

# **Solidification of Metals**

**Class-7/5th SEM**

In systems that contain a eutectic reaction, the Scheil equation can be rearranged to allow the prediction of the fraction eutectic. This is done by setting  $C_s$  equal to  $C_{S_{max}}$ , the solid composition in equilibrium with the liquid eutectic composition,  $C_E$ . By setting  $(1-f_s)$  equal to the fraction eutectic,  $f_E$ , the equation can be rearranged in the following form:

$$f_E = \left( \frac{C_E}{C_0} \right)^{1/(k-1)} \quad (2.4)$$

This relationship can be useful in determining the fraction eutectic in eutectic systems where the solute level is less than  $C_{S_{max}}$ . Since the amount of eutectic constituent can influence solidification cracking resistance, this relationship can be used to predict the weldability of some alloy systems.

Although solid diffusion is not considered in Equation 2.3, it can, in fact, be considered under microscopic solidification conditions by the addition of an  $\alpha$ -factor to the equation, as shown in the following:

$$C_s = kC_0 \left[ 1 - \frac{f_s}{(1 + \alpha k)} \right]^{k-1} \quad (2.5)$$

where  $\alpha = 4D_s t_f / L^2$  and includes the diffusivity of solute in the solid ( $D_s$ ), the local solidification time ( $t_f$ ), and the spacing between the cell and dendrite cores ( $L$ ). When the value of  $\alpha$  exceeds approximately 0.1, diffusion can have a significant influence on solute redistribution since there will be back diffusion into the solid that will alter the solute profile and result in less solute partitioning to the final liquid. In practice, it is difficult to determine the  $\alpha$ -factor since values for  $D_s$  are not well known at elevated temperature and local solidification time must be estimated based on solidification rate. In systems that contain fast diffusing elements such as carbon and nitrogen, diffusion in the solid must be considered in order to accurately approximate the solute gradients and microstructure evolution during solidification.

### 2.3.5 Examples of Fusion Zone Microstructures

The fusion zone microstructure can vary widely based on alloy type and composition and welding process and conditions. This section is included to demonstrate the range of fusion zone microstructures that are observed in a number of common material systems, including steels, stainless steels, aluminum alloys, Ni-base alloys, and Ti alloys.

Most steels, including plain-carbon and low-alloy steels, solidify as bcc ferrite (delta ferrite) and transform to austenite almost immediately upon cooling below the solidification temperature range. The combination of solidification as ferrite and transformation to austenite on cooling tends to eliminate any evidence of the solidification substructure. In addition, these weld metals transform to lower-temperature products (ferrite, bainite, and martensite) upon cooling below the upper critical

(a)



(b)



