Bainite Transformation in Deformed Austenite

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In this work the effect of deformation of metastable austenite on its subsequent isothermal decomposition in bainite regions was investigated in three chromium steels with varying carbon contents. Four different types of treatments were used and the effect of carbon content was studied. Results of the kinetic measurements and of structural changes showed that the most important feature of these treatments was formation of narrow ferrite strips along the slip bands in austenite, which preceded formation of bainite of the usual morphology. A marked dependency of this ferrite strips formation on temperature suggests that the over-all transformation would strongly depend on deformation and recovery processes in austenite and on their mutual relation. A new qualitative model of the bainite transformation in deformed austenite is presented.

$\mathbf{F}$ ROM its earliest development the thermomechanical treatment (TMT) has been regarded as possessing prime importance in the steel strengthening effort. Accordingly, TMT has been the object of much investigation and many papers have been published during the years—for review see for instance Ref. 1. Most of these studies have been aimed naturally at the high-strength steels with martensite structure. Their mostly well experimentally documented conclusions are fairly consistent, allowing thus a basic agreement on a theoretical model explaining origin of better mechanical properties of these steels. On the other hand, data on TMT of medium-strength steels with bainite[2-5] and/or bainite-martensite structures[5-7] are relatively scant and inconsistent. The reasons lie apparently in the complexity of the bainite transformation and in varying modes of austenite deformation used. The latter follows from a detail analysis of some of the works investigating the influence of the austenite deformation on its isothermal decomposition.[2,6,8-10]

Any further data on processes connected with deformation of metastable austenite and its subsequent transformation are of a practical importance for selection of the optimal conditions of TMT. The present work contains results of an extensive study on effects of temperature, carbon content and amount and mode of the metastable austenite deformation on morphology and kinetics of the bainite transformation. A qualitative model of the bainite transformation in deformed steels is also presented.

EXPERIMENTAL

Samples from model chromium steels (Table I) with dimensions $6 \times h \times 150$ mm were deformed by single-pass rolling (roll diam: 60 mm; deformation rate: 8.0 s^{-1}) after austenitizing 15 min/1050°C. The original dimension h was calculated so that after all used degree of deformation the final thickness of samples was 3.5 mm.

The four modes of TMT used are schematically illustrated in Fig. 1. For comparison of structural changes undeformed samples under the same conditions were made (CT for conventional treatment). The

stop-transformation method was applied to both sets of samples. For all types of TMT used the zero time of the isothermal holding was set up in the moment of the sample quenching to the first transformation temperature. Before rolling (with exception of B type of TMT) all samples were tempered for 20 s at the rolling temperature.

Morphology of the resulting structure was studied on the cross-sections either by optical microscopy (etching in mixture of Nital and Vilella-Bain etchants) or by electron microscopy (collodion replica or thin foils techniques). Fractions of the bainitic ferrite were de-

Table I. Chemical Compositions and M_s Temperatures of Used Steels

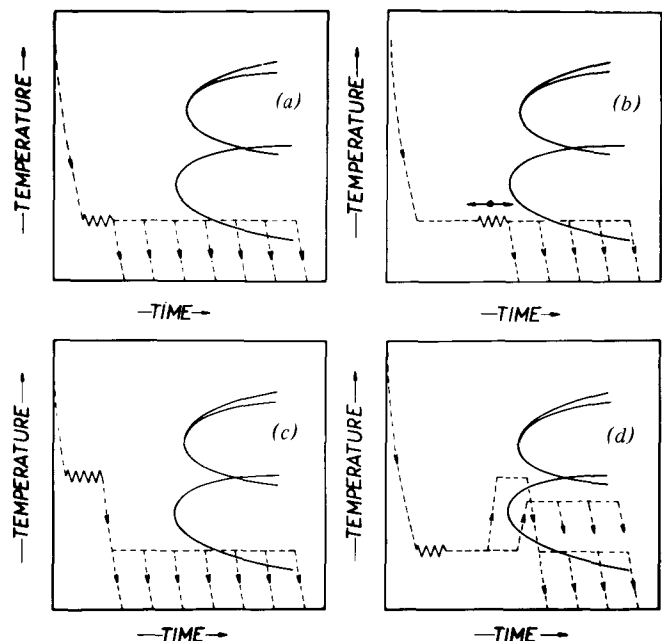| Steel	Concentration, Wt Pct									M_s, °C
	C	Cr	Mn	Si	P	S	Ni	Cu	Al	
L	0.43	3.51	0.38	0.18	0.019	0.020	0.12	0.15	0.045	280
M	0.86	3.74	0.39	0.20	0.041	0.020	0.05	0.05	0.009	160
H	1.02	3.56	0.40	0.19	0.017	0.025	0.15	0.15	0.020	100

Fig. 1—Schematical representation of TMT used.

