Z. Ławrynowicz, S. Dymski

Department of Materials Science and Engineering, Mechanical Engineering Faculty,
University of Technology and Life Sciences in Bydgoszcz, av. Kaliskiego 7,
85-796 Bydgoszcz, Poland

CARBON DIFFUSION DURING BAINITE REACTION IN AUSTEMPERED DUCTILE IRON

ABSTRACT

The paper presents an investigation of the carbon concentration in the residual austenite and the time required for the diffusion of carbon out of supersaturated sub-units of ferrite into the retained austenite. Experimental measurements of volume fraction of bainitic ferrite and volume of the untransformed austenite indicate that there is a necessity of carbides precipitation from austenite. A consequence of the precipitation of cementite from austenite during austempering is that the growth of bainitic ferrite can continue to larger extent and that the resulting microstructure is not an ausferrite but is a mixture of bainitic ferrite, retained austenite and carbides. Additionally, carbon concentration in the residual austenite was calculated using volume fraction data of austenite and a model developed by Bhadeshia based on the McLellan and Dunn quasi-chemical thermodynamic model. The comparison of experimental data with the T_0 , T_0' and Ae_3' phase boundaries suggests the likely mechanism of bainite reaction in cast iron is displacive rather than diffusional. The carbon concentration in retained austenite demonstrates that at the end of bainite reaction the microstructure must consist of not only ausferrite but additionally precipitated carbides.

Key words: carbon diffusion, decarburization, bainite, ductile iron ADI

INTRODUCTION

The development of austempered ductile iron (ADI) is a major achievement in cast iron technology. The starting material for the development of ADI is the high quality ductile or nodular cast iron. It is then subjected to an isothermal heat treatment process known as “austempering”. The attractive properties of ADI are related to its unique microstructure that consists of ferrite and high carbon austenite. Because of this microstructure, the product of austempering reaction in ductile iron is often referred to as “ausferrite” rather than bainite [1,2]. Ausferrite consists of ferrite in a high carbon, stabilized austenite. If ADI is held for long time periods, the high carbon austenite will eventually undergo a transformation to bainite, the two phase ferrite and carbide ($\alpha + \text{Fe}_3\text{C}$). In order for this transformation to occur, longer periods of time are typically needed – much longer than would be economically feasible for the production of ADI. Once the ausferrite has been produced, the components are cooled to room temperature. The cooling rate will not affect the final microstructure as the carbon content of the aus-

tenite is high enough to lower the martensite start temperature to a temperature significantly below room temperature.

Carbides can be suppressed by alloying with elements such as Si and Al. As shown in Fig. 1 bainite growth by a diffusionless mechanism has to occur at a temperature just below T_0^* when the free energy of bainitic ferrite falls below that of austenite of the same composition.

Fig. 1. Schematic illustration of the origin of the T_0 curve on the phase diagram. □ and □ refer to ferrite and austenite, respectively. T_1 is the temperature corresponding to the free energy curves [3]

At T_1 the bainitic reaction will cease when the carbon concentration of the residual austenite (x_γ) approaches the T_0 curve:

$$x_\gamma \rightarrow x_{T_0} \quad (1)$$

Then it is impossible for the displacive growth of bainite to proceed further. It is therefore possible to predict the carbon concentration in the retained austenite at the end of bainite transformation simply from chemical composition if the carbon to distribute evenly in the austenite. On the other hand, if bainite grows in a manner without supersaturation of ferrite, the reaction should be able to precede until the carbon concentration in the retained austenite reaches the paraequilibrium $Ae_3'^{**}$ phase boundary.

During isothermal transformation the excess carbon in the bainite partitions into the residual austenite, forcing the next plate to grow from carbon enriched austenite. The

* The T_0 temperature can be defined [3] such that stress free austenite and ferrite of the same composition (with respect to both the interstitial and the substitutional alloying elements) are in metastable equilibrium. Thus any displacive transformation involving a full supersaturation of carbon (i.e. bainitic ferrite would then inherit the carbon content of the parent austenite) can occur only below the appropriate T_0 temperature. Strain energy would have effect of shifting curve to lower carbon concentration, T_0' curve [4].

