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# The Mechanism of Stress Relief Cracking in 2¼Cr-1Mo Steel

## Correlation exists between embrittlement of quenched base metal tempered at low temperatures and stress relief cracking, which is shown to be closely related to the formation of coherent $Mo_2C$ during stress relieving

BY ROBERT A. SWIFT

ABSTRACT. A study of the mechanism of stress relief cracking in quenched and tempered 2¼ Cr-1 Mo steel has shown the phenomenon to be closely related to precipitates that form during stress relieving. The result is that the cracks initiate during heating to the stress relieving temperature. The results of welding tests have been correlated to the Charpy V-notch toughness of the base metal. Modifications in the postweld heat treating cycle have substantially reduced cracking.

### Introduction

Stress relief embrittlement and stress relief cracking have been the subject of intensive study for the past several years. While not strictly correct, the terms are often used interchangeably. Stress relief embrittlement is a loss in notch toughness within the heat-affected zone and/or the weld metal as a result of stress relieving. Stress relief cracking is the term used to describe the intergranular cracks that develop within the weld zone during stress relieving. It had been presumed that both phenomena occur by the same or similar mechanism and recent data verifies this.[1-3] In addition, a tendency for creep embrittlement and notch weakening during short time elevated temperature stress relaxation tests usually occurs in these same materials. These phenomena are characterized by low ductility fractures along prior austenitic grain boundaries.

Stress relief cracking has been a problem ever since the welding of low alloy steels was first attempted. Creep embrittlement became evident when these low alloy steels were tested for extended periods of time at elevated temperatures. Similarities between these phenomena led investigators to the belief that the identification of the mechanism for one form of embrittlement would aid in the identification of the mechanism for the other form.

One such coordinated study was conducted by Steiner *et al.*[1] They investigated stress relief cracking and the creep rupture properties of several low alloy steels. Creep-rupture tests were performed on both base metal and welded composites. The configuration of the welded creep-rupture specimens was such that the fractures initiated in the coarse grained heat-affected zone and were accompanied by low creep ductility. Also, materials that were susceptible to stress relief cracking were found to exhibit creep embrittlement of the unaffected base metal.

Murray[2] found that materials exhibiting notch sensitivity during short term elevated temperature stress relaxation were also prone to stress relief cracking. His analysis of the relationship is based upon the yield strength of the materials. The high yield strength materials exhibited a more acute notch sensitivity during stress relaxation than did low yield strength materials. Therefore, Murray reasoned that, since a high yield strength material can store higher residual stresses than a low yield strength material, intergranular cracking results when the stress-temperature conditions within the weld zone are such so as to approximate those that induce notch sensitivity.

The role of precipitates in stress relief embrittlement has been studied by Meitzner and Pense.[3] A finely dispersed intergranular precipitate strengthens the matrix, thereby transferring the creep strain to the grain boundaries. An intergranular precipitate restricts grain boundary sliding and migration so that intergranular voids form, resulting in a loss in creep ductility, increased notch sensitivity and finally cracking.

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This present work describes the results of a study of the mechanism of stress relief cracking in quenched and tempered $2^1/_4$Cr-1Mo steel. Experiments were conducted to define the crack initiation temperature (CIT) and the effect of major welding variables on this temperature. Modifications in the post-weld stress relieving parameters were made in an attempt to reduce the propensity to crack. The results of these tests are correlated with the results of Charpy V-notch impact tests of $2^1/_4$Cr-1Mo steel, water quenched and then tempered at temperatures ranging from 300 to 1300°F. Finally, the mechanism is explained by means of a metallographic study.

