# Thermal Dependence of Austempering Transformation Kinetics of Compacted Graphite Cast Iron

K.F. LANERI, J. DESIMONI, R.C. MERCADER, R.W. GREGORUTTI, and J.L. SARUTTI

The evolution of the relative fraction of high-carbon austenite with austempering time and temperature was analyzed in a compacted graphite (CG) cast iron (average composition, in wt pct: 3.40C, 2.8Si, 0.8Mn, 0.04Cu, 0.01P, and 0.02S) at five different austempering temperatures between 573 and 673 K. Samples were characterized by Mössbauer spectroscopy, hardness measurements, and optical microscopy. During the first stage of transformation, the kinetics parameters were determined using the Johnson–Mehl's equation, and their dependence with temperature in the range from 573 to 673 K indicates that the transformation is governed by nucleation and growth processes. The balance between growth-rate kinetics and nucleation kinetics causes the kinetics parameter (k) to have a maximum at  $\approx 623$  K of  $3.9 \times 10^{-3}$ (s<sup>-1</sup>). The evolution of the C content in the high-carbon austenite was found to be controlled by the volume diffusion of carbon atoms from the ferrite/austenite interface into austenite, with a dependence of  $t^{0.40\pm0.05}$  on the austempering time (t).

#### I. INTRODUCTION

THE austempering transformation in cast irons, which leads to end products with the best mechanical properties, is described as a three-stage process.<sup>[1–4]</sup> In *stage I*, parent austenite transforms into acicular ferrite and high-carbon austenite ( $\gamma_p \rightarrow \alpha_{\rm Fe} + \gamma_{hc}$ ), forming a microstructure called ausferrite. Carbon atoms are rejected from the growing ferrite plates, causing the enrichment of the surrounding austenite. The driving force for this transformation arises from the carbon-concentration gradient set up in the austenite as result of local equilibrium at the ferrite/austenite interface. Nucleation and growth processes govern the transformation kinetics.<sup>[5,6]</sup> The solid-solid nucleation theory<sup>[7]</sup> postulates that the driving force for nucleation is proportional to the volume free-energy change and the volume strain energy, while the interfacial free energy of the cluster acts as a barrier to this process. In terms of the temperature dependence of the nucleation process, from expressions reported by Doherty,<sup>[8]</sup> it could be deduced that, as the undercooling increases, the volume free-energy change increases, promoting a higher nucleation rate. Several authors have treated the growth of a ferrite plate in an austenite matrix from a theoretical point of view,<sup>[9,10]</sup> proposing that the rate of growth is controlled by the diffusion of carbon atoms through the austenite away from the tip of the advancing particle.

The study of the thermal dependence of the austempering kinetics in cast irons can contribute to the determination of the times needed to reach the optimum mechanical properties. In the case of the compacted graphite (CG) cast iron, its high thermal-fatigue resistance makes the material optimum to be used in operational conditions of thermal cycling. This feature could be complemented by the improvement of the mechanical properties achieved through the austempering heat treatment.

A previous study<sup>[11]</sup> determined that the evolution of the fraction of  $\gamma_{hc}$  during the stage I transformation of CG cast irons has a sigmoidal behavior and the kinetics of transformations through the *k* and *n* parameters of the Johnson–Mehl's equation was quantified.<sup>[5,6]</sup> The results indicated that the transformation proceeds through a localized nucleation and a phase transformation controlled by an interface reaction.<sup>[6]</sup> After the nucleation sites get saturated, the advance of the transformation is further controlled by a diffusion process.<sup>[12]</sup> However, the dependence of the parameters on the temperature along stage I of the austempering kinetics still needs further understanding.

