Recent Advances in the Deformation Processing of Titanium Alloys

S. Tamirisakandala, B.V. Vedam, and R.B. Bhat

(Submitted 4 August 2003)

Titanium (Ti) alloys are special-purpose materials used for several critical applications in aerospace as well as non-aerospace industries, and extensive deformation processing is necessary to shape-form these materials, which poses many challenges due to the microstructural complexities. Some of the recent developments in the deformation processing of Ti alloys and usefulness of integrating the material behavior information with simulation schemes while designing and optimizing manufacturing process schedules are discussed in this paper. Discussions are primarily focused on the most important alloy, Ti-6Al-4V and on developing a clear understanding on the influence of key parameters (e.g., oxygen content, starting microstructure, temperature, and strain rate) on the deformation behavior during hot working. These studies are very useful not only for obtaining controlled microstructures but also to design complex multi-step processing sequences to produce defect-free components. Strain-induced porosity (SIP) has been a serious problem during titanium alloy processing, and improved scientific understanding helps in seeking elegant solutions to avoid SIP. A novel high-speed processing technique for microstructural conversion in titanium has been described, which provides several benefits over the conventional slow-speed practices. The hot working behavior of some of the affordable α + β and β titanium alloys being developed recently—namely, Ti-5.5Al-1Fe, Ti-10V-2Fe-3Al, Ti–6.8Mo-4.5Fe-1.5Al, and Ti-10V-4.5Fe-1.5Al—has been analyzed, and the usefulness of the processing maps in optimizing the process parameters and design of hot working schedules in these alloys is demonstrated. Titanium alloys modified with small additions of boron are emerging as potential candidates for replacing structural components requiring high specific strength and stiffness. Efforts to understand the microstructural mechanisms during deformation processing of Ti-B alloys and the issues associated with their processing are discussed.

Keywords deformation processing, process design, titanium alloys

1. Introduction

Titanium (Ti) alloys offer a unique combination of mechanical and physical properties and excellent corrosion resistance, which make them desirable for a variety of critical applications.[1] Elevated strength-to-weight ratio is the primary incentive for selection and design into aerospace applications including engine and airframe components. The expansion of titanium applications to non-aerospace industries (e.g, automotive, chemical, energy, marine, biomedical, sports, and architecture) entails improvements in the understanding of titanium metallurgy, advances in processing methods, ability to manufacture components without defects, and development of low-cost alloys.[2] Depending upon the type of alloy (α, α + β, or β), these can be processed to produce a gamut of microstructures with tailor-made mechanical properties to suit the criticality of the application.[3] Shape-forming of engineering components in these materials requires extensive mechanical working often involving complex multistage processing sequences to obtain desired final microstructure. Therefore, any innovation in the design and optimization of deformation processing will lead to large overall cost savings and improved service life of components. The key to these innovations is a thorough understanding of the material's constitutive response to the imposed conditions and detailed process analysis. This is achieved by effective material modeling and integration of microstructural information with process simulation and optimization methodologies.

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It is well known that evolution of microstructure during hot working is very sensitive to process parameters such as temperature, strain rate, strain, and "history" of the material (chemistry and starting microstructural condition). Ti alloys in general are difficult to work compared with steels in view of the microstructural complexities and therefore close control of process parameters is necessary to avoid occurrence of defects. In recent years, the processing maps approach[4] has been established as a useful tool in seeking solutions to problems during bulk metalworking processes such as forging, rolling, and extrusion. Processing maps are developed on the basis of hot compression experimental data over wide temperature (T) and strain rate ($\dot{\varepsilon}$) ranges. The flow stress data, which implicitly contains the materials response to the imposed conditions, is analyzed to evaluate its temperature and strain rate sensitivity for identifying various microstructural mechanisms that operate under different T-$\dot{\varepsilon}$ combinations.[5] The information obtained from processing maps along with hot ductility and grain size considerations can be used to effectively design and control industrial process schedules. This approach has been ap-

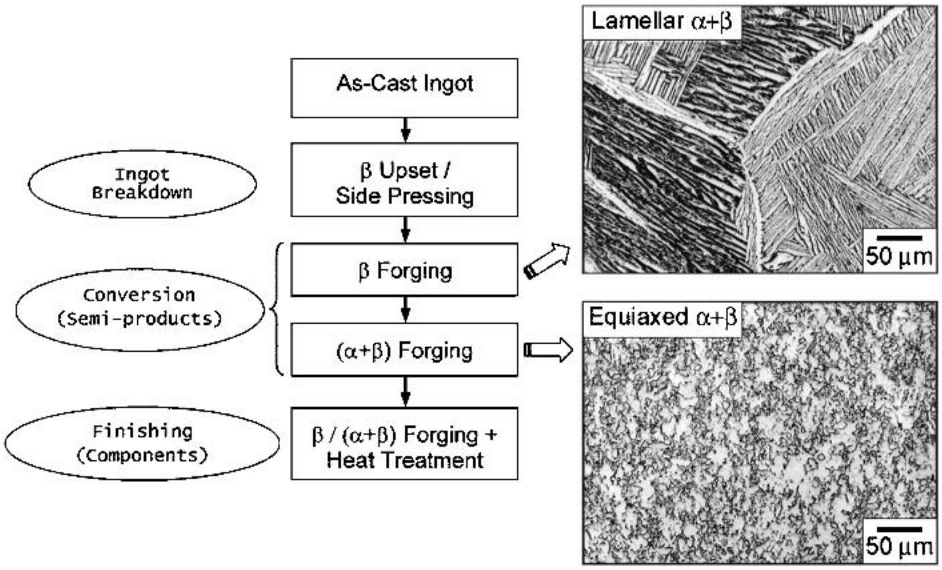


Fig. 1 Typical thermomechanical processing sequence used for Ti-6Al-4V

plied to several important alloys, which enabled optimization of microstructures and also many process innovations.[4] The intent of this paper is to provide an overview of recent developments in Ti alloy deformation processing based on few examples. The improved scientific understanding of the processing and its impact on the technology of Ti alloy component manufacture is discussed.

2. Ti-6Al-4V

Among all titanium alloys, Ti-6Al-4V (all compositions in this paper are given in wt.% unless specified otherwise) is the most widely used and accounts for more than 50% of the worldwide titanium tonnage. While Al addition stabilizes hcp α-phase and increases strength, V addition is intended to introduce bcc β-phase into the α matrix, thereby increasing ductility and fracture toughness. Ti-6Al-4V is characterized as an α-rich α + β alloy in which the particular Al-V balance offers a wide spectrum of mechanical property combinations depending on the thermomechanical processing (TMP) history. The sequence of TMP steps generally adopted to manufacture components in this alloy is shown in Fig. 1, which involves ingot breakdown, billet conversion, and component manufacture. The β transus (temperature at which $\alpha + \beta \rightarrow \beta$ phase transformation occurs) is an important parameter to be considered during the design of TMP schedules since entirely different microstructures evolve depending upon whether the alloy is processed above or below this temperature. During primary processing, the coarse as-cast microstructure is broken down by deformation processing above the β transus. The resulting microstructure after β processing consists of lamellar colonies of α and β in large prior β grains of about 200 μm, as shown in Fig. 1. Secondary processing involves conversion of ingots into semi-products (e.g., billets, plates, and rods) with a concurrent breakdown of lamellar microstructure into equiaxed. This is achieved by extensive plastic deformation well below the β transus. While the lamellar microstructure exhibits high strength and fracture toughness, equiaxed microstructure possesses excellent ductility and resistance to crack initiation under low-cycle fatigue loading.[1] In view of the widely different mechanical properties of lamellar and equiaxed microstructures, selection of the starting microstructure is made depending upon the criticality of the application. Component manufacture involves either α + β or β processing with appropriate heat treatment to obtain the desired final microstructure. Significant differences in the morphology and distribution of microstructural constituents in lamellar and equiaxed starting structures, requires a thorough understanding of their response during hot working, which is essential in designing deformation processing schedules and controlling the microstructural evolution.

Another important variable that significantly affects the mechanical properties in Ti alloys is the oxygen content. Ti-6Al-4V is marketed in two grades, the standard grade which contains 0.16-0.20 wt.% oxygen and the extra-low interstitial (ELI) grade which contains 0.10-0.13 wt.% oxygen. The standard grade is used for strength critical applications (e.g., turbine disks) while the ELI grade is the most common choice for toughness critical applications (e.g., bulkheads).[1] Oxygen as a strong α stabilizer in titanium can significantly influence hot deformation mechanisms, and therefore understanding the behavior of standard and ELI grades is critical for process design. In the following sections, the influence of starting microstructure and the oxygen content on the hot deformation response of Ti-6Al-4V and their impact on industrial processing are presented.

