THE EFFECT OF TEMPERING ON MECHANICAL PROPERTIES OF COLD-ROLLED 42CrMo4 STEEL PIPES

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The effect of tempering on mechanical properties of cold rolled 42CrMo4 steel pipes was studied. The production of pipes consisted of hot rolling + annealing at 700 °C + cold rolling + quenching with subsequent tempering. By annealing of seamless pipes at 700 °C under industrial conditions and using protective atmosphere and cold rolling, the tempered bainite microstructure was observed. Quenching and tempering of pipes in laboratory conditions resulted in a high yield strength (986–1230 MPa), combined with a high impact toughness value at 20 °C (46.1–66.9 $\mathrm{J \cdot cm^{-2}}$). The strength of pipes shows a continuous decrease with the increase in tempering parameter. Decrease in strength is due to the spheroidization of carbides and the recovery of the microstructure. The most important observation is that it is possible to use this technique for the production of pipes in order to optimize the strength and toughness of 42CrMo4 steel pipes.

Key words: steel, cold-rolling, pipes, tempering, mechanical properties, microstructure

VPLYV POPÚŠŤANIA NA MECHANICKÉ VLASTNOSTI ZA STUDENA VALCOVANÝCH OCEĽOVÝCH RÚR 42CrMo4

Študovali sme vplyv popúšťania na mechanické vlastnosti za studena valcovaných oceľových rúr 42CrMo4. Výroba rúr pozostávala z valcovania za tepla, žíhania pri 700 °C, valcovania za studena a kalenia s nasledujúcim popúšťaním. Popúšťaciu bainitickú štruktúru sme pozorovali v bezšvových rúrach, ktoré sme žíhali v ochrannej atmosfére pri 700 °C a valcovali za studena v priemyselných podmienkach. Kalenie a popúšťanie rúr v laboratórnych podmienkach viedlo k vysokej medzi sklzu (986–1230 MPa) kombinovanej s vysokou rázovou húževnatosťou pri 20 °C (46,1–66,9 $\mathrm{J \cdot cm^{-2}}$). Pevnosť rúr sa kontinuálne znižovala so zvyšovaním parametra popúšťania. Zníženie pevnosti zapríčinila sferoidizácia karbidov a zotavenie mikroštruktúry. Použitá metóda je vhodná na výrobu rúr s cieľom optimalizovať pevnosť a húževnatosť oceľových rúr 42CrMo4.

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1. Introduction

It is known that heat treatment technology improves the strength and especially the toughness of low alloy steels [1]. Low alloy CrMo steels are used in quenched and tempered condition for general engineering and petrochemical applications [2]. The main reason for these applications is excellent hardenability combined with high strength and good toughness properties. Conventional heat treatment of low alloy CrMo steel consisted of austenitization at 830–870 °C, followed by oil quenching and tempering at different temperatures. Due to different conditions of application, the optimum combination of toughness, strength and ductility of steels is desirable. For this purpose, many approaches have been proposed [3–6]. Some of them are double austenitizing [3], high temperature austenitizing [4], rapid austenitization [5], and duplex treatment consisting of two stages: austenitizing and tempering at a high temperature, and then at conventional temperature [6]. The second alternative approach involves microalloying of low alloy CrMo steel with vanadium, titanium, niobium, boron, plus variations in chromium and molybdenum contents [7]. Considerable effort has been directed towards the relationship between the microstructure and mechanical properties of fully martensitic low alloy steels [8]. However, systematic study has not been carried out to clarify the production of cold rolled pipes including heat treatment although such pipes are frequently encountered in commercial practice.

The main objective of this work was to develop the production of cold rolled pipes including the heat treatment programme that would result in a high yield strength close to 1100 MPa, combined with a high toughness value at 20 °C. The 42CrMo4 steel was chosen since it is probably the most representative low alloy steel. In the present work, the effect of tempering temperature on both the mechanical properties and microstructure of 42CrMo4 steel pipes was investigated.

