

Behavior of Welded Joints of Creep-Resistant Steels at Service Temperature

Creep-resistant steels were examined in an attempt to reduce the probability of in-service cracking

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ABSTRACT. Power station pipelines and other structural fabrications operating at high temperature are predominantly made of creep-resistant steels. These steels may become martensitic even if the cooling medium was air due to their relatively high Cr, Mo, and V content, which can retard the bainitic transformations. These steels must, therefore, be preheated, and filler materials must be carefully selected. In heterogeneous joints, substantial diffusion takes place during high-temperature service conditions. This paper describes the layered formation in the heterogeneous welded joint and reports on the investigation of the consequences of such diffusion. The principal goal of this investigation was to reduce the probability of cracking during service.

Introduction

A substantial number of creep-resistant steels are used in the fabrication of nuclear and fossil fuel power stations. These steels are alloyed using high-content chromium and molybdenum with the consequence that, after welding, in spite of air cooling, their microstructure can become quenched to a bainitic-martensitic or fully martensitic structure. To prevent this undesirable outcome, these steels are preheated to 200–350°C (based on empirical data) prior to welding. When a welding procedure is being elaborated, the preheat temperature is selected in accordance with wall thickness. This practice is not wholly correct, because the optimum preheat temperature for the concentration range of heats permitted by the relevant steel standards may be different. If the composition is left out of consideration, the crack sensitivity of the steels made of certain heats may be increased.

In many power stations there may be several welded joints in which the base metals and the filler materials are not of the same type. These joints are referred to as

heterogeneous joints. It is well known that in such a joint, diffusion occurs at increased temperatures, and diffusional layers are formed that may cause crack formation within a short service time. Diffusional layers are the zones formed due to the diffusion movement of the carbon on both sides of the interface between the weld and the heat-affected zone (HAZ), typically carbide layers and decarburized zones. The microstructure of the formed diffusional layers significantly differs from the neighboring parts of the joints. The characteristic dimension of these layers is the thickness, i.e., the size perpendicular to the weld interface between the weld and the HAZ. The rate of dimensional change of these layers is relative to the chemical composition: the greater the composition (first of all Cr) difference between weld and base metals, the greater the dimensional change. Temperature also has the same effect on this rate. In a heterogeneous joint of different base and filler materials, the acceptable service temperature cannot be as high as would be possible for homogeneous joints. The allowable service temperature for these heterogeneous joints has not been precisely determined and it is necessary to rely on empirical observations for these values.

In order to prevent welding imperfections, this paper suggests a method that is based on metallurgical knowledge and is in harmony with modern quality assurance principles. During the last decade, many experiments have been carried out on this topic and a number of the results have been published (see references) in addition to the present paper, which deals with the service behavior of heterogeneous joints.

KEY WORDS

Creep-Resistant Steel
Heterogeneous Joints
Crack Sensitivity
Ni-Based Filler Material
Preheating Temperature
Postwelding Tempering
Diffusion
Ferrite Layers
Carbide-Rich Layer
Service Temperature

Preheating and Tempering of Homogeneous Joints

During the last century, characteristic types of creep-resistant steels were developed all over the world. In Table 1, the data of the last column demonstrate that the permitted service temperature increases in the slightly oxidizing smoke gas atmosphere, if the quantity of alloying elements is higher (Refs. 1, 2). Similarly, another effect of alloying, namely the quenching susceptibility, is stronger in alloyed steel grades and, therefore, all steels except the ferrite-pearlitic (F+P) and austenitic (A) groups should be preheated.

According to experimental results (Ref. 3), the effect of chemical composition on the preheating temperatures (T_p) seems as significant as the wall thickness in the case of the bainite-martensitic (B+M) group. On the basis of investigations, it became possible to reduce the widely used temperature range of preheating as follows:

$$T_p = 180 + 200 \left[C + \frac{Cr + V}{5} + \frac{Mn}{6} + \frac{Mo + Ni + Si}{15} - 0.4 \right] \leq 200 \quad (1)$$

Equation 1 is valid with the following provisos:

- If the sum in parentheses is less than zero, preheat is not needed.
- The upper temperature limit is suggested for castings and wrought products of wall thickness greater than 40 mm.
- The lower temperature limit should be applied when the thickness is less than 15 mm.

Metallurgical characteristics of steels of B+M and M (martensitic) groups differ markedly from each other. These differences must be taken into account when creep-resistant steels are classified or standards are prepared (Ref. 4).

The suggested preheat temperature for the martensitic welding of steels belonging to the martensitic group should be equal to the interpass temperature (Ref. 5).