**Table 1—Chemical Analysis, %**

| C | Cr | Mo | Mn | Si | Ni | Cu | Al | P | S |
|---|---|---|---|---|---|---|---|---|---|
| .11 | 2.25 | .96 | .51 | .22 | .11 | .17 | .013 | .011 | .022 |

**Table 2—Mechanical Properties**

| Heat treatment[a] | Orientation | .2% YS,[b] ksi | UTS,[b] ksi | RA, %[b] | E, %[b] | Charpy V-notch transition temperature FATT,[c] °F | Charpy V-notch transition temperature 40 ft-lb, °F |
|---|---|---|---|---|---|---|---|
| 1750° F/1 hr-WQ 1250° F/1 hr-AC | Longitudinal | 105.6 | 119.9 | 72.6 | 22.0 | ND[d] | ND[d] |
| 1750° F/1 hr-WQ 1225° F/1 hr-AC | Longitudinal | 98.7 | 113.5 | 73.1 | 22.0 | —75 | —110 |
| 1225° F/1 hr-AC | Transverse | 96.4 | 110.5 | 62.3 | 18.0 | —50 | — 60 |

[a] WQ—water-quenched; AC—air-cooled.
[b] .2% YS—0.2% yield strength; UTS—ultimate tensile strength; RA—reduction in area; E—elongation in 2 in.
[c] Fracture appearance transition temperature.
[d] ND—Not determined.

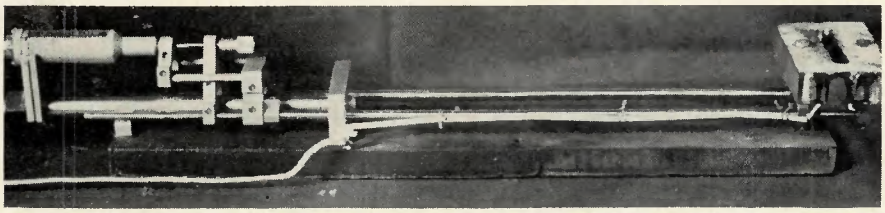

Fig. 1—Experimental set-up showing extensometer and modified Lehigh restraint specimen

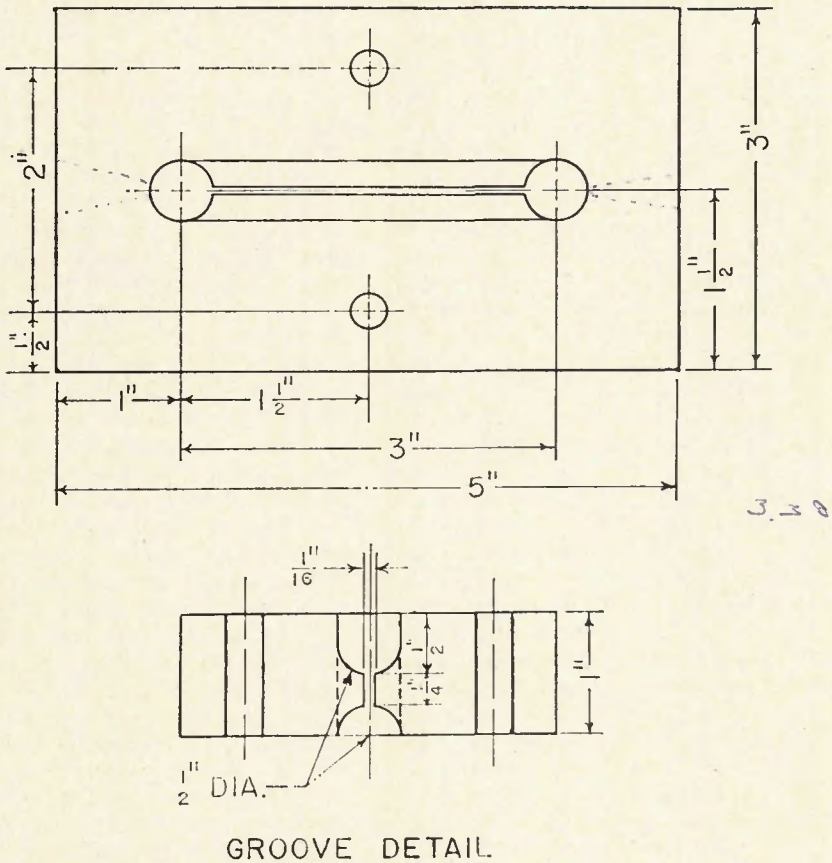


Fig. 2—Modified Lehigh restraint specimen

## Experimental

### Materials and Heat Treatment

The material used for this study was from a 360 x 60 x 1 in. production plate. The chemical analysis is listed in Table 1. Crack initiation tests were performed on material mill quenched and tempered to ASTM A542 Class 2. All other tests were performed on material heat treated in the laboratory to ASTM A542 Class 1. The heat treatments, tensile and Charpy V-notch impact properties of the base metal are detailed in Table 2.