** Hultgren [5], introduced the term "paraequilibrium" to describe the constrained equilibrium between two phases which are forced to have the same substitutional to iron atom ratio, but which achieve equilibrium with respect to carbon. Ferrite formation under conditions of paraequilibrium essentially implies that the substitutional lattice is configurationally frozen and transformation occurs at the rate controlled by the diffusion of carbon in the austenite, the driving force for paraequilibrium transformation is dissipated in this process alone.

process finally ceases as the austenite carbon content reaches T_0 value, leading to the so-called ‘the incomplete reaction phenomenon’ [4]. This also explains why the degree of transformation to bainite is zero at the bainite-start temperature (B_S) and increases with undercooling below B_S .

The purpose of the present paper is to demonstrate how a thermodynamic method can be used for solving a problem of the mechanism of bainite reaction in ADI and determination of the carbon concentration in the retained austenite. This should in principle enable to examine the partitioning of carbon from supersaturated ferrite subunits into adjacent austenite and the carbon content in retained austenite using analytical method.

MATERIAL AND METHODS

The chemical composition of the experimental ductile iron is listed in Table 1. The concentration of alloying elements in the matrix is obtained from the chemical analysis. Ductile iron blocks were produced in a commercial foundry furnace. The melt was poured into a standard Y block sands molds (ASTM A-395), which ensured sound castings. Specimens austenitized at $T_\gamma=950$ for 60 minutes were rapidly transferred to a salt bath at austempering temperatures 250, 300, 350 and 400°C, held for 15, 30, 60, 120 and 240 minutes and then water quenched to room temperature. The microstructure of the as-cast material matrix contains 40% ferrite and 60% pearlite, however graphite nodules in material is 11.5%.

After heat treatment, the samples were prepared for metallographic analysis. The samples were etched using 2% nital. Optical micrographs were taken with a Nikon camera attached to a light microscope.

Table 1. Chemical composition of ductile cast iron ADI, wt-%

C	Si	Mn	P	S	Mg	Cr	Ni	Mo
3.21	2.57	0.28	0.06	0.01	0.024	0.036	0.098	0.015

The X-ray investigations were performed on the specimens heat treated after specific time of the isothermal bainite reaction at the given temperature. The total volume fraction of the retained austenite was measured from the integral intensity of the (111) γ and (011) α peaks. The presence of high silicon content in ADI retards the formation of cementite in ferrite and austenite. The carbon concentration was calculated from measured lattice parameter of the retained austenite. The 2θ values for austenite peaks were used to calculate the d spacing with Bragg’s law and then the lattice parameters. The lattice parameter of austenite (a_γ) is related to the known relationship between the parameter and the carbon concentration [6]:

$$a_\gamma \text{ (nm)} = 0.3573 + 0.0033 x_\gamma \quad (2)$$

where x_γ is the carbon concentration in austenite, in weight %.

The matrix carbon concentration, x_γ^m , of the ductile iron was also determined experimentally with Dron 1.5 diffractometer using $Co K_\alpha$ radiation on specimens austenitized at 950 °C for 60 minutes and quenched to ambient temperature. It was found that after quenching the calculated carbon content in matrix is $x_\gamma^m=1.044\%C$ and measured carbon content is $x_\gamma^m=1.05\%C$, thus, the measured values were taken for further calculation.

PHASE DIAGRAM OF AUSTEMPERED DUCTILE IRON

It is usually assumed that the point where the microstructure of austempered ADI ceases to change represents full transformation. But in case of bainitic transformation, reaction ceases before the parent phase (austenite) has completely transformed. It means that at any temperature below B_s and in the absence of any interfering secondary reactions only a limited quantity of bainitic ferrite forms before the reaction terminates.

The determined carbon concentrations of the residual austenite at the different stages of the formation of bainite are compared with the T_0 , T_0' and Ae_3' phase boundaries for investigated ADI in Figure 2. The diagram was calculated as in Ref. [7] using a model developed by Bhadeshia [8] based on the McLellan and Dunn quasi-chemical thermodynamic model [9]. The martensite reaction starts temperature M_s is also marked on this diagram. The paraequilibrium phase boundary is chosen because no substitutional alloying element partitioning occurs during bainite formation.

