JMEPEG (2005) 14:179-187
DOI: 10.1361/10599490523328

1059-9495/$19.00

Characteristics of Steels in Annealing Combined with Hot-Rolling Preheating

V.M. Khlestov, H.J. McQueen, G.A. Cingara, and E.V. Konopleva

(Submitted October 4, 2004)

The rates of transformation and the morphologies after full annealing and industrial rolling without reheating were determined to establish the degree of similarity. Steels with 0.1-1% C and almost nil to 17% alloying addition (M2 steel) were isothermally transformed in a dilatometer after various thermomechanical processing (TMP) sequences. For annealing, the heat treatment was performed at about 880 °C, depending on C content, and then cooled for transfer to the dilatometer. In some cases, specimens were rolled to 50% reduction in 3 passes before cooling. The rolling schedule was simulated by preheating to 1150 °C followed by slow cooling to 950 °C for finish rolling as above before transfer to the dilatometer. Some specimens were directly cooled to transformation. The high preheat thoroughly homogenized the austenite (except for the M2 steel) and slowed the transformation by a factor >10 as the alloying content rose. In all cases, the hot rolling accelerated the transformation. After the high preheat the rate and microstructure were restored to those of standard annealing for hypo-eutectoid and M2 steel.

Keywords annealing, hardness, microstructure, phase transformations, steels, thermomechanical processing, tool steels

1. Introduction

Combining hot rolling (or forging) and annealing (RA) processes for steel has obvious advantages over the same procedures conducted separately. Power consumption is considerably lower, and decarburization during annealing is excluded (Ref 1). For ferrous alloys, (full) annealing involves heating to above the transformation range, holding to homogenize, and cooling slowly to produce a soft microstructure with enhanced malleability and machineability. However, combined RA processing at the cooling stage can have a notably different effect from conventional annealing, since austenite in the former case may have an altered structure, and sometimes a different chemical composition, both resulting from high heating temperature and deformation. Thus, in M2 steel heated at the annealing temperature (880 °C), approximately 2% Cr and 0.3-0.5% W and Mo are dissolved in austenite. However, at the hot rolling temperature (1150 °C) almost all Cr and 1.5-2% W and Mo are dissolved (Ref 2). As a result, the stability of austenite increased by the factor of 10-12, and by even more in A2 steel similarly treated (Ref 3-5).

Such a large increase in austenite stability, due to high preheating, can render annealing combined with hot rolling impractical. However, an encouraging factor is that deformation significantly lowers the stability of austenite (Ref 6-12), and consequently, it can considerably decrease or even completely neutralize the negative effect of high temperature heating on annealing. Thus, in M2 steel, deformation at 950-900 °C after heating to 1150 °C decreased the stability of austenite by 5-6 times. Nevertheless, such decrease in the stability of austenite did not completely neutralize the effect of high temperature heating. The duration of complete decomposition of austenite after rolling was approximately twice as long as after heating for conventional annealing (Ref 3).

Generally, the times for pearlite formation from austenite under the types of treatment examined should be largely determined by the chemical composition of the steel. In plain carbon and low-alloy steels the stability of austenite after hot rolling evidently would not be higher than after conventional anneal heating because alloying of the γ-phase grows insignificantly under rising temperature. In medium and high-alloy steels, the stability might change considerably, and the change would not be favorable to combined RA processing. However, at present one cannot predict what the relation will be in a specific steel, and consequently, kinetics data of austenite to pearlite decomposition are needed when selecting a cooling schedule during annealing combined with hot rolling.

2. Experimental

Hypoeutectoid, eutectoid, and hypereutectoid steels with medium-stable alloy carbides (MSAC) and with ultra-stable alloy carbides (USAC, persisting above 1150 °C) were subjected to low- and high-austenitizing temperature (T), and to hot rolling, or not, before isothermal annealing. Chemical compositions are shown in Table 1. These steels were selected for study to reveal the peculiarities of annealing processes inherent to each group. The A2 is a cold work tool steel and M2 a high speed one. All steels were processed in five schedules as shown in Fig. 1.

The experiments were carried out on a laboratory mill equipped with two small tube furnaces located near the forming

V.M. Khlestov, Priazovsky State Technical University, Mariupol, Ukraine; **H.J. McQueen** and **E.V. Konopleva,** Mechanical and Industrial Engineering, Concordia University, 1455, de Maisonneuve Blvd. W., Montreal, QC, H3G 1M8, Canada; and **G.A. Cingara,** Materials Science and Engineering, University of Toronto, Toronto, ON, M5S 3E4, Canada. Contact e-mail: mcqueen@vax2.concordia.ca, ekonopl@videotron.ca.

Table 1 Chemical composition of examined steels

Group of steel	Grade	Content of elements, mass %						
		C	Si	Mn	Cr	W	Mo	Other elements
Hypoeutectoid	30CrMnSiNi2	0.28	0.70	1.08	1.13	…	…	1.93 Ni
	55CrSi2	0.56	1.76	0.70	0.92	…	…	…
Eutectoid	80 Mn	0.77	0.26	0.88	…	…	…	…
Hypereutectoid	70 Cr3	0.67	0.23	0.32	3.5	…	…	…
MSAC(a)	A2	1.0	0.30	0.70	5.15	…	1.15	0.31 V
USAC(b)	M2	0.84	0.25	0.30	4.0	6.5	5.0	1.9 V

(a) (MSAC) Medium Stable Alloy Carbide; (b) (USAC) Ultra Stable Alloy Carbide

rolls (Ref 10); this allowed rolling small specimens according to a selected schedule. Deformation of 5 mm thick specimens was conducted in 3 passes with reductions $\varepsilon_1 = \varepsilon_2 = \varepsilon_3 =$ 20% at $\dot{\varepsilon} = 17$ to 20 s^{-1}. Schedule 1 is typical of standard isothermal annealing. Schedule 2 (heating as in 1) was used for determining the effect of deformation on pearlitic transformation under isothermal annealing. Schedule 3 allowed determination of the effect of high-temperature heating on the kinetics of decomposition to pearlite and isothermal annealing. Schedule 4 was used to determine the combined effect of high-temperature heating and deformation with accelerated cooling to the transformation temperature (simulating annealing combined with hot rolling). Schedule 5 is similar to 4 but the specimens after deformation were cooled to 650-780 °C slowly (at 1-1.5 °C/s) before transformation.

