

An Experimental Study of Causes and Repair of Cracking of $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ Steel Equipment

API PUBLICATION 938

MAY 1996

American Petroleum Institute
1220 L Street, Northwest
Washington, D.C. 20005



Documents

Document Sharing Hub

SPECIAL NOTES

(1) API publications necessarily address problems of a general nature. With respect to particular circumstances, local, state, and federal laws and regulations should be reviewed.

(2) API is not undertaking to meet the duties of employers, manufacturers, or suppliers to warn and properly train and equip their employees, and others exposed, concerning health and safety risks and precautions, nor undertaking their obligations under local, state, or federal laws.

(3) Information concerning safety and health risks and proper precautions with respect to particular materials and conditions should be obtained from the employer, the manufacturer or supplier of that material, or the material safety data sheet.

(4) Nothing contained in any API publication is to be construed as granting any right, by implication or otherwise, for the manufacture, sale, or use of any method, apparatus, or product covered by letters patent. Neither should anything contained in the publication be construed as insuring anyone against liability for infringement of letters patent.

(5) Generally, API standards are reviewed and revised, reaffirmed, or withdrawn at least every five years. Sometimes a one-time extension of up to two years will be added to this review cycle. This publication will no longer be in effect five years after its publication date as an operative API standard or, where an extension has been granted, upon republication. Status of the publication can be ascertained from the API Authoring Department [telephone (202) 682-8000]. A catalog of API publications and materials is published annually and updated quarterly by API, 1220 L Street, N.W., Washington, D.C. 20005.



**FINAL REPORT TO
THE AMERICAN PETROLEUM INSTITUTE
OF RESEARCH UNDER THE GUIDANCE OF
THE TASK GROUP ON MATERIALS AND CORROSION RESEARCH
OF THE
COMMITTEE ON CORROSION AND MATERIALS**

**AN EXPERIMENTAL STUDY OF CAUSES
AND REPAIR OF CRACKING OF
 $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ STEEL EQUIPMENT
MAY 1996**

**Contractor:
The Materials Properties Council, Inc.
Investigator: M. Prager**

**Subcontractor:
Department of Materials Science
University of Tennessee
Investigators: C. D. Lundin, P. Liu,
C. Y. P. Qiao, G. Zhou and K. K. Khan**

FOREWORD

The origin of this project was an in-depth study for API of numerous reported incidents of cracking of equipment of 1Cr-½Mo and 1¼Cr-½Mo-Si steels. The report to API is an explanation of the problem and the basis for further work to help prevent and repair such cracking. Metallurgical reports, fabrication records and service histories were reviewed. Worldwide research on the subject by steelmakers and studies of these alloys and similar materials in related applications were considered. In many cases, the cracking was major and cracks propagated in service. Emphasis in the report was placed on the causes of crack initiation during fabrication or of their appearance after only a short time in service.

It was concluded that major contributions to the cracking were from poor design, fabrication and operating practices which should be corrected using reasonable precautions and well known technology. Such action would prevent future vessels from entering service with preexisting cracks or initiating cracks in service. However, there was strong evidence that some of the plates and forgings used for vessel construction were more prone to cracking than others or have disturbingly low toughness. This study was intended to recommend ways to eliminate detrimental fabrication practices and materials.

Fabrication and repair operations must be upgraded because subsurface cracks which cannot be readily detected may occur and then emerge in service. Repairs have been troublesome.

Specifically, the study was developed to address the materials, fabrication and repair issues of greatest concern.

Under API and MPC Task Groups, Chaired by J. McLaughlin. The objectives of the Phase II Study were established as follows:

- 1) Develop an understanding of the fabrication/welding factors that affect cracking of Cr-Mo equipment, including the effects of PWHT and pre-heat temperature.
- 2) Develop an understanding of the inherent material properties that affect cracking of Cr-Mo equipment. This was to include the effects of impurities in the steel and initial condition of the steel (i.e., annealed vs. normalized and tempered).
- 3) Define a controlled deposition (temper bead) procedure for repair and initial fabrication that will produce a fine grain, more damage tolerant, microstructure in the weld heat affected zone (HAZ).
- 4) Determine the effect of using lower carbon, lower strength fillers for repair welds. Experience suggests that depending on conditions, the use of a low carbon filler can either improve or impair the performance of a repaired weld.

Appreciation is expressed to API for support. Portions of the work were cost shared by MPC, PVRC and WRC.

This work resulted in important new physical simulation, weldability and notched bar test methods. Fresh insight was gained into the heat affected zone metallurgy of this important class of materials.

Dr. M. Prager
Executive Director, Materials Properties Council, Inc.

CONTENTS

Executive Summary	1
For Applications of 1¼Cr–½Mo Steel at 825 F and Higher	1
Report Overview and Conclusions	2
Recommendations for Vessel Fabrication and Repair	3
Introduction.....	3
Fabrication Guidelines	3
Guideline Details Class 2 Properties.....	5
Class 1 Properties	6
Repair Guidelines	7
Summary of Program and Results of Testing and Evaluation	8
Introduction.....	8
Section A. Literature Review	8
Section B. Materials	9
Section C. Weld HAZ Transformation Behavior.....	9
Section D. Reheat/PWHT Cracking Assessment	10
Section E. Development of Factors to Predict Reheat/PWHT	
Cracking	11
Section F. Toughness Evaluations as a Function of PWHT.....	14
Section G. Microstructural and Fractographic Evaluations	16
Section H. Creep Rupture Behavior of the Coarse Grained HAZ-	
Notch Bar and Smooth Bar Creep/Stress Rupture Testing.....	17
Section I. Repair Welding Procedures, Behavior of Low Carbon	
Metal and Repaired Weldments.....	17
References	19
Appendix A—Literature Survey—Cr—Mo Steels—Reheat and In-	
Service Cracking.....	20
Appendix B—Chemical Composition of 1¼Cr–½Mo API Materials.....	46
Appendix C—Coarse Grained Transformation Behavior and Associated	
Microstructures	48
Appendix D—Assessment of Reheat Cracking Susceptibility	55
Appendix D1: Gleeble Simulation Smooth Bar Reheat Crack Testing..	55
Appendix D2: Spiral Notch Testing	78
Appendix D3: Development of a New Reheat Cracking Test—Preview	
Test and Evaluation of Reheat Cracking in API Materials	86
Appendix E—Determination of Factors to Quantify Reheat Cracking	
Susceptibility Based on Chemical Composition	101
Appendix F—Toughness Study.....	117
Appendix G—Microstructural and Fractographic Evaluations.....	130
Appendix G1: Fractographic Examination of Notched Creep and	
Gleeble Stress Rupture Samples of API Materials.....	130
Appendix G2: SEM Metallographic Investigation and EDS Analyses	
of UT2 and UT3 Materials.....	136
Appendix G3: Phase I—High Resolution Electron Microscope	
Evaluation on API Materials.....	142
Appendix G3: Phase II—High Resolution Electron Microscopic	
Evaluation on API Materials.....	158

Appendix G4: Transmission Electron Microscopy Evaluation on API	
Materials.....	176
Appendix H—Creep Rupture Behavior of the Coarse Grained HAZ	187
Appendix H1: Notch Bar and Smooth Bar Creep/Stress Rupture	
Testing	187
Appendix H2: Preparation of Extended Length HAZ Simulation	
Specimens by the Gleeble Technique	204
Appendix I—Repair Welding Procedure	210
Behavior of Low Carbon Weld Metal and Repaired Weldments.....	210
Temper-Bead Welding Procedure for Repair Welding.....	211
Appendix J—Program Tasks	214

An Experimental Study of Causes and Repair of Cracking of 1¼Cr-½Mo Steel Equipment

C. D. Lundin, P. Liu, C. Y. P. Qiao, G. Zhou, K. K. Khan and M. Prager

Executive Summary

A multitask experimental study was conducted to provide the petroleum industry with solutions to recurring incidents of cracking in the application of welded 1¼Cr-½Mo steel for hydrogen processing equipment. The principal objective was to develop recommendations for the elimination of cracking that occurred during fabrication or early in operating life, was associated with repairs or was found after extended service exposure at elevated temperature.

Vessel and equipment experience has shown that the majority of weld cracking problems have occurred at temperatures in excess of 850°F. Further, little or no problems have been found for operation at temperatures below 800°F. Thus, a cutoff temperature of 825°F has been suggested for invoking the precautions, considerations and recommendations regarding the potential for coarse grained weld HAZ (CGHAZ) cracking in 1¼Cr-½Mo steels.

The research plan followed was proposed as a Phase II study at the conclusion of a survey and investigation (Phase I) conducted for API by MPC and reported in September, 1990.

The objectives of Phase II were to: determine what compositional and other material issues influence cracking; evaluate controlled deposition repair techniques; determine the suitability of low carbon filler materials; and understand the role of fabrication and welding practices on susceptibility to cracking.

The program succeeded in all objectives. It was found that fabrication, repair- and service-related cracking often have the same roots and are responsive to the same remedial action. Problems arise where there is low heat affected zone ductility. The following conclusions and recommendations are therefore provided.

For Applications of 1¼Cr-½Mo Steel at 825°F and Higher

(a) It is strongly recommended that Class 1 (60/35 ksi tensile, yield strength) materials be specified in

preference to Class 2 (75/45 ksi tensile, yield strength), accelerated cooled materials.

(b) High PWHT temperatures were found to be necessary to improve heat affected zone ductility. PWHT requirements are related to welding variables and material composition. Fabrication guidelines are provided herein with specific PWHT recommendations depending on composition, desired strength and welding variables.

(c) High PWHT temperatures may be used without undesirably impairing creep strength and charpy impact values provided carbon content is not too high.

(d) Certain materials display a high sensitivity to cracking. Materials Composition Factors (MPC-5, MPC-7) have been identified and can be used to screen the materials. All of the elements included in the factors may not normally be included in the specification requirements and thus the range of elements controlled must be especially requested with the accuracy defined in Appendix B.

(e) Design, fabrication and materials specifications may now be prepared to assure freedom from cracking.

(f) Controlled deposition welding techniques and low carbon filler metals may be implemented in repair strategies when performance objectives and materials are identified.

(g) A number of screening tests have been demonstrated as suitable for determining material sensitivity to fabrication-related cracking. These include (Gleeble) simulated heat affected zone cracking, spiral notch rupture and large scale (PREVEW) weldability tests. These tests are not intended as requirements for material purchase. However, if the composition suggests that the material may be sensitive to reheat and in-service cracking, it may be wise to consider these tests to define the extent of anticipated problems.

(h) Studies of smooth and notch bar stress rupture behavior of simulated CGHAZ specimens provided insight into the effect of PWHT, heat input and microstructure on creep rate, ductility and cracking tendency.

(i) The results of this work have shown that the term "creep embrittlement" when applied to the low

Final report to API on *Prevention and Repair of Cracking in Chrome Moly Equipment*, MPC, September, 1990.

ductility in-service cracking in the low Cr-Mo materials is inappropriate. The low ductility behavior is essentially inherent from the initiation of service and is a combined result of the material factors in terms of chemistry, resistance to tempering and the degree of thermal treatment provided prior to service. No embrittlement was found to be caused by service in the creep range and, therefore, the use of the term creep embrittlement to describe service behavior of $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ HAZ is not appropriate.

Report Overview and Conclusions

The attached report documents a comprehensive and complex study of cracking associated with an alloy for use at elevated temperatures. The Research Plan was developed and reported in an MPC document *Final Report to API—Prevention and Repair of Cracking in Chrome-Moly Equipment*. It was presented originally as a two year plan for Phase II to be conducted under the guidance of the API Task Group on Corrosion and Materials Research which prioritized the program tasks as follows:

1. *Effects of Fabrication and Welding*
Develop an understanding of the fabrication/welding factors that affect cracking of Cr-Mo equipment. This will include the effects of PWHT and preheat temperature
2. *Materials Variables*
Develop an understanding of the inherent material properties that affect cracking of Cr-Mo equipment. This will include the effects of impurities in the steel and initial condition of the steel (i.e., annealed vs. normalized and tempered).
3. *Controlled Deposition Repair Procedures*
Define a Controlled deposition (temper bead) procedure for repair and initial fabrication that will produce a fine grain microstructure in the heat affected zone (HAZ).
4. *Filler Metals*
Determine the effect of using lower carbon, lower strength fillers for repair welds. Experience suggests that depending on conditions, the use of a lower carbon filler can either improve or impair the performance of a repaired weld.

Additional work on hydrogen effects originally suggested by MPC received a low priority and was not pursued.

While the objectives of the tasks are defined separately, the work was performed in a testing plan that most efficiently explored the various interrelated issues. Appendix J indicates the relationships of the various tasks as originally described.

The various studies in Phase II that are documented and attached here are:

- (a) Update of the Literature Survey (Appendix A)
- (b) Compositional and Microstructural Studies, Heat Affected Zone Transformation and Metallurgical Characteristics (Appendixes B and C)

(c) Assessment of Reheat Cracking Susceptibility (Appendix D)

(d) Predicting Reheat Cracking Susceptibility Based on Chemical Composition (Appendix E)

(e) Toughness Study (Appendix F)

(f) Microstructural and Fractographic Evaluations (Appendix G)

(g) Notch Bar and Smooth Bar Stress-Rupture Studies (Appendix H)

(h) Repair Welding (Appendix I)

(i) Original Phase II Plan (Appendix J)

The overall logic of the program was as follows:

1. obtain a broad range of materials;
2. select small scale notch tests to screen material variables for susceptibility to elevated temperature cracking;
3. screen and rank materials on the basis of HAZ behavior;
4. validate ranking and test predictions by large scale tests;
5. systematically evaluate material variables using small scale tests;
6. examine repair procedures on sensitive heats with large scale tests;
7. use a notched bar rupture test for simulation of cracking in-service;
8. examine the effects of materials and fabrication variables on in-service cracking probability; and
9. rank materials and heat treatments for in-service cracking tendency.

A total of seventeen commercial heats were obtained and information on others was utilized. Based on analyses of the behavior of the more than twenty heats it has been concluded that the hardest areas in the weld heat affected zones of $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ steel respond relatively slowly to PWHT and may display low ductility at elevated temperatures. Ductility depends on material composition, weld heat input and PWHT conditions. While these qualitative characteristics were not surprising, the quantitative details which emerged from the study were. For example:

(a) heat affected zone ductilities among the materials varied by a factor of ten;

(b) coarse *grained* heat affected zones of high carbon materials tended to display low ductility, perhaps only a fraction of 1% to failure, even after PWHT;

(c) for a given heat input and hardness, creep rates of coarse *grained* heat affected zones varied by as much as a factor of 10 depending on composition (transformation microstructure);

(d) the ductilities of some heats were improved significantly by heat treatment while others reached a plateau and remained relatively notch sensitive;

(e) smooth bar and notched bar stress-rupture lives of the materials were found to vary by as much as a factor of ten;

(f) there is no evidence that the materials become brittle in time (creep embrittlement). Instead, it is concluded that brittleness is a consequence of the

as-tempered microstructure which must be softened significantly before ductility can be observed;

(g) creep rates and smooth bar rupture lives of simulated HAZs were only slightly affected by tempering temperatures from 1250°F–1350°F. However, the ratios of notch to smooth bar lives and ductilities tended to improve;

(h) the PWHT temperature necessary to reduce significantly notch sensitivity in the heat affected zones varied among the heats by as much as 100°F. Fabrication and repair procedures should take this into account;

(i) controlled deposition techniques and low carbon filler metals may be used to reduce the tendency for cracking during heat treatment and service;

(j) materials that were found to display PWHT cracking susceptibility tended to rate poorer expectations for service;

(k) as a result of this work the compositional factors identified as useful are shown below. All quantities are expressed in wt %. (See Appendix E for a more detailed description of these factors);

$$\text{MPC Factor-5} = [\text{Cfn}(\text{Tramp} + \text{Sin})\text{Alfn}] - 1$$

$$\text{Cfn} = (5\text{C} + 1000\text{Nb} + 100\text{V} + 50\text{Ti} - 0.5) + 1$$

$$\text{Tramp}$$

$$= 2[4.3(\text{Sn} + \text{As}) + 150\text{Sb} + \text{Cu} + 50(\text{P} - 0.01)]$$

$$\text{Sfn} = 1 + 30(\text{S} - 0.02 \text{ Tramp});$$

$$\text{For Sfn} < 1, \text{ Sfn} = 1$$

$$\text{Alfn} = 1 + 15(\text{Al} - 0.015);$$

$$\text{For Alfn} < 1, \text{ Alfn} = 1$$

$$\text{MPC Factor-7} = 2(\text{C} - 0.12) - 0.25(\text{Mn} - 0.6)$$

$$+ 150\text{Nb} + 15\text{V} + 15\text{Ti} - 100\text{B}$$

$$+ 40(\text{P} - 0.010) + 5(\text{S} - 0.015)$$

$$+ 10(\text{Al} - 0.010)$$

$$+ 20(\text{Cu}/100 + \text{Sn}/3 + \text{As}/3 + 3\text{Sb})$$

MPC Factor-5 combines a carbon function (strength), a tramp element function (embrittlement), a sulfur function (embrittlement) and an aluminum function. The combination of these functions is shown in the factor as presented in Fig. 3. MPC Factor-7 (Fig. 4) utilizes the concept of a lower limit threshold as a basis for the effect of the elements in an additive fashion. It is believed that the sensitivity to reheat/PWHT cracking can be assessed by either of these factors and a reasonable correlation exists with fabrication behavior. Using MPC Factor-5 the limiting value for the onset of a potential for reheat/PWHT cracking is 2.0 and for MPC Factor-7 the limiting value is 0.5.

(l) it is recommended that carbon content should be in the range of 0.10–0.13% to achieve satisfactory material properties with minimum fabrication problems;

(m) it is also recommended that users specify materials and processes to obtain Class 1 (60–85 UTS) to reduce problems during fabrication, repair and service. There is no difference between allowable stresses for Class 1 and Class 2 in the creep range; and

(n) it is considered that a similar factor concept be applied to 1Cr-½Mo materials. The Cr ranges overlap. The Si content is the basic differential. Excluding this difference in Si, the basic considerations are applicable. However, a slightly different factor may have been derived if a number of 1Cr-½Mo heats had been included.

Recommendations for Vessel Fabrication and Repair

Introduction

The key results of this program on means of mitigating cracking either during PWHT or in-service are presented in the form of guidelines for fabrication and repair. The discussion of the research results that support these recommendations is presented in the next section and the experimental results are contained in the various Appendixes. The guideline flow charts were derived in consideration of the data and the experience of the investigators and those in the petroleum industry. The repair recommendations offered are firm at this time, but tests of the efficacy of the low carbon weld metal continue.

Fabrication Guidelines

The fabrication guidelines are presented in the form of a flow chart (Fig. 1) that will direct the user to the considerations necessary for successful fabrication of vessels and components of 1¼Cr-½Mo steel for use at elevated temperatures into the creep temperature regime (> 825°F). The fabrication guidelines recommend that the users first establish the composition of the material of construction and consider the strength level (Class) to be employed.

Guidelines are offered for both the Class 1 (60–85 ksi) and Class 2 (75–100 ksi) strength levels. The initial material strength (hardness) during fabrication will be dictated by the final strength or Class desired in the vessel or component. The material must be purchased at a specific strength/hardness level so that the application of the required PWHT schedules does not reduce the base metal strength below that for the design Class desired. For example, a quenching and tempering (Q & T) operation may be required to maintain Class 2 strength after the desired PWHT exposure. It is evident from the testing accomplished here that the higher the initial strength of the material the more likely the occurrence of reheat/PWHT cracking in a sensitive material and if the material enters service with the HAZ only moderately tempered to preserve the high end of the strength level, in-service HAZ cracking is also more likely. The research also strongly suggests that the vessel or component should be PWHT high in the

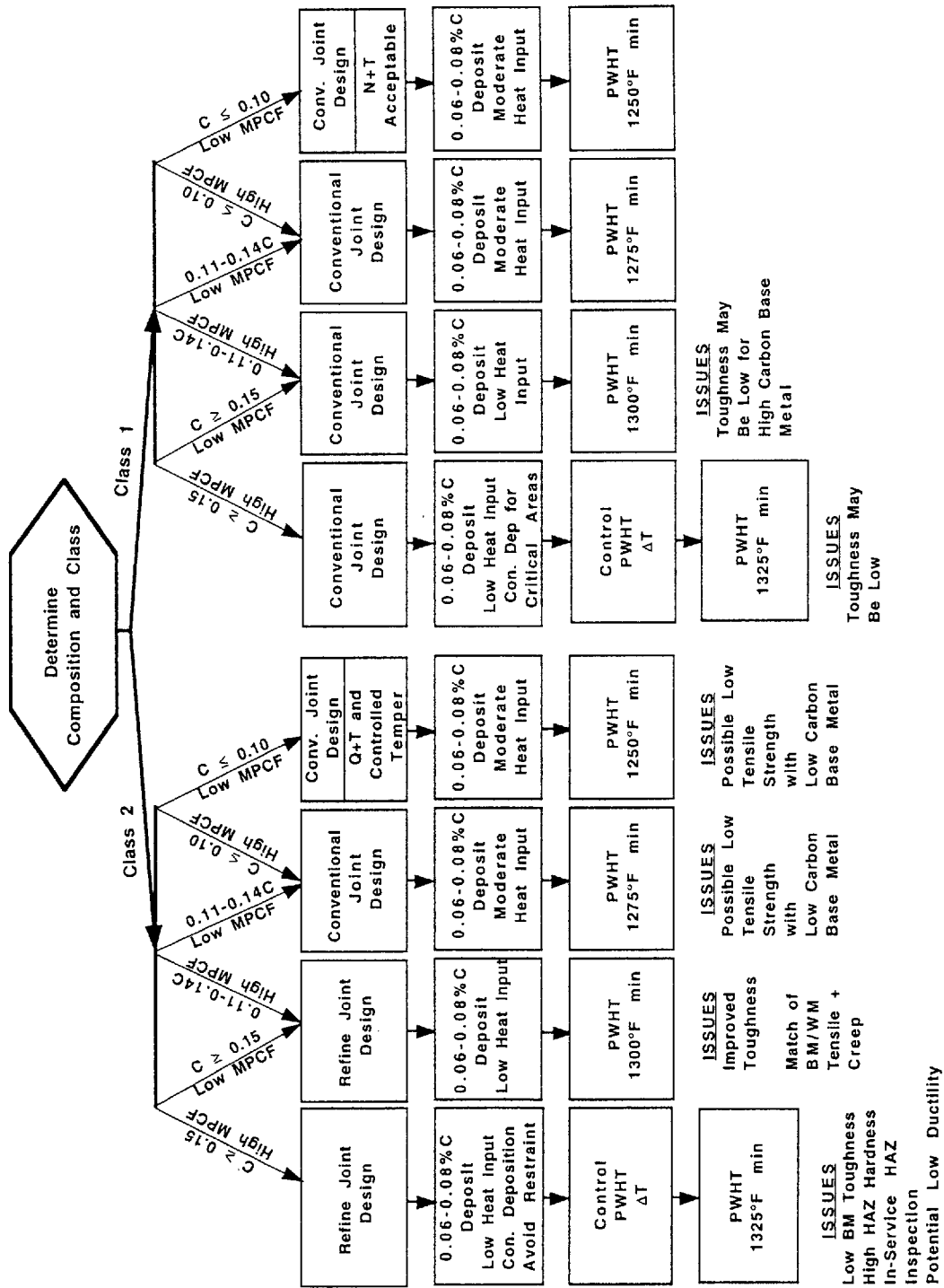


Fig. 1—Fabrication guidelines

PWHT range and certainly above 1250°F in any event in order to impart good HAZ ductility.

PWHT at high temperatures will more rapidly reduce strength levels and may cause loss in toughness, bringing further considerations to bear in terms of properties. This study of the coarse grained HAZ creep properties also showed that the strength of the HAZ for a given material was not significantly affected by the degree of PWHT, but notch sensitivity was greatly increased by relatively low PWHT. Further, in the considerations for design it must be recognized that when time-dependent properties govern, the ASME allowable stresses are substantially the same for both Classes of material (above 900°F). Thus, it would be prudent to select a Class 1 strength level for these applications.

The composition of the material should be considered from several standpoints. Today, most 1¼Cr-½Mo-Si steel is furnished to relatively low J and X-bar factors (J factors of <200 and X-bar levels less than 15) which are readily attainable with current steel making practice. However, to avoid crack susceptible heats, we propose a newly defined MPC Factor. Several were evaluated. Of these MPC Factors, 5 or 7 should be employed, and the satisfactory performance cutoff maximum values for the factors are: 2.0 for MPC Factor-5 and 0.5 for MPC Factor-7. These factors assess the sensitivity toward reheat/PWHT cracking and in-service sensitivity to low ductility HAZ cracking. It is also important to consider the carbon content of the material because of its major effect on hardenability (ability to attain a specific strength [hardness] level during the initial heat treatment; e.g., Q & T or N & T). The carbon content also plays a major role in the level of toughness that can be maintained after PWHT. The higher carbon content materials tend to experience a greater loss in toughness with a given PWHT.

When the strength class, carbon content and MPC Factor have been determined, the flow charts may be entered in the appropriate column to determine the basic considerations in fabrication that will aid in avoiding PWHT as well as in-service cracking and property degradation.

Guideline Details Class 2 Properties

Case 1. The Class 2 properties column, which is characterized by a carbon content $\geq 0.15\%$ and a high MPC Factor, represents the most critical material for fabrication. The initial direction calls for a "refined joint design" that should include all means for reducing the fabrication and operational stresses and should invoke the requirement that all joints be full penetration welded and utilize the least amount of filler metal. The weld crowns should be removed and ground flush with the plate surface for butt welds and the contour of the fillet welds should be ground such as to provide a smooth transition to the base material (blend ground). The grinding scratches should be

transverse to the weld axis and the final grinding should be done with fine wheels. The sequencing of welds should be considered carefully so as to minimize long range residual stresses.

"Refined joint design" is a generic term that reflects the optimum placement and configuration of weld joints in order to avoid excessive long range residual stresses and stress raisers that can trigger the initiation of reheat cracking. The application of butt joints, which have the smallest amount of filler metal added (narrow groove technology), limits the generation of residual stresses that span significant distances and cause distortions that contribute to the stress redistribution which triggers reheat cracking. A secondary consideration is the employment of "refined joint details" such as the surface dressing of the overfill in butt joints, which reduces stress raisers and mitigates the occurrence of reheat cracks in the butt weld HAZ. It is known that properly made butt welds with no surface or internal discontinuities are relatively immune to any exacerbating factors that contribute to reheat cracking. Sit-on nozzles that employ fillet welds are to be avoided as well as fillet welds that are not properly contoured and fared into the base metal at the toes. Lack of fusion at the root of fillet welds is also to be avoided. In general, welds should not be placed in regions of natural stress elevation or where the stress state from both the fabrication and operation standpoints is high or unknown. ASME B & PV Code Section VIII provides some guidance in Part UG, UHA and UHT, in addition to the appendixes. The 1988 Hague Conference paper by Cane also describes "refined joint designs." Naturally the sensitivity of the material to reheat cracking plays a part in the potential for cracking and to this end the MPC factors that are given cutoff limits for sensitivity should be considered when the types of joints to be employed are selected.

The weld deposit carbon content should be aimed at the 0.06–0.08% range as this carbon content range provides sufficient elevated temperature strength without the potential for excessive hardness in the weld deposit. Extensive Pressure Vessel Research Council (PVRC) research has showed that little benefit is gained in terms of elevated temperature strength as the weld metal carbon content is raised much above the 0.06–0.08% level. Preheat should be 300°F minimum, in accordance with ASME recommendations. A postweld hold at the preheat temperature is advisable when the section size exceeds one-half inch. Low weld heat input should be utilized, employing small passes that induce low temperature transformation products to yield a better ductility response during PWHT and higher creep rates to redistribute the strains attendant with reduction in welding residual stresses. Small passes also provide for maximum overlap in the base metal HAZ and thus will limit the extent of the coarse grained region remaining after welding is complete (it is the coarse grained

region that has the maximum potential for cracking). As an example, the number of passes to insure low heat input should exceed approximately 16 for a 1 in. thick plate. Controlled deposition procedures should be considered strongly dependent upon the MPC factor and carbon content. The employment of controlled deposition procedures similar to those defined in Appendix I can be used with success in refining the base metal HAZ microstructure thus minimizing the potential occurrence of reheat/PWHT cracking or cracking in the HAZ in-service.

PWHT should be performed at a minimum of 1325°F with controls placed on the heat-up rate and the ΔT within the vessel during the PWHT treatment to minimize stresses. The high PWHT will greatly desensitize the HAZ tendency toward low ductility cracking in-service.

Several issues must be addressed for the fabrication of the high carbon and high MPC Factor materials. Loss in base metal toughness, upon the high temperature PWHT, is a concern and efforts to combat this loss may involve limited heat treatment and tempering before fabrication in the attempt to maintain toughness after completion of fabrication. However, the higher the strength of the initial base material the higher the potential for cracking problems. High carbon content materials with high levels of carbide forming elements (which contribute to the MPC Factor) may cause the retention of high base metal HAZ hardness after fabrication and influence certain applications for which the hardness level is a consideration. The high hardness HAZs are also more sensitive to HAZ cracking in-service and will cause consideration for a significant level of in-service inspection of the welds, especially early on in operation.

The high carbon and high MPC Factor conditions are the most critical in the fabrication of a vessel and component and will involve the most carefully considered fabrication steps to be invoked.

Case 2. If the MPC Factor is low with a high carbon content or a high MPC Factor with the carbon content in the recommended range of 0.11–0.14%, certain relaxation in fabrication procedures can be invoked. These are basically that controlled deposition welding procedures are probably not necessary and the PWHT temperature can be reduced to 1300°F minimum. The reduction in PWHT temperature will no doubt improve toughness response upon PWHT and the base metal and weld metal tensile and creep strength will most likely have a better match. The degree of in-service inspection can probably be relaxed after the initial start-up, although the attention to joint design details and the contouring of the welds should be followed rigorously.

