Light Non-Magnetic Steels Based on the Fe–25 Mn–5 Ni–Al–C System

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Abstract—The influence of aluminum (5–10%) and carbon (0.04–1.7%) contents on phase transformations, structure formation processes and mechanical properties of Fe–25 Mn–5 Ni–Al–C steels was studied theoretically and experimentally. The authors have estimated intervals of optimal crystallization regimes and subsequent deformation-thermal effects for obtaining austenitic steels with high specific strength. Hardness measurements on the sample section and mechanical tests in a wide interval of temperatures of cold, warm and hot deformation were performed, as well as the phase structure assessment of steels (alloys) on the basis of Fe–25 Mn–5 Ni–Al–C. In a cast state, an alloy with 5% of Al was non-magnetic, i.e., it had austenitic structure; alloys with 10–15% of Al were magnetic with two-phase structure ($\gamma + \alpha$). Aluminum considerably increases deformation resistance. At the same time, values $\sigma_1$ and $\sigma_{\text{max}}$ grow, i.e., also deformation hardening grows and softening processes are slowed down. With deformation rate growth, the Al influence becomes stronger. Austenitic high-manganese alloys with 5% of Al, both with low and high carbon content, have rather high plasticity and durability, and differ in high austenite stability. Alloying with nickel increases plasticity. Alloys with Al less than 10% are rather plastic also in a cast state. High-manganese (from 25% of Mn) alloys with Al content to 5–7% can be considered as high-strength cold-resistant and heat-resistant with thermally and mechanically stable austenite up to carbon content $\sim 1.5\%$.

Keywords: austenitic high-manganese light-weight steels, aluminum content, alloying with nickel, durability, plasticity, stability, impact resistance, abrasion strength, ferrite bond, silicate bond

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INTRODUCTION

High-strength alloys based on the Fe–Mn–Al–C system with a high content of manganese and aluminum are presently considered as promising materials for numerous application areas in nuclear power engineering, space technology, rocket production, transportation and consumption of liquefied gases due to their reduced specific weight and wide range of physical-chemical properties, adjustable by content and thermal treatment [1–3]. The use of steel in the mentioned areas promotes strength, plasticity, corrosive resistance, corrosive cracking resistance, stability of phase composition, brittle fracture resistance, compatibility with operating environment, and technological effectiveness during production and cost effectiveness. Alloys based on the Fe–Mn–Al–C system represent a new group of so-called TRIPLEX alloys with a high content of manganese and aluminum. Changing of chemical composition and thermal treatment of the given alloys make it possible to produce at least five phases in the equilibrium state: austenite, ferrite, $k$-carbide, carbide MeC, and $\beta$-Mn. Austenite is a predominating phase, forms first, and forms further matrix [4–12]. A new alloying complex of high-manganese alloys, containing aluminum, carbon and nitrogen, can obtain new properties and application areas. A high-strength state is achieved at the expense of all the strengthening mechanisms: solid solution, dispersive and structural, including those as a result of phase transformations under load. These alloys are widely studied for composition optimization: basic element content and additional alloying (Ni, Si, N, etc.), as well as technologies of production and thermal treatment [13–24].

Works [25–29] describe systematic theoretical and experimental researches of phase and structural transformations in multicomponent alloys of systems Fe–Mn–Al–C–N and Fe–Mn–Ni–Al–C–N at all stages of obtaining metal products from their liquid to solid states.

Alloys, containing aluminum, have low solubility of nitrogen in liquid and solid states. At 5% content of Al, the highest nitrogen content is solid solution Fe–20% Mn–Al at solidus temperature may constitute $9.3 \times 10^{-4}\%$. This quantity is even less than the possibility of modern industrial technologies. Thus, aluminum nitrides are released at temperatures of smelting and crystallization of the considered alloys
Alloying and microalloying of these alloys by nitrogen at content of ≥5% of Al is impossible [26].

The software package Thermo-Calc was used to plot polythermal section views of phase equilibrium diagrams of the Fe–Mn–Al–Ni–C system with variable content of one of the elements (Ni, Al, C) and constant content of the rest [27]. It was established that austenite (γ) of ferrite (α) steels are obtained depending on the content correlation of manganese, carbon and aluminum. The boundaries of domain γ of these alloys at ordinary heating temperatures at hot deformation and solid solution annealing were defined (Table 1).