The procedure of industrial annealing is determined mainly by the austenite stability. Due to this fact, the kinetics of pearlite transformation in the steel was studied at temperatures most advantageous for carrying out isothermal annealing. Heating temperatures and the deformation conditions for the steel are given in Fig. 2-5 and Table 2. The kinetics of pearlite transformation was studied on a specially designed dilatometer (Ref 13), which allowed quick installation of a specimen between the holders (within 2 s) and measuring austenite decomposition. To cool the deformed 2 mm thick specimens quickly to the isothermal temperature, they were placed for 5 s in a salt-bath furnace and then transferred to the furnace of the dilatometer, both at the same temperature.

Microstructures of some specimens were examined by optical microscopy after completing schedules 1-4 (Fig. 1), mechanical polishing, and etching in 4% HNO_3. Specimens of M2 steel were examined by transmission electron microscopy (TEM) after schedules 1, 3, and 4. Samples were mechanically ground to a thickness of about 100 μm and then electrochemically polished using a solution of 62 ml perchloric acid (70%), 700 ml ethanol, 100 ml butanol, and 137 ml distilled water. The electropolishing was conducted at a temperature between −30 and −38 °C at a voltage of 17 V and at a current of about 0.2 mA. TEM examination was carried out using a conventional HITACHI H-800-TEM/STEM operating at 200 kV.

3. Results and Discussion

As is shown in Fig. 2, the high-heating temperature of hypoeutectoid steels 30CrMnSiNi2 and 55CrSi2 leads only to a relatively small increase in the austenite stability (curves 1-A and 3). This is explained by the fact that the chemical composition of the γ-phase during high-*T* heating of hypoeutectoid steels practically does not change. Consequently, the retardation of ferrite-pearlite formation turns out to be insignificant and is predetermined by the austenite grain size increase. It is known that in steels, grain growth may slow down the austenite transformation to pearlite by a factor of 1.3-1.5, while the increase in alloying content in austenite makes it several or even tens of times more stable (Ref 14). On the contrary, the effect of deformation that introduces a dislocation substructure and grain boundary disturbances proves to be considerable (Ref 15, 16). As a result, the times of complete decomposition of austenite of both steels during RA processing are decreased by approximately half as compared with conventional annealing A (curves 4-RA and 1-A in Fig. 2). However, to achieve such a considerable decrease in the time of austenite decomposition, fast cooling right after rolling is required. If cooling of specimens immediately after rolling is retarded, the acceleration effect is considerably decreased (curve 5 in Fig. 2), which is caused by recrystallization of the deformed austenite (Ref 15, 16). Consequently, to apply the effect of accelerated austenite decomposition in RA processing, it is necessary to ensure quick cooling of products down to the transformation temperature immediately following their deformation.

The effect of overheating austenite and its hot rolling on the kinetics of pearlite transformation in eutectoid steel 80Mn is shown in Fig. 3. Comparison of the curves in Fig. 3 to those in Fig. 2 clearly shows that the effect of the various procedures on decomposition of austenite in eutectoid steel is practically the same as in hypoeutectoid steels. Such similarity is determined by the fact that austenite in eutectoid steel, when heated at high *T*, does not change its chemical composition, and upon retarded cooling after hot rolling, the recrystallization process occurs quickly.

In the alloyed hypereutectoid steels, other patterns from high temperature heating, either without or with hot rolling, on pearlitic decomposition of austenite become apparent. Figure 4(a) shows that raising the heating temperature of 70Cr3 steel from 820 to 1150 °C considerably slows down the pearlite reaction compared with hypoeutectoid and eutectoid steels. However, hot deformation of high *T* heated specimens accelerates this reaction in the same manner but only when specimens are cooled down quickly to the isothermal temperature after deformation. If after hot rolling the specimens (or products) are cooled comparatively slowly (1-1.5 °C/s), then recrystallization occurs, and as a consequence, the stability of the deformed austenite significantly increases (curves 4-RA and 5 in Fig. 4a). Hence, from the point of view of kinetics of austenite decomposition to pearlite, the isothermal holding time of 70Cr3 steel products during combined processing would be shorter than it is during conventional isothermal annealing, or

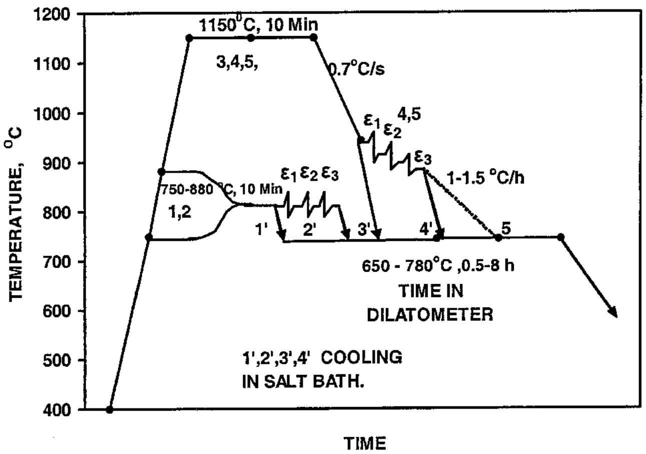

Fig. 1 Thermomechanical processing schedules of heating for annealing (1 and 2) or rolling simulation (3, 4, and 5) provides cooling rates, the presence or absence of rolling and conditions of transfer to the dilatometer. The temperatures vary with the steel, as given in Table 2.