2. Experimental

The 42CrMo4 heat was produced in an electric arc furnace and continuously cast in billets. The billets were hot rolled to seamless tubes (ø 159 × 8 mm) under industrial conditions. Before cold rolling, pipes were annealed at 700 °C under industrial conditions and protective atmosphere ($CO + H_2 = 6$ vol. %). Cold rolling of seamless tubes (ø 130 × 6.2 mm) was performed by pilgering. Chemical composition of pipes is given in Table 1. It is tested by an optical and X-ray emission spectrometry, type ARL 8686. The content of carbon and sulphur was determined by Leco CS-444 device. The heat treatment of pipe samples was carried out in a laboratory electric resistance furnace without protective atmosphere.

The heat treatment of samples consisted of quenching and tempering. The samples were inserted into a heated furnace at 860 °C, austenitized for 25 minutes

Table 1. Chemical composition of steel investigated [wt. %]

Steel	C	Mn	P	S	Si	Cu	V	Mo	Cr	Al
42CrMo4	0.44	0.77	0.011	0.015	0.25	0.27	0.01	0.28	1.06	0.026

and quenched in oil. Tempering was performed in a temperature range from 520 to 600 °C for 30 minutes followed by air cooling.

Before and after heat treatment, mechanical properties were tested on Instron tensile testing machine (type 1196) in accordance with the ASTM procedure [9]. The hardness test was performed by Brinell (HB) and Vickers (HV_{10}) methods. Three Charpy V-notch specimens, taken from the pipes in direction of rolling, were tested at 20 °C.

Microstructure of polished and etched (in 2 pct nital solution) specimens before and after heat treatment was studied by scanning electronic microscope (SEM), type Jeol JXA-50A, operating at a 50 kV voltage.

3. Results

Mechanical properties of pipes before and after heat treatment are given in Table 2. The reported mechanical properties present the average of three determinations. The values of mechanical properties and toughness were reproducible within 3%. Fig. 1 shows microstructure after hot rolling. As it can be seen, it is the bainite microstructure with small carbide particles. Figure 2 shows SEM micrographs of pipes after cold rolling. As shown, the microstructure consisted of

Table 2. Mechanical properties of pipes after different treatment

State of pipes	Yield strength [MPa]	Tensile strength [MPa]	Elongation [%]	Hardness [HB]	Impact energy at 20 °C [J]	Impact toughness at 20 °C [J/cm^2]
Cold-rolled	730	866	16.3	297*	9.6	56.6
Q: 860 °C/25 min, oil	–	–	–	692**	–	–
Q: 860 °C/25 min, oil T: 520 °C/30 min, air	1230	1290	10.5	412*	7.6	46.1
Q: 860 °C/25 min, oil T: 560 °C/30 min, air	1092	1173	13.6	366*	9.9	56.4
Q: 860 °C/25 min, oil T: 600 °C/30 min, air	986	1063	16.1	361*	11.6	66.9

Q – quenching; T – tempering; * During Brinell test the loading of 187.5 kg, ball diameter of 2.5 mm and time of 30 s were used ($HB_{187.5/2.5/30}$); ** Vickers hardness with the loading of 10 kg (HV_{10})

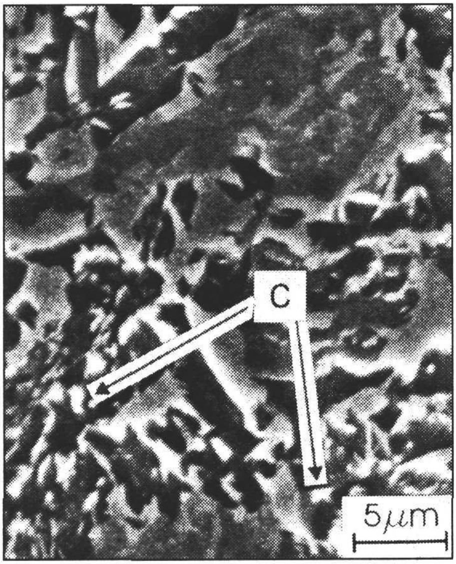

Fig. 1. SEM micrographs of pipes after hot rolling. C – carbide particles in ferrite matrix.

