

## DAMASK STEEL – UNALLOYED CARBIDE CLASS STEEL

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*Once, the question was posed: “What is damask steel – special steel or technology?” and I could not answer. Everything is simple and simultaneously complicated. People have used damask steel for a thousand years. Legends are made up about it. Today secrets of the microworld are being opened up, but damask steel microstructure, determining its properties, has remained in the shade of its reputation. However, only composition and structure govern steel properties. Any steel is prepared by specific technology. There is technology for preparing die steels (X12), and there is technology for preparing high-speed steels (R18), and consequently there should also be technology for preparing damask steel. As a result of searching for technology for preparing damask steel, it has become known that for the very best damask steels the amount of carbon within their composition approaches that of cast iron. With high-temperature treatment of these alloys, there is a risk of forming free graphite, as a result of which there is inevitable loss of toughness. A condition for forming graphite is the presence of crystallization centers (silicon compounds). Composition purity is a starting factor for revealing the main structural components governing damask steel properties. So, in order to understand today why blade sharpness is combined with high elasticity it is necessary to reveal the true structure of damask steel and to establish the underlying features of its individual properties. Experimental study of the structure makes it possible to determine what all of them are in the field of unalloyed carbide class steels.*

The task set in this work is to study the structure and properties of unalloyed carbide class steels.

There is hardly any published work in the field of unalloyed carbide class steels. Today data are only encountered for alloyed carbon steels of type X12, used widely in the tool industry. Within the structure of die steel X12, a eutectic (ledeburite) is observed after crystallization, located within liquation areas (Fig. 1a). In a deformed state, the structure is a metal matrix based on austenite, or its decomposition products, with coarse eutectic carbide inclusions (Fig. 1b) comprising recrystallized ledeburite.

Setting out to study unalloyed material with a carbon content of 1.5 to 3.0%, one is obliged to take as a basis the classical distribution of phases and structures according to the iron–carbon composition diagram. With this carbon content, metal may be either eutectoid steel (1.5–2.0% C) or hypoeutectoid cast iron (2.0–3.0% C). The composition and structure of these alloys are controlled by GOST 22536. The difference only consists in the fact that within the test alloy composition there are no modifiers or decomposition products. Such ingots were melted in a vacuum furnace in the scientific-production area of Bardin TsNIIchermet, and had the following chemical composition:

- alloy No. 1 is 1.62% C; 0.069% Si; 0.026% Mn; 0.002% P; 0.004% S ;
- alloy No. 2 is 2.25% C; 0.065% Si; 0.024% Mn; 0.002% P; 0.004% S ;
- alloy No. 3 is 2.70% C; 0.10% Si; 0.10% Mn; 0.003% P; 0.003% S.

The content of the rest of elements in all three alloys is hundredths or thousandths.

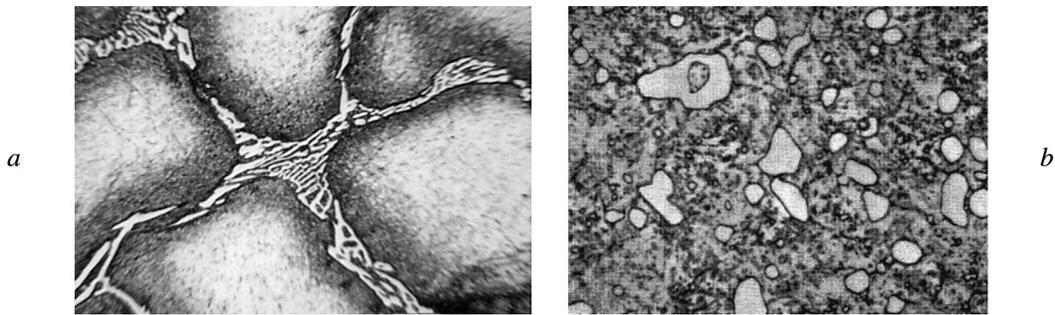


Fig. 1. Steel X12 structure: *a*) cast condition. Primary dendrites of austenite and eutectic  $\text{Fe}-(\text{Cr}, \text{Fe})_7\text{C}_3$  in the form of a network ( $\times 1000$ ); *b*) forged condition. Martensite and eutectic carbides (light inclusions). Hardness 950 HV ( $\times 800$ ).