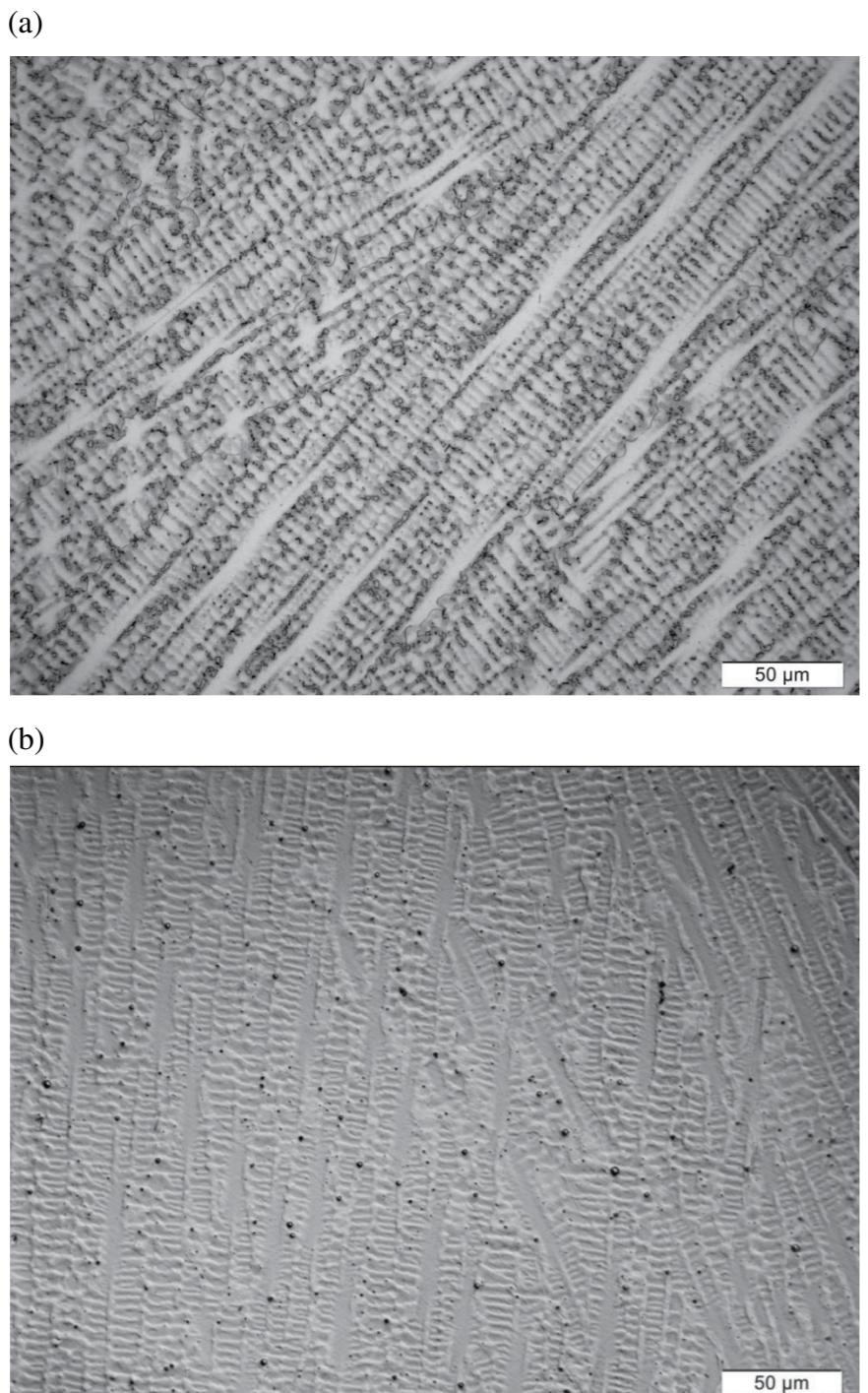
**FIGURE 2.30** Fusion zone microstructure of (a) plain-carbon steel, (b) low-alloy steel.

temperature ( $A_3$ ). The resultant fusion zone microstructures show evidence of a columnar solidification pattern, but SGBs and SSGBs (as shown in Fig. 2.24) are not observed. Examples of the fusion zone microstructure in two such steels are shown in Figure 2.30.

When steels solidify as fcc austenite, the solidification substructure becomes more apparent, as already shown in the austenitic stainless steel fusion zone in Figure 2.24. Other alloy systems, such as Ni-base, Cu-base, and Al-base alloys, also solidify as an fcc phase and exhibit distinct solidification substructure, as shown in Figure 2.31. This occurs since diffusion is relatively sluggish in the fcc phase relative to bcc,

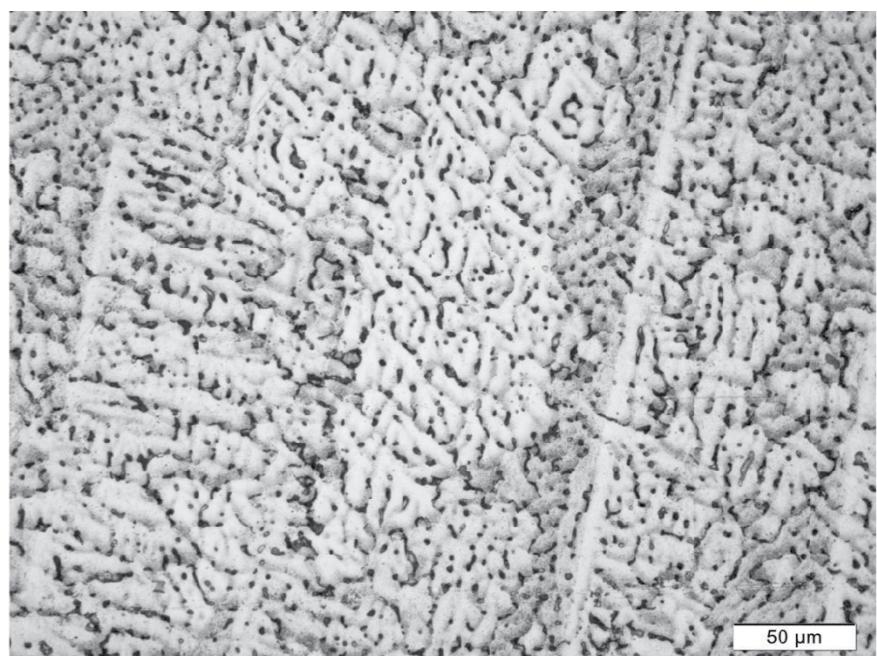
and the original solute segregation patterns established during solidification are preserved. Note that in some aluminum alloys a twinning phenomenon can occur following solidification, giving rise to what are described as “feather crystals” in the fusion zone microstructure.

