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Controlled annealing of sandwich-structured aluminum AA1050 for optimized combinations of strength and ductility

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Abstract:

A heavily rolled AA1050 sample with a microstructurally continuous sandwich structure, characterized by distinct microstructural evolution in the center and subsurface layers, has been annealed at different temperatures for 2 h with the objective of establishing optimized combinations of strength and ductility. It is observed that a large reduction in the fraction of high angle boundaries taking place during recovery in the subsurface layers results in delayed onset of recrystallization compared to that in the center layer, where the change in the fraction of high angle boundaries during recovery is small. The different recrystallization rates in this sandwich structure facilitate control of the overall recrystallized fraction, and can therefore be advantageous in obtaining a desired combination of both strength and ductility. A good combination of moderate strength and intermediate ductility is obtained in the material annealed at 250 °C and 270 °C, where the area fractions of recrystallized microstructure in the center are 7% and 36%, respectively. By analyzing the dependence of mechanical strength on the

microstructure it is found that the mechanical properties can be described by a simple composite model using a rule of mixtures.

Keywords: electron microscopy; hardness; tensile properties; aluminum alloys; cold rolling; annealing.

1. Introduction

Deformation of metals to large plastic strains is accompanied by an extensive refinement of the microstructure, in extreme cases down to the nanoscale, resulting in high mechanical strengths [1–6]. However, such materials typically exhibit a very low ductility, as a result of their limited work-hardening capability. This leads to a low uniform elongation during tensile deformation, which correspondingly restricts the use of such metals in applications where sufficient formability is required.

It is known that a balance between strength and ductility can be sought by the use of annealing treatments, where the main processes involved are recovery of the deformation microstructure and recrystallization [6–10]. For aluminum, the former process can be regarded as uniform structural coarsening [11–15]. In contrast, the process of recrystallization leads in the early and intermediate stages (partial recrystallization) to a microstructure containing strain-free nuclei/grains in a coarsened (recovered) deformation microstructure. Such a structure is an example of a heterogeneous (or composite) microstructure, in which both hard regions providing high strength, and soft regions promoting ductility, and hence formability, are combined in an effort to achieve a required balance in mechanical properties. A similar philosophy underlies the interest and development of dual-phase materials, and of mono-phase systems with a bimodal or even multi-modal grain size distribution. Wang et al. [9], for example, reported that annealing of 93% cryo-rolled copper resulted in good ductility with only a marginal reduction of strength. This combination of high strength and high ductility was attributed to the formation of a bimodal microstructure, in which nano- and ultrafine (<300 nm) crystallites were combined with coarser (1–3 μm) recrystallized grains.

An alternative approach is to explore the use of sample scale variations in microstructure with a view to enhancing overall mechanical properties. This approach is typified by the use of mechanical

surface hardening to generate a microstructural gradient from the nanoscale at the surface to a conventional microstructure in the sample interior. Although a very good combination of strength and ductility has been reported for such samples [16–18], a common limitation is the difficulty in extending the methods to large-scale processing. In contrast, heavily rolled and annealed materials are produced on an industrial scale and are widely used in various applications.

In a recent study of aluminum deformed by heavy cold rolling to an ultrahigh von Mises strain of $\varepsilon_{\text{VM}} = 6.4$, it was observed that annealing at 300 °C resulted in a differing progress of recovery and recrystallization in the center layer and the two sandwiching subsurface layers [14]. Under certain annealing conditions microstructures were obtained consisting of a central partially recrystallized volume enclosed by mostly a recovered deformation matrix. Such a microstructure can be considered as a system exhibiting a simple form of macroscopic (sample scale) heterogeneity. The main motivation for the present work is to investigate the relationship between structural parameters and mechanical properties in such microstructures, with an aim of achieving a good combination of strength and ductility. For this purpose, aluminum cold rolled to $\varepsilon_{\text{VM}} = 6.4$ and annealed for 2 hours at temperatures in the range 130 to 400 °C is studied in this work. The material has been examined using electron backscatter diffraction (EBSD) to allow the characterization of both the recrystallized and recovered microstructure over large areas. The mechanical properties in this material have been probed on the local scale using hardness measurements, and on the sample scale by tensile testing. An important question also addressed in this work is whether the mechanical strength and ductility in such macroscopically heterogeneous systems can be fully accounted for by a simple “rule of mixtures” approach [19], or whether additional effects arising from mechanical constraints between the hard and soft regions affect the mechanical properties.