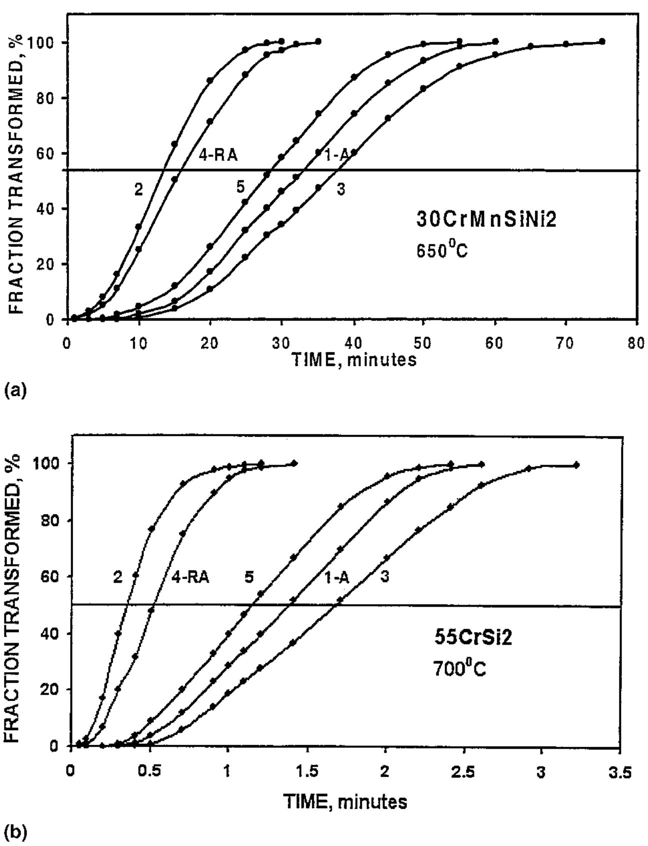

Fig. 2 Kinetics curves of austenite transformation in hypoeutectoid steels 30CrMnNiSi2 at 650 °C (a) and 55CrSi2 at 700 °C (b) after various heating and deformation schedules. (1-A) after heating at 850 °C; (2) after heating at 850 °C and rolling to 50% reduction at 850 °C; (3) after heating at 1150 °C; (4) RA, after heating at 1150 °C and rolling to 50% reduction at 950-900 °C; (5) same as 4, but slow cooling (1-1.5 °C/s) to isothermal temperature

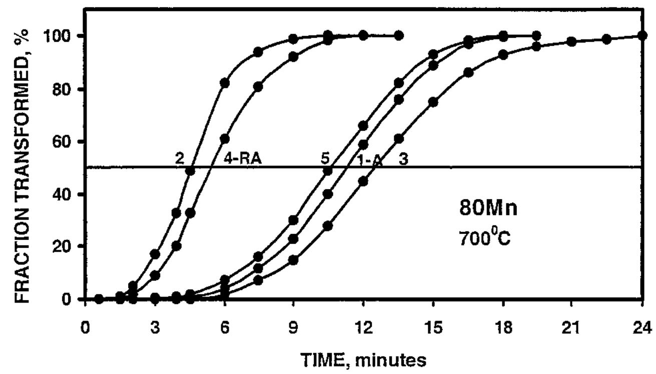

Fig. 3 Kinetics curves of austenite transformation in eutectoid steel 80Mn at 700 °C. (1-5) same processing schedules as in Fig. 1 and 2 with preheating in schedules 1 and 2 at 750 °C and rolling in schedule 4 at 800-750 °C

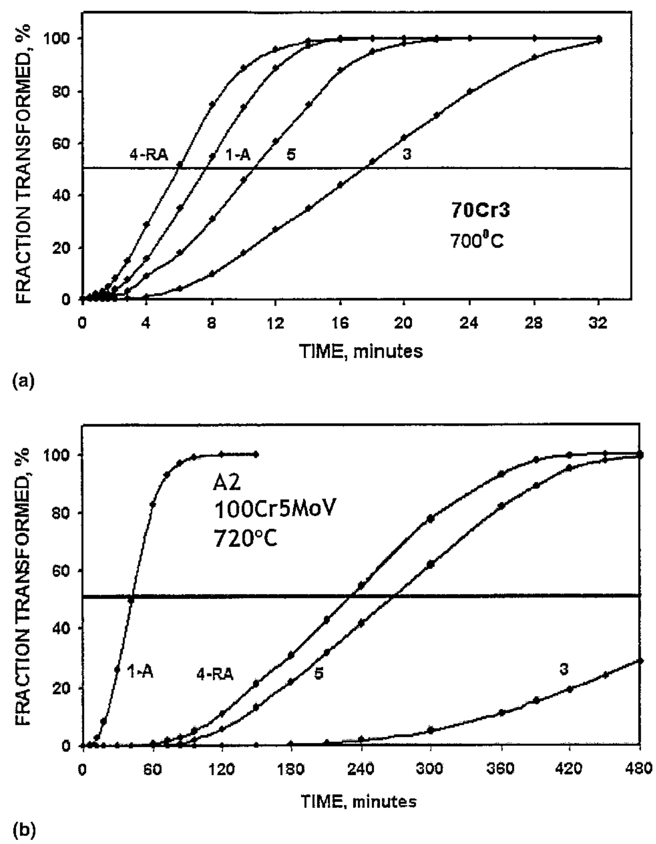

Fig. 4 Kinetics curves of austenite decomposition in hypereutectoid steels: (1-5) same processing schedules as in Fig. 1 and Fig. 2. (a) Steel 70Cr3: with preheating in schedules 1 and 2 at 820 °C and rolling in schedule 4 at 880-850 °C. Transformation at 700 °C. (b) A2 steel with preheating at 880 °C in schedules 1 and 2 and rolling in schedule 4 at 880-850 °C. Transformation at 720 °C

of the same duration if the products are cooled down quickly to isothermal temperature. This is difficult to achieve in production, especially when rolling large cross sections.

When alloying of hypereutectoid steels is increased (MSAC), the noted negative effect of RA processing intensifies (Fig. 4b). Indeed, in the MSAC steel A2 the rise of heating temperature from 880 to 1150 °C causes such an increase in the stability of austenite (approximately 40 times more) that plastic deformation, while considerably accelerating transformation, does not fully compensate for the effect of high-temperature heating.

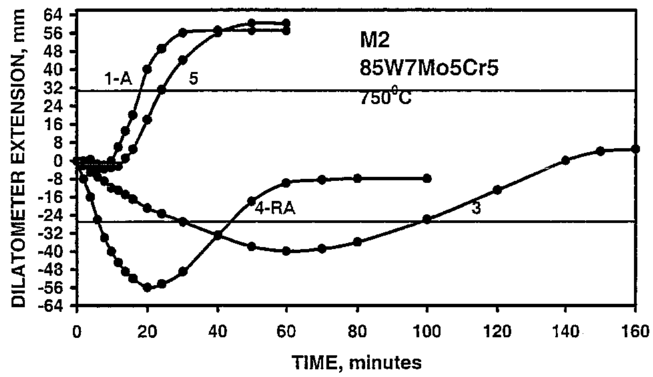

Fig. 5 Dilatometry curves of austenite decomposition in M2 steel at 740 °C. Processing schedules are the same as in Fig. 1 and 2 with preheating in schedules 1 and 2 at 880 °C and rolling in schedule 4 at 950-900 °C.