In this article, the investigation of the temperature dependence of the kinetics parameters for CG cast iron has been undertaken. The transformation kinetics was determined at temperatures between 573 and 673 K by Mössbauer spectroscopy, monitoring the evolution of high-carbon austenite. The ferrite/martensite microstructure and the austenite carbon–enrichment processes are also analyzed in detail. The results are discussed in the frame of nucleation, growth, and diffusion processes<sup>[5,6,13]</sup> and are compared with those reported in the literature on CG cast iron<sup>[4]</sup> and ductile cast iron.<sup>[14]</sup>

#### **II. EXPERIMENTAL**

A base metal of composition 3.40C, 1.5Si, 0.2Mn, 0.04 Cu, 0.01P, and 0.02S (in wt pct), was used to produce CG cast iron in a medium-frequency induction furnace, using the sandwich technique in the ladle to treat the liquid metal. The necessary amount of FeSiMgCeCa was added to obtain a CG morphology, and Mn and Si balances were attained by adding appropriate amounts of FeMn ( $C_{Mn} = 60$  wt pct) and FeSi ( $C_{Si} = 75$  wt pct). The resulting chemical composition was 3.40C, 2.8Si, 0.8Mn, 0.04Cu, 0.01P, and 0.02S (in wt pct).

Samples of 20 mm in diameter and 3-mm thick were taken from "Y-shaped" blocks (ASTM A-395) cast in sand molds. Heat treatments consisted of austenitizing at 1173 K

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Fig. 1—Typical Mössbauer spectra belonging to alloys austempered at 573 K (left) and at 673 K (right) at the austempering times indicated. At the bottom, the arrows indicate the signals corresponding to ferrite + martensite and austenite.

for 30 minutes and then quenching in a salt bath held at different temperatures, rangeing from 573 to 673 K, for times between 1 and 30 minutes.

The samples were specially prepared for Mössbauer spectroscopy by conventional grinding techniques, to reduce their thickness down to  $\approx$ 70  $\mu$ m, using a diamond paste of 6, 1, and 0.1  $\mu$ m for final polishing.

Mössbauer spectra were taken in transmission geometry using a <sup>57</sup>CoRh source of approximately 5 mCi intensity and were recorded in a standard 512-channel conventional constant-acceleration spectrometer. Spectra belonging to two ranges of velocities were taken in order to analyze in detail the different phases present in the samples. One range covered velocities between -8 and +8 mm/s and the other one covered velocities between -2 and +2 mm/s. Velocity calibration was performed against a 12  $\mu$ m-thick  $\alpha$ -Fe foil. All isomer shifts are referred to this standard at 298 K. Spectra were fitted to Lorentzian line shapes with a nonlinear least-squares program with constraints. For the effective thickness of the samples analyzed, no Voigt line-shape correction was necessary.<sup>[15]</sup>

Hardness tests were carried out with a standard Vickershardness machine, using a load of 30 kg. The hardness value obtained for each sample is an average of ten measurements.

In order to characterize the austempering microstructure, a Reichert optical microscope was used. The samples for optical microscopy were prepared by standard polishing techniques and etched with 2 pct Nital solution.

#### **III. RESULTS AND DISCUSSION**

Some typical wide-velocity-range Mössbauer spectra recorded to investigate the austempering kinetics of the stage

I transfomation are shown in Figure 1 for two austempering temperatures and three different times. The six broad lines comprising the sextets characteristic of the ferromagnetic phases,<sup>[16]</sup> *i.e.*, ferrite and martensite, are observed in the spectra together with paramagnetic signals associated with austenite.<sup>[17]</sup> The contribution of the ferromagnetic signal decreases with the austempering time, portraying the progress of the transformation. The ferrite/martensite subspectra were reproduced using three hyperfine interactions, whose average parameters were  $H_1$  (kOe)  $\approx 336 \pm 1$  and  $\delta_1$  (mm/ s)  $\approx 0.02 \pm 0.01$ ,  $H_2$  (kOe)  $\approx 205 \pm 1$  and  $\delta_2$  (mm/s)  $\approx 0.05 \pm 0.01$ , and  $H_3$  (kOe)  $\approx 279 \pm 2$  and  $\delta_3$  (mm/s)  $\approx$  $0.09 \pm 0.01$ . The first interaction is common to iron probes both in ferrite and martensite phases without near-neighbor C atoms.<sup>[16]</sup> The remaining magnetic interactions are associated with Fe atoms, with C atoms placed as first and second neighbors.<sup>[16]</sup> Due to the lack of resolution in the present velocity range, the austenite subspectra were simulated with only two interactions (a single line and a quadrupole doublet) instead of the three interactions usually associated with the different Fe-C configurations in austenite.<sup>[17]</sup> This approximation does not affect the areas associated with austenite and ferrite/martensite phases.

