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THE MECHANISM OF THE MARTENSITE BURST
TRANSFORMATION IN SINGLE CRYSTALS OF IRON
CONTAINING 31.7 PER CENT NICKEL

Berkeley, California

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IRON CONTAINING 31.7 PER CENT NICKEL

Jack C. Bokros

(Ph. D. Thesis)

September 7, 1962

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ABSTRACT

The martensite burst transformation was studied in single crystals of iron containing 31.7% nickel. It was found that the morphology of the transformation was greatly simplified in crystals that had been strained plastically prior to transformation. This simplification made possible detailed crystallographic studies of the morphology of the transformation. These studies showed that the autocatalysis responsible for the burst transformation had its origin in a mechanical stimulation (coupling) between certain variants of the habit plane that were geometrically oriented such that the stress induced by transformation on one aided transformation on the others. In both strained and unstrained crystals, the most effective mechanical autocatalysis was observed to be between groups of four nearly parallel planes whose poles clustered about common $\langle 110 \rangle$ directions.

In the transformation of strained crystals, the operation of certain of these groups was favored. These were the groups whose planes were nearly perpendicular to the active slip plane, but did not include the group whose poles clustered about the Burger's vector of the active slip system. This behavior was taken to be a consequence of the anisotropic substructure introduced during deformation. Because of the enhanced transformation by certain groups in strained crystals, the frequency of secondary coupling was low, and the regions of the crystals that were transformed through the operation of a single group of four variants of the habit plane were large. This effect was responsible for the simplification of the morphology produced in the burst transformation of strained crystals.

Introduction

For many years there was much uncertainty about what constituted the unique and intrinsic characteristics of the martensitic transformation. In 1955 Bilby and Christian¹ set forth two criteria that have proved to be satisfactory for all such transformations thus far observed. These two essential features are: the transformation proceeds in the absence of diffusion, and the transformation is accompanied by a macroscopic shape change. Greninger and Troiano² were first to point out, however, that for steel the homogeneous macroscopic shear does not account for the total atomic movement, because this deformation cannot produce the martensitic crystal structure from that of austenite. An additional heterogeneous deformation due to twinning or slip on a fine scale is required to take place in the martensitic structure in order that the atoms are carried to their correct position on the martensite lattice. Transmission electron microscopy has recently revealed that the heterogeneous deformation can be either slip or twinning.³

Because of the restraint imposed on a freshly nucleated martensitic crystal in austenite, the plates usually develop with a lenticular or ellipsoidal shape with a characteristic habit plane parallel to the large dimensions of the plate. The nature of the interface that connects the displaced martensite and the austenite is particularly significant since it must remain unrotated and undistorted over macroscopic distances. This implies that the shear component of the macroscopic strain must lie parallel to the interface, whereas the dilatational component must be normal to it.⁴

In recent years much progress has been made in describing the crystallographic features of the martensitic transformation and the nature of the interface. A review of these theories will not be given here, but rather, the reader is referred to the original work of Bowles and Mackenzie,⁵ Wechsler et al.,⁶ and to the excellent reviews of Bilby and Christian^{1, 7} for a rationalization of the crystallographic features; and to Frank,⁸ Bilby and Frank,⁹ Bullough and Bilby,¹⁰ and Knapp and Dehlinger¹¹ for the interpretation of the martensite interface in terms of dislocation arrays that move through the lattice and produce the

necessary atomic adjustments involved in the transformation.

As pointed out above, the formation of a martensite plate is not limited by diffusion and can proceed at great speed (reported to be 10^5 cm/sec or one-third the velocity of sound), even at temperatures approaching absolute zero.^{12, 13} These observations have led to the conclusion that the activation energy for the growth mechanism is essentially zero and that the kinetic problem is primarily one of nucleation. Two different theories of nucleation have been suggested. One set forth by Fisher et al.¹⁴ utilizes classical nucleation theory and involves a thermally activated nucleus. The other, presented by Cohen et al.,¹⁵ assumes the presence in the lattice of certain imperfection arrays, termed strain embryos, that already possess a component of the strain necessary in the transformation, and grow spontaneously as the temperature drops below the M_s . Kaufman and Cohen^{4, 16} have recently extended the original strain-embryo concept by utilizing the dislocation model due to Knapp and Dehlinger,¹¹ and show how such an embryo might grow cataclysmically and how it might be thermally activated at constant temperature to form isothermal martensite. Recent transmission electron microscopy has revealed certain arrays of imperfections that may be able to serve as nuclei.^{17, 18} It now appears clear that the nuclei are formed at preferred sites where conditions are favorable; however, no theory of nucleation can be complete until the exact crystallographic relations describing an "embryo" have been determined.

Thermodynamic treatments (refer to the recent review by Kaufman and Cohen⁴) of the transformation have shown that the transformation is initiated, not at the temperature where the chemical free-energy difference between the austenite and martensite becomes negative, but rather at some lower temperature where additional free energy is available to form the necessary surface and to overcome the restraint of the austenite lattice. Recently Brook et al.¹⁹ have examined the amount of transformation as a function of temperature for a number of steels. Using the thermodynamic equation of Fisher²⁰ and substituting the Kaufman and Cohen^{21, 22} value of $\Delta F^{(\gamma \rightarrow \alpha')}$ for Fe-Ni and the value of Fe-Cr

from Kaufman,²² they have found that the slope of the percent transformation versus temperature curve below M_s is proportional to $\frac{\partial \Delta F(\gamma \rightarrow \alpha')}{\partial T}$. In other words, the increment of the transformation occurring in a given temperature interval is directly related to the driving force supplied in the same temperature interval. Although the data for certain steels can be made to conform to this relation, it is well known that in highly alloyed steels the transformation does not follow such a simple relation; instead, the transformation can be highly autocatalytic. That is, large fractions, as high as 80%, of the transformation are observed to occur in a single burst at or somewhat below the M_s temperature. The role of autocatalysis in the transformation in low-alloy steels is not understood.

