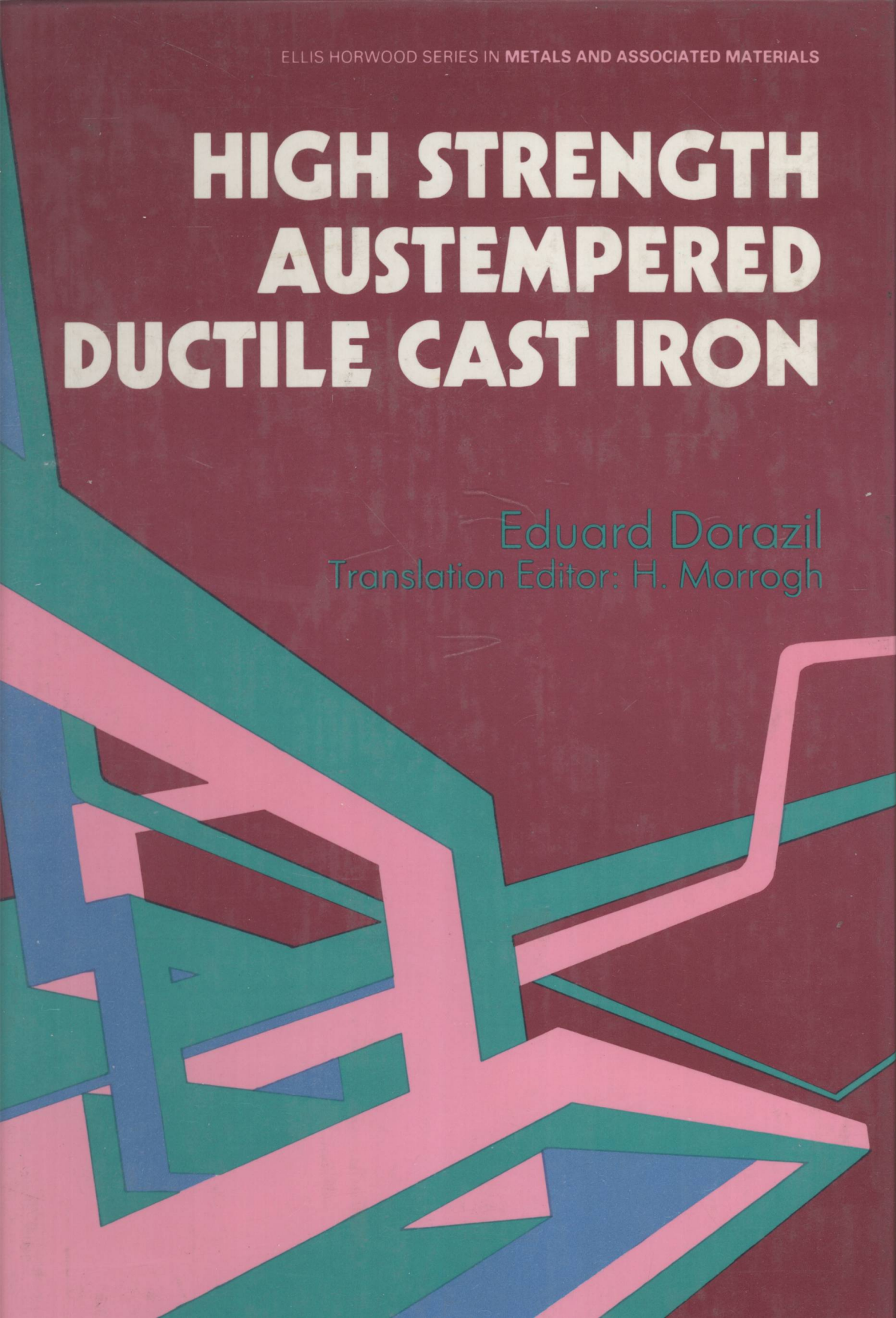


ELLIS HORWOOD SERIES IN METALS AND ASSOCIATED MATERIALS

HIGH STRENGTH AUSTEMPERED DUCTILE CAST IRON

Eduard Dorazil
Translation Editor: H. Morrogh



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Eduard Dorazil

1

Introduction

Cast irons were among the first metallic materials in engineering use and are still widely used even today. They are easily cast and the continuous development of metallurgical production techniques has resulted in improved engineering properties, giving better technical performance and service reliability of iron castings.

