Chapter 1

Weld Solidification*

JOINING METALLURGY has assumed an even greater role in the fabrication of metals within the last few decades, largely because of the development of new alloys with tremendously increased strength and toughness. Therefore, a working knowledge of metallurgy is essential to understanding current engineering structures and, in particular, the mechanisms that control weldment performance. Fundamental to joining metallurgy are the microstructures of a weld joint, which determine the mechanical properties, and welding variables such as weld thermal cycle, chemical reactions in the molten pool, alloying, flux composition, and contaminants, which significantly affect the weld and heat-affected zone (HAZ) microstructures.

Metallurgically, a fusion weld consists of three major zones, namely the fusion zone, the unmelted heat affected zone (HAZ) adjacent to the fusion zone, and the unaffected base metal as shown in Fig. 1. In alloys, there also is a fourth region surrounding the weld pool consisting of a partially melted or liquated zone, where the peak temperatures experienced by the weldment fall between the liquidus and the solidus.

It is well established that solidification behavior in the fusion zone controls the size and shape of grains, the extent of segregation, and the distribution of inclusions and defects such as porosity and hot cracks. Since the properties and integrity of the weld metal depend on the solidification behavior and the resulting microstructural characteristics, understanding weld pool solidification behavior is essential.

In this article, a general introduction is provided on key welding variables including solidification, microstructure, and causes and remedies of common welding flaws.

Solidification Behavior

The integrity and performance of a weld is, to a large extent, controlled by the solidification behavior of the weld metal or fusion zone. It controls weld-metal microstructure, grain structure, inclusion distribution, porosity, hot-cracking behavior, and, ultimately, weld-metal properties.

Current knowledge of weld pool solidification is an active area of research, as fundamental advances in solidification mechanics for castings and ingots have been applied to weldments (Ref 1-4). Fundamental solidification mechanics developed primarily for cast metals have been successfully applied to the solidification of welds. However, major differences do exist between casting and welding including:

- Dynamic nature of welding process
- Unknown pool shape
- Epitaxial growth
- Variations in temperature gradient and growth rates within the pool

In solidification mechanics, the important parameters that influence microstructure are temperature gradient ($G$), growth rate ($R$), undercooling ($\Delta T$), and alloy composition ($C_0$). In the weld pool, $G$, $R$, and $\Delta T$ vary considerably from region to region.

to region and process to process. Therefore, the microstructure that develops in the weld metal varies noticeably from region to region. Figure 2 shows schematically variations in weld-metal microstructure influenced by temperature gradient and growth rate.

Weld Grain Structure

The final weld-metal grain structure has considerable practical significance. It has been shown that a coarse columnar structure is susceptible to hot cracking, whereas a finer grain structure, a distorted columnar structure, or an equiaxed structure is more resistant to hot cracking. In addition, the fusion zone grain structure has a strong influence on the mechanical properties of the weld. To improve both the mechanical behavior and the hot cracking resistance of weld metal, efforts have concentrated on refining the fusion zone grain structure. Several means are available for refining the weld-metal grain structure. Since grain growth is epitaxial in nature, the size of the grains in the base metal controls the final grain size to a limited extent.

The weld pool shape also influences the fusion zone grain structure. For example, in an elliptically or circularly shaped weld puddle, not only does the magnitude of the maximum thermal gradient change continuously from the fusion line to the weld centerline but also the direction of this gradient changes. Since the average growth direction during solidification of a weld pool is approximately normal to the solid/liquid interface, which is also along the maximum temperature gradient, a given grain will not be favorably oriented during the entire solidification process since the direction of this gradient changes. Therefore, many of the grains at the fusion line that are initially of unfavorable orientations may become more favorably orientated
before they are completely eliminated and thus they may survive and continue to grow towards the centerline, resulting in a finer fusion zone grain structure. Such observations are typical in high energy beam welds or arc welds made at low welding speeds. For a tear drop shaped puddle, there is almost an invariant direction of maximum thermal gradient at all points on the pool edge from the fusion boundary to the weld centerline. This will promote large columnar grain growth as a result of the competitive grain selection process since only a few grains survive and grow uninterrupted toward the center. The result is a coarse columnar fusion zone structure.

