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M/A-CONSTITUENT IN BAINITIC LOW-CARBON HIGH-STRENGTH STEEL STRUCTURE. PART 2*

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The microstructure and mechanical properties of steel K65 (X80) with a significant proportion of martensite and residual austenite (M/A-constituent) within the microstructure, prepared in a laboratory rolling mill by controlled rolling using stepwise accelerated cooling, are studied. During accelerated cooling (AC) the temperature for the end of the first stage and the duration of pauses between AC stages was varied. A dependence is revealed for mechanical properties on temperature and time parameters of the AC process and microstructure. It is shown that the microstructure formed during two-stage AC, consisting of a matrix of low-carbon bainitic ferrite and secondary high-carbon phases in the form M/A-constituent "islands", makes it possible to obtain rolled sheet with high strength and increased ductility, with a low σ_y/σ_f ratio, and good cold resistance. Formation of this microstructure with stepwise AC compared with single-stage cooling makes it possible to increase ultimate strength on average by 40 MPa, relative elongation on average by 5% (abs.), and uniform elongation by 3%. The optimum properties are achieved for steel sheet with a microstructure consisting of a matrix of bainitic ferrite and strong phase in the form of M/A-constituent "islands" with predominance of residual austenite.

Keywords: M/A-constituent, retained austenite, etching in LePera reagent, two-stage accelerated cooling, mechanical properties, transmission electron microscopy.

Experimental results are described in [1] performed in a deformation dilatometer. Use of post-deformation cooling made it possible to form in pipe steels a microstructure of bainitic ferrite (BF) and fine martensite-austenite "islands" (M/A-constituent). With stepwise accelerated cooling (AC) depending on soaking temperature the main type of M/A-structure were revealed: islands with predominance of twinned martensite (TM), islands with predominance of residual austenite and a mixed M/A-structure. It has been established that a reduction in pause temperature leads to a reduction in the proportion of martensite and an increase in the proportion of residual austenite within the composition of M/A-constituent areas. With a high pause temperature the steel microstructure ferritic matrix consisted of quasi-polygonal ferrite (QPF), and with a lower temperature it consisted of bainitic ferrite (BF), and with intermediate temperatures a mixture of QPF and BF. A procedure has been developed for studying the morphology and quantitative estimates of steel microstructure components containing a significant proportion of M/A-constituent.

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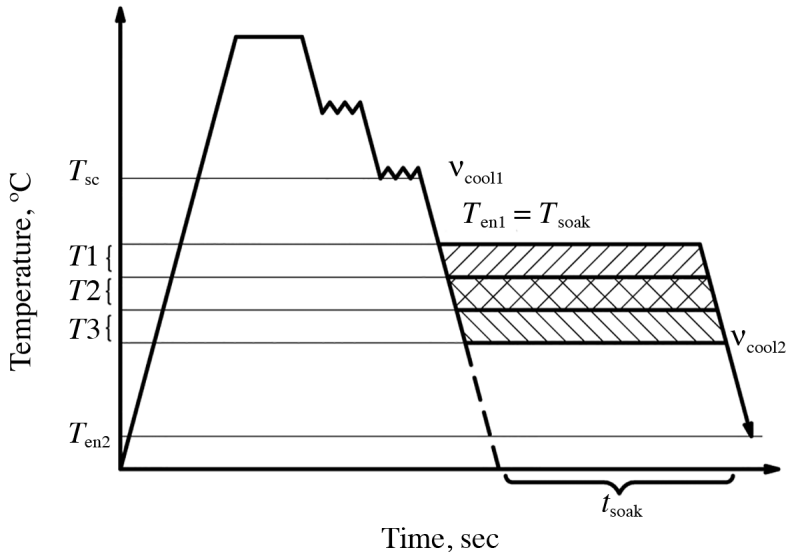


Fig. 1. Experimental scheme for obtaining different forms of M/A-constituent in steel microstructure ($T1$–$T3$ are soaking temperature intervals): T_{sc} is temperature for the start of accelerated cooling; ν_{cool1} and ν_{cool2} are rates of the 1st and 2nd cooling stages respectively; T_{en1} and T_{en2} are temperature for the 1st and 2nd cooling stages respectively. T_{soak} is soaking temperature; t_{soak} is soaking time.

The next task consisted of determining the connection between matrix and secondary phase structural parameters (in accordance with classifications [2–4]) and mechanical properties of low-alloy pipe steel with a significant proportion of M/A-constituent. With this in mind experiments were conducted for preparing workpieces of industrial steel K65 (X80) in a laboratory rolling mill using stepwise cooling. Specimen microstructure and mechanical properties were determined.

Research Procedure

Experiments in deformation dilatometer [1] have shown that for steel K65 with a high element content stabilizing austenite (Mn, Mo) with stepwise cooling the temperature range for M/A-constituent structure formation is more extensive than for steel K60. A reduction in carbon content in steel K65 (0.04% C) also promoted premature precipitation of cementite particles and formation of a carbide-free microstructure consisting of BF and M/A. Proceeding from this in order to perform rolling in a laboratory mill steel K65 produced by AO VMZ was selected, of the following chemical composition, wt.%: 0.04 C; 0.25 Si, 1.76 Mn, 0.006 P, 0.001 S, 0.67 Mo+Ni+Cu, 0.034 Al, 0.003 N_2, 0.07 Nb, 0.021 Ti, 0.003 V, C_{equ}= 0.42.

Rolling of sheet 12 mm thick was conducted in a DUO 300 laboratory mill. Workpieces 80 mm thick were cut from slabs of an industrial melt, heated in a furnace to 1170 °C, then two-stage deformation was performed with fixed reductions. Rolling ceased in the austenitic region with T_{er} = 800 °C, and the temperature for the start of accelerated cooling was 780–790 °C. Stepwise accelerated cooling was conducted in two stages with interruption of the first cooling stage at temperatures (T_{ec1}) in the range 400–650 °C with slow cooling between stages lasting 20–60 sec. The temperature for the end of the second stage of accelerated cooling was 150–250 °C, and the average cooling rate was 20–30 °C/sec. Control sheets were prepared by a similar regime with single-stage accelerated cooling. The experimental scheme is shown in Fig. 1.