An important development in the production of cast irons was brought about by the discovery of methods of controlling solidification, to give graphite in spheroidal form [1]. The technological properties of this type of cast iron – ductile iron (also known as “nodular iron”, “spheroidal graphite iron” and “S. G. iron”) – differ considerably from those of grey cast iron since both its strength and ductility characteristics are primarily determined by the properties of the metallic matrix in a similar manner to those of steels.

Although ductile iron castings are still mostly used with a ferritic, ferritic-pearlitic, or pearlitic matrix, higher-strength matrices can also be used. Taking advantage of the improved metallurgy of production of ductile iron there has also been a trend towards the production of high strength ductile iron with sufficient ductility and toughness to find application in highly stressed castings made in large series. The higher costs of production arising from the more exacting technology used may be outweighed by a more useful combination of properties.

In ductile iron castings high strength can be obtained in various ways. In recent years, much attention has been focussed on ductile iron with a bainitic matrix, which gives a specially favourable combination of strength properties, ductility, and toughness. This new engineering material is the concern of the present book in which the author summarizes the results of his many years' study and complements them with data published in the technical literature.

The introductory part of the book (Chapter 2) deals with the causes of the segregation of alloying elements during the solidification of ductile cast iron.

This segregation results in chemical and structural micro-inhomogeneity of the matrix, which can severely affect the kinetics of the bainitic transformation and thus also the constitution of bainitic structure at various points of the matrix. With an increasing degree of chemical and structural micro-inhomogeneity of the matrix the desirable mechanical properties of austempered ductile iron are less easy to develop.

Chapter 3 gives detailed information about the isothermal bainitic transformation in Fe–C–Si technical alloys. The most significant factors which affect the constitution and morphology of bainite and the kinetics of the bainitic transformation are described. The data on the bainitic transformation in the Fe–C–Si system provides the theoretical foundation for the explanation of the changes in mechanical properties of austempered ductile iron accompanying the changes in its structure. By applying the knowledge of the relations between structure, mechanical properties, and heat treatment conditions the properties may be controlled and the heat treatment procedure best suited for a given application can be chosen.

In Chapter 4 the effects of heat treatment conditions, chemical composition, and chemical micro-inhomogeneity of the matrix on the mechanical properties of austempered ductile iron are discussed. Experimental data have been gathered on the mechanical properties of unalloyed and low-alloy ductile irons of different chemical compositions. This makes it possible to choose the appropriate chemical composition for the ductile iron and an optimum procedure for the austempering heat treatment for a given application and for the required level of mechanical properties.

Chapter 5 is devoted to the hardenability of ductile iron during austempering heat treatment. The methods of evaluating bainitic hardenability are discussed together with the effect of alloying elements on bainitic hardenability. The application of CCT diagrams and data on martensitic hardenability with a view to determining the hardenability during austempering are also dealt with.

The part devoted to the behaviour of austempered ductile iron under mechanical stresses (Chapter 6) deals especially with the effect of static tensile loading on the deformation and fracture characteristics of austempered ductile iron at room temperature and at lower temperatures. The impact and the fatigue properties, wear resistance, and contact-fatigue properties of austempered ductile irons are also dealt with.

The concluding Chapters 7 to 9 give data on some practical aspects of the production of austempered ductile iron castings. Examples are given of industrial applications showing the possibility of employing this new ductile iron in engineering applications. A comprehensive summary and conclusions are given in Chapter 9.

2

Chemical micro-inhomogeneity of the ductile iron matrix

The properties of a flake graphite iron casting are determined on the one hand by the metallic matrix properties, and by the shape, distribution, and amount of graphite on the other. The decisive factor which determines the extent to which the strength and especially the ductile properties of the matrix are made use of is the shape of graphite. In contrast to flake graphite, the spheroidal nodules of graphite in ductile iron have a much smaller internal notch effect, which allows a comparatively wide range of mechanical properties to be developed by variations in the matrix structure. Nevertheless the manufacturing process must be such as to give nodular graphite as nearly perfectly spheroidal as possible, with spheroids uniformly distributed throughout the matrix [2].

