DOI 10.1007/s11041-023-00842-0
Metal Science and Heat Treatment, Vol. 64, Nos. 9 – 10, January, 2023 (Russian Original Nos. 9 – 10, September – October, 2022)

UDC 669.018.298:669.131.4:620.172:620.181

EFFECT OF THERMOMECHANICAL TREATMENT ON THE STRUCTURE AND MECHANICAL PROPERTIES OF PIPE STEEL 38G2F

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Translated from *Metallovedenie i Termicheskaya Obrabotka Metallov*, No. 9, pp. 23 – 28, September, 2022.

Original article submitted May 16, 2022.

The effect of rolling modes and cooling conditions on the structure and mechanical properties medium-carbon manganese low-alloy steels of type 38G2F is investigated. The critical temperatures Ac_1 and Ac_3 are determined. The characteristics of the ferrite, pearlite and bainite structural components (the content and the morphology) and the sizes of the initial austenite grains in pipes with different wall thicknesses are determined. Tensile mechanical tests are performed. It is shown that the high level of mechanical properties ($\sigma_{0.2} > 650$ MPa, $\sigma_r > 900$ MPa) of the pipes with wall thickness 16.0 mm is a result of the presence of bainitic component in the structure. It is suggested that the temperature of heating of the pipe billet for rolling should be reduced from 1230 – 1260°C to 1170 – 1180°C for manufacturing pipes with a size of ∅93.2 × 13.0 mm and ∅127 × 16.0 mm. The combination of their mechanical properties matches the E strength group.

Key words: seamless tubing and casing, medium-carbon microalloyed steels, thermomechanical treatment, mechanical properties, ferrite-pearlite structure.

INTRODUCTION

Development of the oil and gas industry and of new deposits with various conditions of oil and gas extraction demand a wide range of tube products. Hot-rolled tubing and casing of commercial quality, of strength group *D* in particular, when the only controlled parameters are the strength properties, remain in demand until today [1, 2]. At the same time, producers report an increasing interest in oil and gas tubes from steels with enhanced strength level (for example, of group *E*), which are expected to possess a high impact toughness and crack resistance in addition to the high strength characteristics. This requires the presence of low-temperature products of decomposition of supercooled austenite in the structure or conduction of a hardening and tempering treatment.

In the case of thermomechanical treatment (TMT) of tubes from low- and medium-carbon steels microalloyed with strong carbide-forming elements (Nb, V, Ti) the behavior of special carbides of type MC is very important [3 – 5]. In the case of precipitation of MC carbides at a temperature above that of the start of recrystallization of austenite t_{sr}, their role is to suppress the dynamic and static recrystallization for refining the structure of the ready tube. This approach is implemented in controlled rolling of low-carbon steels of type 05G2MFBT (strength class X80) and produces a fine dispersed ferrite-bainite structure (with grain size $d_{f\text{-}b} \sim 3 – 5$ μm) with a unique combination of mechanical properties [6, 7].

Microadditions of Nb, V and Ti in medium-carbon steels used for making hot-rolled seamless tubes serve primarily for raising the yield strength due to precipitation hardening [8, 9]. In addition, the process parameters (the temperature of heating of the tube billet, the temperature and the strain in different rolling stages) are varied to control the structural and phase state of the austenite, i.e., the grain and/or subgrain size, the presence of particles of second phases, the density of crystal lattice flaws required for attainment of the specified level of mechanical properties in the tube in post-deformation air cooling.

The recent researches performed at the Volzhsky Pipe Plant (VPP) in the production of hot-rolled seamless tubes of strength group *E* involve two directions, i.e., (*1*) develop-

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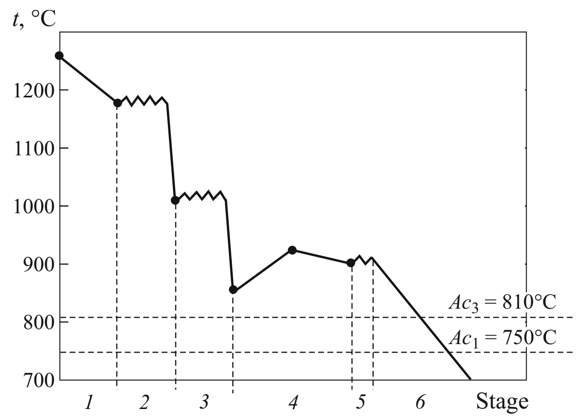