Ti alloys solidify as a bcc beta phase and exhibit little or no evidence of solidification substructure. Since the beta phase is stable over a relatively wide temperature range prior to transformation to the hcp alpha phase (at the beta transus), solid-state diffusion is effective in eliminating solute segregation resulting from solidification. In addition, and as with plain-carbon and low-alloy steels, the transformation to

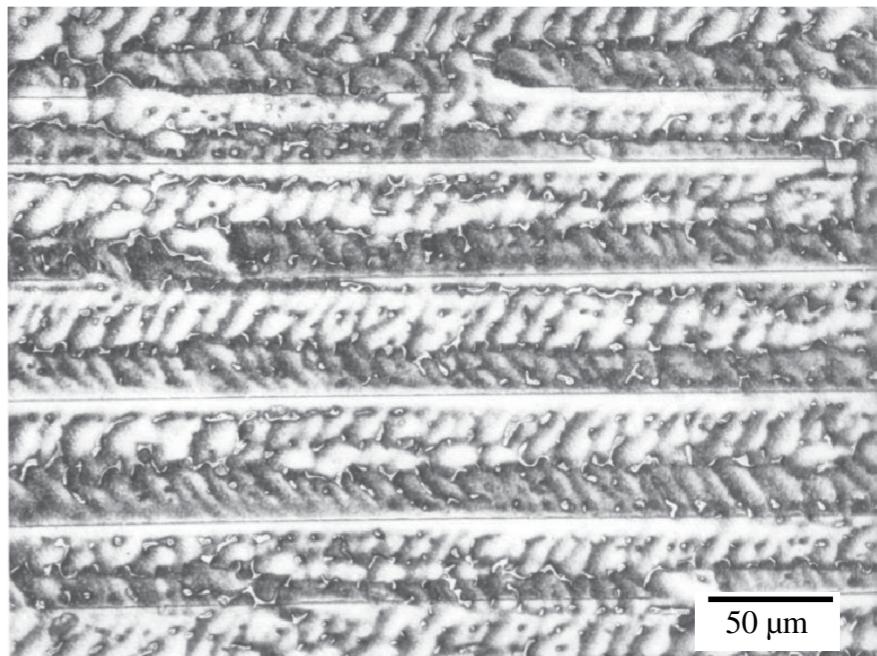


**FIGURE 2.31** Representative fusion zone microstructure of different alloy systems: (a) Ni base, (b) Cu base,

(c)



(d)