To monitor the ferrite-martensite balance, the relative fractions  $(F^{M}_{i})$  of the three magnetic signals used in the fittings were normalized and are shown in Figure 2 for the different temperatures and austempering times. The constancy observed in the  $F^{M}_{i}$  relative fractions, within error, suggests that the sample microstructures do not exhibit any temperature dependence.

Over the temperature range studied, the austenite relative fraction  $(f_{\gamma})$  follows a sigmoidal-type behavior (Figure 3),



Fig. 2—Evolution of the relative intensities of the magnetic subspectra (classified according to the model of Ref. 17) for different austempering times. At each time, the magnetic fractions shown correspond to alloys austempered at 673, 648, 623, 598, and 573 K. The lines are guides for the eyes.  $\blacklozenge - F^{M}_{1}, \bigcirc \cdots F^{M}_{2}$ , and  $\blacktriangle - F^{M}_{3}$ .



Fig. 3—Evolution of the  $f_{\gamma}$  austenite relative fraction with the austempering time. The lines are guides for the eyes:  $\blacksquare$  — 673 K,  $\bigcirc$  … 648 K,  $\triangle$  ---- 623 K,  $\blacklozenge$  --- 593 K, and  $\nabla$  --- 573 K.

typical of a nucleation and growth transformation.<sup>[6]</sup> That kind of process can be quantified using the Johnson–Mehl's equation<sup>[5,6,14]</sup>  $X(t) = 1 - \exp(-kt)^n$ , from whose linear form,

 $\log \log (1 - X(t))^{-1} = (n \log k + \log \log e) + n \log t$ 

the kinetics parameters k and n can be determined. In this equation, the transformed fraction (X(t)) is defined as

$$X(t) = (f_{\gamma}(t) - f_{\gamma}(0))/(f_{\gamma}(f) - f_{\gamma}(0))$$

where  $f_{\gamma}(0)$  is the austenite relative fraction at time zero, while  $f_{\gamma}(t)$  and  $f_{\gamma}(f)$  are the relative fractions of austenite at time *t* and after completion of the transformation, respectively. The *n* parameter determines the type of process that governs the transformation, and *k* involves the nucleation and growth rates.<sup>[6]</sup>



Fig. 4—Linear form of the Johnson–Mehl equation log log  $(1 - X(t))^{-1}$  against the logarithm of the austempering time. The lines are the linear fits for each temperature:  $\blacksquare - 673 \text{ K}$ ,  $\bigcirc \cdots 648 \text{ K}$ ,  $\triangle \cdots 623 \text{ K}$ ,  $\blacklozenge \cdots 593 \text{ K}$ , and  $\nabla \cdots 573 \text{ K}$ .

Table I. Values of k and n Obtained by the Johnson–Meh									
Equation for All Temperatures (Errors in <i>n</i> Are Quoted									
as Subindex)									

Temperature	п	$k (s^{-1})$
673 K	1.62	$2.9 \cdot 10^{-3}$
648 K	1.82	$3.2 \cdot 10^{-3}$
623 K	$2.2_{2}$	$3.9 \cdot 10^{-3}$
598 K	1.5	$3.3 \cdot 10^{-3}$
573 K	1.63	$2.0 \cdot 10^{-3}$

The resulting values of n and k, obtained from Figure 4, are reported in Table I. The n values, close to 1.7, are representative of a diffusion-controlled transformation.<sup>[6]</sup> The variation of the parameter k with temperature is illustrated in Figure 5, which displays a log plot of k as a function of the inverse of temperature. The observed maximum at 623 K indicates a non-Arrhenius-type dependence.