For steels of C=0.2%,

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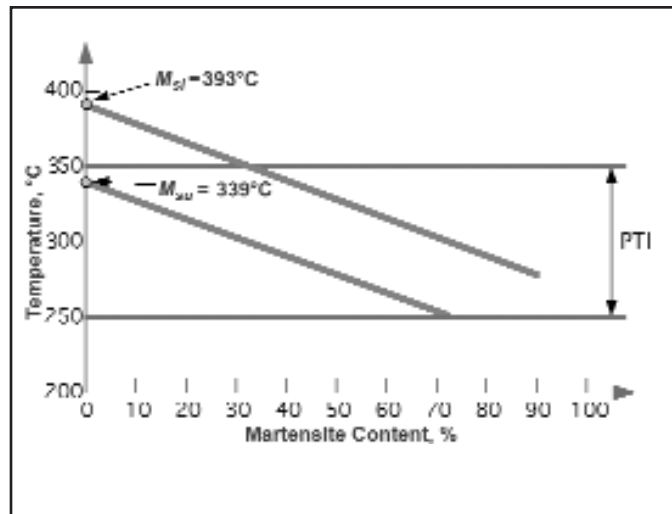
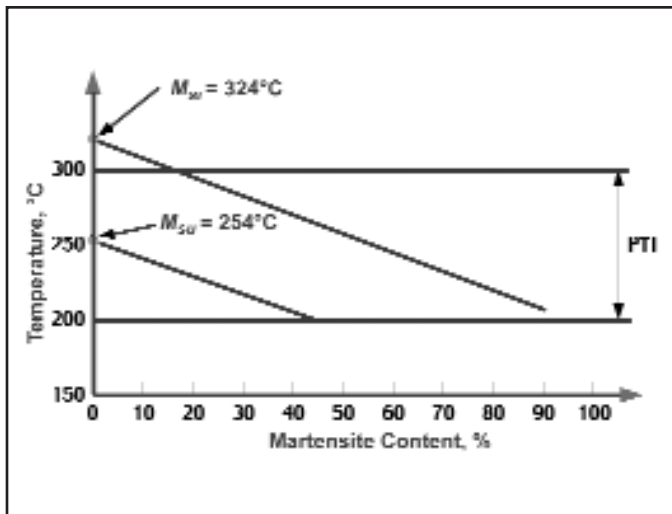


Fig. 1 — Martensite content of 0.2C-12Cr-1Mo-type creep-resistant steel during welding. M_{sl} is M_s temperature for heat of lower limit of the standard ($M_{sl}=324^\circ\text{C}$ [Ref. 5]); M_{su} is M_s temperature for heat of upper limit of the standard ($M_{su}=254^\circ\text{C}$); PTI is preheat temperature interval in the practice.

Fig. 2 — Martensite content of a 0.1C-9Cr-1Mo-type creep-resistant steel during welding. M_{sl} is M_s temperature for heat of lower limit of the standard ($M_{sl}=393^\circ\text{C}$ [Ref. 5]); M_{su} is M_s temperature for heat of upper limit of the standard ($M_{su}=339^\circ\text{C}$); PTI is preheat temperature interval in the practice.

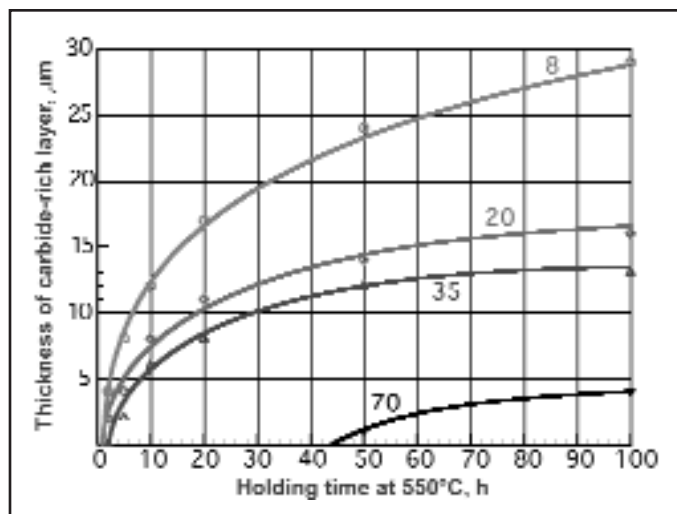


Fig. 3 — Thickness of carbide-rich layer after annealing at 550°C. Base metal: 0.15C-1.1Cr-0.6Mo. Filler material: 8 — 18Cr-8Ni; 20 — 22Cr-20Ni; 35 — 20Cr-35Ni; 70 — 70Ni-20Cr.