2. Experimental

A plate of aluminum AA1050 was cold rolled in multiple passes from 10 cm to approximately 0.4 mm ($\varepsilon_{\text{VM}} = 6.4$). The material was rolled with lubrication, unidirectionally by alternating the top and bottom sides between passes (for more details see [12,14]). The rolled sample was then annealed in air for 2 h at temperatures in the range 130 to 400 °C.

The microstructure and texture were studied in the longitudinal section containing the rolling direction (RD) and the normal direction (ND). After mechanical and electrochemical polishing the

samples were investigated using EBSD in a Zeiss Supra 35 field emission gun scanning electron microscope. A step size of either 25 nm or 30 nm was used for the EBSD analysis of the cold-rolled microstructure. Larger step sizes were applied for studying the annealed samples and for texture analysis. Microstructural parameters for the deformed and recovered structures were investigated at two different depths: in the center, and in the volume 30~130 μm from the surface. For texture analysis of all samples and for microstructural characterization of partially and fully recrystallized samples, the entire sample thickness was covered by EBSD. In this case, to separate the data for the center and subsurface layers, each through-thickness data set was divided equally into three subsets. The data from the two opposite subsurface layers were combined for calculation of microstructural parameters and fractions of different texture components.

A critical misorientation angle of 1.5° was applied to determine the boundary spacing and proportions of different boundary types. Low angle boundaries (LABs) and high angle boundaries (HABs) were defined as those with misorientation angles $\theta=1.5\text{--}15^\circ$ and $\theta>15^\circ$, respectively. Recrystallized grains were identified based on the method described in [20]. In the present work, such grains were defined as regions greater than 3 μm with internal misorientations less than 1° , which were separated from the deformed/recovered matrix by both LABs and HABs. Fractions of different texture components were calculated applying a 15° deviation from the closest $\{hkl\}\langle uvw \rangle$ orientations.

Vickers hardness measurements were performed in the RD-ND section with a load of 10 g both along the mid-thickness and in the subsurface. These measurements were analyzed only for the specimens annealed at temperatures up to 270°C , for which the distance between the specimen edge and the center of an impression in the subsurface layers was at least 3 times the impression diagonal. Tensile tests were carried out at room temperature by pulling specimens with a cross-sectional area of 5 mm^2 and a gauge length of 50 mm along the RD at an initial strain rate of $1.5 \times 10^{-3}\text{ s}^{-1}$ and a constant crosshead speed. Two specimens were tested for each condition.

3. Results

3.1. Cold-rolled condition

The deformation microstructure after cold-rolling has been described in detail in Ref. [14]. Here we give only a short summary of the salient key features. The microstructure is a typical lamellar

structure with extended boundaries almost parallel to the rolling plane. In this microstructure, the average boundary spacing along the ND (d_{ND}) is $\sim 0.2 \mu\text{m}$ and the fraction of HABs is above 50% in both the center and subsurface layers (Table 1). The main difference between the center and subsurface is in the texture associated with these layers (see Fig.1(a), Fig.2(a) and Table 1). The center layer of the cold-rolled sample contains a typical rolling texture with a dominant S $\{123\}\langle 634 \rangle$ component and smaller fractions of the copper “Cu” $\{112\}\langle 111 \rangle$ component and brass “Bs” $\{110\}\langle 112 \rangle$ components (Table 1). In contrast, the texture in the subsurface is dominated by one of the symmetric variants of an orientation near the Cu component (see [14] for more details). Correspondingly, the center layer contains broad texture bands consisting of the S-oriented lamellae, whereas in the subsurface the broad texture bands are of the Cu component, as seen in Fig.2(a).

Table 1. Parameters of the microstructure and texture in the cold-rolled material as measured using EBSD.

Layer	d_{ND} (μm)	Fraction of HABs (%)	Fractions of rolling texture components (%)		
			Cu	S	Bs
Center	0.24	54	28	53	15
Subsurface	0.23	56	50	38	4

3.2 Changes in the microstructure during annealing

Annealing of the cold-rolled material at temperatures from 130°C to 250°C for 2 h results in a process of predominantly recovery as shown in Fig.1(b,c) and Fig.2(b,c). During recovery the microstructure coarsens by triple junction motion [13–15], largely retaining the lamellar morphology of the deformed microstructure. The average aspect ratio in this coarsened recovered matrix remains fairly large (between 1.8 and 3, see Fig.3(b)). A number of almost equiaxed subgrains are also observed within the lamellar structure after annealing at 250°C and above (Fig.2(c,d)). The coarsening of the deformed microstructure is accompanied by a reduction in the fraction of HABs, which is more significant in the subsurface than in the center (Fig.4) as a result of the different local texture evolution during recovery in each of these regions [14]. The availability of fewer crystallites surrounded by

HABs in the recovered microstructure of the subsurface layers results in delayed onset of recrystallization compared to that in the center layer.

