EVALUATION OF SUSCEPTIBILITY OF WELDED JOINTS OF HEAT-RESISTANT CHROMIUM MARTENSITIC STEEL TO CRACKING AT HEAT TREATMENT

V.Yu. SKULSKY, A.K. TSARYUK and **S.I. MORAVETSKY**
E.O. Paton Electric Welding Institute, NASU, Kiev, Ukraine

The mechanisms of cracking of welded joints in tempering used to relieve stresses are considered. It has been found that welded joints on steel 10Kh9MFB with a homogeneous martensitic structure are insensitive to temper cracking. Formation of δ-ferrite in the martensitic structure may lead to cracking of the weld metal. Cracks form in tempering in the range of about 450--550 °C as a result of concentrated deformation within the zone of soft ferrite interlayers in development of secondary hardening of the martensitic matrix. A probable cause of hardening is precipitation of chromium carbide M_7C_3.

Keywords: *arc welding, martensitic steel, welded joints, heat treatment, dispersion hardening, soft interlayers, temper cracking*

Manufacture of welded structures from hardening heat-resistant and high-temperature steels is related to the need to perform heat treatment of welded joints for tempering of quenching structures and lowering of the residual stress level. In some cases temper cracks can form in the welded joints during heating or soaking in certain temperature intervals. The risk of cracking increases at treatment of rigid joints, as well as in the presence of design stress raisers, lacks-of penetration, undercuts and extended inner defects in them.

Temper cracks (or reheating cracks) are defects forming as a result of a non-uniform plastic deformation under the conditions of high-temperature relaxation of inner stresses [1]. The non-uniform nature of relaxation creep of metal at tempering can be related to chemical microinhomogeneity (which is characteristic of weld metal) and development of dispersion hardening of grain bodies at certain temperatures as a result of precipitation of finely dispersed phases, namely carbides, intermetallics. Grain strengthening due to secondary hardening is a factor of «relative softening» of grain boundary regions. As a result, the deformation at relieving of inner stresses is concentrated in the grain boundary zone. A fast increase of density of crystalline structure defects at local deformation, as well as formation of interatomic discontinuities under the impact of embrittling impurities, leads to initiation of microdamage in the form of initial pores [2--4] and to crack propagation. A feature of temper cracks is their intergranular nature.

A susceptibility to hardening and, therefore, to formation of cracks at tempering is found in steels containing strong carbide-forming elements (titanium, vanadium, niobium) and elements strengthening the solid solution (molybdenum, chromium, which also belong to carbide-forming elements) [1, 5--9]. Depending on the alloying system, strengthening can be induced in structural and heat-resistant steels by Cr_7C_3, Mo_2C, V_3C_4 carbides, in austenitic steels ---- by NbC, TiC carbides, in nickel-based alloys ---- by intermetallics of $Ni_3(Al, Ti)$ type [1, 3, 7, 10, 11]. Lowering of high-temperature ductility in the boundary zone and crack formation are caused by impurities of phosphorus, arsenic, antimony, tin and sulphur [1, 5, 12--16]. According to the data of [17], the embrittling action of such impurities as phosphorus and sulphur is due to weakening of the bonds between the metal atoms as a result of formation of electronic bonds on the levels of *s*-orbitals of metal atoms and *p*-orbitals of impurity atoms. Such elements as silicon, manganese, carbon, aluminium and copper [3, 5, 18] also increase the temper cracking susceptibility. They, however, have an indirect influence on embrittlement, for instance, by enhancing the grain-boundary segregation of phosphorus (silicon, carbon, manganese) [5], or ousting carbon from the zone of their clustering with formation of soft microstructural components (silicon, aluminium).

In welded joints the metal in the near-weld sections is more susceptible to cracking, these sections developing a coarse-grained structure as a result of heating to subsolidus temperatures and a high degree of hardening as a consequence of a more complete dissolution of the carbide precipitates and saturation of γ-solid solution by carbon and carbide-forming elements. In the welds cracks can form predominantly in the microsections, in which the solidification boundaries enriched in liquating impurities, coincide with the secondary boundaries ---- the austenite grain boundaries.