2.1 Standard Ti-6Al-4V With Lamellar Starting Microstructure

Processing map for hot working of Ti-6Al-4V with a lamellar α + β starting microstructure[6] identifying various microstructural mechanisms in the temperature range 750-1100 °C

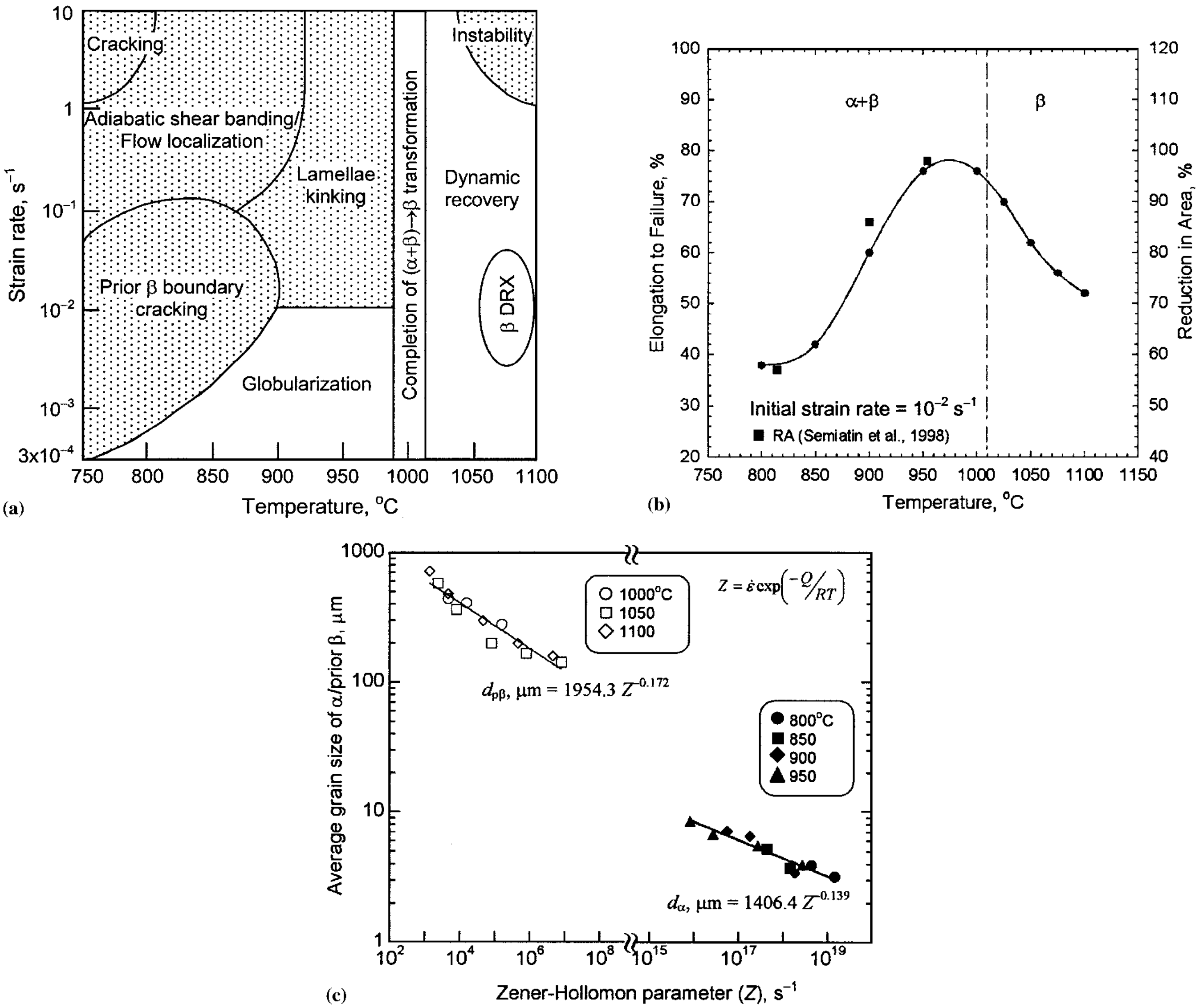


Fig. 2 **(a)** Processing map for hot working of standard grade Ti-6Al-4V with a lamellar α + β starting microstructure, **(b)** variation of ductility with temperature, and **(c)** variation of grain size with temperature-compensated-strain rate parameter (Z)

and strain rate range 3×10^{-4} to 10 s^{-1} is shown in Fig. 2(a). Representative microstructures of these microstructural processes are presented in Fig. 3. Microstructural conversion from lamellar (Fig. 3a) into equiaxed (Fig. 3b) morphology is achieved by deformation processing in the region marked globularization in Fig. 2(a), which occurs at slow speeds corresponding to strain rates below 10^{-2} s^{-1}. The mechanism of globularization involves[6] shearing of lamellae during deformation and subsequent rounding by diffusion processes, which is driven by the urge to minimize the surface area for attaining thermodynamically lowest energy state. As can be seen in Fig. 2(a), precise temperature and strain rate control during conversion is essential to avoid formation of unacceptable microstructures. When lamellar Ti-6Al-4V is hot worked below ~900 °C at slow strain rates, cracking at the prior β grain boundaries (Fig. 3c) occurs due to incompatibility between the deformation behaviors of micro-constituents (grain boundary α, colony α + β, and thin β layer in-between grain boundary and colonies as seen in Fig. 3a) and insufficient accommodation of the stress concentrations developed at the interfaces. With respect to strain rate, deformation processing of lamellar Ti-6Al-4V should be limited to slow speeds due to the occurrence of adiabatic shear banding (Fig. 3d) and flow localization (Fig. 3e) at high strain rates. At fast deformation rates, there is insufficient time for the dissipation of adiabatic deformation heat generated, which causes highly localized flow along the maximum shear stress plane (at an angle ±45° to the stress axis) and ultimately leads to cracking. The intensity of these defects decreases with increase in temperature and decrease in strain rate. Deformation at higher temperatures well below the β transus and higher strain rates causes kinking of lamellae (Fig. 3f) instead of shearing, which is a type of mechanical instability (e.g., buckling). When the deformation temperature is close to the β transus, on the other hand, fully lamellar microstructures form due to rapid increase in the β phase content with increase in temperature, which

