THE KINETICS OF RECRYSTALLIZATION AND GRAIN GROWTH IN WELD HEAT-AFFECTED ZONES

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ABSTRACT

The general kinetics of recrystallization and grain growth under actual welding conditions were characterized for a variety of Ni and Fe-alloys. With the Ni-alloys, recrystallization takes place in a deformed matrix through preferential nucleation and discontinuous growth mechanisms, which may lead to a non-uniform grain distribution. With the Fe-alloys, the final austenitic grain sizes are essentially controlled by the phase transformation on heating and are unaffected by initial grain size and degree of cold-work. The growth kinetics differ in nature between the transformable and non- transformable alloys, e.g. because with the former the austenite grain size distribution is essentially homogeneous. With both classes of alloys, variations in grain sizes along the weld length may occur, depending on the spatial isotherms distribution, existence of local heat-flow disturbances and the thermal conductivity of the weld metal.

INTRODUCTION

The basic recrystallization and growth kinetics are fairly well understood under a wide range of static and dynamic conditions found in material fabrication and processing. That is not the case with welding, due to complications arising from the transient nature of the thermal cycle and the associated microstructural gradients.

In a previous work, Ref. (1), it was found that the relationship between grain sizes and cooling times from 800 to 500 degC (CT8/5) were essentially equivalent for a range of HSLA materials, which was unexpected bearing in mind the wide differences in composition, degree of cold work, etc. A subsequent study, Ref. (2), showed that the grain growth rates could be successfully reduced by chemistry control, but indicated that the nucleation kinetics remained unchanged.

In the present work, the recrystallization and growth kinetics are re-assessed for an even broader range of materials, including some Ni alloys that do not transform in the solid state. Strictly reproducing the same experimental conditions as those in Refs. (1,2), the kinetics of recrystallization for the Ni-alloys were analyzed from the distinct viewpoints of nucleation and growth. The analysis was then repeated for the transformable materials and the main differences were pointed. Finally, the kinetics of growth (i.e. grain growth following recrystallization) are summarized for both classes of materials.

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RECRYSTALLIZATION WITH NON-TRANSFORMABLE ALLOYS

The dominant nucleation mechanisms can be summarized from experiments in which the initial grain size was varied but where the thermal cycle and the degree of previous deformation were maintained constant. Studies on heavily deformed pure Ni, Inconel alloy 600 and Inconel alloy X750 confirmed that successful nucleation was essentially interface-associated, particularly with the three classical sites

i. grain boundaries,
ii. second phases and
iii. twin boundaries,

either individually or in combination. A few representations are pictured in Figure 1. Other mechanisms may had been present, but those were apparently statistically insignificant and will therefore not be discussed in greater detail.

In any given transverse or longitudinal weld section, the temperature threshold for nucleation can be assessed from the position of the farthest successful nuclei with respect to the fusion boundary. So far, all data on a variety of alloys and CT8/5 ranging from 3 to 45 s revealed no clear effect of initial grain size over the temperature threshold. In practical terms, the conditions involved in the formation of individual successful nuclei appeared to be reproduced independently of initial grain size for each constant set of experimental conditions (considering the typical experimental errors). In spite of the apparent equivalence, it is important to consider that the overall nucleation rate is intensified with decreasing grain sizes, resulting in a smaller nuclei spacing. Smaller nuclei spacings limit the extension of growth by accelerating impingement, thence favoring a reduction in final grain size.