### Specimen and Data Collection

Crack initiation was determined by monitoring the width of the weld in a modified Lehigh restraint specimen. The test set-up is shown in Fig. 1 and the specimen is illustrated in Fig. 2. The Lehigh restraint test was selected because of the high degree of restraint which it develops. In this test, restraint is more severe than that occurring normally in fabrication weldments, and accordingly, it will ensure cracking if a susceptibility exists.

The output from an extensometer mounted across the weld and the temperature at the root of the weld were recorded on a strip chart recorder. Figure 3 is a schematic diagram of a typical data chart. The contraction upon welding is attributed to residual stresses from the welding operation. Thermal inertia of the test piece and differences in thermal expansion between the extensometer arms and specimens account for the apparent contractions when the temperature is increased. Once the temperature is stabilized, no dimensional changes occur unless a crack develops. The relatively rapid expansion during heating indicates crack initiation and propagation.

### Welding Conditions

The effect of welding variables on the crack initiation temperature was examined during the first phase of the study. The variables were welding current, welding speed, and heating rate to the stress relieving temperature. The heat inputs developed ranged

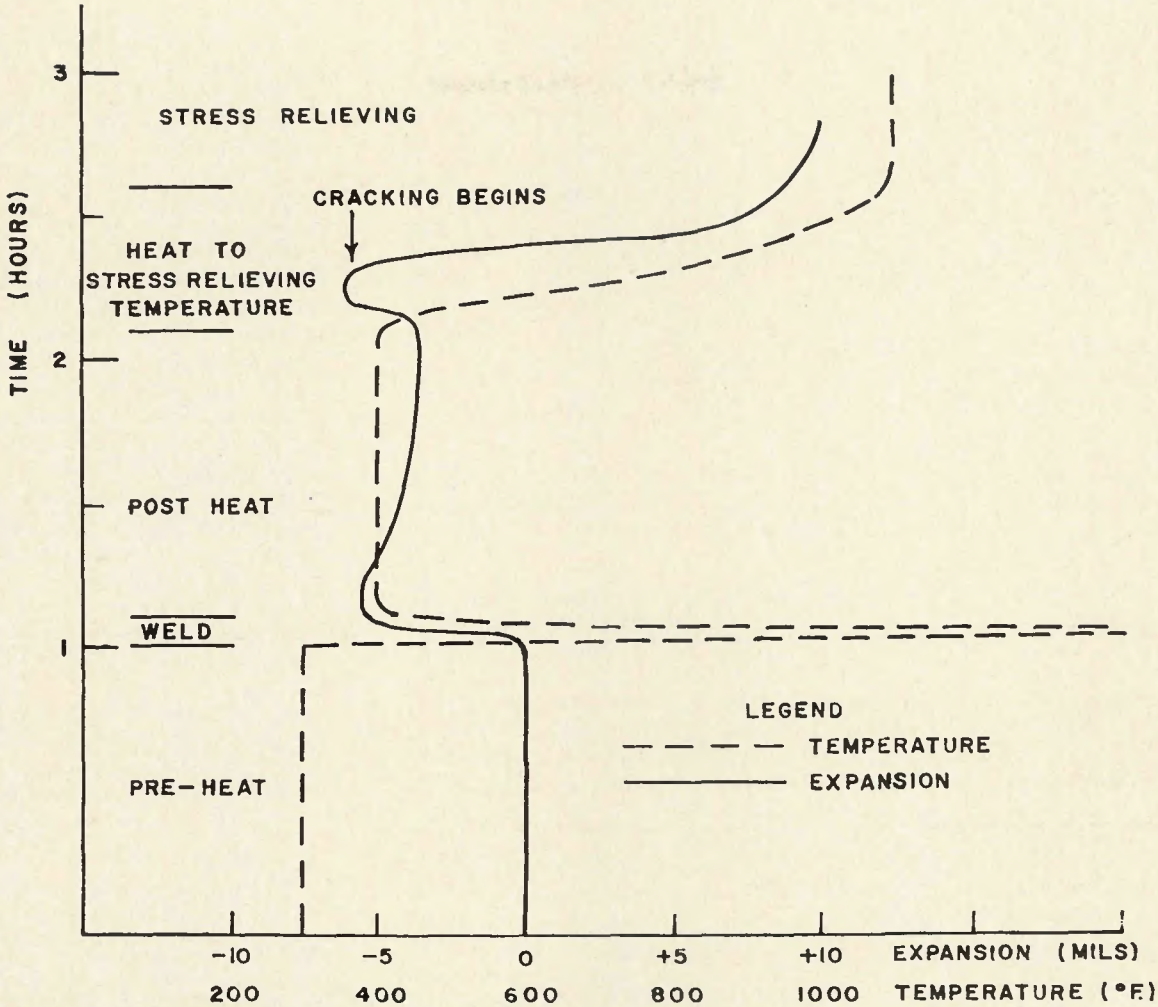


Fig. 3—Schematic expansion and temperature vs. time data chart

from 13.8 to 40.5 kilojoules/in. The range of each variable is listed in Table 3. All specimens were preheated at 300° F for 1 hr prior to welding and postweld soaked at 400°F for 1 hr prior to stress relieving. Welding was done manually with $^1/_8$ in. diameter E9018-B3L electrodes.

### Effects of Heat Treatment on Notch Toughness

To assist in explaining the phenomena of stress relief cracking and aid in identifying the mechanism, a series of Charpy V-notch impact tests were conducted. Plate material was austenitized at 1750 and 2300° F followed by a water quenching. Charpy V-notch blanks were then tempered for 1 hr at temperatures ranging from 300 to 1300° F. Based upon the room temperature impact properties of this material, full fracture appearance transition curves were obtained for tempering temperatures of 700 and 1300° F. In addition, a full transition curve was obtained for material as-quenched from 2300° F.