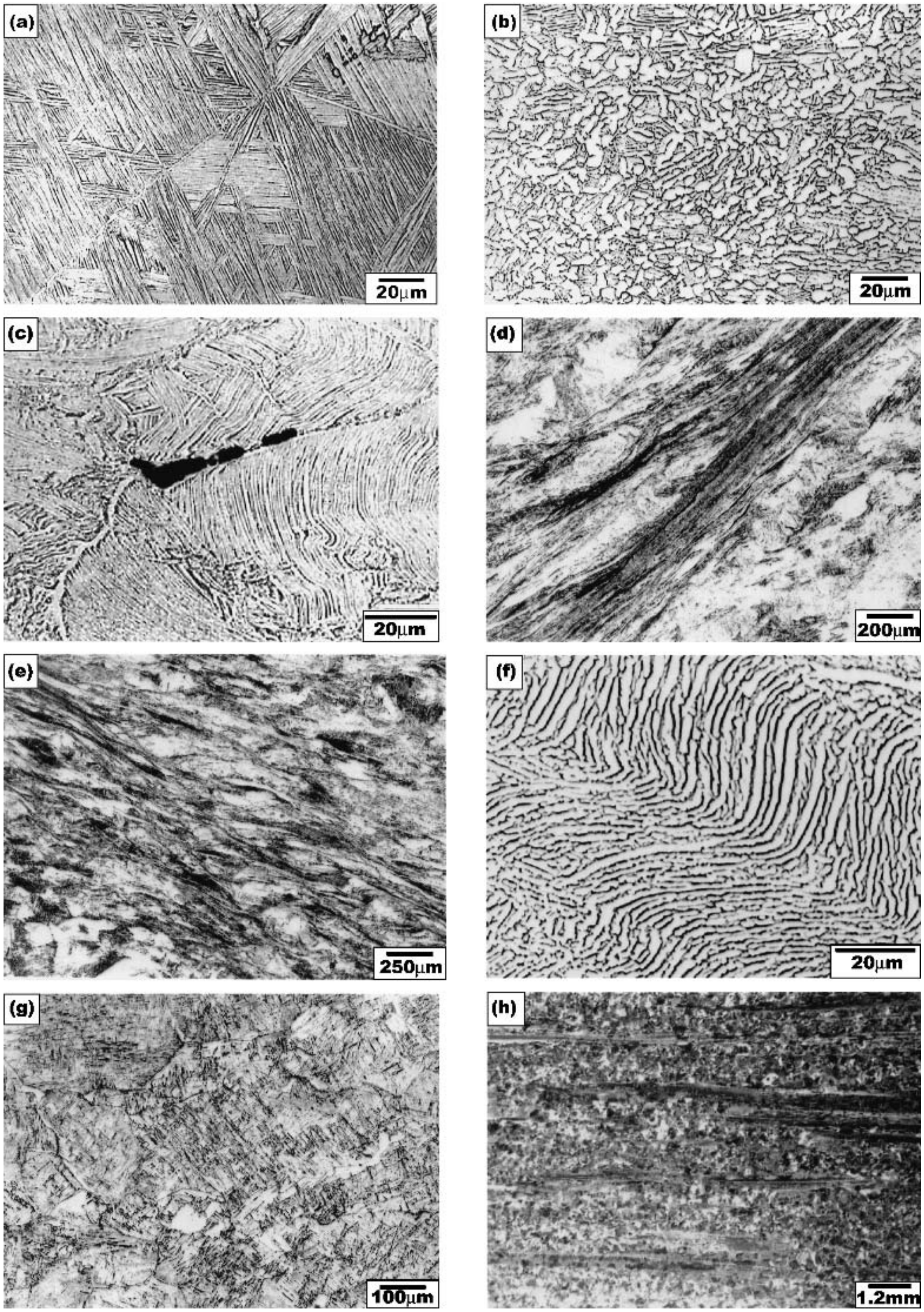


Fig. 3 Representative micrographs of standard grade Ti-6Al-4V: **(a)** lamellar starting microstructure, **(b)** globularization, **(c)** prior β boundary cracking, **(d)** adiabatic shear banding and cracking, **(e)** flow localization, **(f)** lamellae-kinking, **(g)** β dynamic recrystallization, and **(h)** β instability

reverts the microstructure back to fully lamellar condition and ruins the purpose of conversion. The variation of elongation to failure with temperature at an initial strain rate of 10^{-2} s^{-1} for lamellar Ti-6Al-4V is shown in Fig. 2(b), which exhibits good ductility (50-80%) in the conversion temperature range. The optimum temperature for globularization is approximately 960 °C at which the ductility curve exhibits a peak and this condition corresponds to equal volume fractions of α and β phases. The variation of α grain size with the Zener-Hollomon parameter Z which combines the effect of temperature (T) and strain rate ($\dot{\varepsilon}$) on flow stress by the equation

$$Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right) \quad \text{(Eq 1)}$$

is shown in Fig. 2(c). In Eq 1, Q is the activation energy for hot deformation (445 kJ/mol for globularization) and R is the gas constant (8.314 J/K-mol). The α grain size after conversion under various T-$\dot{\varepsilon}$ combinations can be predicted from the power-law relationship

$$d_{\alpha},\ \mu\text{m} = 1406.4\ Z^{-0.139} \quad \text{(Eq 2)}$$

The grain size increases with increase in T and decrease in $\dot{\varepsilon}$ and the kinetics of globularization are controlled by these two individual process parameters.

Ingot breakdown operations (upsetting and side pressing) for lamellar Ti-6Al-4V are generally conducted above the β transus in the β phase field taking the advantages of chemical homogenization due to anomalously high diffusion rates, good workability, and low press loads. Deformation in the β phase field is best conducted in the dynamic recrystallization (DRX) region (Fig. 3g) at slower strain rates or in the dynamic recovery region at moderately high strain rates which produces acceptable microstructures with fine prior β grain sizes. Figure 2(b) shows that β phase exhibits good ductility, and the variation of prior β grain size with Z presented in Fig. 2(c) is useful in predicting and controlling the grain size after β deformation processing. The β processing should be limited to lower temperatures to avoid excessive grain growth and minimize oxidation damage. At fast deformation rates (>1 s^{-1}), flow instabilities occur in the β phase field, which manifest as unstable flow (Fig. 3h), and these conditions should be avoided.

2.2 Standard Ti-6Al-4V With Equiaxed Starting Microstructure

Processing map for hot working of Ti-6Al-4V with an equiaxed α + β starting microstructure (average α grain size ≈8 μm) in the ranges 750-1100 °C and 3×10^{-4} to 100 s^{-1} is shown in Fig. 4(a).[7] In the α + β phase field (750-960 °C) at slow strain rates (<10^{-2} s^{-1}), equiaxed Ti-6Al-4V exhibits superplastic behavior, in contrast to grain boundary cracking or globularization in lamellar Ti-6Al-4V (Fig. 2a) under similar processing conditions. Variation of elongation to failure with temperature at an initial strain rate of 10^{-2} s^{-1} is shown in Fig. 4(b), which exhibits large elongations above 100% confirming the fine-grained superplasticity. The ductility curve exhibits a peak (~200%) at about 825 °C, which is the optimum temperature for superplasticity. The mechanism of superplasticity involves grain boundary sliding with concurrent diffusion accommodated flow. The variation of β volume fraction with temperature taken from Ref. 8 is also plotted in Fig. 4(b), which shows that at the optimum superplastic temperature, the microstructure consists of 25 vol% β. The ductility variation with temperature exhibits a sharp drop in ductility in the temperature range 825-950 °C and the volume percent of β increases from about 25 to 60 in this range. This ductility drop is attributed to the increase in the α grain size as well as β volume fraction and therefore temperature is a critical processing parameter during superplasticity. A smaller β phase content helps in maintaining a stable fine-grained structure and relieving the

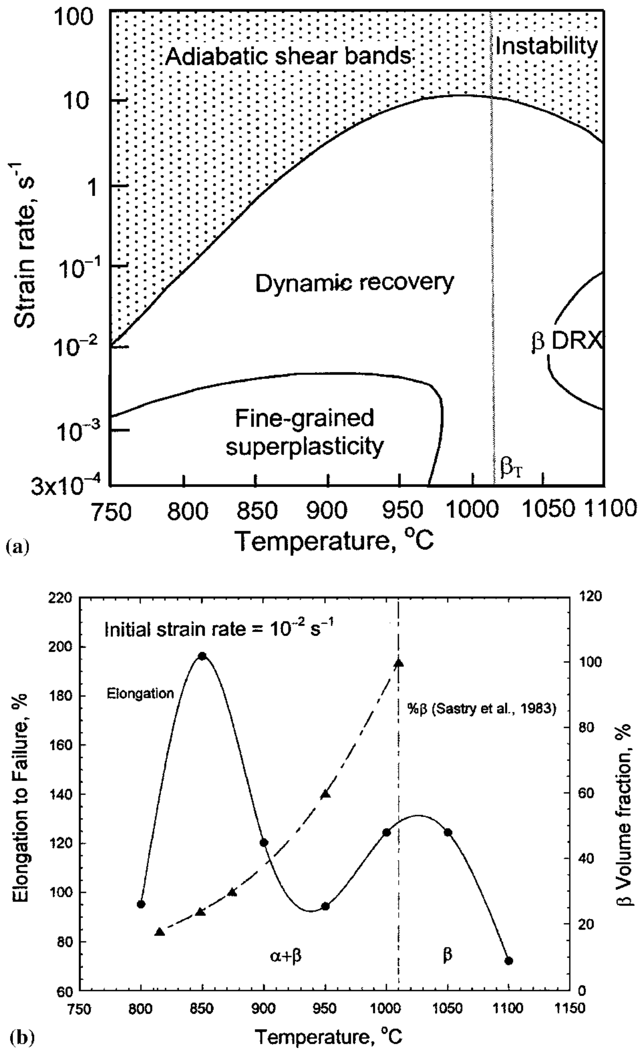


Fig. 4 **(a)** Processing map for hot working of standard grade Ti-6Al-4V with an equiaxed starting microstructure (β_T: β transus), and **(b)** variation of ductility and β phase volume fraction with temperature[8]

stress concentrations developed during sliding via plastic deformation, while a higher β content increases the number of α-β boundaries, which cannot slide due to mismatch in the deformation characteristics of individual phases. As for the strain rate, superplastic characteristics are enhanced with decrease in strain rate, which allows sufficient time for the sliding and diffusional processes. Apart from large ductility, the other advantages of superplasticity include low press loads, good grain size control, and ability to form complex shapes. Superplastic forming (SPF) is a well-established commercial process in Ti-6Al-4V, which is often coupled with diffusion bonding to manufacture near-net/net shape sheet metal components with intricate geometries.[1] However, in view of the slow strain rates associated with SPF, isothermal (workpiece and dies heated to same temperature) forming is necessary to avoid temperature drop during deformation. Equiaxed Ti-6Al-4V can also be hot worked at high speeds in the dynamic recovery region, which produces acceptable microstructures although SPF benefits will be lost. The upper strain rate limit for processing of equiaxed Ti-6Al-4V is set by the onset of micro-