Fig. 2. The calculated phase boundaries Ae_3' , T_0 and T_0' for the investigated ADI together with all the experimental data of the measured carbon contents of the austenite

In presented diagram the reaction is found to stop when the average carbon concentration of the residual austenite is closer to the T_0 , T_0' curves than the Ae'_3 boundary. The presented results can be explained when it is assumed that bainitic ferrite grows without diffusion, but any excess of carbon is soon afterwards rejected into the residual austenite by diffusion [10,11]. This makes more difficult for subsequent bainitic ferrite to grow, when the austenite becomes stabilised by increased carbon concentration. The maximum extent to which the bainite reaction can proceed is therefore determined by the composition of the residual austenite. A stage where diffusionless growth becomes thermodynamically impossible and the formation of bainitic ferrite terminates is where the carbon concentration of the austenite reaches the T_0' or T_0 curves. Thus, the incomplete reaction phenomenon supports the hypothesis that the growth of bainitic ferrite occurs without any diffusion with carbon being partitioned subsequently into the residual austenite. The reaction is said to be incomplete, since the austenite has not achieved its equilibrium composition (given by the Ae'_3 curve) at the point the reaction stops. If on the other hand, the ferrite grows with an equilibrium carbon concentration then the transformation should cease when the austenite carbon concentration reaches the Ae'_3 boundary [4, 10-13].

Thus, it is found experimentally that the transformation to bainite does indeed stop close to the T_0 boundary (Fig. 2). Similar results have previously obtained by Bhadeshia and Christian for other alloys [12].

In Figure 2 the reaction seems to stop closer to the T_0 line than to the T_0' boundary. This might be explained by the fact that the T_0' line accounts for 400 J/mol of stored energy in the bainite. If this energy is reduced by plastic deformation of the surrounding austenite then a higher volume fraction of bainite should be able to form.

The thermodynamic restriction imposed by the T_0 curve on the extent of bainite transformation can result in the formation of pools of retained austenite with a coarse, blocky morphology, Fig. 3.

Fig. 3. Carbon replica taken from sample austenitised at 950 °C and austempered at 400 °C for 120 min. This replica reveals the presence of carbides inside of bainite (arrowed) and shows large pools of blocky retained austenite

When the matrix of ADI only consist of ausferrite, thus:

$$V_{\gamma} + V_{\alpha} = I \quad (3)$$

and the permitted fraction of bainite (V_α) can be determined from Lever rule applied to the T_0 curve, Fig. 4.

Fig. 4. Application of the Lever-rule to the T_0 curve allows the estimation of the permitted fraction of bainite (V_α) at any temperature (where for 950°C $\bar{x} = 1.05 \text{ wt.\% C}$ $x_\alpha = 0.03\% \text{ C}$)

The maximum volume fraction of retained austenite (V_γ) will then equal $1 - V_\alpha$.

Coarse carbides were not observed but it seems from the micrographs that more fine dispersed particles of carbides can be seen inside sheaves of bainite (Fig. 3, arrowed). In case of carbides precipitation the maximum volume fraction of bainitic ferrite (V_α) can be calculated using the following equation [14]:

$$V_\alpha = \frac{x_{T_0} - \bar{x}}{x_{T_0} - x_\alpha - x_C} \quad (4)$$

where V_α is volume fraction of bainitic ferrite, $\bar{x}$ is the average carbon concentration in the matrix of the alloy, x_α is the paraequilibrium carbon concentration in the bainitic ferrite (0.03 wt. %), x_{T_0} is the carbon concentration of the austenite corresponding to the T_0 curve, x_C is the amount of carbon, which is tied up as carbides (cementite).

Thus, the maximum volume fraction of bainite taking into account cementite precipitation can be calculated using the relationship (4).

It is seen in Fig. 5 that precipitation of cementite leads to an increase of volume fraction of bainitic ferrite. Carbides locally reduce the carbon content of the parent austenite and increase the driving force for further ferrite growth.

Cementite can precipitate from supersaturated bainitic ferrite and also from austenite. The growth of bainite is probably diffusionless but any excess carbon in the supersaturated ferrite soon afterwards partitions into the residual austenite or precipitates within bainitic ferrite in the form of carbides [15-17]. When the process of carbon partitioning into the residual austenite is rapid relative to that of carbide precipitation, the transformation product is called “upper bainite”, whereas “lower bainite” is obtained when some of the carbon supersaturation is relieved by precipitation within the bainitic ferrite [15].