The above effects of high T heating and deformation on the pearlitic transformation in 70Cr3 and A2 steels are determined by the fact that carbides in both steels completely dissolve in austenite when heated at 1150 °C. Thus, the solute increases to the level of specified elemental content. 70Cr3 steel is considerably less alloyed than A2 steel, and that is why high T heating of the 70Cr3 slows down the pearlite reaction to a much lesser degree. In contrast, the effect of a postdeformation pause turns out to be more significant and is caused by a faster recrystallization process. Consequently, the duration of isothermal holding required for hot rolling combined with annealing is determined by the alloy level of the hypereutectoid steel to an even greater extent than during conventional annealing. In low-alloyed steels, the time can be approximately the same as that during conventional annealing, i.e., similar to eutectoid steels. However, when the content of alloying components increases, the holding time required for the completion of decomposition to pearlite will quickly increase, which might result in combined annealing being hardly feasible, and thus, ineffective.

The kinetics of pearlite transformation in M2 steel (USAC) was examined quite thoroughly in a previous study (Ref 3). The present research concentrates on the decomposition of austenite in this steel at 740 °C to compare the results with those obtained on steels in other structural groups. Dilatometric curves of austenite decomposition in M2 steel are shown in Fig. 5. The first behavior distinguishing austenite decomposition in USAC steel from the same process in hypereutectoid steels is the precipitation preceding pearlite decomposition of a large amount of secondary carbides from the γ-phase heated to 1150 °C. This is verified by the considerable contraction of specimens (curves 3 and 4 in Fig. 5) prior to their expansion that accompanies pearlite formation from austenite. Furthermore, hot deformation greatly accelerates carbide formation and considerably increases their amount, since contraction of deformed specimens is notably more pronounced than that of undeformed ones.

Another distinction is the fact that raising the heating temperature from 880 to 1150 °C leads to a much greater stability of austenite in hypereutectoid MSAC steel A2 than in USAC steel M2, although the content of alloying elements in the latter is twice as high. This is probably due to the peculiarities of carbide solubility in the two steels during their heating to 1150 °C. In M2 steel, only part of the carbides (approximately 7% of alloying elements) dissolve in the austenite (Ref 2). In MSAC hypereutectoid steel, carbides dissolve completely so that the content of alloy solute in A2 austenite is the same as the alloy composition. Carbides were not found during microstructural analysis of specimens quenched after high-*T* heating. During preheating for hot rolling, the microstructure that formed in the A2 steel is austenite alone, whereas in the M2 steel, a considerable amount of carbides remains and the austenite has very fine (ASTM No. 13) grains (Ref 2). Both factors trigger decomposition to pearlite (Ref 17), and most probably determine that the raised heating temperature enhances the stability of austenite in USAC steel (Fig. 5) to a much lesser degree that in hypereutectoid MSAC steel (Fig. 4b). This conclusion is confirmed by the fact that for A2 steel, a rise in heating from 880 to 980 °C did not lead to the complete dissolution of carbides and only weakly increased the stability of austenite (Ref 3).

One more peculiarity of austenite decomposition in hypereutectoid USAC steel is linked to the unusual effect of a postdeformation pause (i.e., retarded cooling after hot deformation). Figure 5 shows that retarded cooling of specimens after rolling does not decrease the destabilizing effect of deformation, as was observed in all other steels (Ref 10, 16). On the contrary, retarded cooling adds to the destabilization of austenite. This unusual phenomenon that helps accelerate austenite decomposition after 1150 °C preheating and rolling has not been previously examined. The main factor responsible for additional increase in the destabilizing effect of deformation may be the intensive precipitation of secondary carbides from deformed austenite during decelerated cooling of rolled M2 steel specimens to isothermal temperature. The comparison of curves 4-RA and 5 in Fig. 5 points to it. If the specimens are cooled in a salt bath immediately after rolling, the precipitation of large amounts of carbide from austenite (i.e., contraction of specimens, curve 4-RA) precedes pearlite transformation. If they are cooled slowly (1-1.5 °C/s), the isothermal pearlite formation is preceded by very insignificant precipitation of carbides, since contraction in curve 5 is practically absent. This means that excess carbides have time to be released from solution in austenite during slow cooling of the specimens to the isothermal temperature. At the same time, the content of alloy solute considerably decreases in the γ-phase, causing the decrease in the stability of austenite. Furthermore, it is possible that the fine carbides inhibit recrystallization.

Suitability of combined RA processing is determined not only by the stability of the deformed austenite, but also by other factors. To find and evaluate these factors, all the steels were annealed according to schedules shown in Fig. 1. Specific austenitization temperatures, deformation temperatures, isothermal holding temperatures, and times of annealing are shown in Table 2. The results of annealing in compliance with the standard process were evaluated by means of microstructure and hardness of the steels. For hypoeutectoid steels 30CrMnSiNi2 and 55CrSi2 annealing combined with hot rolling does not present any difficulties. In obtaining the required structure and hardness even relatively short (0.5-1 h) holding times proved to be enough to obtain the desired hardness. Moreover, austenite deformation stimulates polymorphic transformation and increases the content of ferrite in the structure, leading to a slight decrease in the steel's hardness as compared with conventional annealing. An increase by a factor of three or four of the time of isothermal holding (during both the conventional and RA processing) results in slight additional de-