### Metallography

After stress relieving, all welded specimens were sectioned transverse to the weld and examined with a light microscope. The specimens were sectioned at the center of the weld and at $^1/_2$ in. intervals to each end of the weld bead. Thin foils and surface replicas of selected Charpy V-notch specimens were examined to assist in identifying the embrittling mechanism.

## Results

The results of the crack initiation temperature (CIT) determinations are tabulated in Table 4. The series of welds deposited parallel to the rolling direction, using identical welding conditions, shows the CIT to be a function of heating rate to the stress relieving temperature. As the heating rate was increased from 10.5° F/min. to 24.0°F/min., the CIT increased from 810 to 905° F. The data indicate anisotropy of CIT with respect to rolling direction since the longitudinal welds cracked at higher heat inputs than the transverse welds, i.e., greater susceptibility for the longitudinal orientation. The cracks that developed in the transverse welds are shown in Fig. 4. The welds deposited with high energy inputs (Figs. 4A and 4C) did not crack at a rate sufficiently rapid to be detected by the extensometer. The welds with low energy inputs (Figs. 4B and 4D) cracked upon heating to the stress relieving temperature.

The effect of soaking transverse welds at 750° F prior to stress relieving is seen in Fig. 5. Similar results were obtained for the longitudinal welds. There was a marked improvement in cracking resistance when a soaking time of 2 or 4 hr was used. However, severe cracking resulted when the specimens were soaked for 7 hr at 750° F (Fig. 5C). The welding conditions were selected based upon the experience gained in the CIT determination. This procedure was known to produce cracks in speci-

**Table 3—Welding Parameters**

| | |
|---|---|
| Welding current, amp. | 125, 150 |
| Welding speed, ipm | 6, 12 |
| Stress-relieving temperature, °F | 1100, 1200 |
| Heating rate to stress relieving temperature, °F/min. | 10.5, 22.2, 24.0 |