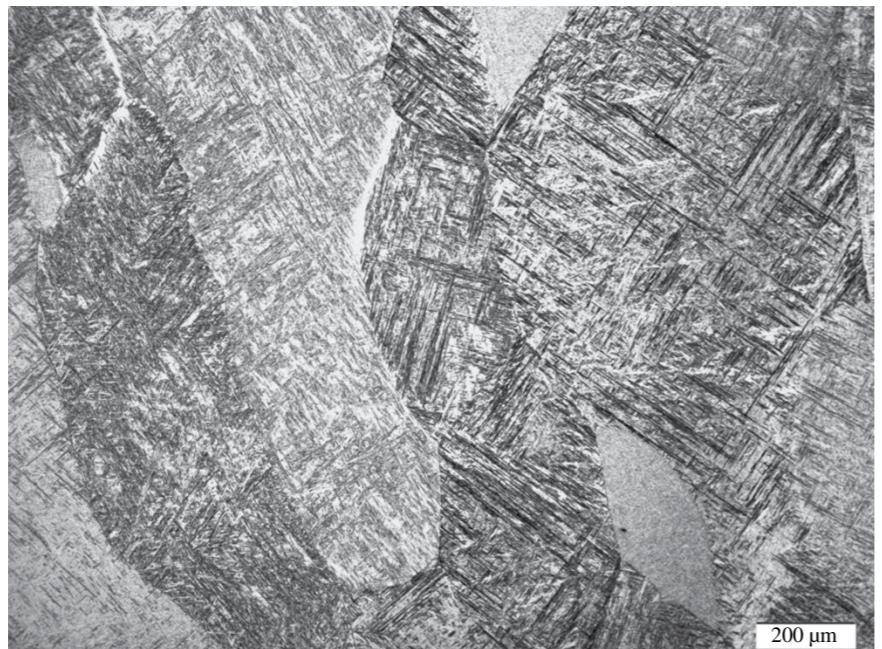


**FIGURE 2.31** (Continued) (c) Al base, and (d) Al base with twinned crystals.

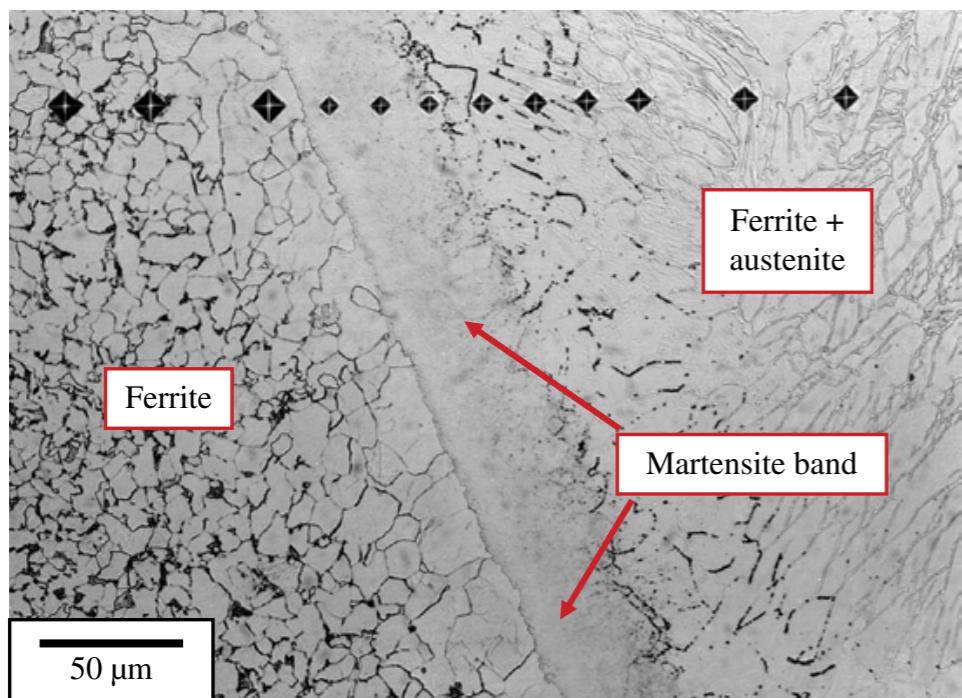
alpha below the beta transus tends to eliminate evidence of the preexisting solidification structure. An example of the fusion zone in Ti–6Al–4V is shown in Figure 2.32.

### 2.3.6 Transition Zone (TZ)

In heterogeneous welds, a composition TZ must exist between the fully mixed (diluted) weld metal and the base metal. If the composition difference between the base and filler metal is not large, this TZ may be undetectable, particularly if the TZ and composite fusion zone have the same microstructure. When the composition