Fig. 1. Temperature distribution during rolling of tubes from steel 38G2F by the main mode in the following stages: *1*) heating of the tube billet; *2*) broaching of the billet; *3*) expanding; *4*) furnace heating of the tube and its transfer to the sizing mill; *5*) sizing of the tube; *6*) post-deformation air cooling at a rate of 0.5 – 1.0 K/sec.

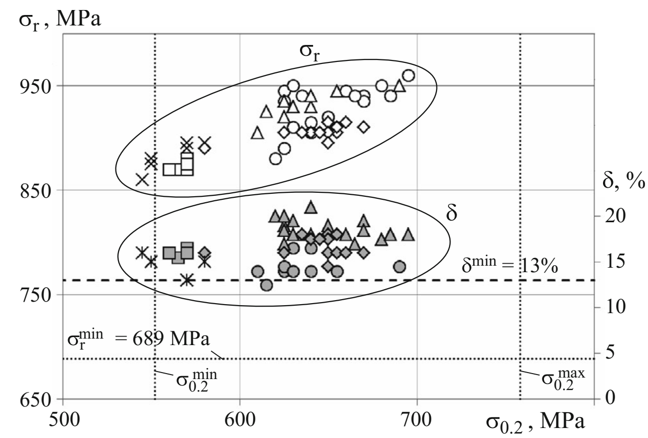


Fig. 2. Mechanical properties of tubes from steel 38G2F and the requirements they should meet to match strength group *E* (rolling by variant *2*): ○, ●) wall thickness 8.9 mm; △, ▲) 12 mm; ◇, ◆) 16.0 mm; □, ■) 20.0 mm; ×, ✳) 25.0 mm.

ment sparingly alloyed compositions of medium-carbon steels and (*2*) optimization of the parameters of TMT (the temperature of heating of the tube billet, the temperatures in the final deformation stages, the intensities of the post-deformation cooling). The interest in these directions is explainable by the possibilities of the VPP to produce tubes of this group with a larger diameter and wall thickness as compared to other productions of the "Tube Metallurgical Company."

The aim of the present work was to generalize the results of the studies of the effect of rolling modes on the structure and mechanical properties of hot-rolled tubing and casing from medium-carbon steels of type 38G2F.

METHODS OF STUDY

We studied the structure and mechanical properties of the metal of tubes of various standard sizes (∅93.2 × 13.0, ∅168.3 × 8.9, ∅110 × 12.5, ∅88.9 × 13.0, ∅127 × 16.0, ∅127 × 20.0, and ∅203 × 25.0 mm) produced by different modes of TMT at the VPP during 2017 – 2020. The chemical compositions of steel 38G2F from the test heats are presented in Table 1.

The main mode of tube rolling (mode *1*) was the one with heating of the tube before the sizing mill in a walking-beam furnace or in an induction device. By the data of the temperature control, the actual sizing temperature of the tubes to $\varepsilon \sim 6\%$ in the TMT by mode *1* was $t_s \sim 890 – 910$°C (Fig. 1).

A special feature of the TMT of tubes at the VPP is absence of recrystallization of their structure before sizing and low strain ($\varepsilon < 10\%$), which narrows substantially the possibility of reduction of the size of austenite grains d_a . To refine the grains, the temperature of heating of the tube billet t_h was lowered from 1230 – 1260°C to 1170 – 1180°C (mode *2*). The value of d_a was determined in the tubes after each operation of hot deformation, i.e., piercing, expanding and sizing. The rate of the post-deformation cooling did not exceed 0.5 – 1.0 K/sec for all the studied standard tube sizes.