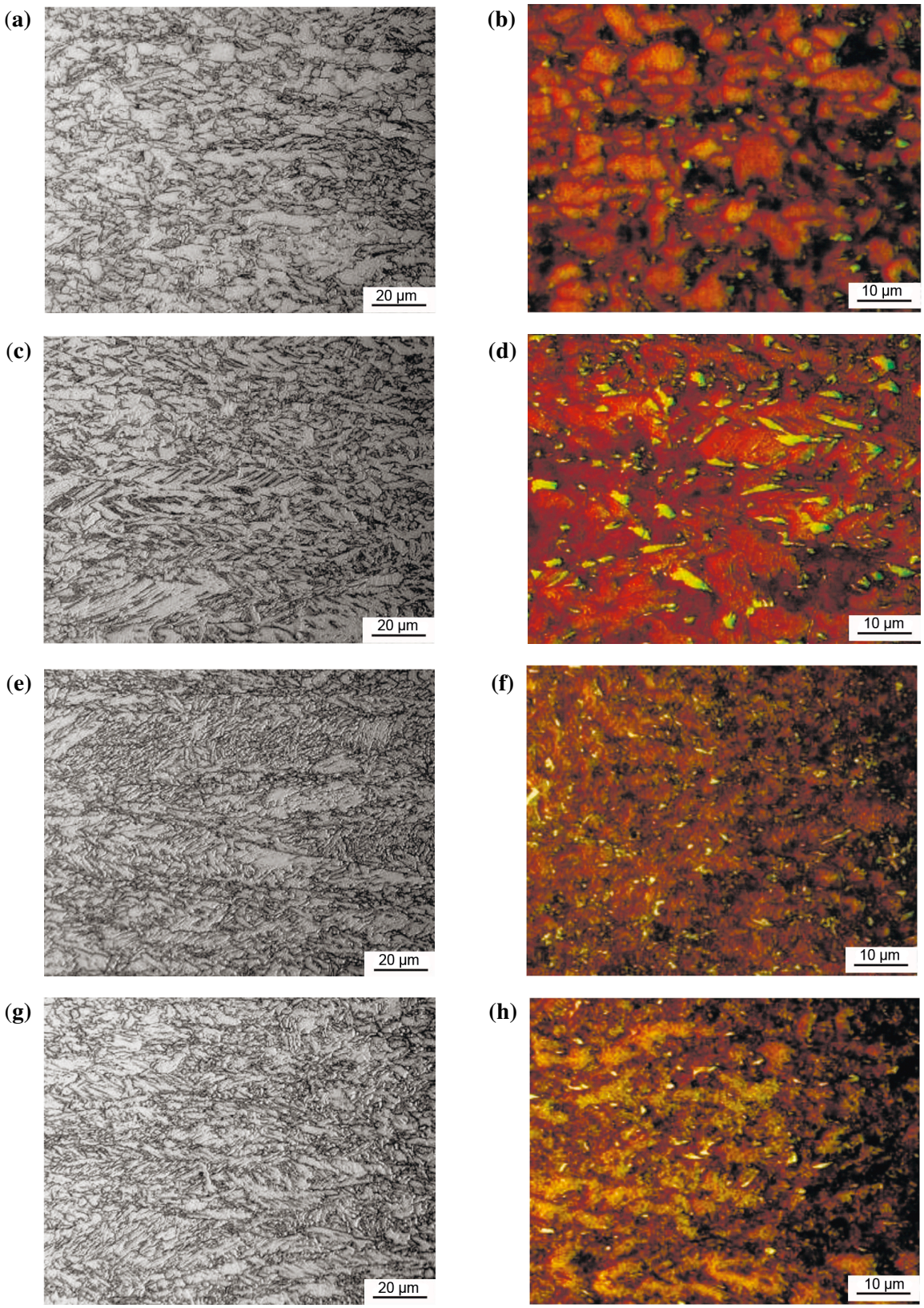
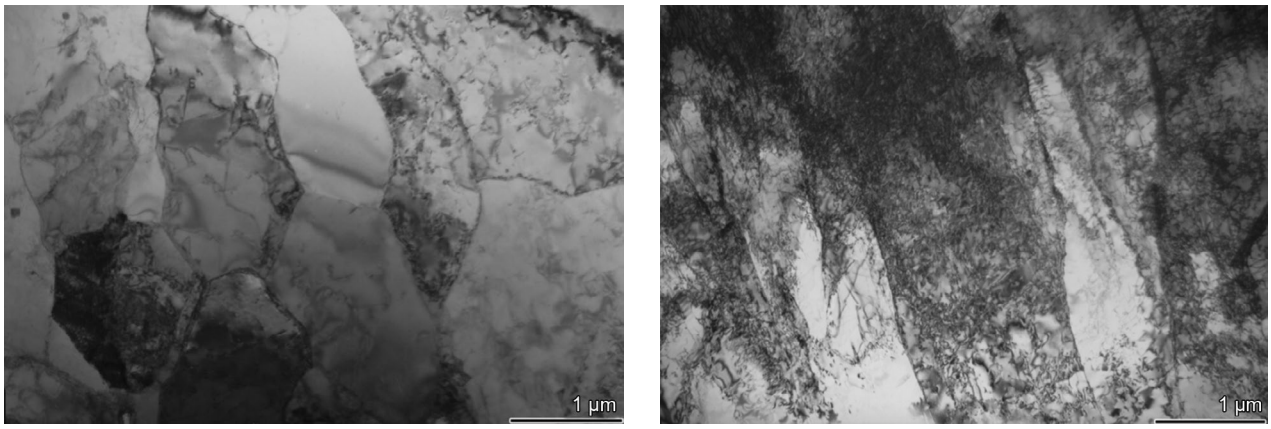


Fig. 2. Microstructure of steel K65 sheet in relation to cooling conditions (OM): (a), (c), (e), (g) etching in 4% nitric acid; (b), (d), (f), (h) etching in Le Pera reagent; sheets with T_{ec1} equal to 600°C (a), (b); 500°C (c), (d); 400°C (e), (f); 200°C (g), (h)



(a) **(b)**

Fig. 3. Rolled product matrix with high-temperature in *T*1 (a) and low-temperature soaking in *T*3 (b) (steel K65, TEM, light field, X15,000) (a) sheet with $T_{en1} = 600$ °C; (b) sheet with $T_{en1} = 400$ °C.

During an experiment values of temperature during rolling and accelerated cooling were recorded by means of an intellectual heat-sensor thermocouple (ATsP) ZET 7020 – ThermoTC – 485 with an RS–485 interface. Specimens were prepared from rolled strip for mechanical tests: in tension in a transverse direction (according to GOST 1497–84, specimen type III), for impact strength at test temperatures of –20, –40, –60, –80, –100 °C (*KC*V according to GOST 9454–78, type II). Tensile testing was performed in an Instron universal test machine, and impact testing was conducted in a pendulum unit 2010 KM-30.

Specimen microstructure of sheets was studied by a procedure developed by the authors described in the first article [1]: the proportion of M/A-structural component after light etching in Le Pera reagent was determined from photographs, and calculation was conducted by an ImageExpertPro 3 program, and then the TEM method was used to determine the type of M/A-structural component and the proportion of residual austenite was determined by an X-ray diffraction (XD) method.

