COMPLETE REAUSTENITISATION IN MULTIRUN STEEL WELD DEPOSITS

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ABSTRACT

It appears that some of the scatter observed in toughness data from steel weld deposits may be a consequence of their inhomogeneous microstructure. This effect, and the need to produce welds of exceptionally high-strength levels, has stimulated the idea that by suitable alloying, multirun welds may be designed in which *all* regions of the weld exhibit a microstructure identical to that of an as-deposited weld. This concept is tested theoretically using a computer model designed to simulate and record the thermal cycles experienced by each point in the weld as a function of welding conditions, joint design and alloy chemistry. The results demonstrate that by increasing the stability of austenite with respect to ferrite, it is in principle possible to go a long way towards achieving a multirun weld with a homogeneous, high-strength microstructure.

INTRODUCTION

The microstructure which evolves as a weld pool solidifies and then cools to ambient temperature is often termed the *as-deposited* or *primary microstructure* of the weld. It may typically consist of a mixture of one or more of the phases allotriomorphic ferrite, Widmanstätten ferrite, acicular ferrite, martensite, retained austenite or degenerate-pearlite, contained within the columnar prior austenite grains typical of the solidification structure of low-alloy steel welds.

In multirun welds, the gap between the components to be joined is filled using a sequence of weld passes, each of which fills only a part of the weld gap. The metal deposited is therefore influenced significantly by the additional thermal cycles induced by the deposition of subsequent layers. Only the final layer to be deposited can then be expected to exhibit a true primary microstructure. The remaining regions of the welds may have undergone transient temperature rises high enough to cause partial or complete reverse transformation into austenite, which on subsequent cooling retransforms to ferrite, but not necessarily to the same microstructure as the primary regions. The regions which do not experience peak temperatures high enough to cause reversion to austenite, are tempered to an extent which is dependent on factors such as the starting microstructure and alloy chemistry.

Given that each regions of a multirun weld is likely to have experienced a different thermal history, the weld microstructure is expected to be inhomogeneous on a scale related to the dimensions of each weld pass, and on the detailed welding conditions (e.g. the heat input). This should necessarily lead to variations in mechanical properties, a feature which is most obviously reflected in the hardness profiles of such welds. There are indications that welds which are not mechanically homogeneous are susceptible to undesirable scatter in toughness [1,2,3], scatter which prevents the achievement of optimum properties and presents difficulties in engineering design. This phenomenon is well established in wrought steels where the microstructure may consist of a mixture of martensite and bainite, in which case the scatter in toughness is observed to be larger than in the corresponding uniform microstructures consisting of just martensite or bainite [4,5].

It can therefore be argued that it is desirable to produce mechanically homogeneous welds although it is difficult to see how this can be achieved for *multirun* welds. One possibility is to anneal the final weld to such an extent that the differences in strength between the variety of microstructures are reduced. Unfortunately, this has the obvious disadvantage that the general level of strength should also drop. The strength may be reduced to a level where only solid solution strengthening and the intrinsic strength of pure iron are the main contributors to strength, whereas the microstructural contribution (*e.g.* grain boundary and dislocation strengthening) is minimal by comparison.

This is in conflict with the increasing industrial demand for exceptionally high-strength, tough steel welds, particularly for use in the manufacture of submarines. There have been many attempts in the past to produce high-strength welds, but there is usually a penalty to be paid as far as toughness is concerned. Progress has been made in designing alloys which give an unusual primary weld microstructure which is essentially an approximately equal mixture of acicular ferrite and low carbon martensite [6]. Strength levels greater than 700 MPa can then be achieved. Whether these welds have reached their optimum level of toughness remains to be demonstrated, since in the multirun condition, they are to some extent mechanically inhomogeneous.

THE CONCEPT

It has been suggested that with these new range of alloys, it should be possible to design a multirun weld that has at all positions, a microstructure which differs little from the as-deposited microstructure. To achieve this, requires the simultaneous fulfilment of three main conditions. The first of these is that the Ae_3 temperature should be as low as is practicable, so that the effect of depositing a new layer is to reaustenitise as much of the adjacent underlying layer as possible. The second condition requires the hardenability of the alloy concerned to be high enough to lead to the retransformation of the reformed austenite to a microstructure resembling the as-deposited regions, *i.e.*

into the required mixture of just acicular ferrite and martensite. The low Ae_3 temperature also ensures that regions which do not reaustenitise are tempered to a lower peak temperature, thereby ensuring a lower loss in microstructural strength. The third condition requires that the alloy concerned should have a high tempering resistance.