There is a sufficient number of publications devoted to studying the problem of temper brittleness of low-alloyed pearlitic and bainitic heat-resistant steels with up to 2--5 % Cr [2, 4--6]. Introduction of new complex-alloyed martensitic steels with increased chromium content leads to the need to study the properties of their welded joints, including temper crack sensitivity. Possible predisposition of such steels to development of processes usually accompanied by cracking, is associated with the presence of carbide-

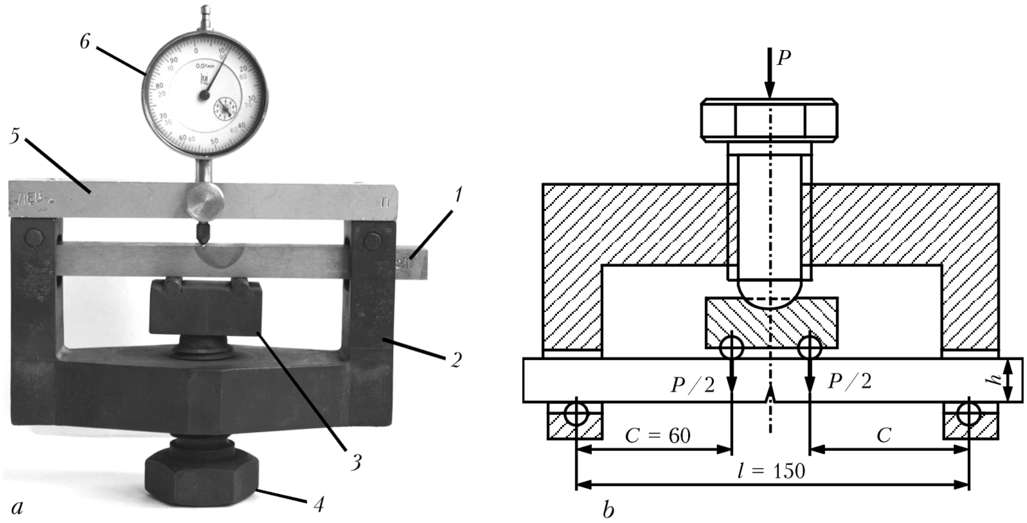

Figure 1. Appearance (*a*) (*1* — sample; *2* — welded bed; *3* — support; *4* — loading screw; *5* — removable plate for indicator fastening; *6* — indicator) and schematic for sample loading (*b*)

forming elements in their composition, as well as with martensite, the non-uniform decomposition of which at tempering (develops quickly at grain boundaries [19]) can be accompanied by non-uniform distribution of deformations and increase of the density of defects on grain boundaries.

The purpose of this work consisted in evaluation of the susceptibility of welded joints of martensitic steel with 9 % Cr to development of temper cracks.

Relaxation testing was conducted by the procedure of the I.I. Polzunov Central Institute for Boilers and Turbines [20], based on performance of heat treatment of welded joint samples loaded in a fixture of a high-temperature nickel alloy up to a certain stress level by four-point bending (Figure 1). Deflection *f* in the zone of a uniform bending moment (between the support contact points) was monitored by a removable indicator. Stresses *s* in the stretched zone on the surface of samples of a rectangular section were calculated using the following relationships:

$$\sigma = M/W, \quad f = \frac{M}{24EJ}(3l^2 - 4c^2),$$

where $M = (P/2)c$ is the bending moment, N·m; $W = bh^2/6$ is the axial moment of resistance of sample section, m^3; *b*, *h* is the section width and height; $J = bh^3/12$ is the axial moment of inertia of sample section, m^4; *E* is the modulus of elasticity equal to $218\cdot10^3$ MPa for steel.