Results and Discussion

The microstructure of sheets prepared in the laboratory mill entirely coincided with results of modelling in a deformation dilatometer [1]. Studies by light microscopy (Fig. 2) and TEM (Figs. 3, 4) confirm that sheet microstructure varies in relation the pause temperature range during stepwise cooling (see Fig. 2). Soaking at high temperature (T_{en1} = 550–650 °C) in the *T*1 range leads to formation of a mixed type matrix with predominance of QPF (see Fig. 2 a, b, Fig. 3 a), and the secondary phases observed are predominantly islands of twinned martensite, often with residual austenite (TM+A) (see Fig. 4 a, b). Color etching revealed that the proportion of M/A-structure with this treatment comprises 5–6% (see Fig. 2 b), TM is etched with dark color, and residual austenite is white. In this case the proportion of residual austenite within the structure determined by the XD method does not exceed 3%.

Soaking in the T2 range ($T_{en1} = 500 \pm 50$ °C) leads to an increase in the proportion of residual austenite within the composition of islands and to appearance of austenite (A) in pure from in the BF matrix (see Fig. 2 c, d). Also observed are coarse islands of M/A-constituent with a size of 6–10 μm represented predominantly by lath martensite, sometime in combination with twinned mar and austenite. From results of color etching and XD the proportion of M/A-constituent within the structure is about 7% of the proportion of residual austenite, i.e.,

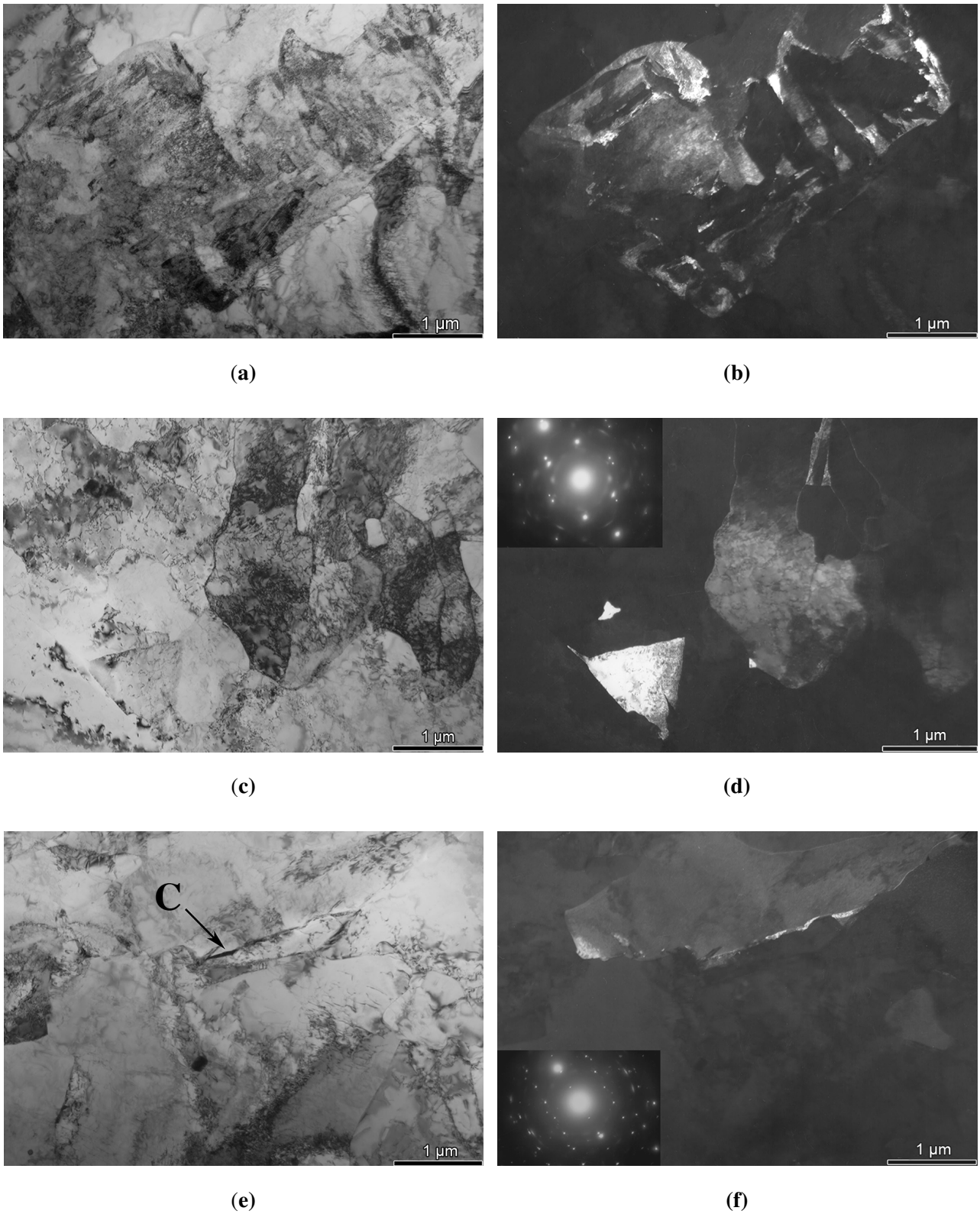
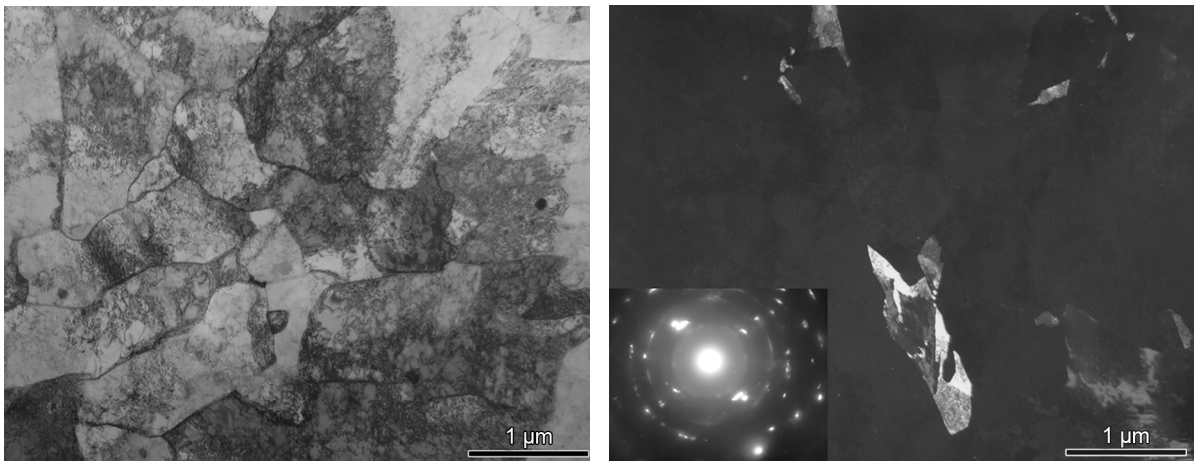


Fig. 4. Typical secondary phases formed within rolled product structure in relation to temperature T_{en1} (steel K65, TEM, ×15,000): (a), (c), (e)) light field; (b), (d)) dark field in austenite reflection; (f) in cementite reflection; (a), (b) twinned martensite (sheet with $T_{en1} = 600$°C); (c), (d) austenite; (e), (f) cementite (sheet with $T_{en1} = 400$°C).