The nature of the solidification process in these materials also determines the structure and extent of chemical inhomogeneity of the matrix, which depends on the type of cast iron [3] and affects subsequent phase transformations in the solid state [4] and therefore the matrix structure and properties. Although in the case of a grey cast iron with a lower strength the matrix may not have a completely dominant role, it is the most important feature of ductile cast irons of higher strength.

2.1 EUTECTIC SOLIDIFICATION OF DUCTILE IRON

Since 1946, when Morrogh and Williams [1] discovered a method of influencing the solidification of liquid cast iron to give a eutectic with nodular graphite, numerous publications have dealt with this subject [5, 6, etc.], and the material itself has been produced industrially for almost forty years. In spite of the many theoretical and experimental studies no generally acceptable explanation of the nucleation and growth of nodular graphite has been proposed.

It follows from a critical analysis of the conditions for the formation of nodular graphite, which was submitted by Ryš [7] and which is still valid, that the occurrence of nodular graphite is promoted especially by additions of desulphurizing and deoxidizing elements (Ce, Mg, Ca, Li, Ba, K, Na). The finding that nodular graphite can form even in a melt of high purity, cooled under controlled conditions but not treated with active additions, seems to suggest that other factors which are effective in the formation of nodular graphite, may also be the purity of melt and the rate of cooling. It can therefore rightly be assumed that the occurrence of nodular graphite in the as-cast condition is conditional especially on supercooling of the melt, which can be achieved in various ways. This theory has been adequately confirmed by experiments [8–10, etc.].

It has also been well established that, in ductile iron, graphite nodules normally nucleate directly in the melt during solidification [9, etc.] and that they are more or less perfect spherulites each consisting of radially arranged fibres growing from a common nucleus [11]. The preferred growth direction of a spherulite nucleus is the [0001] direction of the elementary cell of the hexagonal lattice of graphite. Since, however, the causes of the different growth rates in the two main systems of crystallographic planes of the graphite lattice have not yet been explained satisfactorily, there are still many views about the growth mechanism of a graphite spherulite.

Some authors [12, 13] suggest that the fundamental mechanism in the growth of such a polycrystalline graphite formation may be the branching of radially growing fibres of graphite, which does not proceed in any definite crystallographic direction but rather with low arbitrary angles [14]. A dislocation pattern was suggested [15] which was based on the growth of a graphite fibre along a screw dislocation. Minkoff [13] later developed this theory by suggesting the necessary participation of foreign atoms which, located on the steps of growth spiral, limit the growth in [1010] directions; their presence in the graphite lattice gives rise to stresses and defects in the lattice which cause the branching. These notions have found support in experiments [16, etc.] which suggest the possibility of the nodule of graphite growing with the involvement of screw dislocations. Contrary to these views, which attribute an important role to lattice defects, branching could also be a result of the nucleation [17] made possible by considerable supercooling of the melt.

The formation of spherulites is sometimes explained also by the existence of the constitutional supercooling of the melt ahead of the graphite/melt interface which is due to some elements being pushed through the moving interface into the melt. As a consequence, the interface is split into protrusions the tips of which grow quickly into the constitutionally supercooled melt [14, 18].

The radial internal structure of a graphite nodule may result from foreign atoms affecting the rate of formation of plane nuclei and the rate of their growth

[19, 20]. The difference in the intensity of adsorption processes at different points of the graphite/melt interface is thought to be the reason for the different rates of lengthening and thickening growth of the fibres forming the graphite spherulite.