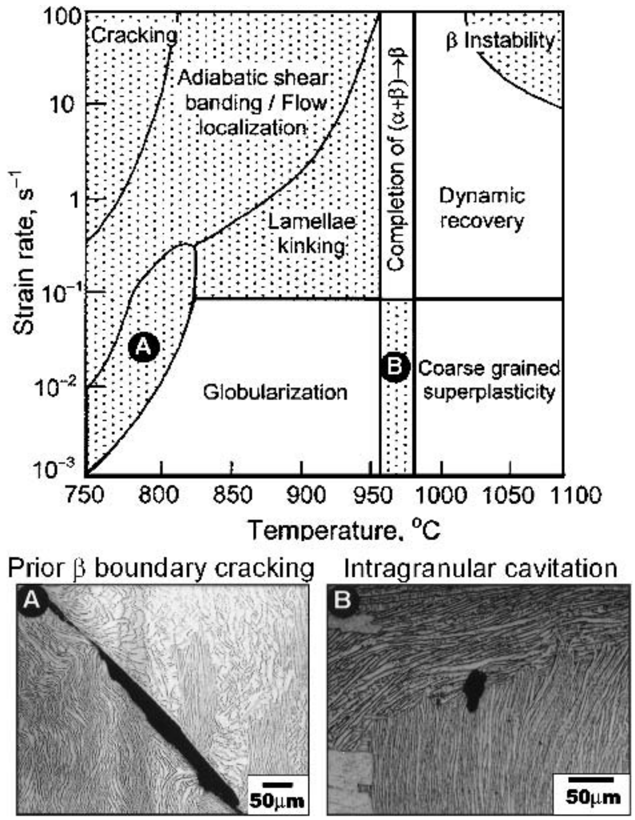


Fig. 5 Processing map for hot working of ELI grade Ti-6Al-4V with a lamellar $\alpha + \beta$ starting microstructure

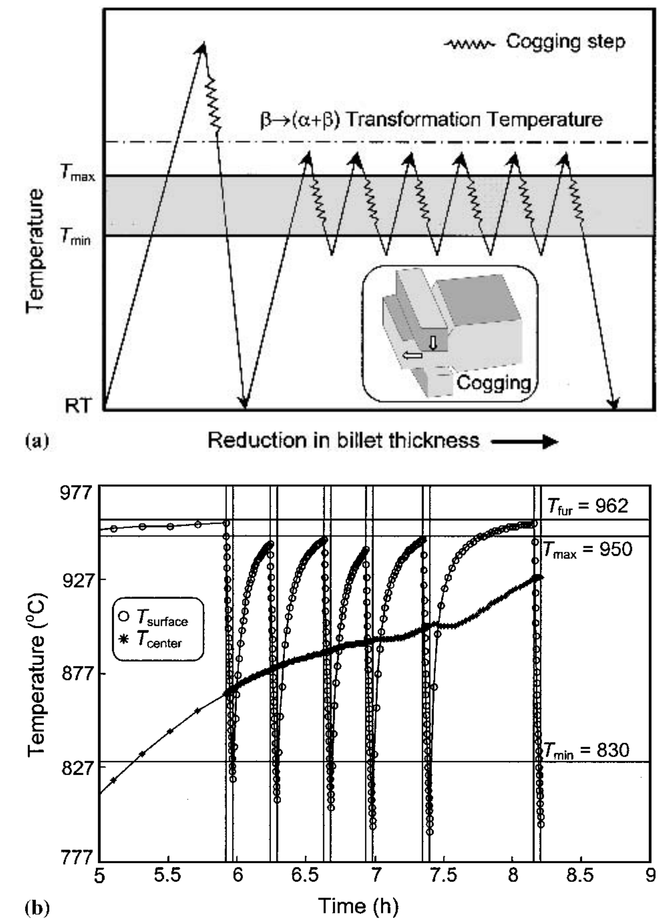


Fig. 6 **(a)** Illustration of a typical sequential cogging cycle used to convert large ELI Ti-6Al-4B billets into plates as well as to breakdown lamellar microstructure into equiaxed. Shaded region represents the "safe" temperature range for processing without the initiation of microstructural defects, **(b)** transient thermal response of the surface and center of the billet during sequential cogging. The vertical bands represent the out of furnace time interval (3 min. each)

structural defect which manifests as adiabatic shear bands. In the β phase field, the deformation behavior of equiaxed Ti-6Al-4V is similar to that of lamellar with safe regions being dynamic recovery or recrystallization and flow instabilities occurring at strain rates above 5 s^{-1}.

In the design of multi-step deformation processing schedule for Ti-6Al-4V starting with a cast ingot, processing map for lamellar starting microstructure (Fig. 2a) is very useful for designing and optimizing ingot breakdown and conversion operations while the process parameters for finishing operations can be designed using the processing map of equiaxed Ti-6Al-4V (Fig. 4a).

2.3 ELI Ti-6Al-4V With Lamellar α+β Starting Microstructure

Processing map for hot working of ELI grade lamellar Ti-6Al-4V[9] is shown in Fig. 5, which reveals three major differences in the deformation behavior compared with the standard grade. First, in the $\alpha + \beta$ range, microstructural mechanisms in ELI grade are similar to that of standard grade with a shift in the upper temperature limits by 35-40 °C, which is equal to the difference in the β transus between the two grades. Secondly, unlike the standard grade, the ELI Ti-6Al-4V exhibits a region of intragranular cavitation near the β transus, typical microstructure of which is shown in Fig. 5. The implication of this damage mechanism is that it sets an upper constraint during microstructural conversion in ELI grade. The occurrence of prior β boundary cracking at lower temperature and cavitation near the transus together constitute the strain-induced porosity (SIP) and these damage processes narrow the $\alpha + \beta$ working processing window of ELI grade. Thirdly, the β phase exhibits coarse-grained superplasticity at slow strain rates, which can be advantageously used to form complex shapes with ease. In summary, processing of ELI grade is more difficult than the standard grade and requires close control of processing temperature.

2.4 ELI Ti-6Al-4V Cogging Sequence Design and Optimization

Among all processing steps of the TMP sequence of Ti-6Al-4V, billet conversion is the most critical and time-consuming operation and the occurrence of SIP has been a serious industrial problem. In this section, we discuss the integration of intrinsic material behavior during conversion with that of process simulation models to optimize the billet conversion sequence design while maintaining microstructural control.

Conventionally, conversion is achieved by a multi-step open-die forging commonly referred to as "cogging," which is illustrated in Fig. 6(a). Cogging consists of repeated furnace heating and plastic deformation of the billet to convert the coarse lamellar microstructure of the ingots into a wrought equiaxed structure in the semi-products. Cogging (illustrated in the inset of Fig. 6a) is performed by series of short bites along the length of the billet wherein the bite size is limited by the press capacity. In view of relatively large section size (e.g., 1000 mm thick) of the starting billet, strain penetration through the thickness is not achievable in a single step. Therefore, several cycles of reheating and cogging are necessary to obtain the final size and desired microstructure. During the processing of large ELI Ti-6Al-4V ingots, it is difficult to keep the surface temperature hot enough to avoid grain boundary cracking, while at the same time preventing the center of the billet from being overheated, thereby increasing the likelihood of cavitation (Fig. 5). It must be noted that the optimum temperature for globularization of ELI grade Ti-6Al-4V is lower than that of standard grade, and if the ELI grade is cogged by heating the billet to a uniform temperature that is optimum for the commercial grade—namely, 960 °C—void nucleation occurs in the center portion, which does not loose temperature during cogging operation due to huge thermal mass. Finite element simulations of cogging revealed that the stress state changes from compressive to tensile from the surface to the center, and a large tensile principal stress locks up at the mid plane. It has been demonstrated that voids nucleated during the deformation step can grow during re-soaking periods under a state of tensile stress by a simple creep relaxation process.[11] If the billet is heated to a lower temperature, on the other hand, surface cracking aggravates due to temperature loss during billet transfer and sluggishness of the cogging operation. One possible solution to avoid both the defects in any portion of the billet is to use a "differential" heating strategy, whereby the entire billet is not heated to a uniform temperature before being taken out from the furnace for cogging.[11] The temperature at the surface of the billet should be higher than the limit for the onset of grain boundary cracking and the mid region temperature should be lower than the limit for the initiation of cavitation. A trajectory optimization algorithm has been developed[12] to determine the optimum re-soaking periods during a typical 6 cycle $\alpha + \beta$ cogging sequence such that the temperatures at the surface and the center of the billet will remain within the limits prescribed by the material behavior and the results of process optimization are shown in Fig. 6(b). The optimization of soaking and resoaking times for the differential heating of the billet not only leads to the mitigation of SIP, but also reduces the total process time by 30% compared with the conventional practice.