(a) **(b)**

Fig. 5. Steel K65 microstructure (control sheet, T_{en1} = 200 °C), TEM, ×15,000: (a) bainitic ferrite with high dislocation density); (b) TM+A (dark field in austenite reflection).

about 2–3%. With low temperature soaking in the T3 range (T_{en1} = 450–400 °C) there is formation of a finer matrix structure of lath bainitic ferrite (LBF) (see Fig. 2 e, f, Fig. 3b) and relatively fine islands of residual austenite (see Fig. 4 c, d). In this case the M/A-constituent is represented mainly by residual austenite, and its proportion within the microstructure is ≈4–5%. Also with low-temperature soaking (T_{en1} = 400 °C) for areas of specimens fine cementite platelet precipitates are observed over the boundaries of ferrite laths (see Fig. 4 e, f).

The microstructure of control sheets (sew Fig. 2 g, h; Fig. 5) cooled without stepwise cooling (T_{en1} = 200 °C) consists of a matrix of fine bainitic ferrite of granular and lath type, and high-carbon phases, i.e., fine areas of M/A-structure: TM+A and residual austenite. The proportion of residual austenite determined by the XD method is about 4%. The main difference from sheet prepared during rolling with high-temperature soaking consists of a higher dislocation density in a ferrite matrix (see Fig. 5 a) and a smaller size for areas of high-carbon phases (see Fig. 5 b). According to TEM data areas of austenite on average have a size of less than 0.8 μm, and islands of TM + A of up to 2–3 μm.

Analysis of the effect of cooling regime on microstructure makes it possible to confirm that the difference ferrite crystal shape (QPF or LBF) and type of M/A-constituent is connected with the temperature T_{en1}, determining the intensity of diffusion processes during stepwise cooling. At T_{en} temperatures of about 600 °C there is formation of ferrite crystals whose shape is close to equiaxed. Carbon is redistributed intensively in zones of incomplete bainitic transformation, saturating relatively large areas of austenite up to a concentration satisfactory for implementing martensitic transformation in the second cooling phase. Individual areas are enriched with carbon to a greater extent (possibly during transformation). Stable austenite is retained within and over TM areas. With low T_{en1} temperature (450 °C and lower) due to retardation of diffusion processes carbon redistribution is difficult, i.e., there is formation of bainitic ferrite of lath morphology. In this case during soaking carbon is expelled into ferrite grain boundaries and laths forming areas and layers of continuous austenite that during soaking coarsen and form relatively coarse islands (with size of > 1–2 μm).

In the absence of soaking austenitic areas do not manage to coarsen, and it is retained in the form of fine areas (with a size of < 1 μm) within bainitic ferrite. At intermediate temperatures (T_{en1} ≈ 500–550 °C) there is formation of both areas of TM and also austenitic areas, and in this case in zones of QBF matrix is predominance of M/A-constituent in the form of TM. In zones of the BF matrix residual austenite is often encountered.

Table 1
Steel K65 Sheet Mechanical Properties (Average Values), Microstructure Type, Proportion of M/A-Constituent (Color Etching, Optical Microscopy, Proportion of Residual Austenite (XD) in Relation to Cooling Regime

Structure type in relation to soaking T	σ_f, MPa	σ_y, MPa	σ_y/σ_f	δ_5, %	δ_p, %	$KCV^{(-40/-80)}$, J/cm^2	M/A and A proportion (TM/XD), %
Soaking in T_1 QPF+(TM+A)+A	670	530	0.80	26	11	368/360	6.5/3.0
Soaking in T_2 QPF/LBF+(TM+A)+A	720	550	0.75	26	12	368/260	5.3/2.1
Soaking in T_3 LBF+A	742	545	0.74	26	12	340/220	4.2/4.1
Without soaking (control specimen) LBF+A+(TM+A)	707	570	0.81	23	9	368/300	0.6/4.0

Formation of cementite (sometimes observed in the form of thin platelets at boundaries with low-temperature soaking) occurs with breakdown of austenite with a lower carbon content than in stable areas [5], or during prolonged soaking.

With relatively fast cooling rates, with which pearlitic transformation is suppressed, in the bainitic transformation temperature range at high temperature there is formation of QBF within the matrix, and at ow temperature there is LBF [4, 6, 7]. As is well known, austenite stability within steel during cooling depends on carbon content [5, 8, 9]. A study of the microstructure (in particular almost always presence of QBF is observed around coarse M/A-constituent areas) makes it possible to conclude that as a result of carbon redistribution in areas of untransformed austenite during prolonged soaking ferritic transformation in the vicinity of M/A-constituent areas proceeds more rapidly and is completed almost entirely at the T_{en1} temperature. Therefore, soaking facilitates formation of higher temperature (i.e., with lower dislocation density) [10] forms of ferrite than those that could form with continuous cooling (for example QBF instead of LBF).

Comparison of mechanical property test results with metallographic data and studies by an X-ray diffractometry (XD) showed that strength properties of rolled sheet metal with static tension are determined mainly by the type of matrix structure obtained (Table 1).