Another way that can also disclose the kinetics of stage I of the austempering transformation is the measurement of the decrease in hardness as a consequence of the reduction of the martensite content with the austempering time. The method is based on the determination of the time at which the hardness is 100 Vickers units higher than the plateau value, when no further transformation occurs. At the time when such a hardness is attained, it is estimated that 60 to 80 pct of the stage I transformation has been completed.<sup>[4]</sup> The CG cast-iron samples analyzed in the present article yielded the measurements illustrated in Figure 6(a), in which the experimental errors were obtained as the standard deviation corresponding to ten measurements for each value. Considering that the plateau value is reached at  $\approx$ 30 minutes, the time at which the hardness is 100 Vickers units above



Fig. 5—Logarithm of the kinetic parameter k vs 1/T.  $\bigcirc$  Values obtained for CG samples of the present work.  $\bullet$  Values obtained by Liu *et al.* for ADI samples.<sup>[14]</sup>

the plateau value was calculated, and the results are shown in Figure 6(b). The shortest time in the figure is seen at  $\approx 623$  K. This value is in agreement with the temperature of the maximum value of the rate constant k, suggesting a faster transformation rate at this temperature. To corroborate this observation, Figure 7 shows the microstructures after 10 minutes of treatment for each temperature. It can be observed that the ferrite needles characteristic of the ausferrite structure are more noticeable at 623 K (Figure 7(c)), indicating that the advance of the transformation is faster at this temperature. This observation supports the results obtained by Mössbauer spectroscopy and hardness methods.

The presence of a maximum in the transformation rate needs an explanation based on the thermal dependence of the nucleation and growth rates. The temperature dependence of the nucleation process suggests that, as the undercooling  $(\Delta T)$  increases, the volume free-energy change increases, promoting a higher nucleation rate.<sup>[8]</sup> Concurrently, according to References 9, 10, and 18, the growth rate is proportional to the diffusion coefficient of C in austenite. Hence, at high transformation temperatures, the growth rate is high and the nucleation rate is slow; consequently, this process mainly controls the transformation rate. The contrary situation occurs at low temperatures, and the reaction becomes mainly dependent on the diffusion. The balance between both trends implies that the transformation-rate parameter k should display a maximum at some temperature (such as 623 K, as shown in Figure 4). This behavior closely resembles the characteristic "C" curve of the time-temperature-transformation (TTT) diagram.<sup>[19]</sup>

The present results, which indicate that the variation of k with temperature does not follow the Arrhenius-type dependence, are different from those reported by Liu *et al.*<sup>[14]</sup> in their work on an austempered ductile iron (ADI) of similar



Fig. 6—(*a*) Hardness evolution with the austempering time for each temperature. The lines are guides for the eyes:  $\blacksquare - 673 \text{ K}$ ,  $\bigcirc \cdots 648 \text{ K}$ ,  $\triangle \cdots - 623 \text{ K}$ ,  $\blacklozenge \cdots 593 \text{ K}$ , and  $\bigtriangledown - \cdots 573 \text{ K}$ . (*b*) Time to achieve 60 to 80 pct of the transformation as a function of the austempering temperature. The line is a guide for the eyes.

compositions. The ADI exhibits a linear dependence of k with the inverse of temperature in the narrower range of 623 to 693 K. This different behavior is also shown in Figure 5, where the k values for the austempered CG cast iron are higher than for the ADI at all temperatures, indicating that the kinetics of the austempering transformation is faster for the CG cast iron. The k values of the ADI samples<sup>[14]</sup> are 4 times smaller than the values determined in the present work for the CG cast-iron samples.



Fig. 7—Austempering microstructures obtained at (a) 573 K, (b) 598 K, (c) 623 K, (d) 648 K, and (e) 673 K, after 10 min of treatment.

Bayati *et al.*<sup>[4]</sup> analyzed the kinetics in a CG cast iron with, in wt pct, 3.5 C, 2.3 Si, 0.02 Mn, and 1 Cu and found longer austempering times at  $\approx$ 623 K. The different results reported in this work could be related to the different chemical compositions and, therefore, to the different C curves of both CG cast irons analyzed.