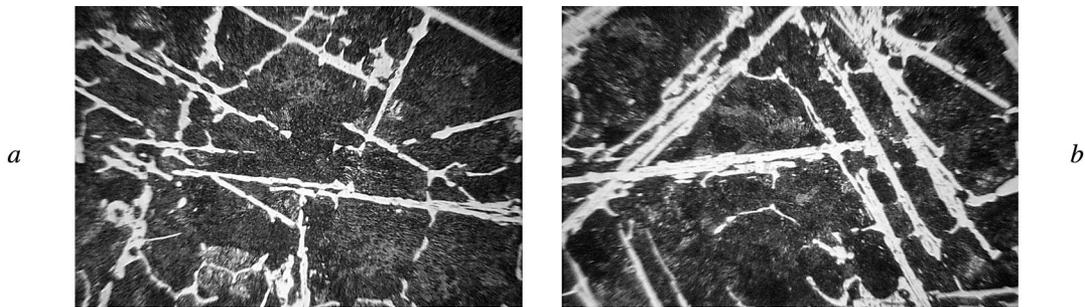


Fig. 2. Alloy structure: *a*) No. 1, cast condition + annealing at  $700^\circ\text{C}$ . Pearlite (dark background) and needles of Widmanstätten cementite 10% ( $\times 800$ ); *b*) No. 2, cast condition + annealing at  $700^\circ\text{C}$ . Pearlite (dark background) and needles of Widmanstätten cementite 20% ( $\times 800$ ).

The aim of this work was to define features of eutectic carbide formation in iron alloys containing from 1.5 to 3.0% carbon (content of the rest of elements hundredths or thousandths). The following problems were resolved in the work:

- 1) a study of features of cementite and ledeburite shape change during heat treatment;
- 2) revelation of regularities of cementite and ledeburite shape change after forging in the temperature range  $820\text{--}650^\circ\text{C}$ ; and
- 3) establishment of thermomechanical treatment regimes providing unalloyed steels of the carbide class.

In pure iron-carbon alloy, the amount, shape, and nature of dendrite distribution is only determined by the degree of melt supercooling. Thus, with slow cooling (crucible cooling in a furnace) a coarser dendritic structure is obtained. During alloy melting in a vacuum, the degree of melt supercooling is higher (a crucible is cooled at room temperature). Therefore, the structure of alloys Nos. 1 and 2 is a finely dispersed matrix of lamellar pearlite with needles of excess Widmanstätten cementite located within it, consisting of individual layers (Fig. 2). Preferred precipitation of cementite of the Widmanstätten type is explained by the fact that with an increase in degree of melt cooling there is a reduction in the critical size of nuclei, and an increase in the number of crystallization centers. In alloy No. 3, ledeburite is added to these structural components in the form of an interdendritic network (Fig. 3). Thus, the structure of alloys Nos. 1 and 2 is typical for ultra high carbon steels, but the structure of alloy No. 3 is typical for hypoeutectic white cast iron cooled at room temperature.

Cementite of the Widmanstätten type is almost entirely dissolved during homogenizing annealing at  $1150^\circ\text{C}$ . In specimens of alloy No. 1, a coarse cementite network forms over grain boundaries of previous austenite grains (Fig. 4*a*). In individual areas, there are coarse conglomerates of cementite precipitates. Metallographic indications do not make it possible to



Fig. 3. Structure of alloy No. 3, cast condition. Pearlite (dark background), needles of Widmanstätten cementite and ledeburite network ( $\times 800$ ).