THEORETICAL ANALYSIS

Analytical approaches to the modelling of heat flow in arc weld deposits are necessarily approximate, and a good way of verifying the general method used is to test its ability to predict the features of single bead-on-plate welds. The configuration tested is as shown in Figure 1. An approximation adopted in this work is to assume that the weld metal solidifies leaving a "spherical cap" on the surface of the plate. The cross-sectional area of this spherical cap is equivalent to the volume of consumable deposited per unit length in the welding direction. It is possible to identify Ac_1, Ac_3 and the solidus isotherms in the heat-affected zone (HAZ, which in the present context actually lies in the weld gap). The region between the Ac_3 and solidus isotherms is completely reaustenitised, whereas that between Ac_3 and Ac_1 isotherms only partially transforms to austenite during heating, and all regions heated to temperatures less than Ac_1 are tempered.

It is necessary to consider next a sequence in which a number of weld beads is deposited side-by-side in a single layer on a plate. The configuration is illustrated in Figures 2a & 3a. It

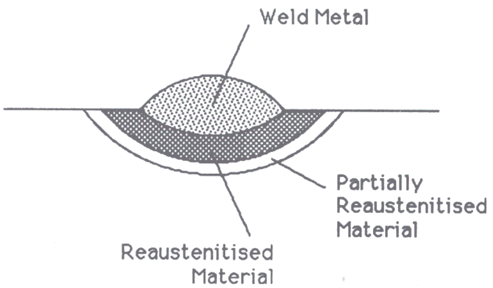

Figure 1: The single bead-on-plate weld.

is again possible to identify the relevant isotherms in the HAZ, but this time for simplicity the overlapping of isotherms from adjacent beads is ignored. The deposition of weld metal is now such as to raise the surface of the workpiece. In reality, after the deposition of these weld beads, the weld deposit is no longer flat, but it is nevertheless possible to define [7] a *reinforcement height* R_1 which corresponds to the "average" distance by which the height of the workpiece is raised.

Similarly, the *reaustenitisation distance* R_2 is defined as the distance between the fusion boundary and the Ac_3 isotherm, as measured beneath the weld centre-line. Again, in reality, there is a variation in this distance across the weld bead, and across the workpiece. When the second layer of weld beads is then deposited, assuming that the new weld beads are placed directly above the centrelines of the beads in the original layer, the condition for the heat input associated with the deposition of the second layer to completely reaustenitise the metal deposited in the first layer, is given by (Figures 2 & 3) $R_1 \leq R_2$.

In practice, the spacing between weld beads in a multirun weld tends to be non-uniform. The weld beads in any given layer are not deposited exactly above those in the previous layer, and after the deposition of the first layer the workpiece surface is no longer flat, but adopts the contours of the individual beads. This can be partly a consequence of the human error associated with the manual component of the welding processes. However, there is also the physical constraint that an integral number of beads must be deposited in any layer of the weld, the space available being determined by the weld geometry. In these circumstances, it is possible for unreaustenitised regions to exist even though the condition $R_1 \leq R_2$ is not broken. Therefore, a more general criterion which must be met if a fully reaustenitised weld is to be approached is that the ratio R_1/R_2 be minimised.

There are two ways that in which this problem can be tackled. The first approach involves the addition of those alloying elements to the welding consumable, which reduce the temperature at which austenite transforms to ferrite. The distance between the melting isotherm and the Ac_3 isotherm will then be larger. Nickel, manganese, chromium and copper all have the effect of lowering the Ae_3 temperature in steels, but it is necessary to consider carefully which additions might be the most appropriate; for example, the use of too much nickel or manganese can lead to increased segregation during solidification [8]. Similarly, carbon cannot be used in high concentrations if toughness and resistance to tempering is to be maintained.

The second approach involves controlling the rate at which the welding electrode (*i.e.* filler wire) is consumed. The requirement that the ratio R_1/R_2 be minimised can also be achieved by minimising the ratio of the rate at which the filler wire is consumed (the "burn-off rate") to the heat input. This has the effect of extending the isotherms to deeper positions below the fusion boundary. For processes such as submerged arc (SA), or tungsten inert gas (TIG) welding, it is possible to adjust the burn-off rate independently of the heat input. This is because the electrode itself is not consumed. However, in other processes such as manual metal-arc (MMA) welding, the filler wire itself acts as the source of heat, so that the burn-off rate and the heat input are no longer independent.

