensite regions. The transgranular nature of the cracks are characteristic of HAC.

Summary of the relative HAC susceptibility of 12Cr-2Mo-Ni. In the UT-Mod HST the 12Cr-2Mo-6Ni, 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni materials exhibited HAC cracking even at the preheat and strain conditions most favorable to prevent HAC (750°F (400°C) preheat, 1% nominal strain). This type of cracking behavior has also been observed during the UT-Mod HST evaluation of 12Cr-2Mo-0.2C-V-W(HT-9) (72). The 12Cr-2Mo-0.2C-V-W material is known to be very susceptible to HAC (preheat temperatures in the range of 600°F (315°C) are employed during fabrication). It is apparent that under the conditions imposed by the UT-Mod HST (high hydrogen content and high level of restraint), the 12Cr-2Mo-Ni materials also reflect a high degree of HAC susceptibility.

The high degree of HAC susceptibility observed in the 12Cr-2Mo-Ni materials may be related to the on-cooling transformation behavior. The range of the on-cooling transformation in the 12Cr-2Mo-Ni materials provides little opportunity for autotempering or self-diffusion of hydrogen, thus the cracking susceptibility is increased. This problem is compounded for the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni materials in which the M_f temperature is below 75°F (24°C).

Evaluation of the HAC Susceptibility of the 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo

The UT-Mod HST results for the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials are summarized in Figure 88. In this figure, the preheat required to prevent cracking is plotted as a function of

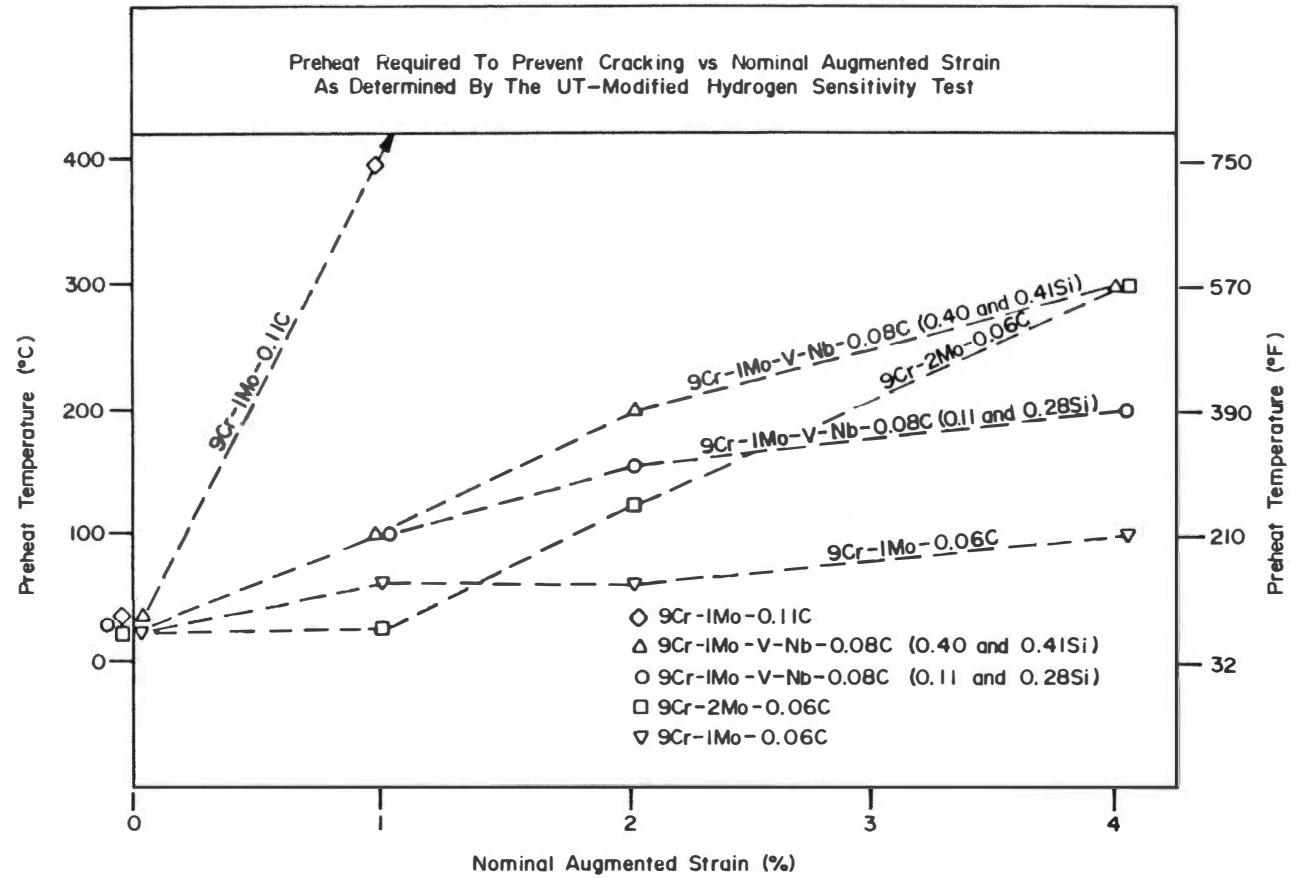


Figure 88. Summary of UT-Mod HST for 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo showing the effect of strain on the preheat required to prevent cracking.

nominal augmented strain at 32°F (0°C). Straining at 32°F (0°C) provides a conservative estimate of the relative HAC susceptibility. From the results shown in Figure 88, it is apparent that as the augmented strain increased, the preheat required to prevent cracking in the 9Cr-1Mo-V-Nb, 9Cr-1Mo-0.11C and 9Cr-1Mo-0.06C materials increased as well.

The most conservative estimate of the relative HAC susceptibility of the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials is obtained by comparing the level of preheat required to prevent cracking when strained to 4% at 32°F (0°C). Based on this approach, the 9Cr-1Mo-0.11C material can be considered the material most susceptible to HAC because a preheat in excess of 750°F (400°C) is required to prevent cracking. For 4% strain, the cracking susceptibility of the higher Si content 9Cr-1Mo-V-Nb materials (383,384) and the 9Cr-2Mo material is comparable, since a preheat of 570°F (300°C) is required to prevent cracking in both materials. The lower Si content 9Cr-1Mo-V-Nb materials (148,176) require a preheat of 390°F (200°C) to prevent cracking and thus, are less susceptible to cracking than both the higher Si content 9Cr-1Mo-V-Nb and the 9Cr-2Mo. The 9Cr-1Mo-0.06C material is the least susceptible to cracking requiring a preheat level of 210°F (100°C) to prevent cracking.

The 9Cr-2Mo, 9Cr-1Mo-V-Nb, and 9Cr-1Mo materials exhibited no cracks when welded at 75°F (24°C) with low hydrogen (100% argon shielding gas) and strained to 4%. This confirms that the cracking behavior observed in these materials is hydrogen related. In contrast,

after welding with 100% argon, the 9Cr-2Mo-V-Nb exhibited cracking during straining to 1% at 75°F (24°C). It is apparent that the 9Cr-2Mo-V-Nb is more sensitive to weld cracking than the 12Cr-2Mo-Ni, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials evaluated in this investigation.

Metallographic evaluation of 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo UT-Mod HST specimens. Metallographic studies of the fusion zone and HAZ of the 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo were conducted to determine if any structural relationships exist between the susceptibility to HAC and observable microstructural variations. Photomicrographs of the UT-Mod HST specimens which are transverse and longitudinal to the welding direction are shown in Figures 89-98.

The fusion zone of the 9Cr-2Mo-V-Nb material contains massive amounts of delta-ferrite (Figures 89 and 90). The delta-ferrite morphology consists of large islands surrounded by an almost continuous band of martensite. The Widmanstätten morphology of the martensite in the fusion zone and HAZ is the result of a partial high temperature transformation of delta-ferrite to austenite. The austenite present in the HAZ during cooling transforms to martensite.