Austenite is supersaturated with respect to cementite precipitation when $x_\gamma > x^{\gamma\theta}$, where $x^{\gamma\theta}$ denotes carbon concentration at the extrapolated A_{cm} boundary. This means for the bainite reaction, that $x_{T_0} > x^{\gamma\theta}$ since the growth of bainitic ferrite stops when x_γ reaches the value x_{T_0} given by the T_0 curve of the phase diagram. A consequence of the

precipitation of cementite from austenite is that its carbon concentration drops below x_{T_0} , so that the growth of bainitic ferrite can continue to an extent larger than would be otherwise possible (see Fig. 5).

Fig. 5. Calculated the maximum volume fraction of bainite in investigated ADI taking into account cementite precipitation in the range from 0% to 15 wt % Fe_3C

METHOD OF CALCULATION OF THE REDISTRIBUTION OF CARBON

The time t_d needed to decarburise the ferrite is intuitively expected at least to be comparable to that required for a subunit to complete its growth. If t_d is small relative to the time required to relieve the carbon supersaturation by the precipitation of carbides within the ferrite, then upper bainite is obtained, otherwise lower bainite forms [4,15].

Kinsman and Aaronson [18] first considered the kinetics of the partitioning of carbon from bainitic ferrite of the same composition as the parent phase. For a plate of thickness w_α the flux of carbon is defined along a coordinate z normal to the α/γ interface, with origin at the interface and z being positive in the austenite (Fig. 6).

The method used to calculate the time of decarburizing of bainitic ferrite subunits is based on the hypothesis that transformation to bainite can only occur in regions of austenite where $x_\gamma \leq x_{T_0}$, where x_γ is the carbon concentration in austenite and x_{T_0} is the carbon concentration corresponding to the T_0 curve. As a subunit of bainitic ferrite forms it partitions its excess carbon into the retained austenite. This creates a carbon

diffusion field around the subunit. Another parallel subunit (of the same sheaf) which forms subsequently can only approach the original subunit to a point where $x_\gamma \leq x_{T_0}$. The method assumes that the interval between subunit formations is larger than the time required to decarburize each subunit.

Fig. 6. Schematic diagram of method used in estimating the time of decarburising the bainitic ferrite subunits. Sub-unit 1 forms first and subunit 2 and 3 and next is allowed to approach it to point where $x_\gamma \leq x_{T_0}$ (distance of this point from subunit 1 is denoted w_γ). This is in fact the thickness of the retained austenite film. The mean thickness of the retained austenite films is almost tenfold thinner (0.01-0.02 μm) than the average thickness of the bainitic ferrite subunits (~0.2 μm)

DECARBURISATION OF SUPERSATURATED BAINITIC FERRITE SUB-UNITS

The problem becomes a calculation of the sum of the decarburization times of all bainite subunits that are existing on the coordinate connecting the nearest graphite nodules Fig. 7).

The average carbon diffusion distances also depend on the mean spacing among the graphite nodules. Figure 7 shows a photomicrograph that contains graphite nodules with diverse distance among them, changing from about 150 to 50 μm (marked z_1 and z_2 in Fig. 7). Thus, the average distance among nodules in examined ADI is assumed about 100 μm.

Fig. 7. Microstructure of ADI austenitized at 950 °C and austempered at 350 °C for 240 min. Etched with 2% Nital

The time needed to decarburize the ferrite matrix between the adjacent nodules of graphite t_{dz} :

$$t_{dz} = \sum_i t_{di} \quad (5)$$

where t_{di} is the time required to decarburise individual supersaturated bainitic ferrite subunit of specific thickness w_{ai} .

Because of the inhomogeneous distribution of carbon and other solutes in the matrix after transformation to bainite the retained austenite is enriched to a greater extent in the immediate vicinity to bainite platelets or in the region trapped between the platelets and in the eutectic cell boundary while other region contains relatively poor carbon [6,12,15]. The above effect can be exaggerated in ADI, since cast iron is usually extremely segregated. Martensite is usually found to be in the cell boundary which solidified last [19,20,21]. It indicates that the austenite in cell boundary is less enriched with carbon, and therefore is thermally unstable.