2.5 High-Speed Microstructural Conversion in Ti-6Al-4V

As discussed in the previous sections, break-up of lamellar colony microstructure developed after processing above the β transus into equiaxed grain structure is the most time-consuming step in the manufacture of Ti-6Al-4V products for engineering applications. Conventional microstructural conversion in Ti-6Al-4V by cogging has to be limited to slow speeds in view of the occurrence of defects, such as cracking and adiabatic shear banding, at high deformation speeds. The cogging process is not only slow, but also demands a close temperature control to avoid formation of unacceptable microstructures: cracking at the prior β grain boundaries at low temperatures in the $\alpha + \beta$ phase field, and lamellar microstructure/cavitation near the β transus. Recent studies[13,14] on high-speed deformation of lamellar Ti-6Al-4V have indicated that it is possible to obtain equiaxed morphology by deforming at strain rates above 1 s^{-1} near the β transus in the window identified in Fig. 7(a). A typical microstructure of a specimen deformed in this new high-speed processing window is shown in Fig. 7(b). This is the highest deformation rate reported for converting titanium alloy microstructures without producing any macro or micro defects. We note that such a difference in the available strain rate leads to quite a large difference in the deformation time: to achieve a strain of 50%; for example, it takes about 30 min at 3×10^{-4} s^{-1}, whereas it takes only 0.5 s at 1 s^{-1}.

Explanation for the microstructural evolution after high-speed deformation near the β-transus comes from the fact that precipitation of the α-phase from the β phase is a nucleation and growth process, the kinetics of which are significantly influenced by the deformation rate.[15] It is well known that statistical dislocations are generated during plastic deformation and the dislocation density is directly proportional to the deformation rate at a given temperature. The excess dislocations produced during high-speed deformation act as effective sites for heterogeneous nucleation of the α-phase in the β matrix grains during cooling. Nucleation site density estimations using experimental stress data plotted as a function of strain rate in Fig. 7(c) revealed that the number of available nucleation sites increases by three to five orders of magnitude at high strain rates (>1 s^{-1}) compared with low strain rates (<10^{-1} s^{-1}). In addition to this, contribution from diffusion through dislocation core (pipe-diffusion) increases the overall diffusion coefficient, which accelerates the kinetics of precipitate growth leading to the evolution of equiaxed morphology of the α-phase. Deformation at slow speeds (<1 s^{-1}) near the β-transus, on the other hand, produces lamellar microstructure, which forms by nucleation of α precipitates on the prior β grain boundaries and their simultaneous growth according to the crystallographic orientation relationships between the parent β and product α-phases. Deformation at higher temperatures outside the high-speed processing window produces coarse lamellar microstructure while deformation below the lower temperature limit causes kinking of lamellae. The benefits of the novel high-speed process over the conventional practice include considerable increase in production rates, affordable component manufacture via reduction in number of processing steps, and the ability to implement this process using conventional metalworking equipment such as hammers.

3. Ti-5.5Al-1Fe

Materials used for permanent implants in the human body must exhibit corrosion resistance, bio-compatibility, amenability to osseo-integration, and bio-functionality (a high ratio of

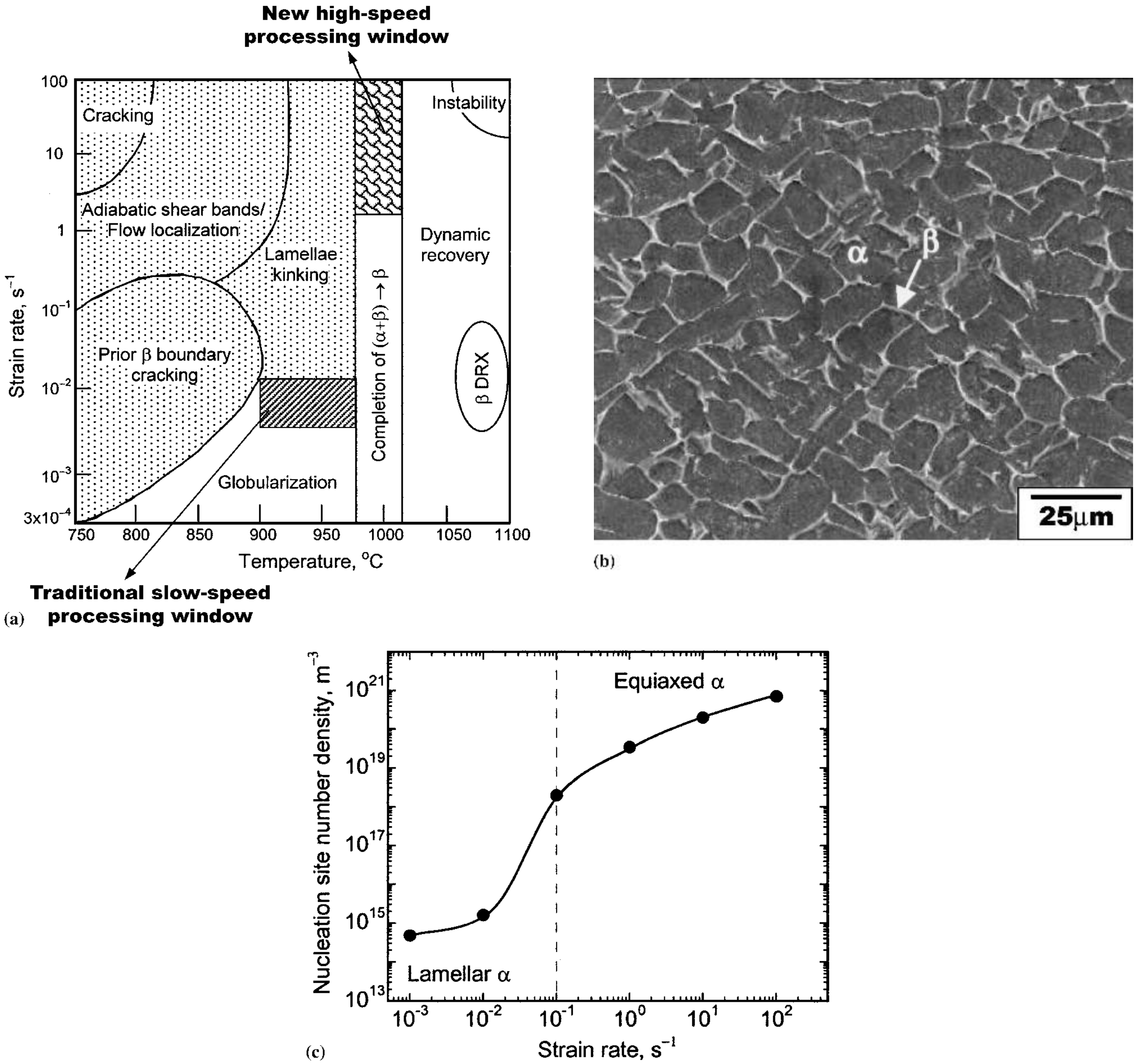


Fig. 7 **(a)** Processing map for hot working of lamellar Ti-6Al-4V (standard grade) with new high-speed conversion window, **(b)** equiaxed α + β microstructure that evolved after high-speed deformation near the β transus, and **(c)** heterogeneous nucleation site density vs. strain rate at the β transus illustrating the change in nucleation kinetics

fatigue strength to Young's modulus). Ti-Al-Fe alloys, which do not contain any toxic (e.g., V) or allergenic constituents or decomposition products, are found to be suitable for these applications. Recently, Ti-5.5Al-1Fe has been developed as a low-cost substitute for Ti-6Al-4V, which meets the essential requirements for medical applications and possesses same level of mechanical properties.[16]

Processing map of Ti-5.5Al-1Fe with an equiaxed α + β starting microstructure (α grain size of ~5 μm) is shown in Fig. 8(a), which exhibits wide processing windows for hot working.[17] Below the β transus (~1010 °C) and at strain rates below 10^{-2} s^{-1}, the alloy exhibits superplasticity. The variation of ductility with temperature at an initial strain rate of 10^{-3} s^{-1} taken from Ref. 16 is shown in Fig. 8(b), which exhibits elongations of several hundred percent under these deformation conditions confirming fine-grained superplasticity in this material similar to Ti-6Al-4V (Fig. 4). A ductility peak of 800% occurs at about 827 °C, which corresponds to α:β phase fraction of 0.86:0.14 and is the optimum temperature to hot work this alloy in the superplasticity region. Superplastic forming characteristics improve with decreasing strain rates due to enhanced accommodation rates, but isothermal forging becomes necessary to avoid temperature loss in the workpiece due to prolonged forming times. Ti-5.5Al-1Fe can also be hot worked at higher strain rates in the dynamic recovery region, although this does not provide the benefits of superplastic forming but