Figures 1 and 5 demonstrate that pronounced recrystallization takes place first in the center layer, occurring at a temperature of 270 °C, although a small volume fraction, $f_{\text{Rex}} = 3\%$ and $f_{\text{Rex}} = 7\%$, of recrystallized grains are also present in this layer in the samples annealed at 220 °C and 250 °C, respectively. After annealing at 285 °C, the center is almost fully recrystallized ($f_{\text{Rex}} = 96\%$), while the subsurface is slightly less recrystallized ($f_{\text{Rex}} = 88\%$). It can be noted also that the recrystallized grains in the subsurface are considerably larger than those in the center of the sample, see Fig.1(e). This difference is also found in the sample annealed at 300 °C, in which the entire sample volume is almost fully recrystallized (Fig.1(f)).

The crystallographic texture, dominated initially by the rolling texture components, is also affected by the annealing treatments (Fig.6). The total volume fraction of the rolling texture components slightly increases during recovery, with a pronounced strengthening of the Cu-component at the expense of the S- and Bs-components in the subsurface (see Fig.6(b)), and then decreases sharply with the onset of recrystallization. In the center, the recrystallized grains have orientations related primarily to the rolling texture, whereas the recrystallized grains in the surface are primarily of P {011}⟨566⟩, ND-rotated cube “Cube_{ND}” {001}⟨310⟩ and random orientations (Fig.1(d-f)). The origin of this difference in texture has been described in Ref. [14].

3.2 Hardness

Softening during annealing investigated by the use of Vickers hardness measurements in the center and subsurface layers is demonstrated in Fig.7. Up to an annealing temperature of 250 °C the subsurface and the center have a similar hardness, which decreases with increasing annealing temperature as a result of coarsening during recovery. However, in the sample annealed at 270 °C, the difference in hardness is significant: the center, with a large number of coarse recrystallized grains, is considerably softer than the subsurface where the microstructure is still mostly recovered.

3.3. Tensile test data

Stress-strain curves for the cold-rolled and several annealed samples are presented in Fig.8. It is evident that the cold-rolled sample has a very high ultimate tensile strength (UTS) of 196 MPa and a low uniform elongation of 1.6%. The uniform elongation remains low in the samples annealed at temperatures below 250 °C. However, the sample annealed at 250 °C shows a significant increase in elongation, with only a very small amount of work-hardening. The samples annealed at temperature of 270 °C and above exhibit extensive elongation, combined with a significant amount of work hardening. The differences in the tensile flow curves for all the samples are summarized in Fig.9, based on which the evolution of mechanical properties of this material during annealing for 2 h can roughly be divided into three temperature ranges:

- (i) a low-temperature range (130–220 °C) characterized by a gradually reduced strength with no increase in elongation. This combination can be described as high strength–low ductility (HSLD);
- (ii) an intermediate-temperature range (250–270 °C) with a continued gradual reduction in strength and an appreciably improved ductility (9~15%). Thus, this range enables a combination of moderate strength and improved ductility (MSID); and
- (iii) a high-temperature range (285–400 °C) resulting in low strength (UTS $\approx$ 90 MPa) and large uniform elongations (23~28%), i.e. low strength–high ductility (LSHD).

4. Discussion

To rationalize the evolution of mechanical properties during annealing, it is necessary to consider the processes taking place in the microstructure in the annealing temperature range. For the present material, these processes include either only recovery, or recovery and recrystallization taking place concurrently. As the tensile test data represent the mechanical response over the entire thickness of the sample, it is useful to consider first how the mechanical behavior varies as a function of the fraction recrystallized throughout the sample thickness. It should be recalled, however, from the experimental observations described above that the progress of recrystallization varies in the different layers and that the overall mechanical properties may therefore also depend on this sample scale heterogeneity.