Machlin and Cohen²³ have studied the burst phenomenon in a 70% iron-30% nickel alloy and have suggested that the burst transformation involves some sort of cooperative movement among several plates--a movement that operates to transform a large fraction of the austenite in a single audible click. However, the speculations were not substantiated by any experimental work or theoretical analysis. Somewhat later Machlin and Cohen²⁴ measured the macroscopic shear direction for the {259} transformation. The shear directions they measured were not in the habit plane, and consequently were evidently slightly in error. They did, however, note that four of these directions grouped about a common [110] direction, and speculated that cooperative shear might occur amongst them. No definitive experiments have been conducted to date to determine the role of cooperative movements in burst formation and, as a result, no satisfactory explanation of the burst phenomenon is available.⁴

It is clear that an understanding of the autocatalytic effect responsible for bursting is important for the development of a thermodynamic-mechanical relationship that describes the amount of transformation as a function of temperature below M_s , and also for any treatment of nucleation below M_s . This investigation was therefore undertaken to study the transformation in an alloy known to exhibit the burst transformation,

in an attempt to arrive at an understanding of the process. Preliminary experiments with single crystals of austenite iron containing 31.7% nickel revealed that plastic strain at room temperature prior to transformation greatly simplified the morphology of the burst transformation. Large regions of strained crystals were observed to transform in single bursts. In each region there were only four variants of the habit plane at acute angles with one another. Unstrained crystals, when transformed by a single burst, had a complex structure composed of many smaller less well-defined regions.

Because of the apparent simplicity of burst formation in strained crystals, it was decided to study first the burst phenomenon in strained crystals and then to extend the work to unstrained crystals, in which the burst mechanism is more complex.

Experimental Details

An iron-nickel alloy containing 31.7% nickel was selected for investigation since considerable data describing the martensitic transformation in this alloy are available and the transformation is known to involve bursting. In addition, the M_s temperature is conveniently below room temperature.

Single crystals, 1/4 inch in diameter and up to 9 inches long, were grown by using a floating-zone technique. Heating was accomplished by means of a five-turn concentric induction coil previously described.²⁵ The atmosphere consisted of slowly flowing helium purified by being passed over activated charcoal at liquid-nitrogen temperature. The growth rate found to be most satisfactory was 2 inches per hour.

The single crystals were cut to a length of 2-1/2 inches and, by using the concentric plate-type coil, were welded in helium to end grips in order that the crystals could be handled and strained uniformly without gripping them directly. The composition of the alloy selected for the specimen ends (iron containing 25% chromium and 20% nickel) was one that was austenitic but would not transform to martensite. Figure 1 shows a section of a typical weld joint. After welding, all crystals were given a short anneal at 1100°C in helium by passing the induction coil over them at a rate of 5 inches per hour. All crystals were prepared in this way, including those which were not strained prior to transformation.

Chemical analysis of a crystal revealed that segregation of about 0.20% nickel occurred during growth. Because of the slight nickel redistribution during melting and refreezing, all experiments conducted to determine the effect of strain on the transformation were carried out on pairs of crystals obtained by splitting (spark cutting) a single crystal along its axis. Thus two crystals with identical composition were obtained, one of which could be transformed in the strained condition and the other in the unstrained condition. A complete analysis of a typical crystal is given in Table I.

ZN-3292

Fig. 1. Cross section showing a typical weld between a grip and a single crystal. 10X.

Table I: Chemical analysis of a typical crystal.

<u>Element</u>	<u>Weight percent</u>
Nickel	31.75
Carbon	0.010
Manganese	<0.01
Silicon	<0.01
Phosphorus	0.002
Sulfur	0.001
Chromium	<0.01

All deformations were carried out at room temperature in tension in an Instron Testing Machine at a rate of 0.02 inch per inch per minute.

Crystals to be used for habit-plane determinations were prepared in the manner described above, then strained to the desired amount. Two faces at approximately 90° to one another and parallel to the axis of the specimen were then cut on the specimen with the spark cutter. Subsequently, an 0.005- to 0.008-inch layer was removed by electropolishing. The specimens were then strained an additional 1% in order that the slip traces could be observed on the polished faces. A typical crystal prepared in the manner described is shown in Fig. 2.

The orientation of the crystals was determined in most cases, both before and after strain, by the Laue back-reflection technique. However, in cases where the orientation or magnitude of the strain was such that a large amount of duplex slip occurred, a poorly defined Laue pattern was obtained after straining. The final orientation was determined by utilizing the $\{111\}$ slip traces on the two faces. In the former case the accuracy was $\pm 1^\circ$ and in the latter about $\pm 2^\circ$.

The course of the transformation was monitored by simultaneously measuring the temperature and resistance of the specimen as it cooled at a rate of $3/4$ to 1 degree centigrade per minute. The procedure was as follows: Current leads of a Kelvin double bridge were clamped onto the stable austenitic ends. Potential leads and the thermocouples were spot-welded onto the end sections adjacent to the crystal weld. The

ZN-3293

Fig. 2. A typical crystal. 2X.

specimen was then inserted into a brass cup containing isopentane and equipped with a stirrer. This assembly was placed in a double styro-foam dewar which could be cooled at a controlled rate by allowing liquid nitrogen to flow in at a controlled rate. The temperature gradient was always less than 1 degree. In cases in which the transformation was carried to completion, the crystal was transferred to liquid nitrogen after it had cooled to -140°C ; i.e., slightly above the freezing point of isopentane. The resistance of the specimen was measured alternately at liquid-nitrogen temperature and at room temperature until the measured values at each temperature remained constant. This was necessary since it was observed that a small amount of transformation occurred on the second quench to liquid nitrogen temperature. The magnitude of this effect is illustrated in Fig. 4.

For cases where the morphology of the burst transformation was to be determined, the crystal was cooled until the burst occurred and was then immediately heated to room temperature. The variants of the habit plane operating during a particular transformation were determined by means of a two-surface analysis.²⁶

ZN-3294

Fig. 3. Macrograph of two austenite single crystals that have been transformed to martensite. 5X.
Left: Unstrained before transformation.
Right: Extended 30% in tension at room temperature before transformation.

MUB-1310

Fig. 4. Resistance vs temperature curves for two single crystals of identical composition. One unstrained, the other strained 29% prior to transformation.