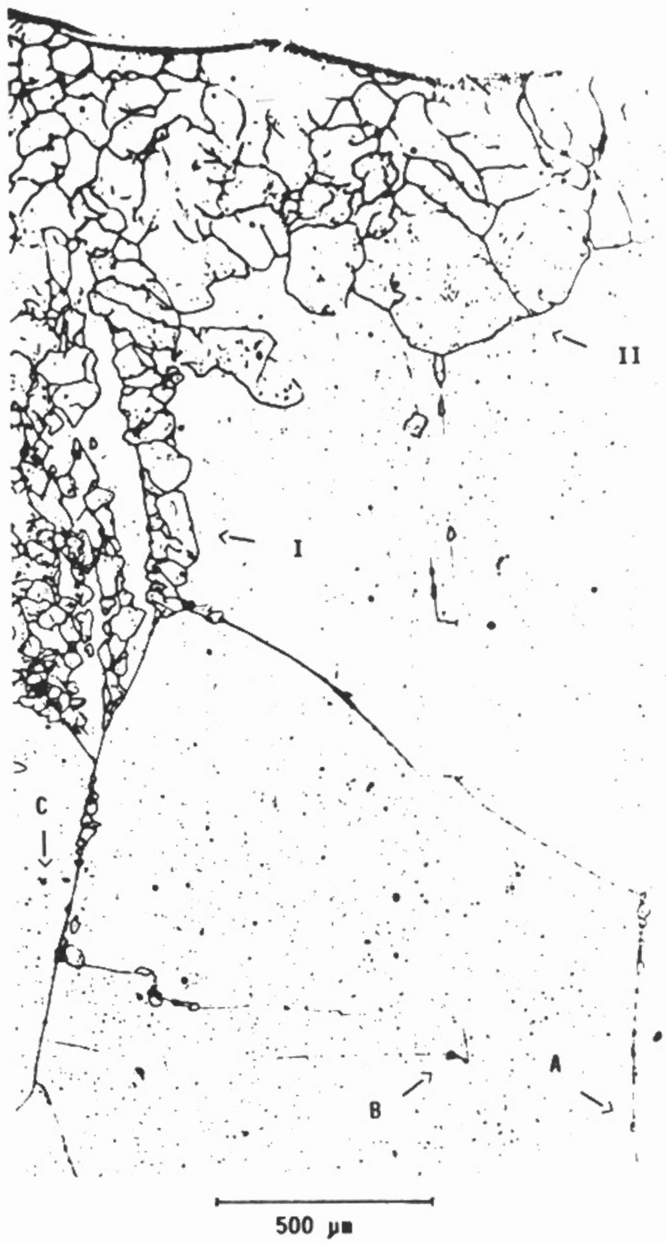

FIGURE 1: HAZ MICROSTRUCTURE WITH INCONEL ALLOY 600, 30% PREVIOUS DEFORMATION

NOTES : (1) Grain boundary, twin boundary and inclusion nucleations at **A**, **B** and **C**, respectively
(2) Observe limited growth along former grain boundary, Position **I**, in comparison with the more unrestricted inwards growth in Position **II**
(3) Fusion boundary is shown at top
(4) Oxalic acid electrolitic etch
(5) Experimental conditions as in Ref. (2), CT8/5 = 45 s

Migration follow nucleation and occurs in two stages. Initially, the arrays of successful nuclei situated along grain and twin boundaries grow until impingement. After impingement, growth is re-directed towards the inside of the deformed grain. Completion of this second stage would correspond to the completion of primary recrystallization. It is interesting to observe that the amounts of growth possible in each of the two stages are not correlated, so in principle the final distribution of recrystallized grain sizes are not uniform, as exemplified in Figure 1. The initial grain size has a very significant influence over the final grain size, whereupon a reduction in the former corresponds to a reduction in the latter.

The fact that nucleation, boundary-growth and grain-inwards growth are distinct and discontinuous processes imply that modeling of the final grain size may not be made on simple terms, e.g. as attempted in Ref. (3). The thermal cycle (or, more specifically, the time available above the temperature threshold) actually determines the stage of progress or completion of the sequence of processes. The HAZ positions where the time above the temperature threshold is insufficient to allow completion of the three stages will then have a bi-modal or even a multi-modal distribution consisting of original and recrystallized grains.

RECRYSTALLIZATION WITH TRANSFORMABLE ALLOYS

The solid-state transformation leads to an essentially different behavior in comparison to the non-transformable counterpart. Several situations may develop between the two extremes corresponding to recrystallization before phase transformation and vice-versa, as analyzed in Ref.(4). The study of the recrystallization kinetics with the Fe alloys is particularly interesting in view of the wide use, e.g. as structural materials. Welding always results in a coarse-grained region (CGR) adjacent to the fusion line, which is a matter for concern because of possible toughness impairment. Earlier works carried out in the 70's have proposed that the large grain sizes at the CGRs would be induced by an abnormal grain growth kinetics, triggered by some modest amounts of plastic strain generated during phase transformation, Ref. (5). Unfortunately, that is not consistent with the indications of uniform nucleation revealed by the uni-modal grain size distributions, so a re-assessment is required.

An analysis of the combined data from Refs. (1,2) reveals that the grain sizes in the CGR tend invariably to the 20-40 micrometer range with CT8/5 below 5 s. That occurred irrespectively of the wide differences in degrees of previous deformation, initial grain

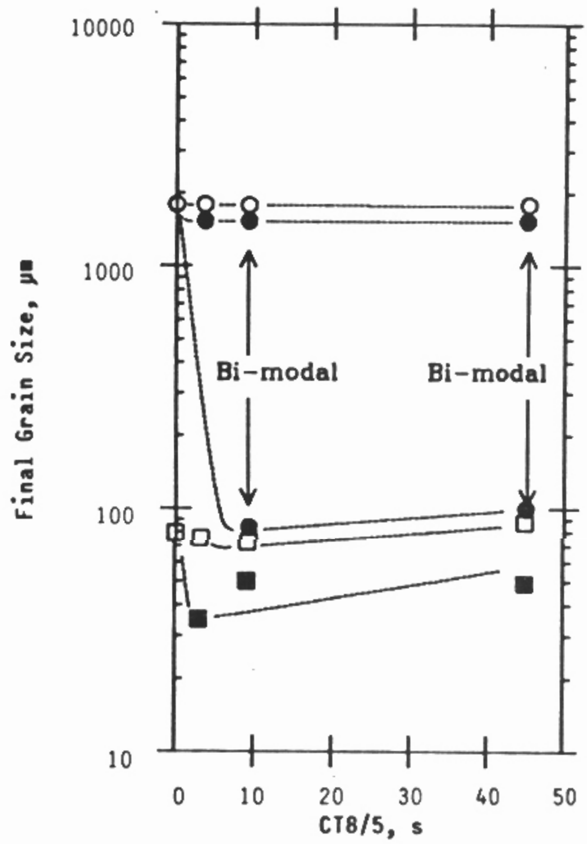

FIGURE 2: VARIATION IN FINAL GRAIN SIZES WITH CT8/5, INCONEL ALLOY 600

NOTES: (1) Legend: □ ○ Fully annealed
■ ● 30% strained (tensile)
(2) Experimental conditions as in Ref.(2)

distribution. The fact that these are all controlling factors for recrystallization induce to the consideration that the final grain size is essentially determined by the precipitation of austenite during heating and not austenite recrystallization. It is hoped that further information will be gained during the course of work being presently carried out, which will be the subject of future publication.