Case 3. With the optimum range of carbon content and a low MPC Factor or a low carbon ($\leq 0.10\%$) content and a high MPC Factor relaxation to conventional joint designs and normal treatment of weld contours and grinding can be accomplished. Moderate

heat input welding can be employed and the PWHT temperature reduced to 1275°F minimum. Better retention of toughness will be affected. However, at the low carbon level the tensile strength may be reduced to values approaching the lower limit of the Class 2 properties if extended PWHT times are employed.

Case 4. For a carbon content ($\leq 0.10\%$) and a low MPC Factor the chance of reheat/PWHT or in-service HAZ cracking is minimal and the PWHT temperature can be reduced to 1250°F minimum. However, this condition of low carbon and low MPC Factor may require a Q & T heat treatment for the base metal prior to fabrication with a controlled tempering temperature to maintain Class 2 properties.

It should be noted that Class 2 offers no advantage as far as allowable stresses for service in the creep range. Thus specification of Class 1 facilitates fabrication with no lower allowable stress and improved ductility.

Class 1 Properties

The columns representing Class 1 properties in the Fabrication Guidelines are less restrictive in fabrication procedures than those for Class 2 properties because the lower strength (hardness) of the base material provides for a greater degree of stress relaxation during PWHT and for better redistribution of stress during transient conditions in-service. This tends to reduce the potential for coarse grained HAZ cracking during PWHT or as a function of time for in-service exposure.

Conventional joint designs and the 0.06–0.08% weld deposit carbon content range are recommended with alteration of welding and PWHT procedures depending on the carbon content and MPC Factor.

For the most restrictive case of the high carbon content ($> 0.15\%$) and high MPC Factor, low weld heat input is recommended with controlled deposition welding procedures employed in critical weld regions. Control of the PWHT process should be undertaken with careful control of thermal gradients (ΔT) and PWHT should be accomplished at 1325°F. As with the Class 2 considerations, the base metal toughness may be significantly reduced by the more severe PWHT schedule and this should be attended to by the overall heat treatment schedule for the base material.

If the MPC Factor is reduced at the high carbon level or the carbon content is reduced to the recommended range of 0.11–0.14% with a high MPC Factor, the PWHT temperature may be reduced and a relaxation in the employment of controlled deposition welding procedures may be effected. Again, a concern for a reduced toughness at the 1300°F PWHT temperature may be evident with the high carbon material.

As the carbon content is further reduced and in combination with the MPC Factor as shown in the guidelines, the PWHT may be reduced to 1275°F minimum.

With a low carbon content and low MPC Factor the chance for HAZ cracking either during PWHT or in-service is virtually eliminated and the PWHT temperature may be reduced to 1250°F. Normalization will most likely be adequate to retain the Class 1 properties even at the low carbon level.

Repair Guidelines

The considerations for repair are presented in the Repair Guidelines flow chart (Fig. 2) which is divided into two categories depending on the required life of the vessel subsequent to repair.

The long term repair scenario (greater than 2 years)

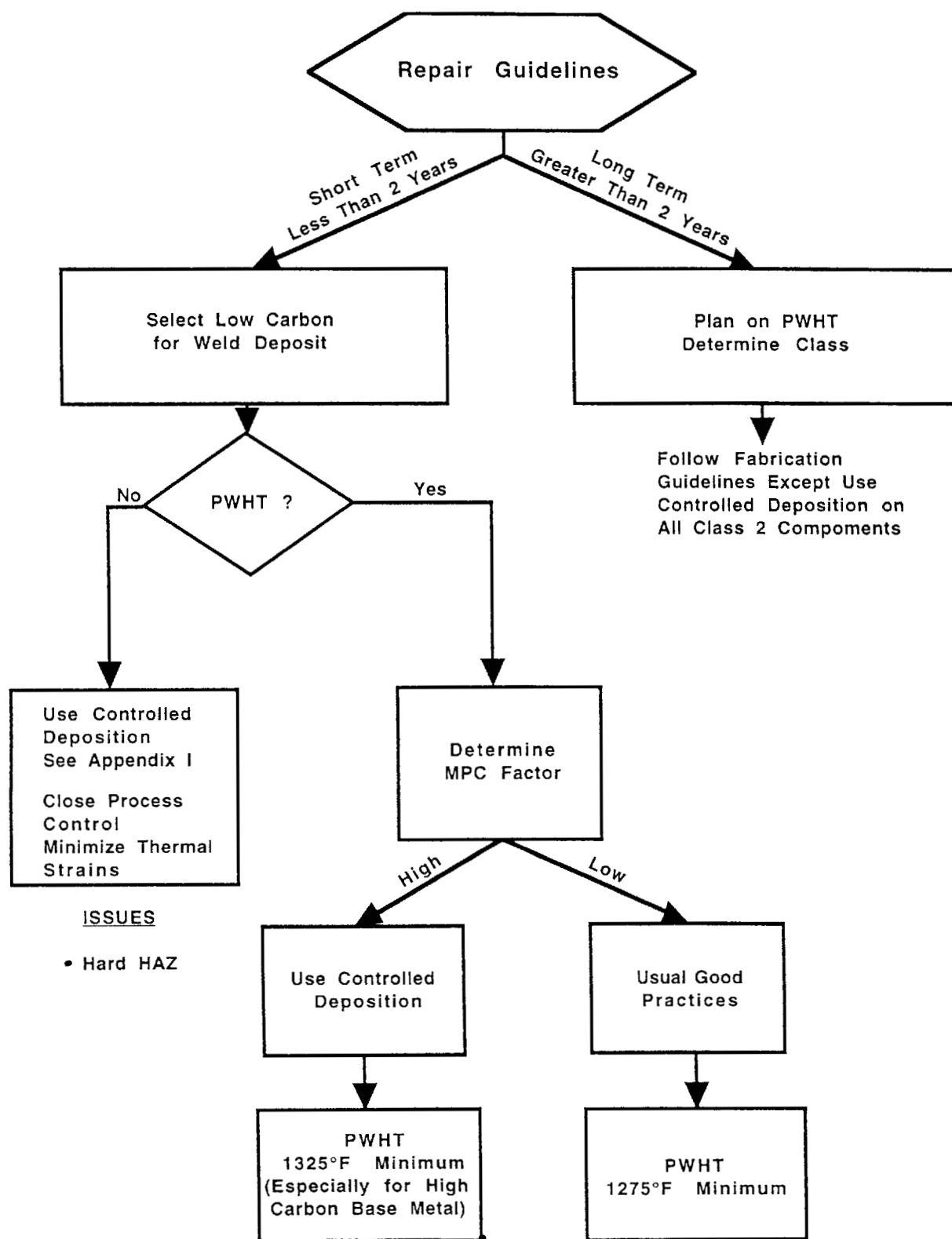


Fig. 2—Repair Guidelines

invokes considerations similar to the original fabrication and requires controlled deposition welding procedures on Class 2 components. The column in the Fabrication Guidelines appropriate to the carbon content and MPC Factor should be employed for long term repairs.

Short term repairs (less than 2 years) are governed by the considerations in the Repair Guidelines and it is recommended that a low carbon weld metal (0.025–0.035% carbon), E8018-B2L, be employed. (The AWS is currently changing the designation for the “L” Grade of the B2 class to reflect that the strength will not normally meet the 80,000 ksi level.) The use of this filler metal will aid in mitigating weld cracking problems (including cold cracking) and reduce the level of residual stresses that will occur as a result of localized repair. Furthermore, the creep properties of the low carbon weld metal will be adequate for short term service after repair.

If PWHT is to be employed, the MPC Factor should be determined and if the Factor is high, controlled deposition weld procedures should be used together with PWHT of 1325°F minimum. If the MPC Factor is low, normal welding practices can be employed and PWHT should be accomplished at 1275°F minimum. These precautions will mitigate the cracking potential and provide adequate creep behavior for the less than 2 year anticipated additional life.

The election to repair without PWHT involves many factors, but significant considerations should be given to this option if the component is to be used at elevated temperatures, especially in the creep regime, since PWHT related metallurgical changes will occur as a function of operational time thus softening the repair and relaxing the residual stresses. If the decision is to repair without the application of PWHT, the controlled deposition procedures defined in Appendix I should be used together with close process control to minimize thermal strains during welding and equipment start-up after repair. The elimination of the coarse grained HAZ by use of the controlled deposition procedures will result in the virtual elimination of the type of cracking responsible for most problems in the $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ system. The issue of hard HAZs must be addressed in terms of the service environment for the component or vessel.

Repairs by their very nature are critical operations for which all precautions and controls should be invoked and considered before the onset of repair. The repair cavity size and configuration should be chosen so that the controlled deposition procedures can be properly implemented and low heat input welding techniques always should be used.

Summary of Program and Results of Testing and Evaluation

Introduction

This program was predicated on the obtaining and testing of $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ materials which represented

wide ranges in composition and thus would cover the majority of materials currently in use or those that are now offered in the marketplace for the construction of petroleum/refinery vessels. A total of 17 lots representing 17 different heats of $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ materials were acquired from both foreign and domestic sources. The time span of the production of the materials was greater than 20 years. It was originally intended that materials from vessels that had experienced cracking problems would be used during the conduct of the program. Unfortunately, no such materials were made available. Nevertheless, it is felt that the range of compositions and behavior obtained within the 17 materials tested is sufficient to formulate approaches to vessel fabrication and repair that are meaningful and technically sound. However, partial compositions of four heats that were reported to have suffered severe service/fabrication problems were obtained. These partial compositions indicated that if the proposed MPC Factors had been used for screening, the materials would have been identified as susceptible to cracking.

The Appendixes that form the basis of this report contain detailed information on all aspects of the effort and provide support for the conclusions and recommendations offered. The report Appendix titles are shown in Table 1 and excerpts/conclusions from this data base are used in the following sections.

It is to be noted that a significant amount of highly complementary research was conducted under the auspices of the Welding Research Council (WRC) and the Pressure Vessel Research Council (PVRC). The results and benefits of that work are included here because of their importance to the conclusions and recommendations offered.

Section A. Literature Review

The literature review covers Cr-Mo steel vessel problem areas related to both reheat/PWHT cracking during fabrication and in-service cracking (“creep embrittlement”) related to weld HAZs. It is believed that these phenomena are closely related and the same metallurgical and fabrication conditions apply to them. The literature review contained in Appendix A covers the physical metallurgy of the Cr-Mo steels together with the transformation characteristics pertaining to weld HAZs. The microstructural evaluation in the HAZ is addressed in terms of carbide evolution during PWHT and service exposure and the relationship between the metallurgical changes and the cracking potential of the material. The elemental effects are discussed together with the various theories for reheat/PWHT cracking and in-service HAZ cracking. Compositional factors found in the literature to characterize material behavior, based on chemistry, and the testing techniques to reveal sensitivity to cracking are discussed. This literature review

Table 1 - List of Appendices

A	Literature Survey - Cr-Mo Steels - Reheat and In-Service Cracking
B	Chemical Composition of 1¼Cr-½Mo API Materials
C	Coarse Grained HAZ Transformation Behavior and Associated Microstructures
D	Assessment of Reheat Cracking Susceptibility
	D1: Gleeble Simulation Smooth Bar Reheat Crack Testing
	D2: Spiral Notch Testing
	D3: Development of a New Reheat Cracking Test - PREVIEW Test and Evaluation of Reheat Cracking in API Materials
E	Determination of Factors to Quantify Reheat Cracking Susceptibility Based on Chemical Composition
F	Toughness Study
G	Microstructural and Fractographic Evaluations
	G1: Fractographic Examination of Notched Creep and Gleeble Stress Rupture Samples of API Materials
	G2: SEM Metallographic Investigation and EDS Analyses of UT2 and UT3 Materials
	G3: High Resolution Electron Microscopic Evaluations on API Materials (Phase I&II)
	G4: Transmission Electron Microscopy Evaluation on API Materials
H	Creep Rupture Behavior of the HAZ
	H1: Notch Bar and Smooth Bar Creep/Stress Rupture Testing
	H2: Preparation of Extended Length HAZ Simulation Specimens by the Gleeble Technique
	H3: Sample Tests
I	Repair Welding Procedure - Behavior of Low Carbon Weld Metal and Repaired Weldments
J	Program Tasks

provides the background upon which the results of this program can be assessed and contrasted.

Section B. Materials

The materials employed in the program were obtained from both virgin heats and service exposed materials extracted from service piping and vessels. The service exposed materials were renormalized at 1650°F for 1 hour and tempered at 1150°F for 1 hour before being tested in this program. This return to the virgin state (RV) of the materials that were in the service exposed condition was considered very important. It had been thought that because the microstructure in the HAZ is changed (transformed and homogenized), during both the HAZ simulation cycles and actual weld tests, that all prior metallurgical (aging) changes that occur during service exposure would be erased. There were strong indications that this was not the case and, therefore, the RV treatment was utilized to provide a standard reference state.

The chemistry of all of the materials studied was obtained from the same laboratories and analysis for 23 elements was made. The analysis techniques and the 23 element chemistry for all materials tested is presented in Appendix B.

Section C. Weld HAZ Transformation Behavior

The transformation characteristics and the resultant microstructures in the coarse grained HAZ are important to the microstructural morphology changes during both PWHT and thermal exposure in-service. Thus, the continuous cooling transformation behavior was determined for two heats of the program materials which had significant differences in carbon content (0.10% vs. 0.18%). The remaining materials are expected to fall between these two extremes in hardenability.

The coarse grained HAZ was characterized because it is in this weld region that the cracking occurs for both the PWHT and in-service thermal exposure regimes. The thermal cycles for the coarse grained region spanned the heat input range of 12–120Kj/in. for a 1 in. plate preheated to 350°F. The CCT diagrams are presented in Appendix C together with representative microstructures for the UT4 and UT6 materials fully characterized. In addition, the microstructures for UT2, 3, 4, 5 & 8 are presented for the 12 and 120Kj/in. heat inputs only. It can be noted from the diagrams that the main constituents formed in the coarse grained region of the HAZ in the

$1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ system are principally martensite and bainite with the lower energy input welds in the higher carbon content material forming higher percentages of martensite than the lower carbon materials. Ferrite begins to form at high energy inputs especially primarily for the lower carbon materials. The range of hardness attendant in the coarse grained HAZ associated with each cooling condition is shown both on the CCT diagrams and in the bar graphs in Appendix C. These hardness levels are as expected with the lower carbon material showing lower hardness for all energy inputs and a greater change in hardness upon an equivalent increase in energy input than the higher carbon material. The optical micrographs clearly show the difference in the bainitic structures (carbide and ferrite) formed with higher energy inputs as contrasted to the martensitic (acicular and needle-like) constituent formed upon more rapid cooling. The transformation temperatures determined for the materials are also presented in a table in Appendix C for ready reference as additional data for use in evaluation of the test results.

Section D. Reheat/PWHT Cracking Assessment

In order to determine the reheat/PWHT cracking response of the 17 heats of $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ materials three test methods were employed. These three tests are described in Appendix D. One of the tests has been used for many years in assessing reheat/PWHT cracking tendency while the other two tests were developed at the University of Tennessee to answer special needs in reheat/PWHT cracking assessments. The Spiral Notch test was developed in order to assess which zone in the HAZ is most susceptible to reheat/PWHT cracking and to define the effect of different weld metal deposition techniques on the sensitivity of the HAZ to reheat/PWHT cracking. The PREVIEW (Petroleum Refinery Vessel Evaluation of Weldability) test was developed in the current program so that the reheat/PWHT cracking sensitivity could be determined from a scaled-up test that employs an actual weld deposit. The test overcomes an additional attribute of the Gleeble and Spiral Notch test in that the stress relaxes as a function of time at the test temperature in much the same manner as the residual stresses relax an actual weldment. Furthermore, the test incorporates natural weld contour conditions that introduce realistic stress concentrations for the initiation of reheat/PWHT cracking.

(a) The Gleeble stress rupture test employing a simulated HAZ has been used for over 15 years in assessing HAZ cracking during PWHT. The details of the test methodology have been defined by the work at The University of Tennessee and other researchers. Criteria have been established and significant correlations have been made with service performance. The test has been used as a basis for the determination of welding conditions necessary to

minimize the occurrence of reheat/PWHT cracking, the ranking of a material's sensitivity to reheat/PWHT cracking and for fundamental studies of the mechanism of reheat/PWHT cracking. The method of conducting the Gleeble simulation test is shown in Appendix D1 along with the test results.

The Gleeble tests were carried out using energy inputs of 12 and 120KJ/in. to span the range of energy inputs utilized for a wide range of welding procedures. The testing was carried out as a function of stress at a temperature of 1150°F and as a function of PWHT over the range of 1150–1350°F. The service exposed and the renormalized conditions were used for the ex-service material heats and the virgin condition was used for the new material heats. The samples were used for fractographic studies and for the determination of the hardness of the HAZ and the hardness changes as a function of PWHT. The full range of data is presented in Appendix D1 together with bar graphs. In addition to the ductility measurements that allow a definition of the material's sensitivity to reheat/PWHT three additional important findings are:

1. the employment of PWHT temperatures below 1250°F have only a minimal effect on the rupture ductility indicating that the material retains its sensitivity to coarse grained HAZ cracking unless a PWHT in excess of 1250°F is employed during fabrication (a PWHT greater than 1300°F significantly improves rupture ductility);
2. the hardness of the coarse grained HAZ does not fall rapidly at temperatures to 1200°F and then falls at rate considered to be slower than that anticipated. This hardness retention was noted in the study on the creep behavior of the coarse grained region as a function of PWHT temperature as explained in a later section, and
3. the high energy input places the coarse grained HAZ in a condition that is more susceptible to reheat cracking than the low heat input.

The significance of this result will also be further amplified in and explained in subsequent sections. The ranking of the materials by means of the Gleeble test will be presented after the discussion of the results of the PREVIEW test.

(b) The Spiral Notch test was introduced in order to more rapidly and better define the region of the HAZ that is susceptible to reheat/PWHT cracking in actual welds and to define the conditions which mitigate the occurrence of reheat/PWHT cracking. The test is conducted at a temperature of 1150°F at a constant stress that is indicative of the residual stress in a welded structure. The location of the cracking and rupture in the notched region defines the HAZ microstructure most sensitive to reheat/PWHT cracking and the stress range over which coarse grained cracking occurs is a secondary indication of a material

reheat/PWHT cracking sensitivity. The Spiral Notch testing methodology and test results, together with macrostructural evidence to show the type of ruptures indicative of both sensitive and insensitive materials is presented in Appendix D2.

The Spiral Notch test results on materials UT1-UT11 were evaluated in terms of verification of the Gleeble assessment. This was a one-to-one correlation with the data from both tests.

A significant amount of work was conducted in a PVRC study on the efficacy of the use of controlled deposition procedures in mitigating reheat/PWHT cracking on the material designated UT6 in this study (it is to be noted that UT6 material is the most sensitive to reheat/PWHT cracking of all of the materials examined in this study). The PVRC (1) study defined the conditions necessary for the mitigation of reheat/PWHT cracking by controlled deposition techniques. This PVRC study revealed that if the coarse grained region of the HAZ is eliminated by grain refinement from successive weld passes the reheat/PWHT cracking sensitivity essentially disappears. The Spiral Notch test was instrumental in defining this behavior for 10 full scale weldments utilizing UT6 material. This same elimination of reheat/PWHT cracking by controlled deposition coarse grained HAZ refinement was also shown for a low carbon precipitation hardening steel by the utilization of the Spiral Notch test. Further, the Spiral Notch test was used in an other study to define the differences in HAZ structure for down-hand and out-of-position welds. Thus, the Spiral Notch test has an inherent utility in the testing of full scale weldments duplicating the procedures to be used in production, in addition to testing simulated HAZ's for sensitivity.

(c) The development of the PREVIEW test was undertaken as a part of this program to provide a correlation with the Gleeble and Spiral Notch tests using a full scale weldment test. The development of the PREVIEW test is presented in Appendix D3. This test has the advantage of the utilization of an actual weld with stress relaxation, during the PWHT duplication, which is similar to actual weldment behavior. The stainless steel fixture is simply constructed and no specialized equipment, except for a suitable furnace, is required to conduct the test. The specimen size $10 \times 4 \times 2$ in. allows for actual welds using appropriate consumables and the employment of NDE methods for a rapid assessment of the test results. The tests were conducted on all materials whose configuration permitted the extraction of samples (thus not all materials were tested in this manner). A significant number of tests were conducted before the standardized procedures were defined and then utilized for all additional testing. Dye penetrant NDE is used after welding, after application of the test strain and after testing at elevated temperature to insure that the cracking found is representative of true reheat/PWHT cracking. The

degree of cracking in a test sample is indicative of the relative sensitivity of the material. The test is recommended with a fillet weld configuration for which all of the tests were conducted in this study. However, a butt weld configuration with a broached notch was also evaluated and found to be appropriate to the assessment of controlled deposition weld procedures.

The tests also showed that the hardness of the base metal (original plate condition) was important to the extent of reheat/PWHT cracking, in that the harder the base plate material the more extensive the cracking, which occurred in the coarse grained HAZ. This is due to the fact that a softer base material allows for more ready relaxation of stresses and thus reduces the magnitude of stresses present as a function of PWHT time in the HAZ. The location of the cracking clearly showed the influence of the condition of the weld toes in exacerbating the initiation of cracking in sensitive material. The more abrupt the fillet weld contour the greater the cracking tendency (the change from the base metal to weld metal should be smooth and gradual).

It is felt that the development the PREVIEW Test is significant as a verification test and as a test that can be utilized for the screening of materials and for the evaluation or selection of welding procedures which influence reheat/PWHT cracking.

The reheat/PWHT cracking sensitivity ranking of the program materials using the Gleeble and PREVIEW test is shown in Table 2. The data is presented in three categories: high, intermediate and low. The correlations are considered quite adequate and while more materials were rated with the Gleeble test than the PREVIEW the rankings are consistent. The Gleeble rankings are given separately for the low and high heat input evaluations and this further illustrates that at the high energy input conditions no material ranks in the low category. This ranking is of significant value when the assessment of the sensitivity based on chemistry of the base material is considered in the following section.

Section E. Development of Factors to Predict Reheat/PWHT Cracking

The literature presents more than 10 factors for predicting the behavior of the Cr-Mo materials based on heat chemistry. At least eight of these factors are in some way related to reheat/PWHT cracking. Appendix E, which describes the utility and determination of compositional factors for predicting behavior in the reheat/PWHT or in-service cracking regimes, presents the literature-derived factors (Appendix E, Table E2). These factors and the material chemistry for all 17 program materials are shown in the graphical presentations in Appendix E, Figs. E1-11. This assessment showed that the extent of scatter was too large to utilize any of the literature factors as an index of reheat/PWHT cracking sensitivity. Thus, a series of elemental factors based on multiple regression analy-

Table 2. Reheat Cracking Sensitivity Ranking

SENSITIVITY	PREVIEW TEST*	GLEEBLE TEST**	
		12 KJ/in	120 KJ/in
HIGH			UT-8
			UT-14
			UT-6
			UT-15
	UT-3	UT-3	UT-12
	UT-14	UT-6	UT-16
	UT-4	UT-14	UT-4
	UT-8		UT-5
			UT-13
			UT-3
			UT-9
			UT-7
INTERMEDIATE		UT-5	
		UT-15	UT-2
	UT-5	UT-4	UT-1
		UT-8	UT-11
		UT-9	UT-10
		UT-17	UT-17
LOW	UT-2	UT-16	
	UT-11	UT-7	
	UT-12	UT-11	
	UT-13	UT-1	
	UT-15	UT-13	
	UT-16	UT-12	
	UT-17	UT-2	
		UT-10	

*PREVIEW:

HIGH
INTERMEDIATE
LOWCRACKING
MINOR CRACKING
NO CRACKING

**GLEEBLE:

HIGH
INTERMEDIATE
LOW0-10% RA
10-20% RA
>20% RA

sis and mechanistically related criteria was undertaken. A statistician was employed to utilize the available ductility data and the 23 element analysis for 16 of the heats evaluated. The data were fitted and this resulted in MPC Factor 3 as shown in Fig. E12 in Appendix E. While the data treatment appeared to show a strong correlation it did not agree with accepted mechanistic-based models that consider elemental effects on behavior. Also, when the data from an additional heat, UT17, became available it did not

fit the statistical regression analysis as shown in Fig. E12 in Appendix E. Thus, the majority of the work to derive a factor that could describe reheat/PWHT cracking sensitivity based on chemistry was directed toward an elemental regression approach and a mechanistic approach as described in this appendix. From these efforts MPC Factors 5 and 7 emerged as the most predictive factors for reheat/PWHT cracking sensitivity. These two Factors are shown in Figs. 3 and 4. MPC Factor-5 combines a carbon function

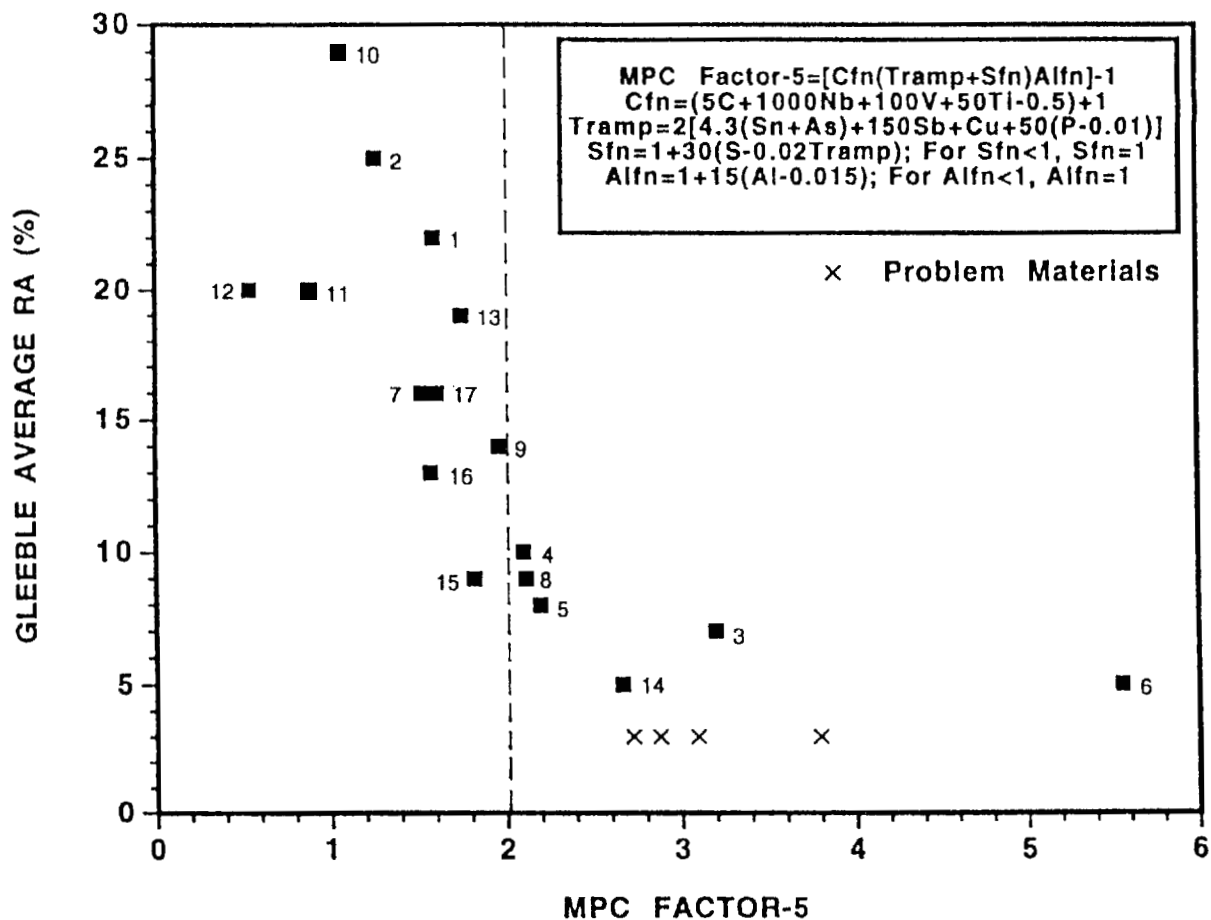


Fig. 3—Correlation of Gleeble reheat cracking results with MPC Factor 5

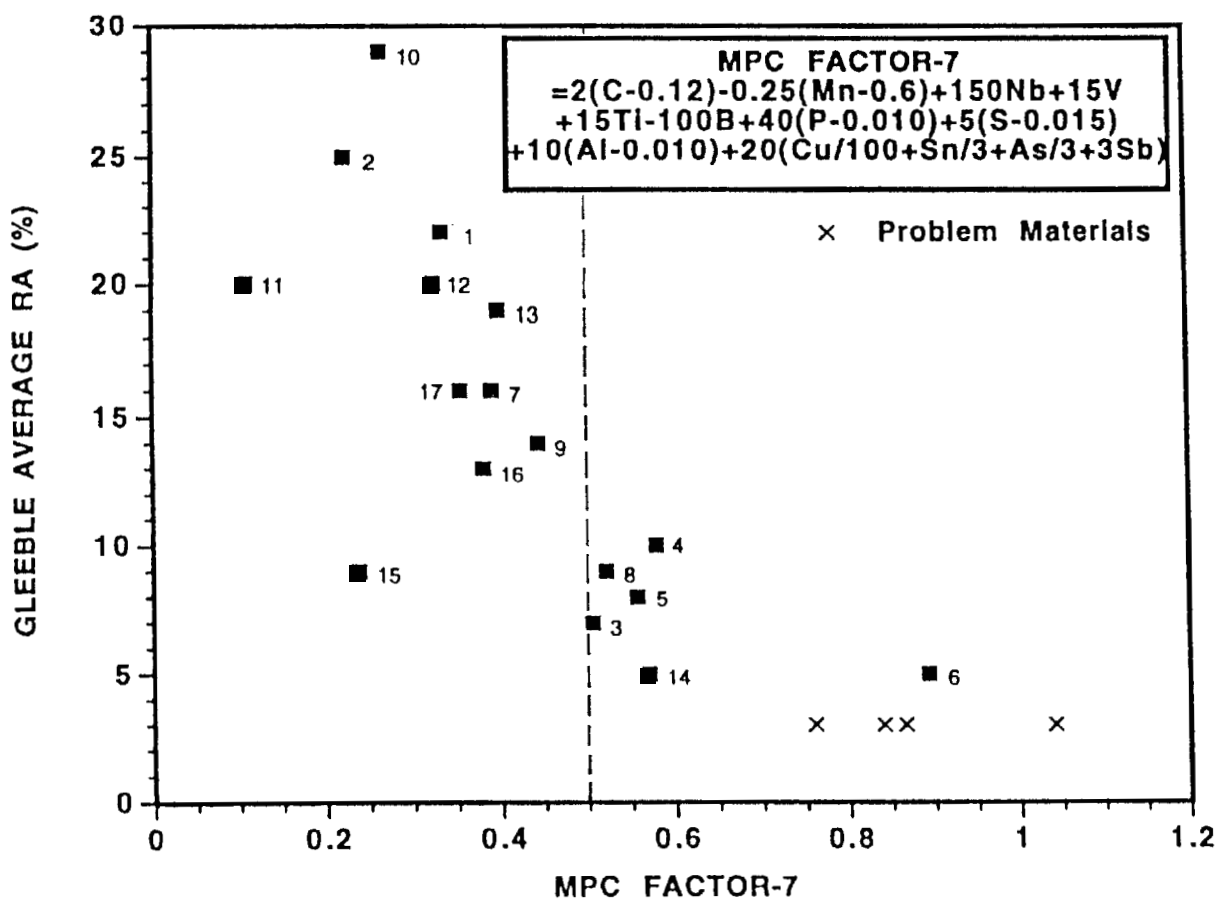


Fig. 4—Correlation of Gleeble reheat cracking results with MPC Factor 7

(strength), a tramp element function (embrittlement), a sulfur function (embrittlement) and an aluminum function. The combination of these functions is shown in the factor as stated in Fig. 3. MPC Factor-7 (Fig. 4) utilizes the concept of a lower limit cutoff as a basis for the effect of the elements in an additive fashion. It is felt that the sensitivity to reheat/PWHT cracking is assessed by either of these factors and a reasonable correlation exists with fabrication behavior. Using MPC Factor-5 the limiting value for the onset of a potential for reheat/PWHT cracking is 2.0 and for MPC Factor-7 the limiting value is 0.5. Thus, the evaluation of a material in regard to its reheat/PWHT cracking behavior can be made utilizing either of these factors. These are the factors to be employed with the recommendations for fabrication and repair presented previously in this report.