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termined directly on the screen of the optical microscope by the point-counting method with a constant step 0.01 mm along the longer axis (relative error δ = 0.10 with reliability ε = 0.90). In all these studies always the central regions only of deformed samples, definitely unaffected by the contact quenching or decarburizing were used. The maximum recorded affected surface layer never exceeded 0.2 mm.

RESULTS

Since results for all three used materials are basically similar (as it will be shown and discussed later), all the results described henceforth are for material M unless otherwise stated. For the type of TMT see Fig. 1. Terms "undeformed austenite" or "undeformed regions" refer to regions with no visible slip bands.

Type A

The transformation curves (Figs. 2 and 3) exhibit a marked reduction of the incubation period and increasing transformation rate with increasing degrees of deformation. Vicker's hardness numbers correspond to transformed fractions, but for the same fraction they slightly increase with increasing degree of deformation. Samples quenched immediately after a deformation higher than 5 pct contain a mixture of the plate and lath martensites with preserved slip bands from the prior deformed austenite (Fig. 4). Since these bands offer a simple diffraction pattern and since they are not confined to a single martensite plate, it is very likely to assume that they are formed rather by a slip in austenite than by deformation twinning. This view is further supported by the fact that these bands often appear simply as places of a different contrast inside of the martensite plates. This difference can
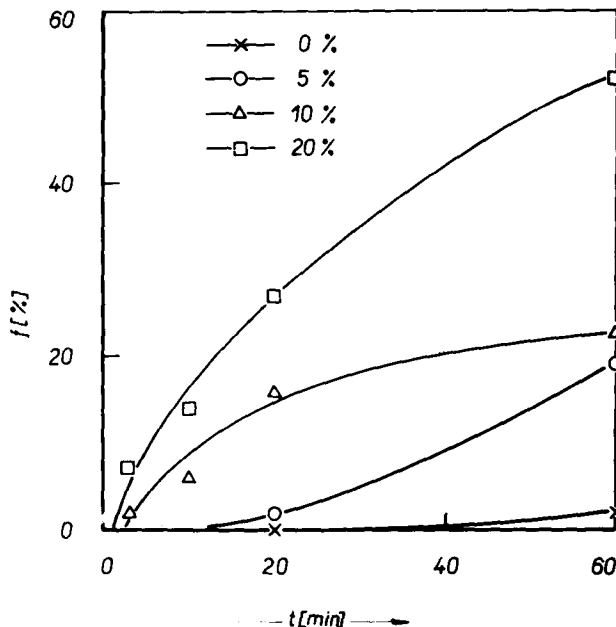
be due to a varying dislocation density or to small local changes in the lattice orientation.

The beginning of deformation accelerated bainite transformation has been then characterized by formation of very narrow dark etching strips along the slip bands with a substructure noticeably different from the matrix (Figs. 5 and 6), but with a very small or no mutual disorientation (single diffraction spots). Very similar microstructural features were already observed by Schmatz and Zackay[6] and by Evans and O'Neill.[2] The first authors described this structure simply as "the carbide decoration of the slipped areas". Evans and O'Neill used for description of this special structure terms as "the strain-induced product" or "the accelerated austenite breakdown product" or simply "the transformed constituent". After unsuccessful attempts to identify this product the authors concluded that "the strain-induced transformation product did not resemble normal bainite. Strain-induced martensite was not detected."
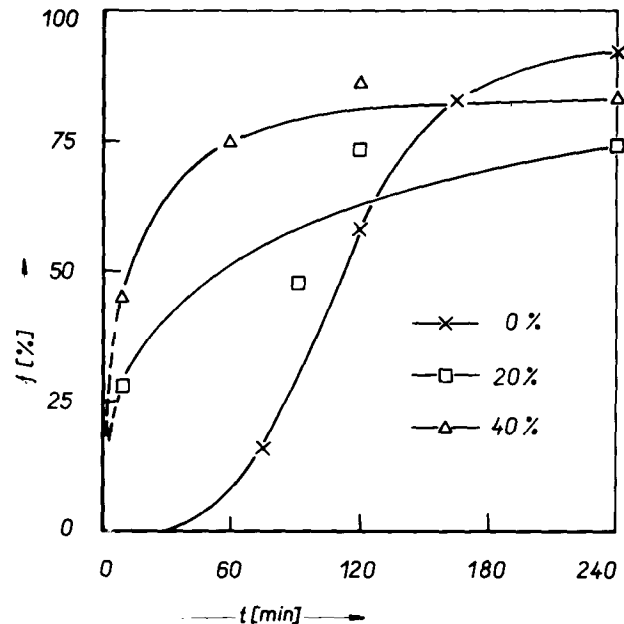
Fig. 3—Effect of the degree of deformation on the transformed fraction f (temperature 325°C; Type A treatment).