GRAIN GROWTH

Although similar in nature, the grain growth kinetics may differ between the transformable and the non-transformable materials for two reasons:

i. in principle, the transformable steels have uni-modal grain distributions during all stages of growth, which is not necessarily the case with the the non-transformable counterparts (see Figure 1) and

ii. with the transformable materials, growth is assumed to take place from inherently fine grains. That is not reproduced with the non-transformable materials, in which case the recrystallized grain size is not constant.

The relevance of a variable recrystallized grain size from which growth takes place with the non-transformable materials lay in the inverse dependency between growth rate and grain size, Ref. (4). The effect is summarized by the curves corresponding to the fully annealed materials in Figure 2, whereby noticeable growth takes place only with fine initial grain sizes. The insensitivity of the large-grained materials even with the slowest thermal cycles studied imply in the absence of an optically defined CGR, a fact often observed in practice, e.g. Ref.(6).

The inverse dependency between growth rate and grain size also imply that growth can be restrained via chemistry control with non-transformable materials only if the initial grain sizes are small. So far, the present experimental data for fine-grained materials in Figure 2 indicate that the ranking in growth rate amongst different alloys mirror that determined isothermally, i.e. decreasing speed in the sequence of commercially pure Ni, Inconel alloy 600 and finally Inconel alloy 750. Further work is being carried out to investigate for possible correlations between welding and isothermal annealing conditions.

The trends in growth with the transformable materials approaches that shown in Figure 2 for the fine grained, annealed material: a nearly-exponential increase in final grain size with CT8/5 above 5s, followed by an asymptotic decrease in final grain size to a constant value with CT8/5 below 5 s. Despite the evidences of unchanged nucleation kinetics , recent work demonstrated that the growth rate could be successfully reduced in Fe-alloys through chemistry adjustments. It is expected that further investigation may be carried out in the future to establish other possible routes for additional improvement.

Both transformable and non-transformable alloys have their growth kinetics significantly altered by some important heat-flow peculiarities. Three pool-associated phenomena could be listed as an example. First, the fusion boundary is characterized by a series of expansions and contractions along the weld length as a result of liquid convection (e.g. as noted in Ref. (7)). Second, the arc energy input is variable to some extent (deliberately as in pulsed welding or unintentionally as with unstable power supplies), eventually enhancing the expansions and contractions. Third, some processes such as submerged-arc produce two distinct and elapsed solidification fronts, leading to the formation of a clear concavity transverse to the weld centerline, Ref.(8).

The relevance of the expansions, contractions and concavities is that they alter the cooling rate at high temperatures (see Figure 3). For instance, concavities and contractions are associated with a thermal saturation effect, which slows down the thermal cycle and causes a tendency to grain coarsening (e.g. as in Ref. (8)). The opposite occurs at the expanded fusion boundaries and root regions, in which case the thermal cycles will be comparatively faster.

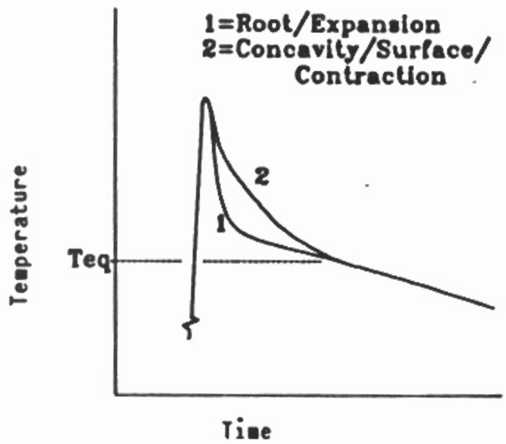

FIGURE 3: HEAT FLOW PECULIARITIES ASSOCIATED WITH BEAD GEOMETRY AND POOL PHENOMENA

NOTE: For example, Teq with a Cr/Mo filler is about 910 degC, which is substantially lower than the 1170 degC value obtained with a commercially pure Ni filler (under equivalent experimental conditions to those in Ref.(2))

The weld metal thermal diffusivity is a determining factor for grain growth in the CGR because it shapes the "high temperature region" of the thermal cycle. The effect of increasing thermal conductivity is similar to a change from curve 2 to curve 1 in Figure 3, i.e. decrease the dwell time at high temperatures. Increasing thermal conductivity also increases the equalization temperature (Teq in Figure 3), reducing the local grain size variations associated to contractions, expansions, etc.

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