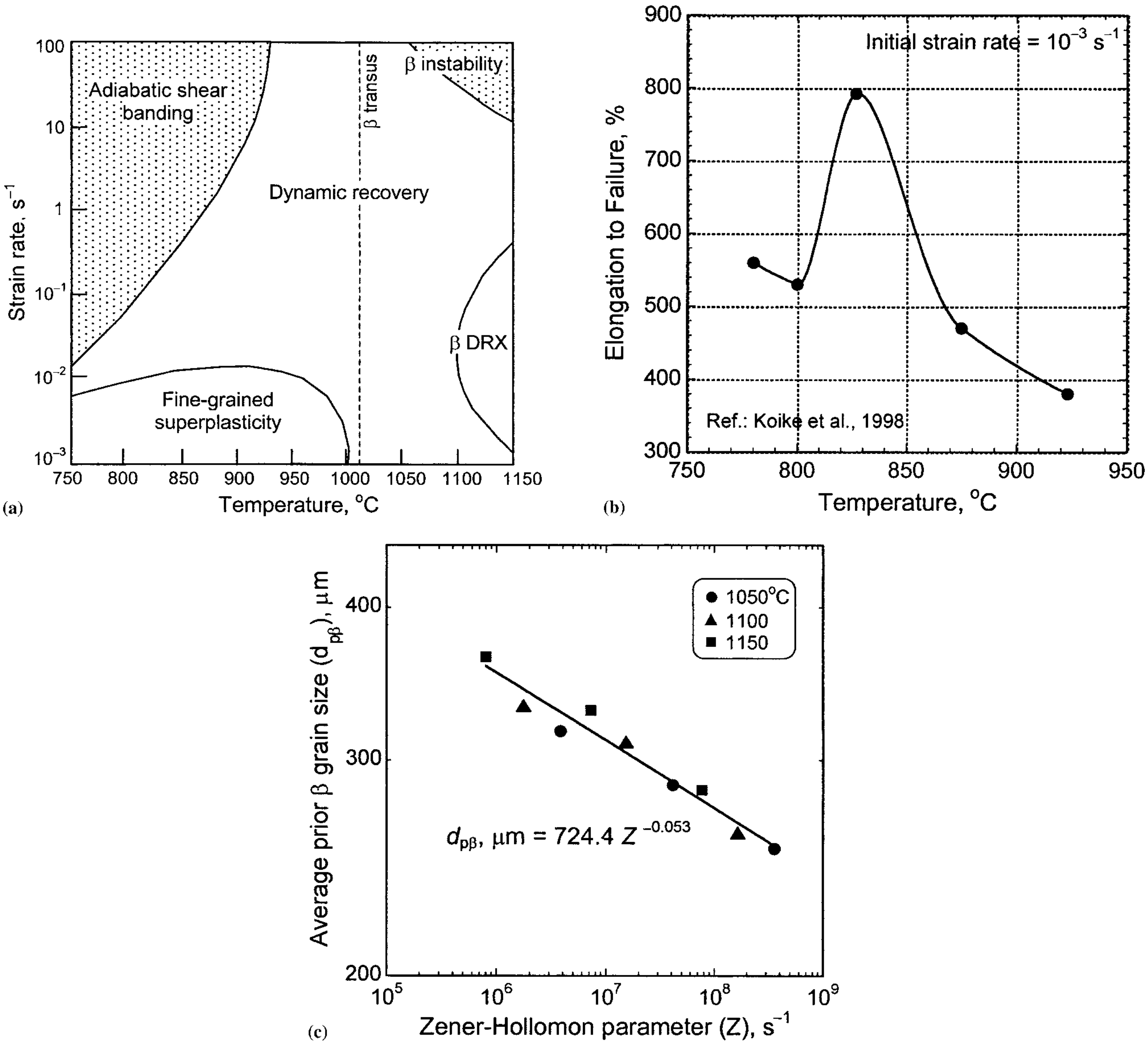


Fig. 8 **(a)** Processing map for hot working of Ti-5.5Al-1Fe with an equiaxed α + β microstructure, **(b)** variation of ductility with temperature in the α + β superplastic forming range, and **(c)** variation of prior β grain size with Z in the β phase field

increases the production rates significantly with acceptable microstructures. At high strain rates and low temperatures in the α + β phase field, flow instabilities manifest in the form of adiabatic shear bands occur and these processing conditions should be avoided to obtain defect-free microstructures. In the β region, Ti-5.5Al-1Fe offers a much wider safe processing window with dynamic softening mechanism being either recovery or recrystallization. Although the alloy exhibits excellent workability in the β range, processing should be restricted to strain rates below 10 s^{-1} to avoid formation of microstructural instabilities. The variation of prior β grain size ($d_{p\beta}$) with Z parameter for different Tr-$\dot{\varepsilon}$ combinations in the β range is presented in Fig. 8(c). The linear relationship between $d_{p\beta}$ and Z given in this figure can be used to predict and control the grain size during β processing. Rapid grain growth and excessive oxidation damage occurs with increase in temperature in the β range and therefore the processing should be restricted to low temperatures. In conclusion, Ti-5.5Al-1Fe seems to be attractive not only from the viewpoint of cost effectiveness and biocompatibility, but also from the ease of processing over wide T-$\dot{\varepsilon}$ ranges and ability to shape-form using a variety of metalworking equipment with controlled microstructures.

4. Ti-10V-2Fe-3Al

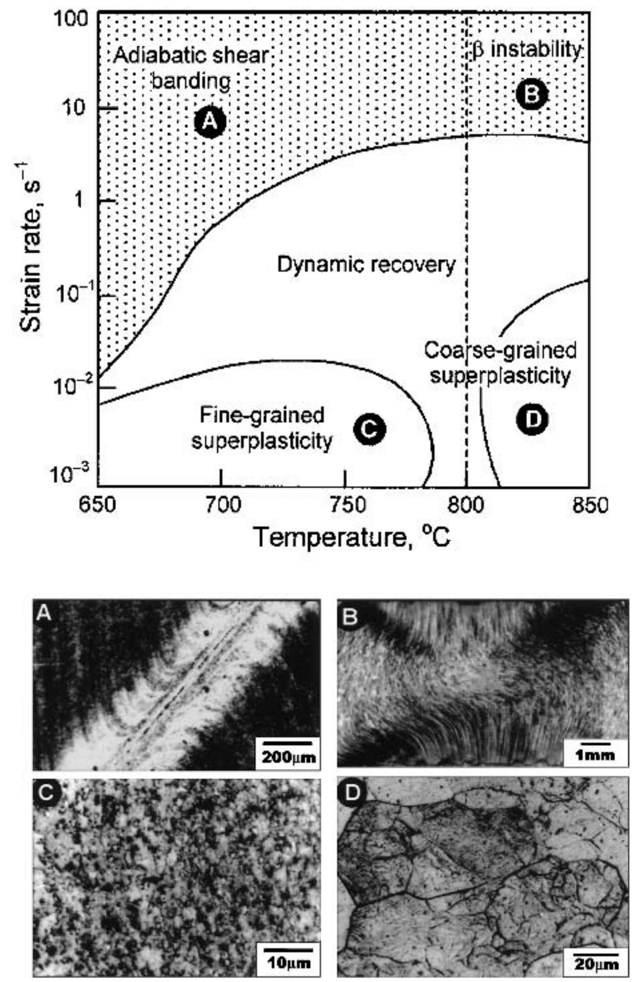
Ti-10V-2Fe-3Al is a near β alloy developed primarily for high-strength and toughness applications at temperatures up to 315 °C and tensile strengths in the range 1200-1400 MPa to provide weight savings over steels in airframe forging applications.[1] The processing map for Ti-10V-2Fe-3Al with a fine duplex $\alpha + \beta$ starting microstructure (fine α particles in an equiaxed β matrix) is shown in Fig. 9(a) along with the microstructures of specimens deformed under different T-$\dot{\varepsilon}$ conditions.[18] In the temperature range 650-775 °C and strain rates below 10^{-2} s^{-1}, Ti-10V-2Fe-3Al exhibits conventional fine-grained superplasticity, which occurs by grain boundary sliding with diffusional accommodation. The elongations to failure in this region at an initial strain rate of 10^{-3} s^{-1} are in the range 400-700% (Fig. 9b). In addition, the alloy exhibits extremely low flow stresses (~50 MPa at 700 °C) in the superplastic forming range. Above the β transus (~800 °C), Ti-10V-2Fe-3Al exhibits coarse-grained superplasticity with moderately high elongations (~300%). Microstructures of samples deformed in this region revealed a stable sub-grain structure (Fig. 9aD) within large β grains, which confirms that the mechanism of β superplasticity involves sliding of subgrain boundaries accommodated by anomalous diffusion in the β phase. For increasing the production rates, hot working can be performed at moderately high strain rates in the dynamic recovery region. As is the case for majority of the Ti alloys, Ti-10V-2Fe-3Al also exhibits large regimes of flow instabilities at higher strain rates (>1 s^{-1}), which are manifested as adiabatic shear bands (Fig. 9aA) and localized flow (Fig. 9aB). In summary, Ti-10V-2Fe-3Al offers the advantages of lower forging temperature, good superplastic forming characteristics, wide processing windows, and lower flow stresses. Moreover, these advantages allow usage of dies and tools of much cheaper alloy steels.