A photomicrograph of a crack representative of those observed in the 9Cr-2Mo-V-Nb material is shown in Figure 90. This crack is primarily confined to the fusion zone and is oriented in the direction of the solidification substructure. A crack of this nature would generally be attributed to hot cracking, however, SEM

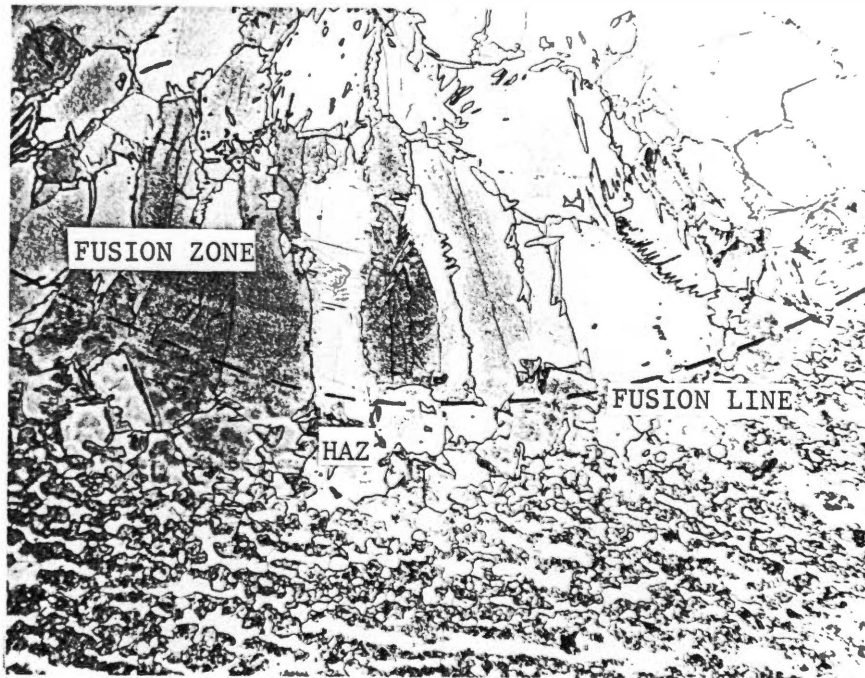


Figure 89. Transverse view of 9Cr-2Mo-V-Nb UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified HCl-Picral.

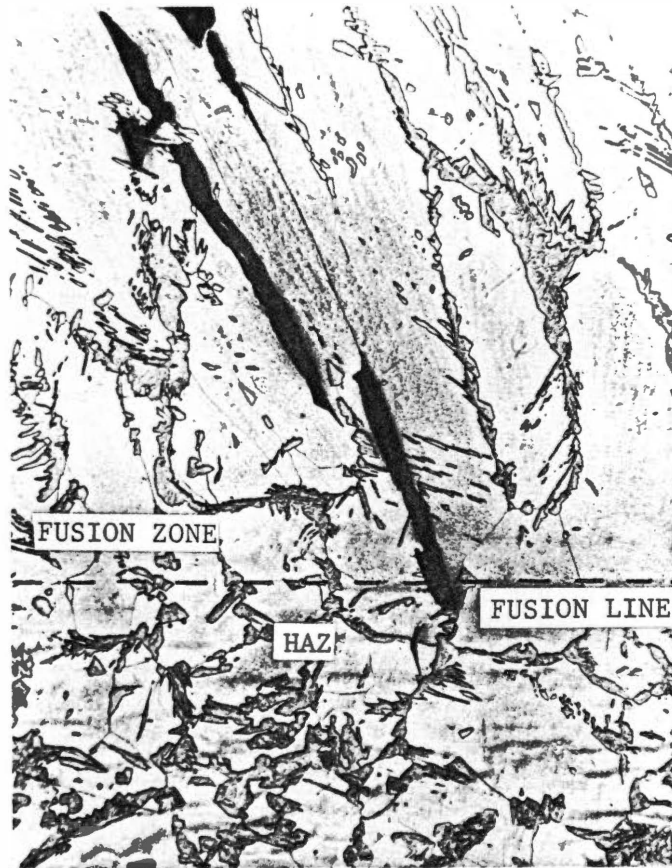


Figure 90. Longitudinal view of 9Cr-2Mo-V-Nb UT-Mod HST specimen. 570°C (300°C) preheat, 4% nominal strain at 75°F (24°C). OLM. 100X. Modified HCl-Picral.

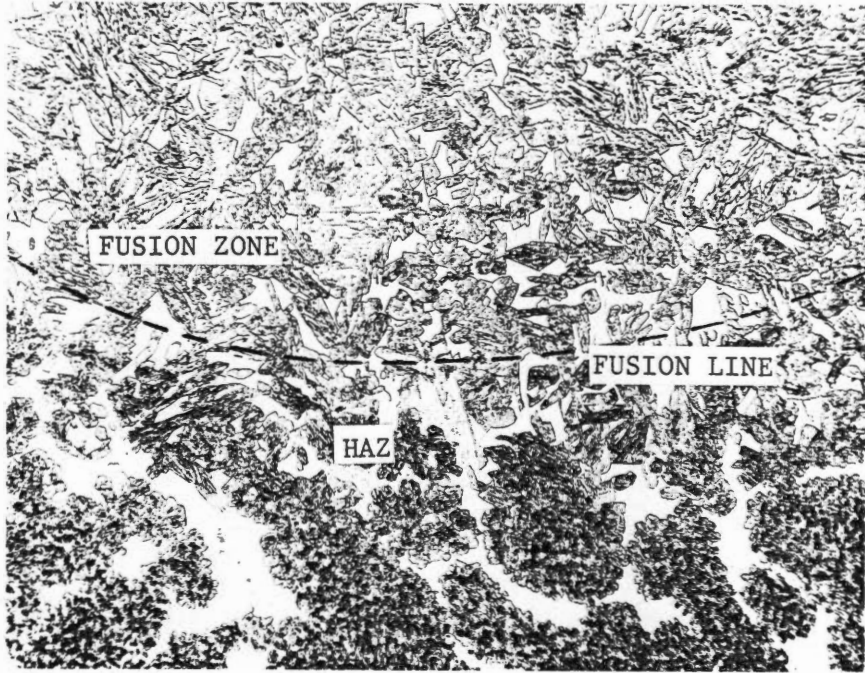


Figure 91. Transverse view of 9Cr-2Mo UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified Wazau's.

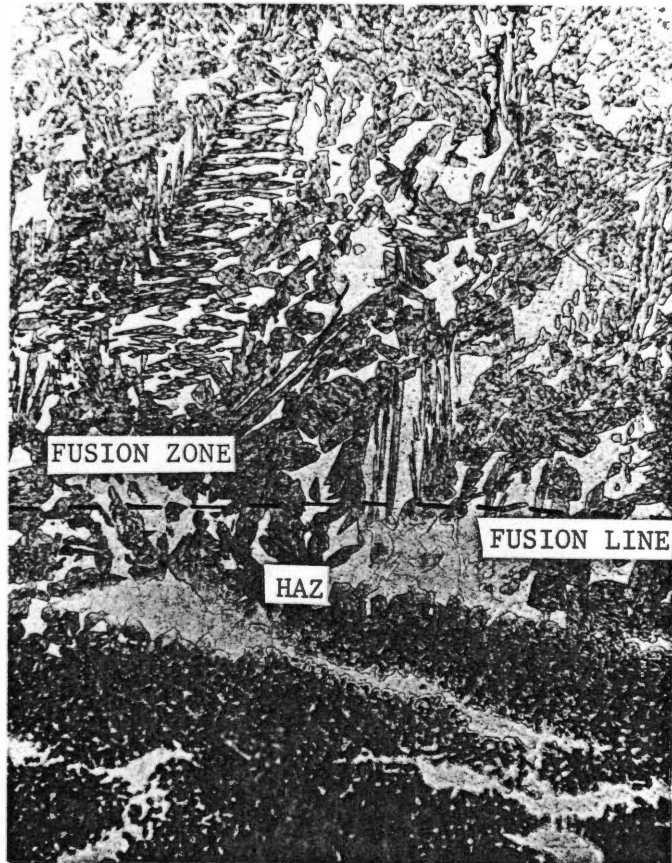


Figure 92. Longitudinal view of 9Cr-2Mo UT-Mod HST specimen. 75°F (24°C) preheat, 2% nominal at 75°F (24°C). OLM. 100X. Modified Wazau's.