**Table 4—Crack Initiation Data**[a]

| Orientation of weld bead with respect to final rolling direction | Crack initiation temperature (CIT), °F | Welding current, amp | Welding speed, ipm | Heat input, kilojoules/in. | Heating rate to stress-relieving temperature, °F/min | Stress relieving temperature, °F |
|---|---|---|---|---|---|---|
| Parallel | 905 | 125 | 6 | 27.5 | 24.0 | 1200 |
| Parallel | 890 | 125 | 6 | 27.5 | 22.2 | 1100 |
| Parallel | 810 | 125 | 6 | 27.5 | 10.5 | 1200 |
| Transverse | — | 150 | 6 | 40.5 | 22.2 | 1100 |
| Transverse | — | 125 | 6 | 27.5 | 10.5 | 1200 |
| Transverse | 800 | 150 | 12 | 20.3 | 22.2 | 1100 |
| Transverse | 1015 | 125 | 12 | 13.8 | 24.0 | 1200 |

Preheat: 300° F—1 hr.
Postheat: 400° F—1 hr.
Electrode: ⅛″ dia. E9018-B3L
Length of weld: 2″

[a] Preheat—300° F, 1 hr; postheat—400° F, 1 hr; electrode—⅛ in. diameter E9018-B3L; length of weld—2 in.

mens heated directly to the stress relieving temperature.

The results of the welding study, both CIT determination and the effect of low temperature isothermal soaking on cracking susceptibility, indicated the existence of a low temperature form of embrittlement. To ascertain this, a series of Charpy V-notch impact tests were performed as outlined previously. The results of these tests are summarized in Table 5. The series of tests on material water quenched from 2300° F shows that there is a loss in notched toughness as measured by the fracture appearance transition temperature (FATT) for the material tempered at 720° F. However, the upper shelf energy increases with increasing tempering temperature. Also, the transition temperatures for comparable tempering treatments are higher for the course grained material than for the fine grained material.

The mechanism of embrittlement was studied by electron microscopy of surface replicas and thin foils of selected Charpy V-notch impact specimens. Figure 6 contains typical structures of material water quenched from 1750° F and tempered. The as-quenched condition (Fig. 6A) shows a sharply defined acicular structure with fine precipitates in the matrix and a prior austenitic grain boundary that is relatively free of carbide precipitates. The precipitates within the matrix could not be identified by selected area electron diffraction (SAED). Upon tempering at 734° F, the structure shown in Fig. 6B is obtained. The coarse precipitates are FeC, but the fine preciptates could not be identified. Tempering at 1157° F results in the structure in Fig. 6C, which consists of fine precipitates that are probably hexagonal MoC and coarse precipitates that are $Fe_3C$.

Typical thin foil structures are shown in Fig. 7. The as-quenched material (Fig. 7A) has a well defined lath structure with a relatively high dislocation density. Tempering at 700° F causes strain markings to appear within the laths as shown in Fig. 7B. The fine structure within the laths appears to be due to lattice straining. The precipitate within these regions could not be positively identified although several diffraction spots which could be attributed to face centered cubic $Mo_2C$ were obtained. The structure in Fig. 7C is from a specimen tempered at 1085° F for 1 hr. Fine acicular precipitates appear within the laths and a cell structure, indicative of recovery, has formed.

The large dark area (A in Fig. 7C) was analyzed by SAED. The diffraction pattern is shown in Fig. 8. The particle is oriented with (001) parallel to the incident beam and has a typical face centered cubic structure. The calculated lattice constant is 4.21Å.

## Discussion

The mechanism of cracking can be explained by considering the process of stress relieving. The coarse grained heat-affected zone is essentially an as-quenched structure. Thus it will have a high concentration of quenched-in vacancies. The initial stages of stress relief occur through vacancy annihilation, which begins as soon as the temperature is raised above ambient. Annihilation of excess vacancies is the