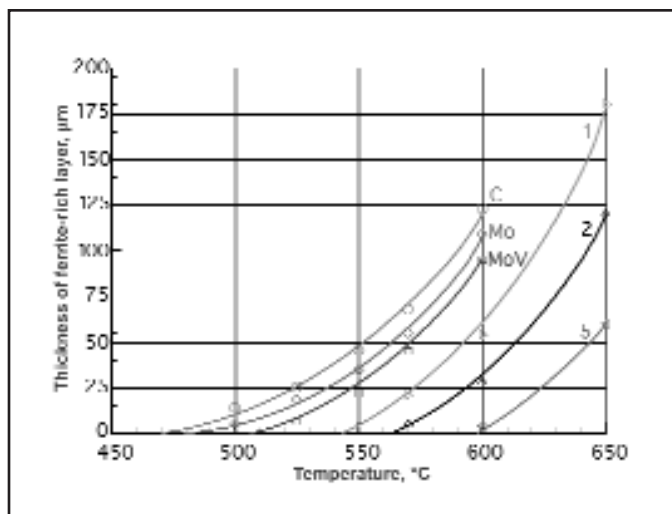


Fig. 4 — Thickness of ferrite layer after a 100-h annealing at different temperatures. Base metals: C — 0.15C-0.8Mn; Mo — 0.15C-0.8Mn-0.3Mo; MoV — 0.15C-0.6Mo-0.3V; 1 — 0.15C-1.1Cr-0.6Mo; 2 — 0.15C-2.2Cr-1Mo; 5 — 0.1C-5Cr-0.6Mo; Filler material: 70Ni-20Cr.

$$T_{p0.2} = (M_s - 60) \angle 10 \square C \quad (2) \quad (\text{Cr} + \text{Mo} + 1.5 \square \text{Si} + \text{W} + \text{V}) - 21 \langle \text{Cu} \quad (4)$$

while for the C=0.1% steel in the same group,

$$T_{p0.1} = (M_s - 90) \angle 10 \square C \quad (3)$$

The M_s temperature in Equations 2 and 3 can be calculated from the composition (Ref. 5):

$$M_s = 454 - 210 \langle C + \frac{4.2}{C} - 9.5 \langle$$

In martensitic steel groups, the effect of wall thickness can be negligible compared to the effect of chemical composition (Ref. 6); therefore, for wall thicknesses more than 40 mm, it is sufficient to add 10°C to the suggested average.

On the basis of their experimental results given in the literature, Steven, Haynes, and Kauhausen state that if the welding was carried out at other than the advised preheating temperature, 10°C dif-

ference from the suggested average value causes 8% difference in martensite content forming during welding (Ref. 5).

These observations are visualized in Figs. 1 and 2. These figures show the widely used preheating interval and the composition effect on the amount of formed martensite during cooling.

It can be seen in Fig. 1 that the microstructure in the HAZ of the welded joint of base metals whose composition equals the lower limit of the relevant standard contains 20% martensite and 80% austenite if the preheating temperature was the maximum possible, i.e., 300°C. When the preheat is lower, e.g., 220°C, the

Table 1 — Classification of Creep-Resistant Steels

| Group of Steels | Composition Type | Chemical Composition, wt-% | | | | | Service Temp., °C |
|----------------------------|-------------------|----------------------------|------|------|------|------|-------------------|
| | | C | Mn | Cr | Mo | Ni | |
| Ferrite-Pearlitic F+P | 0.15C-0.8Mn | 0.12 | 0.40 | — | — | — | 480 |
| | | 0.25 | 1.20 | — | — | — | |
| | 0.15C-0.8Mn-0.3Mo | 0.12 | 0.40 | — | 0.20 | — | 500 |
| | | 0.20 | 1.00 | — | 0.40 | — | |
| Bainite-Martensitic B+M | 0.15C-0.6Mo-0.3V | 0.10 | 0.40 | 0.30 | 0.40 | — | 560 |
| | | 0.20 | 0.70 | 0.60 | 0.70 | — | |
| | 0.15C-1.1Cr-0.6Mo | 0.10 | 0.40 | 0.70 | 0.40 | — | 570 |
| | | 0.20 | 0.70 | 1.30 | 0.70 | — | |
| | 0.15C-2.2Cr-1Mo | 0.07 | 0.40 | 2.00 | 0.80 | — | 600 |
| | | 0.18 | 0.70 | 2.50 | 1.20 | — | |
| Martensitic M | 0.10C-5.0Cr-0.6Mo | 0.08 | 0.30 | 4.50 | 0.45 | — | 650 |
| | | 0.14 | 0.60 | 6.50 | 0.65 | — | |
| | 0.20C-12Cr-1.0Mo | 0.17 | 0.40 | 10.0 | 0.80 | — | 650 |
| | | 0.25 | 0.70 | 12.5 | 1.20 | — | |
| | 0.10C-9.0Cr-1.0Mo | 0.07 | 0.40 | 7.50 | 0.80 | — | 650 |
| | | 0.14 | 0.70 | 9.50 | 1.10 | — | |
| Austenitic A | 0.06C-16Cr-13Ni | 0.04 | 0.80 | 15.0 | — | 12.0 | 750 |
| | | 0.10 | 1.50 | 17.0 | — | 13.5 | |
| | 0.06C-16Cr-16Ni | 0.04 | 0.80 | 15.0 | 1.50 | 15.0 | 750 |
| | | 0.10 | 1.50 | 17.0 | 2.00 | 16.5 | |

microstructure contains 82% martensite and 18% austenite. If the steel heat belongs to the upper limit of the standard, and the preheating temperature is the same (220°C), then only 28% martensite forms, and at the preheating temperature over 254°C, there is no martensite at all.