To follow the carbon concentration evolution  $(C\gamma)$  with temperature and austempering time in the austenite phase, a series of Mössbauer spectra were taken in the velocity range from -2 to +2 mm/s, where the austenite pattern is observed in detail. Typical spectra are shown in Figure 8 for the different austempering temperatures and times indicated in the figure. Genin's model<sup>[17]</sup>—assuming a Fe<sub>8</sub>C<sub>1-y</sub> (0 < y < 1) solid solution and a repulsive interaction between

carbon atoms—was utilized to describe the different Fe-C configurations in the austenite lattice. Following the notation  $F_{\Gamma ij}$  for the normalized fraction of austenite, with *i* and *j* being the first and second C neighbors, respectively, the austenite subspectra were reproduced with three hyperfine interactions associated with (1) iron atoms without near-neighbor or next-near-neighbor C atoms ( $F_{\Gamma 00}$ ), (2) iron atoms with only one near-neighbor C atoms ( $F_{\Gamma 10}$ ), and (3) iron atoms without near-neighbor C atoms but with *n* second-neighbor C atoms (with *n* being a value between 1 and 4) ( $F_{\Gamma 0n}$ ). The results of the fitting procedure are reported in Table II. A simple analysis of the evolution of the normalized relative fractions  $F_{\Gamma ij}$ , reported in Table II, indicates that the C concentration increases with austempering time, since



Fig. 8—Mössbauer austenite pattern obtained using a low-velocity range for alloys austempered at 573 K and alloys austempered at 673 K at the austempering times shown.



Fig. 9—Austenite carbon concentration as a function of the austempering time and temperature assuming a random distribution of C atoms in austenite and a repulsive interaction between C atoms. The lines are guides for the eyes:  $\blacksquare - 673 \text{ K}$ ,  $\bigcirc \cdots 648 \text{ K}$ ,  $\triangle - \cdots - 623 \text{ K}$ ,  $\blacklozenge - \cdots 593 \text{ K}$ , and  $\bigtriangledown - \cdots 573 \text{ K}$ .

the intensity of the  $\Gamma_{00}$  singlet decreases. From the  $F_{\Gamma i j}$  normalized relative fraction and using the usual assumption that the Mössbauer–Lamb factors are the same for all the sites in austenite, the atomic carbon concentration of this phase was determined using the occupation probabilities of Reference 17. The carbon concentrations for the different



Fig. 10—log-log plot of the fraction of carbon atoms  $f_{\gamma}C_{\gamma}$  incorporated in austenite versus austempering time, for differents austempering temperatures. The lines are linear fits for the data.  $\blacksquare$  — 673 K,  $\bigcirc$  … 648 K,  $\triangle$  ---- 623 K,  $\blacklozenge$  ---- 593 K, and  $\bigtriangledown$  ---- 573 K.

temperatures and austempering times are displayed in Figure 9. An increase in the C concentration from  $\approx 1.1$  to  $\approx 1.7$  wt pct from 1 to 10 minutes of austempering time, respectively, is noticed, evidencing the C enrichment of the austenite phase.

In a rough approximation, the C amount  $(f_{\gamma}C_{\gamma})^{[1,2,3]}$  incorporated into the austenite should be directly related to the