Experimental Results

As has already been pointed out, preliminary experiments with single crystals of this alloy revealed that plastic strain at room temperature before transformation had a marked influence on the morphology produced during a burst. This effect is illustrated in Fig. 3, which shows two single crystals: one (left) was transformed in the unstrained condition whereas the other (right) was pulled in tension at room temperature to an extension of about 30% prior to transformation. The orientation of the strained crystal was such that slip occurred primarily on one $\{111\}$ plane. Over 50% of both crystals transformed in a single burst. The traces visible in Fig. 3 are those produced by the macroscopic shear incurred during transformation of austenite to martensite. The surface of the unstrained crystal exhibits traces due to many randomly orientated variants (24 possibilities) of the habit plane, whereas that of the strained crystal exhibits a very regular pattern composed of only a few variants.

Figure 4 shows resistance versus temperature curves for a pair of crystals obtained by splitting a crystal axially so as to ensure identical composition. One crystal had been extended 29% at room temperature before transformation, whereas the other was transformed in the unstrained condition. The M_s temperature was raised from -48.6 to -31.7°C due to the strain. The strained crystal transformed 69%* in the first burst, whereas the unstrained crystal transformed 85% in the first burst. No discernable transformation preceeded these large bursts. A difference in morphology similar to that shown in Fig. 3 was observed in the surface markings of these two specimens.

* The percentage of the specimen transformed was estimated by comparing the resistance drop due to the burst with the total resistance drop of the specimen after it had been cooled to liquid-nitrogen temperature. The comparison was made at the burst temperature. This, of course, involves the assumption that the specimen was transformed 100% at liquid-nitrogen temperature. For this reason the values should be used only for comparisons.

Figure 5 shows resistance versus temperature curves for a pair of crystals similar to those described in the preceding paragraph. In this case, however, the strained specimen was extended only 5.4%. The strained crystal transformed 84% in the first large burst at -48°C , but a very small burst, accompanied by an audible click, preceded the large burst. The unstrained crystal transformed the same amount, 84%, in the first burst at -49°C . The surface markings due to the transformation were similar on these two samples, except that the regions of the crystal showing the regularity produced by the operation of only a few permutations of the habit were distinctly larger in the crystal that had been strained 5.4%.

For comparison, two polycrystalline specimens (0.05-mm austenite grain size), which had been previously annealed in helium at 925°C for 5 hours and air cooled, were transformed. Although audible clicks were heard during the transformation, the resistance versus temperature curves showed no evidence of large bursts (Fig. 6).

A crystal whose orientation was such that two slip planes would be active during deformation was strained 6.4% prior to transformation, and then was prepared for two-surface analysis. Examination of the slip traces on the surface of the crystal indicated that the two slip planes were about equally active during the deformation. This crystal was cooled until a burst occurred; then it was immediately heated to room temperature (Fig. 7). The traces of the habit plane on the surface of the crystal (Figs. 8 and 9) revealed that the crystal could be divided into a number of regions, each of which had transformed through the operation of only a few nearly parallel variants of the habit plane. The results of a two-surface analysis of this crystal showed that each region was made up of four variants of the habit plane that group about a common $[110]$ direction. (Such a set of planes will be termed a "group.") Furthermore, only three different groups could be identified in the crystal, and most of the crystal was transformed through the operation of only two of these three groups. They alternate in position, CDCDCDCD, along the full length of the crystal (Fig. 8). The crystallographic

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Fig. 5. Resistance vs temperature curves for two single crystals of identical composition. One unstrained, the other strained 5.4% prior to transformation.

MUB-1312

Fig. 6. Resistance vs temperature curves for two polycrystalline specimens annealed 5 hours at 925°C in helium prior to transformation.

MUB-1313

Fig. 7. Resistance vs temperature curve for a crystal strained 6.4% at room temperature prior to transformation. The crystal was heated to room temperature immediately after the first burst.

ZN-3289

Fig. 8. Macrograph of a crystal that had been strained 6.4% at room temperature prior to transformation. 2X.

ZN-3295

Fig. 9. Macrophotograph of a crystal that had been strained 6.4% at room temperature prior to transformation. 4.5X.

position of the three groups, as well as that of the two slip planes and slip directions, are plotted on the stereogram of Fig. 10.

A second crystal, prepared as described above for two-surface analysis, was strained 32% at room temperature. The orientation of this crystal was such that slip during deformation occurred on two slip planes, and one plane had been much more active than the other. This crystal was cooled until a large burst occurred (Fig. 11); then it was immediately heated to room temperature. A single burst occurred at -35°C , in which 75% of the crystal was transformed. Visual examination of the crystal after transformation revealed that it could be divided into five regions (Fig. 12), each of which contained four prominent variants of the habit plane. Two regions located at opposite ends of the crystal were transformed by the operation of the same set of four variants of the habit plane. The volume of each of these two regions accounted for about 35% of the total volume of the crystal (region A in Fig. 12). The other three regions, each of which accounted for about 10% of the volume of the crystal (regions B, C, and D in Fig. 12), were grouped in the center of the crystal. Macrophotographs showing the structure of this crystal are shown in Figs. 13 and 14 in order that they might be compared with those of the transformed unstrained crystal shown in Figs. 15 and 16.

The results of a two-surface analysis of the crystal shown in Fig. 12 are plotted on the stereographic projection given in Fig. 17. Each region consisted of four prominent variants of the habit plane grouped about a common $[110]$ direction. The morphology differed from that observed in the crystal strained 6.4% prior to transformation, in that the individual regions were much larger and, consequently, fewer regions formed in the crystal during the burst. The orientation of the groups involved in the transformation relative to the two active slip planes and slip direction was similar in both crystals (compare the relative pole positions in Figs. 10 and 17).

A third crystal prepared for two-surface analysis was strained 28% at room temperature prior to transformation. The orientation of this

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Fig. 10. Results of a two-surface analysis of a crystal that had been strained 6.4% prior to transformation. Groups A, C, and D correspond to the regions indicated in Fig. 8.

MUB-1314

Fig. 11. Resistance vs temperature curve for a crystal that had been strained 32% at room temperature prior to transformation. The crystal was heated to room temperature immediately after the first burst.

ZN-3286

Fig. 12. Macrograph of a crystal strained 32% at room temperature prior to transformation. Five separate regions, each with four variants of the habit plane, are delineated. 2.5X.

ZN-3290

Fig. 13. Macrograph of the surface of a crystal, that had been strained 32% at room temperature prior to transformation. 1.3X.