Consequently, if the chemical composition is not taken into account and the preheating and the interpass temperatures are kept as 220°C, martensite content can vary between 28 and 82%, depending on the actual chemical composition of the steels to be welded. It is understandable, therefore, that a person who has welded creep-resistant steels with 30% martensite content for a long time will be surprised by a new steel heat in which 80% martensite can be formed during welding, because the weld and HAZ show increased crack susceptibility.

After some practice, an experienced, qualified welder can successfully weld the different steel heats by applying the usual preheating temperature interval (in Fig. 1 it is denoted as PTI), but the crack risk can be minimized if the welding is done at a preheating temperature to which the same martensite amount belongs. This preheating temperature is considered an optimal one. The optimal preheating temperature can be calculated by Equation 2. It follows

from Fig. 1 that the martensite content of any heat is approximately 50% if it was preheated to ($M_s - 60$)°C. Applying the advised preheated temperature, joints will be uniform from the viewpoint of the quality, and much time can be saved, mainly in repair or on-site welding, because there is no need for testing.

The structure of Fig. 2, belonging to the 0.1C-9Cr-1Mo steel, is similar to Fig. 1, except the PTI is now 250–350°C. From Fig. 2, the same results can be drawn. Experiments demonstrated that as high as 70% martensite content can be allowed for this steel, because of the lower C concentration and lower hardness. The optimal preheating temperature can be calculated by Equation 3.

Following completion of the welding of steels with B+M and M microstructure the HAZ becomes rather hard and brittle. In order to reduce hardness to the acceptable level, the joints should be tempered. During tempering, a certain upper limit given in Table 2 must not be exceeded, otherwise the austenitized part of the joints could be transformed into martensite. Tempering at lower temperatures than the advised values would reduce the effectiveness of tempering. In the literature, papers unanimously claim (Refs. 6, 7) that the hardness increases by approximately

15 HV when the tempering temperature differs up or down from the suggested value by 10°C. At temperatures 60–80°C below the lower limit of the advised interval, tempering is practically ineffective. To be on the safe side, it is widely accepted that the tempering temperature always should be below the A_{c1} temperature by 20 to 30°C (Ref. 8).

Heterogeneous Joints Containing Austenitic Filler Metal

In homogeneous joints, the hardness of the weld and the HAZ are almost equal, i.e., considerably brittle before tempering. The frequently used austenitic filler material has an advantage, because the weld metal is tough and ductile during welding and cooling.

Since joints containing creep-resistant steels and austenitic filler material differ in composition, diffusional processes occur between the weld and HAZ of the welded joints at high (service or tempering) temperatures.

The Cr content of austenitic filler materials is approximately 20%. This concentration is much more than that of steels belonging to the F+P, B+M, or M group, so they can be applied to joints operating under 300°C and need not be tempered

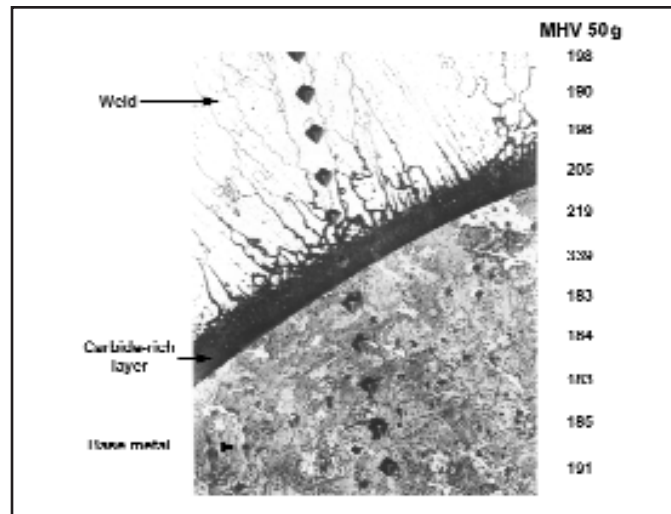
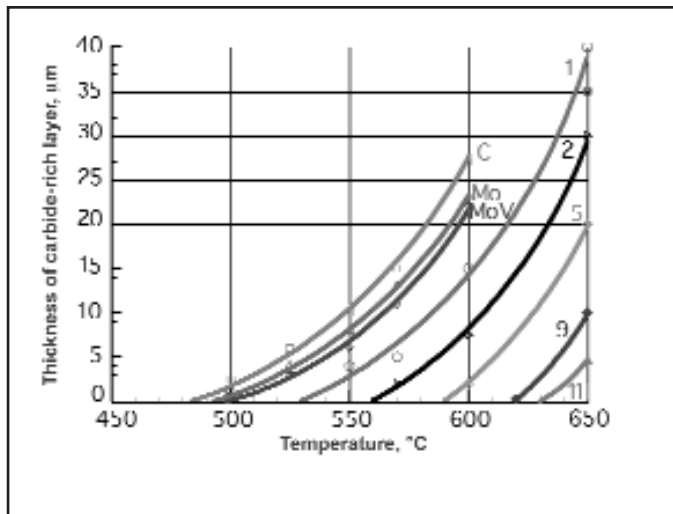