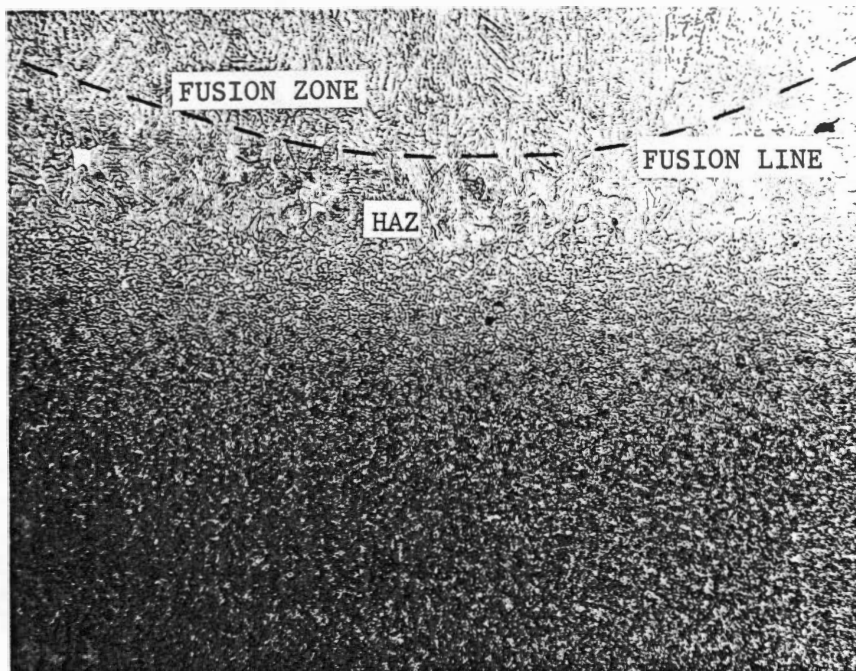


Figure 93. Transverse view of 9Cr-1Mo-V-Nb (176) UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified HCl-Picral.

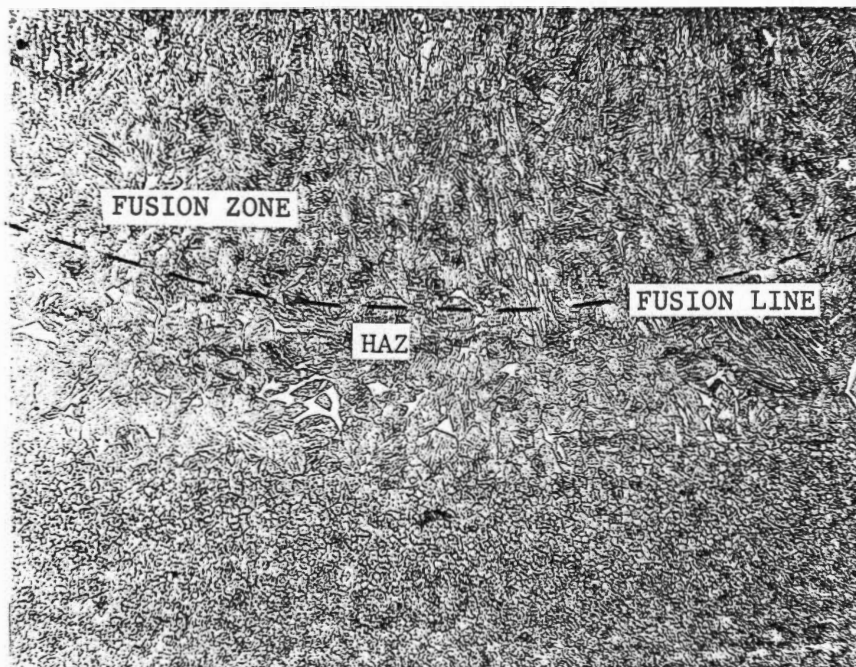


Figure 94. Transverse view of 9Cr-1Mo-V-Nb (394) UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified HCl-Picral.

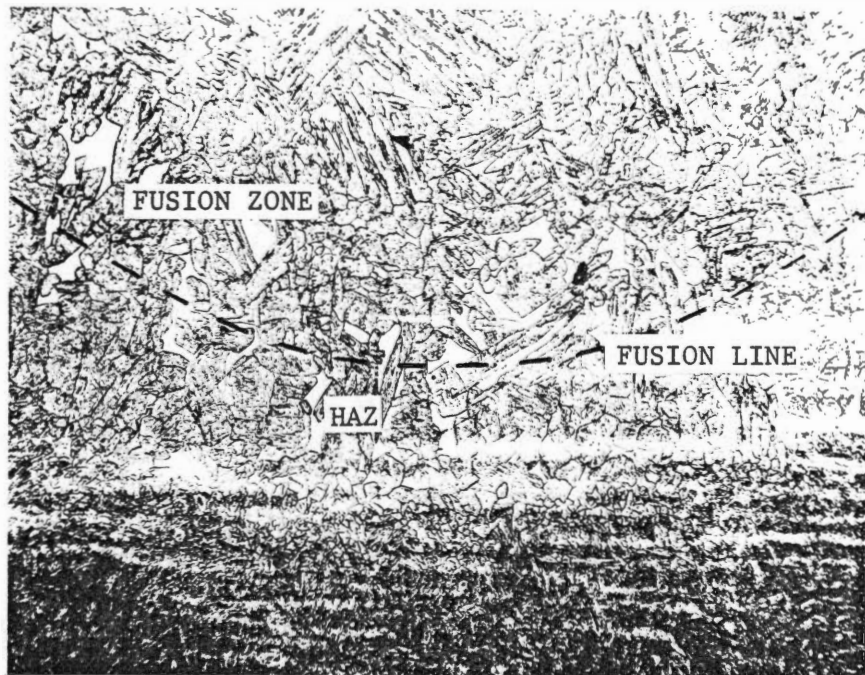


Figure 95. Transverse view of 9Cr-1Mo-0.11C UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified HCl-Picral.

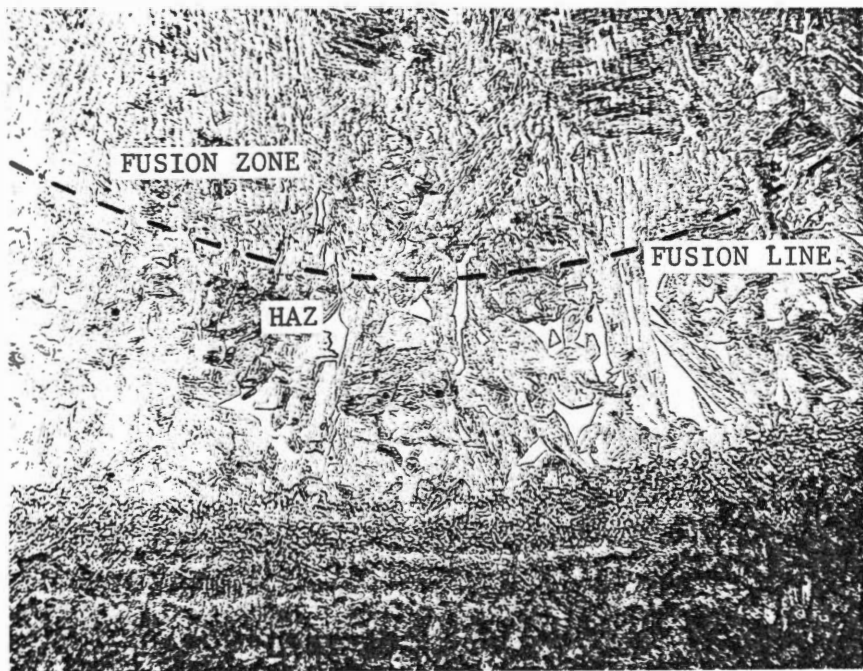


Figure 96. Transverse view of 9Cr-1Mo-0.06C UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Modified HCl-Picral.