Table 2 Hardness and microstructure of steels after isothermal annealing

Grade	Typical hardness HB and microstructure after conventional annealing	Processing schedule: Prior: Heating T, °C	Processing schedule: Prior: Rolling T, °C	Isothermal annealing: T, °C	Isothermal annealing: Time, h	Hardness HB	Microstructure
30CrMnSiNi2	Pearlite + ferrite	850	no def.	650	1	235	Pearlite + ferrite
	255 HB	1150	950-900	650	1	229	Pearlite + ferrite
		850	no. def.	650	2	229	Pearlite + ferrite
		1150	950-900	650	2	223	Pearlite + ferrite
		850	no def.	650	4	223	Pearlite + ferrite
		1150	950-900	650	4	220	Pearlite + ferrite
55CrSi2	Pearlite + ferrite	840	no def.	700	0.5	285	Pearlite + ferrite
	285 HB	1150	950-900	700	0.5	269	Pearlite + ferrite
		840	no def.	700	1	255	Pearlite + ferrite
		1150	950-900	700	1	255	Pearlite + ferrite
		840	no def.	700	3	244	Pearlite + ferrite
		1150	950-900	700	3	248	Pearlite + ferrite
80Mn	Spheroidite(a)	750	no def.	700	2	177	Spheroidite + pearlite
	187 HB	750	750-730	700	2	174	Spheroidite
		1150	no def.	700	2	201	Pearlite
		1150	800-750	700	2	197	Pearlite
70Cr3	Spheroidite	820	no def.	700	3	207	Spheroidite
	217 HB	820	820-800	700	3	207	Spheroidite
		1150	no def.	700	3	285	Pearlite
		1150	880-850	700	3	262	Pearlite
A2 (MSAC)	Spheroidite + undissolv. carbides	880	no def.	720	6	229	Spheroidite + carbides
	241 HB	880	880-850	720	6	229	Spheroidite + carbides
		1150	880-850	720	6	344	Pearlite + martensite
M2 (USAC)	Spheroidite + undissolv. carbides	880	no def.	740	4	228	Spheroidite + carbides
	255 HB	880	880-860	740	4	225	Spheroidite + carbides
		1150	no def.	740	4	266	Mixed carbides
		1150	950-990	740	4	259	Spheroidite + carbides

(a) Spheroidal carbides in a matrix of ferrite

crease in hardness. Due to this fact, one can conclude that during RA processing, the optimal isothermal holding for hypoeutectoid steel is the duration that guarantees complete decomposition of austenite.

As becomes apparent from the data in Table 2 and Fig. 6, RA processing of eutectoid 80Mn and hypereutectoid 70Cr3 and A2 steels produced unsatisfactory results, both with regard to hardness and to microstructure. For these steels, the required structure for these annealed steels should be spheroidal carbides in a matrix of ferrite (spheroidite) but not pearlite. The unsatisfactory results are not caused by deformation but by high-temperature heating, since after such heating, undeformed specimens behave the same. If the heating temperature is such that carbides do not completely dissolve in austenite, annealing (both conventional and combined with deformation) results in the formation of a spheroidite (Fig. 6a and b). In steel 80Mn there is even some pearlite in the microstructure of the normally annealed steel (Fig. 6a). When austenite was deformed at this temperature, the spheroidization of carbides occurred more intensively (Fig. 6b). When the preheating temperature of steel 80Mn was 1150 °C, the microstructure appeared to be mainly pearlite in both nondeformed and deformed specimens (Fig. 6d). The same was observed in hypereutectoid steels. Eutectoid carbides were globular after conventional annealing and lamellar after high-T heating. During manufacturing by hot rolling and forging, high-T heating is used and complete dissolution of carbides in the austenite is inevitable, as well as the appearance of pearlite structure after deformation and cooling. Such a structure is not acceptable for annealed products made of eutectoid and hypereutectoid steels, and consequently, RA processing of rolled and forged products made from these steels is hardly possible.

In earlier work, the possibility of performing RA processing on USAC hypereutectoid steel M2 was partially evaluated (Ref 3). In the current study, the effect of the duration of isothermal holding on hardness and microstructure after annealing has been studied. The data in Table 2 and Fig. 7 not only confirm the previous conclusion that RA processing of M2 steel is quite possible but also permits additional assessments. As shown in Fig. 7, the resulting hardness is higher than during conventional annealing (Table 2). The same result in the earlier study, led to the conclusion that the principal cause of higher hardness attained during combined annealing may arise from aging of the deformed austenite before transformation (Ref 3). However, increase in the duration of isothermal holding by 3 times or in the transformation temperature by 30 °C results in additional decrease of the hardness by 10-15 HB to the specified level.

Microstructural examination of M2 specimens revealed that during annealing with heating at 880 °C (schedules 1 and 2 in Fig. 1), the typical resulting microstructures consisted of fine spheroidite and large globular primary and secondary carbides (Fig. 8a and b). A similar microstructure was observed after high temperature heating at 1150 °C (Fig. 8c and d). In all specimens (schedules 1-4) both primary and secondary car-