Figure 3 shows general growth conditions in various solidification processes which all lead to columnar growth where the liquid temperature gradient $G > 0$. In welding, temperature gradients are typically of the order of $10^5$ K/m and growth velocities in the range of $10^{-3}$ to $10^{-1}$ m/s. With the introduction of laser welding techniques there is a trend to increase the welding speed which also increases the temperature gradient. As will be shown later, this trend may lead to problems in steel welding if the same alloys are used (Ref 4).

Weld-metal grain structure is predominantly controlled by the base-metal grain structure and the welding conditions. Initial growth occurs epitaxially at the partially melted grains. Both growth crystallography and thermal conditions can strongly influence the development of grain structure in the weld metal by favoring a strong grain-growth selection process. Growth crystallography will influence grain growth by favoring growth along the easy growth direction. For cubic metals, the easy growth directions are <100>. The conditions are even more favorable when one of the easy growth directions coincides with the heat flow direction in welds.

Figure 4 shows both epitaxial growth and a columnar grain structure in an iridium alloy weld. The preferred grain selection process that occurs during welding promotes a columnar grain structure. The weld-metal grain structure can be refined by making changes in the process variables and the use of external devices such as magnetic arc oscillator and vibrators. Although not done in practice, inoculants may be used as in castings to refine the weld-metal grain structure. Unlike in castings, the natural occurrence of a columnar-to-equiaxed transition in the grain structure of the weld is not very common. However, depending on the alloy system, such a transition has been observed and is described by H.W. Kerr (Ref 5).

An understanding of the development of grain structure in the fusion zone of polycrystalline welds is limited. This is because details of the growth selection process and the three-dimensional pool shape are obscured by the multitude of grains and crystal orientations that are present in polycrystalline materials.

Epitaxial Growth. The outstanding difference between the solidification of a casting and that of a weld (aside from the relative size and cooling rates) is the phenomenon of epitaxial growth in welds. In castings, formation of solid crystals from the melt requires heterogeneous nucleation of solid particles, principally on the mold walls, followed by grain growth.

For weld solidification, there is no nucleation barrier, and solidification occurs spontaneously by epitaxial growth. The nucleation event in welds is eliminated during the initial stages of solidification because of the mechanism of epi-
Introduction

Fig. 4 Epitaxial and columnar growth near fusion line in iridium alloy electron beam weld. Source: Ref 1

Epitaxial growth wherein atoms from the molten weld pool are rapidly deposited on preexisting lattice sites in the adjacent solid base metal.

As a result, the structure and crystallographic orientation of the HAZ grains at the weld interface continue into the weld fusion zone. In fact, the exact location of the weld interface is very difficult to determine in any weld deposited on pure metals using matching filler metal. Even microstructural features, such as annealing twins located in the HAZ weld joints, will continue to grow epitaxially into the weld during solidification.

Epitaxial growth is always the case during autogenous welding. For nonautogenous processes in which a filler metal is used, epitaxial growth may still occur. Nonmatching filler metals will also solidify epitaxially, particularly if the filler metal and base metal have the same crystal structure upon solidification, e.g., welding Monel (fcc) with nickel (fcc) filler metal.

However, the more classical case of heterogeneous nucleation also applies in this case. In welds, growth of the solid progresses from the unmelted grains in the base metal, and the stability of the solid/liquid interface is critical in determining the microstructural characteristics of the weld metal. The thermal conditions in the immediate vicinity of the interface determine whether the growth occurs by planar, cellular, or dendritic growth. Composition gradients and thermal gradients ahead of the interface are of primary importance. The effect of solute and thermal gradients on the solidification front can be described by the concept of constitutional supercooling criterion (Ref 6), which can be mathematically stated as:

\[ G_l/R > \frac{\Delta T_0}{D_L} \text{ for plane front solidification} \]  

or

\[ G_l/R < \frac{\Delta T_0}{D_L} \text{ for cellular or dendritic solidification} \]  

(Eq 2)

where \( G_l \) is the thermal gradient in the liquid at the solidification front, \( \Delta T_0 \) is the equilibrium solidification range at the alloy composition (Cn), \( D_L \) is the solute diffusion coefficient in liquid, and \( R \) is the solidification front growth rate.

Equations 1 and 2 define the conditions for the development of various solidification substructures in welds. This concept of constitutional supercooling is sufficient to understand the development of microstructures and the influence of process parameters on these microstructures in welds.

Weld Pool Shape

The macroscopic shape of the weld is an important factor that influences not only weld integrity but also grain size and the microstructure of the fusion zone. Therefore, it is critical to understand the dynamics of the weld pool development and its geometry.