Fig. 5 — Thickness of the carbide-rich layer after a 100-h annealing. Base metals: C — 0.15C-0.8Mn; Mo — 0.15C-0.8Mn-0.3Mo; MoV — 0.15C-0.6Mo-0.3V; 1 — 0.15C-1.1Cr-0.6Mo; 2 — 0.15C-2.2Cr-1Mo; 5 — 0.1C-5Cr-0.6Mo, 9 — 0.1C-9Cr-1Mo; 11 — 0.2C-12Cr-1Mo. Filler material: 70Ni-20Cr.

Fig. 6 — Microstructure and hardness distribution of a heterogeneous joint, which had been operated for 35,000 hours in a main steam pipeline. Base metals: 0.15C-0.6Mo-0.3V. Filler material: 70Ni-20Cr. Service conditions: 520°C temperature and 120-bar pressure. Etching with aqua regia.

(Ref. 9). Above this temperature, the thickening rate of diffusional layers significantly increases, which offers a good possibility of determining the increase rate of layer thickness. For these welded joints, the filler metal should be chosen so that martensite does not appear in the diluted first layer (Ref. 10).

In experiments, a 0.15C-1.1Cr-0.6Mo-type creep-resistant steel was used. The dimensions of specimens were 100 mm long, 50 mm wide, and 8–12 mm thick. An 80-mm-long bead was welded onto the surface of the specimen by the shielded metal arc welding (SMAW) process. The specimens were soaked at 550°C for different holding times. After the heat effect, a cross section was made for measuring layer thickness. The specimens were etched by aqua regia, and the applied magnification of the optical microscope was 250.

Discussion of the theoretical basis (Refs. 11, 12) of the diffusional processes is not a chief aim of this paper, but it is well known that the diffusional rate of carbon is much higher than the same of other alloying elements (Ref. 13). Transport of carbon depends, among other factors, on the concentration difference between weld and base metals, the service temperature, and the microstructure of matrix (ferritic or austenitic). The carbon movement is directed from the lower Cr part to the higher since the carbon demand of creating chrome carbides dynamically decreases the carbon concentrations, causing diffusional fluxes. Consequently, a soft ferrite layer will form in the HAZ and a carbide-rich layer in the weld. Layer formation increases crack sensitivity in the welded joint (Ref. 14).

Table 2 — Postweld Tempering Temperature and Hardness of Bainite-Martensitic and Martensitic Creep-Resistant Steels

| Composition Type of Steels | Temp., °C A_{c1} | Temp., °C Tempering | Hardness in HV10 | |
|----------------------------|-----------------------|------------------------|------------------|-----------------|
| | | | Before Tempering | After Tempering |
| 0.15C-0.6Mo-0.3V | 735 | 700–720 | 260–320 | 200–250 |
| 0.15C-1.1Cr-0.6Mo | 745 | 660–700 | 260–320 | 180–230 |
| 0.15C-2.2Cr-1.0Mo | 780 | 690–750 | 270–340 | 200–250 |
| 0.10C-5.0Cr-0.6Mo | 785 | 720–750 | 280–320 | 200–250 |
| 0.20C-12.0Cr-1.0Mo | 820 | 740–760 | 350–450 | 200–260 |
| 0.10C-9.0Cr-1.0Mo | 810 | 740–780 | 300–370 | 200–250 |

The dimensional rate of change of the carbide-rich layer is shown in Fig. 3 as an example. It can be clearly seen that the layer begins to form only after a definite incubation time. The thickening rate (y) follows exponential law in time (t) and can be written as follows:

$$y = a_0 + a_1 \ln t \quad (5)$$

where a_0 is the threshold of the layer formation, and a_1 is the temperature-dependent thickening rate of the ferritic or carbide-rich layer.

Equation 5 shows that the thickening rate slows down as a function of time. This digressive process is characteristic for the ferrite layer as well, and can be explained by increasing of diffusion distances, while the diffusional coefficient of carbon is not significantly changed.

The thickening rate decreases as the Ni concentration increases, but considerable reduction can be achieved by applying Ni-based (70% Ni + 20% Cr) filler materials. It follows from the position and shape of

curves that if the traces of layer formation can be observed by optical microscopic examination as late as 100 h, then it is probable that a dangerous-sized continuous layer will not occur in a time longer by three orders of magnitude.