Since no problem materials were received from cracked vessels the literature description (5, 6) of four reactors that experienced HAZ cracking was used to provide a check on the factors described above. The elemental analysis for the 4 problem materials is shown in Table 3 and also in Appendix E. The entire spectrum of elements was not available in the literature reporting and for the elements that were not reported a value representing the lowest detectable limit was used in calculating the factor (Table 3). The problem materials are shown at an arbitrary selected reference ductility of 3% as Xs on Figs. 3 and 4. It is clear that these materials fall to the high side of the factors describing reheat/PWHT cracking behavior. Indeed, if the actual chemistry was available for the unreported elements in these problem heats, the points would fall at higher factor levels. Thus, the selected factors MPC Factor-5 and MPC Factor-7 appear to be relevant to practical cracking occurrences.

Section F. Toughness Evaluations as a Function of PWHT

The data obtained with regard to the effect of PWHT temperature on reheat/PWHT cracking potential clearly showed that PWHT at as high a PWHT temperature as possible is desirable in mitigating both reheat cracking and in-service low creep ductility (cracking). Thus, a limited evaluation of the changes in toughness attendant upon PWHT was made. Three materials from the 17 lots evaluated for reheat/PWHT cracking were selected spanning the full range of carbon content. Heats UT11 (0.086%C), UT12 (0.10%C) and UT5 (0.17%C) were PWHT over the range of 1250°F–1350°F for times to 8 hours. The full set of data obtained from the Charpy toughness evaluation as a function of PWHT is presented in Appendix F. Figs. 5, 6 and 7 summarize the toughness results for the three materials as a function of PWHT. It is clear by inspection of these figures that the low carbon (UT11) material showed little change in toughness as function of PWHT whereas the high carbon (UT5) heat showed a progressive deterioration with increasing in PWHT temperature. This effect is attributed to the precipitation and growth of carbides along the grain boundaries (not to embrittlement phenomena). The good initial toughness of the UT12 material enabled it to retain at least 40 ft-lbs at –40°F after a 1350°F 8 hour PWHT even though the as-received toughness was degraded by the PWHT.

These data are offered to illustrate the effect of PWHT temperature on toughness and the need to consider this aspect of material properties when selecting the proper material under all aspects of petroleum vessel fabrication and for optimum in-service performance.

Table 3. Chemical Composition of Problem Materials

Material	C	Si	Mn	P	S	Ni	Cr	Mo	Cu	Al	As	Sn	Sb	N	O
PM-1	0.15	0.67	0.57	0.014	0.012	0.15	1.28	0.53	0.25	0.018	0.024	0.025	NR*	NR	NR
PM-2	0.14	0.59	0.60	0.017	0.008	0.18	1.38	0.51	0.24	0.006	0.030	0.032	NR	NR	NR
PM-3	0.15	0.67	0.58	0.015	0.009	0.17	1.21	0.53	0.23	0.015	0.032	0.028	NR	NR	NR
PM-4	0.15	0.51	0.61	0.018	0.019	0.21	1.15	0.49	0.21	0.005	0.018	0.020	.0045	.0111	.0035

NR*: Not Reported

Values assigned to un-reported elements for factor calculations:

Ti - 0.0001%, V - 0.005%, Nb - 0.0001%, B - 0.000005%

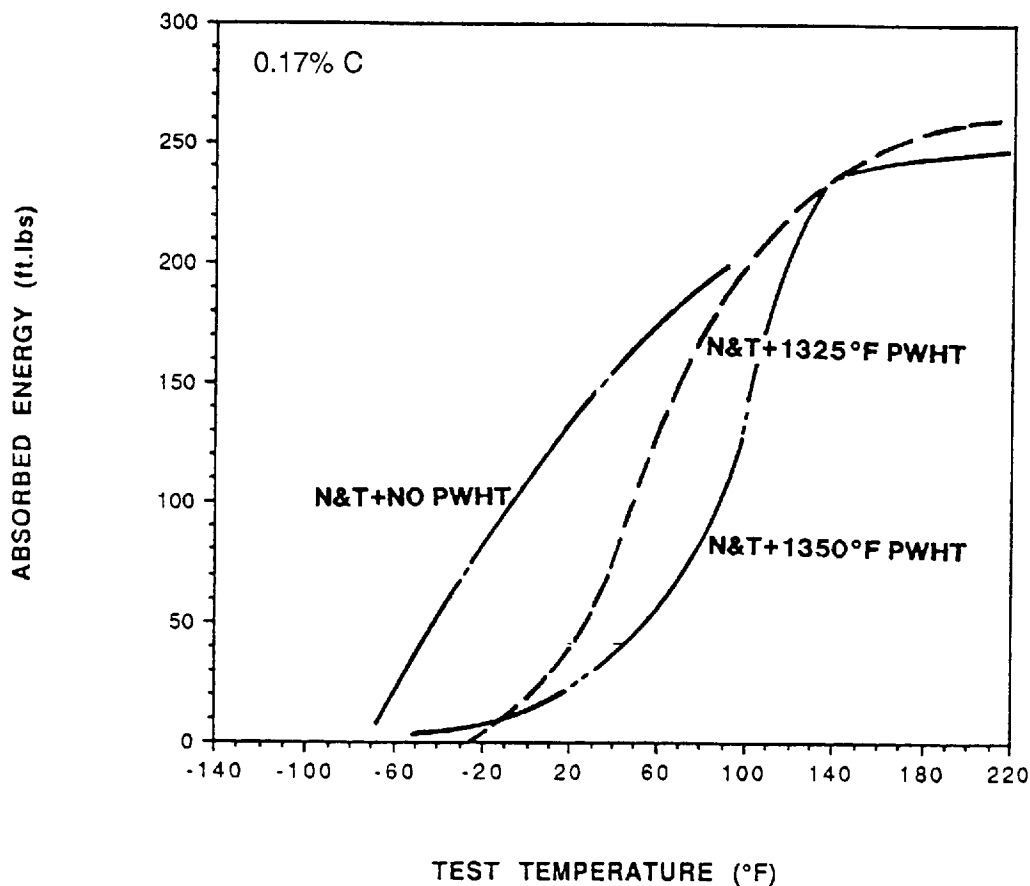


Fig. 5—UT5 CVN-ENERGY N(1650°F-40 min) & T(1320°F-86 min)

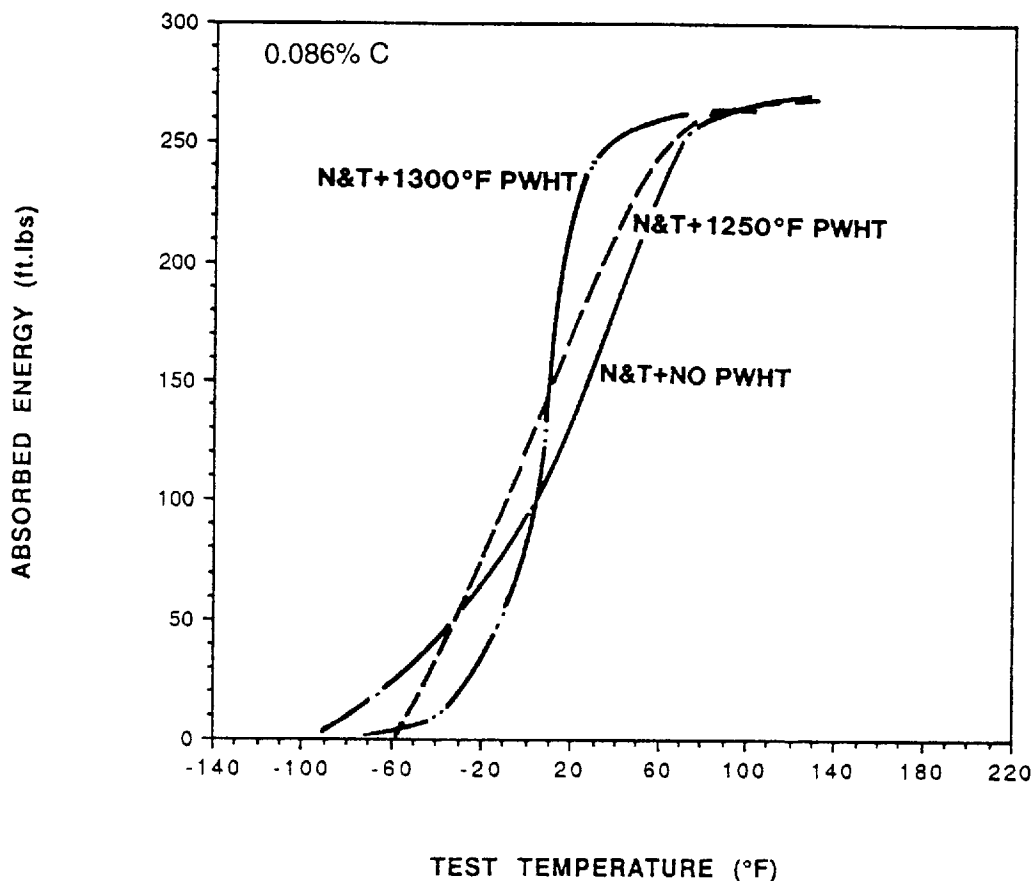


Fig. 6—UT11 CVN-ENERGY N(1650°F-1 hr) & T(1150°F-1 hr)

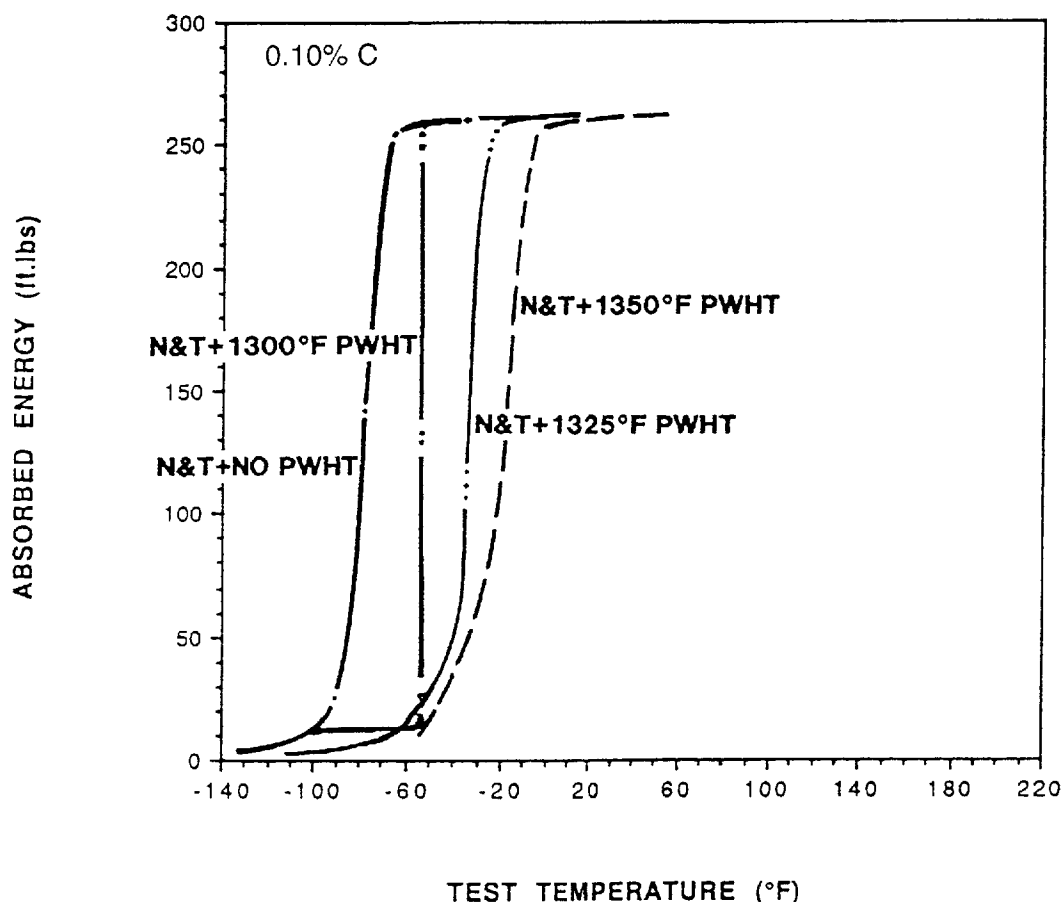


Fig. 7—UT12 CVN-energy, N(1700°F, ? hr) & T(1310°F, 30 min)

Section G. Microstructural and Fractographic Evaluations

Evaluation of the microstructural aspects of the behavior of the coarse grained HAZ was undertaken to provide the mechanistic aspects of the assessment of behavior and to provide a basis for the understanding of the test results. Fractographic examinations of tested reheat/PWHT cracking samples and creep rupture tested samples were also carried out to define the nature of the cracking and to relate fractographic features and elemental changes on the grain boundaries to rupture ductility. The full range of metallographic and fractographic studies is presented in Appendix G.

The reheat/PWHT cracking evaluations clearly showed that the high energy weld heat inputs resulted in lower rupture ductilities than the low weld heat input conditions. It was also found from the creep testing of the coarse grained HAZ that higher creep rates were evident for the low energy input conditions even though the initial HAZ hardness was high (indicating a stronger material). There are literature reports that indicate that higher temperature transformation products such as bainite result in a stronger material in the creep range than the lower temperature transformation product martensite. Extensive OLM, SEM, STEM and TEM work on the high and low heat input samples as a function of

PWHT temperature clearly showed that the carbide evolutionary sequence is distinctly different for the two microstructures. For a martensitic structure the carbides evolve to the more stable forms such as M_7C_3 , $M_{23}C_6$ and M_6C more rapidly as a function of time at equivalent temperature than does the initially bainitic HAZ characteristic of high weld heat input weld conditions. The high heat input weld conditions result in the formation of M_3C - and M_2C -type carbides that persist for long times and thus result in strengthening the material in the elevated temperature regime. The decrease in hardness, upon PWHT, for the coarse grained HAZ microstructures is less than anticipated from studies of normally heat treated base metal and thus the maintenance of strength at higher temperatures speaks to the need to PWHT at high temperatures to place the material in a more creep ductile condition.

These observations in several materials with widely differing reheat/PWHT cracking responses are not in conflict with the traditional assessment of reheat cracking tendency. The cause of reheat/PWHT cracking most likely does not lie in only one aspect of material behavior and thus the approach to the MPC Factors with the mechanistic approach, incorporating a variety of effects, including strengthening, tramp elements, effect of all elements on hardenability, carbide precipitation kinetics and creep rate differences are all important in the ultimate causation.

Elemental species such as manganese and boron appear to influence creep behavior (increase creep rate) by enhancing hardenability and martensite formation in the HAZ for equivalent levels of other elements. This effect on creep rate may be an important ancillary feature that leads to reduction of residual stress without the strain accumulation and rupture of grain boundaries which are the sites for segregation of the tramp elements, P and S.

Section H. Creep Rupture Behavior of the Coarse Grained HAZ—Notch Bar and Smooth Bar Creep/Stress Rupture Testing

An extensive matrix of smooth and notch bar creep and rupture testing established the fact that PWHT temperatures should be above 1275°F to improve the HAZ ductility (Tables H1–H5 in Appendix H). Also, these tests provided further justification for avoiding high heat input welds. Ductilities of equivalent zones were about twice as high at 45 Kj/in as at 120 Kj/in.

It was found that PWHT temperatures of 1350°F and above did not significantly reduce the rupture lives of the hard heat affected zone materials. The remarkably low ductilities and low creep rates observed for the hard portions of 1¼Cr-½Mo heat affected zones goes a long way to explaining the tendency for cracking in-service. There is no evidence that the materials become brittle in time (creep embrittlement). Instead it is concluded that brittleness is a consequence of the as-tempered microstructure that must be softened significantly before ductility can be observed. Some heats tended to show low ductility and notch sensitivity even after extensive tempering. Such persistent crack susceptibility appeared to be due to the same impurity factors that contributed to susceptibility to cracking during PWHT.

Section I. Repair Welding Procedures, Behavior of Low Carbon Weld Metal and Repaired Weldments

The subject of optimum repair procedures for 1¼Cr-½Mo materials have been addressed in a PVRC study (1) dealing with the considerations of the effects of both controlled deposition and PWHT on the efficacy and life of repaired weldments. The initial study was followed by an evaluation of the use of low carbon weld metal as an adjunct to repair for limited life or for repairs that do not employ PWHT.

The PVRC study, which was conducted with industrial involvement from the repair procedure development aspects, tested 10 full scale weldments made with program material UT6 (the most sensitive material to reheat/PWHT cracking) and clearly demonstrated that if the welding procedure used a controlled deposition approach, which was aimed at the elimination of the coarse grained HAZ, the potential sensitivity toward reheat/PWHT cracking was eliminated. In like manner, if a conventional welding procedure was employed which resulted in a signifi-

cant amount of coarse grains in the HAZ, the reheat/PWHT cracking tendency was high.

The replacement of the coarse grained HAZ with a completely refined region adjacent to the weld in the HAZ causes some concern in terms of elevated creep rupture behavior. It is well known that a fine grained material creeps at a greater rate than a coarse grained material. Thus, the total creep life of a controlled deposition repaired weldment might be reduced over that of a conventional weldment. This concern was answered in the PVRC study by numerous creep rupture tests of the controlled deposition weldments. The creep rupture samples behaved in a ductile manner and the life of the weldment, based on a Larson-Miller approach, showed that the failure times fell within the virgin base metal data band (between the minimum and mean). The ductility revealed in these tests was good and thus the potential for in-service low ductility cracking is considered negligible. The PWHT weldments behaved in a similar manner, in the creep regime, provided that the PWHT temperature was above 1250°F. Thus, the controlled deposition methods produced elevated temperature behavior similar to the PWHT weldments. However, it is to be noted that the HAZ hardness in the controlled deposition weldments is significantly higher than in the PWHT weldments but it is below that of the conventional weldments. Therefore, in regard to repairs for which there is a significant consideration for cracking due to the presence of coarse grained regions, either during PWHT or after the structure is returned to service, the controlled deposition methodology should be strongly considered. Further, if the weld metal is to be used in an environment where hydrogen cracking is possible the hardness level attendant with the non-PWHT repair methods must be addressed.

The controlled deposition repair procedure, as stated in Appendix I, has been used to repair ex-service weldments from both petrochemical plants and steam power plants. Long seam welds in these components were repaired using low carbon (0.025%) SMAW filler (E8018-B2L) and tests have been conducted using full scale jumbo creep samples of full thickness, incorporating all of the service exposed material and the weld repair. The initial results of this work have been reported to the PVRC Committee on Welds. The behavior of these repaired weldments, based on a Larson-Miller approach, shows lives in at the mean of the virgin base metal data band. Comparison tests of the PWHT repairs and the original service-exposed weldments are currently in progress.

Testing of the low carbon weld metal is underway and the early results show that the weld metal creep rupture strength, in the as-welded condition, exceeds the minimum Larson-Miller expectations. PWHT weld metals are in test.

The toughness of the low carbon deposit was determined for the as-welded condition and after PWHT at 1350°F for 8 hours. Summary curves are shown in Figs. 8–10, which reveal that the 1350°F/8

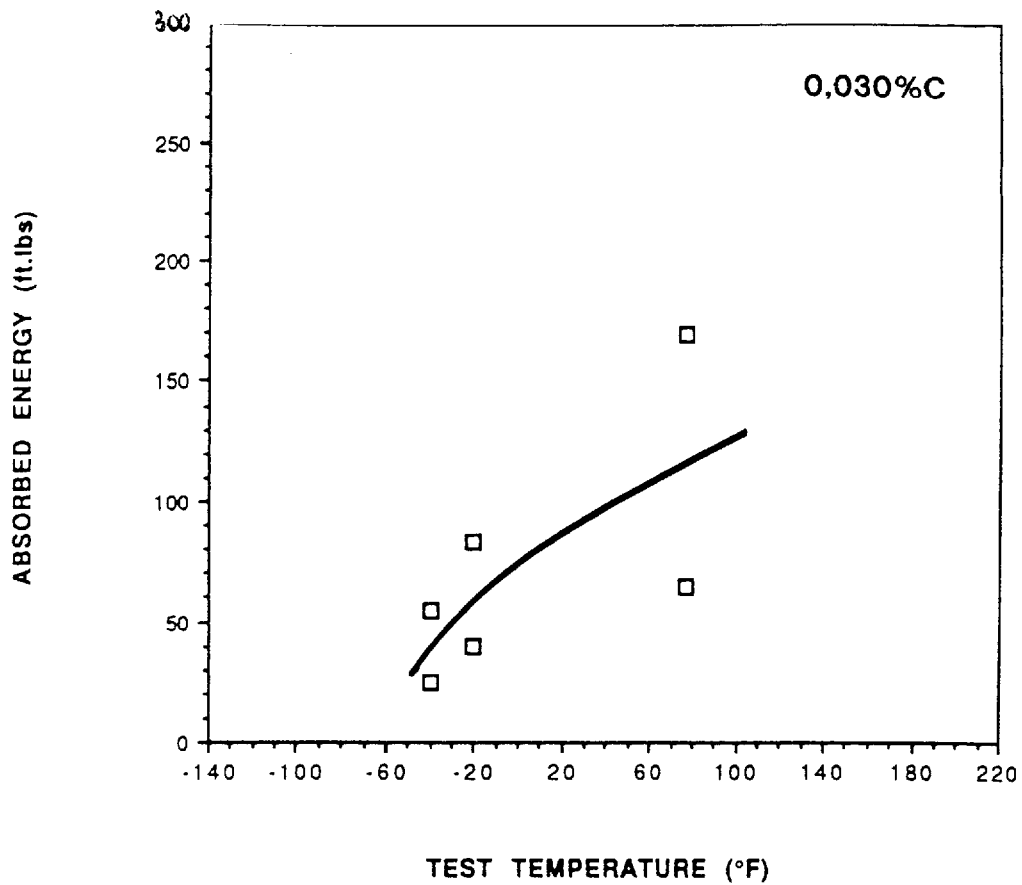


Fig. 8—CVN-absorbed energy, low C SMA repair weld metal (E8018 B2L) in as-welded condition

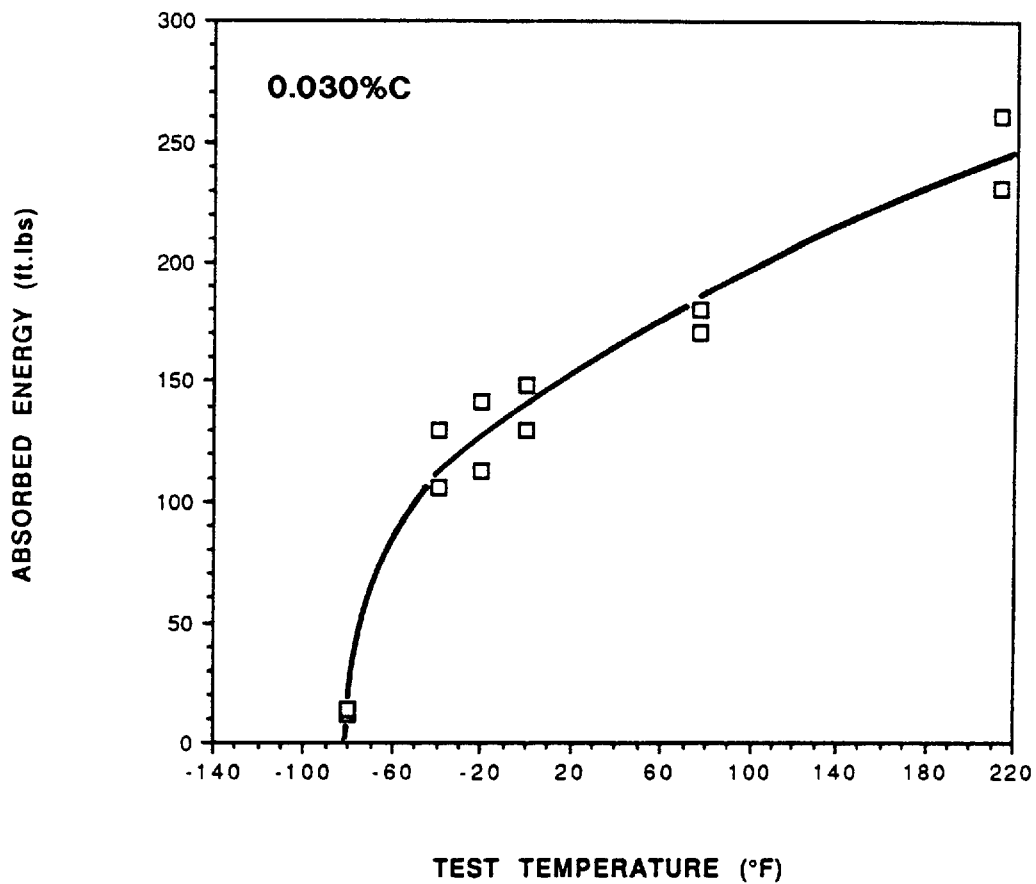


Fig. 9—CVN-absorbed energy, low C SMA repair weld metal (E8018 B2L) PWHT, 1350°F for 8 hrs

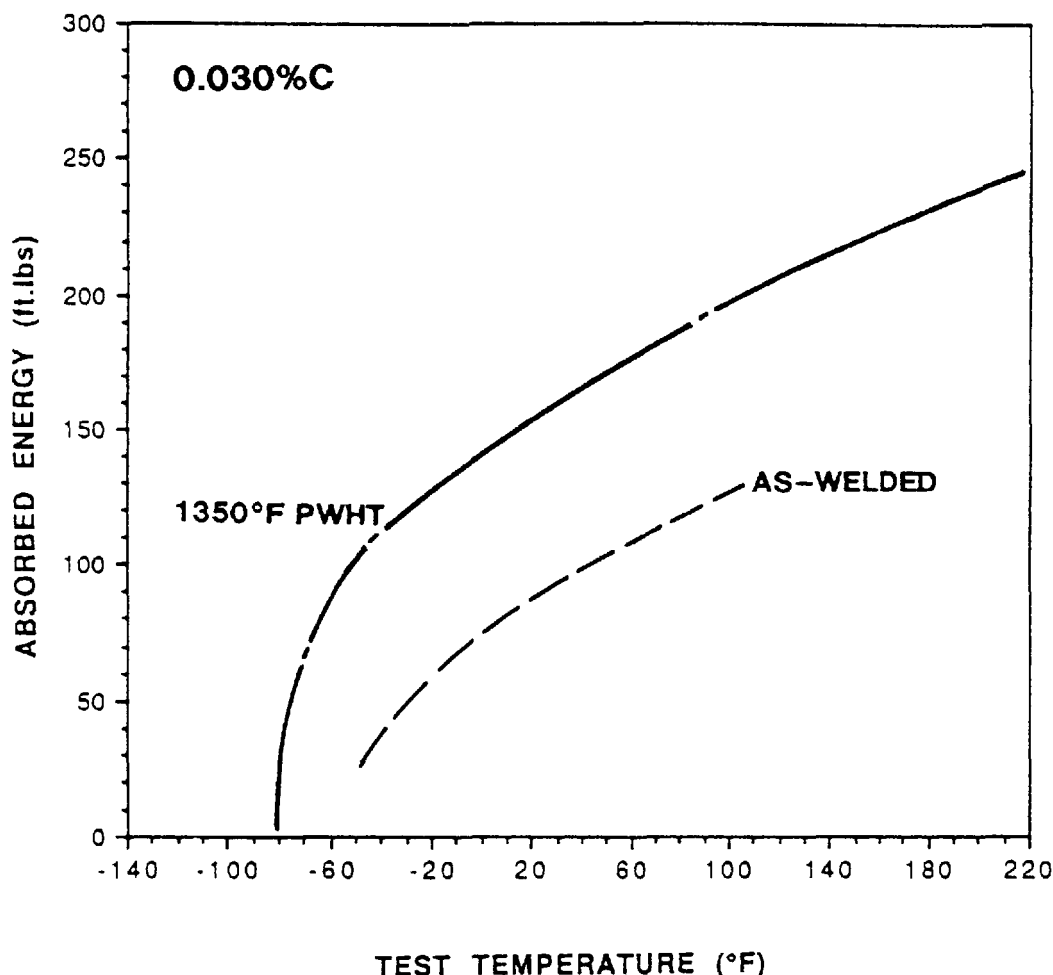


Fig. 10—CVN-absorbed energy low C SMA repair weld metal (8018 B2L) PWHT, 1350°F for 8 hrs/as-welded

hour PWHT significantly improved the weld metal toughness over that of the as-welded condition. The toughness levels for both conditions exceeded 40 ft-lbs at -40°F.