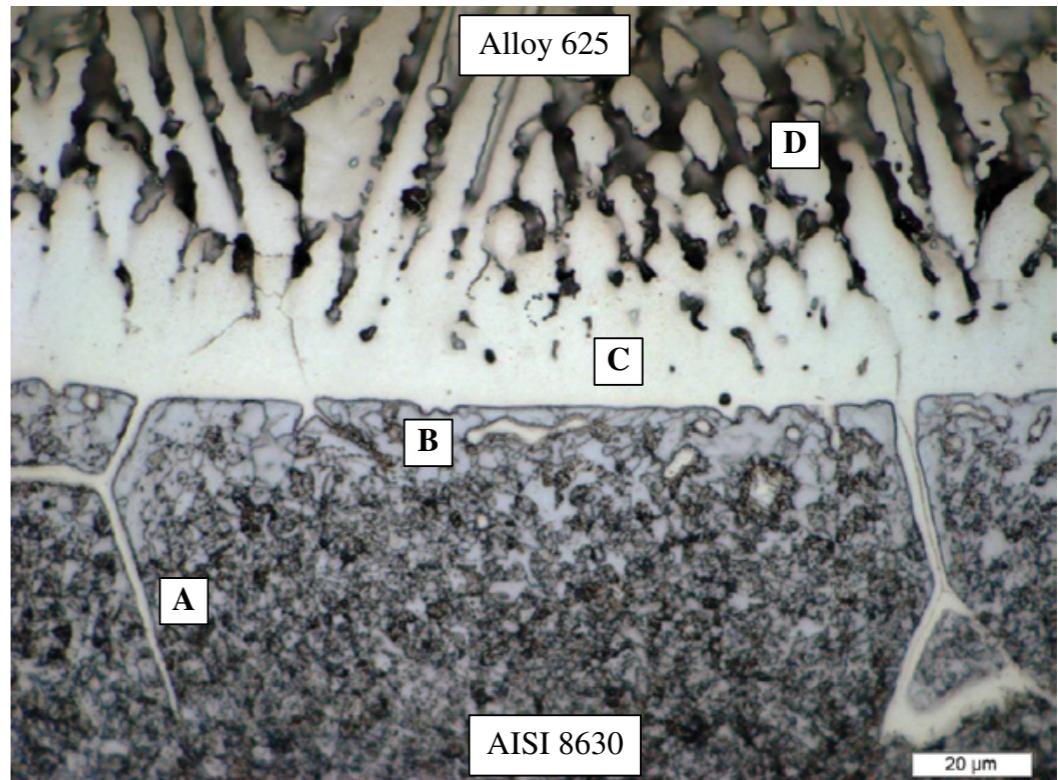
**FIGURE 2.32** Fusion zone microstructure of Ti-6Al-4V.



**FIGURE 2.33** Transition zone in carbon steel clad with Type 308L austenitic stainless steel.

difference is large, such as when austenitic stainless steel or Ni-base alloy cladding is applied to carbon steels, the TZ is very apparent and may exhibit different microstructure and properties relative to the base and filler metals. Two examples are provided here.

When carbon steels are clad with austenitic stainless steels (such as Type 308L) for corrosion protection, the TZ reaches a composition where the austenite that forms during solidification transforms to martensite on cooling to room temperature. This results in a narrow band of martensite close to the fusion boundary that exhibits



**FIGURE 2.34** Transition zone between AISI 8630 steel clad with Ni-base Alloy 625 after PWHT: (a) penetration of weld metal down the grain boundary, (b) carbon-depleted zone, (c) planar growth region, and (d) cellular growth region. (From Ref. [23]. © Springer)

as-welded hardness much higher than either the base metal or composite fusion zone, as shown in Figure 2.33. In some situations, this may require a PWHT to temper this hardened region.

When Ni-base alloys are used to join or clad steels, a similar TZ structure can form, as shown in Figure 2.34. In this case, an apparent planar growth region exists at the fusion boundary within the TZ. This planar growth region quickly breaks down into cellular and cellular dendritic solidification. Because of the high Ni content, the TZ in this combination tends to be austenitic (fcc) rather than martensitic, since the austenite that forms at elevated temperature is quite stable and resists transformation to martensite. When such a combination is subjected to PWHT to temper the HAZ of the steel, carbon migration from the steel to the cladding occurs. This results in depletion of carbon in the steel and subsequent buildup in the cladding at the interface. For the 8630/Alloy 625 combination shown in Figure 2.34, this leads to the transformation to ferrite in the carbon-depleted zone and the formation of a very hard austenite band in the planar growth region of the cladding [23].

## 2.4 UNMIXED ZONE (UMZ)

The UMZ represents the region of the fusion zone immediately adjacent to the fusion boundary, as shown in Figure 2.4. It is normally very narrow relative to the other regions of a weld, which can lead to the notion that it is insignificant. In many systems, it may be difficult to even distinguish the UMZ. For some dissimilar combinations, the

mechanical properties or corrosion properties of the UMZ can be significantly different from those of the base and filler metals. For example, the UMZs in some combinations are subject to cracking or localized corrosive attack.

Theoretically, a UMZ is present in every fusion weld. Because the fluid velocity in the weld must go to zero at the fusion boundary, a stagnant liquid layer of some finite thickness will exist. Even in autogenous and homogenous welds, a UMZ would exist since the adjacent weld metal would be of slightly different composition due to evaporation or contamination effects. In practice, the UMZ is normally indistinguishable in these welds since its microstructure is similar to that of the bulk fusion zone.