The weld-pool geometry is determined mainly by the thermal conditions in and near the weld pool and the nature of the fluid flow.

In general, the weld pool volume is controlled to a large extent by the welding parameters. The volume is directly proportional to the arc current (heat input) and inversely proportional to the welding speed. In addition, the speed of the moving heat source has an influence on the overall bead shape. For arc welding processes, the puddle shape changes from elliptical to tear drop shaped as the welding speed increases. Correspondingly, the isotherms also change in shape. For high energy density processes such as elec-
tron beam or laser welding, the thermal gradients are steeper and as a result the puddles are circular at lower speeds, becoming more elongated and elliptical in shape as the welding speed increases before eventually becoming tear drop shaped at high speeds.

For example, if a single-phase metal is gas tungsten arc welded at a low velocity, the weld pool is elliptical (nearly circular), as shown in Fig. 5(a). The columnar grains grow in the direction of the thermal gradient produced by the moving heat source (arc). The grains grow epitaxially from the base metal toward the arc. Because the direction of maximum temperature gradient is constantly changing from approximately 90° to the weld interface at position A to nearly parallel to the weld axis at position B, the grains must grow from position A and continuously turn toward the position of the moving arc. The process of "competitive growth" provides a means whereby grains less favorably oriented for growth are pinched off or crowded out by grains better oriented for continued growth. The \(<001>\) and \(<00\overline{1}0>\) are the generally favored directions for crystal growth in cubic (fcc and bcc) and hexagonal (hcp) metals, respectively. In fcc metals, for example, the \(<001>\) most favored direction leads each solidifying grain because the four close-packed \(\{111\}\) planes symmetrically located around the \(<001>\) axes require the greatest time to solidify and, therefore, serve both to drag and guide the growth of solidifying grains.

The shape of the weld pool tends to become more elongated with increasing welding speed. In Fig. 5(b), the direction of maximum temperature gradient is perpendicular to the weld interface at positions A and B, but because the weld pool is trailing a greater distance behind the arc, the temperature gradient at position B is no longer strongly directed toward the electrode. Therefore, the columnar grains do not turn as much as in the case of a nearly circular weld pool.

Finally, the weld takes on a teardrop shape at the fast welding speeds that are usually encoun-
Solidification of dendrites in a weld. (a) Solidification of 3% Cu-Al alloy by the growth of dendrites, (b). (c) Solute redistribution occurring ahead of the solid/liquid interface. (d) Constitutional supercooling develops when the actual temperature of liquid in the copper-rich zone is greater than the liquidus temperature.

Fig. 6 Solidification of dendrites in a weld. The weld pool is elongated so far behind the welding arc that the directions of the maximum temperature gradient at position A and B in Fig. 5(c) have changed only slightly. As a result, the grains grow from the base metal and converge abruptly at the centerline of the weld with little change in direction. Welds that solidify in a teardrop shape have the poorest resistance to centerline hot cracking because low-melting impurities and other low-melting constituents tend to segregate at the centerline. Unfortunately, this solidification geometry occurs most frequently in commercial welding applications, because high heat input and fast travel speeds produce the most cost-effective method of welding.

Quantitative Modeling. Computational modeling is a powerful tool for understanding the development of weld-pool geometry. In recent years, significant advances have been made to model and understand the development of pool shape in welds (Ref 7–10). Some of these models can address coupled conduction and convection heat-transfer problems to predict weld-pool geometry and weld penetration. Of the various modes of heat transfer, convection plays a critical role in determining the weld penetration. Convection in the weld pool is driven mainly by buoyancy, electromagnetic, and surface tension forces. Significant work has concentrated on the convective heat transfer and, in particular, the effect of the spatial variation of surface tension (surface tension gradient) on fluid flow and weld penetration (Ref 11–14).

A fundamental limitation to a realistic prediction of the weld-pool shape is the lack of necessary thermophysical data. Unless a comprehensive data base on thermophysical properties and mechanical behavior of solids at very high temperatures (including the mushy zone) during solidification is established, use of computational models for a realistic prediction of microstructure, defects, and stresses will be stifled. In addition, verification of the models using critical experiments is essential.

Solute Segregation

Solute redistribution during weld-pool solidification is an important phenomenon that can significantly affect weldability and, in particular, hot-cracking behavior, weld microstructure, and properties. As alloys solidify, extensive solute redistribution occurs, resulting in segregation of the alloying elements that constitute the alloy.
Segregation on a small scale (few microns) is referred to as microsegregation, and large-scale segregation (hundreds of microns) is known as macrosegregation.

