Effects of Different Thermomechanical Processing on Mechanical Properties and Microstructure of Al-Mg-Si-Cu Alloy Sheet (Postprint)

Authors: Zhang Yan, Guo Mingxing, Xing Hui, Wang Fei, Wang Xiaofeng, Zhang Jishan, Zhuang Linzhong

Date: 2023-03-19T00:00:00+00:00

Abstract

Through tensile testing combined with OM, SEM, TEM observations and EBSD characterization, the influence of different thermomechanical processing routes on the mechanical properties and microstructure (including texture) of Al-Mg-Si-Cu alloy sheets was investigated. The results indicate that variations in thermomechanical processing have essentially no effect on the strength and strain hardening exponent n of T4P pre-aged alloys, but exert significant influence on the average plastic strain ratio r, planar anisotropy Δr, and elongation in different directions. Alloy sheets fabricated by applying a certain amount of cold rolling deformation followed by annealing after hot rolling (Process II) exhibit superior formability after solution treatment compared to those subjected to direct annealing after hot rolling (Process I), with r reaching 0.6187 and a pronounced reduction in anisotropy. Although the PSN effect is significant during solution treatment for Process I alloy sheets, the cold rolling reduction prior to solution treatment and the distribution of various-sized particles in Process II are rationally designed, resulting in recrystallized grains that are essentially equiaxed and contain only weak CubeND, Cube, and H textures. Based on the observed effects of thermomechanical processing on the microstructure of alloy sheets, a schematic diagram of the microstructure evolution model for this alloy system during thermomechanical processing is proposed.

Full Text

Preamble

Acta Metallurgica Sinica, Vol. 51, No. 12, December 2015, pp. 1425-1434

Influence of Different Thermomechanical Processes on the Mechanical Properties and Microstructure of Al-Mg-Si-Cu Alloy Sheets

Yan Zhang, Mingxing Guo, Hui Xing, Fei Wang, Xiaofeng Wang, Jishan Zhang, Linzhong Zhuang
(State Key Laboratory for Advanced Metals and Materials, University of Science and Technology Beijing, Beijing 100083)

Abstract
To reduce the weight of car body, Al-Mg-Si-Cu alloys are becoming increasingly attractive as a candidate for material substitution used to produce the outer body panels of automobiles because of their favorable bake-hardening response. However, the formability still needs to be further improved compared to steels. In this work, the effect of the thermomechanical processing on the mechanical properties and microstructure of Al-Mg-Si-Cu alloy is studied through tensile test, OM, SEM and TEM observation, as well as EBSD characterization. The results reveal that there is almost no change in both strengths and strain-hardening exponent n of the sheets in T4P condition after different thermomechanical processing, but the average plasticity strain ratio r, planar anisotropy Δr and elongations in the three directions show obvious differences. The sheet undergone hot rolling, cold rolling, intermediate annealing, cold rolling and solution (processing II) has a better formability ($r = 0.6187$) and a weaker planar anisotropy than that subjected to hot rolling, intermediate annealing and then cold rolling before solution treatment (processing I). Although the particle stimulated nucleation (PSN) effect of processing I is remarkable during solution treatment, due to the appropriate controlling cold deformation and distribution of second-phase particles with different sizes in processing II, most of the recrystallization grains are equiaxial and the recrystallization texture is only consisted of CubeND, Cube and H with a low intensity. At last, according to the relationship between the microstructure and the thermomechanical processing, the microstructure evolution model during different thermomechanical processes is established.

Keywords
Al-Mg-Si-Cu alloy, thermomechanical processing, formability, recrystallization texture, modelling

Introduction

With increasing emphasis on energy conservation and emission reduction worldwide, automotive lightweighting has become a critical direction for next-generation vehicle development. Aluminum alloys have emerged as important materials for automotive lightweighting due to their low density, corrosion resistance, high specific strength, ease of coloring, and high recyclability. Among them, age-hardenable Al-Mg-Si series alloys are particularly attractive because of their excellent rapid age-hardening response, which ensures relatively low strength during stamping and forming, followed by substantial strength enhancement after high-temperature paint baking, thereby exhibiting superior dent resistance. Consequently, these alloys have been widely adopted for manufacturing automotive body outer panels in recent years.