To summarise, the extent of material reaustenitised by the deposition of a new bead is likely to be maximised if a significant fraction of the arc power enters the workpiece, rather than being used in melting the filler material. These qualitative predictions are tested in the work presented below.

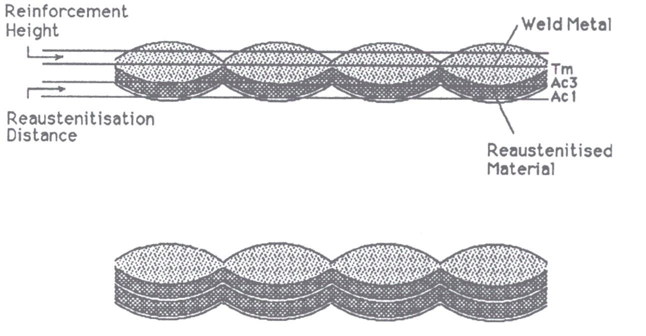

Figure 2:
a) Top, weld beads deposited side-by-side in a single layer on a plate.
b) Bottom, two layers deposited one above the other. Here, the reinforcement height is equal to the reaustenitisation distance.

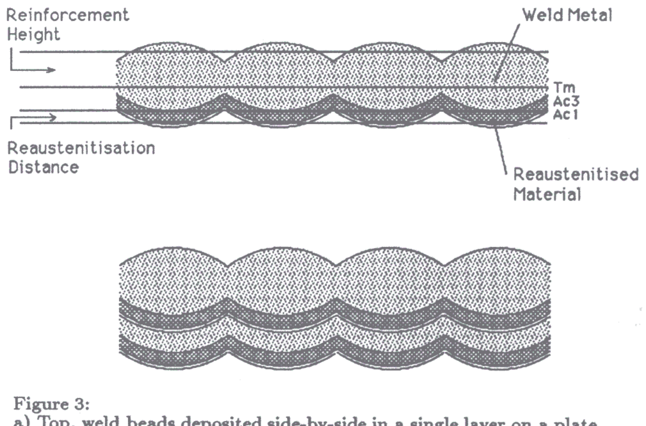

Figure 3:
a) Top, weld beads deposited side-by-side in a single layer on a plate.
b) Bottom, two layers deposited one above the other. Here, the reinforcement height is greater than the reaustenitisation distance.

COMPUTER MODEL

Since this work is a part of a project whose aim it is to provide facility for modelling the complete microstructure of multirun welds and their heat-affected zones, the method used is designed to be flexible in terms of joint geometry, welding procedures *etc.*, with the penalty that the computing times involved become rather large. The entire joint and parent plate is divided into a series of elements, each being a component of a multidimensional matrix. In physical terms, the weld (which includes a sufficiently large extent of the parent plate and backing plate) is divided into a large number of square elements, each of side $100\,\mu m$. As the weld gap is filled in a specified sequence, and under conditions determined by the program input, the thermal cycles associated with each element are recorded, as the beads are deposited, until the specified sequence is completed. The method for computing the thermal cycles is discussed below, but with this program, it is a relatively easy matter to quantitatively determine the volume fraction of all regions (either within the fusion zone or the parent plate or the entire assembly) which have experienced a specified sequence of thermal treatments.

The program requires an input of the weld variables: interpass temperature, electrode recovery (which can be greater than 100% for manual metal arc welding), heat input, arc efficiency, arc current, arc voltage, electrode diameter, arc travel speed and wire burn-off rate. Using these quantitites, it computes the data associated with the single bead-on-plate weld. Then, for the geometry specified by the user, the weld gap is filled by a series of further weld layers, each consisting of a number of weld beads.

The "building block" for the program is therefore, the single bead-on-plate weld. Appropriate geometrical modifications are made when the weld bead is deposited at the end of each layer, next to the near-vertical face of the plate being welded (due to space restrictions, the details of the algorithms used will be published elsewhere). The extended source heat-flow equations of Ashby & Easterling [9] are used, and weld metal is assumed to adopt a spherical cap geometry, although again appropriate modifications are necessary to allow for remelting of beads within each layer.