ZN-3296

Fig. 14. Macrophotograph of a crystal that had been strained 32% at room temperature prior to transformation. 4.5X.

ZN-3287

Fig. 15. Macrophotograph of the surface of a transformed unstrained crystal. The transformation curve is shown in Fig. 4. 3X.

ZN-3288

Fig. 16. Macrophotograph of the surface of a transformed unstrained crystal. The transformation curve is shown in Fig. 4. 8X.

MUB-1306

Fig. 17. Stereographic plot of habit planes identified in regions A, B, C, and D of a crystal (Fig. 12) strained 32% prior to transformation. These regions formed in a single burst (Fig. 11).

crystal was such that a large amount of duplex slip occurred during deformation and, in addition, a number of deformation bands involving a third slip plane were visible on the surface. The transformation curve is shown in Fig. 18. In this case 52% of the transformation occurred in a single burst and the large burst was preceded by several very small bursts. Only one prominent region was identified in this crystal; the associated variants of the habit plane identified in this region are plotted in the stereographic projection shown in Fig. 19.

The orientation of the groups plotted in Figs. 10, 17, and 19 have a number of features in common. Specifically, in each crystal there is a group that contains four variants of the habit plane that are nearly perpendicular to two active slip planes. In addition, all other groups, with one exception, are nearly perpendicular to one of the active slip planes, and in no case was a group formed whose poles clustered about an active [110] slip direction. In all cases, the size of the burst, as measured by the percentage drop in resistance, was reduced by prior plastic strain.

The observation that the groups formed perpendicular to an active plane was also evident in three other crystals that had been strained prior to transformation. Although these crystals had not been prepared for two-surface analysis, the relative orientation of the slip lines and the variants of the habit plane was apparent from their surface traces. The frequency of these observations is considerably higher than that expected by chance.

Two experiments to determine the effectiveness of the autocatalysis responsible for burst transformation were conducted. In the first, a single crystal 2-1/2 inches long was cooled slowly while a temperature gradient of about 10°C per inch was maintained along its length. The transformation (almost 80%) was uniform along the length of the crystal and occurred in a single burst, despite the fact that nearly all of the crystal was above the normal M_s temperature when the burst occurred. The traces of the habit plane on the surface of the cylindrical crystal were similar to that of the crystal on the left in Fig. 3; i. e., the morphology appeared random.

MUB-1315

Fig. 18. Resistance vs temperature curve for a crystal that had been strained 28% at room temperature prior to transformation. The crystal was heated to room temperature after first large burst. Subsequently, it was completely transformed.

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Fig. 19. Stereographic plot of the variants of the habit plane identified in a crystal strained 28% at room temperature prior to transformation. The region plotted was formed in a single burst (Fig. 18).

In the second experiment a crystal was split along its axis in the manner previously described. One crystal was transformed in the unstrained condition, whereas the other was compressed 30% locally near one end in order to provide nucleation above the M_s temperature of the remaining unstrained length (2-1/4 inches) of the crystal. Both crystals were cooled until a burst occurred; then were immediately heated to room temperature. The transformation curves are shown in Fig. 20. The crystal that had been strained locally transformed about 12°C higher than the unstrained crystal, and the size of the burst was considerably reduced. Examination of the surface of the two crystals revealed that the crystal that had been strained locally had transformed uniformly along its total length, but the density of martensite plates was not as high as that in the unstrained crystal. The morphology of the crystal strained locally was somewhat more regular than the unstrained crystal, but did not compare with the regularity observed in crystals strained uniformly prior to transformation.

The habit plane observed in this study was irrational and scattered about the $\{10\ 3\ 15\}$. For purposes of discussion it will be denoted as such.

MUB-1316

Fig. 20. Resistance vs temperature for two crystals of identical composition. One unstrained, the other strained 30% in compression locally near one end prior to transformation.

Discussion of Results

The distinctive geometric grouping of four nearly parallel variants of the habit plane that cooperate to transform large regions of a strained austenite single crystal in a burst suggests that the planes in such a group may mechanically catalyze transformation on one another. This autocatalytic effect is apparently highly restricted to four planes grouped about a common $[110]$ direction, and rarely extends outside the group to the other 20 potential variants of the habit plane. It confers a degree of regularity upon the transformation that has not been previously observed in unstrained austenite crystals.

In addition to the regularity of the morphology of transformed strained single crystals, the orientation relationship between the direction and plane of slip and the groups of four variants of the habit plane which cluster about common $[110]$ directions also has certain distinctive features. First, the groups tend to form that are nearly perpendicular to the slip plane, and the group that is nearly perpendicular to both slip planes is especially favored. This group was observed in all three strained crystals that were transformed (Figs. 10, 17, and 19). Second, those groups that are oriented such that the four variants of the habit plane are nearly perpendicular to a slip plane, and at the same time nearly perpendicular to the slip direction, are conspicuously absent in the three strained crystals that were transformed. These observations are part of a pattern which can be summarized as follows:

1. In strained crystals, the four variants of the habit plane that group about common $[110]$ directions catalyze one another and transform large regions of the crystal during a burst.
2. The higher the prior strain, the larger are the individual regions containing only four variants of the habit plane.
3. The higher the strain, the smaller the amount of transformation that occurs autocatalytically in a burst.
4. Certain groups are favored in the transformation of a strained crystal; these are the groups that are nearly perpendicular to the slip plane.

5. Of the groups that are nearly perpendicular to the slip plane, the one made up of four variants of the habit plane that are nearly perpendicular to the slip direction is restricted. These groups were never observed in strained crystals.