The mechanical data are replotted in Fig.10 as a function of total recrystallized fraction. From this figure it can be seen that a combination of moderate strength and a reasonably high ductility (15%) is achieved when f_{Rex} reaches approximately 20% (in the sample annealed at 270 °C). It is interesting to note that a gain in ductility is first observed after annealing at 250 °C, when the overall recrystallized fraction is only 4.7%, and that a significant drop in mechanical strength occurs as a result of the microstructural changes taking place mostly during recovery. This intermediate-temperature range, where improved ductility is combined with values of both the 0.2% proof stress and UTS above 50% of the values recorded in the cold-rolled material, is observed to be rather narrow in Fig.10.

In the following, we focus on the two most interesting regimes in mechanical behavior, namely the changes occurring during recovery, and the intermediate-temperature regime covering the annealing conditions, where good combinations of strength and ductility are observed.

4.1. Changes in mechanical properties during recovery

Based on our microstructural characterization, it is possible to investigate further the origin of the changes in mechanical properties during annealing of the samples. To determine an average linear spacing d_{Rec} for analysis of the contribution to the yield strength from the recovered microstructure, the subsurface and center regions are considered separately. For each region, the boundary area per unit volume (S_V) is calculated as $S_V = 1/d_{\text{ND}} + \pi/2d_{\text{RD}}$, based on the assumption that the lamellar boundaries sampled along ND line-transects are parallel to the rolling plane, and that the interconnecting boundaries sampled along RD line-transects have random boundary plane inclinations [21]. The volume weighted contributions from both the center (C) and subsurface (SS) regions can then be combined as $S_V^{\text{tot}} = (1/3)S_V^{\text{C}} + (2/3)S_V^{\text{SS}}$, from which a mean equivalent linear spacing is calculated as $d_{\text{Rec}} = 2/S_V^{\text{tot}}$.

A Hall-Petch plot showing the relationship between d_{Rec} and the 0.2% proof stress is presented in Fig.11, using only data for the as-deformed material, and samples annealed up to 250 °C (where the microstructure is predominantly in the recovered state). A linear fitting of the data gives an intercept (at $d_{\text{Rec}} = \infty$) of $\sigma_0 = 23$ MPa with a slope of $83 \text{ MPa } \mu\text{m}^{0.5}$. Note that in accordance with similar calculations made on Ni [4] and Cu [22] this should be described as an effective Hall-Petch slope since the analysis does not include the contribution from incidental dislocation boundaries with misorientations less than 1.5° because of the limited angular resolution of the EBSD data.

The loss of strength during annealing prior to the onset of recrystallization can be explained therefore by the observed structural coarsening, which in this highly deformed sample takes place with a relatively low activation energy by a process of triple junction motion [13–15]. The results highlight the fact that control of this coarsening process is important for maintaining a high strength during low-temperature annealing.

Considering ductility, it is low in the conditions obtained in the low-temperature range (130–220 °C), where the characteristic feature is the lamellar structure. Although coarsening within this structure results in decreased strength, the ductility is not improved as long as this lamellar structure is retained without appreciable development of equiaxed subgrains. The rather small fraction recrystallized in the sample annealed at 250 °C alone is unlikely to result in the considerable improvement of ductility observed in this condition. It is significant that this improvement occurs when large and almost equiaxed subgrains are present in the recovered microstructure, which provides indirect evidence that such subgrains can contribute to the improved ductility in this condition. It is possible that there is also some unseen (using EBSD) evolution in the cell interior dislocation density, or in the dislocation boundary content that also has a positive effect for the improved ductility.

4.2 Mechanical properties in partially recrystallized samples

For any material that can be regarded as a composite of components with different mechanical properties, it is interesting to establish the extent to which the properties can be accounted for by a simple additive (also referred to as a rule of mixtures) model. For the samples annealed in the low- to intermediate-temperature range, a simple test in this regard is shown in Fig.12, where the macroscopic UTS data obtained from tensile testing are compared against an effective sample hardness calculated from Vickers hardness measurements in the center and subsurface layers. The linear fit of the data supports the conclusion that, at least for the chosen temperature range, the sample hardness is directly proportional to the UTS, implying purely additive strengthening contributions from the center and subsurface regions following a rule of mixtures model.