Macrosegregation can occur in welds, but it is not very common. In welds, it is primarily due to a sudden change in the welding parameters and the resultant change in growth velocity. A common manifestation of macrosegregation in welds is banding. Another form of macrosegregation in welds is the formation of a solute-enriched region along the centerline of the weld, which could promote the formation of centerline hot-cracking. However, evidence or proof of this phenomenon is lacking or not entirely clear.

Microsegregation is characterized by a compositional difference between the cores and peripheries of individual cells and dendrites. Cells are microscopic pencil-shaped protrusions of solid metal that freeze ahead of the solid-liquid interface in the weld. Dendrites are more developed than cells and appear to have a "tree-like" shape; the main stalk is called the "primary dendrite arm," and the orthogonal branches are called the "secondary dendrite arms." The cores of the cells and dendrite arms have a higher solidus temperature and contain less solute than the intercellular and interdendritic regions. In actual welding practice, cellular or dendritic microsegregation is virtually impossible to avoid unless the metal being welded is a pure element.

Generally, the important parameters controlling the cellular or dendritic substructures in welds are:

- The equilibrium partition ratio, \( K \), which is an index of the segregation potential of an alloy:

\[
K = \frac{C^*}{C^L} \tag{Eq 3}
\]

where \( C^* \) is the solute content of the solid at the solid-liquid interface and \( C^L \) is the solute content of the liquid at the solid-liquid interface;

- The alloy composition itself, \( C_0 \)

- The temperature gradient, \( G \), in the liquid at the weld interface, and

- The growth rate, \( R \), or velocity of the interface

Considering the solute redistribution in the interdendritic regions, also known as the mushy zone, it may be adequate to apply the solidification models for microsegregation originally formulated for castings. The solute profile in the solid is given by the Scheil equation (Ref 15):

\[
C^* = k C_0 \left( 1 - f \right)^{3/0.17} \tag{Eq 4}
\]

where \( C^* \) is the solid composition, \( C_0 \) is the initial alloy composition, \( F_s \) is the volume fraction solid, and \( k \) is the equilibrium partition coefficient.

This equation was modified by Brody and Flemings (Ref 16) by allowing for diffusion in the solid during solidification. Kurz and Clyne (Ref 17) have further refined Brody and Fleming’s model to account for extensive solid diffusion. These models can be easily adapted to analyze microsegregation in welds.

**Example: Solute Redistribution in the Dendrites of a 3%Cu-97%Al Composition.** Consider an alloy of composition \( C_0 \) equal to 3%Cu-97%Al in Fig. 6(a) so that per Eq 3:

\[
K = \frac{C^*}{C^L} = \frac{1.7\%}{10\%} = 0.17
\]

The first metal to solidify will contain only:

\[
C_0 K = (3 \times 0.17) = 0.51\% \text{ Cu}
\]

while the last liquid to solidify between cells or cellular dendrites is rich in copper:

\[
C_0/K = 3/0.17 = 17.6\% \text{ Cu}
\]

These values represent the short transients at the start and finish of solidification of a cell or cellular dendrite. As the cell or dendrite grows in the weld, a dynamic equilibrium is achieved between the newly forming solid of composition, \( C_0 = 3\% \text{ Cu} \), and the copper-rich liquid containing a maximum of \( C_0/K = 17.6\% \text{ Cu} \) at the solid/liquid interface as shown in Fig. 6(b) and (c). If the actual temperature distribution ahead of the solid/liquid interface is less than the liquidus temperature, constitutional supercooling occurs (Fig. 6d). Supercooling means that the solute-enriched liquid ahead of the solid-liquid interface has been cooled below its equilibrium freezing temperature, and constitutional indicates that the supercooling originated from an enrichment in composition rather than temperature.

Microsegregation results when the copper-rich liquid at the solid/liquid interface solidifies between the cellular dendrites. The interdendritic regions are so segregated with copper (solute) that a small amount of eutectic (α+δ) is frequently observed. Eutectic structures can only occur when the composition of solidifying metal exceeds the maximum solid solubility.

**Effect of Temperature Gradient.** Whether or not a planar, cellular, or dendritic substructure occurs upon solidification is largely determined by \( G \) and \( R \) (Fig. 7) which control the amount of constitutional supercooling. If a weld is deposited at a constant travel speed, \( R \) becomes fixed. By inducing an extremely steep temperature gradient \( G_1 \) (Fig. 8a), no constitutional supercooling occurs and the solidified weld-metal grain structure is planar.