The critical points (Ac_1 and Ac_3) in the heating were determined from the curves plotted with the help of a Linseis RITAL78 dilatometer. The content and the morphology of the structural components (ferrite, pearlite, bainite) over the wall thickness and the size of the initial austenite grains d_a were determined by the ferrite mesh method according to GOST 5639–82 using an OLYMPUSJX 51 optical microscope at different magnifications. To uncover the structure, the laps were etched in a 4% solution of nitric acid in ethyl alcohol. The tensile tests were conducted at room temperature for five-fold cylindrical specimens cut from the middle of the tube wall in accordance with GOST 1497–84.

TABLE 1. Chemical Compositions of Steel 38G2F from Different Heats

Heat	Content of elements, wt.%								
	C	Mn	Si	Cr	Ni	Cu	V	P	S
1	0.36	1.45	0.21	0.15	0.11	0.13	0.11	0.013	0.007
2	0.37	1.33	0.45	0.21	0.12	0.18	0.11	0.013	0.005
3	0.42	1.30	0.35	0.15	0.12	0.18	0.09	0.013	0.007

RESULTS AND DISCUSSION

The mechanical properties of the metal of tubes of various standard sizes after the rolling by variant *1* and the requirements for them for strength group *E* are presented in Fig. 2. It can be seen that the strength properties of the tubes with wall thickness $h \le 16.0$ mm are much higher than the minimum values required by the Standard ($\sigma_r^{min} = 552$ MPa, $\sigma_{0.2}^{min} = 689$ MPa); the margin for the yield strength $\Delta\sigma_{0.2} =$

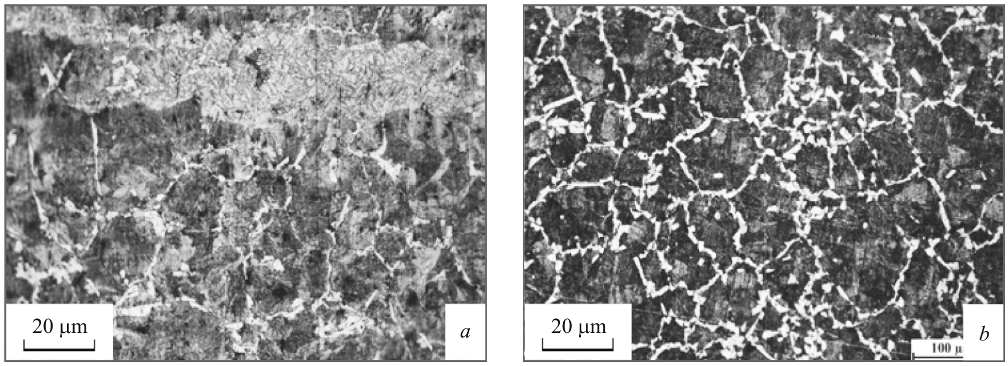


Fig. 3. Microstructure of the tubes with sizes ∅110 × 12.5 mm (*a*) and ∅127 × 16.0 mm (*b*) from steel 37G2F (heat *2*) rolled by variant *1* (optical microscopy).

$\sigma_{0.2} - \sigma_{0.2}^{\min} \sim 60 - 140$ MPa; for the ultimate strength it is even higher, i.e., $\Delta\sigma_r = \sigma_r - \sigma_r^{\min} \sim 190 - 270$ MPa. At the same time, for the tubes with $h > 16.0$ mm the strength margin $\Delta\sigma_{0.2} \sim 20 - 25$ MPa is at the minimum required level, though the margin for σ_r is high enough (σ_r = 860 – 890 MPa).

The variation of the elongation δ is virtually independent of the thickness of the wall; a low ductility margin $\Delta\delta = \delta - \delta^{\min} \sim 3 - 6\%$ is detectable in tubes of all the standard sizes except for ∅165.3 × 8.9 mm, for which the value of Δδ is the bottom required level ($\delta^{\min} = 13\%$).

The results of the metallographic analysis show the presence of low-temperature products of decomposition of supercooled austenite even in the tube with great wall thickness after post-deformation air cooling (see Fig. 3*a*). The appearance of these large extended regions of bainite components ($l > 500$ μm) occurring primarily near the inner surface of the tube (marked by the arrow in Fig. 3*a*) is explainable by the low reduction in sizing ($\varepsilon \le 6\%$), which is responsible for nonuniformity of the plastic strain over the wall of the tube).