The microstructure obtained with T_{en1} in the range $T1$, consisting mainly of QBF with a small proportion of M/A-constituent islands, is represented mainly by TM with some proportion of austenite over the periphery, and lower strength is provided. At the end of the first cooling stage in the $T2$ range a mixed matrix of QBF and LBF and relative coarse islands of TM with fine areas of austenite provides a higher level of strength properties and a reduction of values of σ_y/σ_f ratio with retention of good ductility. With cooling in the $T3$ range the matrix of lath bainite and islands of residual austenite provides even higher values of ultimate strength, lower σ_y/σ_f ratio without a reduction in values of relative, and uniform elongation. Control sheets with a structure of

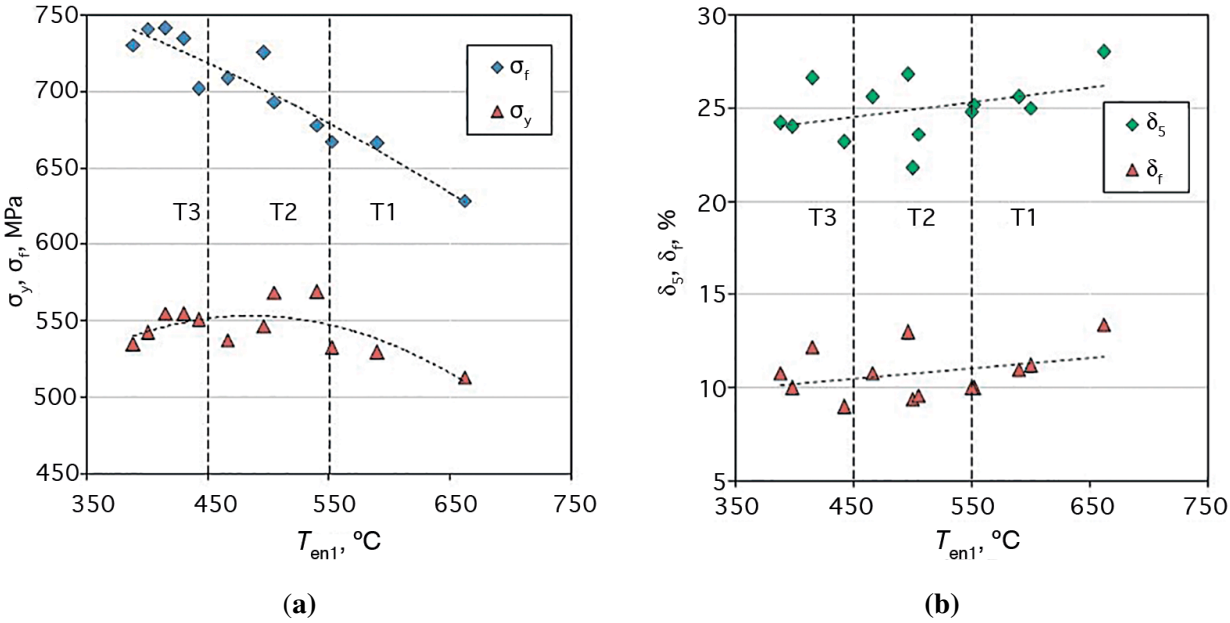


Fig. 6. Dependence of strength properties (a) and ductility (b) on T_{en1} temperature for steel K65 sheets.

low-carbon bainite, consisting of LBF with layers of residual austenite, has lower strength, but better yield strength and lower ductility, which is mainly connected with a low content of string martensite within the structure and a higher dislocation density in LBF. The dependence of strength properties and ductility for sheet on T_{en1} temperature, which determines the type of microstructure (primarily the matrix) is provided in Fig. 6. It is evident that ultimate strength is connected to the greatest extent with T_{en1}. The dependence of yield strength (see Fig. 6a) and values of relative and uniform elongation (see Fig. 6b) are insignificantly defined by this parameter.

Analysis of the effect of steel microstructure type on the ratio of mechanical properties of steel K65 sheets is given in Fig. 7. The ratio of yield and ultimate strengths varies and is dependent on the type of microstructure. The predominant low dislocation quasipolygonal ferrite (QPF) within the composition of ferritic matrix and high-carbon phase in the form of coarse (up to 6–8 μm) of TM+A islands, with a high proportion of twinned martensite in the steel structure with T_{en1} in the $T1$ range, provides a relatively low level of σ_f, ($\approx$ 30–690 MPa) and in this case σ_y is relatively high ($\approx$ 510–570 MPa). Control sheets with a structure of granular and lath bainitic ferrite with high dislocation density and secondary high-carbon phase, consisting of fine (1–2 μm) and islands of TM+A or A, had a higher level of σ_f ($\approx$ 700–745 MPa) and a high value of σ_y ($\approx$ 550–590 MPa).

Sheets with T_{en1} in the temperature ranges $T2$ and $T3$ had a structure for the matrix consisting predominantly of LBF with high dislocation density, and also a smaller proportion of finely dislocated QPF, often encountered alongside high-carbon phase of coarser, thin in control sheets, areas of TM+A and austenite in pure form (A), had a higher level of σ_f ($\approx$ 705–745 MPa). In this case the value of σ_y was relatively lower than for preceding cases ($\approx$ 535–555 MPa). Therefore, the structure of the type LBF+QBF+(TM+A)+A provides a more favorable value of the ratio σ_y/σ_f.

For sheets with a LBF+QBF+(TM+A)+A and LBF+(TM+A)+A structures the effect of level of σ_f on values of relative (δ_5) and also uniform (δ_u) elongation is followed clearly (see fig. 7 b, c). As strength increases