It can be seen that the microstructure of alloys Fe–Mn–Ni–Al–C is mainly determined by contents of Mn, Al, and C. A large amount of aluminum (≥10%), which significantly increases the specific strength of austenite alloys of the Fe–Mn–Ni–Al–C system, achieves the following contents: Mn ≥ 20%, Ni ≥ 5%, C ≥ 1.4%. Low-carbon austenite alloys with 5% of Al can be obtained at contents Mn ≥ 25% and Ni ≥ 5% or at ≥10% Mn and ≥10% Ni.

In order to contain nickel in 5 and 10% amounts, alloys based on Fe–20% Mn–10% Al–(5–10)% Ni–C (C ≥ 1.4%) can be suggested as high-strength non-magnetic cryogenic alloys in both hardened and aged states.

Experimental results confirmed the adequacy of phase composition calculations, as well as solid solution estimates and dispersive alloy hardening of system Fe–(20–25)% Mn–(5–10)% Al–(0–10)% Ni–C. It is possible to achieve high strength and satisfactory plasticity in deformation conditions at room temperature on cast high-carbon alloys Fe–(19.1–23.9)% Mn–(4.0–9.0)% Al–(1.1–2.18)% C. The following strength factor levels occur: σ0.2 = 1000–1200 MPa, σu = 1300–1600 MPa at deformation degree ε = 10% [28].

The aim of this work is to research and compare computed and experimental data regarding: the influence of aluminum and carbon content on phase transformations, the obtained structure of steel strength of basic composition Fe–25 Mn–5 Ni–Al–C, as well as the estimation of rational temperature-temporary conditions of crystallization and thermal-mechanical treatment for obtaining austenite steels with high specific strength. Depending on manganese (25%), alloys refer to high-manganese and additionally nickel (5%) content, sparingly alloyed at least for cryogenic assignment.

### MATERIALS AND RESEARCH METHODOLOGY

Three alloys similarly composed of Fe–25 Mn–5 Ni–C with different content of aluminum (5, 10, 15%) and carbon (from 0.04 to 1.7%) corresponding to low- and high-carbon steels were melt by special technology (Table 2) in order to compare theoretical (computed) and experimental data about phase transformations, structure formation processes and obtained alloy properties of the system under consideration.

In the phase diagrams of Figs. 1–3, polythermal section views of basic composition Fe–25 Mn–5 Ni–Al–C, as well as the estimation of rational temperature-temporary conditions of crystallization and thermal-mechanical treatment for obtaining austenite steels with high specific strength. Depending on manganese (25%), alloys refer to high-manganese and additionally nickel (5%) content, sparingly alloyed at least for cryogenic assignment.

### Table 1. Elements content limiting the single-phase area γ at 1000–1200°C

| Alloy         | Content, % (by mass) | Mn | Ni  | Al  | C  |
|--------------|----------------------|----|-----|-----|----|
| Fe–Mn–Al     | 10                   | 5  | 5   | ≤2.8| –  |
|              | 20                   | 5  | 5   | ≤3.8| –  |
|              | 30                   | 5  | 5   | ≤4.2| –  |
| Fe–Mn–Ni–Al | 10 >10               | 5  | 5   | ≥0.18| –  |
|              | 25 ≥4.3              | 5  | 5   | ≥0.7 | –  |
|              | 10 5                 | 5  | 5   | 6.0 | –  |
|              | 25 5                 | 5  | 5   | ≥0.04| –  |
|              | 25 10                | 5  | 5   | ≥1.37| –  |

### Table 2. Chemical composition of experimental Fe–25 Mn–5 Ni–Al–C alloys, wt %

| Designation of alloy | C     | Mn  | Ni  | Al  | Mo  | Cr  | Si  | Cu  | P    | S    |
|----------------------|-------|-----|-----|-----|-----|-----|-----|-----|------|------|
| 57-1                 | 0.095 | 23.5| 5.21| 4.08| 0.22| 0.04| 0.12| 0.007| 0.0078| <0.0050|
| 57-2                 | 1.706 | 23.9| 5.46| 8.9 | 0.21| 0.03| 0.13| 0.005| 0.0084| <0.0050|
| 57-3                 | 1.383 | 25.0| 5.48| 12.6| 0.30| 0.01| 0.14| 0.006| 0.0063| <0.0050|

* Remaining iron.
Solid solution \( \text{L} + \gamma \) might run into a two-phase domain by mistake. Below hot deformation temperature without the risk of running into a two-phase domain by mistake. Below 500°C steel structure is \( \alpha + \gamma + \text{Mn}_2\text{C} \).