Fig. 4—Transmission electron micrograph of the martensite structure (deformation 20 pct/275°C → water).

Fig. 2—Effect of the degree of deformation on the transformed fraction f (temperature 275°C; Type A treatment).

After a careful study of the over-all character of these newly formed strips, detailed analysis of their morphology and substructure (using extensively the transmission electron microscopy), and regarding their lower microhardness values, behavior under polarized light, time dependency of their formation, *etc.*, we have concluded that these strips are ferritic.
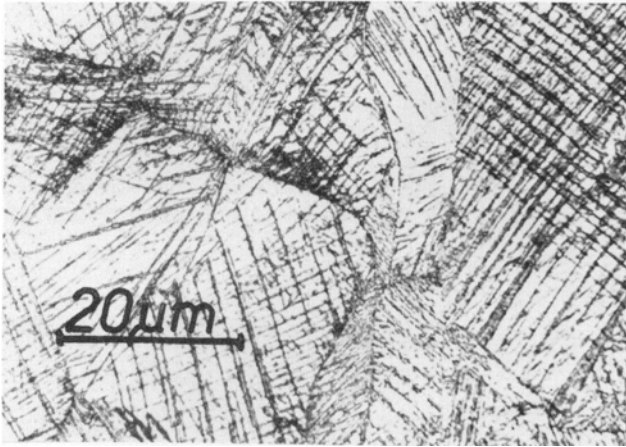

Fig. 5—Optical micrograph of the ferrite strips (deformation 20 pct/275°C → 3 min/275°C → water).
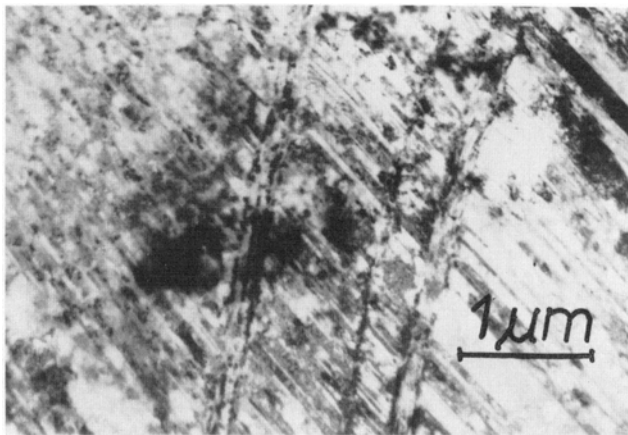

Fig. 6—Transmission electron micrograph of the ferrite strips (deformation 20 pct/275°C → 3 min/275°C → water).
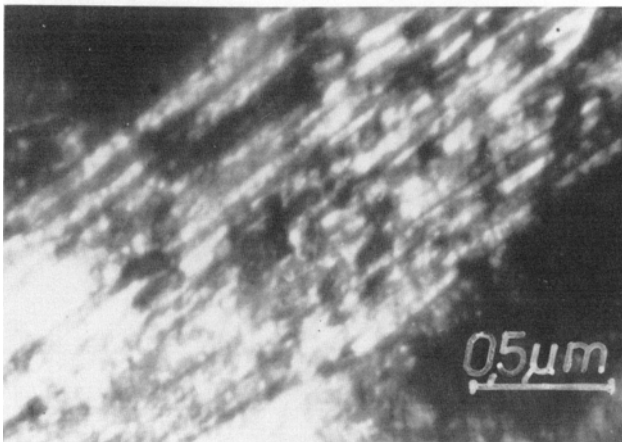

Fig. 7—Transmission electron micrograph of ferrite strips with finely precipitated carbides (deformation 20 pct/275°C → 10 min/275°C → water).

Their width (less than 0.1 μm) changes neither with the isothermal holding time nor with the degree of deformation, their frequency, however, increases with deformation. After a sufficiently long holding time at the isothermal temperature when most of the slip bands are already decorated by the ferrite particles, bainitic ferrite of the usual morphology forms at the ferrite-austenite interface and in the undeformed regions as well. Later on in this process a decomposition inside the ferrite strips occurs and fine carbides precipitate in a direction parallel to the slip band axis (Fig. 7) and the slip lines become less evident.