Thus, for short term repairs, the use of a low carbon filler metal, which should reduce residual stress and enable the weld to be made with less concern for hydrogen cracking, should be adequate from both the toughness and creep resistance standpoints. Therefore, this aspect of weld repair is to be considered in the repair methodology presented in the recommendations accompanying this report.

References

1. Lundin, C. D. and Wang, Y. "Half-Bead/Temper-Bead/Controlled Deposition Techniques for Improvement of Fabrication and Service Performance of Cr-Mo Steels," Draft Final Report Submitted to the Committee on Welds of the Pressure Vessel Research Council (December 1993).
2. Lundin, C. D. and Khan, K. K. "Fundamental Studies of the Metallurgical Causes for Reheat cracking in 2¼Cr-1Mo, 1¼Cr-½Mo and Copper Precipitation Hardenable Steels and Problem Mitigation," Final Report to the Weldability Committee of Welding Research Council (January 1993).
3. Lundin, C. D., Khan, K. K., Zhou, G. and Al-Ejel, K. A. "The Efficacy of the Utilization of Low Carbon Cr-Mo Weld Metal for Repairs in Cr-Mo Vessels and Piping," Progress Report Submitted to the Committee on Welds of the Pressure Vessel Research Council (January 1994).
4. Lundin, C. D., Khan, K. K., Zhou, G. and Liu, P. "The Efficacy of the Utilization of Low Carbon Cr-Mo Weld Metal for Repairs in Cr-Mo Vessels and Piping," Progress Report Submitted to the Committee on Welds of the Pressure Vessel Research Council (May 1994).
5. Nomura, T. et al. "Creep Embrittlement of Structural Components in Catalytic Reformer Reactor," Trans. Japan Soc. of Mechanical Engineers, 1993-9, pp. 2066-2073.
6. Cantwell, J., Private Communication to M. Prager of MPC (May 1993).

Appendix A—Literature Survey: Cr-Mo Steels—Reheat and In-Service Cracking

HAZ Transformation Behavior and Microstructure

The metallurgical transformations that occur during welding affect the final microstructure and therefore can influence many problems that can develop during and after welding. The coarse grained heat affected zone (CGHAZ) is the location of maximum susceptibility for reheat cracking, stress rupture/relief cracking (SRC) or postweld heat treatment (PWHT) cracking. It is also a primary region for reduction in toughness.

In a discussion of reheat cracking by Ito and Nakanishi,⁸ they indicate that in 1Cr-½Mo alloys a HAZ microstructure consisting of martensite or lower bainite was more susceptible to PWHT cracking than upper bainite. In temper embrittlement, a related materials problem, it was found that a martensitic microstructure is more prone to a loss in ductility and toughness than a bainitic microstructure. Thus, the determination of HAZ transformation characteristics is a first step in determining the weldability of a material which may, in turn, provide the key to reducing or eliminating weld HAZ problems.

Easterling⁹ has compared the microstructural regions of a weld with the equilibrium diagram (Fig. A1). However, such a representation is overly simplis-

tic in that it ignores major differences between the weld thermal cycle and the conditions that are utilized in establishing the equilibrium diagram. Welding can induce rapid heating (3000°F/sec) and cooling (500°F/sec) rates resulting in conditions far from equilibrium. Furthermore, the complete homogenization, required for equilibrium, never exists upon welding. Also, equilibrium considerations do not include such nonequilibrium constituents as martensite or bainite.

Many of the objections to the use of the equilibrium diagram to predict weld HAZ transformations also extends to the use of standard continuous cooling diagrams.¹⁰ These diagrams are developed starting with homogeneous austenite. In welding, inhomogeneity occurs due to the inability of alloying elements to diffuse uniformly throughout the austenite and the incomplete solution of carbides, nitrides and other constituents as a result of the rapidity of the welding thermal cycle and the concomitant short austenitizing times. In order to predict accurately the on-cooling transformation temperatures and microstructures, weld HAZ continuous cooling transformation diagrams must be derived using the heating and cooling conditions attendant upon welding.

Fig. A2 shows a conventional continuous cooling transformation diagram for 2¼Cr-1Mo and Fig. A3

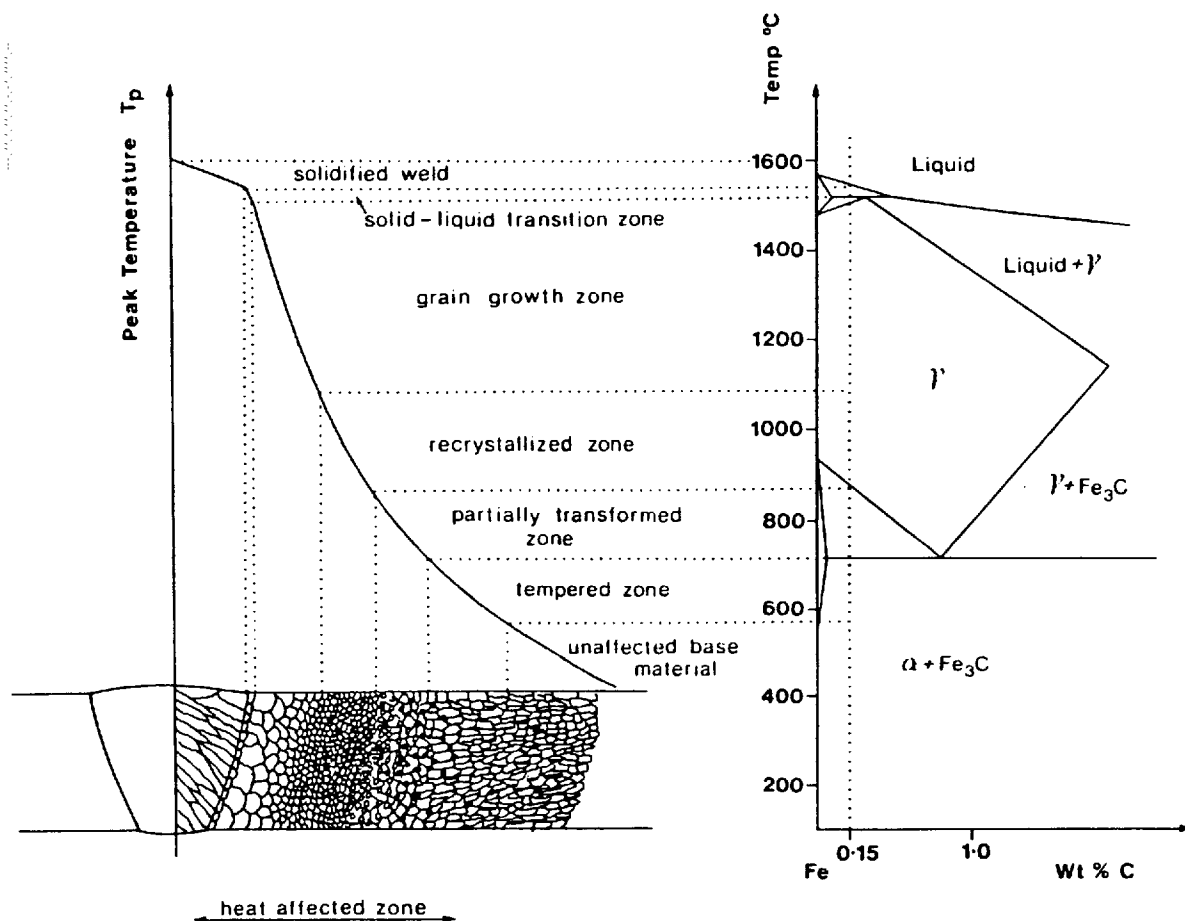


Fig. A1—Schematic diagram of the various regions of the HAZ approximately corresponding to the 0.15% carbon indicated on the Fe-Fe₃C equilibrium diagram. Source: Easterling, K., Introduction to the Physical Metallurgy of Welding, Butterworths, 1983

illustrates a diagram determined under simulated welding conditions for the CGHAZ in 2¼Cr-1Mo.¹¹ As may be seen by comparing Figs. A2 and A3, the depression of the on-cooling transformation temperatures (principally bainite) under welding conditions is approximately 90°F due to the rapid heating and cooling rates and short austenitizing times.

A literature review by Lundin *et al.*¹¹ revealed that only a few continuous cooling transformation diagrams have been determined under welding conditions for the Cr-Mo materials. However, conventional continuous cooling diagrams are available for many of the unmodified and modified Cr-Mo alloys. Continuous cooling transformation diagrams for various Cr-Mo steels are shown in Figs. A4, A5 and A6. The effects of the addition of vanadium, titanium and boron to the 2¼Cr-1Mo and 3Cr-1Mo alloys on the continuous cooling transformation behavior are shown in Figs. A4 and A5 by the superposition of the continuous cooling transformation diagrams for the unmodified and modified materials.

The 2¼Cr-1Mo and 3Cr continuous cooling transformation diagrams (Figs. A4, A5 and A6) give clues to the fact that the resulting microstructure under various welding conditions is complex. Depending on the degree of homogenization and the cooling rate (related to the heat input and preheat for a given

process and material thickness), the on-cooling microstructures in the weld HAZ may consist of martensite, mixed martensite and bainite or bainite coupled with retained austenite.

The microstructure of the 2¼Cr and 3Cr steels may be further complicated by the formation of martensite-austenite islands (a martensite-austenite constituent).¹² The formation of a martensite-austenite constituent is due to the partitioning of carbon to the austenite during the bainite transformation reaction resulting in locking of dislocations which prevents the shear transformation from occurring¹³ or the stabilization of austenite.^{14,15} The last austenite present can be highly enriched in carbon. Carbon contents of the martensite-austenite constituent have been reported by Biss and Cryderman¹⁴ to exceed 0.5 wt% in a nominal 0.15% C alloy and to be approximately 3 at % (approximately 0.7 wt%) in a series of 0.3C-3Cr-0.5Mo as shown by Thomas *et al.*¹⁵

Biss and Cryderman¹⁴ found that slow cooling rates enhanced formation of the martensite-austenite constituent by allowing carbon to diffuse away from the ferrite-austenite interface into the austenite. However, rapid cooling rates resulted in higher ferrite-austenite interface carbon content due to carbon diffusion being slower than interface advancement. This results in enhanced cementite precipitation and

	C	Si	Mn	P	S	Cr	Mo	V	Ti	B	Austenitizing Temperature
2¼Cr-1Mo-V-Ti-B	.10	.02	.51	.005	.005	2.20	.94	.30	.022	.0023	950°C (1742°F)
2¼Cr-1Mo	.10	.34	.48	.017	.013	2.16	.96	—	—	—	920°C (1688°F)

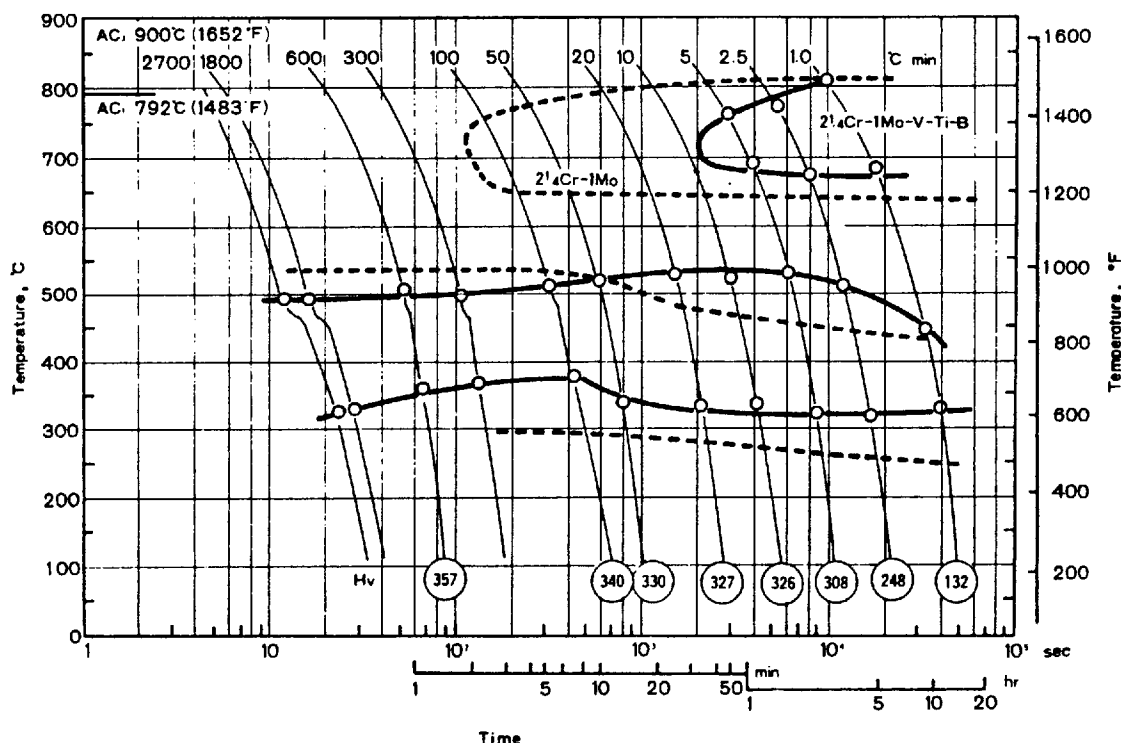


Fig. A4—CCT diagrams for V-Ti-B modified 2¼Cr-1Mo steel and standard 2¼Cr-1Mo steel. Source: Ishiguro, T., Murakami, Y., Ohnishi, K. and Watanabe, J., "A 2¼Cr-1Mo Pressure Vessel Steel with Improved Creep Rupture Strength," Applications of 2¼Cr-1Mo Steel for Thick Wall Pressure Vessels, ASTM-STP 755, 1982, pp. 129-147

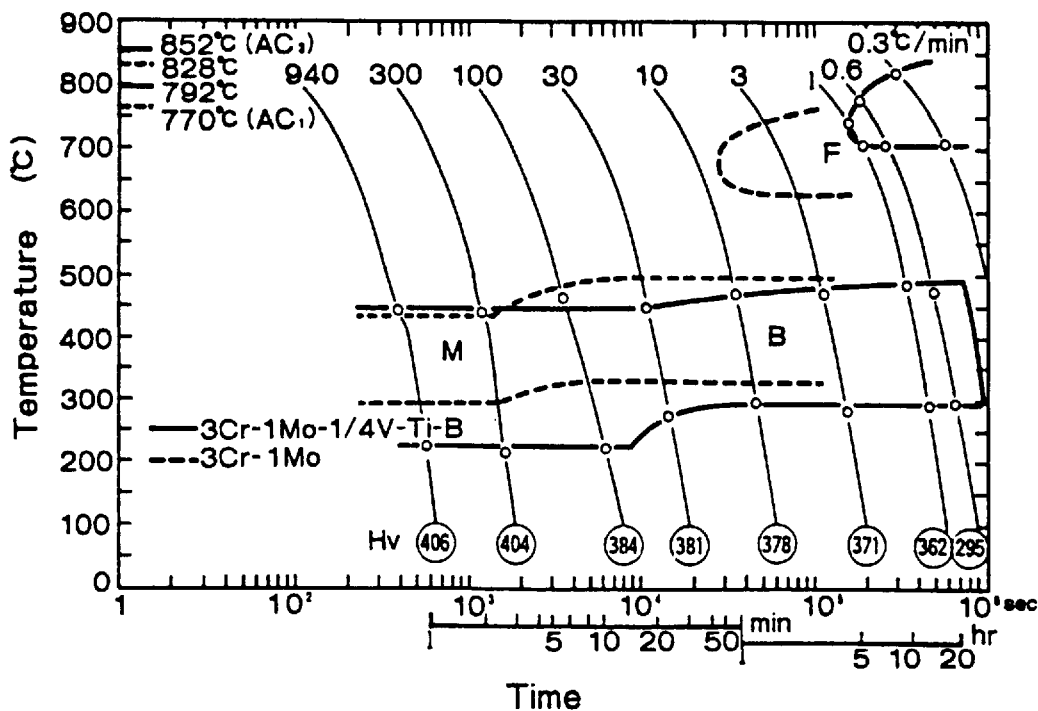


Fig. A5—CCT diagrams of a 3Cr-1Mo and a 3Cr-1Mo- $\frac{1}{4}$ V-Ti-B steels. Source: Ishiguro, T., Murakami, Y., Ohnishi, K. and Watanabe, J., "A 2 $\frac{1}{4}$ Cr-1Mo Pressure Vessel Steel with Improved Creep Rupture Strength," Applications of 2 $\frac{1}{4}$ Cr-1Mo Steel for Thick Wall Pressure Vessels, ASTM-STP 755, 1982, pp. 129-147

suppression of the martensite-austenite constituent. Economopoulos and Habracken¹³ found that the presence of the martensite-austenite constituent was associated particularly with massive, or granular,

bainitic structures formed at slow cooling rates. Wada and Eldis¹² found martensite-austenite islands in 2 $\frac{1}{4}$ Cr-1Mo steel under a slow cooling rate of 4°C/sec. The cooling rate dependence of martensite-austenite

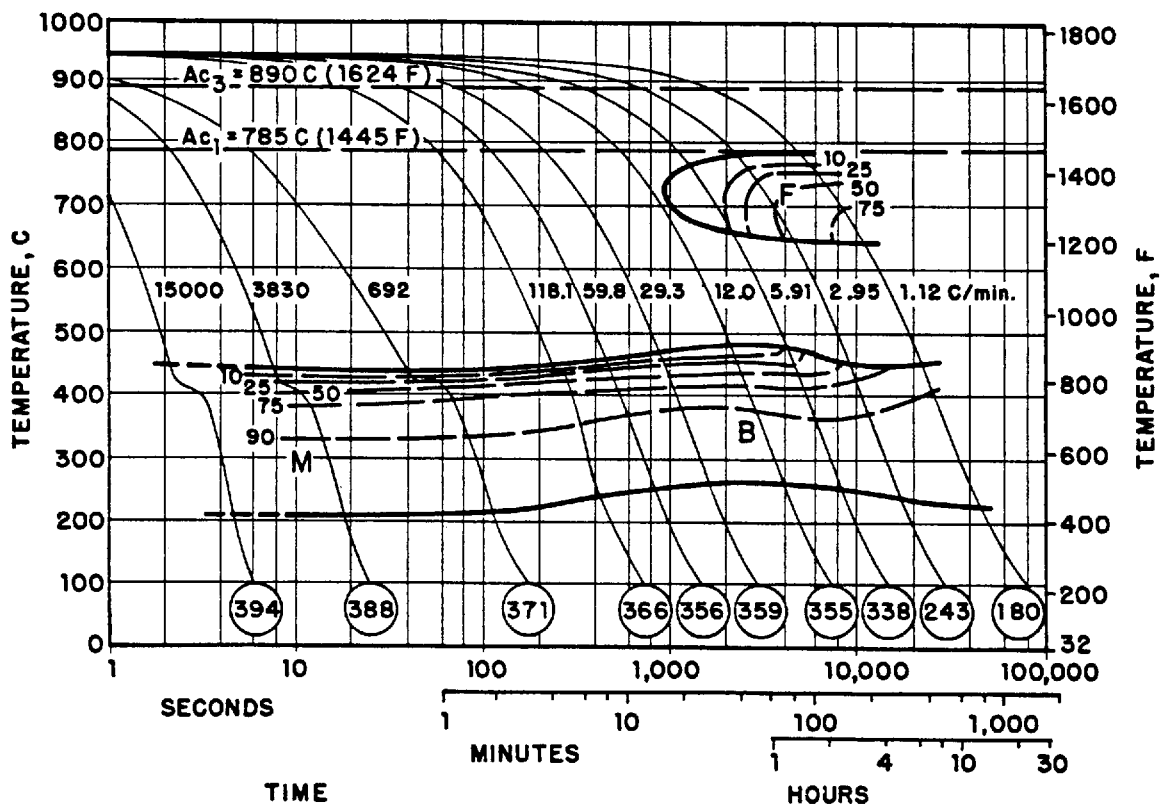


Fig. A6—CCT diagram for a commercial heat of 3Cr-1 $\frac{1}{2}$ Mo steel, Austenite grain Size: ASTM No. %. Source: Wada, T. and Cox, T. B., "A New 3Cr-1 $\frac{1}{2}$ Mo Steel for Pressure Vessel Applications," MPC-21 Research on Chrome-Moly Steels, ASME, 1984, pp. 77-94

constituent formation probably accounts for the inability of Lundin *et al.*¹¹ to detect any martensite-austenite constituent in their study of the transformation characteristics of 2¼Cr-1Mo steels under simulated welding conditions. However, Thomas *et al.*¹⁵ detected thin films of retained austenite along martensite laths by TEM examination. It was claimed that the retention of austenite was due to carbon redistribution during the martensite reaction and that this diffusion of carbon was possible due to the high M_s temperature (572°F (300°C) or higher) and the time to cool through the temperature range for martensite formation.

Apblett *et al.*¹⁶ investigated the transformation behavior of the HAZ in 1¼Cr-½Mo and 2¼Cr-1Mo steels. They found that these two steels essentially transformed to proeutectoid ferrite and bainite and the extent of either constituent varied depending on the peak temperature experienced and the cooling rate. In regions containing homogeneous (or nearly so) austenite, that is, regions which have been heated to peak temperatures of 2000°F (1095°C) and above, the ferrite reaction is suppressed and only a bainitic reaction occurs. The reaction start temperatures are in the vicinity of 1000°F (540°C) depending on the peak temperature and grain size of the austenite. Regions heated between 1750–2000°F (955–1095°C) contain undissolved carbides. These carbides act as nucleating sites for the formation of proeutectoid ferrite in addition to bainite.

In the portions of the HAZ heated in a temperature range between 1450–1750°F (790–955°C), austenitization is limited only to those regions in the immediate vicinity of the grain boundaries. This continuous network of austenite may transform to martensite which can result poor impact toughness. In general, the same trends are found in 2¼Cr-1Mo steel as in the 1¼Cr-½Mo steel. An increase in alloy content only tends to reduce appreciably the amount of proeutectoid ferrite in the microstructure.

Two factors should be evident from the above discussion of transformation characteristics and resulting microstructure. Since the continuous cooling transformation diagrams are important to the understanding of properties and potential cracking susceptibility, there exists a need to determine the continuous cooling transformation diagrams for welding conditions as the development and understanding of the weldability of the Cr-Mo alloys continue. Also, since the possible role of partial austenite transformation on HAZ softening has not been previously addressed due to temperature excursions into the intercritical region, a need exists to evaluate the effect of partial transformation on HAZ softening.

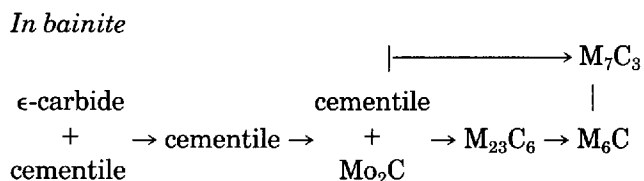
Microstructural Evolution in the HAZ upon PWHT

A major function of PWHT is to restore ductility in the HAZ and weld metal in Cr-Mo weldments.¹⁸ In addition, the PWHT also reduces the residual stresses in the weldment by a creep relaxation process.

Recommended practices for welding Cr-Mo steels are detailed in ANSI/AWS D10.8-86.¹⁸ The recommended postweld heat treatment temperatures and holding times for the various grades of Cr-Mo steels are often given as follows. For 1¼Cr-½Mo the recommended temperatures for PWHT are 1175–1275°F (635–690°C); for components intended for creep service and 1275–1350°F (690–730°C) for components where resistance to corrosion and hydrogen embrittlement are the primary considerations. For 2¼Cr-1Mo the recommended temperature is 1275–1375°F (690–745°C). Holding times are generally one hour per in. of thickness up to two in. and 15 min for each additional inch of thickness.

During a weld thermal cycle all or part of the carbides are taken into solution depending upon the peak temperature experienced, the energy input and the material thickness. During subsequent cooling the transformed matrix (bainite/martensite/ferrite) remains supersaturated with respect to carbon as well as alloying elements that subsequently precipitate as carbides during tempering. The various types of carbides that occur in Cr-Mo steels are MC, M_2C , M_3C , M_4C_3 , M_7C_3 , and M_6C . The carbide types, size, distribution and morphology will depend on the chemical composition, microconstituents present and the tempering temperature and time. The niobium, titanium and vanadium carbides are more stable than the chromium, molybdenum or iron carbides.

Baker and Nutting¹⁹ have shown that the types of carbides present in 2¼Cr-1Mo base metal are dependent on starting microstructure, heat treatment (tempering) and time at the tempering temperature. After normalizing, the microstructure is generally found to consist of a mixture of ferrite and bainite whereas after quenching the microstructure is mainly bainitic. They determined that the carbide evolution in the bainitic regions of both quenched and normalized material was similar, as shown below:



However, in the ferritic regions the Mo_2C type carbides do not undergo all the transition carbides and transform directly to M_6C carbides. Mo_2C , the carbide conferring the greatest resistance to creep deformation, was found to be more stable in ferrite than in the initial microstructures consisting of martensite or bainite. In the martensitic and bainitic microstructures $M_{23}C_6$ grew at the expense of Mo_2C , resulting in a degradation of the creep properties with increasing tempering temperature or time at a tempering temperature.

Because of the stability of Mo_2C within the ferrite, Baker and Nutting¹⁹ recommended the use of normalized and tempered 2¼Cr-1Mo rather than quenched

and tempered 2¼Cr-1Mo. However, the presence of proeutectoid ferrite is generally regarded as unfavorable due to its lower strength and poorer toughness. Current alloy development reflects the need to avoid the formation of primary ferrite in the microstructure during cooling by making increased hardenability one of the principal criteria in alloy design.²⁰

Modification of 2¼Cr by the addition of vanadium results in the precipitation of finely dispersed, slow growing V_4C_3 upon tempering or postweld heat treatment.²¹ The precipitation of V_4C_3 occurs directly from the matrix and, unlike Mo_2C , is not dependent on the formation of other carbides. Vanadium has also been reported to be present in Mo_2C with a metal atom ratio of up to 0.3 in alloys that contain more than 0.1% V. Vanadium in Mo_2C results in stabilization of the Mo_2C .²² The combined effects of precipitation and stabilization result in vanadium being one of the most potent elements in promoting creep resistance.

Similar investigations to determine the carbide evolutionary sequence in other Cr-Mo steels upon tempering have been conducted by several investigators.^{23,24} However it is reasonable to believe that although the carbide evolutionary sequence in the CGHAZ may be similar to that in the base metal, the situation is complicated by the fact that some of the stable carbides (such as TiC , NbC , V_4C_3 , etc.) although generally finer than the chromium-, molybdenum- or iron-containing carbides, may not dissolve upon weld thermal experience. Lundin *et al.*^{110,140}

have found that in the 2¼Cr and 3Cr alloys that contain modifying (stabilizing) elements such as vanadium, titanium and niobium the carbides do not completely dissolve upon a coarse grain HAZ simulation thermal cycle. However, in standard composition alloys all the carbides dissolve upon CGHAZ simulation. Thus, the subsequent reprecipitation of carbides will be affected. The carbide evolution sequence in the CGHAZ of 1CrMoV steel is shown in Fig. A7.²⁵ It can be seen from Fig. A7 that the carbides present in the CGHAZ on tempering will depend on the postweld heat treatment temperature.

Lundin *et al.*^{110,140} have investigated the carbide evolutionary sequence in the CGHAZ of several standard and modified 2¼Cr-1Mo, 3Cr-1Mo and 3Cr-1½Mo alloys as a function of PWHT time at 1250°F (675°C). Their investigation revealed that in the alloys modified with vanadium, titanium and boron, the Mo_2C type carbides persist for longer times compared to the unmodified alloys.