One option is for the user to specify the number of beads deposited in each layer. The program then assumes that these beads are to be spaced evenly within the layer. The other option requires the user must specify a *weld bead overlap*, following Alberry and Jones [7]. This specifies the % overlap of beads deposited next to each other, within the weld layer; the program then attempts to accommodate this value. However, some beads cannot be deposited at the the same distance apart because it is necessary to deposit an integral number of beads within each layer, and in those cases, the program fits an uneven spacing in the central region of the weld.

Once a layer of beads has been deposited, the program is designed to compute the amount of material deposited in that layer, and calculates the *reinforcement height* [7]. The next layer of beads is then deposited at this 'average' height above the previous layer.

CALIBRATION OF ASHBY'S EXTENDED HEAT SOURCE EQUATIONS: The equations presented by Ashby and Easterling [9] and later developed by Ion, Easterling and Ashby [10] represent the best analytical solution to the problem of the moving heat source first considered by Rosenthal [11]. The equations extend the power intensity of the heat source over a finite area, so that, the temperature predicted at the weld centre-line no longer rises to infinity.

However, it is necessary to decide by how much to extend the heat source. In Ashby and Easterling's original reference, they state that the radius of the circular disc source is likely to be proportional to the radius of the electrode itself. It is therefore appropriate to determine a suitable value for the the constant of proportionality relating these two quantities.

This has been done using the data presented by Clark [12]. This work is one of the best available sources of data for the manual metal arc single bead-on-plate weld, and one in which the weld bead dimensions have been carefully measured. Estimated equilibrium melting and Ae_3 temperatures (1520°C and 850°C respectively) were used to define the melting and Ac_3 isotherms for the mild steel baseplate used in Clark's work.

It was found that the best description of the single bead-on-plate weld occurred when the aforementioned proportionality factor was 1.6. The diameter of the heat source is then 1.6 times that of the appropriate electrode diameter. Values between 1.0 and 2.0 were tested against the experimental data. The results of the statistical analysis are presented in Table 1. It can be seen that the heat-flow equations overpredict the reaustenitisation distance by approximately 19%, and it was found that the magnitude of this overprediction was unchanged for all values of the proportionality factor considered.

The overprediction of the reaustenitisation distance can be attributed to the fact that reaustenitisation is unlikely to occur at the equilibrium Ae_3 temperature. This is due to the very rapid heating rates caused by the passage of the arc, and because the time available for transformation to austenite during the weld thermal cycle is exceedingly short. Both these factors act to raise the effective transformation temperature. The analysis was repeated for values of Ac_3 temperatures lying above 850°C and it was found that with Ac_3 equal to 910°C the regression coefficient for the calculation for the reaustenitisation was unity.

The above analysis indicates that the heat-flow equations predict the positions of the fusion isotherm and the Ac_3 isotherm to a satisfactory degree. In view of the inherent difficulty in quantifying manual processes, and the variability associated with the arc itself, it is felt that these equations represent a useful approximation to the problem of heat-flow in thick-plate (3D) welding.

	Correlation Coefficient	Regression Coefficient
Weld Bead Width	0.89	1.02
Weld Bead Penetration	0.67	1.03
Weld Bead Reaustenitisation Distance	0.79	1.19
Weld Bead Height	0.68	0.86
Weld Bead Area	0.93	1.07

Table 1: Results of statistical analysis on single bead-on-plate welds. Data is from Clark [12]. The diameter of the circular disc source is taken as 1.6 times that of the appropriate electrode diameter.

APPLICATION OF MODEL

The program outlined above, together with the quantities deduced from the analysis of experimental data, have been applied to a multipass weld consisting of 69 weld beads deposited in 12 layers. This is typical of welds deposited in the power generation industry. The $2\frac{1}{4}$CrMo weld metal was deposited using a manual metal arc technique in a V-notch weld geometry of $\frac{1}{2}$CrMoV steel. Suprex B electrodes of 5 mm diameter were employed; these gave an effective heat input of 0.987 kJ/mm (assuming an arc efficiency of 0.75) at an interpass temperature of 250°C. The weld is consistent with those used in the steam-pipes in modern steam-generating power plants. All of the essential weld variables are presented in Table 2.

A transverse section of the weld was cut, such that the surface itself was normal to the welding direction. The section was ground, polished and etched in 2% nital. A macrograph of the prepared surface is shown in Figure 4. Optical microscopy was used to confirm the number of weld beads deposited in each of the 12 layers.