Steel of content close to alloy 57-2 at carbon content of 1.7% is crystallized by scheme \( \text{L} \rightarrow \text{L} + \alpha \rightarrow \text{L} + \alpha + \gamma \rightarrow \text{L} + \gamma \rightarrow \gamma \), i.e., without formation of \( \delta \)-ferrite. From \( T_{\text{solidus}} \) to 900°C steel has a structure of single-phase solution \( \gamma \). Below with temperature decrement the structure changes as follows: \( \gamma + \text{Mn}_2\text{C} \) from 900 to 840°C; \( \alpha + \gamma + \kappa + \text{Mn}_3\text{C}_2 \) from 840 to 780°C; \( \alpha + \kappa + \text{Mn}_3\text{C}_2 \) from 780 to 590°C; \( \alpha + \kappa + \text{Mn}_3\text{C}_3 \) from 590 to 500°C; below 500 and to 350°C \( \alpha + \gamma + \text{Fe}_3\text{Al} + \text{Mn}_7\text{C}_3 \).

The increase of aluminum and carbon content at constant contents of manganese and nickel led to the decrease of a single-phase \( \gamma \) domain at the expense of a lower temperature increase of this domain. It also led to the release of carbide phases, in particular \( \kappa \) carbide, already at temperatures below 900°C. However, at the same time, possible temperature intervals of thermal-deformational treatment remains quite wide (950–1200°C).

Steel 57-3 is crystallized by scheme \( \text{L} \rightarrow \text{L} + \alpha \rightarrow \text{L} + \alpha + \gamma \rightarrow \alpha + \gamma \), i.e., with formation of two-phase solid solution \( \alpha + \gamma \), which exists at temperatures from \( T_{\text{solidus}} \) to 900°C. The increase of aluminum and carbon content at variable carbon content from 0 to 1% and constant content of Mn = 25%, Al = 5%, and Ni = 5%.

Fig. 1. Polythermal cross section of Fe–Mn–Al–Ni–C system state diagram at variable carbon content from 0 to 1% and constant content of Mn = 25%, Al = 5%, and Ni = 5%.

Fig. 2. Polythermal cross section of Fe–Mn–Al–Ni–C system state diagram at variable carbon content from 1 to 2% and at constant content of Mn = 25%, Al = 10%, and Ni = 5%.

Domain \( \alpha + \gamma + \text{Mn}_3\text{C}_3 \), below 880 to 350°C, domain \( \alpha + \kappa + \text{Mn}_3\text{C}_2 \) and below 350°C, domain \( \alpha + \kappa + \text{Mn}_3\text{C}_3 \). Steel 57-3 also has quite a wide interval of temperatures (950–1150°C) for choosing regimes of thermal-deformational treatment, but in two-phase \( \gamma + \alpha \) domain.

Thus, the aluminum content increase leads to a significant change of alloy phase composition already after crystallization. Differences in the structure formation processes during further treatment and in the final structure and properties of these steels may only increase.

Special melt technology provided not only closeness of basic alloy content, but also different rate of crystallization and cooling due to different sizes (diameter of 6 and 14–16 mm) of samples and their cooling in water.

The technology of melt, selection and cooling of samples included the following operations. As a mixture, the following source materials made pure by impurities were used: technically pure iron of type 008ZhR, metallic manganese of type Mn998, electrolytic nickel N1, and electrolytic aluminum. In order to create neutral furnace atmosphere, high-purity argon by TC 6-21-12-94 with volume fraction of argon not less than 99.998% as operating gas was used.
Steels 57-1, 57-2, 57-3 were melt in a vacuum induction furnace. Stuffed crucible, material: 79% MgO; 10% Al2O3; 1% CaF2. The mixture was placed in the furnace’s crucible: iron, nickel and carbon. The mixture was heated and melt in the vacuum. After complete melt of the mixture and stopping of gas release, argon was introduced into the furnace until 1 atm. Power providing constant intensive mixing of the melt and heating to 1500–1550°C was supported.

