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On the $\alpha \rightarrow \gamma$ Transformation During Dynamic Thermal Cycles

W.J. Kaluba, R. Taillard*, K. Jozwiak** and J. Foct*

Université du Littoral, C.C.G.T.E., rue Descartes, BP. 99, 62968 Longuenesse, France

** Laboratoire de Métallurgie Physique, URA 234 du CNRS, Bâtiment C6, Université de Lille I, 59655 Villeneuve d'Ascq, France*

*** Laboratoire de Métallurgie Physique, Université de Poznan, Pologne*

Abstract. Rapid cycles of heat treatment were given to various Fe-C alloys with different initial morphology and carbon content between 0.03 and 0.98% C. The specimens were heated up to temperatures between 600 and 1000°C by passing a high-amperage ($10^3 - 10^4$ A) alternating current through them. The microstructural investigation was performed by light and electron microscopy as well as electron microprobe measurements. The results of the study are discussed in terms of the mechanisms of phase transformations, taking account of various parameters like carbon content, primary microstructure, temperature and heating rate.

1. INTRODUCTION

It is well known that the $\alpha \rightleftharpoons \gamma$ transformation by rapid thermal cycles results in the refinement of grains and, consequently in the improvement of mechanical behaviour of steels [1-3].

Attempts have been made to explain different features of the transformation of ferrite to austenite considering the effect of heating rate and structural parameters. Also, new non-conventional methods of heating like laser technique was developed making possible a remarkable increase of heating rate [4,5].

Two major mechanisms of ferrite to austenite transition have been proposed:

- a diffusional one, related to the process of nucleation and growth of the austenitic phase with an overall kinetics controlled by carbon diffusion,
- a diffusionless one (without long-range diffusion) including both massive and shear models.

According to the experimental data the most important factors promoting one individual mechanism to be operative are heating rate, morphology and chemical composition of the material.

It has been proved that a moderate increase of heating rate results in distinct and continuous rise of the A1 and A3 temperatures. Yet, this upward evolution is no longer observed when the rate of heating exceeds some limit: the position of Ac1 and Ac3 appears practically independent on heating rate, which is usually attributed to diffusionless transformation mechanisms [6,7,8].

The presented behaviour including the elevation of Ac1 and Ac3 as well as the particular value of critical rate of heating may be strongly modified by the prior microstructure of steel and, especially by the distribution and morphology of carbides. In addition, it is also affected by the initial deformation of the material, so it would be useful to bring to light all the phenomena related to dislocation structure and their mobility. This statement is of special interest considering the inevitable generation of thermal stresses and strains during a rapid heat cycle.

In spite of the number of papers dealing with the subject it is still difficult to propose a generalized relationship between heating rate and the evolution of Ac1 and Ac3 temperatures. This is however not surprising in view of the complexity of the phenomena, the number of heating methods applied and the variety of alloys able to be treated. Table 1 gives some illustration of this statement. Furthermore, metallographic observations at room temperature were generally insufficient to provide direct proofs of the diffusionless transformation, partially because the analysis of any final microstructure gives not always a clear image of all the events during rapid heating.

This paper deals with structural evolutions occurring in the range of ferrite to austenite transformation in some Fe-C alloys during rapid heat cycles. The initial structure of the material was selected so as to

investigate the effect of three morphological modifications : ferrite, ferrite + globular cementite and martensite.

Table 1. Values of critical heating rate corresponding to $\alpha \rightarrow \gamma$ diffusionless mechanism of the transformation according to different authors.

Heating rate ($^{\circ}\text{C/s}$)	Heating technique	Author
200 - 300	Welding	E. Rasanen et al.[3]
10^3 - 10^5	Laser	I.V.Liasotskii et al.[4]
400 - 500	Direct conduction of electrical current	K.J.Albutt et al. [6]
10^7	Induction, laser	J. Orlich [5]

2. EXPERIMENTAL.

Four industrial Fe- C alloys were employed. They contained 0.03, 0.10, 0.38 and 0.98 %C respectively. The steels were submitted to preliminary treatment so as to obtain different morphologies before rapid heating:

- 0.10 and 0.98 %C steels were annealed to produce coarse-spheroidized form of cementite. The treatment consisted: heating up to temperature of 750°C in furnace, cooling down to 650°C , soaking during 15 h and air cooling.

- 0.38 %C steel was used in martensiting state. The samples were water quenched after 15 minutes of austenitisation at 860°C .

- 0.03 %C ARMCO iron was used in "as received" ferritic state.

Two different heat cycling were employed in this work:

- direct electrical conduction which ensured heating rates of 10 - 300°C/s .

