ite leads to formation of a small amount of eutectoid composition austenite. Diffusion distances for carbon in ferrite remain quite large, and in order to develop the $F \rightarrow A$-transformation considerable overheating is necessary. On heating steel with a higher carbon content above $A_{c1}$ a large amount of eutectoid composition austenite forms in its structure, diffusion distances for carbon are shortened, and with less heating conditions are created for achieving the critical size of new phase nuclei and their growth at the expense of ferrite supersaturated with carbon in diffusion zones. In this way a threshold value for steel carbon content should exist which provides quite small diffusion distances for carbon satisfying conditions for the process of austenite transformation into ferrite grains without overheating. It should be noted that such structural factors as grain size, dispersion, and uniformity of pearlite distribution in the steel in this case will also play a governing role, somewhat changing the threshold value for carbon content. For the steels studied by us containing more than 0.035% C these conditions are accomplished when the $F \rightarrow A$-transformation within ferrite grains develops immediately after the finish of pearlite transformation, and a latent period in the intercritical range is completely absent.

**CONCLUSIONS**

1. On heating low-carbon steels containing 0.025-0.35% C at rates of 7.5-200 deg/min in the intercritical temperature range $A_{c1}-A_{c3}$ it is possible to isolate three basic temperature ranges: transformation of pearlite into austenite $A_{c1s}-A_{c1f}$, absence of transformation, and transformation of ferrite into austenite $A_{c3s}-A_{c3f}$.

2. The range for absence of transformations between temperatures for the finish of $P \rightarrow A$-transformation and the start of $F \rightarrow A$-transformation is greater as carbon content in the steel decreases. With a carbon content in the steel of 0.35% or more $A_{c1f}$ and $A_{c3s}$ almost coincide.

**STRUCTURE AND DUCTILITY OF EUTECTIC TYPE IRON–CARBON ALLOYS**

P. F. Nizhnikovskaya

Cast white cast iron exhibits high hardness and wear resistance, although it has low mechanical properties and fatigue strength. As a result of deformation the strength, impact strength, and fatigue strength of cast iron increase, but wear resistance remains at the previous level [1].

In industry, forming of cast iron by rolling or forging is difficult as a result of its low ductility in the cast condition, and this is due to presence in the structure of eutectic which forms on the basis of a brittle carbide phase. It is only possible to deform cast iron under all-round compression conditions.

In iron–carbon alloys a considerable amount of bicrystalline eutectic is observed which may be separated into two groups [2]: eutectics whose matrix phase is carbide, and eutectics in which carbide reinforces a metal base.

The most typical and widespread eutectic of the first group is ledeburite, i.e., a cellular austenite–cementite eutectic forming in unalloyed and low-alloy cast irons (Fig. la). The second group relates to eutectics based on special carbides, i.e., $M_7C_3$, $M_6C$, $M_5C$, MC, which crystallize in high-alloy cast irons and steels of the eutectic class (fast-cutting, certain forging materials, etc.) (Fig. lb). In hypoeutectic alloys eutectics are located in interdendritic areas and between branches of primary solid solution, forming a continuous shell around them (Fig. lc). In alloys with eutectics of the first group grains of both primary and eutectic solid solution are separated from each other by a carbide coating, and in alloys with eutectics of the second group continuity of the lamellar metal base is maintained. Contact between dendrites and primary solid solution in alloys with ledeburite eutectic only develops with a eutectic level below 30%, and this corresponds to a carbon content in low-alloy cast iron of about 2.5–2.7%.

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As a result of high Pearls–Nabarro barriers, caused by the presence of rigid directional covalent bonds, and the blocking action of carbon, cementite has an inelastic lattice. Stress concentration, under the action of which dislocation movement is possible, is very high, but the rate of their movement in cementite is very low. At low and intermediate temperatures critical shear stress exceeds ultimate strength and it causes brittle failure of cementite. Critical shear stress at elevated temperature is reduced, and it depends on deformation rate. With low deformation rate ($10^{-2}-10^{-5}$ sec$^{-1}$) marked dislocation movement connected with their sliding is observed at temperatures above 550°C. The nature of failure is brittle-ductile. Macroscopic deformation of cementite is only possible at temperatures above 1100-1190°C when dislocation creep commences. However, the temperature for hot deformation of cast iron is limited by the eutectic melting temperature, which for ledeburite cast iron does not exceed 1140°C. In hot torsion tests at 1100-1120°C as a result of local warming up eutectic melting is observed. In view of this it is understandable why cast iron of around-eutectic composition is not subjected to forming under industrial conditions, and specimens of cast iron containing more than 3% C with hot torsion tests in the range from normal to melting temperature fail in a brittle fashion.

Similar to pearlite colonies, with deformation of eutectic and hypoeutectic alloys with eutectics of the second group, and also in alloys with eutectics of the first group, but with very low levels of eutectic, at first solid solution deforms and around carbides there is a strain localization increasing in proportion to the amount of carbide phase. Eutectic carbides are under the action of normal compressive stresses from the surrounding solid solution and shear stresses caused by cohesive forces at the surface separating eutectic carbide and austenite, and they experience elastic deformation. With development of deformation in a metal matrix localized stresses at the interface with eutectic carbides increase, and accumulated energy may be realized in the form of failure or plastic deformation of carbides.

In contrast to ledeburite cast irons, deformation is possible around the eutectic composition in test alloys and is observed for carbides of practically all types: $M_3C$, $M_6C$, $M_2C$, and $MC$. They are drawn out, they bend, and they narrow down to separate into parts. Brittle carbides deform plastically as a result of the fact that compressive stresses from the surrounding metal base hinder crack propagation in them. The shear stress may exceed the stress necessary for dislocation movement.