At the same time, the structure of the steel near the outer surface and in the middle of the wall is represented by a ferrite-pearlite mixture in which 10 – 15% of the excess ferrite segregates in the form of individual grains and a fine closed net with a thickness of 5 – 10 μm over the boundaries of former austenite grains with size d_a = 60 – 65 μm (Fig. 3*b*).

The hardness $HB_{\text{f-p}}$ of the steel with such a structure can be assessed using the mixing rule, i.e.,

$$HB_{\text{f-p}} = q_f HB_f + q_p HB_p , \qquad (1)$$

where $q_f = 0.15$ and $q_p = 0.85$ are the volume fractions of ferrite and pearlite and HB_f and HB_p are the values of their hardness. The ferrite reinforced with particles of VC vanadium carbide has a hardness $HV_f = 210$ kgf/mm^2 [10]; the pearlite has a hardness $HB_p = 280$ kgf/mm^2 [10], which gives $HB_{\text{f-p}} \sim 270$ kgf/mm^2.

It is known that steels obey a direct correlation between the hardness *HB* and the ultimate strength σ_r [11], i.e.,

$$HB \sim 3\sigma_r . \qquad (2)$$

For steel 38G2F in the structural state described $\sigma_r^{\text{f-p}} = 900$ MPa. The state with $\sigma_r \ge 900$ MPa means that the ferrite-pearlite structure acquires low-temperature (bainitic) products of decomposition of supercooled austenite with a level of σ_r (i.e., the hardness *HB*) higher than that of pearlite.

The joint analysis of the structure and mechanical properties of tubes with different wall thicknesses has shown that the yield strength $\sigma_{0.2}$ is the limiting property for strength group *E*. Elevation of the strength properties (the strength margins $\sigma_{0.2} > 60$ and $\sigma_r > 900$ MPa) of the tubes with wall thickness $h \le 16.0$ mm is a result of the appearance of bainitic component in the ferrite-pearlite structure. Formation of bainitic products of decomposition of supercooled austenite, especially of upper bainite, causes lowering of the resistance of the tube to brittle fracture, which makes it necessary to estimate the contributions of different hardening mechanisms into the yield strength of steel 38G2F with ferrite-pearlite structure.

Data on these contributions are well known for low-carbon (up to about 0.2% C) alloy steels with ferrite-pearlite structure [12], in which the value of $\sigma_{0.2}$ is raised by different kinds of hardening, i.e., solid-solution $\Delta\sigma_{\text{s.s}}$, dislocation $\Delta\sigma_d$, precipitation $\Delta\sigma_{\text{p.h}}$, grain-boundary $\Delta\sigma_g$, and aggregate (pearlitic) $\Delta\sigma_p$ ones.

$$\sigma_{0.2} = \sigma_0 + \Delta\sigma_{\text{s.s}} + \Delta\sigma_d + \Delta\sigma_{\text{p.h}} + \Delta\sigma_g + \Delta\sigma_p . \qquad (3)$$

We assumed that the resistance of the lattice to the motion of free dislocations or the Peierls–Nabarro stress $\sigma_0 \sim 30$ MPa.

The evaluation of the contribution of the solid-solution hardening [12] has shown the for steel 38G2F from different heats (see Table 1) the value of $\Delta\sigma_{\text{s.s}}$ changes inconsiderably and amounts to 27 – 30 MPa.

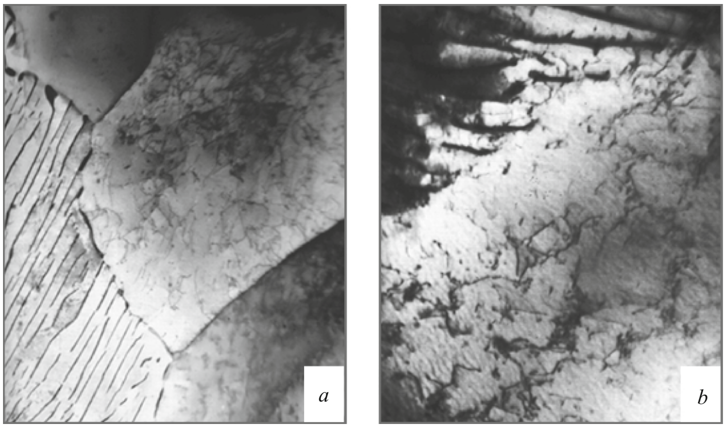


Fig. 4. Microstructure of steel 38G2F after rolling by variant *1* (TEM): *a*) ×12,500; *b*) ×21,000.