The operation sequence to obtain samples of various composition was as follows. Aluminum and then manganese were introduced into the melt in order to obtain 0.1–0.3% of C, 25% of Mn, 5% of Ni and 5% of Al after dissolution and mixing. After that, liquid metal samples were taken from the melt by suction into quartz tubes with a 6-mm inner diameter. Part of the samples was cooled in water, another part, in air. Aluminum, carbon and manganese were added into the remaining solution in order to obtain 1.4–1.7% of C, 25% of Mn, 5% of Ni and 10% of Al after dissolution and mixing. After mixing of the melt metal, samples were taken into quartz tubes again. Then aluminum and manganese were added to the melt in order to obtain a melt with 1.2–1.4% of C, 25% of Mn, 5% of Ni and 15% of Al. After dissolution of admixtures and mixing, metal samples were again taken into quartz tubes and cooled partially in water and air. The remaining metal composed of steel 53-7 was poured from furnace to casting mold. Ingot was cooled from 800 °C in air. Actual composition of the obtained samples is presented in Table 2.

Fast-cooled samples of alloys 57-1 and 57-2 with a 6-mm diameter, as well as alloy samples 44K, 52K, 54K take for comparison and described in work [27], were studied for resistance and changing of phase composition at hot, warm and cold deformations.

Hot and warm deformations were performed by the Gleeble 3800-GTC system, which allowed for the physical simulation of thermal-mechanical processes and treatment. The research was conducted using cylindrical shape samples with a 5-mm diameter and 10-mm height.

Tests were performed in a module operating chamber for high-temperature compression testing (including shock tests) Hydrowedge-II, equipped with auxiliary devices for data capturing and controlling temperature, effort, magnitude and speed of deforming bar motion and sample size measuring during testing.

Researches were performed in a vacuum (about 10−4 Torr). Temperature was controlled using K-type thermocouples, which were directly welded to the central part and ends of a sample. Heating was achieved by a controllable electric-current passage through the sample. At the end of deformation, samples were sharply cooled at the expense of their surface radiation, heat removal to deforming strikers and further due to forced “blowing” by compressed air.

Hot deformation was performed by compression at a temperature of 1000°C at speeds Ʌ = 1 s⁻¹ and 0.1 s⁻¹ to ε = 0.5. Deformation temperature corresponded to γ domain of all samples, and deformation rate 1 s⁻¹ was close to deformation at ordinary rolling.

Warm deformation was also performed by compression at speed Ʌ = 0.1 s⁻¹ to ε = 0.3 at temperatures of 550, 600, 650, 700, and 800°C, which correspond to various phase domains (Figs. 1–3 and Table 3).

Cold deformation of all alloys, except 54K, was performed at room temperature using samples with a 5-mm diameter and 60-mm length on the Instron 3300 test system by controlling for three-point bending according to State Standard 14019–2003. The test was concluded in bending of a sample around a bend previously under static effort.

Alloy 54K samples were cut off from hot-rolled lists with a 4-mm thickness and subjected to a stretching test according to State Standard 1497. Experiments were performed on proportional flat samples on the INSTRON-8801 pull-test machine at a loading rate of 2–10 mm/min.

Phase composition was determined before and after all the tests by magnetometric and X-ray struc-
Table 3. Calculated temperature intervals of phase areas of Fe–25 Mn–5 Ni–Al–C alloys

| Designation of alloy | Temperature, °C | Magnetic | Magnetization at 20°C |
|---------------------|-----------------|----------|----------------------|
|                     | γ               | γ + α    | α                    | Mm   | Md |
| 57-1                | 1200–550 (C > 0)| <550     | –                    | <550 | –  |
| 57-2                | 1200–900 (C ≥ 1.3) | 1200–900 (C = 0.1–1.3) | – | <800 | –  | + |
| 57-3                | –               | 1200–900 (C = 0.05–1.45) | 1200–900 (C ≤ 0.05) | <900 | –  | + |

RESULTS AND DISCUSSION

All the samples of alloys 57-1, 57-2, 57-3 had dendrite structure with typical zone distribution for fast cooled ingots in the initial state. Cast samples of both diameters had defects and pores more thickly in the center. At the same time, samples of alloys 57-1 and 57-2 have a longer directed crystallization zone as compared with 57-3. Alloy 53-7 samples were different for sharper dendrites (likely at the expense of higher liquation), coarse crystalline structure, large pore presence and cracks scattered along the whole cross-section.