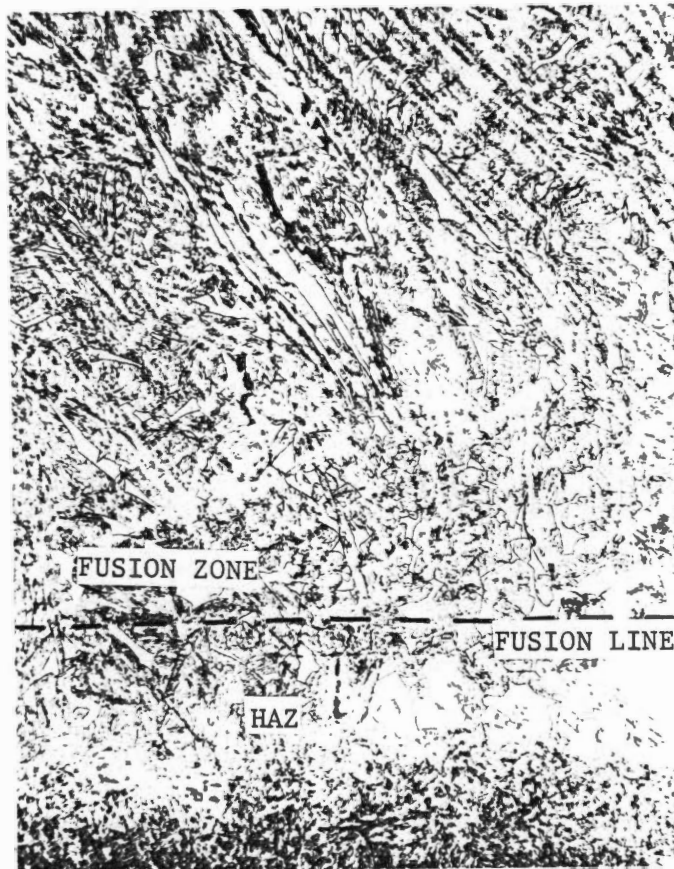


Figure 97. Longitudinal view of 9Cr-1Mo-0.11C UT-Mod HST specimen. 75°F (24°C) preheat, 4% nominal strain at 75°F (24°C). OLM. 100X. Modified HCl-Picral.

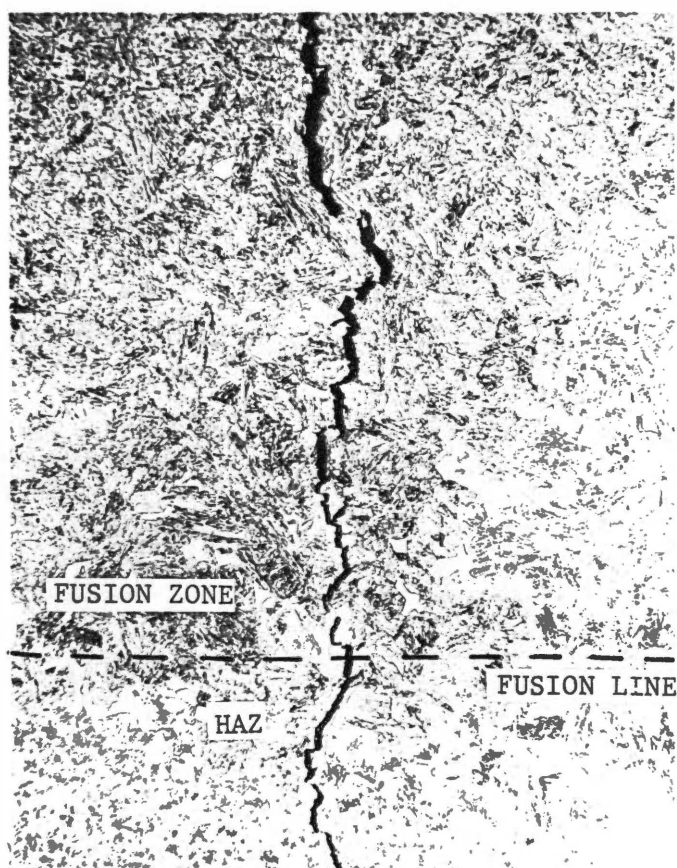


Figure 98. Longitudinal view of 9Cr-1Mo-V-Nb (394) UT-Mod HST specimen. 75°F (24°C) preheat, 4% nominal strain at 75°F (24°C). OLM. Modified HCl-Picral.

examinations revealed no indication of the presence of a liquid film at the cracking temperature (which is usually a characteristic feature of a hot crack). It is apparent that additional work is required to discern the possible elevated temperature cracking phenomenon of the 9Cr-2Mo-V-Nb material.

In the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials (Figures 91-98), the delta-ferrite in the fusion zone and HAZ immediately adjacent to the fusion zone is angular in nature. The 9Cr-2Mo material contains the greatest amount of delta-ferrite. Among the 9Cr-1Mo-V-Nb materials, the higher Si content materials (383,394) contain the most delta-ferrite. There are no obvious differences in the amount of delta-ferrite present in the standard composition 9Cr-1Mo materials.

Photomicrographs which are representative of the type of cracks observed in the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials are shown in Figures 92, 97 and 98. The cracks are generally perpendicular to the direction of the maximum tensile stress on the surface of the UT-Mod HST specimen. The cracks in the 9Cr-1Mo-V-Nb and 9Cr-1Mo materials traverse both the fusion zone and HAZ (Figures 97 and 98), while the cracks in the 9Cr-2Mo material are limited to the fusion zone (Figure 92). The HAZ and fusion zone cracks are generally confined to the martensitic regions because the angular delta-ferrite tends to blunt the crack advancement (Figure 99). The transgranular cracks in the martensite regions are typical of HAC.

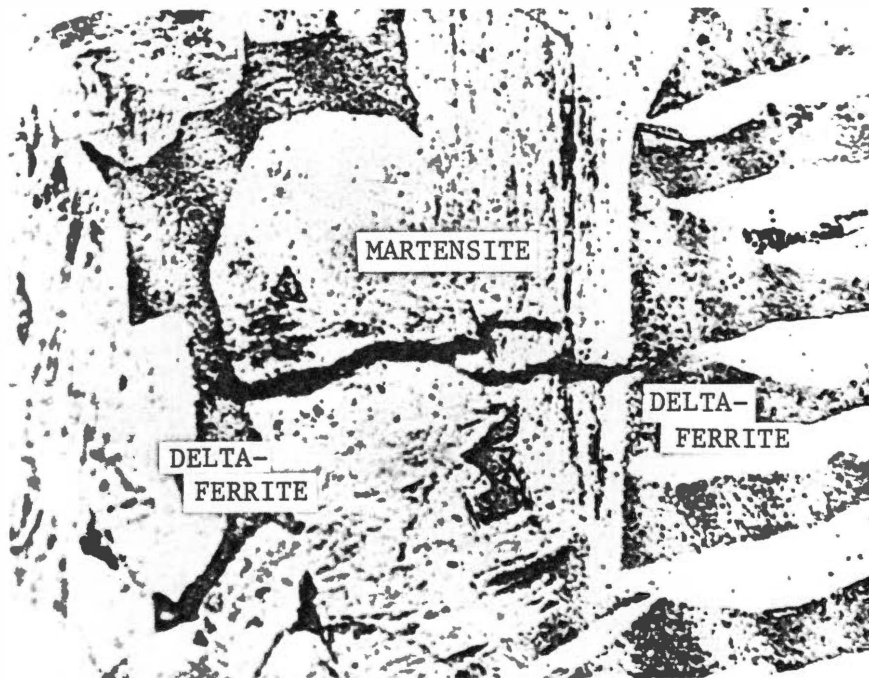


Figure 99. Photomicrograph of fusion zone crack in 9Cr-2Mo. 75°F (24°C) preheat, 4% nominal strain at 75°F (24°C). OLM. 900X. Modified Nital.

Summary of the relative HAC susceptibility of 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo. Based on the UT-Mod HST the 9Cr-2Mo-V-Nb is more susceptible to HAC than the 12Cr-2Mo-Ni, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials. The increased cracking susceptibility of the 9Cr-2Mo-V-Nb may be attributed to the massive amounts of ferrite present in the fusion zone and HAZ and/or to the almost continuous network of crack sensitive, carbon-enriched martensite along the ferrite grain boundaries.