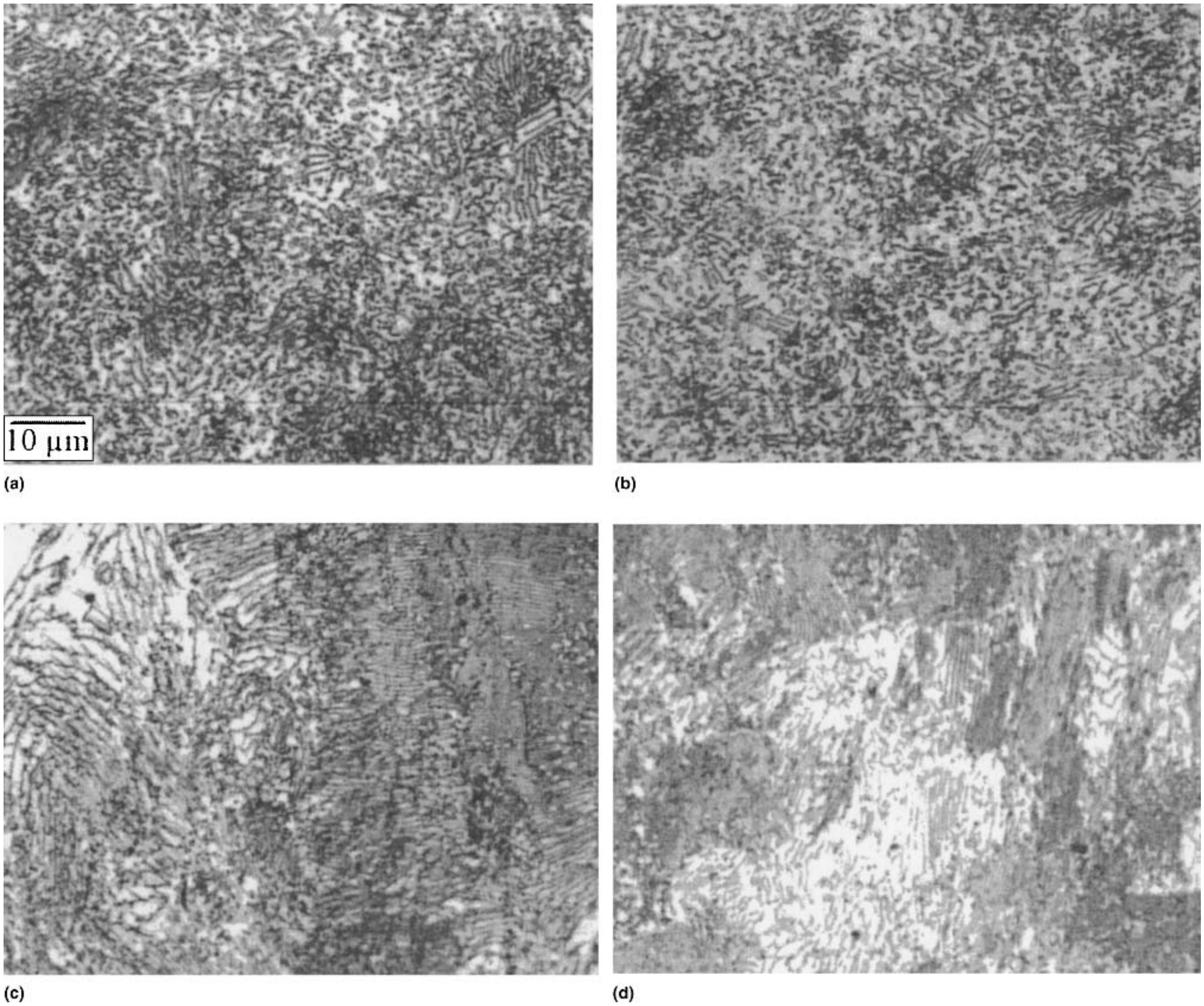

Fig. 6 Effect of heating temperature and rolling on the microstructure of 80Mn steel: (a) conventional annealing (schedule 1 in Fig. 1), (b) rolling after low preheating (schedule 2), (c) high preheating at 1150 °C (schedule 3), (d) RA processing (schedule 4); ×1,000

bides have spherical shape but the latter are smaller. During heating at high temperature, primary carbides remain the same while secondary ones partially dissolve in the austenite and their amount is notably less than after low-*T* heating. Eutectoid carbides are the finest carbides in the microstructure distributed in the ferrite matrix. Their shape and density are affected by the deformation (Ref 18), and conditions of heat treatment as seen from TEM micrographs in Fig. 9. After conventional annealing and RA processing (schedules 1 and 4) eutectoid carbides are mainly globular (Fig. 9a, d, and e), while after high-temperature heating at 1150 °C without deformation (schedule 3), the eutectoid carbides are much less spheroidized. Many carbides are needle shaped (Fig. 9c), or lamellar, similar to pearlite in some areas (Fig. 9b). This may be responsible for the higher hardness after high-temperature annealing without rolling compared with RA processing (Table 2). Figures 9d and 9e show carbides that are much finer than the eutectoid ones and distributed inside grains. These carbides were evidently registered by dilatometry [contraction before the transformation begins (Fig. 5)] and they may contribute to higher hardness than that of

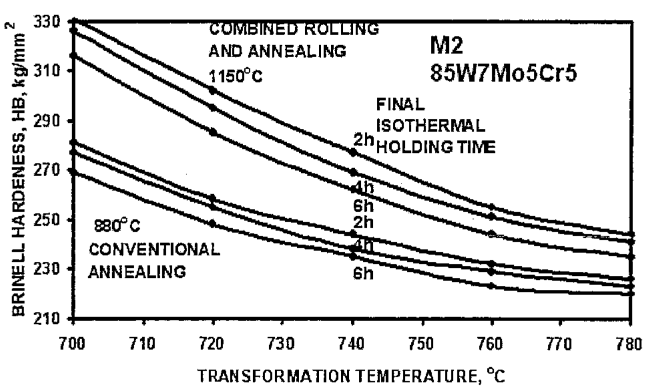

Fig. 7 Hardnesses for M2 steel from different transformation temperatures and holding time after (a) conventional annealing (schedule 1 in Fig. 1) and (b) RA Processing (schedule 4)

conventional annealing. Therefore, the deformation enhances precipitation of excessive carbides inside austenite grains before the transformation, while in undeformed austenite after