When the gradient is decreased slightly to \( G_2 \) (Fig. 8b), any protuberance of solid metal on the interface will grow faster than the remaining flat
interface because the solid is growing into supercooled liquid; that is, the solid protuberance exists at a temperature below that of the liquidus for that alloy. As a result, a cellular substructure develops in each epitaxially grown grain. The liquid ahead and alongside each cell contains greater solute content than the cell core.

If the value of the temperature gradient is decreased further to \( G_3 \) (Fig. 8c), constitutional supercooling becomes so extensive that secondary arms form and cellular dendritic growth is observed. The greatest degree of microsegregation occurs during columnar dendritic solidification, while no measurable segregation is encountered in planar growth. Whether planar, cellular, or cellular dendritic, growth is always anisotropic.

Investigators have found that these solidification substructures can be characterized by the combined parameter \( G/R \). Figure 9 shows that a large value of \( G/R \) combined with a very dilute alloy will result in a planar solidification structure, while a low \( G/R \) and high solute concentration will produce a heavily segregated columnar dendritic structure. Both columnar dendritic and equiaxed dendritic structures, although common in large castings, are not frequently encountered in welds. In practice, cellular and cellular dendritic substructures are most frequently observed in welds. The difference between the cellular dendritic and columnar dendritic structures is related to the length of the constitutionally supercooled zone ahead of the solid-liquid interface. This zone is typically much smaller for cellular dendritic than for columnar dendritic solidification. Therefore, each grain will contain many cellular dendrites, whereas only one columnar dendrite occupies one grain. Unfortunately, it is very difficult to control \( G \) and \( R \) independently in welding practice. As a general rule, a fast welding speed \( (R) \) will produce a steep \( G \). The relative values of \( G \) and \( R \), however, determine the solidification morphology for a given alloy of fixed \( C_0 \) and \( K \).

Solidification Rate. While \( G/R \) controls the mode of solidification, the weld cooling rate, in terms of the parameter \( GR \) (solidification rate in units of \( °F/s \)), determines both the size and spacing of cells and dendrites. Flemings and others have demonstrated that the effect of solidification rate on the dendrite arm spacing \( (d) \) is:

\[
d = a(GR)^{-n}
\]  
(Eq 5)
where \( a \) is a constant and \( n \) is approximately \( \frac{1}{2} \) for primary arms and between \( \frac{1}{2} \) and \( \frac{1}{2} \) for secondary dendrite arms. The dendrite arm spacing of stainless steel in an electroslag weld is often several hundred times greater than that found in a rapidly cooled laser weld.

**Solute Banding.** The phenomenon of solute banding occurs to some degree in all alloy welds. The formation of ripples on the weld surface and solute banding within the weld are both caused by the discontinuous nature of weld-metal solidification and occurs in manual as well as in automatic welds where the travel speed is mechanically constant. During weld-metal solidification, however, \( R \) fluctuates cyclically above and below a mean value of growth rate that is determined by the weld travel speed. Fluctuations in \( R \) result in not only ripple formation, but also solute banding. Because an abrupt increase in \( R \) causes a reduction in the amount of solute that can be held in the solute-enriched liquid (Fig. 15c), excess solute is dumped and appears as a solute-rich band. Similarly, a sudden decrease in \( R \) produces a solute-poor band. Solute banding lines are very helpful in welding research because they always outline the weld-pool shape at a given instant during solidification. For example, the form factor (ratio of width to depth of weld pool), which is so important in electroslag welding (ESW), can be easily measured metallographically using solute band lines.

**Microstructure of the Weld and Heat-Affected Zone**

Through the process of epitaxial growth, the initial columnar grain width of the fusion zone is determined by the size of the base-metal grains adjacent to the weld interface. Because the peak HAZ temperature increases with decreasing distance from the weld interface and grain growth is a function of temperature, the maximum grain
size in the HAZ always occurs along the weld interface. It is this maximum grain size that is transmitted into the weld fusion zone.