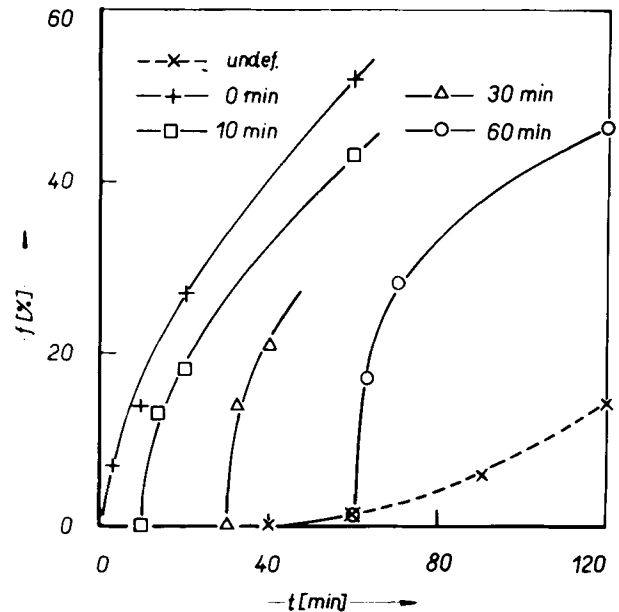The sequence of the above described changes has

Fig. 8—Effect of the isothermal holding before deformation on the transformed fraction f (temperature 275°C, deformation 20 pct; Type B treatment).
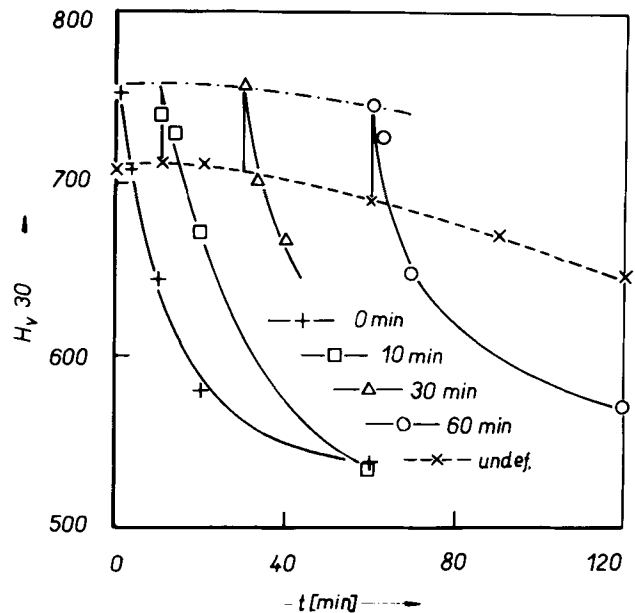

Fig. 9—Effect of the isothermal holding before deformation on the Vickers' hardness Hv (temperature 275°C, deformation 20 pct; Type B treatment).

been the same regardless of temperature but the fraction of ferrite strips markedly increases with decreasing temperature. The lowest deformation used— 5 pct—has not produced any structural changes—only a few isolated narrow ferrite strips have been observed by the optical microscopy.

Type B

A deformation during the isothermal holding generally brings about only a shift in the transformation start but the character of the transformation curves is
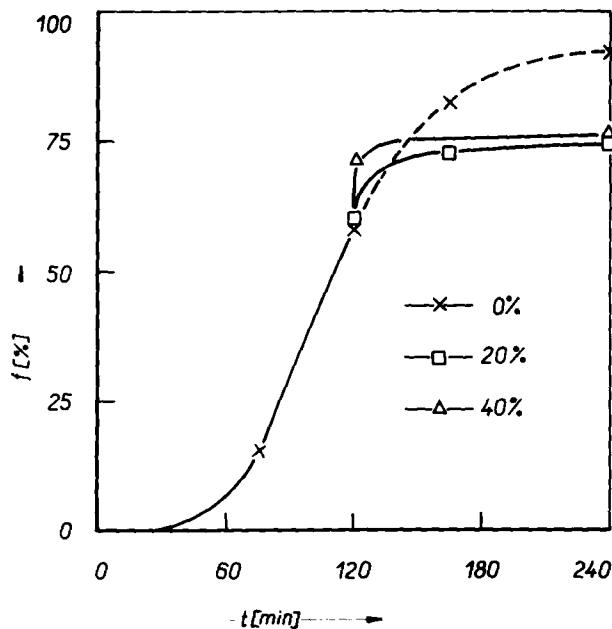
similar to Type A treatment (Fig. 8). Strengthening recorded in this case (= a hardness increase against CT in Fig. 9) does not depend on the very moment of the deformation applications during the incubation period. When a deformation has been applied after the beginning of the bainite transformation, kinetics of the transformation markedly changes (Fig. 10) and hardness again increases (Fig. 11). However, in this case the strengthening has been higher and has remained almost unchanged during further decomposition. Structural features of the resulting structure have shown no dependency on the length of time elapsed before the application of deformation for applications during the incubation period. However, when some bainitic ferrite particles have been already formed before deformation, these do not grow further and act as effective slip barriers inside of the austenite grain (Fig. 12).