Samples fastened in the fixture were placed into a cold furnace, heated up to the required temperature, soaked for the specified time, and then cooled in air. After heat treatment the visual and metallographic inspection of welded joints was conducted to find defects in them. Samples for testing were cut out across the weld made when filling the groove in the plate from 10Kh9MFB steel using automatic submerged-arc welding. A V-shaped stress raiser was made in the fusion zone in the samples. Welding was performed using test flux and flux-cored wire providing the following composition of the deposited metal, wt.%:

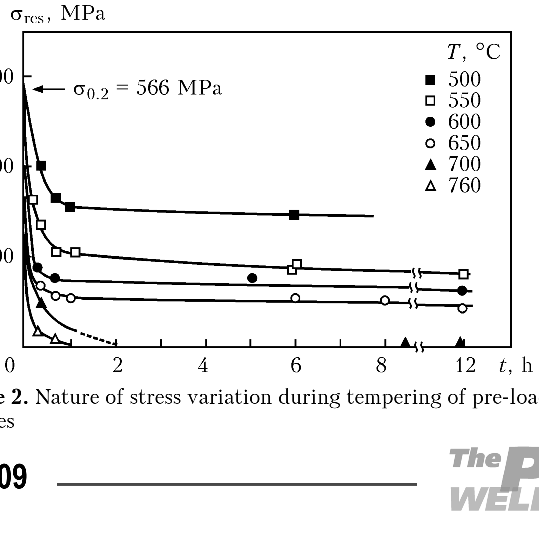

Figure 2. Nature of stress variation during tempering of pre-loaded samples

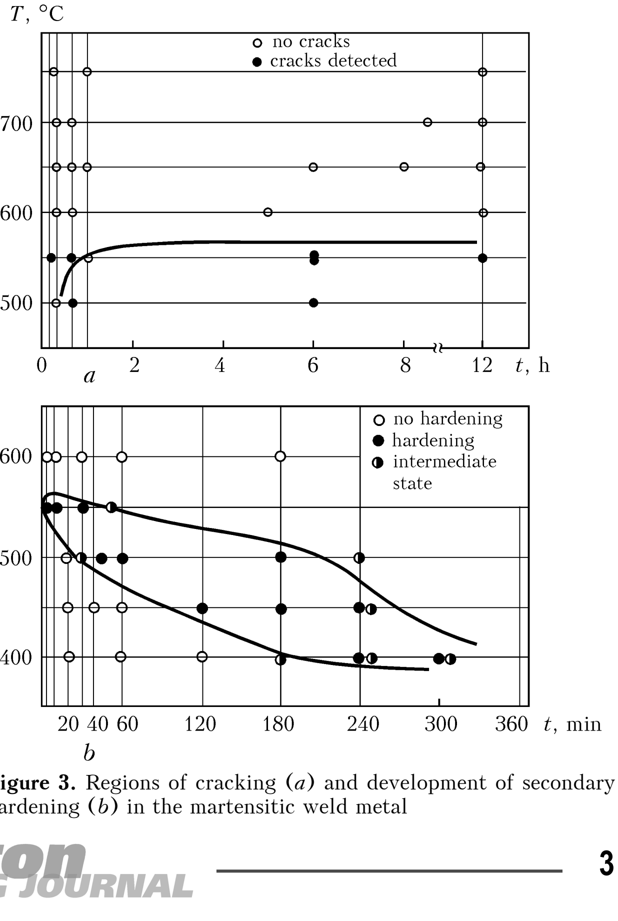

Figure 3. Regions of cracking (*a*) and development of secondary hardening (*b*) in the martensitic weld metal

0.098 C; 0.72 Mn; 0.13 Si; 1.0 Mo; 0.52 Ni; 0.27 V; 0.053 Nb; 0.048 N; 0.01 Al; 0.014 S; 0.015 P, similar to the steel to be welded. Welding was performed in the modes with the heat input of 16 and 23 kJ/cm.

Metallographic investigations showed that in as-welded condition the metal of the welds and HAZ had a predominantly quenched martensitic structure. However, in welding at higher currents (with increased heat input) overheating of weld pool melt and slowing down of the solidification rate lead to formation of a coarser crystalline structure, in which appearance of a small fraction of δ-ferrite is noted as a result of development of chemical inhomogeneity.