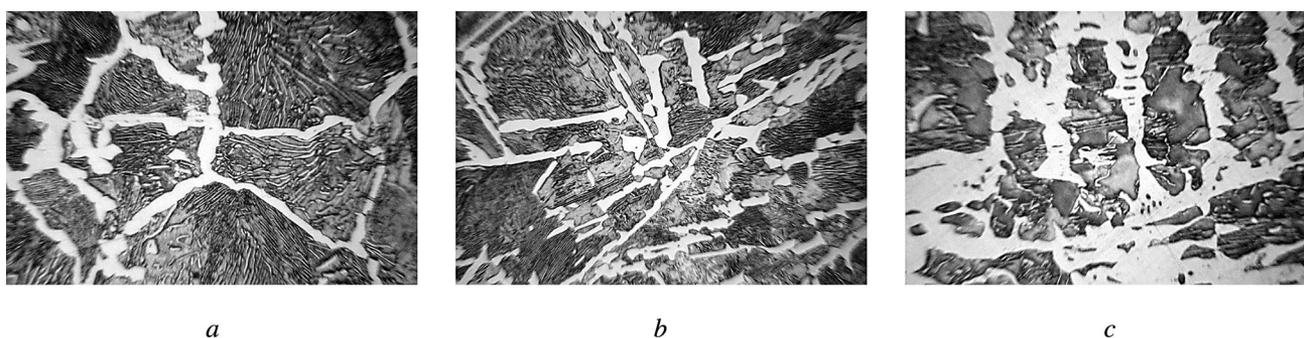


Fig. 4. Alloy structure after annealing at  $1150^{\circ}\text{C}$ : *a*) No. 1. Pearlite (dark background) and cementite network, located over previous austenite grain boundaries ( $\times 1000$ ); *b*) No. 2. Pearlite (dark background) and strips not having a layered structure ( $\times 1000$ ); *c*) No. 3. Pearlite (dark background) and ledeburite network without Widmanstätten cementite ( $\times 1000$ ).

identify eutectic carbides within them, obtained during recrystallization of metastable ledeburite. In specimens of alloy No. 2, Widmanstätten cementite recrystallizes into denser “rods” that precipitate over the same crystallographic directions (Fig. 4*b*). A typical feature of this cementite is the fact that it has a layered structure. In the field of view of a microsection, typical indications are also observed of ledeburite formations. In specimens of alloy No. 3, a ledeburite network thickens due to the excess cementite, and boundaries of ledeburite colony margins become more blurred (Fig. 4*c*). Thus, within the structure of these alloys annealed at  $1150^{\circ}\text{C}$  a specific proportion of eutectic is observed, growing with an increase in carbon content.

Further research has been carried out with the aim of revealing features of cementite shape change during forging in the temperature range  $820\text{--}650^{\circ}\text{C}$ . In this temperature range, excess cementite does not dissolve in austenite, and experiences compressive shear stresses. Under these temperature and rate conditions, deformation of a cementite network in alloy No. 1 is broken down with the formation of globular shaped carbides, retaining orientation along the deformation axis (Fig. 5*a*). In the field of view of a microsection due to the etching contrast of pearlite and carbide components banding is clearly seen, which is greater, the coarser the original austenite grains and coarser cementite network.

During deformation of specimens of alloy No. 2 with excess phase in the form of Widmanstätten cementite, it is necessary to expend more energy in order to obtain carbides of a eutectic type. This is due to the fact that excess cementite obtained by recrystallization of austenite has a layered structure. Cementite splits up over areas of layer contact, recrystallizing again in particles of eutectic type. Cementite particles retain the orientation in the direct vicinity of original layered platelets, forming fine patterns of different configuration. During deformation of specimens of alloy No. 2 obtained by annealing at  $1150^{\circ}\text{C}$  for 2 h, all of the energy accumulated at an austenite–ledeburite or austenite–cementite interface is

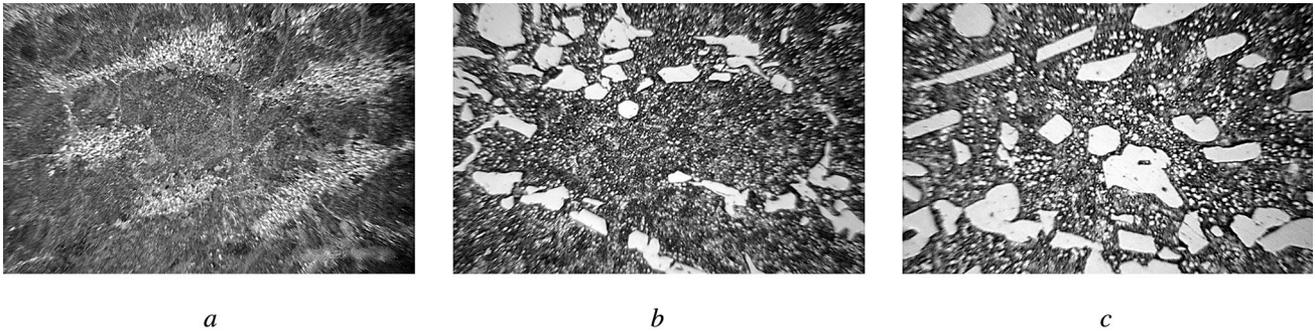


Fig. 5. Structure of alloys after forging at 850–650°C: *a*) No. 1, sorbite ( $d_c$  not more than 0.2  $\mu\text{m}$ ) and excess globular shape carbides orientated along the deformation axis ( $\times 300$ ); *b*) No. 2, matrix of globular pearlite and eutectic cementite 20% ( $\times 1000$ ); *c*) No. 3, matrix of globular pearlite and eutectic cementite 40% ( $\times 1000$ ).