As observed in our experiments, the diffusional layer formation began after 100-h holding time at 520°C when 70/20-type filler metal was used. The curve designated as 70 in Fig. 3 belongs to the annealing at 550°C, would be much less steep at 520°C due to the lower temperature, and the threshold of the layer formation would be postponed to the longer holding time. The aforementioned consequences can be read from Figs. 4 and 5.

On the basis of our theory, the highest safe service temperature (the permitted service temperature) can be defined as the temperature belonging to the traces of the diffusional (ferritic or carbide-rich) layer formation after 100-h soaking time.

The design lifetime of fabrications that use creep-resistant steels is generally 200,000 hours (approximately 20 to 25

Table 3 — Regression Coefficients, Threshold Temperatures, and Rounded Values of Recommended (Maximum) Service

| Composition Type of Base Metals | Symbol of Base Metals in Figs. 4 and 5 | Ferrite Layer (Fig. 4) | | Carbide-Rich Layer (Fig. 5) | | Recommended Service Temp., °C (Lesser of Threshold Temps.) |
|---------------------------------------|--|--|---------------------|--|---------------------|--|
| | | Coefficients | Threshold Temp. | Coefficients | Threshold Temp. | |
| 0.15C-0.8Mn-0.3Mo | Mo | a_0 1411.0 a_1 -5.98270 a_2 0.0063400 | $T(y_F=0)$ 479.6 | a_0 260.90 a_1 -1.13312 a_2 0.0012252 | $T(y_C=0)$ 492.2 | 480 |
| 0.15C-0.8Mn | C | 1503.0 -6.41525 0.0068452 | 471.9 | 239.65 -1.07533 0.001199 | 483.2 | 470 |
| 0.15C-0.6Mo-0.3V | MoV | 1764.8 -7.30860 0.0075500 | 506.8 | 322.8 -1.36065 0.0014313 | 495.3 | 495 |
| 0.15C-1.1Cr-0.6Mo | 1 | 3856.7 -14.43680 0.0135060 | 544.0 | 605.9 -2.34608 0.0022695 | 530.3 | 530 |
| 0.15C-2.2Cr-1Mo | 2 | 3668.6 -13.34390 0.0121286 | 561.7 | 800.6 -2.94639 0.0027084 | 560.3 | 560 |
| 0.10C-5.0Cr-0.6Mo | 5 | 3979.5 -13.77860 0.0119140 | 597.2 | 1053.5 -3.70239 0.0032490 | 589.5 | 590 |
| 0.10C-9.0Cr-1.0Mo | 9 | | | 1752.0 -5.82170 0.0048325 | 619.0 | 620 |
| 0.20C-12Cr-1.0Mo | 12 | | | 914.0 -3.02564 0.0025000 | 629.2 | 630 |

years). In our opinion, the temperature that can be considered as the permitted maximum service temperature is that at which only traces of the layer formation can be discovered after 100 h of soaking.

Heterogeneous Joints Made by 70Ni/20Cr Filler Material

For operation at temperatures higher than 300°C, only nickel-based alloys can be used. Typical of this group is the 70% Ni and 20% Cr (70/20) alloy. Compared with creep-resistant steels, this alloy has life strength greater than martensitic steel, and its elongation is approximately 40% regardless of the heat-treated condition.

One-hundred-hour experiments were carried out at different soaking temperatures with specimens prepared as described above. The formed layer thicknesses were measured using an optical microscope.

After statistical analysis of the results, it was found (Ref. 15) that the best fit could be achieved by a second order polynomial as follows.

$$d = a_0 + a_1 \square T + a_2 \square T^2 \quad (6)$$

where d is the thickness of the ferrite or carbide-rich layer in μm , T is temperature in $^\circ\text{C}$, and a_i are regression coefficients given in Table 3.

Thicknesses as a function of temperature are illustrated in Figs. 4 and 5. Curves intersect the horizontal axis at the temperature at which the traces of layer formation appear. Formation of ferrite and carbide-rich layers does not necessarily begin at the same temperature, e.g., in the steels No. 9 and No. 11, a ferrite-rich layer did not appear at all. The suggested service temperature is the lower of them. We checked the experimental results a few times and stated them precisely. These temperatures are collected and systematized in Table 3.

It is clear from the data in Table 3 that the recommended service temperatures increase sharply with the Cr content of the base metal.

For example, let the base metal be 0.15C-0.6Mo-0.3V-type creep-resistant

steel and the filler metal 70Ni-20Cr-type Ni-based alloy. This case can occur when ferrite-pearlitic and austenitic creep-resistant steels are welded together. Traces of ferrite layer can be observed after 100-h annealing at 506.8°C, just as a carbide-rich layer appears in the joint. It follows from the given data that the allowed service temperature can be chosen as the lower of them, i.e., 495°C.

On the basis of extensive investigation, it is probable that the first layer, diluted by the base metal chemical composition, has not equalized (Ref. 16) and, consequently, the layer hardness is affected by the pre-heating temperature (Ref. 17).