Fig. 10—Effect of the isothermal holding before deformation on the transformed fraction f (temperature 325°C, deformation 0 to 40 pct; Type B treatment).

Fig. 12—Optical micrograph after 60 min/275°C → deformation 20 pct/275°C → 10 min/275°C → water.
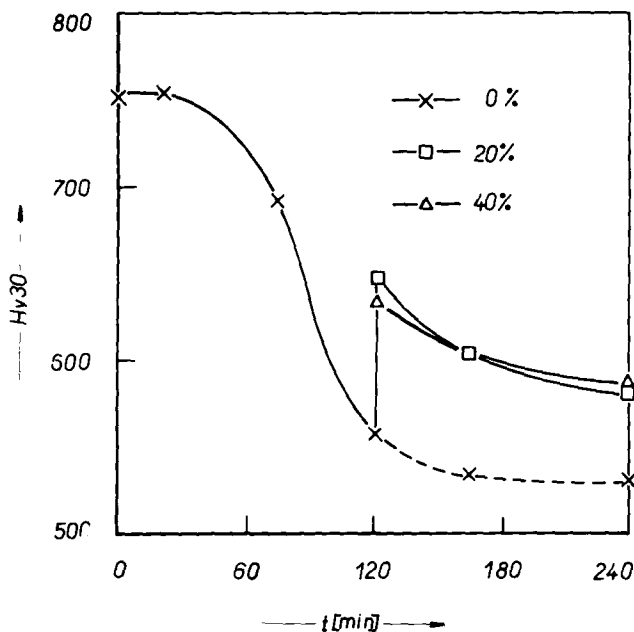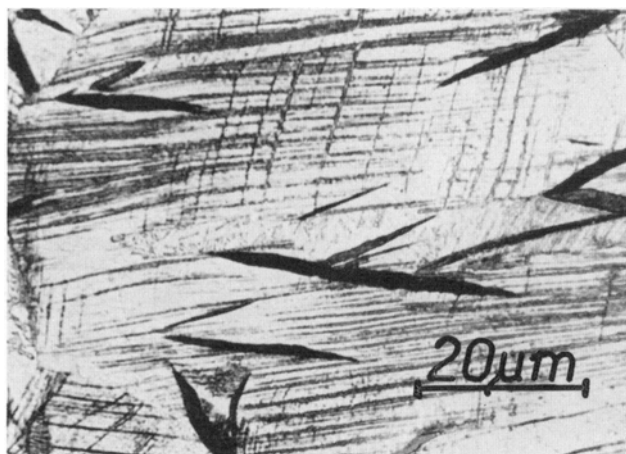

Fig. 11—Effect of the isothermal holding before deformation on the Vickers' hardness Hv (temperature 325°C, deformation 0 to 40 pct; Type B treatment).
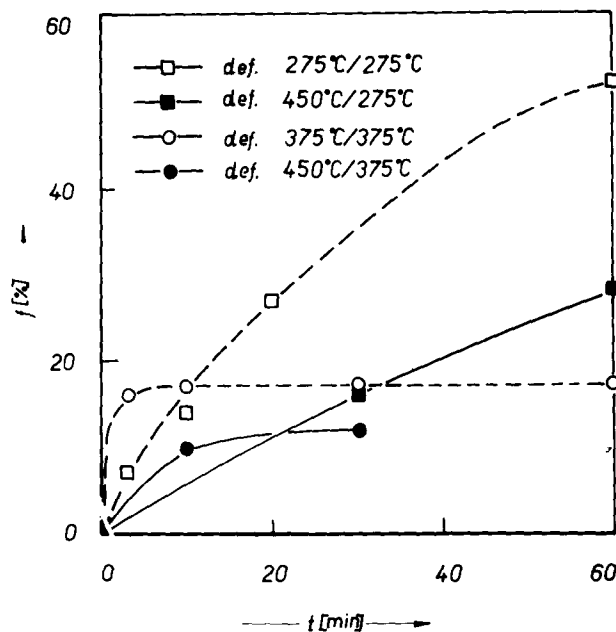

Fig. 13—Effect of the temperature changes after deformation on the transformed fraction f (deformation 20 pct; Type C treatment).