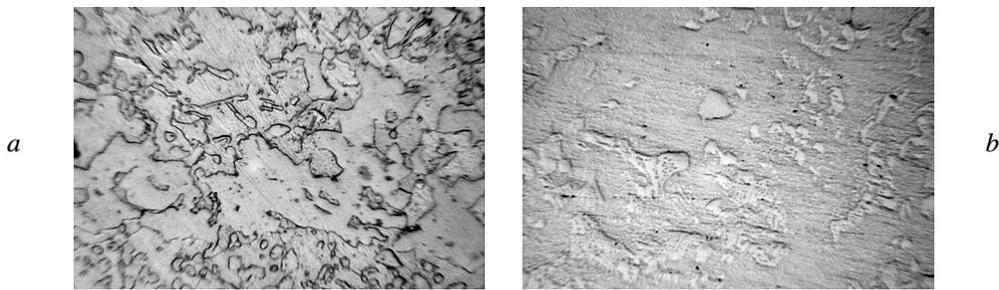


Fig. 6. Alloy No. 3 structure: *a*) forging and annealing at 1000°C. Ferrite, pearlite decomposition products (globular cementite) and coarse eutectic cementite ( $\times 500$ ); *b*) unetched microsection. Markings appear in a microsection surface during polishing formed by eutectic carbides in a pearlite matrix ( $\times 200$ ).

expended in eutectic compaction, or in crushing of cementite fibers into shorter fibers. This is due to the fact that this cementite does not have a layered structure and breaks down due to thinning in the weakest areas (Fig. 5*b*).

During deformation of alloy No. 3, ledeburite colonies at first deform austenite, and around ledeburite there is local plastic deformation. Ledeburite is under the action of normal compressive stresses (austenite grains) and shear stresses, caused by adhesive forces at the ledeburite–austenite interface. The energy accumulated is expended in compaction of cellular austenite–cementite eutectic (ledeburite), which finally leads to its recrystallization and formation of eutectic composition carbides (Fig. 5*c*).

By summarizing this it is possible to note the following metallographic indications of unalloyed carbide steels.

*First*, these steels exhibit forgability, which in principle distinguishes cast iron from steel. Prolonged soaking is required in order that a billet exhibits forgability, i.e., not less than 5 h at temperatures between the liquidus and solidus without bringing the melt to total solidification. As a result of a small degree of melt supercooling, there is a marked reduction in the number of crystallization centers, and interdendritic areas are enriched with carbon up to eutectic content. This production process makes it possible to obtain metal billets for forging over a wide range of carbon content, i.e., from 1.5 to 3.0%. *Second*, the structure of ingots after forging is a globular pearlite matrix with eutectic cementite distributed within it, obtained by dynamic recrystallization of ledeburite. The average size of eutectic cementite particles reaches 20  $\mu\text{m}$ . *Third*, eutectic cementite, obtained by dynamic recrystallization of ledeburite, has an insignificant proportion of dislocations and is a quasicrystal. *Fourth*, eutectic cementite is more thermally stable compared with secondary excess cementite. It does not dissolve at 1000°C, forming an anomalous ferrite-carbide structure (Fig. 6*a*). *Fifth*, the pattern at the surface of alloys is reminiscent of damask

steel, and appears during polishing before etching (Fig. 6*b*). The carbide contour (damask pattern) depends on the plane of inclination with respect to the main axis of previous dendrites and on the forging method. On grinding away maxima and minima after forging, the carbide contour is repeated topographically, forming a typical damask pattern, visible after polishing.

The structure of alloy No. 1 corresponds to that of tool steels of reduced hardenability. Alloys Nos. 2 and 3 with respect to composition, structure, and physical properties correspond to steels of the carbide class.

Production regimes have been established providing preparation of unalloyed steels and alloys of the carbide class with a carbon content of 1.5–3.0% (all the rest of the elements in hundredths or thousandths). Features of cementite and ledeburite shape change have been studied during thermomechanical treatment. It has been detected that particles of eutectic cementite form during dynamic recrystallization of ledeburite. The main metallographic indicators have been revealed for unalloyed steels of the carbide class. It has been established that damask steel is chemically inhomogeneous unalloyed steel of the carbide class.

A study of the physicomechanical properties of unalloyed steels of the carbide class is a separate task for future research.