The dislocation density in a steel of type 38G2F after a standard TMT (Fig. 4*a*) does not exceed about $10^8 - 10^9$ cm^{-2}, which gives the contribution of the dislocation hardening $\Delta\sigma_d \sim 30$ MPa.

Grain-boundary hardening is commonly assessed with the help of the Hall–Petch equation

$$\Delta\sigma_g = k_o\, d_f^{-1/2}, \tag{4}$$

where $k_o = 0.63$ MPa · m$^{1/2}$ is a coefficient dependent on the angle of orientation of the ferrite grains and df is the size of the ferrite grains. For medium-carbon steels, where excess ferrite precipitates in the form of a net over the boundaries of the former austenite grains, df is taken to be equal to the average thickness of the ferrite net tf. By the data of the optical microscopy (Fig. 3*b*), the thickness of the ferrite net $t_f = 5 - 10$ μm; then the contribution of the grain-boundary hardening $\Delta\sigma_g = 200 - 280$ MPa and its average value $\Delta\sigma_g^{av} \sim 240$ MPa.

Hardening with particles of second phase is an important mechanism for raising the strength of structural steels. Structural steels contain a wide set of particles (of second phases) that differ in the hardness, shape, size, distribution, etc. These phases are classified into aggregate ones (where the second phase is surrounded by several matrix grains (as a rule, these are the structural components, i.e., pearlite, bainite and martensite up to several microns in size) and fine ones with a size of 10 – 500 Å, where every particle is fully surrounded with a similarly oriented matrix (carbides, nitrides, etc.).

The best coincidence with experimental results for steels is obtained by assessment of precipitation hardening by the Orowan mechanism, which allows for the interaction between dislocations and incoherent particles (carbides, nitrides, carbonitrides) [12]. The TEM studies of ferrite regions in steel 38G2F have shown that the carbide particles have an average size of about 100 nm and the distance between them $\lambda \sim 500$ nm (Fig. 4*b* on the cover). With allowance for the fact that the volume fraction of the particles does not exceed $f = 0.1$, the contribution of the precipitation hardening amounts to $\Delta\sigma_{p.h} \sim 80$ MPa in accordance with [12].

The leading role of the aggregate hardening in the yield strength of medium-carbon steels consists in the fact that the pearlitic component in the ferrite-pearlite mixture plays the role of a matrix receiving the elastic stresses and shifting the start of plastic flow toward higher stresses. To assess the contribution of the aggregate hardening with pearlite we may use a formula relating $\Delta\sigma_p$ to the fraction of pearlite q_p (%) in the ferrite-pearlite mixture through an empirical coefficient *K*, i.e.,

$$\Delta\sigma_p = 2.72 q_p. \tag{5}$$

By our data, for medium-carbon steels $K = 2.69 - 2.75$ (~2.72). In steel 38G2F the fraction of pearlite qp in the ferrite-perlite mixture does not exceed 80 – 85%. Then the contribution of pearlite into the yield strength will be $\Delta\sigma_p \sim 230$ MPa.

Generalizing the results of the calculation, we established that the main contribution (about 35 – 40%) into the yield strength of medium-carbon manganese microalloyed steels of type 39G2F in made by pearlite, the proportion of which in the ferrite-pearlite mixture is at least 85%, and by the grain-boundary hardening of ferrite precipitating in the form of a thin closed net about 5 – 10 μm thick over the boundaries of the former austenite grains (Fig. 5). The calculated value of $\Delta\sigma_{0.2}^{calc} = 640$ MPa is close to the middle of the range of the experimental values of the yield strength $\sigma_{0.2}$ for tubes with wall thickness 8.9 – 16.0 mm after rolling by variant *1*. These tubes exhibit this very type of ferrite-pearlite structure (Fig. 2).