Furthermore, by using the value for the effective Hall-Petch constant established in section 4.1, it is possible to examine whether the strength of the partially recrystallized samples can also be accounted for by a rule of mixtures model. For the recovered fraction ($1 - f_{\text{Rec}}$), a value for the mean equivalent linear spacing of deformation boundaries, d_{Rec} , can be obtained as in section 4.1 from

experimental measurements of d_{ND} and d_{RD} in non-recrystallized areas. The strength can then be calculated from $\sigma_{Rec} = \sigma_0 + k_{HP}d_{Rec}^{-0.5}$ with $\sigma_0 = 23$ MPa and $k_{HP} = 83$ MPa $\mu m^{0.5}$. For the recrystallized fraction, the strength σ_{Rex} is taken as a fixed value of 26 MPa (based on the experimental 0.2% proof stress data). The predicted strength σ_{ROM} based on the rule of mixtures model is then given by

$$\sigma_{ROM} = (\sigma_{Rex} \times f_{Rex}) + (\sigma_{Rec} \times (1 - f_{Rex})) \quad \text{Eq. (1)}$$

Values calculated using Eq.(1) for three partially recrystallized samples with considerable f_{Rex} are listed in Table 2, together with the experimentally determined values for the same samples. In each case, the yield strength determined from the rule of mixtures calculation is in close agreement with the experimentally measured values. This result is consistent with those obtained by Joshi et al. [19] on materials with bimodal microstructures.

Table 2 Parameters of the microstructure calculated from the EBSD data and strength for several partially recrystallized samples. Here $\Delta\sigma$ is defined as $\sigma_{ROM} - \sigma_{0.2\%}$.

Temperature (°C)	d_{Rec} (μm)	f_{Rex} (%)	σ_{ROM} (MPa)	$\sigma_{0.2\%}$ (MPa)	$\Delta\sigma$ (MPa)
250	0.86	4.7	109	109	0
270	0.98	18	92	86	6
285	1.28	91	33	35	-2

Another interesting observation in the partially recrystallized samples is the large variation in work hardening and ductility. These mechanical properties show a strong dependence on the observed recrystallized fraction. Only a small amount of work hardening during plastic deformation and a moderate uniform elongation of around 9% is observed in the sample annealed at 250 °C (with only a few percent recrystallized volume in both the center and subsurface regions – see Fig.5). As the annealing temperature is increased above 250 °C, both work hardening capability and ductility increase. The sample annealed at 300 °C shows a typical stress-strain curve for a recrystallized material, with a transition from parabolic to linear work-hardening. Qualitatively, the evolution of mechanical properties as a function of recrystallized fraction observed in our study is similar to that in

95% rolled and annealed AA1050 studied by Sun et al. [8], though our material, due to the higher imposed strain, is characterized by a higher strength in the as-rolled and recovered conditions. Additionally, it should be noted that f_{Rex} was measured in [8] based on optical microscopy images, where only comparatively large recrystallized grains could be distinguished. It can be noted also that there is a large difference in f_{Rex} between the center and subsurface layers (38% and 8%, respectively) in our material annealed at 270 °C, suggesting that both the overall fraction and spatial distribution of recrystallized grains may also be important in determining the balance between strength and ductility, which is discussed further in the next subsection.

4.3 Balance of strength and ductility

All combinations of UTS and uniform elongation measured in the present experiment are plotted in Fig.13 along with the data for heavily rolled/ARB-processed and annealed samples of the AA1050 alloy presented by Sun et al. [8] and Su et al. [5]. It is apparent that despite different thermo-mechanical histories and different microstructures produced, each data set falls into one of the three combinations defined in the present work, i.e. LSHD, MSID or HSLD, resembling a so-called “banana” curve. There seems to be a general inverse correlation between strength and ductility for bulk aluminum samples processed by rolling/ARB and annealing. Thus, these aluminum samples follow the general inverse correlation between strength and ductility seen in a wide range of systems [7,10,23,24].

Implicit in this relationship with regard to the optimization of properties is the ability to control both the coarsening kinetics of recovery in the deformation microstructure, and the onset and extent of recrystallization. These processes are naturally interconnected since recovery of the deformation microstructure removes part of the stored energy of deformation that acts as a driving force for recrystallization and also softens the material. This is seen in the present study from the significant decrease in hardness of samples annealed at 220 °C, where the microstructure has coarsened substantially and where f_{Rex} remains at only a few percent. One possible approach for improved control of this coarsening is through introduction of particles, which have been reported to act as pinning centers to slow down coarsening by triple junction motion [15].

The analysis of the strength contributions from the center and subsurface layers in the partially recrystallized conditions suggests that the mechanical properties can be described by a simple composite model using a rule of mixtures, without significant additional gradient plasticity effects.