In the last decade, new results were published about the construction of diffusional layers (Refs. 18, 19), which helped to reveal the mechanism of the layer thickening.

It is clear from the previous works that experimental investigations into the service life of the creep-resistant steels are expensive and time consuming. For this reason, a computer model to simulate the diffusional processes taking place at in-

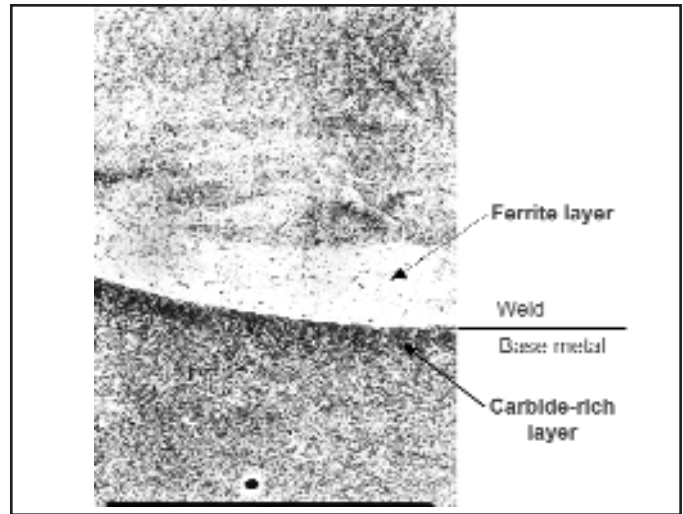
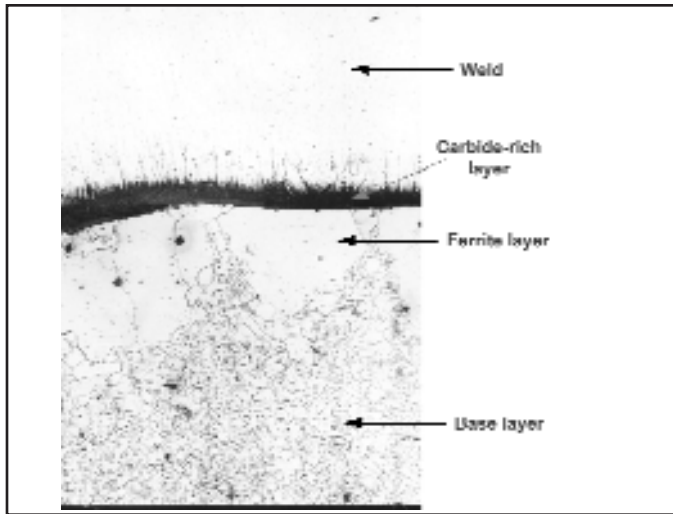


Fig. 7 — Microstructure of the welded joint after tempering of 750°C/3 h. Base metals: 0.15C-1.1Cr-0.6Mo. Filler material: 70Ni-20Cr. Etching with aqua regia.

Fig. 8 — Microstructure of the welded joint after tempering of 750°C/2 h. Base metals: 0.1C-9Cr-1Mo. Filler material: 0.15C-1.1Cr-0.6Mo. Etching with aqua regia.

creased temperature in the equipment of power stations was constructed using the finite volume method (Ref. 20).

To illustrate the reliability of data, several cases are reported below. In an industrial case, a welded joint operated in the main steam pipe of a power station at a temperature 25°C higher than recommended is considered. In the microphoto taken from the section of the pipe (Fig. 6), indentations of a micro-Vickers diamond pyramid loaded by 50-g weight can be seen. The diffusion process directed into the weld metal penetrated much deeper at the grain boundaries than into the grains themselves.

The hardness of the weld was not greater than 200 MHV. Proceeding toward the fusion zone, the hardness increased because more and more carbide grains formed during service life. The hardness of the darkly colored carbide-rich layer is 339 MHV. From the neighboring heat-affected zones, the carbon went off and a ferrite layer (183 MHV) began to form. The next indentation is located at the border of the ferrite layer, and thereafter the hardness of base metal ranges between 185 and 191 MHV.

This photograph shows the fine grains in the ferrite layer. This is due to the V content of the steel. In the vanadium-free steels, the ferrite layer has a coarse grain structure.

The investigated joint was available for inspection because of the reconstruction of the pipeline system. This joint had been in service for 35,000 hours. According to our conclusion, a crack should have formed in the next three to four years in the ferrite layer. This crack-sensitive structure could not have formed if the ser-

Table 4 — Recommended Service Temperature of Welded Joints Made of Quenchable Filler Material

| Composition Type of Base Metals | Cr Content in Base Metal, wt-% | Cr Content in Weld, wt-% | Recommended Service Temp., °C |
|---------------------------------|--------------------------------|--------------------------|-------------------------------|
| 0.15C-2.2Cr-1Mo | 2.2 | 9 | 490 |
| 0.15C-0.8Mn-0.3Mo | 2.2 | 11 | 490 |
| 0.15C-2.2Cr-1Mo | 2.2 | 5 | 560 |
| 0.10C-9.0Cr-1.0Mo | 9.0 | 5 | 600 |
| 0.20C-12Cr-1.0Mo | 12.0 | 5 | 590 |
| 0.10C-9.0Cr-1.0Mo | 9.0 | 2.2 | 560 |
| 0.20C-12Cr-1.0Mo | 12.0 | 2.2 | 540 |

vice temperature had not exceeded the permitted 495°C.