Elevated Temperature Intergranular Cracking. Elevated temperature intergranular cracking, referred to as reheat, SRC/PWHT cracking, may occur in Cr-Mo steels containing less than 3 percent chromium.²⁷ However, reheat cracking has been occasionally observed in steels containing 3% and more chromium. Cracking is manifested by low rupture ductility and intergranular fracture along prior austenite grain boundaries, typically occurring in the coarse grained heat affected zone and occasionally in

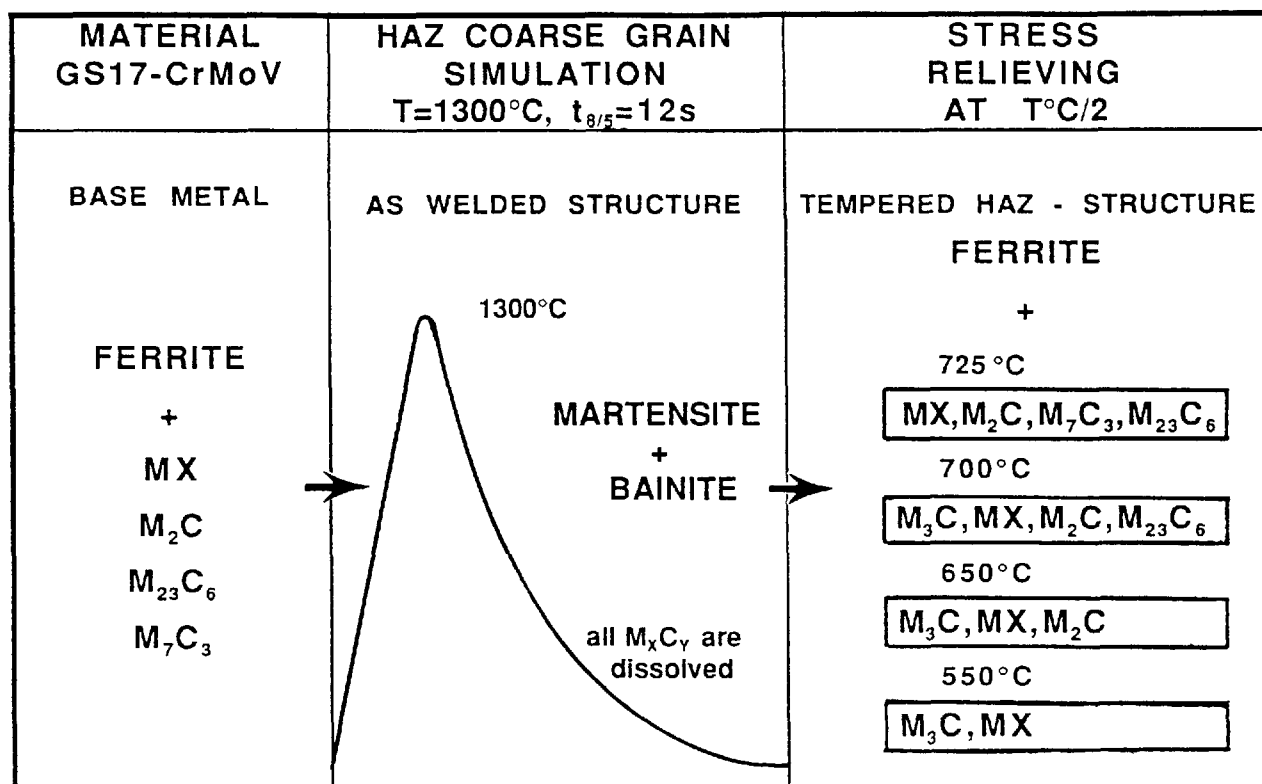


Fig. A7—Carbides in the coarse grained HAZ of 1CrMoV steel upon PWHT. Source: Buchmayr, B., Cerjak, H. and Fauland, H. P., "The Effect of the Precipitation Behavior on the HAZ-Properties of 1%Cr-Mo-V-Steel," 2nd International Conference on Trends in Welding Research, Gatlinburg, Tennessee, May 1989

the weld metal after an initially sound weldment has been subjected to PWHT or elevated temperature service.⁴ Heat-to-heat variations in cracking susceptibility indicate a dependence on residual elements that adds to the worrisome nature of this problem as bulk chemistry of an alloy may not be a reliable predictor of cracking susceptibility.⁵ If undetected this type of cracking may lead to failure of pressure vessels and piping in service.⁶

Large localized creep deformations may occur during PWHT to relax the residual stresses generated by the welding operation. Vinckier and Dhooge^{28,29} state that the extent of plastic strain required to relax residual stresses is directly proportional not only to the level of residual stresses but also to the size of the component. Notches and sharp transitions that are under high residual tensile stresses will act as severe strain raisers once stresses begin relaxing at elevated temperature and can easily cause fissuring in susceptible material.

Examination of unoxidized fracture surfaces under the SEM reveals primarily smooth grain boundary facets with no definable fracture characteristics. Only with high resolution have some portions of the grain boundary fracture surface of cracks been shown to consist of numerous small dimples surrounded by ductile tear ridges. According to Debiez and Granjon³⁰ brittle intergranular fracture is characteristic of low temperatures and high stresses and the ductile intergranular dimpled fracture occurs under low stresses at high temperatures. TEM, using carbon extraction replicas of fresh fracture surfaces, reveals an almost complete absence of grain boundary precipitation and only in a few instances small carbide particles are detected in the small dimples on the grain faces (microcavitation).

There is no longer any doubt that the reheat/stress rupture cracks follow the prior austenite grain boundaries in those regions in the HAZ that have been heated to temperatures well in excess of 2000°F (~1100°C) and that have subsequently undergone plastic deformation either during deposition of subsequent beads or during PWHT. In some cases the grain boundaries show only a row of small voids and in other cases the cracks are well developed, readily visible under the optical microscope and extending for several grain boundaries.

Briant and Bannerji³¹ in their review of the existing theories and mechanisms of intergranular failure in steel report that the circumstances under which steels exhibit intergranular fracture can be classified into four general categories:

1. the presence of certain secondary phases at the grain boundaries;
2. thermal treatments that cause impurity segregation to the grain boundaries without precipitation of an observable second phase;
3. the action of certain environments; and
4. combination of stress and high temperature.

It is well known that certain tramp elements at grain boundaries are a major cause of intergranular fracture. These elements are believed to lower the cohesive energy of the boundaries and at a given concentration, which depends on the yield strength, grain size and microstructure of the material, can change the fracture characteristics from cleavage to grain boundary separation. The most common embrittlors are from groups IV, V and VI in the periodic table presented below.

<i>Common Grain Boundary Embrittlors</i>		
<i>Group IVa</i>	<i>Group VA</i>	<i>Group VIA</i>
Si	P	S
Ge	As	Se
Sn	Sb	Te

Segregation of the alloying elements can occur in any of three ways:

1. equilibrium segregation during tempering;
2. equilibrium segregation during austenitization; and
3. carbide rejection during tempering.

Elements such as sulfur, phosphorus, nitrogen, boron, etc. are known to segregate to the prior austenite grain boundaries during austenitization with the extent of segregation decreasing with increasing austenitization temperature. Carbide precipitation along the grain boundaries is also a possible mean of increasing the impurity concentration. Certain impurities are more soluble in ferrite than in carbides and thus they build up at the ferrite-carbide interface.

Alloying elements or impurity elements can be placed in five broad categories:

1. promoters of segregation that act by co-segregating with the impurities such as manganese, nickel, etc.;
2. promoters that do not themselves segregate, such as chromium;
3. scavengers prohibiting segregation such as titanium, molybdenum, etc.;
4. embrittling elements such as hydrogen, nitrogen, silicon, phosphorus, sulfur, germanium, arsenic, selenium, tin, antimony, tellurium, bismuth, etc.;
5. improve grain boundary cohesion, such as carbon.

Although mechanisms of reheat/stress rupture cracking are not completely understood, it is now generally believed that several conditions must be fulfilled: first, a susceptible microstructure such as a coarse prior austenite grain size as readily occurred in the CGHAZ of welds; second, the presence of residual stresses; and third, discontinuities or notches that act as stress concentrators, for example weld configuration, slag/lack of fusion, cracks, etc. It has been clearly demonstrated that if a material with a CGHAZ is postweld heat treated so that no plastic strain

occurs during heat treatment, then it will show no loss in elevated temperature bend or tensile ductility. However, tests on simulated HAZ microstructures show that the impact transition temperature is raised even without plastic strain. PWHT and stress rupture cracks form because relaxation strains exceed the creep ductility of the CGHAZ at elevated temperature.

According to the existing theory of reheat cracking, carbides (vanadium, molybdenum, chromium, etc.) are taken into solution in the HAZ when the temperatures exceed 2200°F (1200°C). In addition, grain growth occurs in the HAZ. Due to rapid cooling from these high temperatures and the low temperatures at which transformation starts for such steels, the reprecipitation of carbides is almost completely suppressed and martensitic and lower bainitic transformation products are formed. PWHT or service cause the carbide forming elements to precipitate as carbides from the supersaturated solid solution in a manner similar to that which occurs in normal tempering. These precipitates form in the grain interior as submicroscopic platelets only a few Angstroms in diameter, causing considerable stiffening of the grain interior and, depending on temperature, produce an increase in hardness (secondary hardening). The grain boundaries, however, remain generally free of precipitates of sufficient size to prevent grain boundary sliding and it has been observed that areas adjacent to grain boundaries may be denuded (no precipitates).³² Any relaxation of residual stresses or plastic deformation imposed upon such a microstructure will be resisted by the stronger grain interiors and deformation will be concentrated along the weaker grain boundaries or in the narrow denuded zones causing grain sliding. Although the overall strains are small, local high shear and tensile strains develop at the grain boundaries, especially if the grain size of the material is large (less grain boundary area). The resultant significant deformations lead to the formation of voids at steps or other discontinuities on the grain boundary interfaces. Such cavities, when linked up, form the final grain boundary cracks.

Another theory holds that precipitates or inclusions on the grain boundaries nucleate voids or promote boundary decohesion when grain boundary sliding occurs. According to Kanazawa *et al.*,^{33,34} however, the stress relaxation characteristics and strength of the HAZ are more important for cracking susceptibility than the secondary hardening behavior. If the HAZ resists stress relaxation and if the fracture strength of the HAZ is low then the material will exhibit higher susceptibility to reheat cracking.

The type and morphology of precipitates occurring in the HAZ has been a topic of considerable research. Several investigators have contended that the most critical time at elevated temperature is during the formation of coherent or preprecipitate clusters at about 500–550°C (932–1022°F) which corresponds to the maximum in secondary hardening.^{35,40} Orr *et al.*³⁹

have noted that due to the strong lattice correspondence of Mo₂C with bainite or martensite, the nucleation energy is relatively low and therefore Mo₂C forms quickly as a finely divided slow growing precipitate. Swift *et al.*^{35–38} have proposed that coherent precipitates, Mo₂C or V₄C₃, nucleate in the matrix and at dislocation jogs and intersections. Coherent Mo₂C yields a nonuniform, highly strained matrix with decreased dislocation mobility and dislocation locking and results in a reduction of the ability of the grain interiors in the CGHAZ to plastically deform during PWHT or in-service.

Temperature is an important factor in that coherent Mo₂C persists for more than 500 hours at 1100°F (590°C) without decomposition to more complex carbides and has a correspondingly long coherency duration. However, at 1250°F (680°C) coherency is quickly lost as evidenced by the formation of incoherent Mo₂C within one-half hour at temperature.³⁸

A carbide denuded zone has also been reported to exist adjacent to the grain boundaries.^{22,41–44} These zones are apparently a result of the depletion of alloying elements due to carbide precipitation in the grain boundaries. Several researchers have considered the carbide denuded zones to be detrimental as they provide a narrow soft region in which strain can preferentially accumulate.^{21,44} However, Meitzner and Pense²⁸ found the presence of denuded zones did not contribute to stress relief cracking. Swift³⁸ found that although denuded zones were present they formed only at times beyond those corresponding to a minimum in ductility. In investigations conducted by Lundin *et al.*,^{110,140} no correlation could be determined between the denuded zone width and the SRC/PWHT cracking susceptibility in a variety of modified and standard 2¼Cr and 3Cr alloys.

Thus, from the above theory it can be assumed that susceptibility to reheat/stress rupture cracking increases with an increase in the grain size in the CGHAZ and higher initial hardness of the HAZ before PWHT/service. Also, a higher strain hardening rate in the base metal, which forces deformation into the HAZ, may promote cracking. Slow strain rates, which allow grain boundary sliding, also may promote cracking unless they are slow enough to permit softening of the microstructure during elevated temperature exposure.

Another factor is the presence of notches, either surface irregularities or internal defects, at which cracking almost invariably initiates.⁴⁵ The notches in the CGHAZ are particularly detrimental, as they act as intense stress concentrators in the very region of the weld already under a high tensile stress and with a low rupture ductility microstructure.^{28,41,42,46} The notch acuity further inhibits deformation by the creation of a triaxial stress state.³⁸

Another theory deals with the effect of trace elements or impurities. Some investigators have shown that high purity heats of the same bulk chemistry do not show a ductility loss for simulated CGHAZ micro-

structures. It is known, however, that various interactions among alloying elements and trace elements exist and can affect ductility either by a direct influence in the nonmetallic inclusion content/chemistry/shape or by embrittlement caused by a segregation of trace impurities to grain boundary interfaces at high temperature. Thus, interest in the heat-to-heat variability and the development of new techniques, particularly Auger spectroscopy, to study grain boundary composition led to intense evaluation of residual element effects on cracking.^{41,47} Segregation of residual elements (phosphorus, sulfur, tin and antimony) was found to play a critical role by causing embrittlement of grain boundaries.^{29,48,49} The effects of grain boundary embrittlement due to impurity segregation and impurity interactions during precipitation are superposed on the matrix strengthening effects by coherent precipitation.³⁵ Hipsley *et al.*⁵⁰ have noted that grain boundary segregation of impurities can be due to equilibrium segregation from the grain matrix or may be due to solute rejection from grain boundary carbides, the solubility of impurities being higher in the matrix material than in the carbides. Vinckier and Dhooze^{29,48} state that loss of ductility caused by segregation during elevated temperature is certainly a major factor in cracking but that the true mechanism is most likely a combination of these two ideas (segregation and precipitation strengthening) and is more complex.

Other researchers have proposed additional factors that increase the intricacy of the previous cracking mechanisms. Vinckier⁶ proposed that the decomposition of martensite needles during heating creates high localized strains at the grain boundary interfaces. These strains plus external restraint can result in the formation of microcavities, particularly when the matrix is strengthened by intragranular precipitation. Hipsley *et al.*⁵¹ have proposed that segregation of less mobile embrittling elements such as phosphorus, tin and antimony occurs at elevated temperatures in a fashion similar to that proposed by Troiano⁵² for hydrogen. Segregation is locally enhanced in regions of maximum triaxial stresses, such as at the root of a notch or in the region of a crack tip, by diffusion along the strain gradient. Dislocation pile ups at grain boundaries and grain boundary carbides plus the strain induced segregation of impurities reduce the cohesive strength of the grain boundaries and carbide-matrix interfaces sufficiently to allow the development of microcracks. Thus, according to Hipsley *et al.*⁵¹ the factors involved in cracking are not only segregation and precipitation but also the amount of plastic strain which, for a given load, increases as the yield strength falls with increasing temperature.

The necessary factors for cracking were summarized by Ito and Nakanishi:⁸

1. the material must have undergone a thermal cycle that results in solution of alloying ele-

ments and that retains the elements in solid solution after cooling;

2. grain growth must have occurred as a result of thermal cycling;
3. heat treatment between 850–1300°F (450–700°C) resulting in significant precipitation strengthening;
4. grain strength and internal stress must exceed the strength of the grain boundaries; and
5. a stress riser must be present to initiate cracking.

In addition, the material must be one that has a composition that is susceptible to cracking with regard to major alloy content and residual or impurity elements.

Origin of Residual Stresses

Residual stresses are developed in the weld HAZ and fusion zone during cooling due to restrained shrinkage and transformation volume changes as a result of austenite decomposition.^{53,54} On cooling, those areas of the base metal that experienced thermal expansion due to heating must contract or plastically flow. The bulk of the base metal that has experienced no significant heating (and therefore no decrease in strength) prevents or restrains the contraction of the cooling material. Above approximately 1200°F (650°C) the weld fusion zone and those regions immediately adjacent to the weld accommodate the thermal contraction by plastic deformation without developing any significant stress as the yield strength is low above this temperature. Cooling below 1200°F (650°C) results in significant increases in yield strength with decreasing temperature. Plastic deformation only occurs when the stresses due to thermal contraction exceed the yield stress and therefore cooling to the preheat temperature results in increasing residual tensile stress concomitant with the increased yield strength in the fusion zone and the HAZ. The resultant residual tensile stresses occurring in the HAZ and fusion zone are in equilibrium with compressive stresses in the bulk of the base material.

Transformational stresses are a result of the volumetric expansion that occurs during the decomposition of austenite. The material being transformed attempts to expand but expansion is hindered by the cooler material not undergoing transformation. Therefore the material being transformed experiences a compressive stress and the cooler material a tensile stress. If the transformation temperature is high the transformation stresses will be overridden by the effects of subsequent bulk shrinkage. However, if the transformation temperature is low, the transformation stresses will lower the overall tensile stress in the HAZ and fusion zone.

Superposition of the components of the residual stress developed during welding leads to an extremely complex final residual stress state. The CGHAZ, that

portion of the HAZ adjacent to the fusion zone, is under a significant tensile stress both in the direction of and perpendicular to the weld. This biaxial residual stress state as well as the unfavorable metallurgical characteristics of the CGHAZ exacerbates the susceptibility of this zone to reheat cracking.

Microstructural Effects

Meitzner and Pense⁴³ have found that martensite and lower bainite are more susceptible to cracking than upper bainite although the authors state that the difference is probably due to precipitation processes rather than optically resolved microstructure. Ito and Nakanishi⁸ also found that martensite and lower bainite microstructures are more susceptible than upper bainite. They related the increased susceptibility to the supersaturation of the bainitic and martensitic microstructures with alloying elements and carbon which results in intense secondary precipitation. Debiez and Granjon³⁰ found that bainitic structures appear to be more susceptible to cracking than martensitic structures upon implant testing. However, they do not state whether the bainitic microstructure obtained was upper or lower bainite.

Effect of Composition

The composition obviously plays a major role in susceptibility to cracking. Alloying elements added to structural and pressure vessel steels to increase tensile and creep strength form carbide or carbonitride precipitates in ferrite. Widely used alloy additions to steels are chromium, molybdenum and vanadium in high temperature steels. According to many authors the presence of these elements increases the susceptibility of a steel to reheat cracking. However, the restriction of these elements as low-alloy additions, cannot be considered in practice, as their presence is vital to the hardenability, strength and creep resistance of these steels.

In extensive literature reviews Meitzner⁴¹ and Dhooge *et al.*⁴⁸ defined the effects of specific alloying elements. The elements generally considered to be detrimental to cracking are: carbon, vanadium, molybdenum (individually and in concert with vanadium), niobium, aluminum, copper and the residual elements: phosphorus, sulfur, tin, antimony and arsenic. The effects of chromium, boron and titanium are not clearly defined. Nickel was found to be one element that appears to have no effect on cracking. In general, the elements found to be deleterious are either those that promote the formation of carbides of the M_2C or M_4C_3 type or those that are known to have general grain boundary embrittling effects.

Many investigators have tried to quantify the effects of alloying elements on the cracking susceptibility. Nakamura *et al.*⁵⁵ attempted to determine the effect of alloy additions on cracking susceptibility in Cr-Mo steels by development of a cracking susceptibility parameter (ΔG). Variations in the levels of chromium (0.1–1.5%), molybdenum (0.3–0.6%), nickel

(0–3%) and vanadium (0–0.08%) resulted in the ΔG parameter relationship:

$$\Delta G = Cr + 3.3Mo + 8.1V - 2$$

when ΔG is greater than zero the material is considered to be susceptible to cracking. In 1972, Ito and Nakanishi⁸ extended the work of Nakamura. The alloying elements (manganese (0.5–1.4%), nickel (0.5–1.5%), chromium (0.5–1.5%) and vanadium (0.05–0.12%)) were varied and additions of copper (0.15–0.26%), niobium (0.06%) and titanium (0.02–0.07%) were made to steels containing nominally 0.3% silicon and 0.5% molybdenum. This work resulted in the development of the cracking parameter, P_{SR} :

$$P_{SR} = Cr + Cu + 2Mo + 10V + 7Nb + 5Ti - 2$$

when P_{SR} is greater than zero the material is deemed to be susceptible to cracking. The applicable range of the P_{SR} parameter is limited to alloys that contain less than: 2% Mo, 1.5% Cr, 1% Cu and 0.15% V, Ti and Nb. However, Ito and Nakanishi found that chromium contents in excess of 2% eliminated cracking.

Subsequent to development of the P_{SR} and ΔG parameters several investigators have found poor correlation between the parameters and actual susceptibility of different alloys. Pense *et al.*⁵⁶ found that ΔG was a poor predictor of cracking in A 535-A, A 517-F and A 543 steels. McMahon *et al.*⁵⁷ found that both P_{SR} and ΔG did not accurately predict cracking susceptibility in multiple heats of SA 533-B and SA 508-2. Also, many investigators^{26,36–38,51,58–61,110,140} have found that 2¼Cr-1Mo alloys are susceptible to cracking although a chromium content of 2% or greater was considered to eliminate cracking susceptibility by Ito and Nakanishi.⁸

McMahon *et al.*⁵⁷ have suggested an additional parameter, CERL, with the addition of chromium:

$$CERL + Cr = Cr + 0.2Cu + 0.44S \\ + P + 1.8As + 1.9Sn + 2.7Sb.$$

The greater the CERL + Cr value the greater the cracking susceptibility. This parameter clearly emphasizes the effects of embrittling elements over the effect of carbide formers. The authors state that individual alloy content will affect the parameters and that increasing the former carbide content may necessitate their inclusion in a manner similar to that of the P_{SR} parameter. Similarly, in reviewing results of cracking susceptibility tests of 2¼Cr-1Mo weld metal, Boniszewski⁶² recommended use of the metal composition factor (MCF) to rank cracking susceptibility. The metal composition factor:

$$MCF = Si + 2Cu + 2P + 10As + 15Sn + 20Sb$$

combines the relative overall potency of grain boundary embrittling elements present in a material. An increase in the MCF was found to correlate with a decrease in rupture ductility as measured by elongation in hot tensile tests.

Since the ΔG and P_{SR} parameters were published, several similar equations evolved for other steel compositions. T factor was developed for SA 508 C12 and SA 533 Gr B steels:

$$T = 20V + 7C + 4Mo + Cr + 1.5 \log X - 0.5Mn$$

where

$$X = Al \text{ for } Al \leq 2N$$

or

$$X = 2N \text{ for } Al > 2N.$$

Steels are susceptible if $T > 0.9$.

$\Delta G1$ parameter was formulated to consider the effect of carbon in the original ΔG parameter:

$$\Delta G1 = Cr + 3.3Mo + 8.1V + 10C - 2.$$

When the $\Delta G1$ parameter is greater than 2 the material is considered susceptible to cracking.

Considering the effect of tramp elements alone R value was developed for 0.5CrMoV steels:

$$R = P + 2.43As + 3.57Sn + 8.16Sb.$$

Susceptibility to PWHT cracking increases with increase in R value.

Bruscato⁶³ devised an embrittling factor relating weight percent of impurity elements (in ppm), based on 50% shear fracture appearance transition temperature:

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

where susceptibility increases with $\bar{X}$ -values.

In the subsequent part of this section an attempt has been made to review the effect of some alloying elements individually and in combination with other elements.

Effect of Chromium and Molybdenum. In the ΔG and P_{SR} crack susceptibility parameters chromium is an element that increases the susceptibility to SRC/PWHT cracking. However, it has been pointed out by both Nakamura and Ito that steels that contain greater than 1.5% chromium are not susceptible to cracking. Published literature reveals that this is not true and that steels containing up to 3% chromium have been found susceptible to SRC/PWHT cracking. However, this may be because of the effect of other elements such as vanadium, niobium and titanium in the steels tested. Some results show that chromium between 0–2% decreases high temperature ductility and when above 2% increases it markedly.

Molybdenum increases the susceptibility to SRC/PWHT cracking and its effect is greater than that of chromium. In the early stages of tempering the Mo_2C type carbides precipitate and cause hardening of the grain interiors. Other carbide forming elements (such as vanadium, titanium, niobium) that have more affinity for carbon than molybdenum and tend to

form more stable carbides. Even in such cases, molybdenum is a potent solid solution hardening element.

The recent work of Tamaki^{27,64–68} attempts to determine the separate effects of chromium and molybdenum on SRC/PWHT cracking. His papers represent extensive work on materials of varying chromium (0–5%) and molybdenum (0.3–1.5%) content. The effects of chromium and molybdenum, independently and in concert, were studied using a modified implant test. The modified implant test was employed to determine the minimum stress that would cause a specimen to fracture within 20 hours while postweld heat treating the specimen at a temperature of 1112°F (600°C). The susceptibility to cracking was related to the magnitude of the critical stress to rupture ($\sigma_{AW-crit}$). The lower the minimum critical stress to cause rupture, the greater is the susceptibility to cracking.

Alloys containing chromium in the range tested but low in molybdenum (0.25%) were found to be susceptible to cracking. The 1%Cr-0.25%Mo alloy showed the lowest critical stress for failure and therefore the greatest susceptibility to cracking. Increasing molybdenum at any level of chromium increased susceptibility, but the lowest critical stress for any particular molybdenum level occurred for alloys containing 1% chromium. The data from Tamaki's studies are shown in Fig. A8. Note that for low molybdenum content (0.25% and 0.5%) when no chromium is present no cracking occurred over the range of stresses employed indicating that these materials are not susceptible.

When the results are expressed as a function of chromium and molybdenum for different stress levels the evaluation shown in Fig. A9 results. Susceptibility to SRC/PWHT cracking with a change in alloying element content is a maximum on these diagrams where the stress contours are closest together. Susceptibility for a particular alloy may be judged by the magnitude of the critical stress to failure ($\sigma_{AW-crit}$). The plotted data is divided into four regions labelled I, IIa, IIb and III (Fig. A9). The materials in region I, those with less than 1% Cr and less than 0.5% Mo, are relatively insensitive to cracking. Materials in region IIa, comprised of alloys with 0–1% Cr and 0.5–1% Mo, have rapidly increasing sensitivity to cracking with increasing chromium or molybdenum content based on relatively large decreases in the critical stress with small changes in alloy content. Region IIb, comprised of alloys with greater than 2% Cr and 0.5–1% Mo, characterizes behavior of decreasing sensitivity with increasing chromium content. Region III, comprised of alloys with approximately 1% Cr and greater than 1% Mo, represents the highest sensitivity to cracking. The Nakamura parameter⁵⁵ for PWHT cracking susceptibility, ($\Delta G > 0$), was found to predict cracking principally in fields IIa and III and extending somewhat into field IIb. The cracking parameter due to Ito⁸ ($P_{SR} > 0$) predicted cracking principally in field IIa. These parameters are limited to chromium contents less than 1.5%. Since fields IIa and III indicate

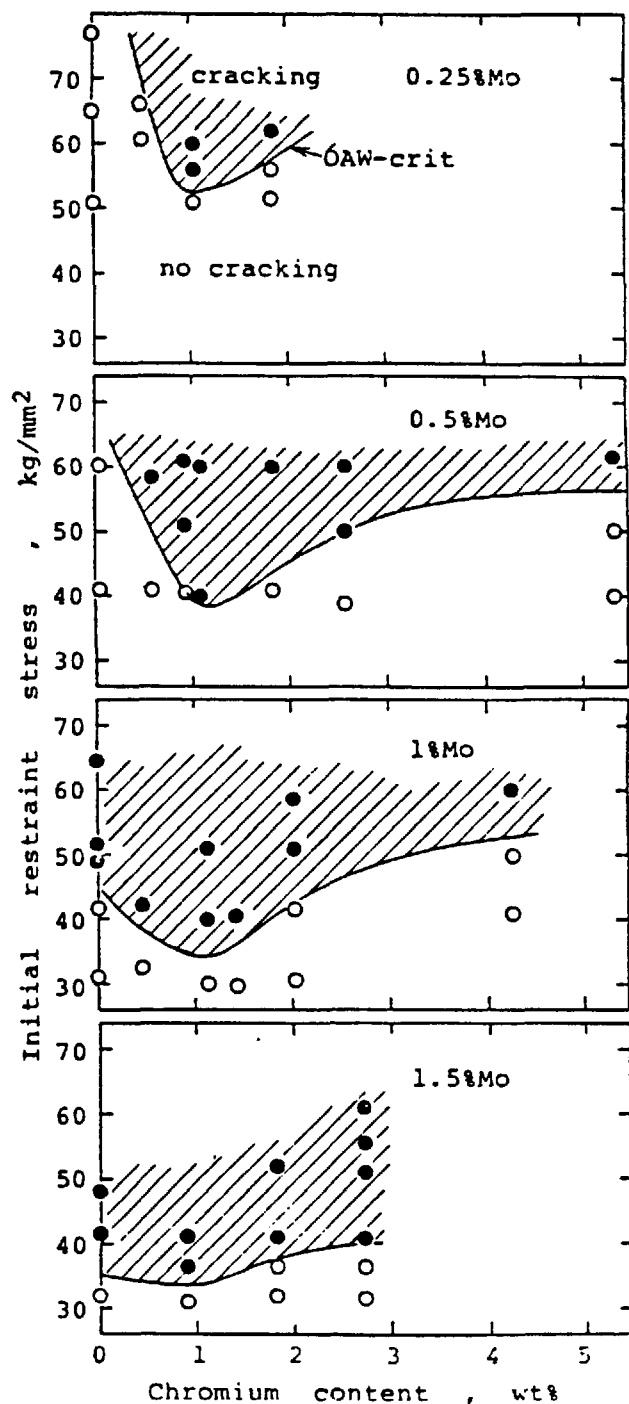


Fig. A8—Effect of chromium on the critical restraint stress, $\sigma_{AW-crit}$. Source: Tamaki, K. and Suzuki, J. "Effect of Chromium and Molybdenum on Reheat Cracking Sensitivity of Steels," Transactions on the Japan Welding Society, Vol. 14(2), October 1983, pp. 39-43

the alloys of maximum reheat cracking sensitivity, the agreement is excellent, with Tamaki's diagrams being more discriminating than either index (Fig. A10).