In the selected samples free of defects, their hardnesses were measured by its cross-section, they were mechanically tested in a wide range of temperatures of cold, warm and hot deformations, and finally their phase compositions were estimated.

In the cast state, alloy 57-1 was non-magnetic, i.e., had austenite structure, while alloys 57-2 and 57-3 were magnetic. Results of X-ray structural analysis confirmed austenite structure of alloy 57-1 and two-phase structure (γ + α) of alloys 57-2 and 57-3.

Table 3 presents temperature ranges of phase domains.

Crystallization rate and cooling samples of nickel-free alloys with close composition and size constituted ≤1 × 10⁻² K/s by interdendritic parameter [27].

Magnetization of alloy 57-2 even after the highest cooling rate (6-mm-diameter sample, cooling in water) implies possible liquations and high speed of austenite decomposition.

Alloy 57-3 with 15% of Al already has two-phase (γ + α) matrix by its average composition. Due to high-temperature ageing of austenite, the greatest hardness is achieved even at fast cooling. At ordinary and average cooling rates, additional strengthening is possible due to γ → α transformation and decomposition (ageing) of already α solid solution. However, in all treatment cases, alloy 57-3 remains fragile. Thus, aluminum alloying in an amount greater than 10% is slightly promising.

The results of measuring the hardness distribution over the cross-section of cast alloy samples are presented in Table 4. Alloy 57-1 samples have hardness HV₀.₅ = 133–156 and average value HV₀.₅ = 152. Hardness weakly depends on cooling rate (water, air, different diameters of cast samples). In case of maximal diameter (16 mm) and cooling in water, hardness is minimal. Overall, smaller scattering of values is observed and minimal values are typical for the sample’s central zone. This can be connected with liquation. The contribution of ageing processes is not excluded. However, it is insignificant due to the small carbon content in the alloy.

Alloys 57-2 and 57-3 have significantly higher hardness in cast state at the expense of greater content of aluminum and carbon as well as two-phase (γ + α) structure. Hardness level for alloy 57-2 constitutes 329–495 HV, for alloy 57-3, 517–566 HV. Greater hardness inhomogeneity along the cross-section is observed as compared to alloy 57-1, in which the influence of cooling rate (water, air, different diameters of cross-section) is more significant, as well as being connected with superposition of phase transformation processes (γ → α and ageing).

Diagrams of hot, warm and cold deformation of the studied alloys look ordinary. Examples of the diagrams are presented in Fig. 4.

All test results on nickel-containing alloys, as well as the comparison of nickel-free analogue (alloy 44К [27]), are summarized in Table 5. Hot deformation diagrams allowing 5% of Al, deformed in the austenite state, look ordinary and may have a maximum (ε = 0.1 s⁻¹) or may not even have a distinct one. Diagrams of alloys 57-2 with ~9% of Al, for which deformation might flow in an inhomogeneous multiphase domain (γ + α + κ) due to liquation and insufficient homogenization, showed jaggedness (flow instability) and the presence of α phase after hardening.

Analysis of hot deformation parameters clarifies the following. Nickel alloying with other conditions being equal accelerates the softening processes, which
leads to slight decrement of resistance against hot deformation $\sigma_{\text{max}}$ (see Table 5 for alloys 57-1 and 44K).

An increase in carbon content (comparison of alloys 57-1 with 52K and 54K) barely lead to change of $\sigma_{\text{max}}$. Aluminum (see alloys 57-1 and 57-2) significantly increases the resistance against deformation. At the same time, values of $\sigma_1$ and $\sigma_{\text{max}}$ grow, i.e., deformation strengthening increases and softening processes are slowed down. The aluminum influence appears stronger with deformation rate growth (see data for $\dot{\varepsilon} = 1 \text{ s}^{-1}$ and $\dot{\varepsilon} = 0.1 \text{ s}^{-1}$). This may be connected with the slowing down of diffusion processes and related recrystallization process changes.