For the welding conditions simulated by the UT-Mod HST (high hydrogen content, high restraint), the relative HAC susceptibility of the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials is dependent on carbon content and delta-ferrite content. An increase in the carbon content in the predominantly martensitic 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo results in an increase in HAC susceptibility as would be expected. For comparable carbon contents (0.06C—9Cr-2Mo, 9Cr-1Mo-0.06C and 0.08C—9Cr-1Mo-V-Nb (148, 176, 383 and 394) the materials containing the greatest amount of delta-ferrite (9Cr-2Mo and 9Cr-1Mo-V-Nb (383, 394)) are the most susceptible to HAC. In the CCT evaluation, it was shown that an increase in the delta-ferrite content increases the carbon content of the martensite. The martensite is more susceptible to HAC as a result of an increase in carbon content.

Evaluation of the HAC Susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo

In Figure 100, the UT-Mod HST results for the 3Cr-1.5Mo and 2 1/4Cr-1Mo are summarized in the same manner as those for the

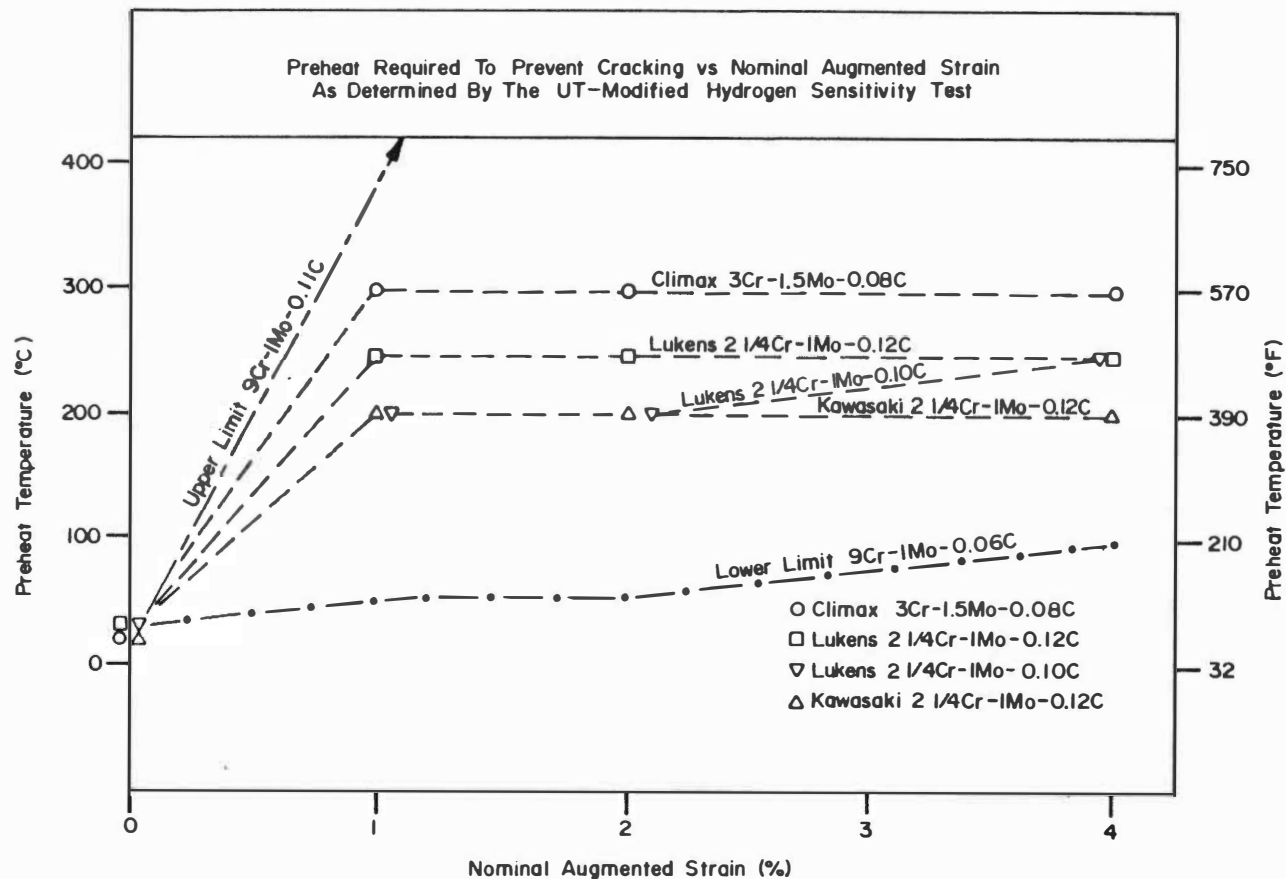


Figure 100. Summary of UT-Mod.HST for 3Cr-1.5Mo and 2 1/4Cr-1Mo showing the effect of strain on the preheat required to prevent cracking.

9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo (Figure 88, p. 171). For comparison, the upper and lower limit of preheat temperatures to prevent cracking determined for the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials are also included in this plot. Again, the 32°F (0°C) straining temperature was selected to produce the most conservative estimate of HAC susceptibility. There was little change in the preheat required to prevent cracking in the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials as the augmented strain increased from 1% to 4%. This is in sharp contrast to the increase in preheat to prevent cracking observed over the same strain range for 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo.

Comparing the relative HAC susceptibilities of the 3Cr-1.5Mo and 2 1/4Cr-1Mo at 4% nominal strain, it is apparent that the Climax 3Cr-1.5Mo material is the most susceptible to HAC, requiring a preheat of 570°F (300°C) to prevent cracking. For 4% strain, the HAC susceptibility of the Lukens 2 1/4Cr-1Mo-0.12C material is approximately the same as that of the Lukens 2 1/4Cr-1Mo-0.10C material, both requiring a preheat of 480°F (250°C) to prevent cracking. The Kawasaki 2 1/4Cr-1Mo-0.12C material is the material least susceptible to HAC, requiring a preheat of 390°F (200°C) at 4% strain to prevent cracking.

No cracking was observed when the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials were welded with 100% argon and strained to 4%. This confirms that hydrogen was responsible for the cracking observed in the UT-Mod HST evaluation of these materials.

Metallographic evaluations of 3Cr-1.5Mo and 2 1/4Cr-1Mo

UT-Mod HST specimens. Metallographic studies of the fusion zone and HAZ of the Climax 3Cr-1.5Mo, Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C and Kawasaki 2 1/4Cr-1Mo-0.12C were conducted to discern microstructural differences which were responsible for the range of HAC susceptibility observed in the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials. Photomicrographs of the UT-Mod HST specimens which are transverse and longitudinal to the welding direction are shown in Figures 101-106. The microstructures observed in the fusion zone and HAZ of the 3Cr-1.5Mo and 2 1/4Cr-1Mo UT-Mod HST specimens are similar to those observed in Figures 36, 38 and 40 (pp. 101, 102 and 103) of the CCT HAZ transformation behavior evaluation. In the CCT HAZ transformation investigation, it was shown that this type of microstructure was a mixture of martensite and bainite. It was also shown that the HAZ of the 3Cr-1.5Mo was more likely to contain martensite than the HAZ of the 2 1/4Cr-1Mo. The presence of more martensite in the HAZ of the 3Cr-1.5Mo would account for the increased relative HAC susceptibility observed in the Climax 3Cr-1.5Mo material.