From the mass balance for carbon it follows that [12]:

$$(0.5w_\alpha)(\bar{x} - x^{\alpha\gamma}) = \int_{z=0}^{\infty} [x_\gamma\{z, t_d\} - \bar{x}] dz \quad (6)$$

where $\bar{x}$ is the average mole fraction of carbon in the alloy and $x^{\alpha\gamma}$ and $x^{\gamma\alpha}$ are the paraequilibrium carbon concentration in ferrite and austenite respectively. Since the diffusion rate of carbon in austenite is slower than in ferrite the rate of decarburization will be determined by the diffusivity in the austenite and the concentration of carbon in austenite at the interface remains constant for times $0 < t < t_d$ after which it steadily decreases as the austenite becomes homogeneous in composition. The equation (6) corrects an error in the original treatment, the error had the effect of allowing $t_d \rightarrow 0$ as the upper integration limit $\rightarrow \infty$. The function x_γ is given by [12]:

$$x_\gamma = \bar{x} + (x^{\gamma\alpha} - \bar{x}) \operatorname{erfc}\{z / 2(Dt_d)^{0.5}\} \quad (7)$$

This assumes that for $t < t_d$, the concentration of carbon in the austenite at the interface is given by $x^{\gamma\alpha}$.

The diffusion coefficient of carbon in austenite $D\{x\}$, is very sensitive to the carbon concentration and this has to be taken into account in treating the large concentration gradients that develop in the austenite. It is clearly necessary to know $D\{x\}$ at least over a range $\bar{x} \rightarrow x^{\gamma\alpha}$, although experimental determinations of $D\{x\}$ do not extend beyond $x = 0.06$. The value of D was calculated as discussed in Ref. [22,23]. The good approximation of the dependent diffusivity of carbon in austenite can be a weighted average diffusivity $\bar{D}$ [24]. Taking into account carbon concentration gradients it has been demonstrated that for most purposes a weighted average diffusivity $\bar{D}$ can adequately represent the effective diffusivity of carbon [22-25]. Weighted average diffusivity $\bar{D}$ is calculated by considering the carbon concentration profile in front of the moving ferrite interface as given by the following equation [22]:

$$\bar{D} = \int_{\bar{x}}^{x^{\gamma\alpha}} \frac{Ddx}{(x^{\gamma\alpha} - \bar{x})} \quad (8)$$

The calculated diffusion coefficients of carbon in austenite are listed in Table 2.

Table 2. The calculated diffusion coefficients of carbon in austenite $D\{x\}$ and a weighted average diffusivity $\bar{D}$ after austenitisation at 950 °C and austempering at 400, 350, 300 and 250 °C

Diffusion coefficients	Austempering temperature, °C			
	250	300	350	400
D [m ² /s]	0.2544×10^{-18}	0.4328×10^{-17}	0.4688×10^{-16}	0.3574×10^{-15}
$\bar{D}$ [m ² /s]	*	*	0.5013×10^{-15}	0.1672×10^{-14}

* Diffusion calculation outside of permitted range. Siller-McLellan model fails at high carbon concentrations evaluate $\bar{D}$.

On carrying the integration, the time required to decarburise a supersaturated bainitic ferrite subunit of thickness w_α is given by [12]:

$$t_d = \frac{w_\alpha^2 \pi (\bar{x} - x^{\alpha\gamma})^2}{16 \bar{D} (x^{\gamma\alpha} - \bar{x})} \quad (9)$$

where: $\bar{x}$ is the average carbon concentration in the alloy, $x^{\alpha\gamma}$ and $x^{\gamma\alpha}$ are the carbon concentrations in ferrite and austenite respectively, when the two phases are in paraequilibrium.

THE CALCULATION OF DECARBURISATION TIMES

For investigated ductile cast iron ADI our calculations show that t_d increases sharply as the thickness of the ferrite subunits increases, see Table 3.