As follows from the test results (Figure 2), relaxation creep develops quite intensively in the first 40--50 min of tempering, which is indicated by the rate of lowering of stress in the loaded samples. The residual stress level determined by metal softening rate and its deformability, decreases with temperature increase. Complete removal of stresses during short soaking (from 2 to 1 h) is achieved at the temperature above 600 °C. It is established that under these conditions the metal with a purely martensitic structure features a high cracking resistance. No cracks were found in the base metal in the overheated section in the stress raiser vicinity. On the other hand, in the weld metal containing δ-ferrite, cracking was found at tempering at temperatures of 500 to 550 °C (Figure 3, *a*). No cracks were formed at temperatures above 550 °C.

In parallel with relaxation testing heat treatment of welded joint samples was conducted in a free (unloaded) state with subsequent measurement of hardness in the weld metal. It is established that in the temperature range of ~400--550 °C (Figure 3, *b*) metal of 10Kh9MFB type develops secondary hardening. Taking into account the data of [21], it can be assumed that the factor causing hardening in the weld metal

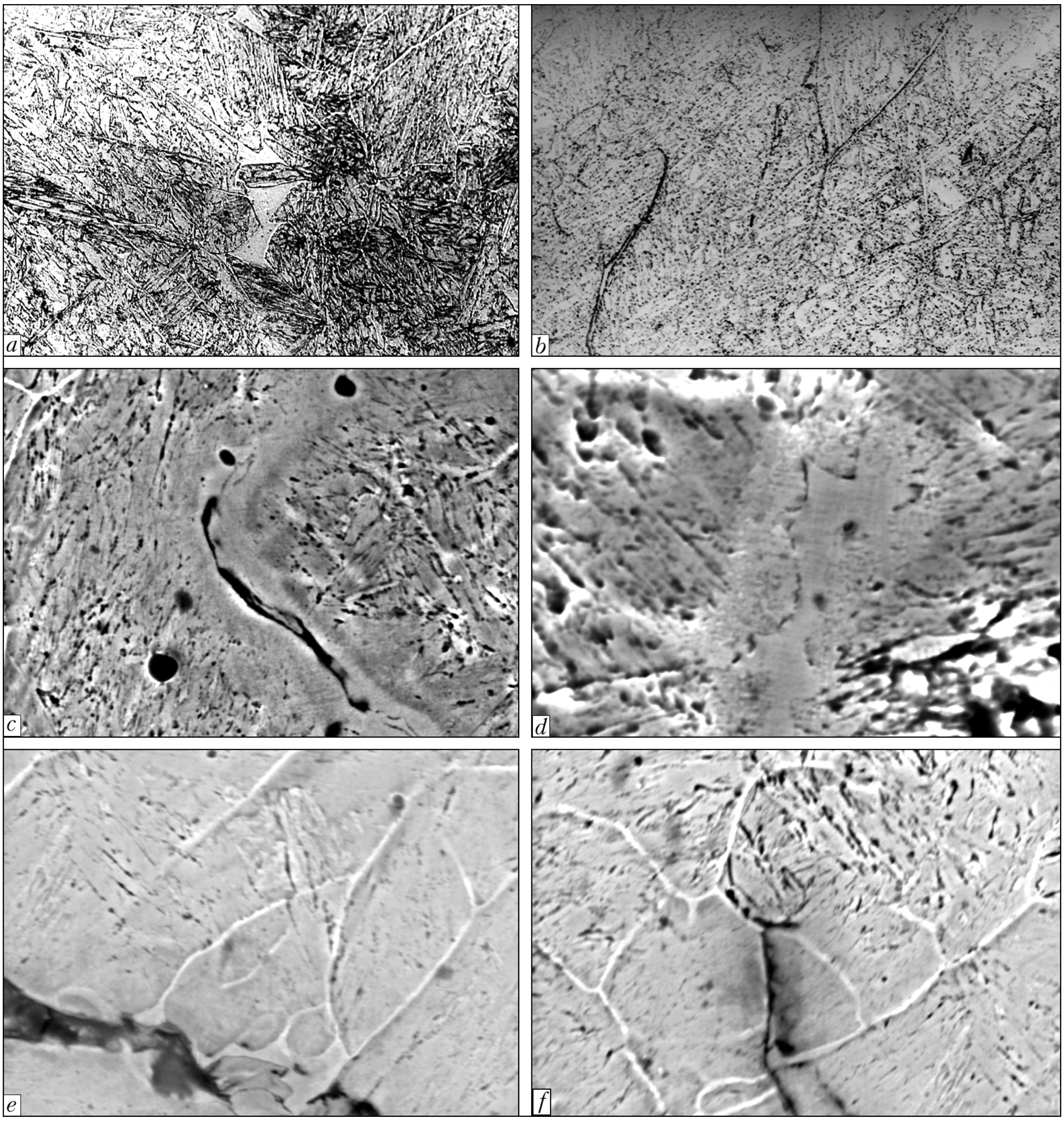

Figure 4. Microstructure of martensitic welds with δ-ferrite (*a*, *b*) and cracking in welds formed at relaxation testing (*c*–*f*): *a*, *b* — ×500; *c*, *e* — ×2020; *d* — ×4780; *f* — ×1690

of this type, is formation of alloyed carbide M_7C_3 ((Fe, Cr)$_7C_3$ type). This carbide can form in high-chromium steels both as a result of substitution of cementite ($Fe_3C \rightarrow$ (Fe, Cr)$_7C_3$), and irrespective of cementite in the form of additional carbide.

Considered cracking in the welds is the consequence of deformation concentration in the zone of soft interlayers of δ-ferrite located in a more rigid (insufficiently tempered) martensite metal, further strengthened by chromium carbide precipitates. δ-ferrite proper features a high ductility and no cracks form in it. However, plastic flow of metal in δ-ferrite area leads to a higher density of dislocations along the boundary with the stronger martensite. As a result, cracks develop in the section of δ-ferrite--martensite transition, or along the thin solidification interlayers, which are a continuation of coarser sections of δ-ferrite (Figure 4). Interphases are a continuation of secondary boundaries of austenite grains, which could be detected by their decoration by dense carbide precipitates at tempering. Therefore, fracture in welds, similar to HAZ metal, develops along the grain boundaries.