Photomicrographs which are representative of the cracks observed in the bainitic materials are shown in Figures 105 and 106. In general, the cracks were perpendicular to the maximum tensile stress direction on the strained surface of the UT-Mod HST specimen. The cracks were transgranular in nature extending from the fusion zone to the HAZ. The transgranular nature of the cracks is distinctive of HAC. Sometimes transgranular cracks which were delayed in nature were

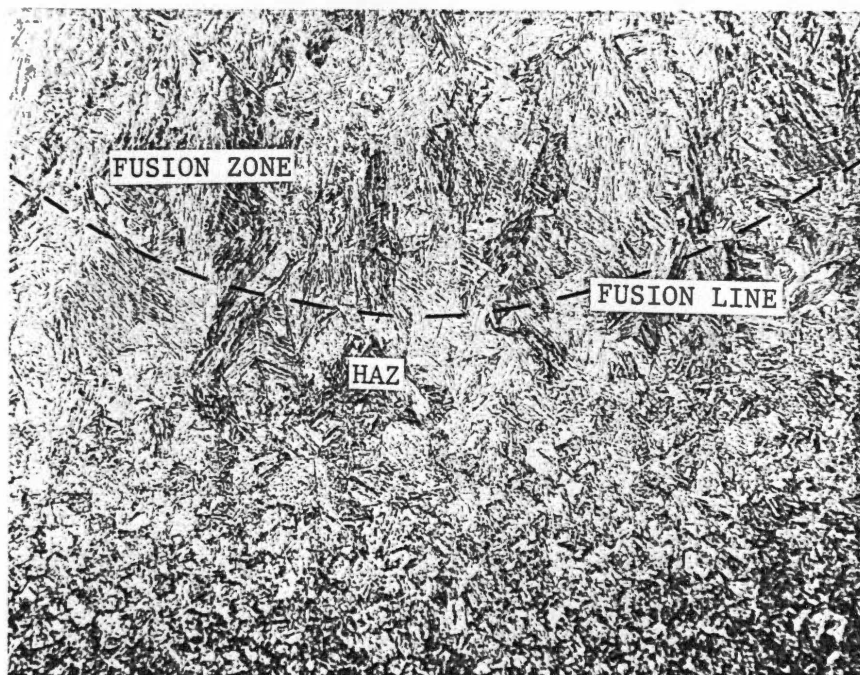


Figure 101. Transverse view of Climax 3Cr-1.5Mo UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Picral-Nital.

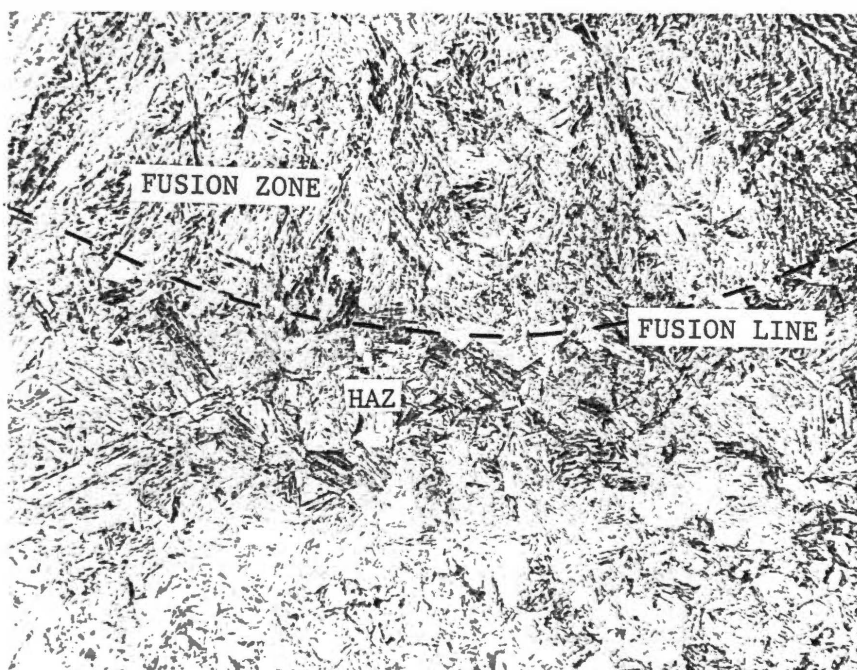


Figure 102. Transverse view of Lukens 2 1/4Cr-1Mo-0.12C UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Picral-Nital.

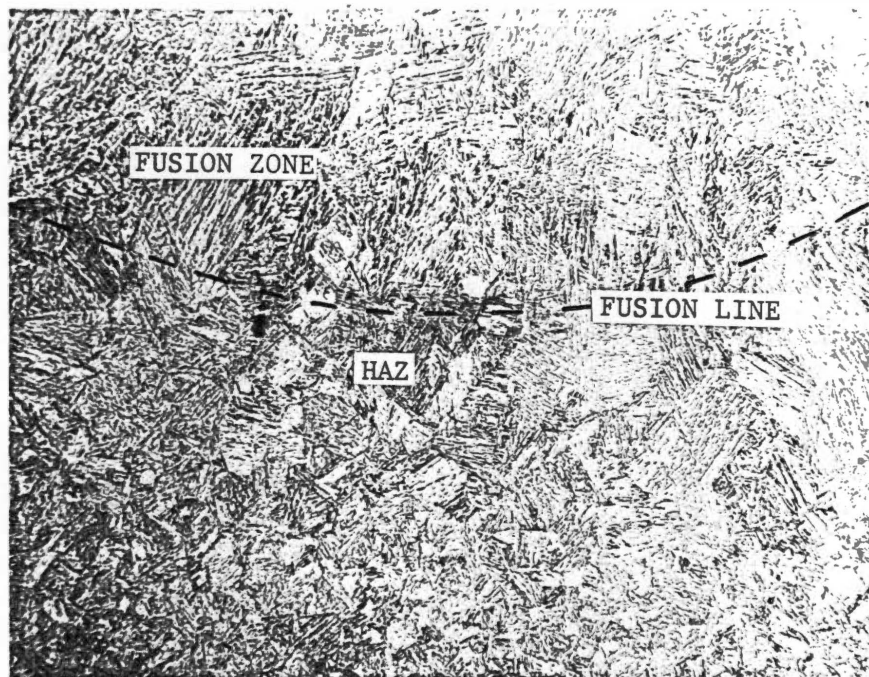


Figure 103. Transverse view of Kawasaki 2 1/4Cr-1Mo-0.12C UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Picral-Nital.

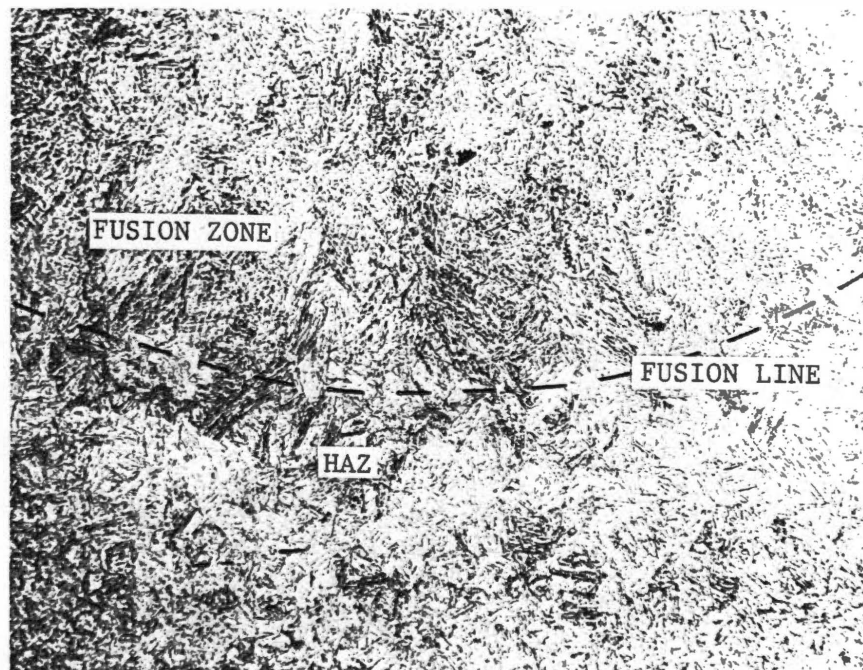


Figure 104. Transverse view of Lukens 2 1/4Cr-1Mo-0.10C UT-Mod HST specimen. 75°F (24°C) preheat. OLM. 100X. Picral-Nital.