Type C

A deformation in the region of the higher austenite stability (450°C) has not altered significantly either character of the transformation curves (Fig. 13) or the morphology, but the transformation has been much slower compared to Type A.

Type D

Fig. 14 illustrates changes in the kinetics for temperature increase from lower to upper bainite regions. In this case the transformation continues primarily by
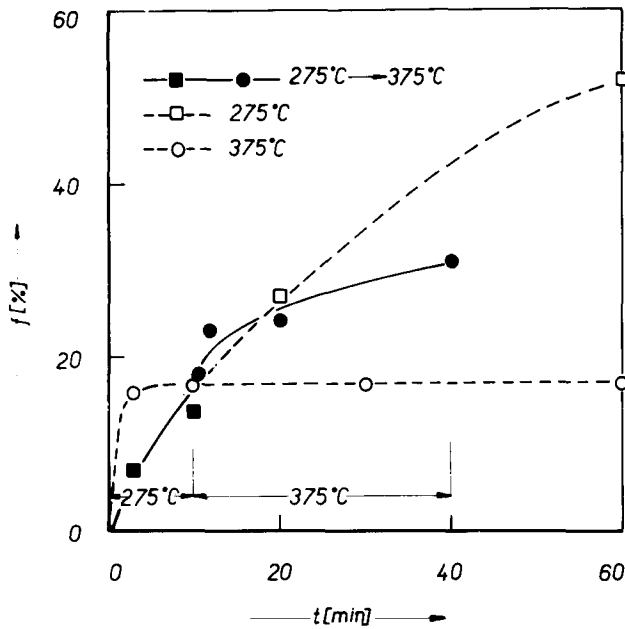

Fig. 14—Effect of the temperature changes after deformation on the transformed fraction f (deformation 20 pct; Type D treatment).
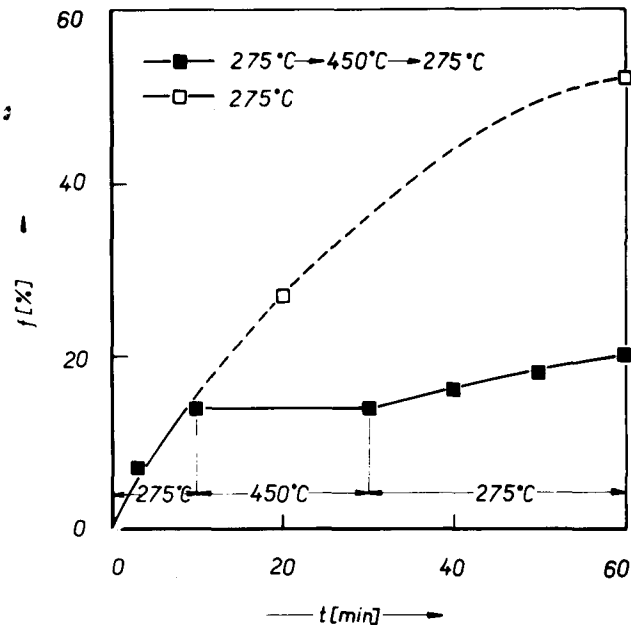

Fig. 15—Effect of a temporal temperature increase on the transformed fraction f (deformation 20 pct; Type D treatment).

growth of the upper bainite at the narrow ferrite strips-austenite interface. Temporary holding at the metastable austenite region leads to a temporary cessation of the transformation which would continue only after cooling to the original temperature, but with a much slower rate (Fig. 15). The resulting structure has consisted mainly of ferrite strips formed before the temperature increase with a small amount of bainitic ferrite formed after the temperature returned to its original value.

Varying Level of Carbon

To assess the influence of the carbon content, treatments of A and B types have been also applied to steels L and H (Table I). The character of the transformation curves is similar (Fig. 16) but the overall transformation rate decreases with increasing carbon content. Furthermore, two other changes have been observed in steel L (low C): 1) in the A-type treatment the narrow ferrite strips formed already during the deformation; 2) B-type treatment caused an increase in the final fraction of bainite in the resulting structure.