Time	δ	Δ	Г	$F_{\Gamma 10}$	δ	Г	$F_{\Gamma 0n}$	$\delta^*$	Г	$F_{\Gamma 00}$	$f_{\gamma}$	
(Min)	(mm/s)	(mm/s)	(mm/s)	(Pct)	(mm/s)	(mm/s)	(Pct)	(mm/s)	(mm/s)	(Pct)	(Pct)	
					673	K						
1	$0.00_{1}$	0.672	0.393	33 <sub>3</sub>	$0.06_{1}$	0.397	$22_1$	-0.1	0.362	45 <sub>2</sub>	$20_1$	
4	0.011	0.661	0.382	434	0.061	0.456	25 <sub>2</sub>	-0.1	0.362	322	$28_1$	
5	$0.01_{1}$	$0.64_{1}$	0.461	52 <sub>3</sub>	$0.06_{1}$	0.413	$21_{1}$	-0.1	0.372	$27_{1}$	$29_1$	
10	0.011	$0.67_{1}$	0.361	47 <sub>3</sub>	$0.08_{1}$	0.434	241	-0.1	0.352	29 <sub>2</sub>	321	
20	0.011	$0.68_{1}$	0.341	45 <sub>3</sub>	$0.08_{1}$	$0.44_{4}$	272	-0.1	0.352	$28_{2}$	351	
30	0.011	0.671	0.362	$44_{4}$	0.071	0.485	27 <sub>2</sub>	-0.1	0.362	29 <sub>2</sub>	35 <sub>1</sub>	
	648 K											
1	$-0.01_{1}$	$0.66_{1}$	$0.40_{2}$	39 <sub>2</sub>	$0.07_{1}$	0.302	$17_{1}$	-0.1	0.341	441	171	
3	$0.00_{1}$	0.651	$0.40_{1}^{-}$	$47_{2}^{-}$	0.061	$0.33_{3}$	191	-0.1	0.31	341	231	
4	0.011	0.651	$0.40_{2}$	482	$0.07_{1}$	0.363	$20_{1}$	-0.1	0.321	$32_1$	$25_{2}$	
5	0.01	0.67	$0.36_{1}$	$46_{2}^{-}$	0.071	0.363	231	-0.1	0.331	331	$26_{1}^{-}$	
10	$0.01_{1}$	0.671	0.362	513	$0.07_{1}$	$0.40_{4}$	$22_1$	-0.1	0.322	272	301	
20	0.011	0.661	0.381	542	$0.07_{1}$	0.433	$23_1$	-0.1	0.341	$23_1$	321	
30	0.011	0.671	0.351	462	0.071	$0.46_4$	$20_{1}$	-0.1	0.352	$27_{1}$	321	
					623	К						
0	$-0.05_{1}$	0.753	0.466	363	0.071	0.26*	$11_{1}$	-0.1	0.362	53 <sub>2</sub>	141	
1	$-0.03_{1}^{1}$	0.65	$0.40_{1}^{\circ}$	391	$0.04_{1}$	$0.26_{2}$	$17_{1}^{1}$	-0.1	0.331	441	181	
3	$0.00_{1}$	0.661	0.382	442	0.061	0.324	181	-0.1	0.31	381	232	
4	$0.00_{1}^{1}$	$0.67_{1}$	$0.35_{1}^{-}$	$44_{1}^{-}$	0.06	0.37*	$21_{1}$	-0.1	0.33	351	$24_{1}^{-}$	
5	0.001	0.67	0.34	441	0.061	0.393	231	-0.1	0.33	331	261	
10	$0.01_{1}$	$0.67_{1}$	0.351	$48_{1}$	$0.05_{1}$	0.382	$22_{1}$	-0.1	0.321	$29_1$	29 <sub>1</sub>	
20	$0.00_{1}$	0.671	0.391	58 <sub>1</sub>	$0.07_{1}$	0.332	19 <sub>1</sub>	-0.1	0.331	$23_1$	29 <sub>1</sub>	
30	0.011	0.661	0.381	612	0.061	0.37*	19 <sub>1</sub>	-0.1	0.311	$20_{1}$	282	
					598	К						
1	-0.01*	0.664	0.416	244	0.072	0.449	$23_{1}$	-0.1	0.423	533	161	
4	$-0.00_{1}$	0.682	0.353	303	0.05	0.489	$26_{2}$	-0.1	0.382	442	214	
5	$-0.00_{1}^{1}$	$0.67^{\tilde{*}}$	0.412	362	$0.07_{2}^{1}$	0.507	$25_{2}^{2}$	-0.1	$0.41_{2}^{-1}$	392	$22_{1}^{-}$	
10	0.01	0.69*	0.351	382	$0.07_{1}^{-}$	0.526	$29_{2}^{-}$	-0.1	0.361	$33_{2}^{-}$	262	
30	0.01	0.69*	0.442	$45_{3}^{-}$	0.101	0.442	$23_{2}^{-}$	-0.1	0.443	$32_{2}^{-}$	$26_{1}^{-}$	
					573	К						
1	$-0.01_{1}$	0.594	0.474	423	0.061	0.295	$17_{1}$	-0.1	0.363	412	161	
2	$-0.02_{1}^{1}$	$0.71_{3}$	0.347	244	$0.06_{3}$	0.479	$24_{4}$	-0.1	0.394	$52_{4}^{2}$	161	
4	$-0.01_{1}$	0.64	0.412	$42_{2}^{\cdot}$	0.061	0.323	$20_{1}$	-0.1	0.32	381	191	
10	0.01	$0.62_{2}$	$0.49_{2}^{-}$	$52_{4}^{-}$	0.051	0.406	$19_{2}^{\cdot}$	-0.1	$0.39_{3}^{-1}$	$29_{2}^{\cdot}$	231	
20	0.01	$0.66_{1}^{-}$	$0.38_{1}^{-}$	502	0.061	0.393	$23_{1}^{-}$	-0.1	0.32	$27_{1}^{-}$	$27_{1}^{1}$	
30	0.01	0.661	0.371	52 <sub>2</sub>	0.061	0.393	23 <sub>1</sub>	-0.1	0.321	25 <sub>1</sub>	251	