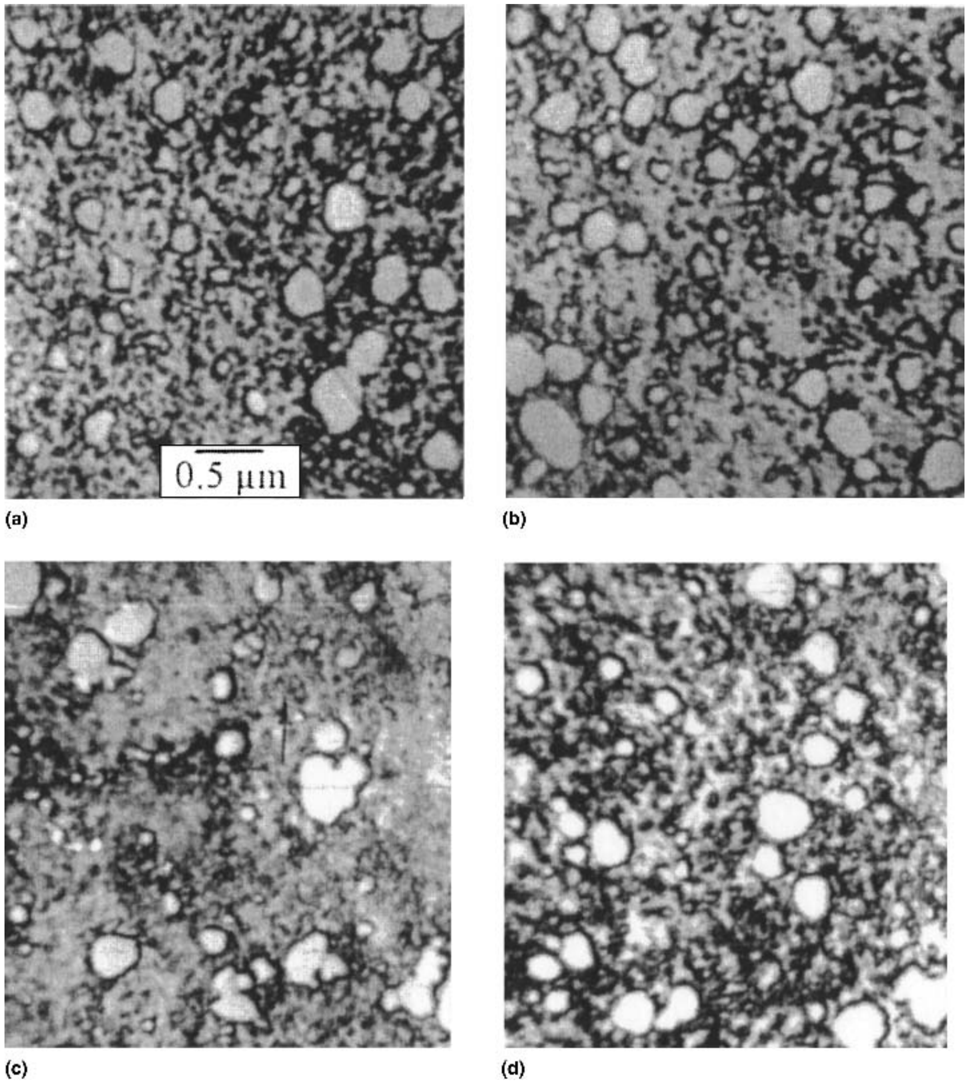

Fig. 8 Effect of heating temperature and rolling on the microstructure of M2 steel: (a) conventional annealing (schedule 1 in Fig. 1), (b) rolling after low preheating at 880 °C (schedule 2), (c) high preheating at 1150 °C (schedule 3), and (d) RA processing (schedule 4); ×1,400

high T heating (schedule 3), they form at already existing secondary and primary carbides.

4. Conclusions

High-temperature heating, and subsequent deformation simulation of the final stage of hot rolling, have diverse effects on the kinetics of austenite transformation to pearlite in steels of different structural classes.

In hypoeutectoid and eutectoid steels, deformed after high-temperature heating, transformation to pearlite occurs at a higher rate than after heating for conventional annealing. The augmented speed of the pearlite reaction decreases during slow cooling of the deformed specimens to the isothermal holding temperature as a result of recrystallization.

In hypereutectoid steels, both the magnitude and direction of change from combined high-temperature heating and hot deformation are related to the alloying content. In low-alloy steels during high-temperature heating, the solute content slightly increases so that the austenite after deformation has a lower or similar stability in the pearlite range as nondeformed austenite after heating for annealing. When there is a considerable alloy content in MSAC hypereutectoid steel, the austenite heated to give complete dissolution and then deformed can be many times more stable than that heated for annealing.

In USAC hypereutectoid steel, as a result of incomplete dissolution of carbides during heating to 1150 °C, the stability of austenite increases; however, it is less than in hypereutectoid steel with MSAC. Hot deformation has a strong destabilizing effect on austenite decomposition. Therefore, the combined effect of high-temperature heating and hot rolling on the duration of the decomposition of austenite of USAC hypereutectoid steel turns out to be rather small.

Combined hot rolling and annealing of hypoeutectoid and USAC hypereutectoid steels provides the required standard structure and hardness under isothermal holding for 1-2 h. The same RA processing of eutectoid and hypereutectoid steels does not lead to a positive result.

At RA Processing of USAC steel M2, precipitation of fine carbides within austenite grains occurs before the transformation starts, resulting in an increase in hardness. To obtain the