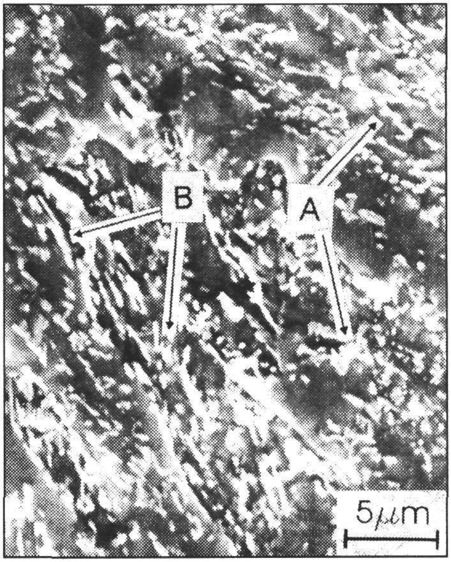

Fig. 2. SEM micrographs of pipes after cold rolling. A – globular carbides; B – rod carbides.

tempered bainite with a considerable amount of carbides resulting from annealing at 700 °C before cold deformation. Two types of carbides were observed: globular carbides and rod shaped carbides. Figure 3 shows SEM micrographs of pipes after quenching. The microstructure is composed of lath martensite packets and few undissolved carbide particles.

It is well known that it is not possible to use martensitic high strength steels in quenched condition (Table 2, $HV_{10} = 692$), i.e. without subsequent tempering [10]. Therefore, pipes were tempered at 520, 560 and 600 °C, immediately after quenching. SEM micrographs of specimens tempered at 520 and 600 °C are given in Figs. 4 and 5, respectively. The microstructure in Fig. 4 shows tempered martensite with precipitated carbides (arrows indicate carbides in the ferrite matrix). Density of carbide particles is low. SEM micrograph (Fig. 5) shows the presence of numerous carbides for samples tempered at 600 °C.

4. Discussion

Bainite is a complicated microstructure of steel [11, 12]. It represents a mixture of acicular bainitic ferrite and carbides. Arrows in Fig. 1 indicate that carbides are present within the ferrite matrix. This microstructure was obtained by continuous

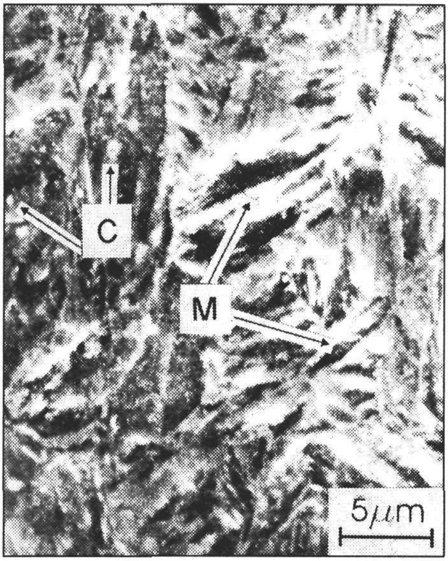

Fig. 3. SEM micrographs of samples after quenching in oil. M – lath martensite; C – carbide particles.