The first of these observations can be interpreted in terms of a cooperative shear, provided that the angles between the planes as well as the shear directions in each are known. The former are, of course, known; however, the shear direction in each must be arrived at indirectly. The $\{10\ 3\ 15\}$ habit has also been observed in two alloys similar to that used in this work. Greninger and Troiano² made precision measurements of the habit plane, orientation relations, and direction and magnitude of the macroscopic shear in an iron alloy containing 22% nickel and 0.8% carbon. The observed habit was $(10\ 3\ 15)$ with a macroscopic shear direction of $[\overline{0.7315},\ 0.3828,\ 0.5642]$.^{*} The theory of Wechsler et al.,⁶ as well as the prism-matching technique of Bilby and Frank,⁹ predicts a $(10\ 3\ 15)$ habit and a $[\overline{0.7660},\ 0.2400,\ 0.5964]$ shear direction. The latter shear direction differs from the experimental value by 8.6° ; however, because of the experimental difficulties involved in the measurement of the direction, the agreement is considered good.⁶ Machlin and Cohen²⁴ measured the macroscopic shear and habit plane for iron containing 30% nickel. The habit planes they observed scattered about the $(9\ 22\ 33)$, which is 3.3° from the $(3\ 10\ 15)$; the shear direction they measured was $[\overline{1}\ \overline{5}\ 6]$. They assumed the direction of motion to be in the (111) plane rather than the habit plane, and therefore the measured direction can be taken to be the projection of the true direction in this plane. Although there is some ambiguity in the Machlin-Cohen analysis, the directions they measured are consistent with the theoretical and experimental evidence quoted above,^{2, 6, 9} in that the directions group closely (within 10°) of a common $[110]$

* For convenience in computation, the direction is expressed as a unit vector having non-integral indices.

direction. In addition, Greninger and Troiano²⁷ determined that the habit plane for iron containing 32.5% nickel was $\{10\ 3\ 15\}$. Both theories^{6, 9} predict the same habit plane and a $[\overline{0.7660}, \overline{0.2400}, 0.5964]$ macroscopic shear direction. Based on the evidence quoted, we will use $(10\ 3\ 15)\ [\overline{0.7660}, \overline{0.2400}, 0.5964]$ as the habit plane and homogeneous shear direction in the following discussion and in the interpretation of the results reported here.

By considering the influence of a stress on the transformation, the autocatalytic effect observed in the burst transformation of the strained crystals used in this investigation can be rationalized. Patel and Cohen²⁸ pointed out that interaction between mechanical energy supplied by an applied stress and the chemical free energy of the reaction can influence the transformation significantly. They took the component of the applied stress acting parallel to the shear and dilatational displacements of the transformation, and algebraically added the mechanical work done by the load stress on the transformed region to the chemical free-energy change of the reaction. The total was used to calculate and check the alteration in the M_s temperature produced by an externally applied stress. In our case, however, no external stress was involved. The stress that must be considered is induced in the austenite crystal by the formation of a plate of martensite, rather than by an externally applied stress.

In an unstrained crystal, we need only consider the stress pattern introduced by the formation of a martensite plate. A strained crystal, on the other hand, already contains a complex residual stress pattern which, in the vicinity of dislocation pile-up or tangle, may be substantially above the macroscopically observed yield stress²⁹ (2.5×10^4 psi). We shall assume that dislocation tangles develop as a crystal is strained, and estimate that they give rise to local residual stresses on the order of 10^5 psi.^{29, 30, 31} It is these local stresses that are assumed to be responsible for the observed increase of M_s temperature due to prior plastic strain. The local residual stresses, though they may be substantial, have strains associated with them that are small (10^{-2}) when

compared with the strains introduced into the matrix by the formation of a martensite plate. The shear strains introduced by such a plate are about 0.2.^{6, 9, 28} We shall proceed, therefore, under the assumption that the residual stresses introduced by plastic strain prior to transformation are easily overwhelmed by the stress introduced by the formation of a martensite plate, and that it is this latter stress that produces the mechanical autocatalysis.

Let us select a potential habit plane at random, e. g., (10 3 15), and consider the shear stress induced on other potential habit planes in the surrounding crystal by the formation of a martensite plate on this plane. To aid in visualizing the 24 permutations of the {10 3 15} habit plane, together with the corresponding direction of homogeneous macroscopic shear, refer to Fig. 21. The macroscopic shear strain associated with the formation of a martensite plate on (10 3 15) will induce a large shear stress in the surrounding austenite. Those variants of the habit plane which lie either nearly parallel to or perpendicular to the plate will have the largest resolved shear stress induced upon them. If, for any of these planes, the direction of the macroscopic shear strain associated with transformation is close to that of the direction of the induced shear stress, the mechanical energy available can effectively raise the M_s temperature and induce spontaneous transformation on them. If we resolve the induced shear stresses, σ_{xy} , and σ_{yx} onto each potential habit plane in the direction of the macroscopic transformation shear, we obtain a shear stress on each plane:

$$\sigma_{xy} \cos \theta \cos \phi + \sigma_{yx} \cos \rho \cos \delta, \quad (1)$$

where

θ is the angle between the x plane on which the shear stress σ_{xy} acts and the plane of interest,

ϕ is the angle between the y direction in which σ_{xy} acts in the x plane and the direction of interest in the given plane,

ρ is the angle between the y plane on which σ_{yx} acts and the plane of interest, and

δ is the angle between the x direction in which σ_{yx} acts in the y plane and the direction of interest in the given plane.

Fig. 21. Schematic illustration of the 24 permutations of $\{10\ 3\ 15\}$ and the macroscopic shear direction associated with the martensite transformation on each.

For the case at hand, the x plane is the (10 3 15), the y direction is the [0.7660, 0.2400, $\overline{0.5964}$], the y plane is the (0.7660, 0.2400, $\overline{0.5964}$), and the x direction is the [10 3 15]. The given planes and directions are those of Table II.

Since σ_{xy} and σ_{yx} are equal, the variant of the habit plane oriented such that the factor*

$$\cos \theta \cos \phi + \cos \rho \cos \delta \quad (2)$$

is a maximum will be catalyzed most effectively by transformation on (10 3 15); i. e., these planes will be most closely "coupled" to the first.

Using the data of Table II, we find that two types of strong mechanical coupling are possible. One type occurs on planes nearly parallel to the particular variant of the habit plane on which transformation was initiated, and the other occurs on planes perpendicular to this plane. If the surfaces of unstrained and strained crystals that have been transformed in a burst are examined closely, we find, however, that parallel coupling is always more effective. That is, the structure of a strained crystal as well as that of an unstrained crystal is composed of regions in which a large number of nearly parallel plates are formed before the sequence is interrupted by the formation of a plate making large angles to the first. When this happens, a second region of nearly parallel plates is nucleated. (This will be termed secondary coupling.) In strained crystals this effect is particularly evident and single regions involving only parallel coupling have been observed to consume about 80% the total volume of the crystal. We are led therefore to the conclusion that, in spite of the high coupling factor to the perpendicular planes, the importance of perpendicular coupling in burst transformation in strained crystals is in some way limited.