Contemporary ideas on the formation of a eutectic with nodular graphite can be summarized as follows: after the formation of graphite nucleus its radial symmetrical growth takes place in the melt which loses carbon not only in the space between radially growing fibres but also from the layer around the nodule of graphite. Due to the reduced carbon content this layer of melt suffers severe constitutional supercooling and, at a certain stage of growth, gives rise to an austenitic envelope around the graphite nodule. It has been suggested [18] that the thickness of the austenitic envelope will depend on the rate of solidification. In the case of slow solidification the carbon can diffuse from the more distant points of the melt and the layer deficient in carbon and the austenitic envelope will be formed later thus giving relatively large graphite nodules. Conversely, at a higher rate of solidification there is an earlier end to the growth of graphite nodule in direct contact with the melt.

This mechanism of solidification of the graphitic eutectic of ductile iron takes into account the nucleation of the graphite nodule in the melt and its growth, initially in direct contact with the melt, and is relatively adequately supported by experiments [9, etc.]. It can also be assumed that the nuclei are crystallographically suitable foreign particles [10, 21] since contemporary knowledge does not provide sufficient ground to postulate another mechanism of graphite nucleation. Some findings [22] suggest the possibility of graphite nucleation being affected by primary austenite dendrites, which can also be observed in ductile irons of eutectic composition.

The existence of a more or less regular austenitic envelope appearing around the graphite nodule at a certain growth stage can be inferred from the results of interrupted solidification tests. Using this method, Jolley [23] made some interesting observations: in quenched samples he found carbide filaments, in the austenite envelopes, joining the graphite nodule to the unsolidified melt. He used this finding as the basis for a hypothesis about the growth of graphite nodules remaining in permanent contact with the melt through narrow channels in the austenitic envelope but this theory requires further verification.

As can be seen, the mechanism of the nucleation and growth of a eutectic with nodular graphite has not yet been explained satisfactorily. Although problems of the solidification of cast irons have attracted continued attention for decades, the knowledge obtained is still only of a qualitative nature. This emphasizes the complexity and difficulty of studying eutectic transformations in the Fe-C-Si system and the much more complicated industrial cast irons.

2.2 SEGREGATION OF ALLOYING ELEMENTS DURING DUCTILE IRON SOLIDIFICATION

As suggested by experimental data [24, 25] it can be assumed that in alloys based on the iron-carbon system the segregation of alloying elements is greatly affected on the one hand by the energy demands of the transfer of individual elements from the melt to the solid phase, on the other hand by their relationship to the thermodynamic activity of carbon. During solidification we therefore cannot expect the segregation of all the elements in phases in which they have maximum solubility, i.e. in the melt, but their progressive segregation either in the separating solid or in the melt on the basis of the thermodynamic requirements of the equilibrium between the melt and the solid solution forming from the melt.

If we consider generally an alloy Fe-Y, in which the element Y is soluble without limitation in both the melt and the solid phase, then the condition of equilibrium between the two phases at a constant temperature and pressure can be expressed by the equality of partial molar free enthalpies of the element Y in melt (\bar{G}_Y^L) and in solid phase (\bar{G}_Y^S):

$$\bar{G}_Y^L = \bar{G}_Y^S \quad (2.1)$$

This condition can be rewritten in the form:

$${}^0G_Y^L + RT \ln a_Y^L = {}^0G_Y^S + RT \ln a_Y^S \quad (2.2)$$

where: ${}^0G_Y^L, {}^0G_Y^S$ is molar free enthalpy of element Y in the fluid and solid state, respectively,

a_Y^L, a_Y^S is thermodynamic activity of element Y in the melt and in the solid phase, respectively.

Since it holds in general: $a_Y = \gamma_Y x_Y$, it can further be written:

$${}^0G_Y^L + RT \ln \gamma_Y^L + RT \ln x_Y^L = {}^0G_Y^S + RT \ln \gamma_Y^S + RT \ln x_Y^S \quad (2.3)$$

and also:

$$RT \ln \frac{x_Y^L}{x_Y^S} = {}^0G_Y^S - {}^0G_Y^L + RT \ln \frac{\gamma_Y^S}{\gamma_Y^L} \quad (2.4)$$

where: x_Y^L, x_Y^S is the molar fraction of element Y in the melt and in the solid phase, respectively,

γ_Y^L, γ_Y^S the activity co-efficient of element Y in the melt and in the solid phase, respectively.