Variable	Value
Interpass Temperature	250 °C
Recovery	113.7 %
Heat Input (Efficiency=0.75)	0.987 kJ/mm
Arc Current	22.0 A
Arc Voltage	231 V
Electrode Diameter	5 mm
Arc Travel Speed	3.76 mm/s
Wire Burn-Off Rate	3.20 mm/s

Table 2: The weld variables recorded for the multipass weld modelled.

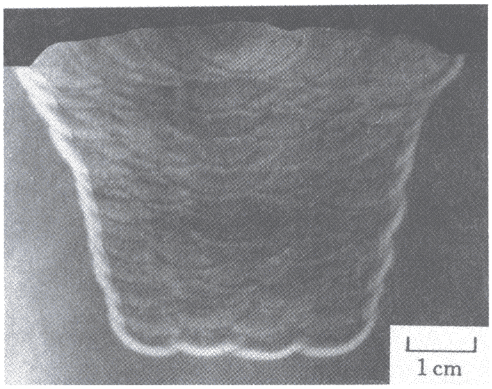

Figure 4: Optical macrograph of the multipass weld studied in this work.

The multipass weld has been modelled using the program described, and the data recorded, and tabulated in Table 2. The melting and Ae_3 temperatures of the weld metal were estimated [13] as 1516° C and 867° C respectively, using a method due to Kirkaldy, Thomson and Baganis. However, for the modelling it is necessary to know the Ac_3 temperature. In the absence of this data, a series of temperatures was assumed in the modelling (50° C intervals between 817 and 1017° C). This approach has the additional advantage of allowing the effect of the altering Ac_3 and Ac_1 transformation temperatures to be studied.

The output from the program is presented in Figure 5a. As the Ac_3 temperature is depressed, the volume fraction of unreaustenitised weld metal is decreased. The variation in calculated volume fraction of unreaustenitised weld metal with Ac_3 temperature is shown in Figure 6. The match between the experimental weld and the computed output, assuming that the etching contrast in the macrograph presented in Fig. 4 can be related directly to the temperature fields predicted by the computer, appears to occur at an Ac_3 temperature of 100 to 150° C above the calculated Ae_3 temperature of the weld metal. This difference is considered to be reasonable given the high heating rates associated with the process. A full confirmation of this is in hand via a detailed study of the reaustenitisation kinetics of weld metal. The results also confirm the qualitative prediction [6,14] that the amount of reaustenitised material should increase as the Ac_3 temperature is reduced; it further demonstrates that the relationship between the volume fraction of reaustenitised metal and the Ac_3 temperature is non-linear, with diminishing returns as the alloying element concentration is increased.

There is, however, the other proposed way in which it is possible to design a completely reaustenitised weld. This involved designing a consumable in which the amount of metal deposited per pass is low, but the heat input kept high.

In order to test this proposition, the modelling has been repeated, but this time the burn-off rate is decreased steadily from the measured value of 3.60 mm/s, to 2.06 mm/s. All other variables are held constant. Decreasing the burn-off rate has the effect of reducing the amount of metal deposited in each pass. Consequently, it is necessary to deposit more weld beads to create the weld. For a constant weld bead overlap of 35%, the program has been used to decide the number of beads deposited in the weld.

The output is shown in Figure 5b. The variation in calculated volume fraction of unreaustenitised weld metal with burn-off rate is shown in Figure 7. A smaller volume fraction of unreaustenitised weld metal is observed at low burn-off rates, again confirming the prediction discussed earlier.