All the samples with Al $\leq$ 10% content were deformed without the temperature destruction of warm deformation (550–800°C) to $\varepsilon = 0.3$ at $\dot{\varepsilon} = 0.1 \text{ s}^{-1}$, i.e., such alloys are ductile enough even in cast state.

Table 4. Distribution of hardness $HV_{0.5}$ along cross section of cast samples of the alloys 57-1, 57-2, 57-3

| Designation of alloy | Diameter of sample, mm | Cooling medium | $HV_{0.5}$ near surface | $1/2$ of radius | center |
|---------------------|------------------------|----------------|-------------------------|----------------|--------|
| 57-1                | 6                      | Air            | 153 ± 7                 | 154 ± 6        | 146 ± 5 |
|                     |                        | Water          | 157 ± 4                 | 148 ± 2        | 152 ± 4 |
| 57-1                | 16                     | Air            | 154 ± 5                 | 152 ± 2        | 156 ± 2 |
|                     |                        | Water          | 156 ± 3                 | 153 ± 6        | 133 ± 21|
| 57-2                | 6                      | Air            | 487 ± 5                 | 493 ± 7        | 485 ± 6 |
|                     |                        | Water          | 488 ± 9                 | 490 ± 7        | 329 ± 103|
| 57-2                | 16                     | Air            | 495 ± 10                | 485 ± 16       | 435 ± 9 |
|                     |                        | Water          | 492 ± 8                 | 487 ± 12       | 466 ± 41|
| 57-3                | 6                      | Air            | 524 ± 8                 | 537 ± 17       | 566 ± 19 |
|                     |                        | Water          | 544 ± 9                 | 530 ± 10       | 536 ± 20 |
| 57-3                | 16                     | Air            | 518 ± 16                | 539 ± 12       | 524 ± 40 |
|                     |                        | Water          | 543 ± 20                | 518 ± 6        | 517 ± 16 |

Alloy 57-3 is fragile at all the studied temperatures. Apparently, alloying by 15% of Al is technologically unfeasible for high-manganese compositions. Thus, a more detailed conclusion requires additional research.

Alloy diagrams deformed strictly in austenite domain (alloy 57-1, 600°C) did not have jaggedness, resistance against deformation $\sigma_1$ and $\sigma_{30}$, nor deformational strength decreased with temperature growth (see Table 5). The resistance level of hot deformation is naturally higher. After sharp cooling at the end of deformation, alloys remained nonmagnetic and austenitic.

Hardness increased more intensively for alloy 57-1 after warm deformation at 600°C, $\dot{\varepsilon} = 0.1 \text{ s}^{-1}$, $\varepsilon = 0.3$ and hardening $HV_{0.5} = 261$, but after hot deformation at 1000°C, $\dot{\varepsilon} = 0.1 \text{ s}^{-1}$, $\varepsilon = 0.5$ and hardening $HV_{0.5} = 171$.

![Fig. 4](source-image-url) Compress deformation diagrams: (a) hot deformation, $T = 1000°C$: (1) steel 57-2, $\dot{\varepsilon} = 1 \text{ s}^{-1}$; (2) steel 57-1, $\dot{\varepsilon} = 0.1 \text{ s}^{-1}$; (b) warm deformation, $\dot{\varepsilon} = 0.1 \text{ s}^{-1}$: (3) steel 57-2, $T = 650°C$; (4) steel 57-1, $T = 600°C$. 

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Table 5. Summary table of deformation resistance parameters at different test temperatures