If the physical situation is examined, we find that, while parallel plates may shear in sympathy with one another by translating the intervening block of austenite in the same direction, this is not physically possible for perpendicular plates. In order for effective "perpendicular

* This will be termed the coupling factor.

Table II. Tabulation of angles pertinent to the burst transformation initiated on the (10 3 15) plane.ⁱ

Plane	Macroscopic shear direction ⁱⁱ	θ	ϕ	$\cos \theta \cos \phi$	ρ	δ	$\cos \rho \cos \delta$	$(\cos \theta \cos \phi + \cos \rho \cos \delta)$
$\bar{1}\bar{5} \ 3 \ 10$	$\bar{c} \ b \ \bar{a}$	$91^{\circ} 32'$	$86^{\circ} 42'$	-0.0015	$174^{\circ} 6'$	$156^{\circ} 17'$	+0.9107	+0.9092
$\bar{1}\bar{5} \ 3 \ 10$	$\bar{c} \ b \ \bar{a}$	$88^{\circ} 28'$	$93^{\circ} 18'$	-0.0015	$5^{\circ} 54'$	$23^{\circ} 43'$	+0.9107	+0.9092
$15 \ 3 \ 10$	$c \ b \ \bar{a}$	$29^{\circ} 23'$	$13^{\circ} 47'$	+0.8462	$74^{\circ} 45'$	$105^{\circ} 15'$	-0.0692	+0.7770
$15 \ 3 \ 10$	$c \ b \ \bar{a}$	$22^{\circ} 19'$	$31^{\circ} 7'$	+0.7919	$70^{\circ} 1'$	$109^{\circ} 59'$	-0.1168	+0.6751
$\bar{1}\bar{5} \ 10 \ 3$	$\bar{c} \ \bar{a} \ b$	$102^{\circ} 58'$	$119^{\circ} 50'$	+0.1117	$126^{\circ} 32'$	$130^{\circ} 28'$	+0.3864	+0.4981
$3 \ \bar{1}\bar{5} \ 10$	$b \ \bar{c} \ \bar{a}$	$77^{\circ} 2'$	$60^{\circ} 10'$	+0.1117	$130^{\circ} 28'$	$126^{\circ} 32'$	+0.3864	+0.4981
$15 \ \bar{1}\bar{0} \ 3$	$c \ a \ b$	$60^{\circ} 24'$	$38^{\circ} 24'$	+0.3871	$66^{\circ} 27'$	$75^{\circ} 13'$	+0.1019	+0.4890
$3 \ 15 \ 10$	$b \ c \ \bar{a}$	$60^{\circ} 24'$	$38^{\circ} 24'$	+0.3871	$104^{\circ} 47'$	$113^{\circ} 33'$	+0.1019	+0.4890
$3 \ \bar{1}\bar{5} \ 10$	$\bar{b} \ \bar{c} \ \bar{a}$	$66^{\circ} 10'$	$82^{\circ} 32'$	+0.0525	$113^{\circ} 26'$	$149^{\circ} 5'$	+0.3411	+0.3936
$\bar{1}\bar{5} \ \bar{1}\bar{0} \ 3$	$\bar{c} \ a \ b$	$113^{\circ} 50'$	$97^{\circ} 28'$	+0.0525	$149^{\circ} 5'$	$113^{\circ} 26'$	+0.3411	+0.3936
$15 \ 10 \ 3$	$c \ \bar{a} \ b$	$47^{\circ} 39'$	$65^{\circ} 25'$	+0.2803	$48^{\circ} 32'$	$89^{\circ} 48'$	+0.0024	+0.2827
$3 \ 15 \ 10$	$\bar{b} \ c \ \bar{a}$	$47^{\circ} 39'$	$65^{\circ} 25'$	+0.2803	$90^{\circ} 12'$	$131^{\circ} 28'$	+0.0024	+0.2827
$\bar{1}\bar{0} \ 15 \ 3$	$a \ c \ b$	$90^{\circ} 52'$	$29^{\circ} 11'$	-0.0131	$108^{\circ} 40'$	$71^{\circ} 20'$	-0.1024	-0.1155
$10 \ \bar{1}\bar{5} \ 3$	$\bar{a} \ \bar{c} \ b$	$72^{\circ} 35'$	$125^{\circ} 56'$	-0.1757	$82^{\circ} 52'$	$135^{\circ} 34'$	-0.0888	-0.2645
$\bar{1}\bar{0} \ \bar{1}\bar{5} \ 3$	$a \ \bar{c} \ b$	$107^{\circ} 25'$	$54^{\circ} 4'$	-0.1757	$135^{\circ} 34'$	$82^{\circ} 52'$	-0.0888	-0.2645
$10 \ 15 \ 3$	$\bar{a} \ c \ b$	$55^{\circ} 20'$	$107^{\circ} 30'$	-0.1710	$58^{\circ} 47'$	$121^{\circ} 13'$	-0.2685	-0.4395
$3 \ \bar{1}\bar{0} \ 15$	$\bar{b} \ a \ c$	$47^{\circ} 39'$	$110^{\circ} 50'$	-0.2396	$119^{\circ} 41'$	$61^{\circ} 4'$	-0.2396	-0.4793
$3 \ 10 \ 15$	$b \ \bar{a} \ c$	$47^{\circ} 39'$	$110^{\circ} 50'$	-0.2396	$118^{\circ} 56'$	$60^{\circ} 19'$	-0.2396	-0.4793
$3 \ \bar{1}\bar{0} \ 15$	$b \ a \ c$	$119^{\circ} 36'$	$89^{\circ} 19'$	-0.0059	$138^{\circ} 18'$	$41^{\circ} 42'$	-0.5574	-0.5633
$3 \ 10 \ 15$	$\bar{b} \ \bar{a} \ c$	$31^{\circ} 26'$	$136^{\circ} 20'$	-0.6172	$103^{\circ} 26'$	$76^{\circ} 34'$	-0.0540	-0.6712
$\bar{1}\bar{0} \ 3 \ 15$	$a \ b \ c$	$66^{\circ} 21'$	$80^{\circ} 1'$	+0.0696	$150^{\circ} 20'$	$29^{\circ} 40'$	-0.7550	-0.6854
$\bar{1}\bar{0} \ \bar{3} \ 15$	$\bar{a} \ b \ c$	$69^{\circ} 41'$	$73^{\circ} 13'$	+0.1003	$161^{\circ} 27'$	$18^{\circ} 33'$	-0.8987	-0.7984
$10 \ \bar{3} \ 15$	$\bar{a} \ b \ c$	$18^{\circ} 54'$	$152^{\circ} 14'$	-0.8372	$96^{\circ} 18'$	$83^{\circ} 42'$	-0.0121	-0.8493