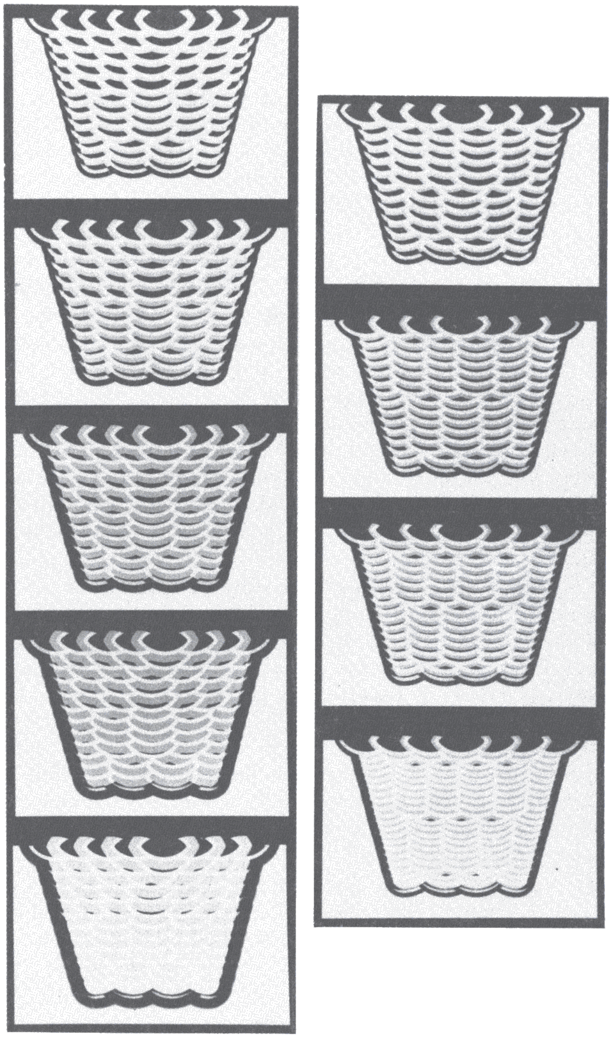

Figure 5:
a) Left, modelling the effect of lowering the Ac3 transformation temperature. From the top, Ac3 takes values of 1017, 967, 917, 867 and 817°C respectively.
b) Right, modelling the effect of lowering the lowering the burn-off rate. From the top, the burn-off rate takes values of 3.60, 2.92, 2.55 and 2.06 mm/s respectively.

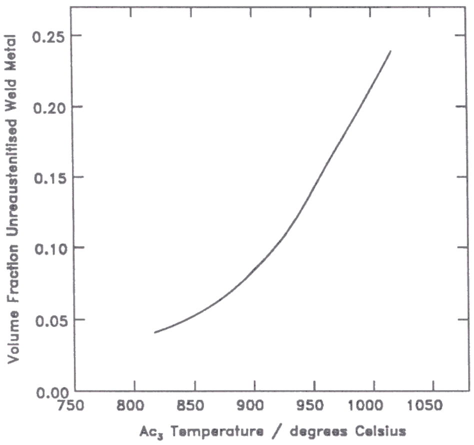

Figure 6: The variation in calculated volume fraction of unreaustenitised weld metal with Ac3 temperature.

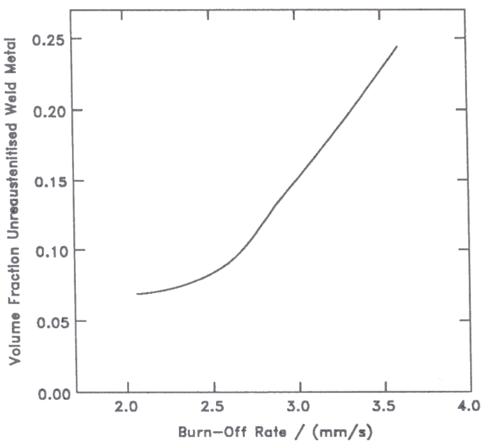

Figure 7: The variation in calculated volume fraction of unreaustenitised weld metal with the burn-off rate.

CONCLUSIONS

A computer model capable of estimating the temperature fields in a multirun weld deposit has been designed, in a way which permits the mapping of all regions which undergo a specific heat treatment. The idea that by suitable alloying, multirun welds may be designed in which the deposition of weld metal on to a previously deposited layer causes the latter to almost completely reaustenitise has been confirmed theoretically using this model. The results provide a basis for much further work towards the design of multirun weld deposits which are microstructurally and mechanically homogeneous.

Given that the basic concept of almost complete reaustenitisation has now been established theoretically, future work towards the design of mechanically homogeneous multirun welds will focus on ensuring that the reaustenitised regions transform back to a microstructure of acicular ferrite and low carbon martensite (*i.e.* similar to the primary regions). It is also vital to ensure that third layer effects can be minimised, so that any tempering beneath the reaustenitised regions has a minimal effect on mechanical properties of those regions. This in turn requires alloys designed for tempering resistance, and work is in progress on modelling such resistance [15].

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