| Designation of alloy, content by main alloying elements | Hot deformation, compression, 1000°C, ε = 0.5 | Warm deformation, compression, ε = 0.3 | Cold deformation, three-point bending, 20°C |
|--------------------------------------------------------|---------------------------------------------|---------------------------------------|------------------------------------------|
|                                                        | speed 1 s⁻¹ | speed 0.1 s⁻¹ | speed 0.1 s⁻¹ | speed ~0.01 s⁻¹ | PC: at deformation temperature/after hardening | PC: after deformation/before deformation | PC: at deformation temperature/after hardening | PC: after deformation/before deformation |
|                                                        | σ₁, MPa | σₘₐₓ, MPa | εₘₐₓ | σ₁, MPa | σₘₐₓ, MPa | εₘₐₓ | σ₁, MPa | σₐ₀₂, MPa | σₐᵤ, MPa | φ, deg |
| 57-1 Fe–23.5 Mn–5.2 Ni – 4.1 Al–0.09 C                 | 146 | 209 | – | 129 | 168 | 0.18 | 187 | 577 (600°C) | 357 | 663 | 131 |
| 44K** Fe–24.1 Mn–5.3 Al–0.04 C                         | – | – | – | 180 | 0.18 | – | 870 (550°C) | 350 | 1095 | 128 |
| 52K** Fe–24.5 Mn–4.8 Ni–4.4 Al–0.15 C                 | 162 | 214 | – | 123 | 170 | 0.18 | 169 | 489 (700°C) | – | – | – |
| 54K** Fe–24.1 Mn–4.7 Ni–4.9 Al–1.47 C                 | 171 | 232 | – | 132 | 169 | 0.15 | 281 | 635 (700°C) | – | – | – |

** CD, stretching

| Designation of alloy, content by main alloying elements | Hot deformation, compression, 1000°C, ε = 0.5 | Warm deformation, compression, ε = 0.3 | Cold deformation, three-point bending, 20°C |
|--------------------------------------------------------|---------------------------------------------|---------------------------------------|------------------------------------------|
|                                                        | speed 1 s⁻¹ | speed 0.1 s⁻¹ | speed 0.1 s⁻¹ | speed ~0.01 s⁻¹ | PC: at deformation temperature/after hardening | PC: after deformation/before deformation | PC: at deformation temperature/after hardening | PC: after deformation/before deformation |
|                                                        | σ₁, MPa | σₘₐₓ, MPa | εₘₐₓ | σ₁, MPa | σₘₐₓ, MPa | εₘₐₓ | σ₁, MPa | σₐ₀₂, MPa | σₐᵤ, MPa | φ, deg |
| 57-2 Fe–23.9 Mn–5.5 Ni–8.9 Al–1.71 C                  | 299 | 324 | 192 | 197 | (yield drop and yield plateau) | 726 | 880 (650°C) | – | – | – |

* PC—phase composition. At deformation temperature—computed data, after hardening—experimental data.

** Alloys taken for comparison [26, 28, 29].
When ageing (alloys 44K, 550°C; 57-2, 650°C) and additional polymorphic $\gamma \rightarrow \alpha$ transformation (alloy 57-2, 650°C), hot deformation at temperatures could develop significantly and was further accompanied by jaggedness, yield plateau on deformation diagrams, as well as a stronger deformational hardening. The higher was the content of carbon and aluminum in the alloys, the stronger were the effects of hardening. After deformation and hardening, alloy 57-2 was magnetic. Arguably, at deformation temperatures, they also had structure $\alpha + \gamma + k$.

Testing of cast alloys for three-point bending at room temperature showed that high-carbon alloys with ~10% of Al (57-2) are characterized by brittle fracture, i.e., they are not appropriate for low-temperature exploitation in conditions of even small-stretch loadings.

Austenite high-manganese alloys with 5% of Al with low (57-1, 44K, 52K) or high (54K) carbon content possess quite high plasticity and strength (see Table 5) and are characterized by high austenite stability.

Nickel alloying increases plasticity and, possibly, decreases deformational softening. In this way, at equal values of $\sigma_{0,2}$ and $\varphi$, $\sigma_{0}$ of alloy 44K is greater than $\sigma_{0}$ of alloy 57-1. At the same time, alloy 54K could be deformed by stretching by 39%.

CONCLUSIONS

It was shown experimentally that high-manganese alloys (with 25% of Mn) containing 5–7% of Al can be considered as high-strength cold- and heatproof materials with thermally and mechanically stable austenite up to carbon content ~1.5%. Steels can be produced as cast or deformable. Additional alloying by 5% of Ni increases plasticity at cold deformation and amplifies the ageing effects at warm deformation temperatures.

For aluminum content greater than 8–10%, an additional correction of the composition and technological regimes of treatment are required.

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