When two steels with different Cr content are welded, service temperature is determined by the side with the lower Cr. Tempering temperature should also be lower than A_{c1} temperature by 20 to 30°C.

Nowadays, a few constructional plans can still be found for which 0.15C-1.1Cr-0.6Mo- and 0.1C-9Cr-1Mo-type steels must be welded. This is now seen to be incorrect since there is no suitable tempering temperature for both sides: 700°C is too low for 9% Cr steel, and 750°C is very high for 1.1% Cr steel, because a thick ferrite layer may form even during tempering — Fig. 7. Welded joints of these steels must not be used operationally. To prevent the problem of incompatibility, a minimum 2-m-long pipe made from a 0.15C-2.2Cr-1Mo-type creep-resistant steel should be inserted. The microstructure of Fig. 7 points out that the duration of the tempering cannot not be too long since the hardness of the joint decreases significantly in the first hour, and later the soft-

ening slows down, but the diffusional layers thicken considerably from the second hour.

In cases when a steel of B+M or M groups is to be welded to austenitic or 0.15C-0.8Mn unalloyed steel grade, an 8- to 10-mm-thick buttering layer should be made using 70/20 type Ni-based alloy. After tempering, the joint welding between the buttering layer and the austenitic or ferrite-pearlitic steel should be carried out with the same filler material, without preheating (Ref. 12). With this method, we can prevent the dilution that would be formed between the Ni-based alloy and any other type of filler material, as well as the hot crack sensitivity caused by the dilution.

Heterogeneous Joint with Quenchable Filler Material

In such a joint, the composition difference between parts of the joint is lower than the same in joints made with austenitic filler material. However, the

weld metal does not contain nickel, hampering diffusion movement; therefore, danger of layer formation is increased.

In the course of the experiments it was found that the thickening rate of the diffusional layers can be estimated by the same type of regression equation as was shown for the joints of austenitic filler material. For the quenchable joints, more filler material can be used, so the number of variations is larger. Table 4 shows likely combinations that can be used.

It was concluded from the experimental results that if the Cr difference between weld metal and base metal is less than 2.5%, diffusional layers harmful for strength parameters of the joint are not formed either during tempering or service. This is true for circumstances where the service temperature does not exceed the permitted temperature valid for the less-Cr-containing steel part of the joint, according to Table 1.

Using a combination in which the Cr difference between base and filler metals is larger than given in Table 4, thick layers can appear in the joint even during tempering, which can crack after a relatively short service time.

This phenomenon is illustrated by the photo in Fig. 8. A few years ago, there was a design between a 0.1C-9Cr-1Mo and a 0.15C-1.1Cr-0.6Mo steel pipe for which a 0.15C-1.1Cr-0.6Mo-type filler material was planned. If this joint had been made, a very thick diffusion layer would have been formed at one side of the joint during tempering, and this layer would have cracked in the first few years.

In this example, the ferrite layer developed in the weld. If the Cr content of the base metal is lower, the ferrite layer will form in it.

A conclusion can be drawn from the examined cases relating to the joints with different Cr steels as follows: joints should be welded by a filler material having Cr content near the average Cr percentage of the base metals.

These joints should be tempered at temperature lower by 20–30°C than the A_{c1} temperature of the steel of lower Cr content part (Table 2). In cases when highly different Cr-containing steels have to be welded, a 1-m-long pipe made from 0.15C-2.2Cr-1Mo steel should be inserted, similarly to the example above.

Our experimental results are in harmony with the theory published in Refs. 21 and 22.

Conclusions

It follows from the principles of quality assurance that welded joints should be reproducible and of the best quality at any time. Accordingly, homogeneous joints

should be welded at a broad preheating temperature interval determined on the basis of chemical composition. It is important to minimize crack sensitivity during welding, because postweld tempering is not able to equalize the difference in hardness in parts of the joint.

It is suggested that there is no such tempering temperature optimal for the elements of a given heterogeneous joint. Due to the concentration differences, strong diffusional movement takes place between the weld and heat-affected zone. This diffusional process accelerates when the service temperature of new power stations has been increased. Both ferrite and carbide-rich zones are undesirable, because of their high-level crack sensitivity. On the basis of detailed metallurgical investigations, attention is directed to requirements that can be prescribed for the welds and heat-affected zones of these joints and methods to prevent the in-service cracking of heterogeneous welded joints.

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