Recent Research on the γ - α Transformation in Alloy Steels*

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I. Introduction

The $\gamma \rightarrow \alpha$ transformation in steels is perhaps the most intensively studied phenomenon in physical metallurgy, and in more practical terms it occurs in millions of tons of steel each week of the year. It controls the microstructures obtained in the majority of steels and through these, is the main factor in reaching the required mechanical properties. There is therefore very good reason indeed for studying this transformation in great depth. The decomposition of austenite to ferrite and pearlite has received much attention with the result that we have a good basic understanding of these reactions. Likewise, the diffusionless transformation to martensite at lower temperatures has been very effectively explored, particularly in Japan, because of its importance in the achievement of high strength and toughness by quenching and tempering.

There is however a third important aspect of the $\gamma \rightarrow \alpha$ transformation which concerns many alloy steels, both micro-alloyed and those with higher alloying concentrations. This applies to steels in which the presence of strong carbide-forming elements leads to the partial or complete elimination of cementite from the structure. This, of course, means that the familiar pearlite no longer dominates the microstructure. I shall be dealing primarily with this aspect of the $\gamma \rightarrow \alpha$ transformation, showing how the microstructure is modified as the formation of pearlite is suppressed, and how alloy carbides then play a significant part in determining the mechanical properties achieved.

II. Alloy Carbides in Transformed Steels

In the tempering of quenched alloy steels, at lower temperatures (100~500°C), martensite decomposes to ferrite by the precipitation of iron carbides. However, above 500°C, cementite is often replaced by one or more alloy carbides *e.g.*, Mo_2C , VC, Cr_7C_3 , W_2C , which usually occur as a much finer dispersion (Photo. 1), both within the ferritic grains and at the grain boundaries.¹⁾ If similar steels are directly transformed from austenite in the same temperature range, *viz.* 550~800°C, alloy carbides are likewise formed in association with ferrite; however, there are very substantial differences in microstructure arising from the basically different heat treatment routes.

The salient feature that allows strengthening of carbon steels is that carbon is more soluble in austenite than in ferrite, consequently leading to cementite pre-

cipitation during the $\gamma \rightarrow \alpha$ transformation. Similarly, all the familiar alloy carbides are much more soluble in austenite than in ferrite²⁾ (Fig. 1), and thus tend to precipitate during the phase transformation or shortly afterwards. We can distinguish three situations which arise during the direct transformation of many alloy steels:

1) Precipitation at the γ/α boundary

This is readily recognized by interrupting the transformation by quenching, and then by demonstrating that the alloy carbides grow in association with the γ/α interface.³⁾ At least two basically different morphologies have been found in these circumstances, *viz.*, interphase precipitation (Photo. 2) and growth of fibrous carbide (Photo. 3).

2) Precipitation from supersaturated ferrite

In many cases, for example, when the transformation takes place rapidly as in micro-alloyed steels, ferrite can form without rejecting carbon as alloy carbides at the interfaces. However, if held longer at the transformation temperature, the carbides are precipitated from supersaturated solid solution, and tend to nucleate within the ferrite grains on dislocations and other structural imperfections⁴⁾ (Photo. 4).

3) Supersaturated ferrite

Clearly, the third possibility is that the ferrite is formed in circumstances where precipitation is difficult *e.g.*, at low transformation temperatures, or during rapid cooling. In these circumstances, the ferrite remains supersaturated to room temperature,

Photo. 1. A 6%W steel tempered 45 min at 700°C. W_2C precipitation in ferrite.

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Fig. 1. Solubility of alloy carbides and nitrides in austenite as a function of temperature (after Aronsson).

Photo. 2. An iron-4%Mo-0.2%C steel transformed 15 min at 700°C.
Coarse interphase precipitation of Mo_2C needles.
Thin foil EM. (Ricks)

only precipitating carbides on subsequent re-heating in the ferrite range.

All three situations are obtained during the heat treatment of alloy steels. We shall now examine some of the factors determining the final microstructure after transformation, and to do this effectively we have first to consider how ferrite grows from austenite.

III. The Growth of Ferrite in Austenite—Effect on Precipitate Morphology

Ferrite nucleates primarily at austenite grain boundaries but also at sites within the grains.⁵⁾ Many of the γ/α boundaries are planar (Photo. 5), particu-

Photo. 3. An iron-1V-0.4C steel transformed at 680°C. Region of fibrous vanadium carbide.
Extraction replica EM. (Parsons)

Photo. 4. A 0.26%V-0.9%Mn-0.05%C steel transformed 5 sec at 720°C. Widmanstätten plate with fine vanadium carbide precipitation on dislocations.
Bright field EM. (Balliger)

larly those where the ferrite is related crystallographically to the austenite in which it grows by the Kurdjumov-Sachs relation:

$$\{111\}_\gamma // \{011\}_\alpha \langle 101 \rangle_\gamma // \langle 111 \rangle_\alpha$$