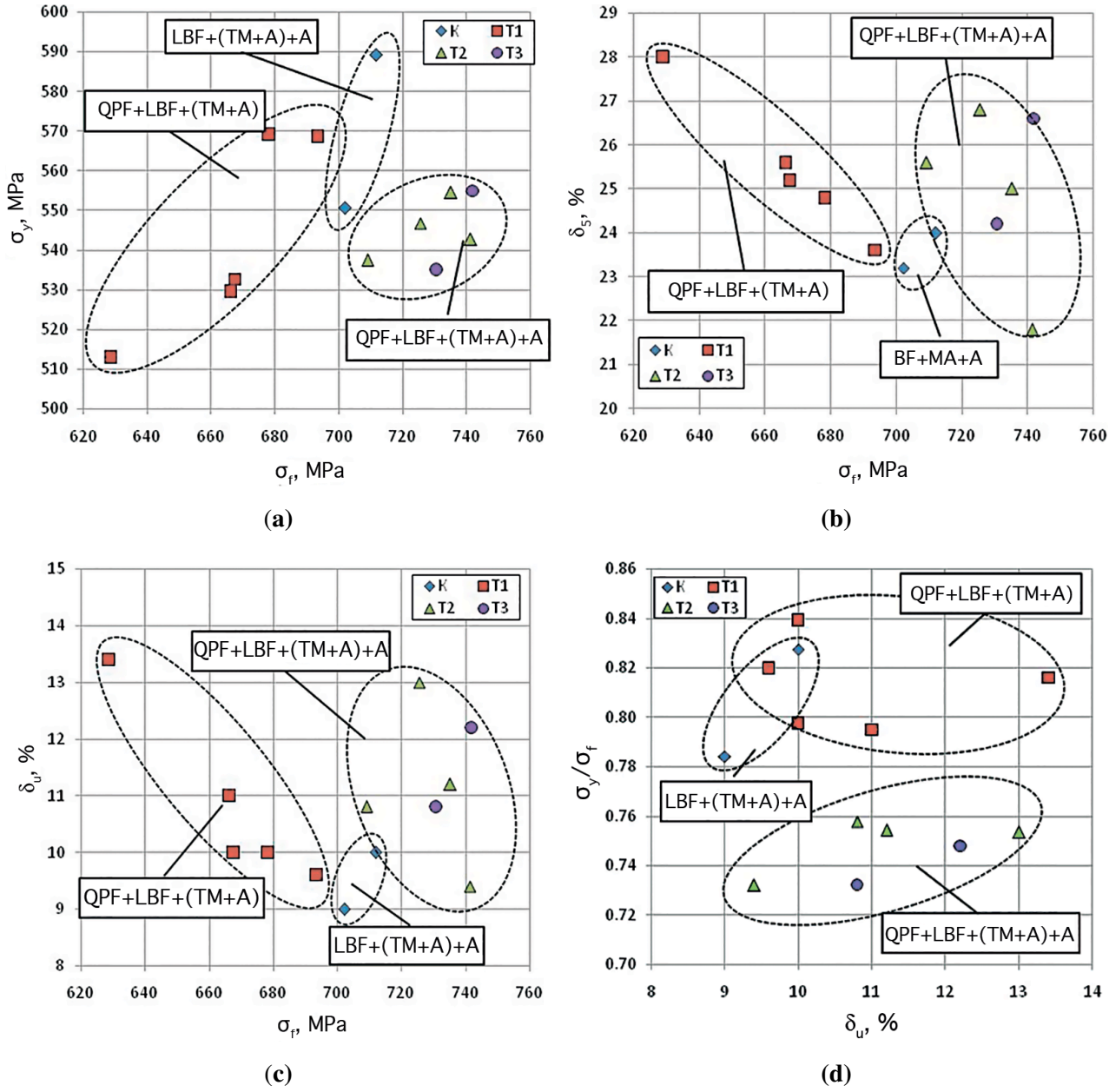


Fig. 7. Effect of steel microstructure type on steel K65 sheet mechanical property ratio: (a) σ_y and σ_f; (b) δ_5 and σ_f; (c) δ_u and σ_f; (d) σ_y/σ_f and δ_u.

ductility properties decrease uniformly almost linearly. With values of $\sigma_f \approx 630$ MPa δ_5 and δ_u are 28 and 13.4% respectively. An increase in σ_f to $\approx$690–715 MPa leads to a redu ction in the values of δ_5 to 23–24% and δ_u to 9–10%. However, sheets with a structure of the LBF+QBF+(TM+A)+A type, in spite of better strength, demonstrate relatively greater ductility. With values of σ_f from 705 to 745 MPa the value of δ_5 is 21.8–27%, and δ_u is 9.5–13%.

The connection is shown in Fig. 7 d of the ratio σ_y/σ_f and uniform elongation δ_u, important from the point of view of providing improved metal deformation capacity. With approximately equal values of δ_u (9–13%)

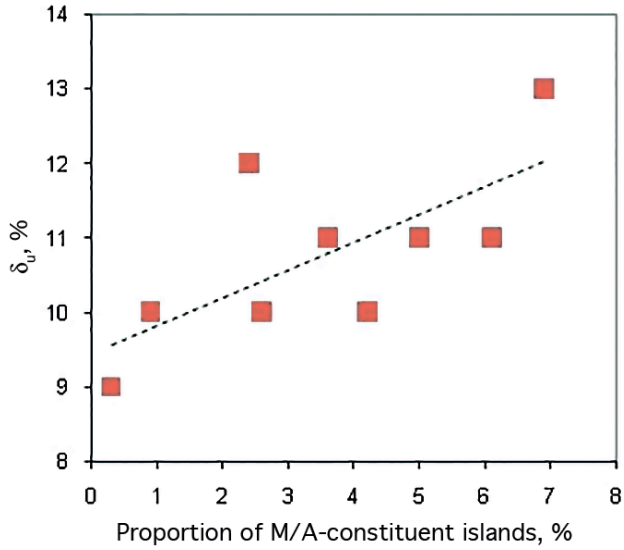


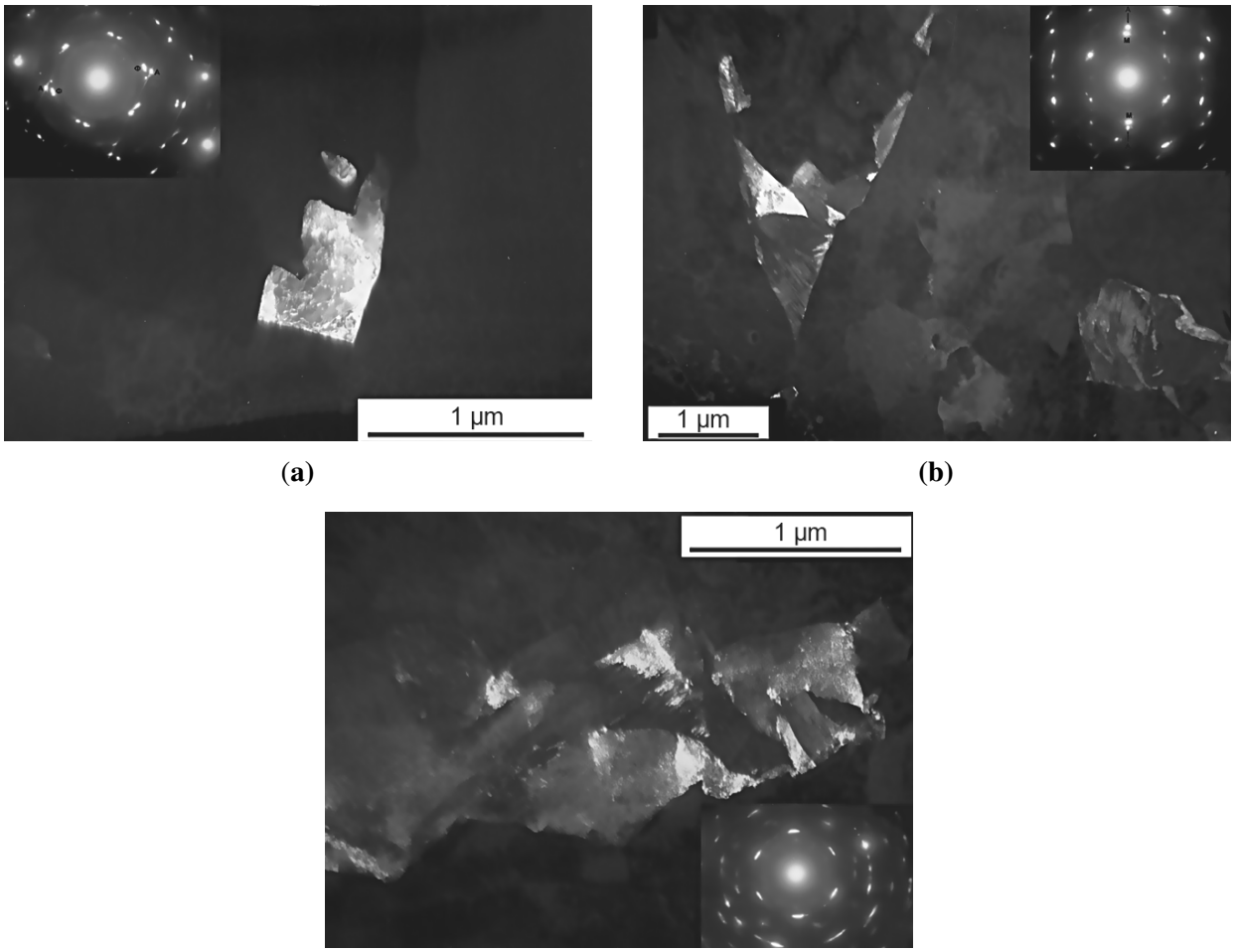
Fig. 8. Effect of proportion of M/A-constituent areas (according to OM data) in microstructure on steel K65 specimen uniform elongation.