 

 Table II.
 Hyperfine Parameters and Normalized Fractions Associated with Austenite and Corresponding to Austempering Temperatures: 673, 648, 623, 598, and 573 K\*x

<sup>x</sup>The term  $\delta$  is the isomer shift,  $\Delta$  is quadrupole splitting,  $\Gamma$  is the line width,  $F_{\Gamma ij}$  are the normalized fractions of the austenite phase for different configurations of Fe, and  $F_{\gamma}$  is the total fraction of austenite. Errors are quoted as subindex. \*Parameter held fixed while fitting.

diffusion length<sup>[13]</sup> and consequently, be proportional to  $t^{0.5}$  if the carbon enrichment of the austenite phase is controlled by diffusion. From the results of  $f_{\gamma}C_{\gamma}$ , the dependence on austempering time is displayed in Figure 10, where a plot of ln  $(f_{\gamma}C\gamma)$  vs ln (t) is presented. The dependence found was  $t^{0.40\pm0.05}$  in the temperature range studied. The exponent, almost equal to the ideal one, indicates that the process is governed by C diffusion. The slight difference might be explained if the dependence of the diffusion coefficient on the carbon concentration<sup>[8]</sup> (that in the present analysis was considered to be constant) were taken into account.

## **IV. CONCLUSIONS**

The kinetics of austempering transformation in a CG cast iron of composition 3.4 C, 2.8 Si and 0.8 Mn (in wt pct), in the range from 573 to 673 K, has been studied. The results, analyzed in the framework of the Johnson–Mehl's equation, indicate that nucleation and growth processes govern the transformation. The evolution of the rate constant k with temperature displays a maximum at  $\approx 623$  K. This result is also conveyed by the measurement of the time needed to reach a hardness of 100 Vickers units above the plateau value, which exhibits a shortest time at  $\approx 623$  K. The current results suggest that, at low temperatures, the transformation is controlled by the growth process, while at higher temperatures, nucleation becomes the controlling process. This evolution is in close relation to the characteristic "C" curve of the TTT diagrams.

The analysis of the C concentration in austenite confirms the C enrichment with austempering time, but the C content  $(f_{\gamma}C_{\gamma})$  does not depend strongly on the austempering temperature. The results suggest that this process is controlled by the diffusion of C atoms from the ferrite/austenite interface into austenite, as shown by the  $t^{0.40\pm0.05}$  dependence of the C content.

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