The plane of the boundary is $\{111\}_\gamma // \{011\}_\alpha$ and represents a good-fit interface with a high degree of coherency. Work on model alloys in which both γ and α are retained to room temperature indicates that these interfaces contain no resolvable dislocations. The mobility of these boundaries is low, so the interface grows by the successive movement of steps along it. These have been observed in partially transformed, quenched out specimens⁶⁾ (Photo. 6), but also dynamically at the transformation temperature by use of photo-emission electron microscopy.⁷⁾

Precipitation of alloy carbides can occur during this stepwise movement of the γ/α boundary as shown in Photo. 7 of a high chromium steel in which M_{23}C_6 is precipitating along the newly formed interface, but not on the rapidly moving steps.⁸⁾ The result of such a mechanism is that the carbide is formed in sheets

Photo. 5. A 0.27%V-0.05%C steel transformed 10 min at 720°C. Mainly Widmanstätten ferrite. Optical micrograph. (Balliger)

Photo. 7. A 12%Cr-0.2%C steel isothermally transformed at 650°C. Cr_{23}C_6 precipitation associated with steps in the γ/α interface. Thin foil EM. (Campbell)

Photo. 6. Steps (arrowed) on a Widmanstätten ferrite / austenite interface in an iron-10%C alloy. (Howell)

Photo. 8. Fine interphase precipitation of VC in a 1%V-0.2%C steel transformed 10 sec at 700°C. Thin foil EM, dark field. (Ricks)

on the γ/α interface, and the sheet spacing is equivalent to the height of the steps observed on the interface.

In the example chosen, the steps are large, about 1 000 Å ($\sim 0.1 \mu\text{m}$ in height), however in other steels, *e.g.*, vanadium steels, the step heights have been observed as small as 50 Å.⁹⁾ Photograph 8 shows a vanadium carbide dispersion formed during the repeated movement of 50 Å steps along the γ/α interface. It is for this reason that these planar carbide dispersions, now fairly familiar in alloy steels, have been called interphase precipitation.

Examination of partially transformed steels indicates that many γ/α interfaces are not planar. Instead they are curved high energy boundaries which do not move by step migration, but by continuous migration normal to the boundary. This is common at high transformation temperatures, where there is adequate thermal activation for this type of migration. Precipitation of carbides occurs at these interfaces during the transformation, and indeed the precipitates act as useful markers for the progressive

movement of the interface. One example is shown in Photo. 9, of a 12%Cr steel transformed at 625°C, where the Cr_{23}C_6 precipitate follows the intricate curved movement of the boundary.

Many new ferrite grain boundaries possess both planar and curved boundaries, (Photo. 10(a)) often associated with one ferrite crystal. In these circumstances, the precipitate morphology frequently changes from curved to planar sheets within a short distance, as exemplified by Photos. 10(b) and (c) of an Fe-Cu-Ni alloy in which copper precipitates at the γ/α interface during the transformation.¹⁰⁾ In Photo. 10(d) the ferrite was initially supersaturated, and the copper has precipitated subsequently on dislocations. The transition between the several types of precipitation can be clearly seen. The proportion of curved sheets of precipitate to planar sheets is determined by such factors as the temperature, and the transformation kinetics. In general, the curved sheets will be obtained at higher transformation temperatures, and in alloy

steels which tend to transform slowly, for example high chromium steels which can take many minutes to transform isothermally in the range 700~800°C. In contrast, micro-alloyed steels containing concentrations of niobium, titanium or vanadium below 0.1 wt.% transform rapidly, and exhibit a very high proportion of planar sheets of precipitate, formed as

Photo. 9. A 12%Cr-0.2C steel transformed at 625°C. Cr_{23}C_6 nucleated on an irregular γ/α interface. Thin foil EM. (Campbell)

small steps move along largely planar γ/α interfaces.

The nature of the γ/α interface also largely controls the morphology of carbide particles formed during the transformation. Banded interphase carbide sheets are frequently observed in the same ferrite grain as fibrous carbide aggregates as shown in Photo. 11, a low-alloy steel containing 0.26V, 0.9Mn and 0.05C. It is now known that the fibrous carbide occurs when the γ/α boundary changes from a planar semi-coherent interface to one which is curved and of high energy. This transition has been shown to occur in several different steels;^{8,11} Photo. 12 shows the transition in a 5%Cr steel as the γ/α boundary changes its orientation, and thus its atomic character and its energy. The growth of the ferrite/carbide aggregate in the fibrous regions is "pearlitic" in character, with the boundary migrating normal to the boundary plane, and the carbide fibres growing side by side with the ferrite matrix.¹² In contrast, the other section of the interphase boundary is planar with steps leading to the characteristic banded precipitation.