In order to discern the microstructural causes for the differences in cracking susceptibility for various materials, Tamaki⁶⁴ undertook an extensive study of the carbides in the alloys using X-ray diffraction techniques (extracted carbides) and transmission electron microscopy of carbide extraction replicas. It was found that the materials most susceptible to PWHT cracking showed the greatest fraction of M_2C type

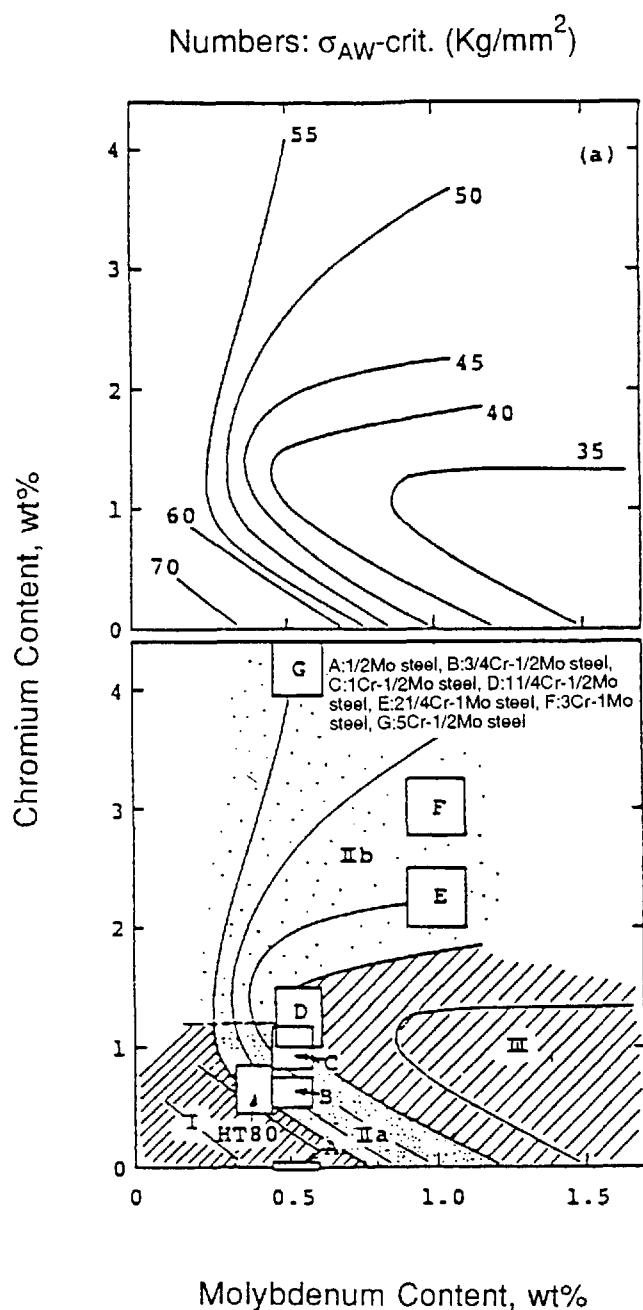


Fig. A9—Contour lines of critical restraint stress ($\sigma_{AW-crit}$) shown on the Cr-Mo content diagram. Source: Tamaki, K. and Suzuki, J. "Effect of Chromium and Molybdenum on Reheat Cracking Sensitivity of Steels," Transactions on the Japan Welding Society, Vol. 14(2), October 1983, pp. 39-43

carbides after PWHT. With a smaller amount of M_2C (or a larger amount of M_7C_3 or $M_{23}C_6$), the susceptibility to cracking decreased. Figs. A11 and A12 depict these results graphically. Fig. A11 shows the relative amount of carbides present for different alloys on a chromium vs. molybdenum content diagram with the curved lines being constant weight percent M_2C . Fig. A12 superimposes these constant weight percent M_2C lines (solid lines, Fig. A12) on the chromium vs. molybdenum content diagram of Fig. A9. Figs. A11 and A12 thus show that the greatest susceptibility to cracking coincides with the largest fraction of M_2C . The only exception noted is confined to below the a-a' line in Fig. A12 where, according to Tamaki, phospho-

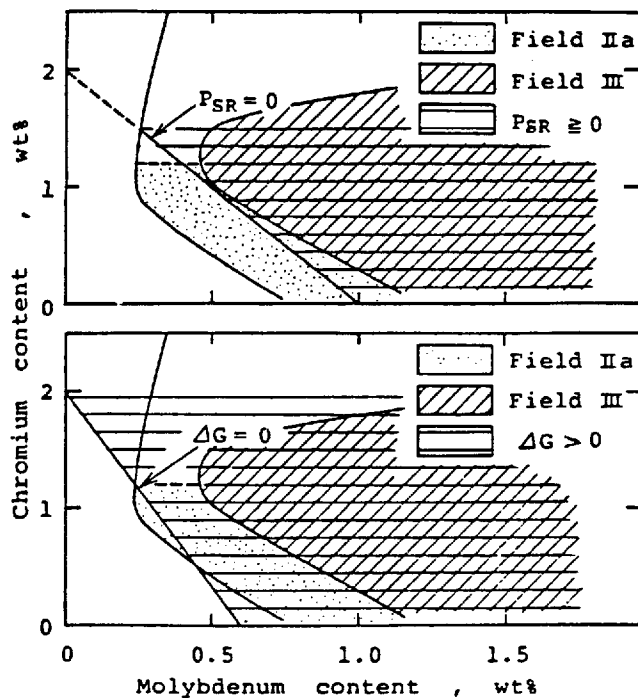


Fig. A10—Comparison between the critical restraint stress ($\sigma_{AW-crit}$) and cracking sensitivity indexes, P_{SR} and ΔG . Source: Tamaki, K. and Suzuki, J., "Effect of Chromium and Molybdenum on Reheat Cracking Sensitivity of Steels," Transactions on the Japan Welding Society, Vol. 14(2), October 1983, pp. 39–43

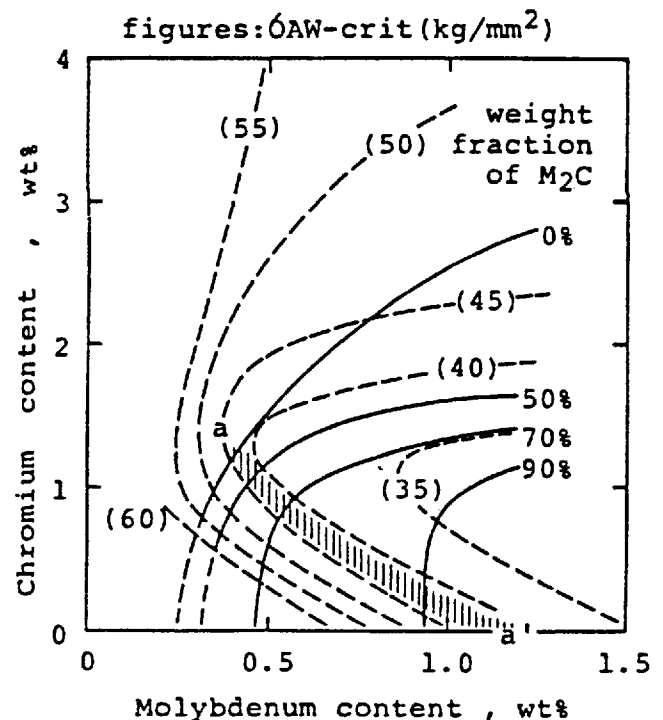


Fig. A12—Weight fraction of M_2C shown on the $\sigma_{AW-crit}$ diagram. Source: Tamaki, K., Suzuki, J., Nakaseko, Y. and Tajiri, M., "Effect of Carbides on Reheat Cracking Sensitivity," Transactions on the Japan Welding Society, Vol. 15(1), April 1984

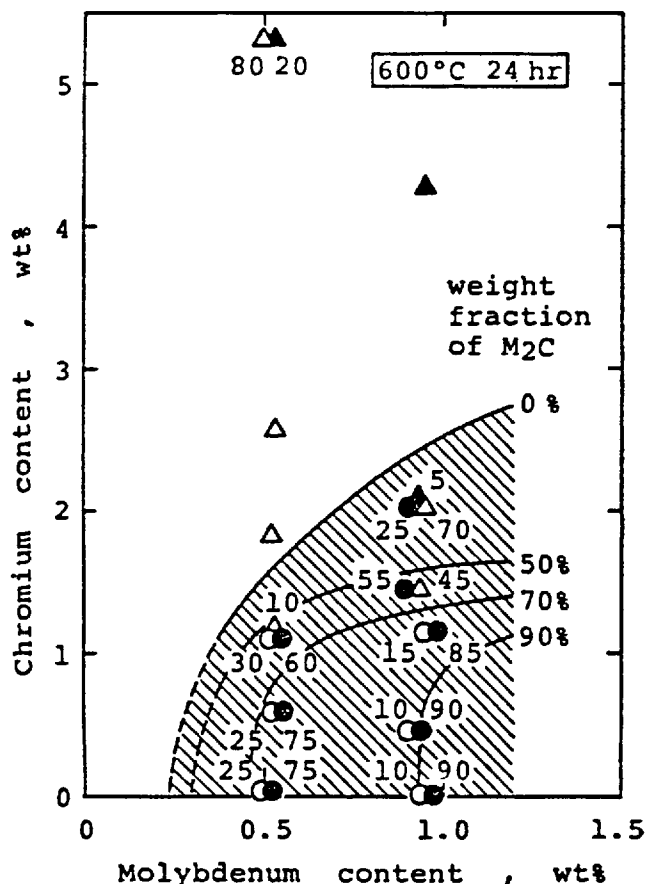


Fig. A11—Carbides present in Cr-Mo steels tempered at 600°C for 24 hours. ○ M_3C ● M_2C △ M_7C_3 ▲ $M_{23}C_6$. Source: Tamaki, K., Suzuki, J., Nakaseko, Y. and Tajiri, M., "Effect of Carbides on Reheat Cracking Sensitivity," Transactions on the Japan Welding Society, Vol. 15(1), April 1984

rus segregation is inhibited in alloys with less than 1% Cr and less than 1% Mo.

Since both M_2C and M_7C_3 strengthen the matrix by precipitation, Tamaki investigated the effects of secondary strengthening and high temperature hardness by making hardness measurements on samples held at temperature for one hour both at the holding temperature and at room temperature after cooling. While secondary hardening is manifested by an increase in hardness at room temperature, Tamaki found that at high temperature the phenomenon is represented only in a delay in softening. That is, softening continues to occur but at a lower rate than that which occurs at lower temperatures. Fig. A13 illustrates these results with the room temperature hardness shown as filled circles and the elevated temperature hardness as open circles. This is in accordance with the findings of Bauford.⁶⁹ In hot tensile tests Bauford found that at temperature there is no increase in strength with time but yield strength remains constant over a long period followed by a slow loss in strength.

It was found that in alloys in which the principal precipitate is M_7C_3 , the delay in softening occurred at lower temperatures than for alloys in which the precipitate consists of large fractions of M_2C . Tamaki⁶⁷ postulated that the grain boundary embrittlement would be of a similar nature in either type of alloy and therefore embrittlement of the grain boundaries would initiate at the same temperature and proceed in a similar fashion for both types of alloys. As shown schematically in Fig. A14, if a delay in

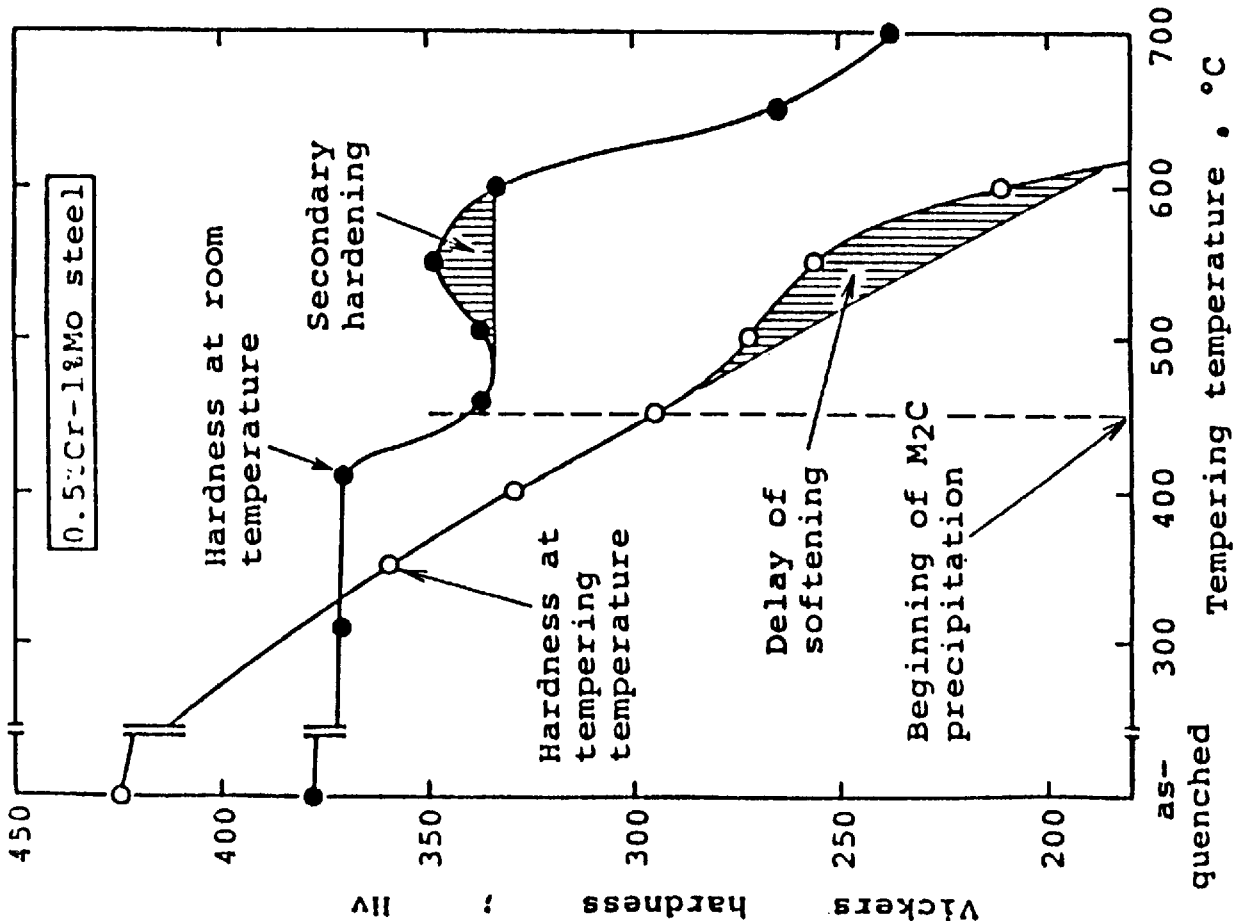


Fig. A13—Hot hardness and hardness at room temperature versus tempering temperature. Source: Tamaki, K., Suzuki, J., Nakaseko, Y. and Tajiri, M., "Effect of Carbides on Reheat Cracking Sensitivity," Transactions on the Japan Welding Society, Vol. 15(1), April 1984

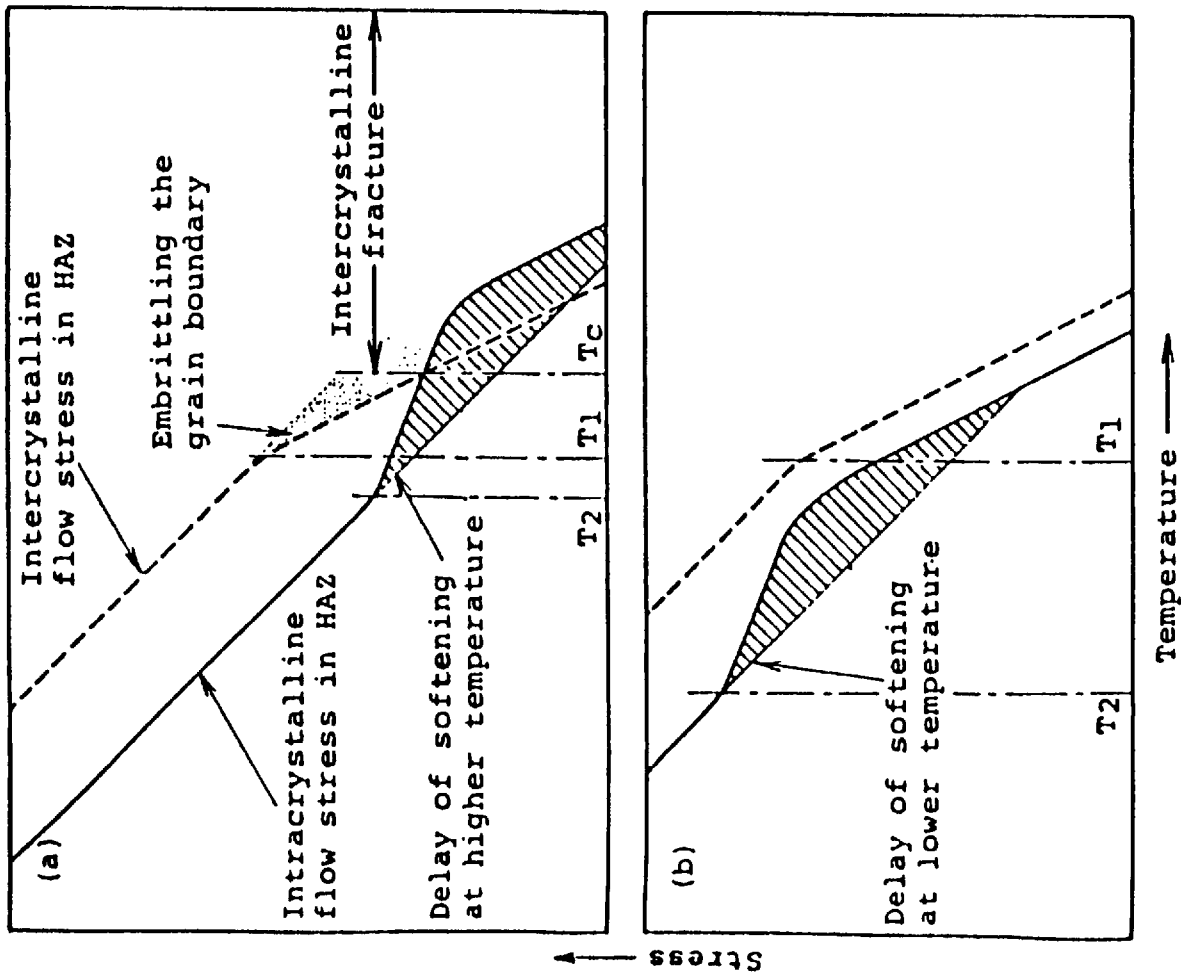


Fig. A14—Schematic representation of Tamaki's mechanism of reheat cracking. Source: Tamaki, K., Suzuki, J., Nakaseko, Y. and Tajiri, M., "Effect of Carbides on Reheat Cracking Sensitivity," Transactions on the Japan Welding Society, Vol. 15(1), April 1984

softening occurs at high temperatures, as in M_2C type precipitation, the embrittlement of the grain boundary may cause the intercrystalline flow stress to be less than the intracrystalline flow stress resulting in fracture. However, as shown in Fig. A14B, if a delay in softening occurs at lower temperatures as with M_7C_3 precipitation, the intercrystalline flow stress exceeds the intracrystalline flow stress at all temperatures and thus intercrystalline fracture does not occur.

Effect of Vanadium. In Cr-Mo alloys vanadium additions dramatically increase elevated temperature strength. Unfortunately, the addition of vanadium to Cr-Mo steels has been found almost universally to result in an equally dramatic increase in SRC/PWHT cracking susceptibility.^{6,8,26,30,38,40,42,43,46,70} The importance of vanadium in increasing cracking susceptibility can be seen in the P_{SR} and ΔG parameters in that the multipliers for vanadium are the highest.^{8,55}

The addition of vanadium results in a uniform and fine precipitation of V_4C_3 in the matrix, resulting in significant grain matrix strengthening and accumulation of strain in the grain boundaries.⁴⁶ Bently⁴⁰ has noted that early in the heat treatment cycle, intense V_4C_3 precipitation occurs at the ferrite-bainite interfaces due to segregation effects, the bainite having a higher carbon content and the ferrite having a higher vanadium content. At temperatures between 930–1020°F (500–550°C) coherent precipitation of V_4C_3 occurs in the ferrite similar to Mo_2C formation and is concurrent with the development of maximum hardness and strength. At higher temperatures (1300°F (700°C)) and longer time periods (10 hours) carbide precipitation occurs in the grain boundaries and large carbides and a grain boundary denuded zone are formed.

In a study of $\frac{1}{2}Cr-\frac{1}{2}Mo$ with varying vanadium

content, Meyers⁷¹ determined that vanadium below 0.22–0.27% did not appreciably increase cracking susceptibility. It was speculated that vanadium could be increased if residual elements were restricted in order to limit grain boundary embrittlement. However, Meyers noted that attention must also be paid to the effects of chromium, manganese and nickel, which increase initial hardness, and to the effect of molybdenum, which increases secondary hardening. Restrictions may have to be placed on them as well as vanadium.

Jones⁷² noted that the vanadium-to-carbon ratio must be considered. In a study of welds in $1Cr-\frac{1}{2}Mo-1\frac{1}{2}V$ materials with vanadium-to-carbon ratios between 3.5–4.5, a high susceptibility to cracking was found. Stone and Murray²¹ noted that a minimum in creep ductility was apparent at vanadium-to-carbon ratios of 3 to 4 and the reduction of this ratio to 1.5 markedly increased ductility. Thus, a vanadium-to-carbon ratio of 1.5 to 2 was recommended to mitigate cracking.

Tamaki *et al.*⁶⁸ found that small additions of vanadium (0.06%) reduced the critical stress to fracture in the implant test as shown in Fig. A15. The maximum effect was found to occur in low chromium and low-to-high molybdenum alloys. The increase in cracking susceptibility was said to be related to a decrease in the rate of stress relaxation in a similar fashion to that experienced in the Cr-Mo alloys previously studied. The decrease in the rate of stress relaxation due to a vanadium addition has been suggested to be brought about by the precipitation of vanadium carbides in addition to molybdenum carbides. Ito and Nakanishi⁸ found that cracking in the Y-groove restraint test increased from 0–95% as vanadium was increased from 0–0.08%. They also showed that the

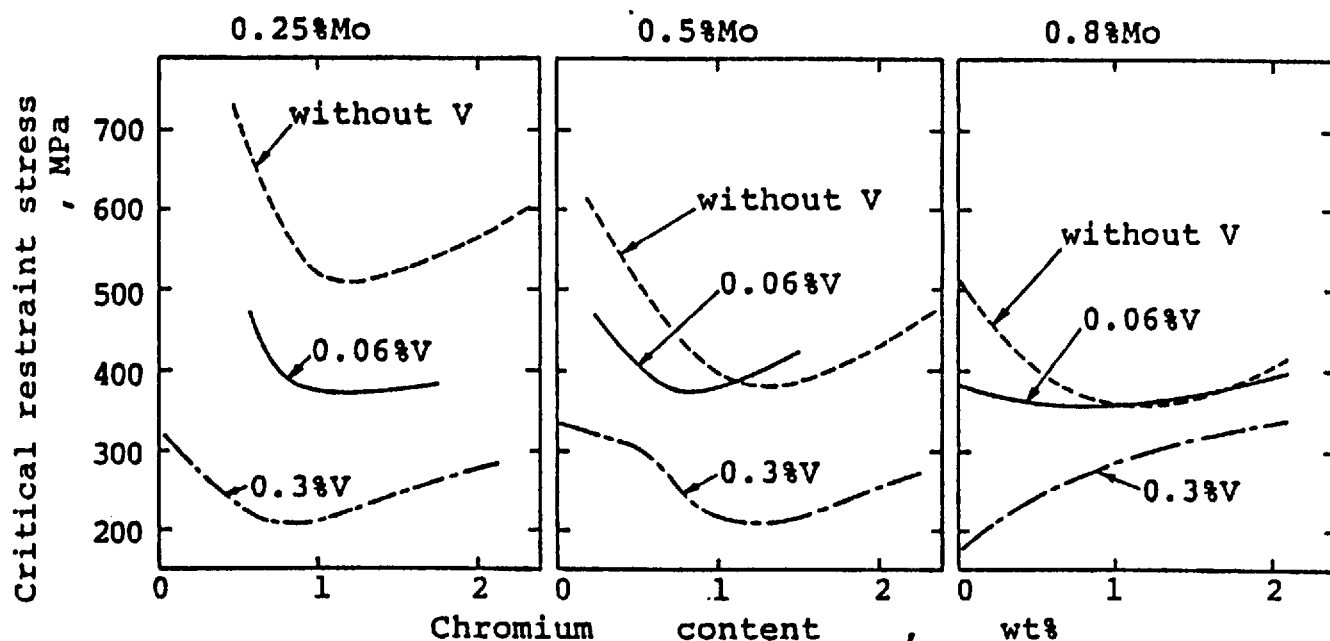


Fig. A15—Effect of vanadium additions on the critical restraint stress, $\sigma_{AW-crit}$ of Cr-Mo steels. Source: Tamaki, K., Suzuki, J. and Kojima, M., "Combined Influence of Chromium, Molybdenum and Vanadium on Reheat Cracking of Steels," IIW Document IX-1518-88

base composition was important, in that no cracking was observed for several steels with up to 0.14% vanadium.

Fractographic examination of PWHT cracks in CrMoV steel (0.37% Cr, 0.56% Mo, 0.29% V) conducted by Boniszewski and Eaton³² revealed that the grain boundary cavities were faceted and contained V_4C_3 type particles 200–600Å in diameter. This suggests a heterogeneous cavity nucleation mechanism. Nonmetallic inclusions were not found to be associated with cavity nucleation.

Effect of Nickel. Nickel is one element that appears to have the least detrimental effect on the Cr-Mo alloys with respect to SRC/PWHT cracking susceptibility. In establishment of the ΔG parameter nickel was varied between 0–3.4% without increasing susceptibility to cracking.⁵⁵ The same is true of the P_{SR} parameter where nickel was varied between 0–1.5%.⁸ Thus nickel does not appear in either cracking susceptibility parameter. Tamaki and Suzuki⁶⁵ found that a nickel content below 1.5% had no effect on the critical stress to fracture but that above 1.5% nickel content greatly increased cracking susceptibility.

The effect of nickel is probably related to the fact that nickel alters the kinetics of carbide precipitation. In a study of the tempering characteristics of 3Cr-1½Mo-Ni alloys, Ritchie *et al.*⁷³ found that $M_{23}C_6$ replaces M_2C , M_3C and M_7C_3 after only one half hour at 1290°F (700°C) as compared to 400 hours for 2¼Cr-1Mo. In accordance with Tamaki's model,⁶⁷ carbides that cause the greatest strengthening of the matrix would form while the overall strength of the grain boundaries is high and before significant embrittlement could occur. The rapid formation of carbides at lower temperatures and the probable very short times of coherency would result in the lowering of SRC/PWHT cracking susceptibility.

Effect of Manganese and Silicon. The effect of manganese on the susceptibility to SRC/PWHT cracking is controversial. A high manganese-to-silicon ratio has been observed to decrease susceptibility of 2¼Cr-1Mo SA weld metal to PWHT cracking.⁷⁴ When the Mn:Si ratio was increased from 2.5 to 4, accelerated stress rupture tests revealed a fourfold increase in the elongation and a twofold increase in the reduction in area. Mullery and Cadman⁴⁶ found lower cracking susceptibility with increased manganese in CrMoV steels. However, according to Bodnar *et al.*⁷⁵ and Weng and McMahon⁷⁶ manganese cosegregates with phosphorus to prior austenite grain boundaries and thus manganese is also a potent embrittling element reducing intergranular fracture strength.

In nickel-bearing steels, silicon behaves as an embrittling element and cosegregates with nickel to prior austenite grain boundaries. In low nickel or nickel-free steels, silicon promotes segregation of phosphorus to austenite boundaries. Boniszewski⁶² states the effect of silicon reduction is an increase in ductility and a lower cracking susceptibility. Vinckier⁷⁴

found that silicon has been reported to enhance Mo_2C formation in Cr-Mo materials.^{70,77} Ratliff and Brown⁷⁸ state that increased silicon appears to enhance cementite dissolution and thereby precipitation of alloy carbides containing chromium, molybdenum and vanadium resulting in increased secondary strengthening which ultimately impairs stress rupture ductility. According to Yu⁷⁹ silicon appears to accelerate the precipitation of Mo_2C type carbides in 2¼Cr-1Mo steel.

Effect of Titanium. Titanium in steel is used as an alloying element in addition to its role as a deoxidizer and grain refiner. Research on the effects of titanium on SRC/PWHT cracking susceptibility have had mixed results. When titanium has been added for deoxidation purposes (Meyers,⁸⁰ 0.016–0.055% and Harris and Jones,⁸¹ 0.021–0.030%) cracking susceptibility was reported to diminish in comparison to the same material deoxidized with other elements, particularly aluminum. Spaeder and Plodder⁸² found that addition of 0.048% titanium to T-1 steel did not increase susceptibility while Ito and Nakanishi⁸ found that additions of 0.03–0.07% titanium in 1Cr-½Mo alloys slightly increased susceptibility. Tamaki and Suzuki⁶⁴ found that the addition of a small amount of titanium, 0.07%, increased the critical stress to fracture in low chromium high molybdenum alloys. Harris⁸³ pointed out that, while small amounts of titanium may appear beneficial, larger amounts as deliberate alloying additions may increase cracking susceptibility due to matrix strengthening.

Effect of Niobium. The effect of niobium has not been investigated extensively in the constructional and ferritic creep resisting alloys, probably because it is not commonly used in many of the steels. However, the work of Ito and Nakanishi⁸ shows that niobium in the range of 0.06–0.10% can have a strong detrimental effect in 1Cr-½Mo alloys. From a study to determine the effect of niobium and vanadium in 2¼Cr-1Mo steel^{70,84} it was found that a steel with niobium (0.013%) and vanadium (0.005%) resisted PWHT cracking whereas another heat (Nb, 0.001%, V, 0.108%) exhibited cracking susceptibility in the Y-groove test.

Effect of Aluminum, Nitrogen and Boron. The presence of aluminum, used as a deoxidizer and grain refiner in Cr-Mo steels, has been found to enhance SRC/PWHT cracking susceptibility.^{49,81} Meyers⁸⁰ found that increasing amounts of aluminum, up to 0.035%, used for deoxidation progressively lowered resistance to cracking in a ½Cr-½Mo-¼V alloy. Aluminum was thought to prevent grain boundary mobility by the presence of aluminum-nitrogen precipitate clusters rendering the grain boundaries unable to recrystallize local regions of strain induced shear. Ratliff and Brown⁷⁸ found that when aluminum content exceeded 0.010% in a 1Cr-½Mo-¼V alloy a marked degradation in stress rupture properties occurred. Viswanathan and Beck⁸⁵ have suggested that

presence of aluminum in solid solution in CrMoV steel reduces the rupture ductility due to increased precipitation of vanadium carbides. Aluminum was proposed to reduce the stability of cementite thereby enhancing the formation of Mo_2C and V_4C_3 . Edwards *et al.*⁴⁴ found segregation of aluminum, silicon and nitrogen occurred at the grain boundaries by Auger spectroscopy. They speculated that if the combined aluminum and silicon concentration exceeded the austenite solubility, formation of ferrite at the grain boundaries could result and cause the rejection of carbon and the creation of a soft, carbide denuded region.