5. Low-Cost Beta Alloys

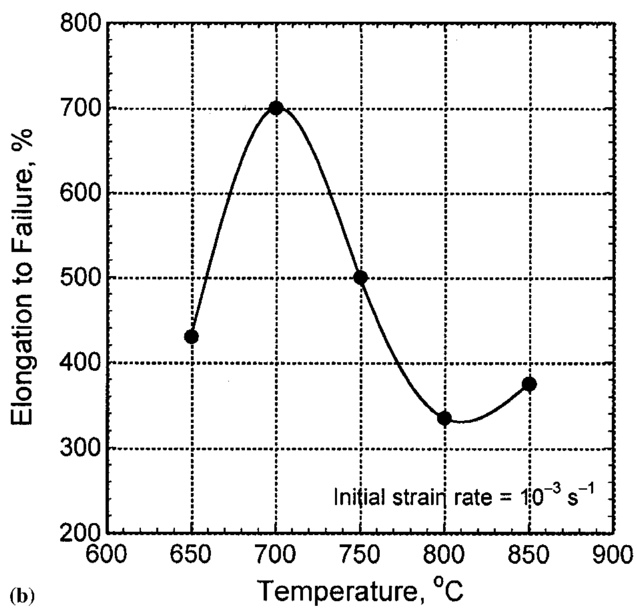
Low cost beta (LCB) alloys are being developed which can be processed to have high strength and low modulus and potential applications include springs and fasteners in aerospace and automobile industries, and armor components.[1] The alloy Ti-6.8Mo-4.5Fe-1.5Al (Timetal LCB) can be produced at less than half the cost of typical β alloys due to lower formulation cost from using ferro-molybdenum master alloys. Processing map obtained on the LCB with a single phase β starting microstructure in the ranges 700-950 °C and 10^{-3}-10 s^{-1} is shown in Fig. 10(a).[19] LCB alloy exhibits a wide safe processing window of DRX with excellent ductility (Fig. 10b). As for the microstructural control, processing should be restricted to strain rates below 5 s^{-1} to avoid flow instabilities. The β grain size versus deformation temperature is also plotted in Fig. 10(b), which follows a typical sigmoidal variation. Another variant of LCB is Ti-10V-4.5Fe-1.5Al, which has been developed by replacing Mo with equivalent amount of V. The processing map for hot working of modified LCB alloy[20] is shown in Fig. 11(a), which also exhibits a wide processing window of DRX similar to LCB. Variations of ductility and β grain size with deformation temperature are plotted in Fig. 11(b), which are very useful in optimizing the processing conditions. Predictions of average β grain size during hot working of LCB alloys can be made using the relationships

$$d_\beta,\ \mu\text{m} = 100\ Z^{-0.032} \text{ for LCB} \quad \text{(Eq 3)}$$

$$d_\beta,\ \mu\text{m} = 151.3\ Z^{-0.029} \text{ for modified LCB} \quad \text{(Eq 4)}$$



(a)



(b)

Fig. 9 **(a)** Processing map for hot working of Ti-10V-2Fe-3Al and **(b)** variation of elongation to failure with temperature in the superplastic forming range

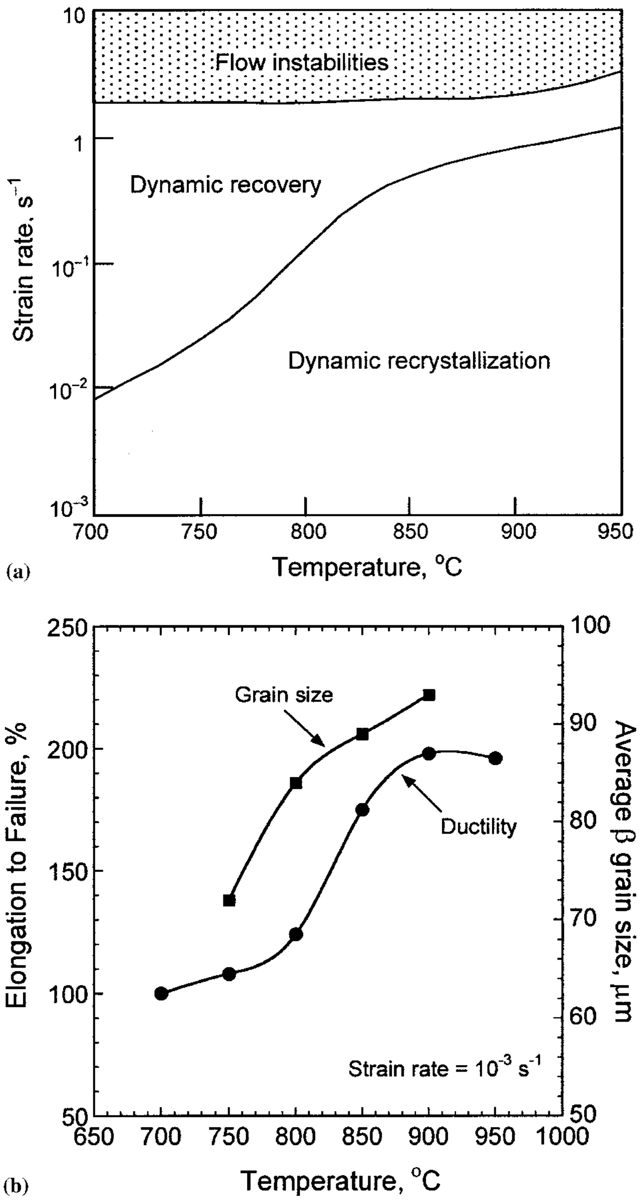


Fig. 10 **(a)** Processing map for hot working of Ti-6.8Mo-4.5Fe-1.5Al (LCB alloy) and **(b)** variation of ductility and grain size with temperature at a strain rate of 10^{-3} s^{-1}

where Z is defined in Eq 1. Q is 210 kJ/mol for LCB and 180 kJ/mol for modified LCB. These alloys can also be worked in the dynamic recovery region at higher strain rates away from DRX, but care should be taken to avoid instability regimes. The presence of DRX regions extending over wide ranges of temperature and strain rate indicates that LCB alloys possess excellent hot workability.

6. Ti-B Alloys

Titanium alloys modified with small additions of boron are emerging as potential candidates for replacing structural components requiring high specific stiffness and strength at room as well as moderately elevated temperatures.[21] The property