Although Al-Mg-Si series alloys possess these excellent comprehensive properties, their stamping formability still requires further improvement compared to conventional automotive steel sheets [?]. Numerous factors influence the stamping formability of aluminum alloy sheets, but alloy composition and thermomechanical processing are the most critical. The typical manufacturing process for automotive outer panel Al-Mg-Si series alloys follows this sequence: casting → homogenization → hot rolling → intermediate annealing → cold rolling → solution treatment → pre-aging → paint baking. This lengthy process flow means that variations in any stage can significantly affect the microstructure and mechanical properties of the final sheet [?]. Previous studies [?] have demonstrated that when a certain number of coarse particles (diameter $d > 1\,\mu$m) are distributed in the alloy matrix, substantial strain energy accumulates around these particles during deformation. During subsequent high-temperature heat treatment, recrystallization nucleation preferentially occurs around these particles, producing the so-called particle stimulated nucleation (PSN) effect. Fully exploiting the PSN effect can yield favorable recrystallized microstructures and texture distributions, leading to improved stamping formability during processing. Therefore, for Al-Mg-Si-Cu alloy sheets, obtaining micron-scale coarse particles is key to utilizing the PSN effect for regulating microstructure and mechanical properties.

Our previous research [?] found that adding Fe and Mn solute elements to Al-Mg-Si-Cu alloys can form Fe-rich coarse particles, and as the Fe and Mn content increases, the number of these Fe-rich phase particles continuously increases, gradually enhancing the PSN effect during recrystallization. In high-concentration alloys, the recrystallized grain size becomes refined after recrystallization, and the plastic strain ratio r, which characterizes stamping formability, also improves to a certain extent (maximum $r_{\max} = 0.578$). However, due to the high Fe and Mn content in these concentrated alloys, coarse Fe-rich phase particles easily form a network distribution during melting and casting [?]. Although these particles can be partially broken during subsequent thermomechanical processing, microcracks readily form around them, which is extremely detrimental to improving the stamping formability and hemming performance of the alloy sheets [?]. Therefore, significantly increasing the concentration of Fe-rich phase particles to regulate the final microstructure, texture, and formability involves numerous issues and still requires further optimization of the manufacturing process, particularly the melting and casting process.

Given this situation, if the thermomechanical processing of Al-Mg-Si-Cu alloys can be controlled to form a certain number of particles with different sizes (both coarse and fine) during processing, the synergistic effects of these differently sized particles on recrystallization nucleation and grain growth can be fully utilized to optimize the microstructure, texture, and formability of the alloy sheets. In this work, we selected an Al-Mg-Si-Cu alloy with relatively low Fe and Mn solute content and varied the thermomechanical processing parameters. We found that a certain number of differently sized particles (such as Mg2Si, Fe-rich phases, and Si particles) indeed formed in the alloy matrix. When these particles were appropriately distributed, they could effectively influence the mechanical properties and microstructure evolution, ultimately imparting excellent stamping formability to the alloy sheets.

Through systematic investigation of the influence of different thermomechanical processes on the mechanical properties and microstructural evolution of Al-Mg-Si-Cu alloys, this work aims to provide a theoretical basis for the development, processing, and application of aluminum alloys for automotive lightweighting.