DISCUSSION

A basic conclusion can be drawn from all the experimental results of this work: a deformation of the metastable austenite strongly affects the transformation characteristics of its subsequent isothermal decomposition.

A shift in the beginning of transformation to shorter times regardless of the temperature and degree of deformation used (Figs. 2 and 3) shows that there is a sufficient amount of potential nuclei of the new phase in deformed austenite which have favorable conditions for a fast growth. A significant drop in the lattice
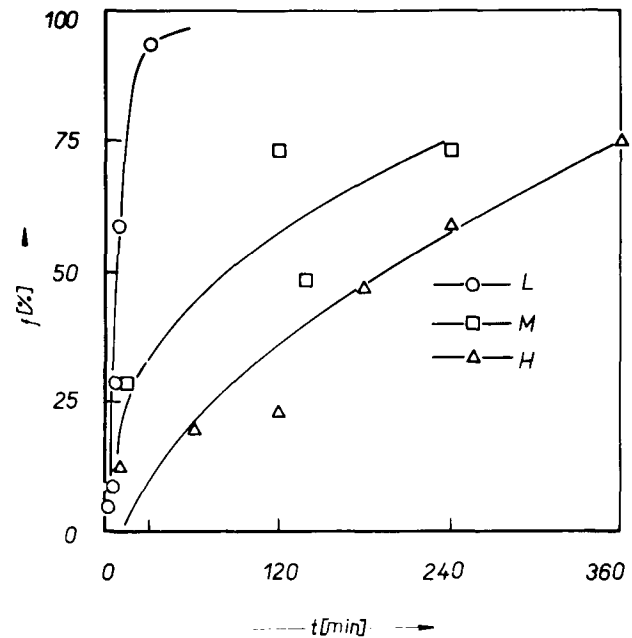

Fig. 16—Effect of the varying carbon concentration (Steel L, M, H) on the transformation kinetics (deformation 20 pct, 275°C; Type A treatment).

parameter of residual austenite immediately after deformation[11] suggests a considerable redistribution of carbon to places with a higher lattice defect density in the vicinity of the slip bands which are then likely to act as places of enhanced nucleation. However, no carbide precipitation during or immediately after deformation has been found. First carbides have appeared together with the first ferrite strips inside of these strips. This finding rather confirms the work of Škarek et al[22] then that of Thomas et al.[23] In the latter work the authors have found special carbides immediately after deformation while Škarek et al have not.

The incubation period does not change much with the degree of deformation. Higher deformations bring an increase in the slip bands density which causes a faster transformation from the beginning, but the width of these bands, e.g. the diffusion paths of carbon atoms, remains the same (about 0.1 μm) and the incubation periods do not change. This conclusion also explains similar results of some other authors.[2,12]

An exception is the 5 pct deformation, when the incubation period is again fairly shortened and the transformation slightly accelerated, but the factor accelerating transformation in this case is probably the stress originated from the contact quenching of the sample surface by cold rolls. This presumption is justified by the fact that practically no slip bands or narrow ferrite strips have been found and the bainitic ferrite has exhibited morphology typical for undeformed samples. Similar results have been obtained for instance by Jepson and Thompson[9] and other authors,[10,13-15] for transformation under a stress.

The entire transformation can be then divided into three steps:
● very fast formation of ferrite in the form of narrow strips;
● slower growth of bainitic ferrite at the interface between the ferrite strips and the deformed austenite;
● even slower growth of bainitic ferrite of the usual ("conventional") morphology in the undeformed regions.

The transformation rate in the first and partially also in the second stage is determined above all by the degree of deformation (Figs. 2 and 3) while the transformation rate in the third step depends basically only on the transformation temperature. On the other hand, higher temperatures lower the transformation rate in the first stage, since the higher the transformation temperature, the more extensive the relaxation and recovery of the deformed austenite and the slower the transformation. At higher temperatures with larger extent of recovery (e.g. less imperfections and lattice defects) also the carbon redistribution will be smaller. These conclusions are in good agreement with experimental data of Baranov et al.[12] It must be noted, however, that a certain redistribution of carbon takes place during the incubation period before deformation. This redistribution, always preceding the start of the bainite transformation,[16-18] can lead to a slight carbon deprivation of martensite matrix and to an increase in the amount of retained austenite in the carbon-enriched regions. Otherwise the transformation is not affected by the length of the incubation period before deformation. In any case—even after deformation at the very end of the incubation period—bainite of the usual morphology has not been found. That is probably

due either to changes in the dislocation substructure of deformed austenite or to changes in the carbon concentration of the matrix caused by very fast nucleation and growth of the ferrite strips. Both processes will reduce the driving force for nucleation and growth of the "usual" bainitic ferrite.