UMZs are almost always associated with heterogeneous welds, particularly where the relative compositions and physical properties of the base and filler metals are quite different. The size and nature of the UMZ can vary tremendously depending on a number of material and process variables.

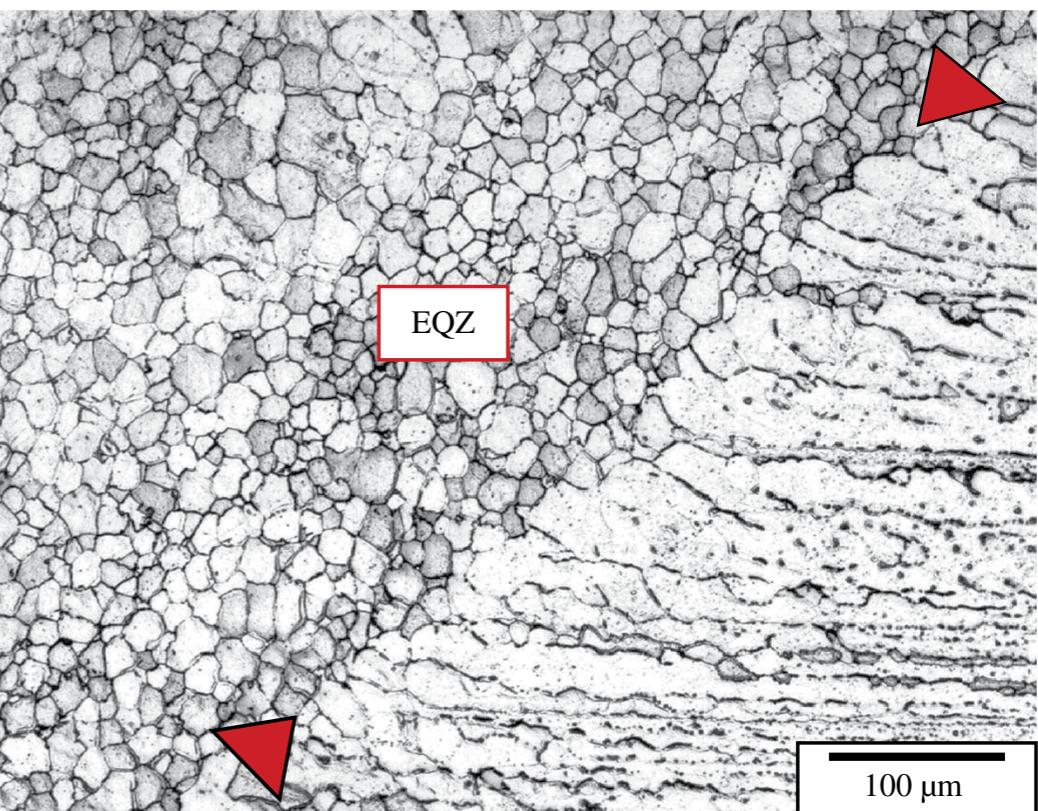
There are many factors that affect the formation of the UMZ. Large relative differences in the composition of base and filler metals may result in important differences in melting temperature and fluid properties. Base metals with a higher melting temperature than the weld metal are probably more prone to UMZ formation, although this effect is complicated by other factors. Differences in fluid viscosity may also be important. For example, if the molten base metal is more viscous than the weld metal, the UMZ is less likely to be disturbed. The miscibility, or ability of the two fluids to mix, may also influence UMZ formation.

Welding process has a significant effect on UMZ formation. High-energy-density processes, such as EBW and LBW, almost never exhibit a UMZ, probably because of the low heat inputs, steep fusion boundary temperature gradients, and vigorous weld pool stirring associated with these processes. Arc welding processes, such as GTAW, GMAW, and PAW, are much more likely to exhibit a UMZ in certain alloy systems.

Both fluid flow in the weld pool and the temperature gradient along the fusion boundary have an important influence on UMZ formation. If fluid flow is vigorous, the UMZ will be “stirred” into the weld metal. Regions of arc welds where fluid flow is sluggish often exhibit distinct UMZs. The temperature gradient at the fusion boundary can influence the width of the UMZ, since it affects the distance over which the base metal is molten.

UMZs have been observed in a number of systems. In their original paper proposing the expanded regions of a fusion weld, Savage *et al.* [4] studied the UMZ in a high-strength low-alloy (HSLA) steel, HY-80, designed for submarine hulls. Baeslack and Lippold determined that a UMZ formed in Type 304L stainless steel when welded with GMAW using either Type 312 or 310 filler metal was more susceptible to corrosion than either the weld metal or base metal [24]. Similar corrosion issues have been found with the UMZ of the high-Mo, high-N “superaustenitic” stainless steels when they are welded with Ni-base filler metals [25].