ⁱAngles, θ , between the (10 3 15) plane and indicated variants of {10 3 15}; angles, ϕ , between the [0.7660, 0.2400, 0.5964] direction in (10 3 15) and the appropriate permutation of this direction in the indicated {10 3 15} plane; angles, ρ , between the (0.7660, 0.2400, 0.5964) plane and the indicated variants of {10 3 15}; angles, δ , between the [10 3 15] direction in the (0.7660, 0.2400, 0.5964) and the appropriate permutation of [0.7660, 0.2400, 0.5964] in the indicated variant of {10 3 15}.

ⁱⁱThe shear directions have been abbreviated as follows: $a = 0.7660$, $b = 0.2400$, and $c = 0.5964$.

coupling" to occur, the second plate would have to shear through the first, thus producing a structure at the intersection that is neither austenitic or martensitic (refer to Fig. 22). Because of this interference, a new perpendicular plate must form on the side of the original plate. Since the stress field induced by the formation of a martensite plate will be disc-shaped, a variant of the habit plane that is nearly perpendicular to the original plate will have only a small portion of its area in the highly stressed region adjacent to the original plate, and will therefore be less effectively coupled to it (Fig. 22; mode depicted on right). On the other hand, those variants of the habit plane that are nearly parallel to the original plate will have a relatively large area in the stress field and, as a result, will be more effectively coupled to it.

We shall proceed therefore by examining parallel coupling in detail. Returning to the example of initial nucleation on the $(10\ 3\ 15)$ plane and referring to Table II, we find that the two planes nearly parallel to the $(10\ 3\ 15)$ that are most strongly stimulated by the shear stress induced by transformation on the $(10\ 3\ 15)$ are the $(15\ \bar{3}\ 10)$ and the $(15\ 3\ 10)$. These two planes have coupling factors of 0.7770 and 0.6751, respectively. Continuing with the same reasoning, a similar calculation shows that transformation induced on the $(15\ \bar{3}\ 10)$ and $(15\ 3\ 10)$ by the formation of the first plate on $(10\ 3\ 15)$ will induce transformation on the $(10\ \bar{3}\ 15)$ and on the original plane, $(10\ 3\ 15)$. It will aid at this point to refer to Figs. 23 and 24, which illustrate the relative shear on these four planes and their position on the stereogram. Both the $(10\ \bar{3}\ 15)$ and $(10\ 3\ 15)$ couple strongly (the factors are 0.7770 and 0.6751) with the $(15\ 3\ 10)$ and $(15\ \bar{3}\ 10)$, thus forming a cycle that involves only four nearly parallel habits that group about a common $[110]$ direction. We shall call this type of coupling "parallel coupling." A schematic illustration of parallel coupling is shown in Fig. 25. This type of coupling is observed to be the most effective in autocatalysis. A macrophotograph illustrating this type of coupling is shown in Fig. 26.

If perpendicular coupling were to occur in this case, a second group of four variants of the habit plane would be nucleated. These are

MU-27985

Fig. 22. Schematic illustration of two cooperative shears involving perpendicular coupling. The mode depicted on the left is restricted because of the interference of the first plate. The mode on the right is ineffective because the second plate lies predominantly outside the stress field of the first.

MU-27986

Fig. 23. Schematic illustration of a group of four planes on which cooperative shear occurs during the burst transformation of an austenite crystal.

MUB-1309

Fig. 24. Stereographic projection showing the macroscopic shear directions associated with four variants of the $\{10\ 3\ 15\}$ plane. The four variants whose great circles are depicted are the $(10\ \bar{3}\ 15)$, $(10\ 3\ 15)$, $(15\ \bar{3}\ 10)$, and $(15\ 3\ 10)$. The small circular dots, $\circ$ and $\ominus$, on the great circles are the associated macroscopic shear directions ($\ominus$ are negative). The poles of the planes as well as the shear directions group about common $\langle 110 \rangle$ directions.

MU-27987

Fig. 25. Schematic illustration of the cooperative shear involved in parallel coupling.

a

b

ZN-3291

Fig. 26. Surface tilts due to a burst transformation in a crystal strained 32% prior to transformation. 250X. (a) Illumination from right, (b) illumination from left. The traces are due to four variants of the habit plane whose poles group about a common $[110]$ direction.

the $(\bar{1}5 \bar{3} 10)$ and the $(\bar{1}5 3 10)$, which are stimulated by perpendicular coupling to the $(10 3 15)$; and the two other planes, the $(\bar{1}0 \bar{3} 15)$ and $(\bar{1}0 3 15)$, which are stimulated by perpendicular coupling to the $(15 \bar{3} 10)$ and the $(15 3 10)$. These four planes group about a common $[110]$ direction (Fig. 24) and are coupled amongst themselves by parallel coupling. It is seen therefore that perpendicular coupling would occur between pairs of groups whose common $\langle 110 \rangle$ directions lie in a $\{001\}$ plane.

The ineffectiveness of perpendicular coupling in strained crystals is further demonstrated by examining the morphology of the crystal that had been strained 6.4% prior to transformation. The traces of the habit plane on the surface of this crystal (Fig. 8) revealed that about 90% of its volume was transformed through the operation of two groups, C and D, which alternate, CDCDCDCD, eight times down the length of the crystal. The relative orientation of these two groups is shown in Fig. 10. The coupling between these two groups is not perpendicular coupling, but rather coupling to variants of the habit plane of intermediate orientation. This type of coupling to groups located adjacent to common $[111]$ directions will be termed "cross coupling." The coupling factors involved in this type of coupling are less than 0.5.