The samples were heated by passing the electrical current through them in a special device described in details earlier [9]. Water cooling was executed immediately when the nominal temperature of cycle was reached. Air cooling was carried out for better understanding of structural evolutions. The corresponding cooling rates inside of the samples were about 15°C/s and 1°C/s .

- common convectional heating in an electrical resistance furnace, before water quenching or air cooling

Microstructural investigation was conducted by light (LM), scanning (SM) and thin foil electron microscopy. Electron microprobe analysis was applied to some qualitative measurements of carbon concentration. Hardness tests were performed by the Vickers method.

3. RESULTS AND DISCUSSION.

3.1. Monophased structure.

3.1.1. 0.03%C Ferrite

In as-received condition, and as illustrated by the TEM micrograph in figure 1, the 0.03%C steel contains a significant density of dislocations that exhibit some tendency to form tangles in certain areas or more well defined walls in other zones. These combinations of dislocations form along (110) planes and their average spacing is close to $0.8\ \mu\text{m}$. The rapid heat treatment gives no rise of this spacing while the maximum temperature remains below 850°C . This distance increases markedly for higher nominal temperatures as the result of recrystallization phenomena inducing the annihilation of subgrain boundaries. The weakness of this annihilation on the one hand, and the evidence of strain induced grain boundary migration on the other (see Fig. 4b for inst.) prove that the specimens heated at 250°C/s up to 950°C are only partially recrystallized. Yet, whichever was the heating rate, a significant effect of

recovery was visible at nominal temperatures as low as 500°C. The presence of this phenomenon is proved by numerous observations showing an increase of sharpness of dislocation tangles with temperature. Figure 2 proves that this behaviour originates from dislocation rearrangements inside the primary tangles.

Figure 1: Dislocation substructure of 0.03%C steel in the as-received state.

Figure 2: TEM evidence of recovery in the specimen heated at 10°C/s up to 700°C

Figure3: General views of samples heated at 250°C up to 700°C.

The optical micrograph in figure 3 shows that concomitant increment of desorientation across a wall gives rise to parallel contrasts in the interior of grains. As it is visible in figure 3b, a significant grain boundary serration may also be noted. In this particular case, it seems to be recrystallisation influence rather than shear mechanism effect, especially because the nominal temperature of the cycle is below the A_{c1} value. The micrograph in figure 4a shows an irregular grain boundary which was frequently observed for nominal temperatures of heating higher than A_{c3} . Such a morphology is sometimes related to the massive transformation [10]. However, this kind of morphology should not be taken as the proof, because of the close similarity between the mechanisms of the massive and recrystallisation phenomena.

Figure 4: Irregular morphology of grain boundaries in the samples heated up to 950°C at : a) 10°C/s, b) 250°C/s.

In this study, the numerous observations of not recrystallized grains (Fig.4b) show relatively rapid displacement of grain boundaries towards the regions of high dislocation density. This means, the irregular form of grain boundaries arise from recrystallization rather than from the transformation to austenite by massive reaction. In addition, it is important to note that this conclusion is also confirmed by scarce observations of in some specimens heated under Ac1.

This discussion easily demonstrates the difficulty to identify a mechanism of austenitization when the recrystallization takes place simultaneously with the phase transformation. The systematic investigation of the temperature range of austenitization by means of continuous series of heat cycles with various origin microstructures seem to be necessary in order to avoid misleading and premature conclusions.

3.1.2. 0.38%C Martensite

The evolution of initially martensitic structure during heat cycles is shown in figures 5, 6 and 7. The rate of heating was 250°C/s and maximal temperature of the cycles were 760, 830 and 900°C, respectively. A distinct variation of morphology appears when heating to 760 °C (Fig.5b). The typically martensitic structure (Fig.5a) has lost a little of its needle-shaped feature, although the initial orientation was still visible. Moreover, a strong grain refinement is observed. The SEM technique reveals two distinct zones corresponding to different stages of transformation :

Dark irregular areas (fig.5c) represent the product of martensite to austenite reaction at its early stage. The acicular zone corresponds to this volume of the material that was not affected by austenitization. The TEM image (fig.5d) of this zone gives an impression of ferritic morphology rather than of a martensitic one because of low density of dislocations, although its hardness of 480HV is relatively high. It is surprising enough, regarding that no carbide precipitations were detected in this zone. However, the carbides are observed after slow air-cooling of the samples together with numerous small pearlitic areas (Fig.7).

In the light of these facts two conclusions are to be made. First, at the given heating conditions the nominal (760°C) temperature of treatment exceeds the actual dynamic value of Ac1.