Table3 . Decarburisation times (t_d) in seconds of distance of 50 µm, consisted of subunits with thickness: 0.1 µm; 0.2 µm; 0.5 µm; 1.0 µm; 10 µm; 50 µm

Ti, °C	Decarburisation times (t_d) in seconds of distance of 50 µm					
	50 µm	5x10 µm	50x1 µm	100x0.5 µm	250x0.2 µm	500x0.1 µm
400	110600	22125	2212	1128	451	225
350	234500	46905	4690	2361	950	470

The calculated times of partitioning are shown in Fig. 8 for thickness of bainitic ferrite phase equal 50 µm but consisted of subunits with different wideness: 0.1 µm; 0.2 µm; 0.5 µm; 1.0 µm; 10 µm and 50 µm.

The decarburisation time t_d as a function of α phase width increases with decreasing austempering temperature because the diffusion coefficient of carbon also decreases with temperature (Table 2). The decarburization time also increases as the thickness of the ferrite sub-unit increases (Fig. 8).

Furthermore, it is generally observed (Fig. 7) that the width of ferrite sub-units is highly diverse [15]. This reflects the possibility that cementite can precipitate in thicker bainite sub-units (when t_d is a long period of time) and in thinner laths has not during isothermal transformation. It is also consistent with the fact that upper and lower bainite often form at the same temperature in a given steel [4, 14, 15, 26, 27].

Fig. 8. The calculated times for decarburisation of ferrite plate with widthness of $50 \mu\text{m}$, consisted of subunits with thickness: $0.1 \mu\text{m}$; $0.2 \mu\text{m}$; $0.5 \mu\text{m}$; $1.0 \mu\text{m}$; $10 \mu\text{m}$ and $50 \mu\text{m}$ after austenitization at 950°C . Enlarged of an area in the bottom part in Fig. 8 corresponds to the marked area "a"

CONCLUSIONS

The paper presents an investigation of the carbon concentration of the residual austenite and the time required for the diffusion of carbon out of supersaturated subunits of ferrite into the retained austenite. This should in principle enable to examine the partitioning of carbon from supersaturated ferrite plates into adjacent austenite in ADI ma-

trix using analytical method. The results are discussed in the context of displacive mechanism of bainite transformation. The following conclusions were reached:

1. The bainite transformation ceases before the austenite achieves its equilibrium composition given by Ae'_3 boundary, so that the effect is called the "incomplete-reaction phenomenon". The T_0 restriction means that equilibrium, when the austenite has a composition given by the Ae_3 phase boundary, can never be reached, as observed experimentally.
2. The extent of transformation to bainite in ductile iron, as in steels, decreases when increasing the isothermal transformation temperature towards the bainite start temperature (B_s). This is because the austenite can only transform to bainite if its carbon concentration is less than a value x_{T_0} given by the T_0 curve.
3. The bainite transformation in cast iron is essentially identical to that in steel. In steel, it has been demonstrated that the carbon concentration of the residual austenite reaches the critical value represented by the T_0 curve will render the displacive bainite reaction to cease. Therefore the carbon concentration of austenite can be estimated by the thermodynamics principles described here. Since cast iron is extremely segregated, x_γ determined by X-ray diffraction is richer than that corresponds to the T_0 curve in ADI.
4. The carbon concentration of the residual austenite increases during bainitic transformation as a consequence of the increasing volume fraction of bainitic ferrite.
5. Precipitation of cementite leads to an increase of volume fraction of bainitic ferrite. Carbides locally reduce the carbon content of the parent austenite and increase the driving force for further ferrite growth.
6. Analytical calculations of the time required for the diffusion of carbon out of supersaturated subunits of ferrite into the retained austenite indicate that there is a necessity of carbides precipitation from ferrite.
7. A consequence of the precipitation of cementite from ferrite or/and austenite during austempering is that the growth of bainitic ferrite can continue to larger extent and that the resulting microstructure is not a pure ausferrite but is a mixture of bainitic ferrite, retained austenite and carbides.

REFERENCES

1. Chang L. C.: Carbon content of austenite in austempered ductile iron. *Scripta Materialia* 39 (1998), pp. 35-38.
2. Pietrowski S.: Nodular cast iron of bainitic ferrite structure with austenite or bainitic structure. *Archives of Materials Science* 18 (1997), pp. 253-273 (in Polish).
3. Christian J.W.: *Theory of transformations in metals and alloys*, 778, Oxford, Pergamon Press, 1965.
4. Bhadeshia H.K.D.H.: *Bainite in Steels*, Institute of Materials, 1-458, London, 1992.
5. Hultgren A.: *Jernkontores Ann.*, 135 (1951) 403.
6. Ławrynowicz Z., Dymski S.: Mechanism of bainite transformation in ductile iron ADI. *Archives of Foundry Engineering* 6 (2006), pp. 171-176 (in Polish).