Experimental

The experimental Al-Mg-Si-Cu alloy had a composition of Al-0.6Mg-0.9Si-0.2Cu-0.1Mn (mass fraction, %). Raw materials included high-purity Al (99.99%), commercial-purity Mg, commercial-purity Zn, and master alloys Al-20%Si, Al-50%Cu, Al-10%Mn, and grain refiner Al-5%Ti-1%B (mass fraction). The materials were melted in an SG2-12-10 resistance crucible furnace in a specific order, and the alloy melt was poured into a water-cooled steel mold with dimensions of 180 mm × 110 mm × 90 mm. The ingots were cropped and surface-machined, then subjected to a two-stage homogenization treatment at 485 °C for 3 h + 555 °C for 16 h. The hot rolling start temperature was 550 °C, with a finish temperature below 300 °C and a final thickness of 4 mm. The samples were then divided into two groups: the first group underwent intermediate annealing at 400 °C for 1 h, cold rolling to 1 mm, followed by salt bath solution treatment at 555 °C for 2 min and pre-aging at 80 °C for 12 h, designated as processing I; the second group was cold-rolled to 2 mm (50% reduction) after hot rolling, then subjected to intermediate annealing at 400 °C for 1 h, followed by cold rolling to 1 mm and the same solution treatment and pre-aging, designated as processing II. Alloy samples processed by both routes were stored at room temperature for 14 days (T4P condition) before mechanical property testing in different directions.

Tensile specimens were cut at 0°, 45°, and 90° to the rolling direction according to GB/T3076-1982 specifications and tested on an MTS810 electro-hydraulic servo material testing machine at a crosshead speed of 3 mm/min. Metallographic specimens were mechanically polished, etched with Keller' s reagent, and examined using an Axio Imager A2 optical microscope (OM). Phase analysis was performed using a SUPRA 55 scanning electron microscope (SEM) equipped with an energy-dispersive spectrometer (EDS). Microstructural observation was conducted on a Tecnai G2 F30 field-emission transmission electron microscope (TEM); samples were prepared by mechanical thinning followed by electrolytic double-jet polishing (electrolyte volume ratio HNO_3:CH_3OH = 1:2). Texture measurements were performed on an LTRA55 SEM equipped with an electron backscatter diffraction (EBSD) system at an accelerating voltage of 20 kV, working distance of 15–20 mm, and 70° tilt; data acquisition was computer-controlled, and three-dimensional orientation distribution functions (ODF) were calculated using a two-step method and presented as constant Euler angle ϕ_2 ($\Delta\phi_2 = 5°$) sections. Samples were prepared by mechanical polishing followed by electrolytic polishing (electrolyte: $HClO_4$:CH_3CH_2OH = 5:95, voltage 20 V).

2.1 Effect of Thermomechanical Processing on Mechanical Properties

[Figure 1: see original paper] shows the stress-strain curves of T4P-treated Al-Mg-Si-Cu alloy sheets processed by the two different thermomechanical routes. The detailed mechanical properties are listed in . It can be seen that for both processing routes, the highest strength and elongation occur along the rolling direction, while the lowest values are observed along the transverse direction, indicating certain anisotropy. However, the alloy sheet processed by route II exhibits significantly lower anisotropy. Additionally, processing II yields a higher average plastic strain ratio r ($r = (r_0{}^\circ + 2r_{45}{}^\circ + r_{90}{}^\circ)/4$, where $r_0{}^\circ$, $r_{45}{}^\circ$, and $r_{90}{}^\circ$ are the plastic strain ratios at 0°, 45°, and 90°, respectively) and lower planar anisotropy Δr ($\Delta r = (r_0{}^\circ + r_{90}{}^\circ - 2r_{45}{}^\circ)/2$), indicating better formability. Both processes have essentially no effect on the strain-hardening exponent n (Table 1).

Furthermore, the stamping formability of alloy sheets can be characterized by the limiting drawing ratio (R_{LD}). Numerous studies [?] have established a quantitative relationship between R_{LD}, r, and n:

$$R_{LD} = e^n \cdot \sqrt{\frac{1+r}{2}}$$

It is generally accepted that when the stretching efficiency factor $f = 0.9$ in this equation, the predictions are more accurate. Based on this relationship, the calculated R_{LD} values for processing I and processing II are 1.97657 and 2.00737, respectively. Although the latter is only slightly higher, it still reflects the influence of thermomechanical processing on formability. Actual R_{LD} values require further experimental measurement.