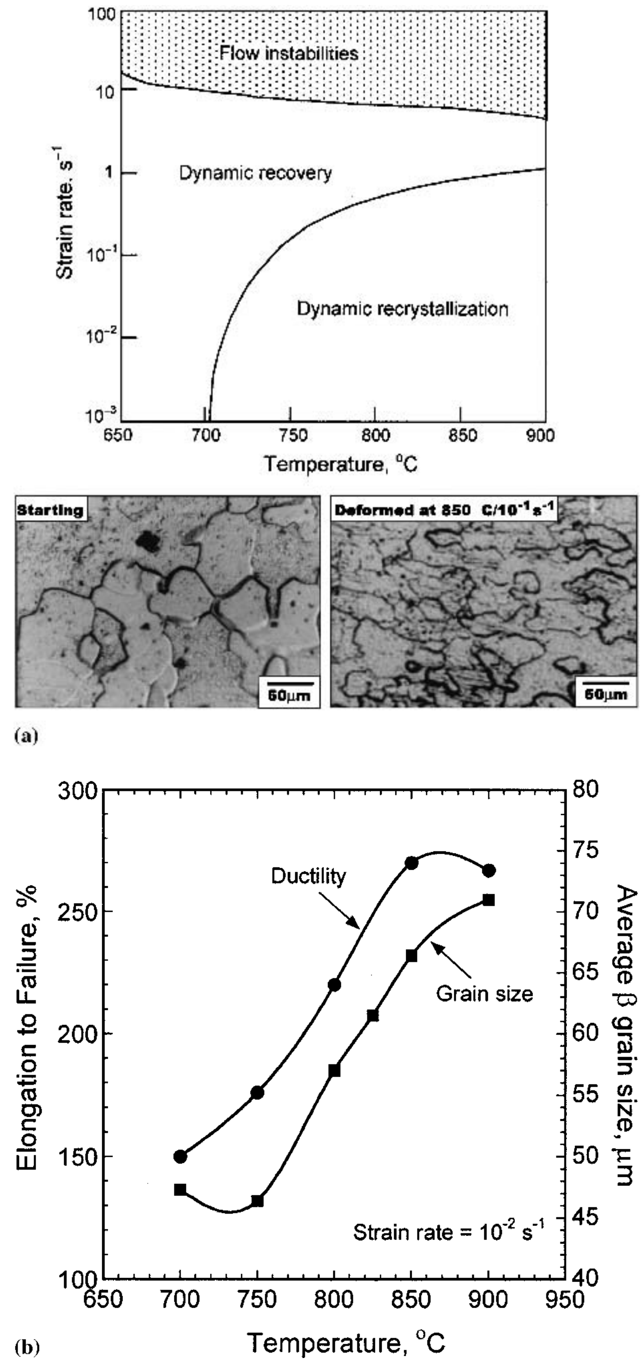


Fig. 11 **(a)** Processing map for hot working of Ti-10V-4.5Fe-1.5Al and **(b)** variation of ductility and grain size with temperature

enhancements are through the formation of fine, dispersed TiB precipitates. These precipitates in the boron-modified alloys are whiskers (TiB_w) formed *in situ* in the titanium matrix, which distribute uniformly and discontinuously and provide nearly isotropic properties, reaction-free interface, and ease of workability. Recent advances in synthesizing techniques enable cost-effective production of these alloys. A variety of techniques such as conventional casting, powder metallurgy, rapid solidification, and mechanical alloying have been used to produce these materials, and the final microstructural characteristics (e.g., grain size and morphology, TiB size, morphology, and distribution) sensitively depend on the processing method. Deformation processing is not only an essential step in the shape-forming of engineering components, but also causes significant modifications in the microstructure and a wide range of enhanced mechanical property combinations can be obtained in these materials. In this section, deformation processing of the most important Ti alloy Ti-6Al-4V modified with boron prepared by two different powder metallurgy approaches is discussed.

The processing map for hot working of Ti-6Al-4V-1.6B compact (equivalent TiB_w volume fraction = 10%)[22] produced by the pre-alloyed powder approach in the ranges 900-1200 °C and 10^{-3} to 10 s^{-1} is presented in the Fig. 12(a). The map reveals that at slow strain rates in the $\alpha + \beta$ range, Ti-6Al-4V-1.6B exhibits superplasticity marked by large elongations to failure and the behavior is very similar to Ti-6Al-4V (Fig. 4). A peak ductility of 335% was recorded at 950 °C (initial strain rate = 10^{-3} s^{-1}), which is the optimum temperature for superplastic forming this alloy. In the β phase field Ti-6Al-4V-1.6B also exhibits superplastic behavior with moderately high elongations (peak ductility of 250% at 1150 °C). The presence of TiB_w enables β superplasticity by stabilizing fine grain size, which otherwise rapidly grows to the order of few mm in Ti-6Al-4V. Grain boundary sliding as well as α or β/TiB_w interfacial sliding, with simultaneous diffusional accommodation are found to contribute to the superplasticity mechanisms. At high strain rates (>1 s^{-1}), Ti-6Al-4V-1.6B exhibits adiabatic shear banding at temperatures below 1000 °C and cavitation at the interfaces above 1150 °C, and these processing conditions must be avoided.

Processing map for hot working of Ti-6Al-4V-2.9B compact (equivalent TiB_w volume fraction = 20%)[23] produced by blending of powders is shown in Fig. 12(b). Ironically, the safe processing window for deformation processing of this alloy is very restricted. The only region to hot work this alloy is in the β phase field at slow strain rates where the deformation mechanism is either superplasticity or dynamic recovery, which is similar to Ti-6Al-4V-1.6B. A large regime of instability manifested in the form of cavitation at the TiB_w ends, leading to whisker fracture in some instances occurs. The intensity of these defects increases with increase in strain rate and decrease in temperature. The key differences between Ti-6Al-4V-2.9B and Ti-6Al-4V-1.6B are increased volume fraction of TiB, coarser TiB size, and coarse α grain size in the former. The occurrence of cavitation and whisker fracture in Ti-6Al-4V-2.9B is attributed to insufficient accommodation of the stresses generated at the rigid TiB_w by the matrix flow. Therefore, care must be taken while processing boron-modified Ti alloys by taking into account the microstructural characteristics.

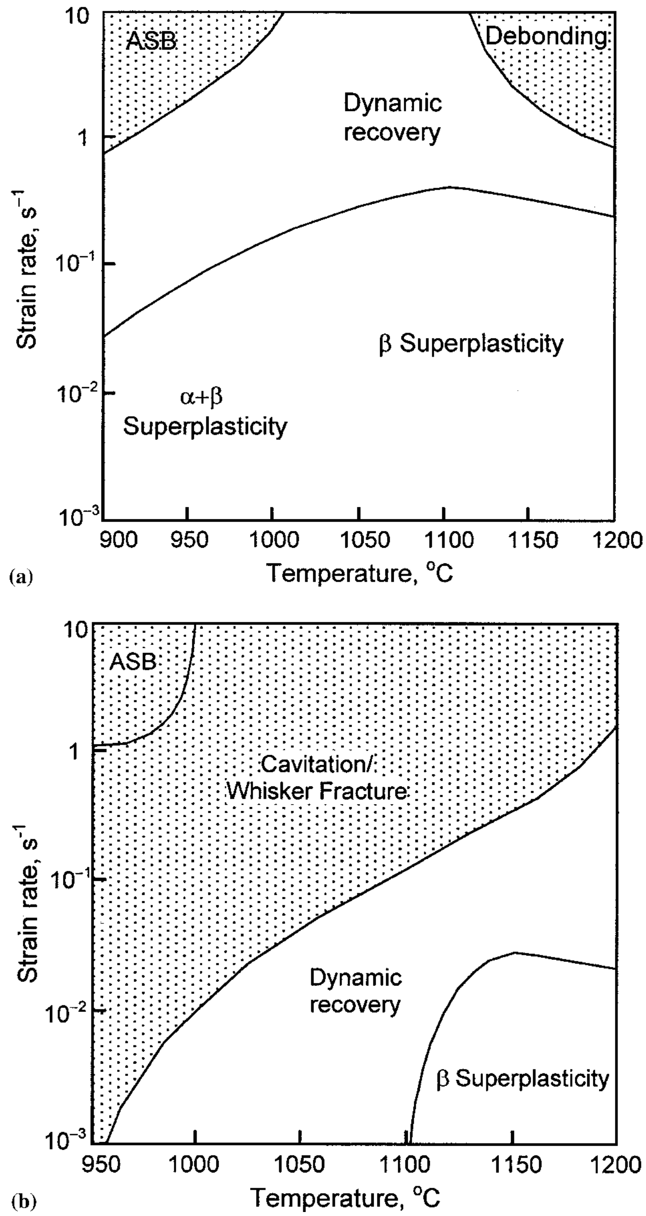


Fig. 12 Processing maps for hot working of Ti-6Al-4V modified with boron: **(a)** Ti-6Al-4V-1.6B prealloyed powder compact and **(b)** Ti-6Al-4V-2.9B blended powder compact

7. Summary

In summary, there have been a number of areas of Ti alloy deformation processing where significant progress has been made over the past decade. These areas include, but are not necessarily limited to the following:

- Application of process mapping and modeling to different titanium alloys has not only provided solutions to many processing problems but also insights into the mechanisms that lead to the optimization of processing conditions.
- The oxygen content and starting microstructure significantly change the deformation mechanisms of Ti-6Al-4V during hot working, and utmost attention on these two variables is required while designing process schedules.
- Careful control of process variables viz. temperature and strain rate is essential to avoid strain-induced porosity during $\alpha + \beta$ working of Ti-6Al-4V, and processing maps are very useful for identifying the safe processing limits. Integration of the intrinsic material behavior information with simulation and optimization schemes provides potential opportunities for improvement in process design while maintaining microstructural control.

- High-speed processing of titanium by exploiting the effect of deformation rate on the kinetics of phase transformation provides alternate conversion scheme which enhances the production rates significantly while maintaining microstructural control.

Acknowledgments

The authors would like to express their gratitude to Prof. Y.V.R.K. Prasad for introducing them to the field of deformation processing and for many stimulating discussions. Two of the authors (S.T. and R.B.B) are supported under the auspices of Air Force contracts F33615-99-C-5809 and F33615-03-D-5801.

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