7. Bhadeshia H.K.D.H.: Bainite: Overall Transformation Kinetics, *Journal de Phys.*, 43 (1982) C4 pp. 443-447.
8. Bhadeshia H.K.D.H.: A Rationalisation of Shear Transformations in Steels, *Acta Metallurgica* 29 (1981) pp. 1117-1130.
9. McLellan R.B., Dunn W.W.: *J. Phys. Chem. Solids*. Vol.30 (1969) 2631.
10. Ławrynowicz Z.: Mechanism of bainite transformation in Fe-Cr-Mo-V-Ti-C steel. *International Journal of Engineering*, 12 (1999) pp. 81-86.
11. Ławrynowicz Z., Barbacki A.: The mechanism of bainite transformation in Fe-Cr-Mn-Si-C steel. *Proc. of the Scientific Con. AMTECH'95*, Rousse, Bułgaria, 19-21 April 1995, pp. 1-8.
12. Bhadeshia H.K.D.H., Christian J.W.: Bainite in Steels. *Metallurgical Transactions A* 21A (1990), pp. 767-797.
13. Eric O. at al.: The austempering study of alloyed ductile iron. *Materials & Design* 27 (2006), pp. 617-622.
14. Takahashi M., Bhadeshia H.K.D.H.: A Model for the Microstructure of Some Advanced Bainitic Steels, *Materials Transaction, JIM*, 32 (1991) pp. 689-696.
15. Ławrynowicz Z.: Transition from upper to lower bainite in Fe-Cr-C steel. *Materials Science and Technology*. 20 (2004) pp. 1447-1454.
16. Ławrynowicz Z.: A discussion on the mechanism of bainite transformation in steels. *Technology and Materials*, Gdańsk, Politechnika Gdańska, 4, (2006), pp. 149-155 (in Polish).
17. Thomson R.C. at al.: Modelling microstructural evolution and mechanical properties of austempered ductile iron, *Materials Science and Technology*. 16 (2000) pp. 1412-1419.
18. Kinsman K. R., Aaronson H.I.: The transformation and hardenability in steels. *Climax Molybdenum Company*, Ann Arbor, MI, p.39, 1967.
19. Guzik S. E.: Austempered cast iron as a modern constructional material. *Inżynieria Materiałowa* 6 (2003), pp. 677-680. (in Polish).
20. Ławrynowicz Z., Dymski S.: Application of the mechanism of bainite transformation to modelling of processing window in ductile iron ADI. *Archives of Foundry Engineering* 6 (2006), pp. 177-182.
21. Kutsov A. at al.: Formation of bainite in ductile iron. *Materials Science and Engineering A* 273-275 (1999), pp. 480-484.
22. Bhadeshia H.K.D.H.: Diffusion of carbon in austenite. *Metal Science* 15 (1981), pp. 477-479.
23. Ławrynowicz Z.: Criticism of selected methods for diffusivity estimation of carbon in austenite. *Zeszyty Naukowe ATR nr 216, Mechanika* 43, (1998), pp. 283-287 (in Polish).
24. Siller R.H., McLellan R.B.: The Application of First Order Mixing Statistics to the Variation of the Diffusivity of Carbon in Austenite. *Metallurgical Transactions* 1 (1970), pp. 985-988.

25. Ławrynowicz Z.: Bainitic transformation: estimation of carbon diffusivity in austenite on the basis of measured austenite film thickness. *Zeszyty Naukowe ATR nr 216, Mechanika 43*, (1998), 289-297 (in Polish).
26. Ławrynowicz Z.: Carbon partitioning during bainite transformations in low alloy steels. *Materials Science and Technology 18* (2002) pp. 1322-1324.
27. Ławrynowicz Z., Barbacki A., Analiza mechanizmu izotermicznej przemiany balitycznej w stali Cr-Mn-Si. *Archiwum Nauki o Materiałach 17* (1996) pp. 127-147.