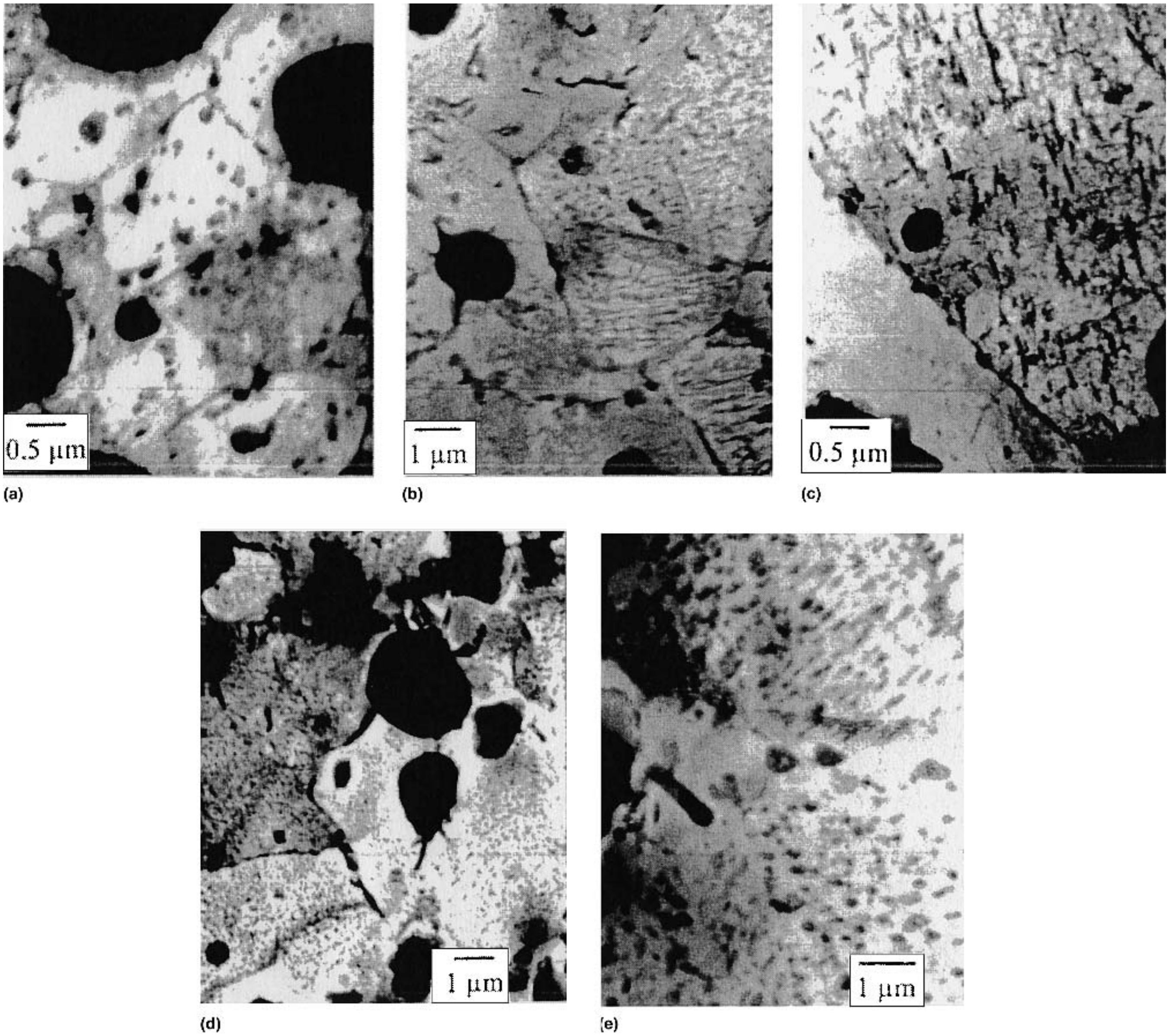

Fig. 9 TEM micrographs of M2 steel after (a) conventional annealing (schedule 1 in Fig. 1), ×12,000; (b, c) high preheating at 1,150 °C (schedule 3), (b) ×6,000, (c)×12,000; (d, e) RA processing (schedule 4), (d) ×5,000, (e) ×15,000

hardness of conventional annealing, the isothermal annealing temperature in RA processing has to be increased by about 30 °C.

References

1. V.V. Polyakov, *Resource Economy in Ferrous Metallurgy,* Mashinostroenie, Moscow, 1993 (in Russian)
2. Yu.A. Geller, *Tool Steels,* Metallurgia, Moscow, 1983 (in Russian)
3. V.M. Khlestov, E.V. Konopleva, and H.J. McQueen, Effect of Deformation and Heating Temperature on the Austenite Transformation to Pearlite in High Alloy Tool Steels, *Mater. Sci. Technol.,* Vol 18, 2002, p 54-60
4. H.J. McQueen, E.V. Konopleva, C.A.C. Imbert, and V.M. Khlestov, Hot Working and Effects on Austenite Transformation to Pearlite in M2 Tool Steels, in *Thermomechanical Processing: Mechanics, Microstructure & Control,* E.J. Palmiere et al., Ed., Univ. Sheffield, UK, 2003, p 368-373
5. H.J. McQueen, V.M. Khlestov, and E.V. Konopleva, Influence of Preheating (850-1150°C) and Hot Rolling on Transformation to Annealed Structure of Hypo-, Hyper-, and Eutectoid Steels, *2nd Int. Conf. on Thermomechanical Processing of Steels (TMP'2004, Liege),* M. Lamberights, Ed., Verlag Stahleisen, Dusseldorf, 2004, p 152-156
6. E.V. Konopleva, V.M. Khlestov, and H.J. McQueen, Hot Deformation Effects on Austenite Decomposition in Alloy Steels, *Phase Transformations During Thermal/Mechanical Processing of Steel,* E.B. Hawbolt and S. Yue, Ed., Met. Soc. CIM, 1995, p 243-258
7. Y.E. Smith and C.A. Siebert, Continuous Cooling Transformation Kinetics of Thermomechanically Worked Low-Carbon Austenite, *Metall. Trans. A,* Vol 2, 1971, p 1711-1725
8. D.J. Walker and R.W.K. Honeycombe, Effects of Deformation on the Decomposition of Austenite: Part 1–The Ferrite Reaction, *Met. Sci.,* Vol 12, 1978, p 445-452
9. J.J. Jonas, R.A. do Nascimento, I. Weiss, and A.B. Othello, Effect of Deformation on the $\gamma \rightarrow \alpha$ Transformation in Two High Silicon Dual-Phase Steels, in *Fundamentals of Dual Phase Steels,* R.A. Kot and B.L. Bramfitt, Ed., TMS-AIME, 1981, p 95-112
10. V.M. Khlestov, E.V. Konopleva, and H.J. McQueen, The Hot Working and Transformation to Ferrite of V-Mo, Nb and Nb-V Steels, *Can. Metall. Quart.,* Vol 35 (No. 2) 1996, p 169-180
11. V.M. Khlestov, E.V. Konopleva, and H.J. McQueen, Effects of Hot

Deformation on Austenite Transformation in Low Carbon Mo-Nb and C-Mn Steels, *Mater. Sci. Technol.,* Vol 14, 1998, p 783-792
12. V.M. Khlestov, E.V. Konopleva, and H.J. McQueen, Kinetics of Austenite Transformation During Thermomechanical Processes, *Can. Metall. Quart.,* Vol 37 (No. 2), 1998, p 75-89
13. V.M. Khlestov, G.K. Dorozhko, G.Y.A. Betin, and N.V. Yarosh, Instrument for Measuring Specimen Expansion, Author's Certificate 551497 (USSR), *Bulletin Discoveries, Inventions,* Vol 11, 1977, p 47-48 (in Russian)
14. L.E. Popova and A.A. Popop, *Transformation Diagrams of Austenite in Steels,* Metallurgiya, Moscow, 1991 (in Russian)
15. H.J. McQueen, S. Yue, N.D. Ryan and E. Fry, Hot Working Characteristics of Steels in Austenite State, *J. Mater. Proc. Technol.,* Vol 53, 1995, p 293-310
16. V.M. Khlestov and G.K. Dorozhko, *Transformation of Deformed Austenite in Steel,* Priazovsky State Technical University, Mariupol, 2002 (in Russian)
17. *Physical Metallurgy: Part 1,* 3rd ed., R.W. Cahn and P. Haasen, Ed., Elsevier Science Publishing, 1983
18. M.I. Sinelnikov and E.A. Titarenko, *Izvestia AN USSR, Metally,* Vol 3, 1979, p 126-130 (in Russian)