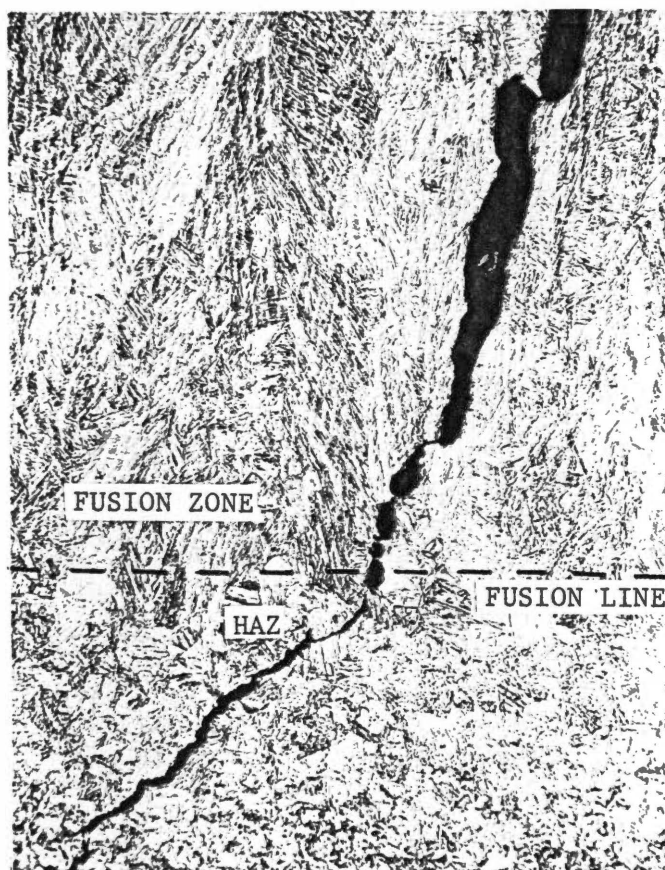


Figure 105. Longitudinal view of Climax 3Cr-1.5Mo UT-Mod HST specimen. 390°F (200°C), 4% nominal strain at 75°F (24°C). OLM. 100X. Picral-Nital.

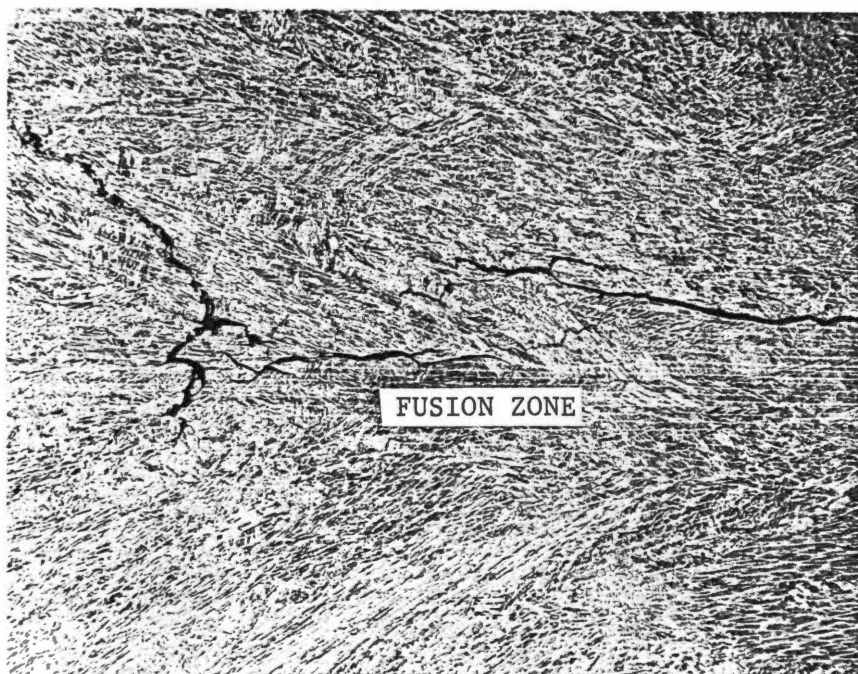


Figure 106. Longitudinal view of Kawasaki 2 1/4Cr-1Mo-0.12C UT-Mod HST specimen. 300°F (150°C), 4% nominal strain at 75°F (24°C). OLM. 100X. Picral-Nital.

observed parallel to the welding direction (centerline cracks, Figure 106). It is possible that such centerline cracks initiated as a small hotcrack, and then proceeded to propagate along the centerline in the presence of hydrogen.

Summary of the relative HAC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo. For the welding conditions simulated in the UT-Mod HST (high hydrogen content, high restraint), the Climax 3Cr-1.5Mo is more susceptible to HAC than the Lukens 2 1/4Cr-1Mo-0.12C, Lukens 2 1/4Cr-1Mo-0.10C, and Kawasaki 2 1/4Cr-1Mo-0.12C because the HAZ of the Climax 3Cr-1.5Mo material is more likely to contain martensite. Among the three 2 1/4Cr-1Mo materials, the preheat required to prevent cracking differs by 90°F (50°C). There is no obvious correspondence between the composition or the microstructure of these three materials which adequately explains the slight difference in their HAC susceptibility.

Based on the preheat required to prevent cracking at 4% nominal strain in the UT-Mod HST, the relative HAC susceptibility of the 3Cr-1.5Mo and 2 1/4Cr-1Mo is approximately the same as that determined for the 9Cr-1Mo-V-Nb and 9Cr-2Mo. For nominal strain levels less than 1%, the 9Cr-2Mo and 9Cr-1Mo-V-Nb is less susceptible to HAC than the 3Cr-1.5Mo and 2 1/4Cr-1Mo. It is apparent that at lower strain levels the martensite/ferrite HAZ microstructure of the 9Cr-2Mo and 9Cr-1Mo-V-Nb is less susceptible to HAC than the bainite/martensite microstructure in the 3Cr-1.5Mo and 2 1/4Cr-1Mo.

Summary of the HAC Susceptibility of Cr-Mo Steels

Based on the foregoing results, it is possible to directly compare the relative HAC susceptibilities of the 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-1Mo-V-Nb, 9Cr-1Mo, 3Cr-1.5Mo and 2 1/4Cr-1Mo materials in terms of preheat, level of restraint and microstructure. Overall, the 9Cr-2Mo-V-Nb material is the material most susceptible to cracking during welding. Even under conditions of very low hydrogen (100% argon shielding gas) and low restraint (1%), this material exhibited cracking on straining. The enhanced cracking susceptibility of the 9Cr-2Mo-V-Nb material may be a result of: (1) massive amounts of crack sensitive ferrite present in the fusion zone and HAZ, (2) continuous network of carbon-enriched, crack sensitive martensite and (3) an elevated temperature crack phenomenon.

While the 12Cr-2Mo-Ni steels are less susceptible to HAC than the 9Cr-2Mo-V-Nb, they are more susceptible to HAC than the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials. All three 12Cr-2Mo-Ni materials cracked in a delayed manner when welded at a preheat of 750°F (400°C) and strained 1% at 75°F (24°C). The difference in the HAC susceptibility of the 12Cr-2Mo-Ni, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials may be related to the range of their on-cooling transformation temperatures. The range of the on-cooling temperatures of the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials encourage autotempering of martensite and promote the diffusion of hydrogen out of the weld region, both of which reduce the HAC susceptibility. In contrast, the range of the on-cooling transformation temperatures of the

12Cr-2Mo-Ni materials are below those of the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials, providing little opportunity for autotempering or self-diffusion of hydrogen. In the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni, the problem of diffusing hydrogen out of the HAZ is further compounded by the presence of austenite below 300°F (150°C). If the austenite transforms below 300°F (150°C), there is little opportunity for the hydrogen to escape and it saturates the newly formed martensite which, in turn significantly increases the HAC susceptibility.

The relative HAC susceptibilities of the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials are primarily related to carbon content and delta-ferrite content. In general, as the carbon content increases, the HAC susceptibility increases as well. No beneficial effects of delta-ferrite on the HAC susceptibility of the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials were observed, in fact, at comparable carbon contents (0.08 and 0.06% C), the materials containing greater amounts of delta-ferrite (9Cr-2Mo and 9Cr-1Mo-V-Nb (383,394) exhibited an increased susceptibility to HAZ. In the CCT evaluation, it was shown that the presence of a sufficient amount of delta-ferrite can possibly increase the carbon content of the martensite which, in turn, results in an increase in HAC susceptibility.

The preheats required to prevent HAC cracking in 3Cr-1.5Mo and 2 1/4Cr-1Mo at various strain levels lie in the mid range of those determined for the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials. For a low strain level (1%), the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials show a greater HAC susceptibility than the 9Cr-2Mo, 9Cr-1Mo-V-Nb and

9Cr-1Mo-0.06C materials. When the nominal strain level increases, the HAC susceptibility of the bainitic materials shows little change. For a nominal strain level of 4%, the HAC susceptibility of the 3Cr-1.5Mo, 2 1/4Cr-1Mo, 9Cr-2Mo and 9Cr-1Mo-V-Nb materials are approximately the same.