It is apparent that in strained crystals mechanical autocatalysis is by far most effective between variants of the habit that are stimulated by one another through parallel coupling. Observations of occasional coupling outside such a group in cases where either cross coupling or perpendicular coupling would involve groups perpendicular to a slip plane (the crystal strained 6.4%) indicates that cross coupling is more likely than perpendicular coupling.

We have yet to consider the influence that prior plastic strain has on the morphology of the burst transformation. As has been pointed out, plastic strain prior to transformation plays a striking role in burst transformation by enhancing nucleation on certain variants of the habit plane and/or inhibiting growth on others. The net effects that were observed were: (a) an increase in the M_s temperature, which undoubtedly is a nucleation effect, (b) a decrease in the total number of variants

of the habit plane that operated during the burst; and (c) a decrease in the amount of transformation in a burst--a phenomenon that may be associated with the growth process.

In order to completely rationalize these effects, it would be necessary to know the exact nature of the defect structure and the associated stress pattern introduced by the plastic deformation. Although we cannot hope to know the exact details of this defect structure, there are some general statements which can be made about it. In all face-centered cubic metals that have been studied, broad three-dimensional tangles of high dislocation density are formed at high strains.³² These three-dimensional tangles are made up predominantly of dislocations with the Burger's vector of the active slip system. It is clear therefore that the defect structure is not an isotropic one but rather highly anisotropic. Furthermore, the associated stress pattern will also be highly anisotropic. It is this substructure that we have assumed to be responsible for the increase in M_s temperature with plastic deformation. Thus we are forced to conclude that, because of the anisotropic nature of the substructure, the enhanced nucleation responsible for the increase in M_s temperature is also anisotropic. Because of the anisotropic nature of the enhanced nucleation, the formation of certain groups will be favored.

We must also consider that in strained crystals the growth of martensite plates might also be anisotropic. Studies of the morphology of martensite plates in polycrystalline austenitized iron-nickel alloys, as well as in steel, have revealed that decorated slip bands and grain boundaries are likely to be impermeable barriers to growing martensite plates. Twin interfaces and undecorated slip bands may be semipermeable to martensite and, as was observed in these experiments, are most easily penetrated at right angles.³³ Thus it is evident that the growth along those variants of the habit plane which lie most nearly perpendicular to the slip plane will be favored, whereas those planes nearly parallel to the slip plane will encounter maximum resistance by the high concentration of debris lying in the slip plane.

It is seen therefore that two factors can account for preferential transformation by certain groups in strained crystals. First, preferred nucleation may occur on certain variants of the habit plane due to the anisotropic substructure, and second, restriction of growth on others that lie nearly parallel to the active slip planes may occur due to the high concentration of debris in the slip plane.

Summary

The following picture of the burst transformation in iron containing 31.7% nickel emerges from the results obtained. The autocatalysis responsible for the burst transformation has its origin in a mutual mechanical stimulation between variants of the habit plane that are geometrically oriented such that the stress induced by transformation on one aids transformation on the others. In both strained and unstrained crystals, the most effective coupling is between groups of four nearly parallel variants of the habit plane whose poles cluster about common $\langle 110 \rangle$ directions. The average size of the regions transformed through the operation of such a group depends on the frequency of less effective secondary coupling (perpendicular or cross coupling) to other similar groups in the crystal. In the case of unstrained crystals, the number of potential groups to which secondary coupling can occur is high, and as a result the probability of nucleating a new group is also high. This leads to a microstructure made up of many different small regions, each of which is formed autocatalytically by means of parallel coupling. On the other hand, in deformed crystals, transformation by means of certain groups is favored because of the internal anisotropic substructure. The most favored groups are those whose planes are nearly perpendicular to the active slip planes, but these groups do not include the group whose planes are nearly perpendicular to the Burger's vector of the active slip system. The number of possibilities for secondary coupling in strained crystals is therefore reduced, and the favored groups are able to grow to a very large size before their growth is interrupted by secondary coupling. Thus, in a strained crystal the transformation proceeds by means of only a few of the most favored groups, and the morphology is greatly simplified.

Conclusions

1. The morphology produced by the burst transformation of single crystals of iron containing 31.7% nickel that have been plastically deformed at room temperature prior to transformation is markedly different from that produced in unstrained crystals.

2. Large regions of strained crystals transform in a burst through the operation of four variants of the habit plane whose poles group about a common $[110]$ direction. The autocatalysis in such a group is shown to be mechanical; that is, the macroscopic transformation shears mutually assist one another.

3. In the transformation of a strained crystal, the operation of certain groups is favored; these are the groups whose planes are nearly perpendicular to the active slip plane, but these groups do not include the group whose poles cluster about the Burger's vector of the active slip system. This behavior is taken to be a consequence of the anisotropic substructure introduced during deformation.

4. The larger the prior plastic deformation, the larger are the regions made up of a single group. This is a consequence of conclusion 3. That is, there are fewer favorable possibilities for secondary coupling; once a group is nucleated, it can grow large before it is interrupted by secondary coupling.

5. The higher the prior plastic strain, the smaller is the amount of transformation that occurs autocatalytically in a burst.

6. In strained crystals, the effectiveness of the three types of coupling that were discussed is as follows, in order of decreasing effectiveness:

- i) parallel coupling within a group,
- ii) cross coupling between groups, and
- iii) perpendicular coupling between groups.

7. In unstrained crystals, parallel coupling is most effective in autocatalysis. The relative effectiveness of cross and perpendicular coupling between groups is not clear.

8. Once the transformation has been nucleated, mechanical autocatalysis is sufficiently strong to enable the transformation to proceed into material which is above its M_s temperature.

Proposed Future Work

It may be possible to extend the results obtained in this investigation to elucidate the process responsible for the mechanical stabilization of austenite. It was observed that in deformed crystals, transformation on certain variants of the habit plane, geometrically related to the Burger's vector of the active slip systems, did not occur. If this was due to a restriction of growth on these variants of the habit plane, then a deformation of an austenite crystal that activated all possible slip systems should restrict the transformation on all variants of the habit plane and lead to mechanical stabilization. Experiments to clarify this point have been initiated.

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