obtained for specimens of rolled product with a different type of microstructure, the value of the σ_y/σ_f ratio for sheets with a LBF+QBF+(TM+A) and QBF+(TM+A)+A microstructure comprises 0.78–0.84, and for sheets with a structure of the type LBF+QBF+(TM+A)+A it is 0.73–0.76. Therefore, sheets with a bainitic ferrite structure with a smaller amount of QBF and medium-dispersed high-carbon phase in the form of islands (TM+A) and A demonstrate the most favorable of mechanical properties: high strength, low σ_y/σ_f ratio, and a high (with respect to the ultimate strength value) level of relative and uniform elongation, which confirms data obtained by other researchers [11–13]. It has been noted that specimens with T_{en1} in the range $T2$ to $T3$ are distinguished by a better strength/ductility ratio than for control specimens and specimens with T_{en1} in the range $T1$.

An increase in the proportion of M/A-constituent areas within the microstructure by color etching (size more than 1–2 μm) has a favorable effect on steel ductility properties, in particular on the value of δ_u (Fig. 8).

In this case presence of residual austenite [14] does not always facilitate an increase in ductility properties, for example ≈4% residual austenite was determined by the XD method in control specimens with very low values of ductility properties. In order to improve ductility properties it is important to have within the microstructure adequate coarse areas of M/A-constituent (size more than 1–2 μm) and a ferritic matrix with dislocation density. A favorable effect of a significant proportion of M/A-constituent on the ductility properties of steel may be explained not only by a reduction in dislocation density within a ferrite matrix and formation of high-temperature forms of ferrite due to additional carbon redistribution within solid structures during soaking, but also by the influence of the TRIP-effect during deformation of steel containing residual austenite [15].

Results of studying the effect of deformation on the phase composition of the M/A-constituent point to the occurrence in areas of M/A-constituent of martensitic transformation, initiated by deformation. Specimens were selected for the study from groups $T2$ and $T3$ with an original proportion of coarse areas of about 5%. Studies were performed by the TEM method. Foil for study was prepared from failed specimens after tensile testing. One foil was cut from an area of each specimen closely adjacent to a fracture, and another was prepared from the same specimens at a distance of 12–13 mm from a fracture.



(a) **(b)**

(c)

Fig. 9. Secondary (high-carbon) phases in steel K65 specimen, TEM: (a) austenite island (dark field in austenite reflection), ×30,000; (b) TM + A islands (at distance of 12–13 mm from fracture, dark field in austenite reflection), ×15,000; (c) twinned martensite (specimen below fracture, dark field in martensite reflection), ×30,000.

These studies have shown that there are differences in the M/A-structure observed and in relation to the area of foil cutting. In specimens cut at a distance of 12–13 mm from a fracture austenite islands are often observed (Fig. 9 a), and austenite within the composition of islands with high-carbon there is twinned martensite (see Fig. 9 b). In specimens prepared from an area of metal immediately above a fracture austenite is not detected, neither within the composition of twinned martensite, nor in pure form among ferrite (see Fig. 9 c). In a diffraction pattern (see Fig. 9 c) only martensite reflections are seen, and austenite reflections are absent. Part of martensite blocks in this island have an orientation [1 1 1] with some misorientation, appearing in tangential blurred martensite reflections, which is apparently a consequence of deformation and shows that the martensite observed appeared during formation of specimen metal during occurrence of mechanical tests.

Results of studying impact strength and cold resistance of sheets of steel K65 are provided in Fig. 10. All of the sheets prepared in a laboratory rolling mill had a high level of impact strength with a test temperatures of -40 and -60 °C (on average not less than 300 J/cm^2). The average proportion of ductile component (PDC) was not less than 90%. Depending on cooling regime (T_{en1}) a different nature of change was observed for impact strength and PDC at reduced temperature. For sheets with T_{en1} in the range $T1$ (i.e., predominance within

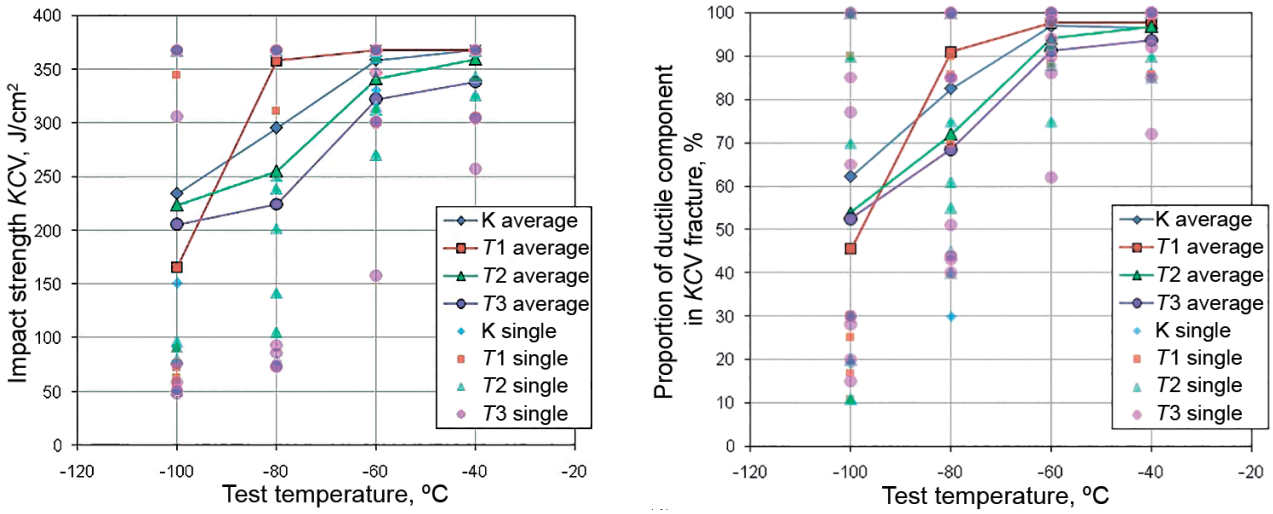


Fig. 10. Impact strength (a) and proportion of ductile component in *KC*V steel K65 specimens (b) prepared by different regimes and tested at different temperatures: $T1$ – T_{en1} = 650–550 °C; $T2$ – T_{en1} = 550–500 °C; $T3$ – T_{en1} = 450–400 °C; K is control specimen without soaking.