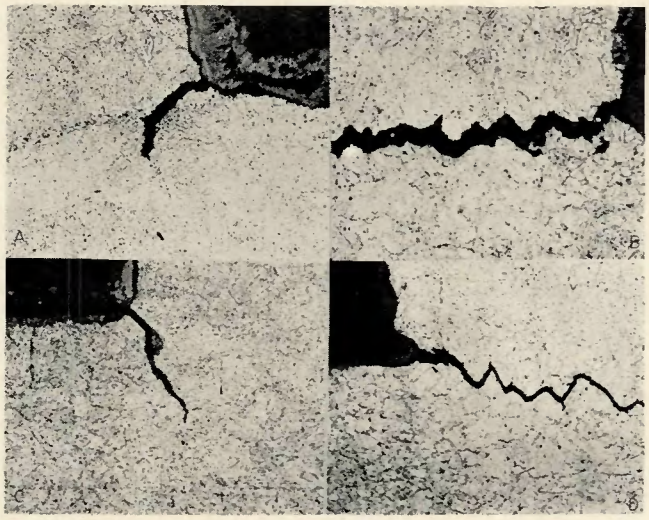


Fig. 4—Stress relief cracks in welds transverse to final rolling direction. A (top left)—125 amp, 6 ipm welding speed, 27.5 kilojoules/in.; B (top right)—125 amp, 12 ipm welding speed, 13.8 kilojoules/in.; C (bottom left)—150 amp, 6 ipm welding speed, 40.5 kilojoules/in.; D (bottom right)—150 amp, 12 ipm welding speed, 20.3 kilojoules/in. Nital-picral etch. X250 (reduced 44% on reproduction)

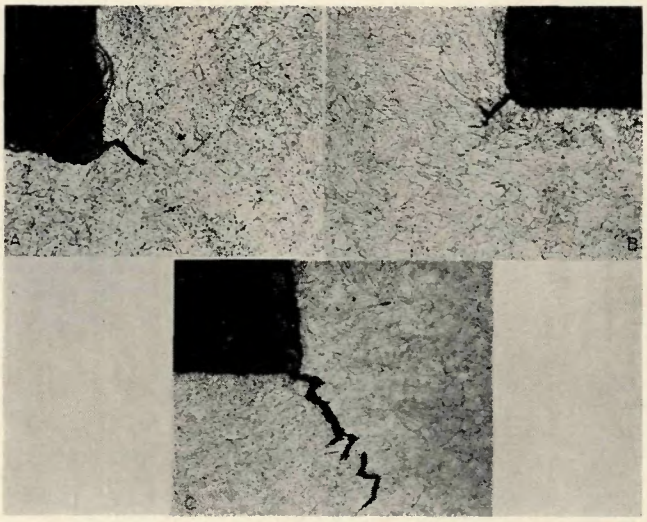


Fig. 5—Effect of soaking time at 750° F on stress relief cracks in welds transverse to final rolling direction. Preheat —300° F/hr; 150 amp, 12 ipm welding speed, 20.3 kilojoules/in. A (top left)—2 hr; B (top right)—4 hr; C (bottom)—7 hr. Nital-picral etch. X250 (reduced 44% on reproduction)

controlling mechanism of stress relieving until the temperature is sufficiently high to permit dislocation climb.[4] The period of most rapid stress relaxation is during the initial stages of stress relieving. The time requirement is a function of both composition and degree of restraint. For example, the period of most rapid relaxation in a C-Mn steel is approximately 2 hr regardless of the temperature providing that the conditions of restraint are constant.[5]

During the heating of $2^1/_4$Cr-1Mo steel to the stress relieving temperature a sequence of carbide precipitations can occur that may counterbalance stress relaxation. Precipitates identified as FeC, when tempering is performed at 700°F, change to $Fe_3C$ when the material is tempered at 1085° F. $Mo_2C$ forms within this same temperature range.[6]

The precipitate whose diffraction pattern is shown in Fig. 8 is probably $Mo_2C$. Nagakura and Oketani[7] report the existence of an $MoC_x$ (x = 0.54) with a NaCl structure and an $a_o$ of 4.22Å. However, it is reported to be unstable below 2200° C (3992° F). The pattern in Fig. 8 is a simple face centered cubic structure and matches the ASTM pattern for $Mo_2C$ (Table 6). The intensities for the (200) and (220) reflections are almost equal in $Mo_2C$ and the intensities in Fig. 8 are similar. Therefore, the precipitate in Fig. 7C is probably face centered cubic $Mo_2C$.

When it first forms, $Mo_2C$ is coherent with ferrite. The $d_{(200)}$ (Fe) is 1.435Å. The fine structure observed in Fig. 7B could be the result of lattice straining since the mismatch between precipitate and matrix is 3.9%. A coherent precipitate will impede dislocation climb and cross slip so that stresses cannot be relaxed during low temperature stress relieving. At higher temperatures (1100° F or higher), much of the coherency is lost and recovery by dislocation motion can occur.