Second, that in spite of the relatively high temperature, the " ferritic " zones remain still supersaturated with carbon. In addition, the electron microprobe analysis has pointed out an essential difference of carbon concentration in both zones. The measured ratio of carbon concentration was about 1.3 in favour of as transformed areas. This means, that ferrite to austenite transformation starts in some areas of the matrix by diffusion bringing about an increase of local carbon concentration. The other structural variant (acicular zone) is formed by recrystallization of martensite that takes place at this temperature

level. Some preliminary tests have pointed out that this original morphology of "recrystallized martensite" shows interesting mechanical properties, like a good compromise between strenght and toughness.

Figure 5: Evolution of initially martensitic (0.38%C) structure after heating to 760°C at the rate of 250°C/s and water quenching: a,b - LM; c,d - TEM

The analysis of the micrographs in figure 6 gives some additional information on the discussed topic. The microstructures correspond to the thermal cycles with nominal temperatures of 830 and 900°C, respectively. It is rather clear that in both cases a "secondary" martensitic transformation has taken place during water quenching, thus complete austenitization of the matrix is evident. Nevertheless, the localized zones of twinned martensite suggest a significant heterogeneity of carbon concentration. Otherwise, it is our strong impression that there is a close similarity between the morphologies in figures 5b and 5d on the one hand and figures 6a and 6b on the other. These micrographs show the same shape and size of crystallites, but with very different dislocation density. One possible reason of such structural

behaviour could be the diffusionless transformation of recovered martensite into austenite over the range of 760 - 830°C. However, this supposition needs some additional proofs.

Figure 6 : Evolution of initially martensitic (0.38%C) structure after heating to 760°C at the rate of 250°C/s and air cooling.

3.2. Two-phase structure 0.98 %C : Ferrite + globular cementite

The results of the metallographic examination of the 0.98%C steel with the initial structure of spheroidized cementite are shown in figure 8. The heating rate was 10°C/s and the nominal temperature of the cycle was 800°C. The employed heating technique enabled to detect the structural evolution at different temperature levels and for the same time of effective austenitization. By this way, transformation progress was easily followed. Moreover, One should notice that practically the same sequence of morphological behaviour with the rate of 250°C/s.

Figure 7: Evolution of the initially martensitic (0.38%C) structure after heating with the rate of 250°C/s and water quenching. a,b - 830°C ; c,d - 900°C ; a,c - LM ; b,d - TEM

An early stage of the transformation is presented in figure 8b. White irregular regions of size a little larger than that of initial particles of cementite (Fig.8a) represent the as-transformed austenite here. This is hardly perceptible by light microscopy which gives an impression of well spheroidized cementite rather than of the transformation product. Yet, the TEM image of such a zone (Fig. 9) points out that diffusional transformation has started at the cementite / ferrite interface producing austenite shells around the cementite particles. The concentration of carbon in these shells seems to be relatively high since they were not transformed to martensite during water quenching. According to theoretical and experimental works on cementite dissolution during austenitization [10,11] the subsequent diffusion-controlled growth occurs by migration of the α / γ interface followed by

Figure 8: Evolution of initially double phased structure of ferrite+cementite after heating to 800°C at 10°C/s: a,b,c,d,e - LM; f - TEM.

Figure 9: An early stage of the formation of austenite by diffusional mechanism. TEM.

partial or total cementite dissolution in ferrite and austenite. Figures 8 c,d,e are a good illustration of this transformation model. In addition, strong etching effect in the neighbourhood of interfaces (dark regions around the as-transformed zones) points out the probability of some rise of carbon concentration. This is particularly evident at advanced stages of the process (Fig.8d). One of the possible reasons is slower dissolution of cementite compared with the interface mobility. When the interface moves, the great number of cementite particles can be directly incorporated into growing austenite without being dissolved. This trend is keeping on until the transformation is completed (Fig.8e,f) giving the heterogeneous product of the reaction with irregular carbide size and distribution. The final hardness of 950 HV is surprisingly high, because of the extremely fine structure of martensite that contains cementite precipitates.

4. CONCLUSIONS

1. In the conditions of this work two distinct mechanisms of austenitization are noted which depends on the initial microstructure. Only diffusion control transition occurs in the case of spheroidized steel. A combined mechanism of transformation is observed with an initial martensitic structure: at first, some localized areas transform by diffusion while the martensitic matrix is recovering; then shear is able to occur.

2. Rapid heating at rates up to 250°C/s does not induce any diffusionless transformation in 0.03%C steel. All microstructural features which were previously related to such phenomena are simply attributed to recrystallisation.

3. The practical advantage of rapid heat treatment is proved by the enhancement of mechanical properties: recovered martensite presents a good compromise of strength and toughness, while extremely high hardness is measured in high carbon steel.

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