Non-uniform distribution of chemical elements can be regarded as one of the possible metallurgical factors causing lowering of strength and ductility of the interphases. So, for instance, in the cracking zone in the section of δ-ferrite to martensite transition X-ray microprobe analysis recorded an increase of manganese concentration (~1.04--1.06 % at the interphase, 0.6--0.8 % in δ-ferrite zone, ~0.8 % in martensite). δ-ferrite usually contains more ferritizers (for instance, 1.23 % Mo; 10.15 % Cr) than the adjacent martensite (0.7 % Mo; 9.48 % Cr). In addition, δ-ferrite formation is associated with depletion of this phase in carbon, which accumulates on the interphases [22] and in the adjacent matrix microsections. Increased concentration of carbon near δ-ferrite is indicated by a denser arrangement of carbides after tempering, which was noted at metallographic examination of the structure of welds of this type. Carbide formation on the interphases and possible enhancement of impurity segregation (first of all phosphorus [5]) under the influence of local increased concentrations of alloying elements may cause boundary embrittlement. At fracture, carbides are the centers of dislocation accumulation and microcrack initiation [2]. However, on the whole, the mechanism of embrittlement of the studied metal with a multicomponent alloying system (Fe--9Cr--Mo--V--Nb--Ni--N taking into account manganese, silicon, aluminium, sulphur and phosphorus content) is quite complicated and requires a special study. In this case δ-ferrite may be regarded as the main structural factor causing cracking in the welds.

Thus, it is established that dispersion hardening develops in the martensitic metal of 10Kh9MFB type in the temperature range of ~450--550 °C, the probable cause of which is precipitation of Cr_7C_3 carbide. During tempering the welded joints with a martensitic structure do not demonstrate any susceptibility to cracking. However, as shown on the example of welds, cracking is possible only in the presence of soft δ-ferrite interlayers in rigid martensite. In this case cracks form at soaking of the hardened welded joints in the dispersion hardening temperature interval. A condition of high resistance of 10Kh9MFB steel welded joints to temper cracking is ensuring their uniform martensitic structure.