Using this as background, the mechanism for crack initiation may be rationalized. During heating to the soaking temperature, some precipitation will occur because of the increasing thermal energy. In addition, there will also be stress relaxation due to vacancy annihilation. Slow heating rates will allow for diffusion of Mo through the matrix to the coherent nuclei due to the duration of the heating cycle. At the same time, vacancy annihila-

**Table 5—Charpy V-Notch Impact Data**

| Heat treatment[a] | ASTM grain size | Upper shelf energy, ft-lb | Fracture appearance transition temperature (FATT), °F |
|---|---|---|---|
| 1750° F/1 hr-WQ<br>700° F/1 hr-WQ | 6–7 | 45 | +165 |
| 1750° F/1 hr-WQ<br>1300° F/1 hr-WQ | 6–7 | 140 | − 80 |
| 2300° F/1 hr-WQ | 2–3 | 30 | +100 |
| 2300° F/1 hr-WQ<br>720° F/1 hr-WQ | 2–3 | 45 | +180 |
| 2300° F/1 hr-WQ<br>1310° F/1 hr-WQ | 2–3 | 200 | − 20 |

[a] WQ—water-quenched.

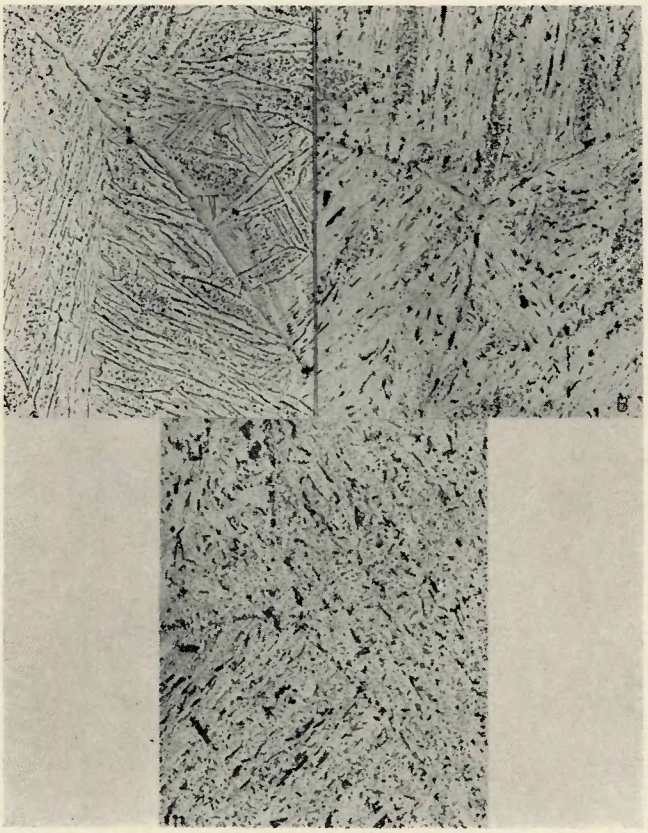

Fig. 6—Electron micrographs of 2¼Cr-1Mo steel water quenched from 1750° F and tempered as indicated. Two stage plastic replica. A (top left)—as-quenched; B (top photo)—1 hr at 734° F; C (bottom)—1 hr at 1175° F. Nital-picral etch. X4.2K (reduced 38% on reproduction)

Fig. 7—Thin foil transmission micrographs of 2¼Cr-1Mo steel water quenched from 1750° F and tempered as indicated. A (top left)—as-quenched; B (top right)—1 hr at 700° F; C (bottom)—1 hr at 1085° F. X16K (reduced 38% on reproduction)