2.2 Tensile Fracture Analysis

[Figure 2: see original paper] presents SEM images of the fracture surfaces of T4P-treated Al-Mg-Si-Cu alloy sheets processed by both routes after tensile testing along longitudinal and transverse directions. The fracture morphology exhibits typical ductile fracture characteristics, featuring numerous dimples with many smaller dimples distributed around larger ones, regardless of testing direction. Moreover, since the alloys were in a recrystallized state during tensile testing and contained numerous equiaxed recrystallized grains in the matrix (as shown in [Figure 3: see original paper]g and h), the fracture morphologies along longitudinal and transverse directions show minimal differences for both processing routes, which explains why their elongation values are essentially similar.

2.3 Effect of Thermomechanical Processing on Microstructure Evolution

[Figure 3: see original paper] illustrates the microstructural evolution during the two thermomechanical processing routes. After homogenization, the as-cast alloy exhibits a relatively uniform microstructure with the dendritic network completely eliminated. However, due to furnace cooling after high-temperature homogenization, certain amounts of precipitates (primarily Mg2Si and minor Fe-rich particles) are distributed within grains and near grain boundaries (Fig. 3a). After hot rolling to 4 mm, grains are significantly elongated along the rolling direction, and the precipitates are fragmented and aligned along the rolling direction (Fig. 3b). When intermediate annealing at 400 °C for 1 h is applied directly after hot rolling, incomplete recrystallization occurs, with some elongated recrystallized grains visible in the matrix (Fig. 3c). In contrast, when 50% cold rolling to 2 mm is performed before the same intermediate annealing treatment, numerous recrystallized grains are clearly observed (Fig. 3d). When the annealed sheets from both routes are further cold-rolled to 1 mm, the final cold-rolled microstructures show no significant differences, both exhibiting fibrous structures (Figs. 3e and f), despite different deformation amounts. After solution treatment at 555 °C for 2 min, the alloy processed by route I still shows slightly elongated grains along the rolling direction with numerous coarse particles remaining in the matrix (Fig. 3g), whereas route II yields approximately equiaxed grains with only a few undissolved coarse particles (Fig. 3h).

[Figure 4: see original paper] shows SEM images and EDS analysis of coarse second-phase particles in the intermediate-annealed sheets processed by both routes. In the route I material, the coarse particles are primarily Mg2Si and Fe-rich phases containing Al, Fe, Mn, and Si (Figs. 4a, c, d). Although route II also contains Mg2Si and Fe-rich coarse particles, the distribution uniformity is significantly improved and the particle size is reduced (Fig. 4b). This is mainly attributed to the additional cold rolling before intermediate annealing, which more effectively fragments the coarse second-phase particles (primarily Fe-rich phases). The degree of fragmentation of Fe-rich phases before annealing also directly affects the distribution of coarse particles in the solution-treated state: lower fragmentation (route I) results in more numerous and larger coarse particles (Fig. 3g), while severe fragmentation (route II) leads to fewer and smaller coarse particles (Fig. 3h).

Thermomechanical processing affects not only coarse particles but also fine particles, which play a crucial role in controlling recrystallized grain growth and thus significantly influence microstructure, texture, and properties regulation. Therefore, analyzing the distribution of fine particles in cold-rolled alloys is essential. [Figure 5: see original paper] presents TEM images of 1 mm cold-rolled alloys processed by both routes. In route I material, fine particles are relatively small (mostly below $0.2\,\mu$m) with high distribution density, and SAED analysis identifies them as primarily Si particles (Fig. 5a). Although route II also contains Si particles, the number of particles smaller than $0.2\,\mu$m is significantly reduced (Fig. 5b). Detailed particle distributions are summarized in . This distribution difference arises because most of these fine particles form during the 400 °C, 1 h intermediate annealing stage. In route I, intermediate annealing follows hot rolling directly, whereas in route II, it follows 50% cold rolling. The larger strain energy in route II promotes rapid growth of Si particles during annealing, resulting in larger particle sizes. The subsequent cold rolling deformation, being relatively small and performed on a softer annealed matrix, does not effectively further fragment the particles. Consequently, the particle distribution in the final cold-rolled state remains essentially the same as in the intermediate-annealed state and can be directly used to analyze its effect on recrystallization during solution treatment.