Several researchers have found that the combination of boron and aluminum was particularly detrimental in promoting cracking.^{5,44,49} Presser and McPherson^{49,86} speculated that aluminum scavenges nitrogen, allowing boron to remain in solution and segregate to the grain boundaries. Boron, when segregated to the grain boundaries, forms $\text{M}_{23}(\text{C},\text{B})_6$ when held between 1100–1650°F (600–900°C) during cooling and upon reheating above 1200°F (659°C) results in de-embrittlement. Presser and McPherson^{49,86} also speculated that other strong nitride formers such as titanium or zirconium could enhance embrittlement due to boron. Boron alone has been reported to cause a slight increase in cracking susceptibility.^{15,43,49,86} However, Paju and Moller⁸⁷ suggest that boron reduces grain boundary embrittlement induced by phosphorus segregation during austenitization. This is due to boron occupying the grain boundary sites in preference to phosphorus.

Effect of Copper. Copper has been found to increase the HAZ SRC/PWHT cracking susceptibility of steels. Ito and Nakanishi⁸ noted increased cracking as copper was increased to about 1% and Harris and Jones⁸¹ report that higher-than-usual Cu + Sn trace levels significantly affect cracking in steels that represent a variety of deoxidation practices. Copper was also reported to increase susceptibility in $\frac{1}{2}\text{Cr}-\frac{1}{2}\text{Mo}-\frac{1}{4}\text{V}$ alloy.^{57,62} Konosu and Maeda¹²⁶ have found an increase in the susceptibility to creep embrittlement in the HAZ at 1022°F (550°C) when the copper content was raised to 0.2% in $\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}$ steel.

Copper also has a deleterious effect in promoting weld metal SRC/PWHT cracking. Hunter^{88,89} and Wolstenholme⁹⁰ found a direct correlation between weld metal copper content in $\frac{2}{4}\text{Cr}-1\text{Mo}$ and cracking susceptibility. This susceptibility increased with increasing copper content. However, at a given copper content, weld metal deposited with a basic agglomerated flux was not as extensively embrittled as that deposited under less basic fused flux.

Effect of Carbon. Since carbide formation is involved in all elevated temperature phenomena, carbon obviously plays an important role in SRC/PWHT cracking. Few researchers seem to have paid any significant attention to the influence of carbon and thus carbon has not appeared in the crack susceptibility parameters. However, remember that carbon in

solid solution is known to increase intergranular cohesion.⁷⁶ Dolet *et al.*⁷⁴ have determined that increasing carbon decreases ductility at 1112°F (600°C) for a steel with the same ΔG parameter. In an investigation of Cr-Mo steels with two vanadium levels, Ito and Nakanishi⁸ found that cracking increased markedly as carbon was increased from 0.05–0.10% but was not changed by further increase to 0.25%.

Auger analysis of grain boundaries in an Fe-Ti-P alloy has shown that both titanium and phosphorus segregate to the grain boundaries.⁹⁴ An addition of 0.025% carbon in an Fe-0.26Ti-0.04P alloy showed 2.5 times more phosphorus at the grain boundaries than the alloy without carbon. Tamaki and Suzuki⁹⁵ have also suggested an increase in phosphorus segregation with an increase in carbon content.

Effect of Tramp Elements. Tramp elements have also been reported to affect SRC/PWHT cracking susceptibility. According to Drinnan *et al.* an increase in total percentage of trace elements (P + Cu + Sn + Sb + As) times a factor (considering the grain size) from 0.5–1.3 resulted in an increase in the percentage of cracking from 0–100%.

Tin was reported to increase susceptibility in $\frac{1}{2}\text{Cr}-\frac{1}{2}\text{Mo}-\frac{1}{4}\text{V}$.⁸¹ Townsend⁸³ reported that the addition of tin to $\frac{1}{2}\text{Cr}-\frac{1}{2}\text{Mo}-\frac{1}{4}\text{V}$ significantly lowered ductility in weld-simulated specimens. However, Meyers⁸⁰ determined that tin additions decreased susceptibility in $\frac{1}{2}\text{Cr}-\frac{1}{2}\text{Mo}-\frac{1}{4}\text{V}$ while both copper and antimony additions increased cracking susceptibility.

Among the residual elements, phosphorus is the most potent in embrittling the grain boundaries and therefore in enhancing SRC/PWHT cracking susceptibility. Tamaki *et al.*⁹⁵ have found that the solubility of phosphorus in ferrite and austenite decreases markedly upon addition of chromium and molybdenum. Thus, chromium and molybdenum can enhance the segregation of phosphorus to grain boundaries and can even cause phosphide precipitation at grain boundaries. Auger analysis of fracture surfaces has provided significant evidence of phosphorus segregation in many studies.^{51,58,76,77,96–99} Phosphorus segregation has been reported to be affected by carbide formation and element interaction,^{51,77,99–102} microstructure,^{100,102} and grain size.¹⁰⁰ Wilkinson *et al.*¹⁰³ have suggested that phosphorus, when segregated to prior austenite grain boundaries, enhances nucleation of cavities.

Phosphorus has been reported by Yu and McMahon,^{77,99} Hipsley *et al.*,^{51,101} Wada *et al.*¹⁰⁴ and Eyre *et al.*¹⁰⁰ to interact with molybdenum in solution in the matrix. Molybdenum in solution scavenges or ties up phosphorus in the form of Mo-P clusters or compounds preventing the segregation of phosphorus to the grain boundaries. Precipitation of Mo_2C during heat treatment releases phosphorus allowing segregation to the grain boundaries to occur. It has been suggested that the change in phosphorus concentration in ferrite is closely related to the formation of M_2C carbides.

Silicon enhances the formation of Mo_2C and has been reported to have a repulsive interaction with phosphorus similar to one that occurs with carbon. Silicon content was also shown by Auger analysis to be lower at the grain boundaries than within the grains.^{77,99} Thus, silicon would appear to enhance grain boundary embrittlement by phosphorus.

Precipitation of alloy carbides has also been reported to decrease the activity of carbon in solution thus allowing greater phosphorus segregation.^{98,100} The precipitation of molybdenum-rich carbides increases the relative amount of chromium in solution and appears to enhance phosphorus segregation through Cr-P cosegregation.^{77,99,102} Also, the formation of carbides in the grain boundaries results in the rejection of phosphorus to the grain boundaries as phosphorus solubility in the carbides is less than the solubility in the matrix.^{51,100,101}

Viswanathan and Joshi¹⁰⁵ and Eyre *et al.*¹⁰⁰ have found that phosphorus segregation is greater in martensitic microstructures than in bainitic microstructures. Viswanathan and Joshi¹⁰⁵ also reported that segregation increased with increasing hardness for a given microstructure in a $1\frac{1}{4}\text{Cr}-\frac{1}{2}\text{Mo}-0.3\text{V}$ material. Eyre *et al.*¹⁰⁰ demonstrated that grain size also affected embrittlement in that increased grain size decreases the area available for segregation.

Tamaki also undertook a study of the effect of phosphorus in Cr-Mo alloys.^{64,66,95,106} Several notable results were obtained from Tamaki's study.

1. Plots similar to those of Figs. A8 and A9 but made for alloys low in phosphorus (0.010–0.013%) and high in phosphorus (0.016–0.020%) showed that increased phosphorus moved the critical stress curve to lower molybdenum contents.
2. Phosphorus had already segregated, to some extent, in the HAZ during the weld thermal cycle.
3. For a particular alloy there exists a critical phosphorus level below which embrittlement is not apparent. The minimum critical value of phosphorus of 0.008% coincides with the $1\text{Cr}-\frac{1}{2}\text{Mo}$ composition.
4. For the range of alloys tested, (0–2%Cr, 0.5%Mo), as long as the phosphorus content was below a critical value for a particular alloy, the critical stress level remained approximately the same.

Thus, phosphorus did adversely affect the cracking susceptibility of the Cr-Mo alloys.

Tamaki surmised that the transformation to austenite played a vigorous role in the segregation of phosphorus as phosphorus did not segregate in materials subjected to heat treatment below the A_1 temperature but did segregate when heat treated in the austenite region. It was assumed that the equilibrium distribution of phosphorus was established at the ferrite/austenite interface during the course of transformation and that the diffusion rate is so low that it diffuses little from the ferrite/austenite interface.⁹⁵

These assumptions, combined with the knowledge that the solubility of phosphorus is approximately $2\frac{1}{2}$ times greater in ferrite than in austenite, led Tamaki to conclude that phosphorus was carried along on the ferrite/austenite interface. This interface, which is the prior austenite grain boundary on cooling, is therefore enriched with the phosphorus. The welding cycle thus initiates the segregation and embrittling processes.

An Auger electron spectroscopic study⁹⁴ revealed that grain boundary concentration of phosphorus was much lower in the Fe-C-P alloy than in the Fe-Cr-C-P alloy thus indicating that the addition of chromium to an Fe-C-P alloy considerably enhances grain boundary segregation of phosphorus. This is probably due to the fact that addition of chromium decreases the activity of carbon in Fe-Cr-C-P alloy thus influencing the segregation of phosphorus to grain boundaries.

Sulfur is another residual element that has been found to segregate on fracture surfaces by Auger analysis.^{97,98,107} The segregation of sulfur is associated with dislocation tangles along the boundaries generated by impurity penetration.^{108,109} Thus, impurity segregation to grain boundaries may be associated with the formation of dislocation tangles with which the impurities become interlocked. Such a grain boundary state will be conducive to sliding, will inhibit migration and thus cavity nucleation at suitable particles may occur. Further, the adsorption of embrittling elements on a cavity surface would have a stabilizing effect.

Imanaka *et al.*⁶¹ and a study at Kawasaki Steel Corporation⁸⁴ of the effects of sulfur, rare earth metal (REM) and calcium additions in $2\frac{1}{4}\text{Cr}-1\text{Mo}$ found that if free sulfur exceeded 25 ppm the material showed a high sensitivity to SRC/PWHT cracking but below 25 ppm exhibited no susceptibility. Free sulfur was calculated by the equation:

$$(S) = [\%S - (32/40) \% \text{Ca} - (32/140) \% \text{REM}] \times 10^4 \text{ ppm.}$$

Ladle calcium treatments appear to improve susceptibility to SRC/PWHT cracking. Lundin *et al.*¹¹⁰ have also found low susceptibility in a heat of modified $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel treated with calcium compared to heats of similar composition not treated with calcium. Nikhitin and Kreshchanovsky³² found that treatment of molten CrMoV steel with BaCl_2 and NaCl improved ductility and toughness and at the same time rendered the prior austenite grain boundaries resistant to etching.

Sulfur in the form of grain boundary sulfides has been linked with the initiation of cavitation on the grain boundaries.^{58,111,112} Middleton and Fujii *et al.*¹¹³ have suggested that submicron ($\sim 0.1 \mu\text{m}$) $\alpha\text{-MnS}$ inclusions at prior austenite boundaries in HAZ regions could act as nucleation sites for cavitation in CrMoV steels. Similar observations were also made

by Hipsley *et al.*^{58,101} on 2¼Cr-1Mo steels. Ayres *et al.*⁴⁸ also report that MnS inclusions are responsible for SRC/PWHT cracking due to liquation. Thus, the cracking susceptibility is increased with increase in energy input and a decrease in Mn/S ratio especially at higher carbon levels. However, it is generally believed that while liquation cracks may help to initiate SRC/PWHT cracks, their presence is not a requirement for the formation of cracks.

Hipsley *et al.*^{58,111} have shown that at PWHT temperatures in excess of 1110°F (600°C), materials susceptible to cracking exhibit a change from smooth intergranular fracture surfaces to fracture surfaces with extensive intergranular microvoid coalescence. The microvoids were said to have nucleated on sulfides which were thought to have low interfacial strength with the matrix. The presence of sulfides was thought to be the result of solution of MnS during the thermal cycle and subsequent preferential reprecipitation at grain boundary sites during cooling. Shin and McMahon¹¹⁴ and Hipsley *et al.*^{115,117} have also suggested stress induced segregation of sulfur to the grain boundaries ahead of the crack tip region during stress relief by grain boundary diffusion and also sulfide dissolution resulting in low crack growth energy requirements.

However, Dolby¹¹² reports that the relationship between grain boundary particles and SRC/PWHT cracking is unclear in that some researchers have found that grain boundary particles have no effect on cracking. Further, both Horn and Kunze⁹⁸ and Kikuchi and Nakao⁹⁷ have found that even in relatively high sulfur materials a reduction in the phosphorus content eliminated cracking in susceptible materials. Thus, the effect of sulfur is apparently minor in comparison to that of phosphorus.

Sun *et al.*,¹⁰⁷ in a study of ultra pure heats of A 508-2 type steel with controlled addition of various impurities, conclude that A 508-2 steel normally susceptible to SRC/PWHT cracking, a very high resistance to cracking was evident in high purity laboratory heat. Thus, SRC/PWHT cracking depends on trace impurities. They found that boron (2 ppm) increased susceptibility markedly even when the sulfur level was 10 ppm. They also suggest that in the presence of boron and sulfur, phosphorus additions decreased cracking susceptibility, which is generally considered to be contradictory to the popular belief.

In their study to investigate the PWHT cracking susceptibility of 2¼Cr-1Mo and ½Cr-Mo-V steel Batte and Murphy¹¹⁸ conclude that by an appropriate choice of welding process and parameters PWHT cracking can be prevented. According to them it is unnecessary to reduce the residual element content below the levels generally obtained in current low-alloy steel-making practice.

In summary, it is evident that composition is probably the most significant factor in SRC/PWHT cracking of the Cr-Mo alloys. Of the alloying elements, molybdenum and vanadium have the greatest

effect due to the preferential formation of coherent or fine precipitates in the grain matrix which significantly increases the flow strength of the grain matrix over that of the grain boundaries. Of the residual elements, phosphorus has the greatest effect due to segregation to and embrittlement of the grain boundaries. The effect of composition is complicated by precipitation kinetics, element interactions and the superposition of the effects of other elements such as sulfur, titanium, tin and copper.

Testing Techniques for SRC/PWHT Cracking

To determine the susceptibility to SRC/PWHT cracking of a certain steel, a suitable test must be chosen or developed.^{48,74} Preferably the specimen should incorporate a welded joint similar to, and made with the same heat inputs as, the joints used in the structure. Furthermore, the thermal treatment should duplicate as closely as possible the PWHT/service exposure. The complex interactions of residual stress and strain, their magnitude and distribution, triaxiality and relaxation together with the progressively changing microstructures during PWHT or service can hardly be integrated in a small weld sample. Baker⁸³ has enumerated some of the factors that should be included for the ideal PWHT cracking test. The specimen should include some defect or notch in a microstructure representative of the actual HAZ. The HAZ should be strained in a manner similar to that experienced in an actual weld. Finally, the test method employed should preferably incorporate stress relaxation since constant load testing may introduce misleading factors into the evaluation. Dhooze *et al.*⁴⁸ note that the incorporation of all the factors involved in an actual weld into a small specimen would be difficult.

A common theme among papers reviewing SRC/PWHT cracking is that there exists a multitude of tests that have been employed and therefore comparison of test results is difficult and sometimes confusing.^{41,47,48,74} Dhooze *et al.*⁴⁸ cite the use of 26 different testing techniques prior to 1978. In order to simplify the overall number of tests three categories were proposed to classify the tests.

A. Tests on complete weldments

- (1) Lehigh restraint test
- (2) Restrained fillet weld test
- (3) Y-groove test
- (4) Stellite bead test
- (5) BWRA ring test
- (6) Steampipe butt-weld test
- (7) H-type restraint test
- (8) Circular patch restraint test
- (9) Strained root bead test
- (10) Restrained butt weld test
- (11) MRT test
- (12) NRC Regulatory Guide cladding test

B. Tests on specimens containing a weld

- (1) Jigged stress relaxation test
- (2) Isothermal constant load rupture test

- (3) Implant test
- (4) Controlled heating rate, constant load test
- (5) Vinckier test, Stainless backing bar test
- C. Tests on specimens containing a thermally simulated HAZ
 - (1) Isothermal stress relaxation test
 - (2) Isothermal constant load rupture test
 - (3) Isothermal slow strain rate tensile test
 - (4) Controlled heating rate stress relaxation test
 - (5) Controlled heating rate constant load test
 - (6) Stainless sleeve test
 - (7) Pre-cracked bend test
 - (8) Slit tube test
 - (9) Jigged stress relaxation test

Meitzner⁴¹ in 1975 separated tests into only two categories: direct weldment tests (those containing an actual weld or portions of an actual weld) and simulated weld tests. Vinckier^{47,74} in 1974 used three categories to divide some 17 different testing techniques.

1. Tests of a welded specimen: Specimens are welded with various degrees of restraint which depends on specimen geometry and subjected to PWHT followed by metallographic examination for cracks.
2. Tensile tests at high temperatures on welded or simulated specimens: The various parameters used to rank the susceptibility of materials in such tests are the rupture time, stress, reduction in area, elongation, etc.
3. Simulation of the PWHT cycle on simulated specimens: Time to rupture, temperature at failure, stress, reduction in area, etc. are the criteria in these tests to rank the susceptibility of materials.

The use of tests on complete weldments or direct weldment tests have the obvious advantage of being directly related to the actual weld and PWHT/service conditions as well as joint geometry. However, both Meitzner⁴¹ and Dhooge *et al.*⁴⁸ note that reproducibility is a significant problem with Meitzner citing erratic results in which one specimen may crack extensively while another may not crack at all. Also, a small test specimen will experience a significantly smaller amount of creep strain than a large welded structure.^{41,74}

Weld simulation tests have the advantage of reproducibility, known stress level and control of microstructure.^{6,41} In simulated specimens it is also possible to locate accurately a notch in a well defined microstructure and that a number of such specimens can be reproduced. The disadvantages are that the strains associated with weld contractional stresses are not duplicated and in general, only one region of the HAZ is tested and the effects of adjacent weld metal and base metal are not present.^{41,48} Therefore, the results of simulation tests must be evaluated carefully before being directly applied to full scale weldments.

Although no standard exists the most popular type of test is the tensile or stress rupture test.⁴⁸ Several means of evaluating this type of test have been proposed. Vinckier⁶ suggested that in hot tensile tests the minimum acceptable reduction in area for an alloy to be considered not susceptible to cracking was 20%. Meyers and Price¹¹⁹ suggested that, in constant load stress rupture testing HAZ, base metal and weld metal rupture strengths be compared. If the HAZ rupture strength was lower than the base metal and weld metal strength, the material was to be considered susceptible and the greater the difference the greater the susceptibility. Spaeder and Plodder⁸² have proposed a stress rupture parameter in which the reduction in area was multiplied by the stress to rupture in 10 min. If the stress rupture parameter was less than 15×10^4 the material was determined susceptible to cracking. Vinckier and Pense^{48,74} in extensive Gleeble simulation have determined the following criteria for simulated PWHT at 1110°F (600°C):

1. extremely susceptible if the reduction in area is less than 5% and peak temperature of the weld simulation exceeds 2190°F (1200°C);
2. highly susceptible if the reduction in area is below 10% and peak temperature exceeds 2370°F (1300°C);
3. slightly susceptible if the reduction in area is below 15% and the peak temperature exceeds 2460°F (1350°C); and
4. not susceptible if the reduction in area exceeds 20% at all peak temperatures.

Thus, tensile type or stress rupture tests have been shown to yield a general determination as to the susceptibility of a material to PWHT cracking and allow a ranking of various materials to be determined.⁴¹

In accordance with the factors enumerated by Baker⁸³ Lundin *et al.*^{91,110,133} have developed a C-ring test that appears to incorporate all the factors for an ideal PWHT cracking test. For this test a square bar (1 in. × 1 in.) is machined with a flat bottomed groove at the center of one of the faces parallel to the plate surface. The groove geometry can be modified in order to accommodate different diameter SMAW electrodes. A single pass SMA weld metal of matching chemistry is used to fill the groove. The bar is then machined into hollow cylinder (1 in. outside diameter, wall thickness 0.125 in.). The cylinder is then sliced to obtain pieces 0.75 in. long. A V-notch is then broached in the CGHAZ and holes drilled at 90° from the notch axis. The back of the ring opposite to the notch is removed in order to allow compression of the ring by tightening of a bolt placed through the drilled holes. The rings are then stressed by tightening the bolt (an equation was developed to relate the reduction in diameter due to tightening the bolt and stress at the root of the notch) and postweld heat treated. To determine cracking the specimen is metallographically or fractographically examined after the test.

Thus, this test includes a weld, a discontinuity and stress relaxation during PWHT similar to that experienced in a weldment. In addition, this test is simple and requires no special equipment. They found that the results of this test correlated well with the Gleeble simulation test results in terms of ranking heats according to PWHT cracking susceptibility.

Another test developed at the University of Tennessee is the Spiral Notch Test. This test has been found to be very useful in evaluating a controlled deposition techniques in HSLA steel for successful prediction of in-service behavior. This test is described in detail in the following chapters.

The ideal test described by Baker⁸³ can perhaps be accomplished by the use of two or more tests that make more efficient use of materials and funds. Utilizing Gleeble simulation to make an initial ranking of a material's SRC/PWHT cracking susceptibility will determine if further testing is indicated. If the material is deemed susceptible to cracking further testing may be employed with specimens incorporating a weld or full scale weldments to determine if the material is susceptible under welding conditions employed in actual weldment fabrication.

Ways to Mitigate SRC/PWHT Cracking

The most obvious remedy for SRC/PWHT cracking, as suggested by Stout,¹²⁰ is to avoid the use of susceptible alloys. Murray²⁶ has recommended the use of alloys with limited molybdenum and vanadium. Mullary and Cadman⁴⁶ have recommended the use of materials in which the sulfur, phosphorus, and vanadium contents are minimized and manganese content is increased.

When the avoidance of susceptible alloys is not practical, process changes may be beneficial. Murray²⁶ and Pense *et al.*⁵⁶ have recommended using a low strength weld metal with high strength base metals to allow deformation to occur in the fusion zone rather than in the HAZ. Mullary and Cadman⁴⁶ have proposed buttering susceptible alloys to promote the formation of coarser carbides in the CGHAZ of the parent material. Muraki *et al.*⁶⁰ have suggested preferential welding of the larger side of unsymmetrical, double-V groove butt joints.

Increasing heat input or preheat has been seen to bring about favorable microstructural changes in the HAZ.^{56,60,83,121} Higher energy inputs and preheating results in slower cooling rates and softer transformation products.¹²⁰ Overaging of carbides in the HAZ during cooling,⁸ or an autostress relief.³⁶ Preheating may also be beneficial in decreasing the temperature range over which internal stresses build up.¹²² However, Vinckier⁶ reported that higher heat input yielded mixed results. Higher heat input causes two competing effects to occur: a coarser grain structure results which increases intergranular embrittlement and a softer HAZ is formed which decreases intergranular embrittlement.

An alternative to increasing heat input is to de-

crease heat input in multiple pass welds. Miller and Batte⁵⁹ found that using small gas tungsten arc weld beads in a multiple pass weld increased the amount of grain refinement due to overlapping of the HAZs and that even highly susceptible materials could be welded with this technique with no evidence of cracking after PWHT. Similarly, the use of temper beads, small stringer beads placed over the last pass to refine the grain structure of the HAZ, have been found to benefit in eliminating stress relief cracking.⁶

Complete austenitization after welding^{6,123} or normalizing^{30,47} has been seen to increase the HAZ toughness and prevent cracking. Bentley⁴⁰ recommended using high PWHT temperatures and high heating rates to avoid prolonged coherent precipitate formation. Debiez and Granjon³⁰ also found that increasing the heating rate is effectual in reducing the risk of cracking. However, Hipsley¹²⁴ and Debeiz and Granjon³⁰ also indicate that high heating rates produce marginal benefits in reducing cracking but speculated that high heating rates might exacerbate cracking in actual weldments due to the introduction of additional tension stresses due to temperature gradient. Kanazawa *et al.*¹²¹ have found no effect with a heating rate from 80°C/hour–400°C/hour on PWHT cracking susceptibility. Ueda¹²³ has suggested the use of higher temperature and longer time PWHT for the prevention of grain boundary embrittlement.

Weld dressing to remove discontinuities at weld toes has been found effective in reducing cracking by elimination of crack initiation sites.^{26,46,60,120} Peening has also been suggested to eliminate residual tensile stresses at the surface of the weld.¹²⁰ Weld stresses can also be reduced by attention to weld sequencing, the use of techniques such as backstepping, the use of interstage PWHT, and simplification of the design of weldments to lower overall levels of restraint.²⁶

Thus, it may be possible to mitigate SRC/PWHT cracking in susceptible alloys by utilizing the techniques described above. Perhaps the most economical means is to eliminate stress risers through careful dressing of the surface of the weld. If performed correctly, temper beading combined with weld dressing should provide a microstructure and surface unfavorable to the initiation and propagation of cracks. One important factor however, is that oxidation during PWHT/service may provide initiation sites by oxide penetration and wedging along grain boundaries.³⁰ Therefore inherently susceptible materials may not be prevented from cracking by simple measures and it becomes necessary to know the susceptibility of individual alloys to SRC/PWHT cracking in order to ascertain the usefulness of preventive measures.

Vinckier⁶ has recommended the following checks if susceptible steels are to be used and thermal stress relieving must be carried out.

1. Tensile tests should be carried out at PWHT temperatures on weld simulation specimens

heated to 2462°F (1350°C). A minimum ductility of 20% reduction in area should be required.

2. The design should be checked to avoid as much as possible built in stress raisers and to allow adequate nondestructive inspection of all joints.
3. By modifying the welding procedure (weld dressing, temper beads, raising preheat temperature, etc.) cracking in susceptible microstructures can sometimes be avoided.
4. Careful nondestructive inspection should take place after stress relieving the structure and again after final pressure testing.

The complexity of SRC/PWHT cracking is demonstrated in Fig. A16 by the interrelationship of the many factors. As has been shown previously the superposition of several factors may result in the occurrence of cracking but when the same factors are considered alone cracking may not occur.

From the review presented thus far one very important point is that SRC/PWHT cracking susceptibility of a steel is a complex interaction of variables such as composition, residual element levels, welding parameters, joint design, etc. In addition, there are other factors to be taken into account such as the persistence of Mo_2C carbides or other similar carbides of vanadium, titanium and niobium that are in coherence with the matrix. Thus, higher chromium steels such as 5Cr and 9Cr steels even with high levels of residual elements are not susceptible since the carbides that form during PWHT/service do not cause secondary strengthening. Other factors are austenite grain growth and the extent of carbide dissolution during HAZ thermal excursion. For example, in case of incomplete dissolution of carbides it is possible that during PWHT/service additional carbides precipitate on the preexisting carbides such that there is no significant secondary hardening or it may

be possible that secondary hardening behavior is completely eliminated. Thus, it is not possible to rate a steel or a heat of steel as susceptible based on the composition or tramp element concentration alone. According to many researchers it is possible to weld any steel without SRC/PWHT cracking if proper design and welding procedures are followed and that it is unnecessary to reduce the level of impurity elements below the levels that are achieved today with good steelmaking practices.

Creep Embrittlement/Stress Rupture Cracking

A variety of Cr-Mo steels operating in the temperature range of 850–1050°F (450–565°C) show low ductility intergranular failure similar to that seen in the case of PWHT cracking.^{4,31,37,63,125–130} The term creep embrittlement has been used to describe this behavior. It has been suggested that creep embrittlement occurs when hardening mechanisms within the grains prevent relaxation of stress such that grain boundary sliding occurs. It is apparent that PWHT may not be adequate to relax the residual welding stresses and stabilize the microstructure so that microstructural evolution in the HAZ can be in progress at the operating temperature. System stresses and residual stresses can thus cause creep embrittlement/stress rupture cracking (SRC).

However, creep embrittlement should not be confused with temper embrittlement which is a loss in toughness associated with a shift in the transition temperature. The toughness is not necessarily low at the operating temperature in a temper embrittled material. Also, temper embrittlement is reversible unlike creep embrittlement. In a creep embrittled material the tensile ductility or fracture toughness may be adequate at ambient temperatures. It is only at elevated temperatures that a creep embrittled material shows low ductility intergranular fracture.

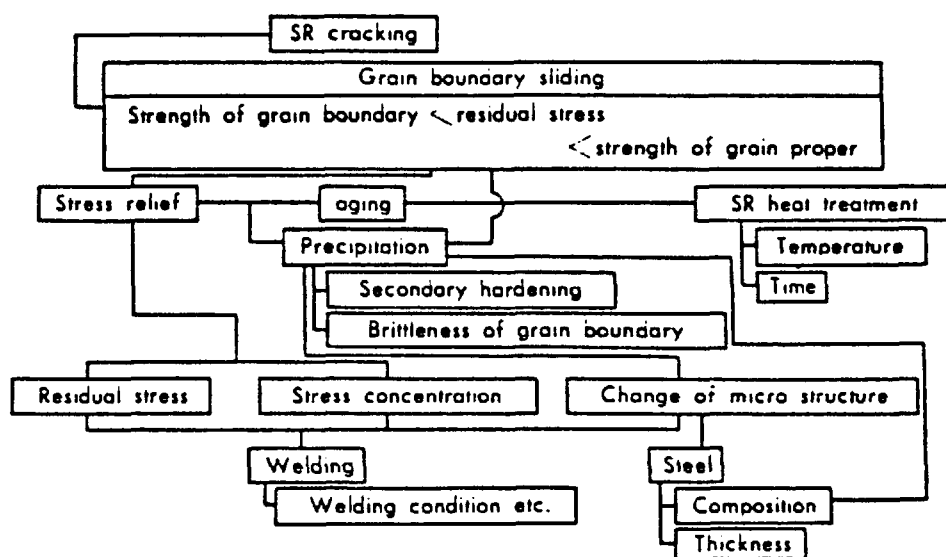


Fig. A16—Block diagram showing factors involved in stress relief cracking. Source: Ito, Y. and Nakanishi, M., "Study on Stress Relief Cracking in Welding Low Alloy Steels," The Sumitomo Search No. 7, May 1972, pp. 27–36

The term creep embrittlement as described by various researchers diverges somewhat depending on their particular experimental method of study or other circumstances. Some investigators apply the term creep embrittlement to stress rupture notch sensitivity, while some are describing actual observable creep damage (e.g., voids). Some researchers have used this term to describe simply low ductility creep failures (i.e., stress rupture cracking (SRC)). Such embrittlement may be due to weakened grain boundaries caused by either precipitation hardening of the grain interiors or by deleterious grain boundary segregates or both. Although creep embrittlement is a phenomenon that normally occurs in-service and is associated with metallurgical changes over time at operating temperature, it is important to remember that all microstructural changes occurring in Cr-Mo steels during service are not necessarily detrimental.