1. Zemzin, V.N., Shron, R.Z. (1978) *Heat treatment and properties of welded joints*. Moscow: Mashinostroenie.
2. Nawrocki, G.J. (2001) Stress-relief cracking of a ferritic alloy steel. *IIW Doc. IX-2001--01*.
3. Boniszewski, T., Eaton, N.F. (1969) Electron fractography of weld-reheat cracking in CrMoV steel. *Metal Sci. J.*, **3**, 103--110.
4. Gotoh, A., Nakagawa, T., Hatano, H. et al. (1999) The effects of vanadium on carbide precipitation in 11/4--3 % Cr--Mo steel weld metals. *IIW Doc. IX-708--99*.
5. Lundin, C.D., Khan, K.K. (1996) Fundamental studies of the metallurgical causes and mitigation of reheat cracking in 11/4 Cr--1/2 Mo and 21/4 Cr--1 Mo steels. *WRC Bulletin*, **409**, 117.
6. Tamaki, K., Suzuki, J., Nakaseko, Y. (1980) Effect of molybdenum carbide on reheat cracking sensitivity of Cr--Mo steels. *IIW Doc. IX-1159--80*.
7. Tamaki, K., Suzuki, J., Tajiri, M. (1984) Effect of vanadium and titanium on reheat cracking sensitivity. *Transact. of JWS*, **154**, 17--24.
8. Tamaki, K., Suzuki, J. (1983) Effect of chromium and molybdenum on reheat cracking sensitivity of steels. *Ibid.*, 14(**2**), 123--127.
9. Tamaki, K., Suzuki, J. (1983) Reheat cracking test on high strength steels by a modified implant test. *Ibid.*, 14(**2**), 117--122.
10. Tamaki, K., Suzuki, J., Nakaseko, Y. et al. (1984) Effect of carbides on reheat cracking sensitivity (Study on reheat cracking of Cr--Mo steels: Report 3). *Ibid.*, 15(**1**), 8--16.
11. Tamaki, K., Suzuki, J., Li, M.-L. (1993) Influence of vanadium carbide on reheat cracking of Cr--Mo steels ---- study on reheat cracking of Cr--Mo steels: Report 10. *Ibid.*, 24(**2**), 87--93.
12. Nawrocki, J.G., Dupont, J.N., Robino, C.V. et al. (2000) The stress-relief cracking susceptibility of a new ferritic steel. Pt 1: Single-pass heat affected zone simulation. *Welding J.*, 79(**12**), 355--362.
13. Tamaki, K., Suzuki, J., Tate, H. (1988) Combined influence of sulfur and manganese on reheat cracking of Cr--Mo steels. *Transact. of JWS*, 19(**1**), 46--52.
14. Tamaki, K., Suzuki, J. (1985) Combined influence of phosphorus, chromium and molybdenum on reheat cracking of steels. *Ibid.*, 16(**2**), 117--124.
15. Frumin, I.I. (1975) Surfacing in nuclear machine-building. *Avtomatich. Svarka*, **10**, 66--73.
16. Guo, A.-M., Wang, Y.-H., Shen, D.-D. et al. (2003) Influence of phosphorus on the ductility of 2.25Cr1Mo steel. *Mat. Sci. and Techn.*, **19**, 1553--1556.
17. Utevsky, L.M., Glikman, E.E., Kark, G.S. (1987) *Reversible temper brittleness of steel and iron alloys*. Moscow: Metallurgiya.
18. Ito, Y., Nakanishi, M. (1972) Study on stress relief cracking in welded low alloy steels: Report 2. *J. JWS*, 41(**1**), 59--64.
19. Sarrak, V.I., Entin, R.I. (1960) Irreversible brittleness in tempering. *Metallovedenie i Termich. Obrab. Metallov*, **10**, 14--19.
20. Demiantsevich, S.V., Zemzin, V.N. (1979) Procedure for evaluation of cracking sensitivity of welded joints in heat treatment. *Trudy TsKTI*, Issue 169, 22--27.
21. Lanskaya, K.A. (1976) *High-chromium heat-resistant steels*. Moscow: Metallurgiya.
22. Skulsky, V.Yu. (2006) Influence of alloying of filler material and welded steel on fusion zone structure. *The Paton Welding J.*, **1**, 9--15.