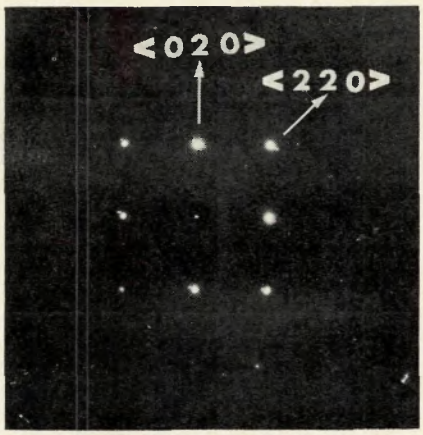


Fig. 8—Selected area electron diffraction pattern from region A in Fig. 7C. Incident beam is parallel to (001). $a_o = 4.21A$

tion will occur and stresses will be relaxed. However, the dissolving of FeC and precipitation of $Fe_3C$ will lock dislocations,[9] prevent their movement and thereby impede stress relaxation. At temperatures between 700 and 800° F, $Mo_2C$ begins to form stable nuclei causing lattice straining. Since stress relaxation is impeded, the additional stresses set up by the coherency strains increase the residual strains. The stresses must be relaxed but, since the matrix cannot plastically deform, an intergranular crack forms. At faster heating rates, the rate of vacancy annihilation is more rapid and the total mass of Mo diffused through the lattice is reduced. Therefore, coherency straining will occur at higher temperatures and consequently, cracking will occur at higher temperatures.

Stress relief cracks do not develop as readily when high heat inputs are used because of the concomitant slower cooling rates. When the cooling rate after welding is reduced, an "auto stress relief" can occur, thereby reducing the residual stresses than must be subsequently relaxed.

Isothermal soaking at 750° F affects the susceptibility to stress relief crack by the following mechanism. Stress relaxation at 750° F is primarily by vacancy annihilation. Coincident with this, is the formation of coherent nuclei of $Mo_2C$. However, stress relaxation is initially more rapid than that nucleation of $Mo_2C$ so that there is a net reduction in the internal stresses and no cracks develop. Longer intervals (greater than 4 hr) do not result in an appreciable increase in stress relaxation, but there is an increase in the internal stresses due to the lattice strains set up by the coherent nuclei of $Mo_2C$. Now, there is a net increase in internal stresses and the matrix cannot plastically deform to relief the stresses. An intergranular crack then forms during a subsequent stress relief treatment.

## Conclusions

Stress relief cracking of highly restrained welds in quenched and tempered $2^1/_4$Cr-1Mo steel has been found to be initiated during heating to the stress relieving temperature. The temperature of crack initiation is a function of the heating rate to the stress relieving temperature and is higher at faster heating rates. The severity of stress relief cracking controlled by the heat input during welding. As the heat input increases, up to 40 kilojoules/in., the severity of cracking decreases.

There is a close correlation between embrittlement of quenched base metal when tempered at low temperatures and stress relief cracking. The mechanism that operates on the low tempering temperature embrittlement can describe the mechanism for stress relief cracking. Apparently, dislocation locking and the formation of coherent precipitates, both of which impede stress relieving, result in cracking along prior austenitic grain boundaries within the coarse grained heat-affected zone.

**Table 6—*d*-Spacings of $Mo_2C$ and Unknown Precipitate**

| hkl | ASTM 15-457 $a_o = 4.14$ | $I/I_o$ | Fig. 8 $a_o = 4.21$ |
|---|---|---|---|
| 111 | 2.38 | 100 | (2.43)[a] |
| 200 | 2.07 | 80 | 2.11 |
| 220 | 1.47 | 70 | 1.49 |
| 311 | 1.255 | 70 | (1.27)[a] |
| 222 | 1.20 | 40 | (1.22)[a] |

[a] Not present due to orientation.

*Acknowledgment*

The author would like to acknowledge J. E. Boyer, D. R. Metzgar and J. R. Lohr for assistance in carrying out the experiments.

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## *Invitation to Authors . . . Coming in June . . .*

The 53rd Annual Meeting of the AMERICAN WELDING SOCIETY will be held in Detroit, Michigan, during April 10-14, 1972. In anticipation of this meeting, the June 1971 issue of the Welding Journal will contain an invitation to authors to participate in this event. Included with the invitation will be detachable application form which authors can submit with the 500-word abstracts of papers which they propose to present. The deadline for submitting application forms and attached 500-word abstracts is expected to be September 15, 1971.