IV. Interaction of Precipitates and Deformation in HSLA Steels

The modern development of control rolled steels, in particular, high strength low alloy steels, has brought general appreciation that hot working in the austenitic

(a) 20 min at 710°C: optical micrograph
(c) 20 min at 710°C: planar interphase precipitation, thin foil EM
(b) 20 min at 720°C: curved interphase precipitation, thin foil EM
(d) 5 min at 700°C: precipitation from supersaturated ferrite, thin foil EM

Photo. 10. Precipitation of copper in an iron-2Cu-2Ni alloy. (Ricks)

(a) polygonal ferrite grain
(b) area (b). interphase precipitation of VC

(c) area (c). fibrous VC precipitation
Photo. 11.

A 0.26%V-0.9%Mn-0.05%C steel transformed 15 sec at 730°C. (Balliger)

Photo. 12. A 5%Cr-0.2%C steel transformed 30 min at 650°C. Change in precipitate morphology with change in orientation of γ/α boundary arrowed. Thin foil EM. (Campbell)

range has profound effects on the final structure of the steel.¹³⁾ These steels, which contain small concentrations of strong carbide forming elements such as niobium and vanadium, usually precipitate carbide dispersions in the austenite, as it cools during hot working. While this type of precipitation (in contrast to that occurring during the transformation) does not have a large direct effect on mechanical properties because the precipitation is relatively coarse, it does have a profound influence on the recrystallization of the hot worked austenite.

The carbides, typically NbC, TiC, V(CN), tend to precipitate on dislocation networks, and exert a strong pinning effect, which substantially reduces the tend-

Fig. 2. Effect of deformation temperature and initial grain size on the critical amount of deformation needed to recrystallize austenite in a niobium containing steel. (Tanaka *et al.*)

ency of the austenite to recrystallize. This is shown by the fact that the critical strain for nucleation is greatly increased. This point has been well demonstrated in the work of Tanaka and colleagues¹⁴⁾ as shown in Fig. 2, where the critical strain for recrystallization of a plain carbon steel at 1100°C is less than 20%, while in a niobium steel it rises to 40%. When the deformation temperature is dropped to 950°C, the critical strain rises to 80~100%. This means that when recrystallization finally occurs, the recrystallized grain size is much finer, partly as a result of a high concentration of nuclei and partly because of the impedance of boundary movements by the carbide dispersion.

The alloy carbides will have an increasing effect as the deformation temperature is lowered, because the decreasing solubility of the carbides in austenite ensures a higher density of precipitate, and therefore a more effective pinning role. At the same time, the thermal activation needed to re-arrange dislocation arrays as a pre-cursor to recrystallization, is decreasing. All these factors increase the difficulty of recrystallization, with the result that frequently the austenite is still in a heavily deformed condition when it reaches the γ/α transformation range. When transformation occurs, the resulting ferrite is very fine grained, nucleating not only at the austenite grain boundaries, but also extensively within the grains. As a result, control rolling of micro-alloyed steels leads to extremely fine ferrite grain sizes, often in the range 5 to 10 μm , which is the most important source of strength and toughness in these steels.

V. The Effect of Deformation on the γ/α Transformation

Little systematic basic work has been done on the effect of prior deformation on the γ/α transformation in alloy steels. The beneficial effects of ausforming are well known, but the mechanisms involved in achieving the high strengths and ductilities in complex alloy steels are not yet fully understood.

Recent work on a 1.5Ni-1Cr-0.25Mo-0.36C steel (AISI4557) involving systematic deformation in the metastable austenite bay at 550°C, followed by isothermal transformation in the range 600~700°C, has shown that the γ/α reaction is very substantially accelerated (Fig. 3).¹⁵⁾

Similar effects were obtained with a simple 5Cr-0.2C steel, and a 1V-0.2C steel. Microscopic examination confirmed that the deformation had increased both the rates of nucleation and growth of ferrite. Photograph 13(a) shows an optical micrograph of the Ni-Cr-Mo steel after 30% tensile deformation at 550°C followed by transformation at 650°C. The ferrite has nucleated not only at austenite grain boundaries but intragranularly as well, which is an important point in the achievement of finer grain sizes. Pearlitic areas are etched dark and often contain ferrite nuclei (Photo. 13(b)). Turning to a 1%V steel, the effect of as little as 8% deformation in tension on the transformation kinetics is shown in Fig. 4. It thus seems clear that the much heavier deformations of austenite which take place during

controlled rolling of micro-alloyed steels, must have a profound effect on the nucleation of ferrite, and must greatly accelerate the phase transformation. As this occurs partly as a result of an increase in the rate of nucleation of ferrite, it is clear that the ferrite grain size will be substantially refined.