The Climax 3Cr-1.5Mo material is more susceptible to HAC than the three 2 1/4Cr-1Mo materials because of its higher alloy content which promotes the formation of martensite, thus increasing HAC susceptibility. The preheat required to prevent cracking in the three 2 1/4Cr-1Mo materials differs by 90°F (50°C). There is no obvious differences in the composition or the microstructure of the three 2 1/4Cr-1Mo materials which adequately explain the slight difference in HAC susceptibility.

In the CCT evaluation, it was shown that the transformation behavior of Cr-Mo steels could be categorized in terms of different subclasses of Cr-Mo steels (12Cr-2Mo-Ni, 9Cr-2Mo and 9Cr-1Mo, 3Cr-1.5Mo and 2 1/4Cr-1Mo). From the previous discussion, it is evident that the relative HAC susceptibility of Cr-Mo steels can also be categorized in terms of the same subclasses of Cr-Mo steels. It is apparent that the differences in HAC among the Cr-Mo steels is directly related to differences in the on-cooling transformation behavior in the HAZ.

CHAPTER V

CONCLUSIONS

The study of the weldability of 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb, 9Cr-1Mo, 3Cr-1.5Mo and 2 1/4Cr-1Mo has produced an extensive amount of fundamental and pragmatic data on the transformation behavior, microstructure and weld cracking susceptibility (SRC and HAC) of the Cr-Mo steels. The conclusions drawn from the information obtained in this study can be summarized as follows:

I. Transformation Behavior.

A. Transformation behavior of 12Cr-2Mo-Ni.

1. The coarse grained HAZ microstructure of 12Cr-2Mo-6Ni is predominantly martensite. Under commonly employed welding conditions, the HAZ microstructure is independent of the cooling rate.
2. The coarse grained HAZ microstructure of 12Cr-2Mo-3Ni and 12Cr-2Mo-1Ni is predominantly martensite and delta-ferrite. Under commonly employed welding conditions, the HAZ microstructure is independent of the cooling rate.
3. The absence of a detectable M_f temperature for the 12Cr-2Mo-6Ni and 12Cr-2Mo-3Ni materials suggests that austenite may be retained in the HAZ at ambient temperature.

B. Transformation behavior of 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb, and 9Cr-1Mo.

1. The coarse grained HAZ of 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb, and 9Cr-1Mo contains martensite and delta-ferrite. The HAZ microstructure is independent of the cooling rate for the common welding conditions investigated.
2. The amount of delta-ferrite present in the HAZ depends on the Cr.E. An increase in the Cr.E. results in an increase in the amount of delta-ferrite retained in the HAZ microstructure.
3. In the 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials, the M_s and M_f temperatures of the coarse grained HAZ range from 760°F to 540°F (405°C to 280°C) and are higher than those observed for the 12Cr-2Mo-Ni alloys.
4. The HAZ M_s and M_f temperatures of the 9Cr-1Mo materials, which contain less than 10% delta-ferrite, are primarily dependent on the carbon content. An increase in the carbon content reduces the M_s and M_f temperature.
5. In the 9Cr-2Mo materials the transformation behavior is influenced by the presence of more than 30% delta-ferrite. The M_s and M_f temperatures of the 9Cr-2Mo materials are lower than expected for the respective

bulk carbon contents. Apparently, in the 9Cr-2Mo-V-Nb and 9Cr-2Mo materials delta-ferrite is present in sufficient quantities to partition carbon into the surrounding martensite, thus reducing the M_s and M_f temperatures.

6. The martensite hardness in the 9Cr-2Mo-V-Nb and 9Cr-2Mo is higher than the martensite hardness in the higher carbon 9Cr-1Mo material because of this aforementioned partitioning.

C. Transformation behavior of 3Cr-1.5Mo and 2 1/4Cr-1Mo.

1. The B_s and B_f temperatures in 2 1/4Cr-1Mo vary with nominal alloy content. The Kawasaki 2 1/4Cr-1Mo-0.12C material, which is slightly richer in C, Cr, Mo and Mn, exhibits the lowest B_s and B_f temperatures of the three 2 1/4Cr-1Mo materials investigated. The B_s and B_f temperatures observed in the 3Cr-1.5Mo material are similar to the B_s and B_f temperatures of the Kawasaki 2 1/4Cr-1Mo-0.12C material.
2. The B_s and B_f temperatures of the coarse grained HAZ in 3Cr-1.5Mo and 2 1/4Cr-1Mo is dependent on the cooling rate. A decrease in the cooling rate results in an increase in the B_s and B_f temperatures.
3. A decrease in the HAZ cooling rate in the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials produces a microstructure which consists of: (1) A mixture of acicular

martensite and lenticular autotempered martensite.

(2) A mixture of lenticular martensite and bainite consisting of elongated, discontinuous carbide particles. (3) A fully bainitic structure composed of irregular carbide particles. (4) A mixture of polygonal ferrite and bainite composed of blocky carbide particles.

4. The 3Cr-1.5Mo HAZ microstructure is more likely to contain martensite than the 2 1/4Cr-1Mo HAZ microstructure.

II. Susceptibility to SRC.

A. SRC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo.

1. The 3Cr-1.5Mo is less susceptible to SRC than 2 1/4Cr-1Mo.
2. Both 3Cr-1.5Mo and 2 1/4Cr-1Mo are more susceptible to SRC at the 1150°F (620°C) PWHT than at the 1250°F (680°C) PWHT.
3. An increase in the Si or C content increases the SRC susceptibility of 2 1/4Cr-1Mo.
4. Based on stress-rupture strength and stress-rupture ductility, the 3Cr-1.5Mo and 2 1/4Cr-1Mo materials are less susceptible to SRC than the martensitic high strength steels examined in a previous investigation.

5. The SRC fracture mode of 3Cr-1.5Mo and 2 1/4Cr-1Mo often exhibits micro-ductility associated with the intergranular fracture which is similar to that observed in a previous investigation.

B. SRC susceptibility of 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo.

1. Based on stress rupture ductility and fractographic examinations, the 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo materials should not be susceptible to SRC when normal PHWT temperatures are employed.

III. Susceptibility to HAC.

A. HAC susceptibility of 12Cr-2Mo-Ni, 9Cr-2Mo-V-Nb, 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo.

1. The 9Cr-2Mo-V-Nb is the material most susceptible to HAC.
2. The 12Cr-2Mo-Ni materials are more susceptible to HAC than the 9Cr-1Mo-V-Nb and 9Cr-1Mo materials because the austenite transforms to martensite at a lower temperature. Hydrogen has a reduced ability to diffuse away from the HAZ at the lower transformation temperature, increasing the HAC susceptibility.
3. The HAC susceptibility of the 9Cr-2Mo, 9Cr-1Mo-V-Nb and 9Cr-1Mo is dependent on carbon content and

delta-ferrite content. An increase in the carbon content results in an increase in HAC susceptibility. The materials containing more delta-ferrite are more susceptible to HAC because the delta-ferrite partitions carbon to the martensite.

B. HAC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo.

1. The 3Cr-1.5Mo material is more susceptible to HAC than the 2 1/4Cr-1Mo because of the increased tendency to form martensite in the HAZ.
2. An increase in the level of simulated restraint results in little change in the HAC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo.
3. For high levels of simulated restraint (2% and 4% augmented strain, the HAC susceptibility of 3Cr-1.5Mo and 2 1/4Cr-1Mo is approximately the same as the 9Cr-2Mo and 9Cr-1Mo-V-Nb.
4. For lower levels of simulated restraint (1% augmented strain), the HAC susceptibility of the 3Cr-1.5Mo and 2 1/4Cr-1Mo is greater than that of 9Cr-2Mo and 9Cr-1Mo-V-Nb.
5. The 2 1/4Cr-1Mo material is less susceptible to HAC than the 9Cr-1Mo material at an equivalent carbon content.
6. There is currently no adequate explanation for the slight differences observed in the HAC susceptibility of the 2 1/4Cr-1Mo materials evaluated.

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VITA

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