The parameters of the ferrite-pearlite structure (the proportion and type of the excess ferrite, the size and the distribution of the pearlite colonies) at a constant rate of post-deformation air cooling $v_{cool} = 0.5 - 1.0$ K/sec are determined by the stability of the supercooled austenite (SSA) in the range of the diffusion transformation. The value of the SSA depends in the first turn on the structural and phase condition of the strained austenite, i.e., the grain and/or subgrain size,

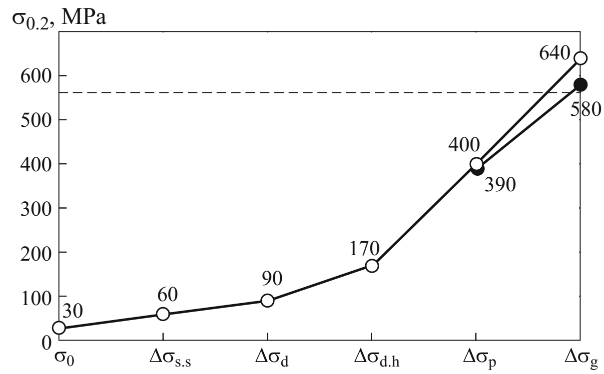


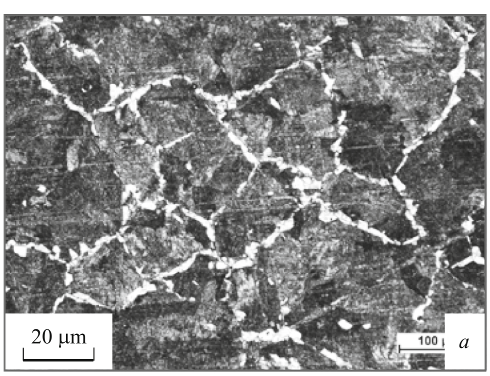
Fig. 5. Contribution of different hardening mechanisms into the yield strength of steel 38G2 with ferrite-pearlite structure (the dashed line presents the values of the yield strength standardized for strength group *E*): ○) rolling mode *1*; ●) rolling mode *2*.

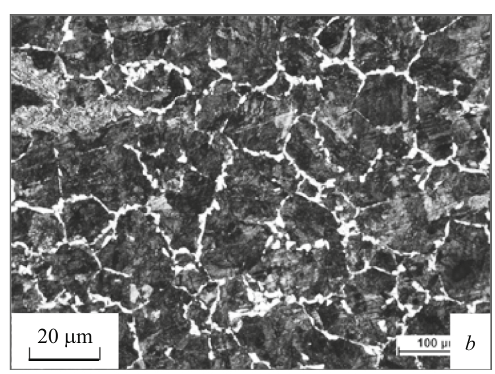
the density of the crystal structure defects and the presence of particles of second phases. With growth in the number of places of nucleation of a new phase in the austenite (grain and subgrain boundaries, incoherent twin boundaries and strain bands, etc.) the resistance of the supercooled austenite to decomposition in stage I decreases. This causes lowering of the strength parameters and elevation of the ductility parameters of the steel with ferrite-pearlite structure due to some growth of the fraction of excess ferrite and changes in its morphology as well as lowering of the dispersion of the pearlite colonies [13].

When the tubes from steel 38G2F are treated thermomechanically by variant *2*, intense plastic deformation occurs at a temperature above that of the start of precipitation of VC carbides (850 – 900°C). Therefore, the austenite grains grow intensely after piercing of the tube billet at 1150 – 1200°C (the draw ratio $K_d \sim 2.5$) and deformation of the tube in the expansion mill at 950 – 1050°C ($K_d \sim 1.5$). The large enough size of the austenite grains ($\bar{d}_a = 60 – 90$ μm) results in formation of a fine closed net of excess ferrite over the boundaries of the former austenite grains simultaneously with precipitation of bainitic products of decomposition of supercooled austenite (Fig. 6*a* and *b*).