Other alloy systems in which UMZ formation has been reported in the literature include aluminum alloys [26], Ni-base alloys [27], and some dissimilar combinations, particularly those involving steels and Ni-base alloys. Many other



**FIGURE 2.36** Unmixed zone in an Al–Li alloy (Arrows indicate the fusion boundary)  
(From Ref. [26]. © AWS)

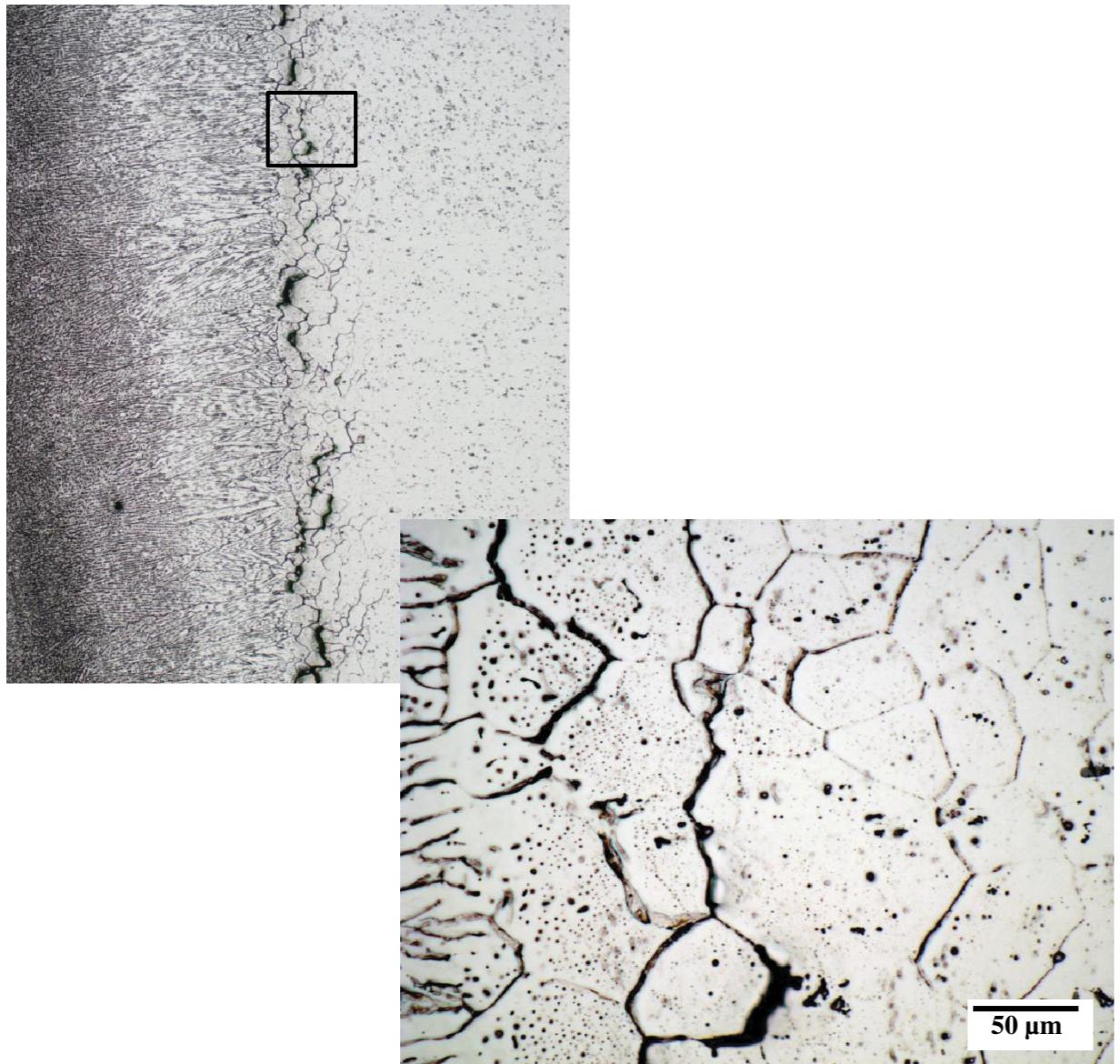
## 2.5 PARTIALLY MELTED ZONE (PMZ)

The PMZ represents a transition region between 100% melting in the fusion zone (or the UMZ at the fusion boundary) and the 100% solid region of the weld (the T-HAZ). In a pure metal, a PMZ will not exist because there is no liquid–solid temperature range. In an isotropic alloy (one in which no segregation or local variations in composition exist), the PMZ represents the temperature range between the alloy liquidus and solidus temperature. For most alloys, this range is typically narrow (25–100°C in most iron- and nickel-base alloys) and would predict a narrow PMZ.