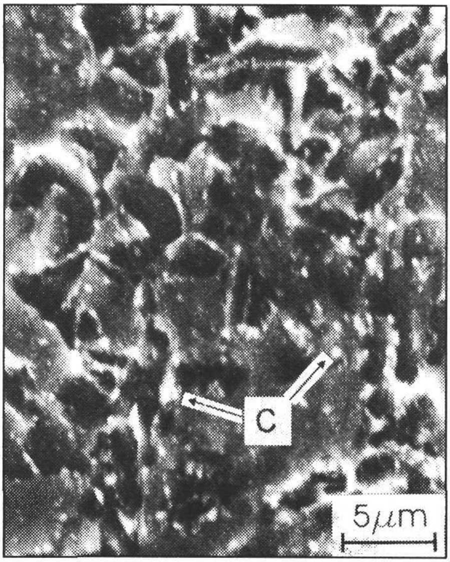

Fig. 4. SEM micrographs of samples after tempering at 520 °C. C – carbide particles in the ferrite matrix.

cooling with a final rolling temperature of 880 °C. Tempered bainite microstructure (Fig. 2) was obtained after annealing at 700 °C and cold rolling. Both globular and rod shaped carbides were observed. At 45° the rod carbides lay parallel with the longitudinal axis of rolling. After cold rolling, the pipes with desirable properties were obtained. To obtain higher mechanical properties of pipes, quenching and tempering were carried out. The cooling rate of samples during quenching in oil was just sufficient to avoid phase transformation. The microstructure presented in Fig. 3 confirms this finding. As it can be seen, the microstructure was lath martensite. The average width of martensitic laths is about 0.2 μm. Similar observations have been reported by other investigators [14, 15].

From a thermodynamical point of view, tempering is an irreversible process during which quenched pipes reach a more stable, the so-called stationary state. Major changes which can be expected on tempering are outlined by Speich and Leslie [16]. The morphology of carbides changes with tempering temperature and coarsening is significant above 600 °C. During tempering in a temperature range from 520 to 600 °C, a recovered microstructure with carbide precipitation evolved (Figs. 4 and 5). The results obtained show that presumably only the M_3C carbide particles are present. Carbides were observed within the ferrite matrix and on grain boundary of martensitic packets and/or on lath interfaces. The M_3C carbide is thermodynamically more advantageous in comparison to pure Cr or Fe carbides. As shown on the carbide phase stability diagram [17] for 2.25Cr-1Mo steel in a temperature range from 520 to 600 °C, molybdenum does not form its own carbide but has a tendency to alloy the existing stable M_3C carbides.

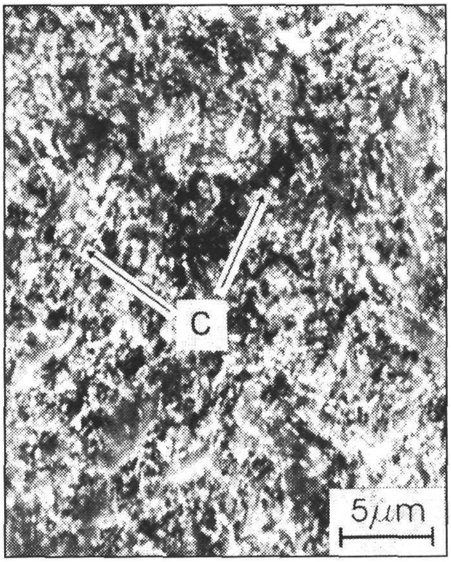

Fig. 5. SEM micrographs of samples after tempering at 600 °C. C – the spheroidized carbide particles.

Due to temperature variations, the tempering conditions were expressed in terms of a tempering parameter (TP), calculated according to [19]:

$$TP = T(20 + \log t) \times 10^{-3}, \qquad (1)$$

where T is temperature in K and t is time in hours.

Figure 6 shows the yield strength (YS) and tensile strength (TS) plotted against tempering parameter. The strength of samples is continuously decreasing with increase in tempering parameter. Decreased strength depending on the increase in tempering parameter between 520 and 600 °C is consistent with the work of Winter and Woodward [20]. They showed that above 500 °C the AISI 4130 steel-type shows a significant decrease in strength due to the spheroidization of carbides and the recovery of microstructure. This was also observed in Figs. 4 and 5. The most important observation of this investigation is that it is possible to use this technique for the production of pipes in order to optimize strength and toughness of 42CrMo4 steel pipes. In particular, the samples quenched and tempered at the lowest tempering parameter ($TP = 15.62$) had a high strength ($TS = 1290$ MPa) and a good impact toughness at 20 °C (46.1 J/cm^2).