These experiments have also shown that the alloy carbides precipitate in the austenite prior to transformation, and can be readily distinguished from carbides forming during or after the transformation. For example, in a vanadium steel, vanadium carbide is observed to nucleate on grain boundaries, and on dislocation networks in the austenite. This undoubtedly occurs during cooling in the γ range in the absence of deformation, (Photo. 14) but the imposition of plastic deformation leads to the formation of disloca-

Photo. 13 (a). Ni-Cr-Mo steel transformed at 650°C after deformation at 550°C showing intragranular nucleation of ferrite.
Optical micrograph
(b). Same steel and treatment as Fig. 16 (a).
Nucleation of ferrite in pearlitic area.
Thin foil EM. (Walker)

Fig. 3.
Effect of deformation at 550°C on the transformation kinetics of a 1.5Ni-1Cr-0.25Mo-0.36C steel at 650°C. (Walker)

Fig. 4. TTT diagram for 1%V-0.2C steel, showing the effect of 8% deformation at 875°C on the kinetics of the transformation. (Walker)

Photo. 14. 1%V-0.2C steel aged at 875°C prior to transformation at 700°C. Precipitation of VC in sub-boundaries in austenite. Thin foil EM, dark field. (Walker)

tion sub-boundaries, which are then pinned by carbide precipitation.

VI. The Strengthening Mechanisms in Micro-alloyed Steels

Having outlined the major structural changes which can occur before and during the γ/α transformation in alloy steels, I would like to summarize how strength is achieved in the familiar micro-alloyed steels. The diagram (Fig. 5) shows the various contributions to the yield strength in these steels:¹⁶⁾

- 1) solid solution hardening—mainly manganese and silicon
- 2) ferrite grain size refinement
- 3) dispersion strengthening due to alloy carbides.

The extent of solid solution hardening is determined by the composition, and is practically unaffected by heat treatment. The contribution to the yield stress from grain size refinement is according to the Hall-Petch relationship, and extrapolation to finer grain sizes (Fig. 6) shows clearly the high yield strength achieved by this means in the 5~10 μm ranges.¹⁷⁾ At the same time, the figure indicates that the presence of strong carbide formers, *e.g.*, Nb, raises the linear curve above the plot for a plain carbon steel. This rise in yield stress results from the precipitation of carbides in the ferrite, not the austenite. While there is much evidence for precipitation of carbides and

Fig. 5. Contributions to the yield strength of 0.15%V-0.2%Si-0.2%C-0.015%N steels with increasing manganese content. Normalized from 900°C. (Gladman *et al.*)

Fig. 6. The yield strength of carbon-manganese and carbon-manganese-niobium steels as a function of ferrite grain size. (Le Bon and Saint-Martin)

carbo-nitrides in austenite during controlled rolling, they are usually on too coarse a scale to appreciably alter properties by dispersion strengthening. Their role, as has been stated above, is essentially in grain

refinement by control of recrystallization and grain growth. The dispersions responsible for the observed strengthening increments arise either during the γ/α transformation, or very soon afterwards by precipitation from supersaturated ferrite. In many micro-alloyed steels, both types of dispersion occur simultaneously in the ferrite, but some degree of control is possible.⁴⁾ For example, steels transformed at the higher transformation temperatures are less likely to produce super-saturated ferrite, and thus precipitation occurs during the transformation mainly by interphase precipitation. Transformation in the lower temperature range, achieved by more rapid cooling, usually leads to supersaturated ferrite in which precipitation on dislocations quickly occurs.

It so happens that the structures transformed in the higher temperature range comprise largely equi-axed ferrite and those in the lower, mainly Widmanstätten ferrite. However it is not the form of the ferrite which determines the mode of precipitation, *e.g.*, interphase precipitation has been observed in association with Widmanstätten ferrite, while precipitation on dislocations can occur in equi-axed ferrite. Nevertheless, the equi-axed ferrite frequently results in sharp yield points in stress strain curves (often undesirable in practice), while the presence of a proportion of Widmanstätten ferrite eliminates the yield point, which is often a desirable technological objective. It seems clear that the more heavily dislocated micro-structure obtained in a steel transformed in the lower temperature range, not only has a predominantly different dispersion morphology, but also has a large effect on the stress-strain curve. However the relationship between mechanical properties and the micro-structure of micro-alloyed steels needs further investigation before it can be fully elucidated.

VII. Conclusion

The formation of ferrite from austenite is not a new subject, but it is obvious that it is not yet fully understood, particularly with reference to the effects of alloying elements. Modern techniques of electron microscopy have given much new information about significant microstructural changes; however, we have not yet completely explained the various mechanisms involved. I have tried in this lecture to outline some of the important phenomena, and in particular have stressed their relevance to the now widely used micro-

alloyed steels. However, similar phenomena occur in higher alloyed steels, and there is considerable advantage to be gained from a fuller understanding of their behaviour during direct transformation from the high temperature γ phase to the ferrite condition.

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