For deformations applied during the transformation it has been found that growth of the already existing bainite stopped very likely as a consequence of changes in the quasiequilibrium conditions[19-21] at tips of linearly growing particles. As a very efficient obstacle to the growth of bainite can act also the newly generated dislocation network. On the other hand, bainite particles owing to their higher strength inhibit the slip in the austenite matrix (Fig. 12), the actual amount of strain in untransformed austenite is much higher than for the same degree of deformation applied in the incubation period and thus the ferrite strips and carbides form much sooner—already during the deformation.

The temperature dependency of the strain-induced ferrite strip formation can be expressed by the following rule: the higher the temperature, the smaller the transformed fraction. In accordance with this rule—which is a result of the process of recovery in deformed austenite—the largest decrease in the transformed fraction has been observed for a temporary increase in the transformation temperature to 450°C (Type D treatment). Similarly can be explained the slower transformation after the C-type treatment (deformation at the austenite bay). In addition to the effect of a higher temperature in this case also the longer times needed for cooling samples down to the transformation temperature help to reach a higher extent of recovery and thus still slower transformation.

It has been shown thus far that very important in all the studied processes are the local changes in the carbon concentration. In this respect it is interesting to note that changes in the over-all carbon concentration have not brought any important qualitative changes in the morphology of the resulting structure. Differences in the carbon level however materialize in the process rates and in the austenite stabilization in similar way as in CT. For instance, in the steel with the lowest carbon level, the ferrite strip form already during the deformation.

A following general qualitative model of the austenite transformation to bainite after thermomechanical treatment can be drawn from the obtained results.

After deformation of the metastable austenite first of all a redistribution of carbon atoms in the slip regions takes place connected with formation of narrow ferrite strips along the slip bands and precipitation of metastable carbides. At the interface between the ferrite strips and the strained austenite, bainite starts growing and finally also in the undeformed austenite regions bainite of the usual morphology forms. At the same time marked redistribution and coarsening of the prior formed metastable carbides occurs, probably connected with transformation of these carbides to stable cementite.[2] Fraction of the transformed bainite, its morphology and the over-all transformation kinetics would depend on the carbon content and above all, on the mutual interactions between deformation and recovery processes. In general, the resulting bainitic structure would be a mixture of the primary formed narrow ferrite strips and the conventional bainitic

ferrite (this work, [2,12]). However, under certain circumstances (special combinations of deformation and transformation temperatures) the recovery would suppress the deformation effect and structure will then consist of bainite of the usual morphology and the transformation curves will exhibit the "conventional type" character.[9,10,13-15] Effect of deformation in this case would materialize only through a higher nucleation rate resulting in a faster overall transformation.

This qualitative model basically fits to all, up-to-now-known experimental data while the previous one by Evans and O'Neill[2] does not. These authors have based their model on formation of ϵ carbide in the slip planes immediately after deformation, followed later by transformation of carbon depleted regions to ferrite strips. If it were so, formation of the ferrite strips would have to increase with increasing temperature while the experimental evidence proves just the opposite. This disagreement is obviously due to the fact that authors (making most of their conclusions from the hardness measurements only) have not taken in account any possible large-scale carbon redistribution or the relaxation processes (recovery) in austenite.

CONCLUSIONS

The main results of this work can be summarized in following conclusions:

1) Deformation causes a shift of the bainite transformation start to very short times eventually leading to incubation periods which are practically negligible fractions of the total processing time.

2) Shortly after deformation narrow ferrite strips form along the slip bands. Later, bainite of the usual morphology transforms in the undeformed regions.

3) The character of the transformation curves for deformation affected transformation alters significantly in comparison to curves for CT. In the regions under investigation the sigmoidal type of these curves changed to the parabolical one.

4) Length of the isothermal holding before deformation during the incubation period does not affect the following transformation.

5) Bainite particles formed before deformation act as an efficient slip barrier. Their growth after deformation stops and it is replaced by kinetically faster formation of narrow ferrite strips.

6) Austenite deformation increases the influence of carbon concentration on the resulting fraction of bainite in the final structure.

7) Any increase in temperatures of deformation and/or of the isothermal decomposition leads to a larger extent of recovery in austenite and thus to a decrease in the number of ferrite strips and to slower transformation.

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