The results of present studies demonstrated that the toughness of samples can be improved by tempering (Table 2). More important, excellent toughness at 20 °C with a relatively high strength can be obtained by tempering at higher temperatures. Figure 7 shows the effect of tempering temperature on impact toughness of samples tempered at 520, 560 and 600 °C. The values of toughness in the direction of rolling are increasing with increase in tempering temperature. This improvement in toughness observed at 600 °C can be explained by a change in microstructure. The observation that more carbides are formed at 600 °C in tempered samples

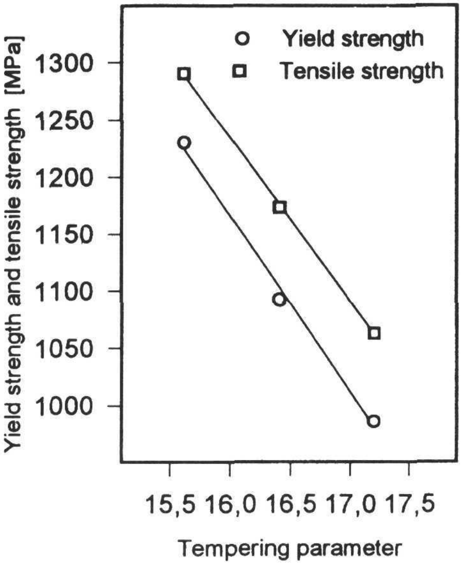

Fig. 6. Relationship between yield strength and tensile strength of pipes vs. tempering parameter.

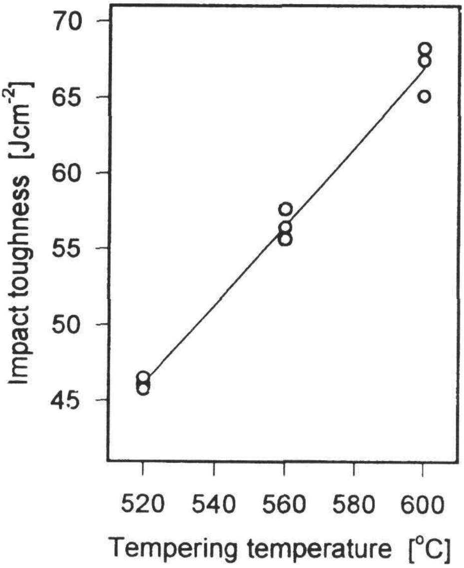

Fig. 7. The effect of tempering temperature on particular values of impact toughness for samples taken from pipes. Symbols present the particular measurements of toughness.

is interesting. Figure 5 shows a higher density of carbides in the ferrite matrix as compared to the samples tempered at 520 °C (Fig. 4). This higher density of fine precipitates in tempered samples at 600 °C may lead to improved toughness properties over those of samples tempered at 520 °C. However, the quantitative relationship between microstructure and toughness is complex and not completely understood [21]. It has not been determined whether these differences caused the increase of toughness at relatively low tempering parameters. An attempt to clarify the problem will be made in future work.

5. Conclusion

The hot rolled pipes with bainite microstructure were obtained. The annealing of seamless pipes at 700 °C and cold rolling under industrial conditions gave desirable mechanical properties. The lath martensite (hardness of 692 HV_{10}) was obtained by quenching of samples in oil. After tempering in the temperature range from 520 to 600 °C, the yield strength of samples from 986 to 1230 MPa was attributed to the spheroidization of carbides and to the recovery of the microstructure. It was found that by means of combination of a cold rolling and heat treatment (annealing at 700 °C before cold deformation and quenching + tempering) it is possible to obtain pipes with optimized yield strength (or tensile strength) and toughness.

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