In most engineering alloys, segregation of alloying and impurity elements increases the “effective” melting temperature range of the base material. The temperature range between the liquidus and “effective” solidus temperatures is generally used to describe the extent of the PMZ. There are a number of phenomena that influence the magnitude of this range by promoting liquation reactions under nonequilibrium thermal conditions.

As indicated earlier, the PMZ in isotropic alloys would simply represent the temperature range between the solidus and liquidus on the phase diagram. In actuality, solute and impurity elements are not distributed uniformly in the base metal, and further segregation may occur during the weld thermal cycle. The net effect is that local variations in composition in the HAZ adjacent to the fusion boundary will promote melting at temperatures below that of the bulk microstructure.

Grain boundaries typically have a higher concentration of alloy and impurity elements than the grain interiors. This concentration gradient lowers the melting



**FIGURE 2.46** (Continued) and (c) aluminum Alloy 6022.

## 2.6 HEAT AFFECTED ZONE (HAZ)

The T-HAZ according to Savage *et al.* [4] separates the PMZ from the unaffected parent, or base material. By definition, all reactions in the T-HAZ occur in the solid state. No melting or liquation reactions occur in this region. For simplicity, this region will be referred to as the HAZ in this text.

Microstructure evolution in the HAZ can be quite complex, depending on both composition and thermal factors. Heating and cooling rates will influence the reactions in this region and can often have profound microstructural effects within the same alloy or alloy system.

The HAZ microstructure surrounding a fusion weld is influenced by many variables. Welding process and heat input, material type, and material condition prior to welding are all important variables that can affect the HAZ thermal history and hence the resultant microstructure. Because the HAZ experiences a spectrum of thermal

cycles (peak temperatures and cooling rates), a wide range of microstructure is possible within the same weld and local variations may be large.

The reactions that occur in the HAZ can be quite complex. There are many possible metallurgical reactions, and any given area in the HAZ may undergo one or more of the following possible reactions:

- Recrystallization
- Grain growth
- Phase transformations
- Dissolution/overaging of precipitates
- Precipitate formation
- Residual stress and stress relaxation

The equilibrium phase diagram is a guideline for determining possible reactions in most material systems. During welding, rapid heating and cooling may suppress these equilibrium reactions and/or promote others not predicted by the phase diagram. For example, continuous cooling transformation (CCT) diagrams have been developed for steels that predict transformations to metastable phases that are not present on the equilibrium phase diagram, such as austenite transformation to bainite and martensite.

Heat input and heat flow conditions can influence the dimensions and nature of the HAZ. These dimensions are controlled by the temperature gradient from the fusion boundary into the surrounding base metal and the nature of the metallurgical reactions that occur over that temperature range. In situations where heat input is low and/or heat flow is effective (high thermal diffusivity), a narrow HAZ will result. This is the case in copper and aluminum alloys that possess high thermal conductivity allowing heat to readily flow away from the weld region. The use of low heat input processes and procedures will also minimize the HAZ, since less heat is introduced into the weld. When heat input is high and/or heat flow away from the weld is restricted, a wider HAZ will result. In stainless steels and nickel-base alloys, which exhibit low thermal conductivity, heat builds up around the weld resulting in a shallow temperature gradient. The net effect is to form a wide HAZ. Similarly, a wide HAZ will exist in welds in thin-sheet materials, since heat flow is controlled by section thickness.

### 2.6.1 Recrystallization and Grain Growth

When welding is conducted on materials that have been strengthened by cold work, the HAZ will be softened by recrystallization and grain growth of the cold-worked microstructure. This process actually occurs in three stages as shown in Figure 2.47.

During the *recovery stage*, internal energy is reduced by the rearrangement of dislocations. This often results in what is described as a cellular dislocation structure that produces strain-free regions in the structure. It is these strain-free regions that act as nuclei for newly formed grains. Since the mechanism involved results in