Further decrease of the size of austenite grains to $\bar{d}_a = 40$ μm observed after sizing ($K_d < 0.1$) lowers the resistance of the supercooled austenite to decomposition in stage I and thus changes the morphology of the ferrite-pearlite mixture; the fraction of excess ferrite in the form of an open net and individual grains 10 – 15 μm in size increases to 20% and the bainitic component disappears (Fig. 6*c*). This lowers somewhat the yield strength of steel 38G2F, which should be $\Delta\sigma_{0.2}^{\text{calc}} = 580$ MPa according to the calculation (Fig. 5).

Thus, when the temperature of heating of tube billets from steel 39G2F is lowered from 1230 – 1260 to 1170 – 1180°C, TMT makes it possible to manufacture tubes ∅93.2 × 13.0 mm and ∅127 × 16.0 mm in size with a high set of mechanical properties ($\sigma_{0.2} = 585 – 590$ MPa, $\sigma_r = 870 – 895$ MPa, $\delta = 15.0 – 17.5$%) without bainite component in the structure. These properties exceed the values of the strength (by $\Delta\sigma_{0.2} = 30 – 40$ MPa, $\Delta\sigma_r = 180 – 200$ MPa)





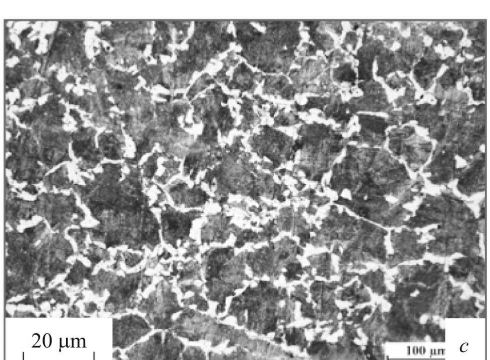


Fig. 6. Microstructure of a rough tube with size ∅127 × 16 mm in different stages of TMT (air cooling): *a*) after piercing (austenite grain size $d_a = 90$ μm); *b*) after deformation in an expanding mill ($d_a = 60$ μm); *c*) after sizing ($d_a = 40$ μm).

and ductility (by $\Delta\delta = 2.0 - 4.5\%$) characteristics standardized for strength group *E*.

CONCLUSIONS

Analysis of the mechanical properties of test batches of tubes from medium-carbon manganese microalloyed steels of type 38G2F obtained by conventional thermomechanical treatment (rolling at 1230 – 1260°C) has shown that the tubes of all the standard sizes studied (diameter 89 – 203 mm, wall thickness 8.9 – 25.0 mm) meet the requirements of the GOST 632(633)–80 Standard for strength group *E* ($\sigma_{0.2} = 552 - 758$ MPa, $\sigma_r \geq 698$ MPa, $\delta \geq 13.0\%$). However, for the tubes with wall thickness $h > 16$ mm the limiting property for reaching the standardized level of properties is the yield strength $\sigma_{0.2}$, because its value is close to the minimum required level $\sigma_{0.2}^{\min} = 552$ MPa.

Calculation of the yield strength with the use of the parameters of the ferrite-pearlite structure of steel 38G2F has shown that the main contribution into the value of $\sigma_{0.2}^{\text{calc}} = 640$ MPa is introduced by the hardening with pearlite, the fraction of which is 85 – 90%, and by the grain-boundary hardening with excess ferrite precipitating in the form of a thin closed net with a thickness of about 5 – 10 μm. The higher level of the strength properties ($\sigma_{0.2} \geq 640$ MPa, $\sigma_r \geq 900$ MPa) detected in the tubes with $h \leq 12.5$ mm is connected with formation of bainitic products of decomposition of supercooled austenite.

The study of the structure of tubes with a size of ∅127 × 16 and ∅93 × 13 mm from steel 38G2F after rolling of the tube billets to a lower temperature (1170 – 1180°C) than in the standard variant shows that at a mean size of the austenite grains $d_a \sim 40$ μm excess ferrite in an amount of about 80% precipitates in the form of a discontinuous net and of individual grains about 10 – 15 μm in size; the structure does not contain regions of bainite component. The mechanical properties obtained in this case ($\sigma_{0.2} = 585 - 590$ MPa, $\sigma_r = 870 - 895$ MPa, $\delta = 15.0 - 17.5\%$) are surely higher than the required ones for strength group *E*.

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