Nevertheless, the sandwich structure of the aluminum sample with the different coarsening and recrystallization rates in the different layers can facilitate control of the overall recrystallized fraction, and can therefore be advantageous in obtaining a desired combination of both strength and ductility. In the present material, only moderate, yet still significant, differences in terms of recovery and recrystallization are seen between the center and subsurface layers. It is expected that other rolling procedures, e.g. rolling with high surface friction [25,26] or asymmetric rolling [27–29], can result in other sandwich structures that can take further advantage of sample scale heterogeneities.

Concluding remarks

Rolling of aluminum to a large plastic strain of $\varepsilon_{VM} = 6.4$ results in a microstructurally continuous sandwich structure, in which the center (middle 1/3) and subsurface layers (upper and lower 1/3) exhibit differing microstructural coarsening during annealing. In particular, recovery in the subsurface layers leads to a large decrease in the fraction of high angle boundaries, resulting in delayed onset of recrystallization compared to that in the center layer. The effect of the sample scale variation in the microstructure on mechanical properties in this sandwich-structured aluminum has been explored by mechanical tests of samples annealed for 2 h in the temperature range 130–400 °C. A good combination of moderate strength and intermediate ductility is obtained in the range 250–270 °C. The higher end of this temperature range corresponds to a regime where recrystallization takes place preferentially in the center layer of the sample.

Analyses of the dependence of mechanical strength on the contributions from the center and subsurface layers and on the contributions from the recovered and recrystallized microstructures shows that the strength can in each case be described by a simple rule of mixtures. The different coarsening and recrystallization rates in the different layers of the continuous sandwich structure of the aluminum sample nevertheless facilitates control of the overall recrystallized fraction, and can therefore be advantageous in obtaining a desired combination of both strength and ductility.

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Fig.1. Through-thickness orientation maps illustrating the spatial distribution of different texture components in the cold-rolled sample (a) and after annealing for 2 h at 130 °C (b), 250 °C (c), 270 °C (d), 285 °C (e) and 300 °C (f): Orientations which do not belong to any orientation listed in the color key are represented by different shadings of gray. White and black lines show misorientations greater than 3° and 15°, respectively. The rolling direction is horizontal.

Fig. 2. EBSD maps showing the evolution of lamellar structures in the center and subsurface layers: (a) cold-rolled sample; (b) sample annealed for 2 h at 130 °C; (c) sample annealed for 2 h at 250 °C; (d) sample annealed for 2 h at 270 °C. Orientations which do not belong to any orientation listed in the color key are represented by different shadings of gray. White lines correspond to 1.5–15° misorientations. Bold black lines show misorientations greater than 15°. The rolling direction is horizontal.

Fig. 3. Parameters for subgrains in the non-recrystallized microstructure: (a) average boundary spacing along the ND; (b) average aspect ratio.

Fig. 4. Fractions of HABs measured after annealing for 2 h at different temperatures.

Fig. 5. Area fractions of the recrystallized microstructure in the subsurface and center layers as a function of the annealing temperature.

Fig.6. Fractions of different texture components in the center (a) and subsurface (b).

Fig. 7. Evolution of Vickers hardness measured in the center and subsurface layers after annealing at different temperatures for 2 h.

Fig. 8. Representative engineering stress-strain curves for the cold-rolled (CR) sample and several samples annealed at different temperatures for 2 h.

Fig. 9. Evolution of the 0.2% proof stress, UTS and uniform elongation during annealing at different temperatures for 2 h.

Fig. 10. The 0.2% proof stress, UTS and uniform elongation plotted against the area fraction of recrystallized grains in the entire sample thickness.

Fig. 11. Dependence of the 0.2% proof stress on average boundary spacing in the as cold-rolled and predominantly recovered samples (f_{Rex} for the four data points from left to right are 4.7%, 2.6%, 0% and 0%).

Fig. 12. Relationship between the UTS measured after annealing in the low- to intermediate-temperature range and effective hardness of the entire sample calculated from Vickers hardness measurements in the center (C) and subsurface (SS) layers as $(1/3)HV^C + (2/3)HV^{SS}$. The solid line is a linear fit to the data points represented by gray symbols.

Fig. 13. Relationship between UTS and uniform elongation for rolled/ARB-processed and annealed AA1050 samples. LSHD, MSID and HSLD ranges correspond to low strength-high ductility, moderate strength-intermediate ductility and high strength-low ductility, respectively.

Fig. 1

Fig. 2

Fig. 3

Fig. 4

Fig. 5

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Fig. 7

Fig. 8

Fig. 9

Fig. 10

Fig. 11

Fig. 12

Fig. 13