Emmer *et al.*¹³⁰ have suggested the following characteristics of creep embrittlement and parenthetical portions of the following are added to show the relevance to the HAZ.

1. Creep embrittlement is dependent on the strength level of the matrix, but is expected to be more severe for materials with high (> 110 ksi) tensile strength. (In Cr-Mo steels the coarse grained region of the HAZ in the as-welded condition exhibits tensile strength usually in excess of 200 ksi.)
2. Creep embrittlement is also dependent on grain size, microstructure, and austenitizing temperatures. A coarse grained material is more susceptible than a fine grained material. Use of unnecessarily high austenitizing temperatures tends to accentuate creep embrittlement. (In the HAZ close to the fusion boundary where the peak temperature experienced is of the order of the melting temperature of the alloy, grain growth cannot be avoided. Thus, in this region the combination of the high austenitizing temperature and coarser grain size presents ideal situations for susceptibility to creep embrittlement.)
3. Creep embrittlement can be most severe under conditions that permit small amounts of creep strain to occur over long periods of time. While creep embrittlement may occur in the temperature range 800–1100°F (425–595°C) it is more severe at lower temperatures. (The situation is even more complicated in welded structures. Different weld configurations such as nozzle welds, girth welds or longitudinal welds presents different degrees of restraint and residual stress states that tend to relax at the operating temperature. In addition, during start up and shut down cycles additional stresses are introduced into the structure. Because of the obvious mismatch between the strength levels (creep rate mismatch) of the various regions of the weld and even within the various regions of the HAZ the

stress relaxation rates are different. As a result stresses may be concentrated at certain regions, especially in those regions that have higher strength and lower ability to accommodate creep strains to relax because of the low ductility.)

4. Creep embrittlement is manifested by a reduction in stress rupture ductility. When ductility is plotted as a function of time to rupture, there is typically a ductility minimum and then, with increasing time, an increase in rupture ductility. (In welded structures in addition to strength mismatch of the various regions as explained above, there is also ductility mismatch. Thus, if the ductility of various regions of welds is plotted as a function of time the ductility minimum in some regions may extend over the entire design life of the component.)

Swift and Rogers³⁷ have related the phenomenon of creep embrittlement to partially coherent precipitation of Mo₂C carbides that form during PWHT. Otoguro *et al.*^{128,131} and Konosu and Maeda¹²⁶ have found segregation of antimony, copper and phosphorus at the prior austenite grain boundaries in CGHAZ of creep embrittled 1¼Cr-½Mo and 2¼Cr-1Mo weldments by Auger analysis. They have indicated that creep embrittlement is a premature transition from notch strengthening to weakening. They have also suggested the use of rare earth additions to mitigate this problem. In his study to determine effects of boron, antimony, tin and phosphorus on creep and temper embrittlement of normalized and tempered 1¼Cr-½Mo steel, Viswanathan¹²⁷ determined that boron additions cause an appreciable decrease in the rupture ductility compared to additions of large amounts of the other impurities. However, in the case of temper embrittlement the other elements were found to be more detrimental than boron. Nagae and Abe¹²⁵ found an increase in the tin and arsenic concentration along the prior austenite grain boundaries in a creep embrittled CGHAZ of a reactor vessel.

Rosenstein¹³² has suggested an increase in the PWHT temperature resulting in tempering to a lower strength level to resist creep embrittlement. Ueda *et al.*¹²³ have also suggested decrease in the amounts of impurities, a full annealing treatment or an increase in the PWHT temperature to improve resistance to creep embrittlement. Konosu and Maeda¹²⁶ have found that an increase in the PWHT temperature considerably increased the ductility in small impurity content heats whereas in heats containing large amounts of impurities the effect of increasing PWHT temperature was small. However, a study at the University of Tennessee revealed improvements in short time stress rupture ductility of the CGHAZ if PWHT at higher temperatures even in heats containing higher-than-usual levels of impurities.

In general, little research has been done to determine creep embrittlement behavior in Cr-Mo steels. However, it is apparent that creep embrittlement is

similar to PWHT cracking but occurs at lower (operating) temperatures over a longer period of time. If temperature and time dependence of stress relief cracking are considered C-curve as has been pointed out by various researchers, then creep embrittlement can be considered to be a form of PWHT cracking at low temperatures where the intersection with the C-curve is at longer times. Thus, creep embrittlement will most likely occur in structures not properly PWHT or in structures where PWHT has been omitted for various reasons. Thus, stress relaxation continues at the operating temperature and the situation is further worsened by the fact that during operation, system stresses add to the residual stresses.

Summary

From the literature review it is clear that although a significant amount of investigation has been carried out to determine reheat cracking susceptibility of various Cr-Mo, Cr-Mo-V and other pressure vessel steels, little information is available on the mechanism of SRC/PWHT cracking. In addition, conflicting ideas further complicate the available information.

As vanadium, titanium, niobium and boron are being used in the new generation of Cr-Mo alloys, with vanadium known to increase SRC/PWHT cracking susceptibility and the influence of titanium, niobium and boron unknown, it is evident that the susceptibility of these alloys to cracking needs to be investigated further.

It has been recognized that refinement of the prior CGHAZ by a subsequent weld pass improves the reheat cracking resistance of the HAZ in these steels which is usually taken advantage of during controlled deposition welding techniques such as temper-bead or half-bead techniques. However, no quantitative data or microstructural investigation in the published literature is available in this regard.

Therefore, it is necessary that the microstructural changes must be distinguished from segregation effects which may be occurring simultaneously if an understanding of the precise mechanism or mechanisms of reheat cracking is to be achieved.

References to Appendix A

1. "Welding Metallurgy," *Welding Handbook*, 7th Edition, Vol. 1, Chapter 4, American Welding Society, 1982.
2. Luther, G. G., Jackson, C. E., and Hartbower, C. E. "A Review of Weldability Test of Carbon and Low Alloy Steels," *Welding Journal, Research Supplement*, October 1949, pp. 376s-396s.
3. Coe, F. R. "The Avoidance of Hydrogen Cracking in Welding," *The Welding Institute*, 1973.
4. Erwin, W. E. and Kerr, J. G. "The Use of Quenched and Tempered 2 1/4Cr-1Mo Steel For Thick Wall Reactor Vessels in Petroleum Refinery Processes: An Interpretive Review of 25 Years of Research and Application," *WRC Bulletin* 275, February 1982.
5. McPherson, R. "Compositional Effects on Reheat Cracking of Low-Alloy Ferritic Steels," *Metals Forum*, Vol. 2(3), 1980, pp. 175-186.
6. Vinckier, A. "Use of Weld Simulation Tests to Study Susceptibility to Reheat Cracking," *Proceedings of International Conference on Welding Research Related to Power Plant*, University of Southampton, England, 17-21 September, 1972, pp. 336-355.
7. Weymueller, C. R. "Chrome-Moly Steel Builds Big Vessels," *Welding Design and Fabrication*, July 1983, pp. 47-52.
8. Ito, Y. and Nakanishi, M. "Study on Stress Relief Cracking in Welding Low Alloy Steels," *The Sumitomo Search*, No. 7, May 1972, pp. 27-36.
9. Easterling, K. *Introduction to the Physical Metallurgy of Welding*, Butterworths, Boston, 1983.
10. Harrison, P. L. and Farrar, R. A., "Application of Continuous

Cooling Transformation Diagrams for Welding of Steels," *International Materials Reviews*, No. 1, Vol. 34, 1989, pp. 35-51.

11. Lundin, C. D., Richey, M. W., and Henning, J. A. "Transformation, Metallurgical Response and Behavior of the Weld Fusion Zone in Cr-Mo Steels for Fossil Energy Application," Final Technical Report No. ORNL/sub/81-7685/01&77, 1985.
12. Wada, T. and Eldis, G. T. "Transformation Characteristics of 2 1/4Cr-1Mo Steel," *Application of 2 1/4Cr-1Mo Steel for Thick-Wall Pressure Vessels*, ASTM STP 755, 1982, pp. 343-362.
13. Habracken, L. J. and Economopoulos, M. "Bainitic Microstructures in Low Carbon Alloy Steels and their Mechanical Properties," *Symposium on Transformation and Hardenability in Steels*, Climax Molybdenum Co., 1967, pp. 69-108.
14. Biss, V. and Cryderman, R. L. "Martensite and Retained Austenite in Hot-Rolled, Low-Carbon Bainitic Steels," *Metallurgical Transactions*, Vol. 2, August 1971, pp. 2267-2276.
15. Thomas, G., Sarikaga, M., Smith, G. D. W., and Barnard, S. J. "Microstructure, Retained Austenite and Mechanical Properties of Experimental 0.3%C Steels," *Proc. Conf. Advances in the Physical Metallurgy and Applications of Steel*, University of Liverpool, 21-24 September, 1981, Metals Society, London, pp. 259-265.
16. Appleby, W. R. Jr., Dunphy, R. P., and Pellini, W. S. "Transformation of Cr-Mo Steels During Welding," *Welding Journal, Research Supplement*, January 1954, pp. 57s-64s.
17. Deb, P., Challenger, K. D., and Burna, R. F. "Microstructural Characterization of Shielded Metal Arc Weldments of a Copper-Bearing HSLA Steel," *Materials Science and Technology*, November 1985, Vol. 1, pp. 1000-1006.
18. American Welding Society, Recommended Practices for Welding of Cr-Mo Steels, *ANSI/AWS D 10.8-86*, AWS, Approved November 18, 1985.
19. Baker, R. G. and Nutting, J. "The Tempering of 2 1/4Cr-1Mo Steel After Quenching and Normalizing," *Journal of Iron and Steel Institute*, July 1959, pp. 257-267.
20. Swindeman, R. W., Thomas, R. D., Nanstad, R. K., and Long, C. J. "Assessment of the need for an Advanced High-Strength Chromium-Molybdenum Steel for Construction of Third-Generation Gasifier Pressure Vessels," ORNL/TM-8873, ORNL Report for DOE Under Contract No. DE-AC05-84OR21400, October 1984.
21. Stone, P. G. and Murray, J. D. "Creep Ductility of Cr-Mo-V Steels," *Journal of Iron and Steel Institute*, Vol. 203, November, 1965, pp. 1094-1107.
22. Woodhead, J. H. and Quarrel, "The Role of Carbides in Low-Alloy Creep-Resisting Steels," Climax Molybdenum Company Ltd.
23. Carruthers, R. B. and Collins, M. J. "Carbide Transformations in Microstructurally Unstable Low Alloy Ferritic Steel," *Metal Science*, Vol. 17, March 1983, pp. 107-110.
24. Todd, J. A. "The Early Stages of Tempering in a 3Cr-1.5Mo Steel," *Scripta Metallurgica*, Vol. 20, 1986, pp. 269-274.
25. Buchmayr, B., Cerjak, H., and Fauland, H. P. "The Effect of the Precipitation Behavior on the HAZ-Properties of 1%Cr-Mo-V-Steel," *2nd International Conference on Trends in Welding Research*, Gatlinburg, Tennessee, May 1989.
26. Murray, J. D. "Stress-Relief Cracking in Carbon and Low Alloy Steels," *British Welding Journal*, Vol. 14(8), August 1967, pp. 447-456.
27. Tamaki, K. and Suzuki, J. "Effect of Chromium and Molybdenum on Reheat Cracking Sensitivity of Steels," *Transactions on the Japan Welding Society*, Vol. 14(2), October 1983, pp. 39-43.
28. Vinckier, A. and Dooge, A. "Reheat Cracking in Welded Structures During Stress Relief Heat Treatments," *Journal Of Heat Treating*, Vol. 1(1), ASM, 1979, pp. 72-80.
29. Vinckier, A. and Dooge, A. "Reheat Cracking in Welded Structures During Stress-Relief Heat Treatments," *Proceedings of Conference on Residual Stresses in Welded Constructions and Their Effects*, London, 15-17 November, 1977, Welding Institute, Cambridge, 1978, pp. 375-385.
30. Debiez, S. and Granjon, H. "Work Carried out by the Institut de Soudure on Reheat Cracking," Paper 37 in *Proceedings of Conference on Residual Stresses in Welded Construction and Their Effects*, London, 15-17 November, 1977, Welding Institute, Cambridge, 1978, pp. 231-239.
31. Briant, C. L. and Banerji, S. K. "Intergranular Failure in Steel: The Role of Grain-Boundary Composition," *International Metals Reviews*, No. 4, Review 232, 1978, Vol. 23, pp. 164-199.
32. Boniszewski, T. and Eaton, N. F. "Electron Fractography of Weld-Reheat Cracking in CrMoV Steel," *Metal Science*, Vol. 3, 1969, pp. 103-110.
33. Kanazawa, S., Yamato, K., Takeda, T., and Hasimoto, K. "Study of Reheat Cracking in Weldment (Report 2)—Relation Between Cracking Susceptibility and Some Properties of HAZ at High Temperature," *Transactions of the Japan Welding Society*, Vol. 8(2), September 1977, pp. 39-45.
34. Kanazawa, S., Yamato, K., Takeda, T., and Hasimoto, K. "Study of Reheat Cracking in Weldment (Report 3)—Relation Between Cracking Susceptibility and Some Stress Relaxation Characteristics," *Transactions of the Japan Welding Society*, Vol. 9(1), April 1978, pp. 22-27.
35. Swift, R. A. and Rogers, H. C. "A Critical Review of Weld Metal Embrittlement," *Welding Journal, Research Supplement*, August 1971, pp. 357s-373s.
36. Swift, R. A. "The Mechanism of Stress Relief Cracking in 2 1/4Cr-1Mo Steel," *Welding Journal, Research Supplement*, May 1971, pp. 195s-200s.
37. Swift, R. A. and Rogers, H. C. "Embrittlement of 2 1/4Cr-1Mo Steel Weld Metal by Postweld Heat Treatment," *Welding Journal, Research Supplement*, April 1973, pp. 145s-153s.
38. Swift, R. A. "Creep-Rupture Tests on 2 1/4Cr-1Mo Simulated Heat-Affected Zone," Lukens Steel Company Research Division Report Serial No. RDR69-11, RP1008, June 1969.
39. Orr, J., Beckitt, F. R., and Fawkes, C. D. "The Physical Metallurgy of Cr-Mo Steels for Fast Reactor Boilers," Paper 16 of *Proceedings of Conference on Ferritic Steels for Fast Reactor Steam Generators*, British Nuclear Energy Society, London, 1978, pp. 91-109.
40. Bentley, K. P. "Precipitation During Stress Relief of Welds in Cr-Mo-V Steels," *British Welding Journal*, 11(10), 1964, pp. 507-515.
41. Meitzner, C. F. "Stress-Relief Cracking in Steel Weldments," *WRC Bulletin* 211, November 1975.
42. Nichols, R. W. "Reheat Cracking in Welded Structures," *Welding in the World*, 7(4), 1969, pp. 36-44, *Welding Research Abroad*, March 1970, pp. 36-44.
43. Meitzner, C. F. and Pense, A. W. "Stress-Relief Cracking in Low-

Alloy Steel Weldments," *Welding Journal, Research Supplement*, Vol. 48(10), 1969, pp. 431s-440s.

44. Edwards, R. H., Barbaro, F. J., and Gunn, K. W. "Stress Relief Cracking in Cr-Mo-V Steels," *Metals Forum*, Vol. 5(2), 1982, pp. 119-129.

45. Lundin, C. D. and Khan, K. K. "PWHT Cracking Susceptibility of 2 1/4Cr-1Mo MPC Materials, Report Submitted to MPC, June 1987.

46. Mullery, F. and Cadman, R. O. L. "Cracking of Welded Joints in Ferritic Heat-Resisting Steels," *British Welding Journal*, April 1962, pp. 212-220.

47. Vinckier, A. and Pense, A. W. "A Review of Underclad Cracking in Pressure-Vessel Components," *WRC Bulletin* 197, August 1974.

48. Dhooze, A., Dolby, R. E., Seville, J., Steinmetz, R., and Vinckier, A. G. "A Review of Work Related to Reheat Cracking in Nuclear Reactor Pressure Vessel Steels," *International Journal of Pressure Vessels and Piping*, Vol. 6(5), September 1978, pp. 329-409.

49. Presser, R. I. and McPherson, R. "The Role of Trace Elements in Reheat Cracking," *Proceedings of Conference on Residual Stresses in Welded Construction and Their Effects*, London, 15-17 November, 1977, Welding Institute, Cambridge, 1978, pp. 387-398.

50. Hipsley, C. A., King, J. E., and Knott, J. F. "Toughness Variations and Intergranular Fracture During the Tempering of Simulated HAZ Structures in a Mn-Mo-Ni Steel," *Proceedings of International Conference on Advances in the Physical Metallurgy and Application of Steels*, University of Liverpool, 21-24 September, 1981, Metals Society, London, England, pp. 147-155.

51. Hipsley, C. A., Knott, J. F., and Edwards, B. C. "A study of Stress Relief Cracking in 2 1/4Cr-1Mo Steel-II. The Effects of Multi-Component Segregation," *Acta Metallurgica*, Vol. 30, 1982, pp. 641-654.

52. Troiano, A. R. "The Role of Hydrogen and Other Interstitials in the Mechanical Behavior of Metals," *Trans. ASM*, Vol. 52, 1960, pp. 54-80.

53. Linnert, G. E. *Welding Metallurgy*, American Welding Society, Miami, Florida, 1983.

54. Udin, H., Funk, E. R., and Wulf, J. *Welding for Engineers*, John Wiley and Sons, New York, 1954.

55. Nakamura, H., Naiki, T., and Okabayashi, H. "Fracture in the Process of Stress-Relaxation Under Constant Strain," *1st International Conference on Fracture*, September 1965, Vol. 2, Sendai, Japan, pp. 863-878.

56. Pense, A. W., Galda, E. J., and Powell, G. T. "Stress Relief Cracking in Pressure Vessel Steels," *Welding Journal, Research Supplement*, August 1971, pp. 374s-378s.

57. McMahon, C. J. Jr., Dobbs, R. J., and Gentner, D. H. "Stress Relief Cracking in MnMoNi and MnMoNiCr Pressure Vessel Steels," *Materials Science and Engineering*, Vol. 37, 1979, pp. 179-186.

58. Hipsley, C. A., Knott, J. F., and Edwards, B. C. "A Study of Stress Relief Cracking in 2 1/4Cr-1Mo Steel-I. The Effects of P segregation," *Acta Metallurgica*, Vol. 28, 1980, pp. 869-885.

59. Miller, R. C. and Batte, A. D. "Reheat Cracking in 1/2CrMoV/2 1/4CrMo Welded Joints," *Metal Construction*, Vol. 7(11), November 1975, pp. 550-558.

60. "Reheat Cracking of 2 1/4Cr-1Mo Pressure Vessel Steel," Japan Steel Works Research Laboratory Internal Report, Ref No. PV 76-78, July 25, 1975.

61. Imanaka, T., Sato, S., Shimomura, J., and Aso, K. "Improvements in the Stress Relief Cracking of 2 1/4Cr-1Mo Steels," Kawasaki Steel Corporation, Mizushima, Japan, 1985.

62. Boniszewski, T. "Reheat Cracking in 2 1/4Cr-1Mo SA Weld Metal," *Metal Construction*, 14(9), 1982, pp. 495-496.

63. Bruscato, R. "Temper Embrittlement and Creep Embrittlement of 2 1/4Cr-1Mo Shielded Metal-Arc Weld Deposits," *Welding Journal, Research Supplement*, April 1970, pp. 148s-156s.

64. Tamaki, K. and Suzuki, J. "Assessment of Reheat Cracking Sensitivity of Steels from their Chemical Composition," *Metals/Materials Technical Series*, 85 ASM's International Welding Congress, Toronto, Canada, 14-17 October, 1985, pp. 1-8.

65. Tamaki, K. and Suzuki, J. "Reheat Cracking Test on High Strength Steels by a Modified Implant Test," *Transactions on the Japan Welding Society*, Vol. 14(2), October 1983, pp. 33-38.

66. Tamaki, K. and Suzuki, J. "Assessment of the Reheat Cracking Sensitivity of Cr-Mo Steels," *The fourth International Symposium of the Japan Welding Society*, November 1982, Osaka, 4JWS-II-17, pp. 467-472.

67. Tamaki, K., Suzuki, J., Nakaseko, Y., and Tajiri, M. "Effect of Carbides on Reheat Cracking Sensitivity," *Transactions on the Japan Welding Society*, Vol. 15(1), April 1984.

68. Tamaki, K., Suzuki, J., and Kojima, M. "Combined Influence of Chromium, Molybdenum and Vanadium on Reheat Cracking of Steels," *IIW Document IX-1518-88*.

69. Cadman, R. O. L. Discussion Session 4, Part 1, "Heat Treatment Cracking," Paper 11, *Metal Construction and British Welding Journal*, Vol. 1(2), 1969, pp. 125-128.

70. "Improvement in High Temperature Properties of Heavy Section 2 1/4Cr-1Mo Steel," Kawasaki Steel Corporation, January 1982.

71. Myers, J. "Influence of Alloy and Impurity Content on Stress-Relief Cracking in Cr-Mo-V Steels," *Metals Technology*, Vol. 5(11), November 1978, pp. 391-396.

72. Jones, K. E. "Stress-Relief Cracking of Welded Pipe/Casting Joints in Cr-Mo-V Steel," Paper 5 in *Proceedings of Conference on Welding Creep Resistant Steels*, The Welding Institute, Abington Hall, Cambridge, 17-18 February, 1970, pp. 66-78.

73. Ritchie, R. O., Parker, E. R., Spencer, P. N., and Todd, J. A. "A New Series of Advanced 3Cr-Mo-Ni Steels for Thick Section Pressure Vessels in High Temperature and Pressure Hydrogen Service," DOE/UC/07843-1, Topical Report No. UCB/RP/84/D1020, Feb. 1, 1980-Jan. 31, 1984, September 1984.

74. Anon. "Testing Techniques to Study the Susceptibility to Reheat Cracking of Carbon-Manganese and Low Alloy Steels," *Welding in the World*, Vol. 12(11/12), 1974.

75. Bodnar, R. L., Ohhashi, T., and Jaffee, R. I. "Effects of Mn, Si and Purity on the Design of 3.5NiCrMoV, 1CrMoV, and 2.25Cr-1Mo Bainitic Alloys," *Metallurgical Transactions A*, Vol. 20A, August 1989, pp. 1445-1460.

76. Weng, Y. Q. and McMahon, C. J. Jr. "Interaction of Phosphorus, Carbon, Manganese, and Chromium in Intergranular Embrittlement of Iron," *Materials Science and Technology*, Vol. 3, March 1987, pp. 207-216.

77. Yu, J. and McMahon, C. J. Jr. "The Effect of Composition and Carbide Precipitation on Temper Embrittlement of 2.25Cr-1Mo Steel: Part II. Effects of Mn and Si," *Metallurgical Transactions A*, Vol. 11A(2), February 1980, pp. 291-300.

78. Ratliff, J. L. and Brown, R. M. "The Deleterious Effect of Small Quantities of Aluminum on the Stress-Rupture Properties of a Cr-Mo-V Steel," *Transactions of the American Society of Metals*, Vol. 60, 1967, pp. 176-186.

79. Yu, J. "Carbide Stability Diagrams in 2.25Cr-1Mo Steels," *Metallurgical Transactions A*, Vol. 20A, August 1989, pp. 1561-1563.

80. Myers, J. "Effects of Deoxidants and Impurities on Simulated Stress Relief Cracking of 1/2Cr 1/2Mo 1/4V Steel," *Proceedings of International Conference on Welding Research Related to Power Plant*, University of Southampton, England, 17-21 September, 1972, pp. 356-368.

81. Harris, P. and Jones, K. E. "The Effect of Composition and Deoxidation Practice on the Reheat Cracking Tendencies of 1/2Cr 1/2Mo 1/4V Steel," *Proceedings of International Conference on Welding Research Related to Power Plants*, University of Southampton, England, 17-21 September, 1972, pp. 369-380.

82. Spaeder, G. J. and Plodder, G. "Stress-Rupture Properties of the Simulated Grain-Coarsened Weld-Heat-Affected Zone of USS "T-1" Steel," Technical Report, Project No. 40.02-004(1), Progress Report No. a-AS-IP-8.

83. Edmondson, B. Rapporteur's Remarks and Discussion of Paper 24-27 of *Proceedings of Conference on Welding Research Related to Power Plant*, University of Southampton, 17-21 September, 1972, CEBG, London, 1972, pp. 435-448.

84. "Developments of Advanced 2 1/4Cr-1Mo and 3Cr-1Mo Steels-On the Effects of Elements," Kawasaki Steel Corporation, May 13, 1985.

85. Viswanathan, R. and Beck, C. G. "Effect of Aluminum on the Stress Rupture Properties of Cr-Mo-V Steels," *Metallurgical Transactions A*, Vol. 6A, November 1975, pp. 1997-2003.

86. Presser, R. I. and McPherson, R. "Boron Segregation and Elevated Temperature Embrittlement of Ferritic Steel," *Scripta Metallurgica*, Vol. 11, 1977, pp. 745-749.

87. Paju, M. and Moller, R. "The Effect of Boron on Phosphorus Segregation in Austenite," *Scripta Metallurgica*, Vol. 18, August 1984, pp. 813-815.

88. Hunter, A. N. R. "Reheat Cracking in 2 1/4Cr-1Mo Submerged Arc Weld Metal-Part 1-Materials, Welding and Testing," *Metal Construction*, 14(4), 1982, pp. 198-201.

89. Hunter, A. N. R. "Reheat Cracking in 2 1/4Cr-1Mo Submerged Arc Weld Metal-Part 2-Results and Discussion," *Metal Construction*, 14(5), 1982, pp. 266-270.

90. Wolstenholme "Transverse Cracking and Creep ductility of 2CrMo Weld Metals," CEBG Research Division Report No. R/M/R266.

91. Menon, R., Lundin, C. D., and Chen, Z. "Postweld Heat Treatment Cracking in High Strength Low Alloy Steels," *WRC Bulletin* 349, December 1989, pp. 22-30.

92. Lundin, C. D., Khan, K. K., and Gill, T. P. S. "PWHT/Reheat/Stress Rupture Cracking and HAZ Toughness in Cu-Precipitation Hardenable Steel-ASTM A 710," *Proceedings of the International Conference on The Metallurgy, Welding, and Qualification of Microalloyed (HSLA) Steel Weldments*, Cosponsored by Microalloying International and AWS, Nov. 6-8, 1990, Houston, Texas, pp. 250-275.

93. Balaguer, J. P., Wang, Z., and Nippes, E. F. "Stress Relief Cracking of a Copper Containing HSLA Steel," *Welding Journal Research Supplement*, April 1989, pp. 121s-131s.

94. Ehart, H., Grabke, H. J., and Onel, K. "Grain Boundary Segregation of Phosphorus in Iron Base Alloys: Effects of Carbon, Chromium and Titanium," *Proceedings of Conference on Advances in the Physical Metallurgy and Application of Steels*, University of Liverpool, The Metals Society, London, England, 21-24 September, 1981, pp. 282-285.

95. Tamaki, K. and Suzuki, J. "Effect of Phosphorus on Reheat Cracking of Cr-Mo Steels," *Research Reports of the Faculty of Engineering, Mie University*, Vol. 9, December 1984, pp. 8-16.

96. Wittig, J. E., Sinclair, R., and Viswanathan, R. "Precipitation of Phosphorus and Tin in Temper Embrittled Low Alloy Steel," *Scripta Metallurgica*, Vol. 19(1), January 1985, pp. 111-116.

97. Kikuchi, T. and Nakao, Y. "Effects of Impurity Elements on the Reheat Cracking in the Weld Zones of Steels," 4JWS-II-15, *The Fourth International Symposium of the Japan Welding Society*, November 1982, Osaka.

98. Horn, H. and Kunze, H. D. "Reheat Cracking Studies by Weld Simulation and Scanning Auger Investigation," 4JWS-II-14, *The 4th International Symposium of the Japan Welding Society*, November 1982, Osaka, pp. 449-454.

99. Yu, J. and McMahon, C. J. Jr. "The Effect of Composition and Carbide Precipitation on Temper Embrittlement of 2.25Cr-1Mo Steel: Part I. Effects of P and Sn," *Metallurgical Transactions A*, Vol. 11A(2), February 1980, pp. 277-289.

100. Eyre, B. L., Edwards, B. C., and Titchmarsh, J. M. "Physical Metallurgy of Reversible Temper Embrittlement," *Proceedings of International Conference on Advances in the Physical Metallurgy and Application of Steel*, University of Liverpool, 21-24 September, 1981, The Metals Society, London, England, pp. 246-258.

101. Hipsley, C. A. "Precipitation Sequences in Heat-Affected Zone of 2.25Cr-1Mo Steel During Stress-Relief Heat Treatment," *Metal Science*, Vol. 15(4), April 1981, pp. 137-147.

102. Muhammed, A. B., Szkopiak, Z. C., and Waldron, M. B. "Effect of Types of Carbides on Temper Embrittlement in Commercial 2.25Cr-1Mo Steel," *Proceedings of International Conference on Advances in the Physical Metallurgy and Applications of Steels*, University of Liverpool, 21-24 September, 1981, Metals Society, London, England, pp. 340-348.

103. Wilkinson, D. S., Abiko, K., Thyagarajan, N., and Pope, D. P. "Compositional Effects on the Creep Ductility of a Low Alloy Steel," *Metallurgical Transactions A*, Vol. 11A, November 1980, pp. 1827-1836.

104. Wada, M., Fukase, S., and Nishikawa, O. "Role of Carbides in the Grain Boundary Segregation of Phosphorus in a 2.25Cr-1Mo Steel," *Scripta Metallurgica*, Vol. 16(12), December 1982, pp. 1373-1378.

105. Viswanathan, R. and Joshi, A. "Effect of Microstructure on the Temper Embrittlement of Cr-Mo-V Steels," *Metallurgical Transactions A*, Vol. 6A(12), December 1975, pp. 2289-2297.

106. Tamaki, K. and Suzuki, J. "Recent Studies on Reheat Cracking of Cr-Mo Steels," *Proceedings of International Conference on Stress Relieving*