**Grain Size.** The relationship used to calculate the grain size in the HAZ is:

\[ D - D_0 = be^{-Q/2RT}t'^a \]  

(Eq 6)

where \( D \) is final grain diameter; \( D_0 \) is the original grain diameter; \( e \) is the natural base for logarithms; \( T_P \) is the peak temperature, which would be the solidus temperature at the fusion line; \( t' \) is the time at temperature; \( Q \) is the activation energy for grain growth of the alloy; \( R \) is the universal gas constant; and \( b \) and \( a \) are constants determined by the materials. Both temperature and time at temperature produce grain growth in the HAZ, and as stated previously, the maximum grain size always occurs immediately adjacent to the weld interface. Because all welds experience the same spectrum of peak temperature from \( T_0 \) to the solidus temperature, the only significant variable in Eq 6 is the residence time, \( t' \). As the cooling rate decreases, residence time increases, substantially coarsening the maximum HAZ grain size.

The process of competitive grain growth may cause further lateral growth of the weld-metal grain size. Columnar grains emanating from the HAZ continue to widen as they grow into the weld fusion zone.

The maximum columnar grain width in the weld metal is limited only by the physical size of the weld bead and the arc energy input. For example, it is virtually impossible for a gas tungsten arc weld deposited on a coarse-grained copper casting to exhibit any distinguishable HAZ because the residence time is insufficient to cause noticeable grain growth; for example, \( D \) in Eq 6 is insignificantly greater than \( D_0 \). Furthermore, because the large base-metal grains that grow epitaxially into the weld must squeeze into a bead of limited volume, lateral growth of the columnar grains in the weld is not possible.

When cold-worked alloys are welded, the HAZ experiences both recrystallization and grain growth reactions. The hardness and strength properties of the recrystallized HAZ lose the benefits derived by cold working, and joint strength approaches that of an annealed alloy. Although weld-metal properties can always be controlled by judicious alloying, HAZ properties can only be controlled by regulating heat input or by changing the base-metal composition.

In steels and other metals that undergo allotropic phase transformation, the HAZ is conveniently divided into two regions:

- The grain growth region, which lies adjacent to the weld interface
- The grain-refined region, which is farther away from the weld interface

Because the grain growth region of the HAZ has experienced peak temperatures approaching the solidus of the base metal, coarse grains develop in accordance with Eq 6. The grain-refined region of the HAZ has been thermally cycled only briefly into the low-temperature portion of the austenite region, resulting in significant grain refinement. This grain-refining reaction occurs by the nucleation of new grains each time the \( A_1 \) and \( A_3 \) lines are crossed, either upon heating or weld cooling. The general structure of a steel weld will always appear fine-grained when compared to similar welds deposited on single-phase metals, such as pure nickel, copper, \( \alpha \) brass, and ferritic stainless steel.

The grain size distribution in precipitation-hardening alloys—which include maraging steels; precipitation-hardening stainless steels; 7xxx, 6xxx, and 7xxx series aluminum alloys; cobalt- or nickel-based superalloys; copper, titanium, and magnesium alloys; and many others—is generally similar to that of the single-phase alloys. The majority of precipitation-hardening alloys develop coarse grain structures in both the weld and HAZ, and the small amount of second-phase transformation is insufficient to produce any grain refinement. For example, welding and slow cooling a typical nickel-based superalloy containing small additions of titanium and aluminum result in a coarse-grained weld and HAZ structure with small amounts of Ni3(Al,Ti) phase along the \( \gamma \) (nickel solid solution) grain boundaries. If the weld cooling rate is fast, as in an electron beam weld, the Ni3(Al,Ti) does not form at all on cooling, but remains in a supersaturated solid solution. Subsequent aging treatments only precipitate Ni3(Al,Ti) as microscopic particles throughout the weld and the HAZ.

**Multiple-Pass Welds.** Grains of single-phase metals continue to grow without obstruction through each succeeding weld pass of multiple-pass welds, until all of the required passes are complete. Such interpass epitaxial growth leads to coarse columnar grain structures and extreme anisotropy of mechanical properties. Peening or cold working each weld pass prior to deposition of the subsequent pass helps mitigate the problem. The peening action sufficiently cold works the columnar grains of a freshly deposited pass to cause development of a refined or recrystallized grain structure in the new HAZ of this weld pass. Through epitaxial growth, these refined grains grow into the weld. By interpass peening, columnar grains are restricted to growth only within each weld pass, thus greatly reducing the overall grain size and anisotropy of multiple-pass welds of single-phase alloys. Peening is not recommended by most welding codes for the first and last (surface) passes because of the likelihood of fracturing the first pass and heavily distorting the surface of the last pass.