the matrix microstructure of QPF and secondary phase (TM+A) with predominance of TM) higher values of *KC*V and PDC were recorded at temperatures to –80 °C, but a sharp reduction was observed at –100 °C, and the cold brittleness threshold T_{50} was –95 °C. With a reduction in T_{en1} in the range $T2$ the matrix microstructure consists of QPF and LBF, secondary phase (TM+A) and A, average values of *KC*V and PDC in the test temperature range –40 and –80 °C are somewhat reduced, but a sharp reduction in properties at a test temperature of −100°C was not observed. The calculated T_{50} temperature was about –105 °C. Sheets with T_{en1} in the range $T3$ (matrix microstructure predominantly LBF, secondary phase islands of A) have lower average *KC*V values and the calculated T_{50} temperature also comprised abut – 105 °C. The lowest calculated T_{50} temperature (about –115 °C) applied to control sheets with an LBF structure with layers of A and fine islands (TM+A). The results obtained confirm the well-known dependence of the effect of structural component sizes on cold resistance [8]. In this case the nature of the reduction in series cold resistance curves for metal with a lath type matrix is flatter than for metal with a granular matrix.

In the course of work a study was made of the effect of the proportion of residual austenite (according to XD data) on the value of impact strength for sheets at temperatures from –60 °C to –100 °C (Fig. 11). It is apparent that with presence of residual austenite in an amount of more than 4% a reduction is observed in impact strength at –80 °C and –100 °C, which agrees with data of other researchers [9, 16] about the unfavorable effect of M/A-constituent on cold resistance. However, it should be noted that the standard test temperature for pipe steels in Charpy specimens is –20 °C, for cold resistance –40 °C, and therefore this effect for fine-grained pipe steels is unimportant.

The possibility was also considered of austenite transformation into martensite under action of negative temperatures during specimen cooling. Sheet was selected for the study with T_{en1} in the range $T2$ with a proportion of residual austenite at room temperature of about 4%, from which specimens were selected that were subjected to cold treatment by immersion in a cooling medium (alcohol cooled with liquid nitrogen) for 15 min (at temperatures from –40 °C to –100 °C) and liquid nitrogen for 45 min. After cold treatment in specimens the austenite content was measured by the XD method. This study showed a weak effect of cold treatment on the proportion of austenite in specimens (Table 2).

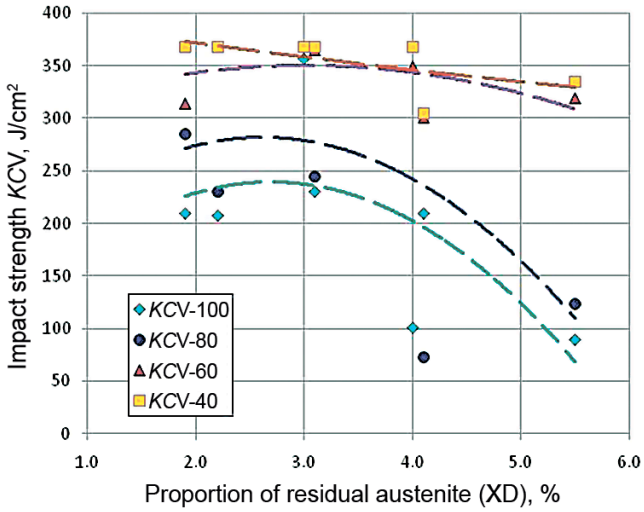


Fig. 11. Effect of residual austenite volume fraction on *KC*V impact strength at reduced temperature.

Table 2
Proportion of Residual Austenite Determined by XD Method in Steel K65 Specimens at Different Temperatures

Temperature, °C	A proportion, %	Soaking, min
+20	4.2	15
–40	3.8	15
–80	3.6	15
–100	3.9	15
–196	3.1	45

Dependences established for mechanical properties on steel K65 microstructure, obtained with variation of modified cooling regimes (see Figs. 7, 8, 10) made it possible to draw conclusions about the effect of different types of structural component on steel mechanical properties and cold resistance. The type of ferritic matrix, determined by the transformation temperature, affects steel properties as a result of a change in dislocation density and grain size (or laths) [5, 8, 9]. As there is a change in the type of matrix (QPF → GBF → LBF) there is an increase in dislocation density, σ_y and σ_f increase (and σ_y increases more rapidly), and ductility is reduced; with refinement of structural components there is a reduction in $T50$ transition temperature. Hard high-carbon components within the microstructure in the form of M/A ($HV10g$ = 420–520) [17] increase steel strength. At the same time as a result of an increase in temperature for ferritic grain transformation connected with M/A-constituent particles (due to transfer of carbon into untransformed austenite) and formation of forms of ferrite with lower dislocation density, there is an increase in matrix ductility. In addition, austenite within the M/A-structure composition also may increase ductility properties somewhat as a result of the TRIP-effect.

CONCLUSION

Experiments performed in a laboratory mill have confirmed the principle of structure formation with use of two-stage AC similar to that prescribed in [1]. A study of the effect of cooling regime on mechanical properties and microstructure has shown that performance of conditionally isothermal soaking in the temperature range 390–500 °C with the optimum pause time rolled product specimens are obtained with a structure consisting of lath, granular and quasi-polygonal ferrite with islands (TM + A) and/or residual austenite. The proportion of austenite according to XD results is about 4%. Specimens with this microstructure exhibit improved ductility properties (δ_5 higher by 2–4%, abs.) with a higher level of ultimate strength (by 25–40 MPa) compared with control specimens.

The level of sheet mechanical properties with the optimum microstructure: δ_5 = 25–27%, δ_u = 12–13%, σ_f = 725–740 MPa, $\sigma_y/\sigma_f \le 0.76$, impact strength $KCV^{-60} \ge 300$ J/cm^2.

The principles in the article for preparing pipe steels with a significant proportion of M/A-constituent within the microstructure have already passed industrial approval under AO VMZ conditions [18], and they may be used for producing rolled product with good deformation capacity and improved strength and ductility ratio.

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