2.4 Effect of Thermomechanical Processing on Recrystallization Texture

[Figure 6: see original paper] shows EBSD grain orientation and size distribution maps of the solution-treated alloys processed by both routes. Different colors represent different grain orientations, with lighter colors indicating greater deviation from ideal orientations. The alloy processed by route I exhibits relatively small average recrystallized grain size with a narrow distribution range, attributed to the numerous fine particles that strongly inhibit grain growth. However, the grains show a large aspect ratio and slight elongation along the rolling direction, which is unfavorable for reducing anisotropy (as shown in Table 1). The large aspect ratio primarily results from the greater cold rolling deformation before solution treatment.

Although the experimental alloy has low Fe content, both processing routes produce cold-rolled sheets containing various amounts of micron- and submicron-sized particles. According to extensive literature [**?**], micron-sized Fe-rich or Mg2Si particles generate substantial strain energy and deformation zones around them during cold rolling, promoting preferential recrystallization nucleation around particles (PSN effect) during high-temperature heat treatment. However, successful PSN depends on particle size and prior deformation amount; particles must generally exceed a critical size d_c to exert the PSN effect, which can be expressed as [**?**]:

$$d_c = \frac{2\gamma_{gb}}{P_D - P_Z}$$

where γ_{gb} is the specific grain boundary energy, P_D is the stored strain energy, and P_Z is the Zener pinning force from fine particles; α is a constant; ρ is dislocation density; G is the shear modulus; b is the magnitude of the Burgers vector; and F_V and d_p are the volume fraction and diameter of fine particles, respectively. Equation (2) shows that P_Z significantly affects the critical size d_c for PSN, varying with the volume fraction and size of fine particles. For the two processing routes used in this work, although route I contains a higher volume fraction of fine particles after cold rolling (Fig. 5a), the PSN effect can still operate during solution treatment due to the larger size of coarse particles and greater final cold rolling deformation, resulting in smaller recrystallized grain size (Fig. 6a). Conversely, although route II has fewer and smaller coarse particles in the cold-rolled state, the PSN effect can also occur during solution treatment because the number and size of fine particles are significantly lower than in route I, resulting in less inhibition of PSN. Therefore, the final recrystallized grains are approximately equiaxed (Fig. 6b).

[Figure 7: see original paper] shows the ODF maps of the solution-treated alloys processed by both routes, with corresponding texture components and their intensities and volume fractions listed in Table 3. The route I material exhibits numerous recrystallization texture components with relatively high intensities and contents (Fig. 7a), whereas route II shows reduced texture components and overall lower intensity, with recrystallized grains being primarily randomly oriented (Fig. 7b).

The recrystallization texture components in Al-Mg-Si-Cu alloy sheets are generally determined by competition between Cube-oriented grains and grains formed through the PSN effect [?]. During recrystallization, Cube-oriented subgrains nucleating in cube bands generally recover and grow faster than other oriented subgrains due to their high symmetry. Moreover, their $40° < 111 >$ orientation relationship with the S orientation of the cold-rolling texture gives Cube-oriented grains strong growth capability [?], leading to the presence of Cube texture in both processed materials. If the cold rolling deformation is large, shear bands with high local strain can also promote Goss-oriented nuclei formation, and the Goss orientation grows preferentially due to its $40° < 111 >$ orientation relationship with the $B\{011\} < 211 >$ rolling texture. Therefore, the solution-treated alloy processed by route I, which involves larger cold rolling deformation, exhibits a certain amount of Goss texture. Coarse particles generate random orientations or CubeND and P textures through the PSN effect. Since CubeND typically has a $40° < 111 >$ orientation relationship with the $\{112\} < 111 >$ rolling texture, and P texture has a $40° < 111 >$ relationship with the S rolling texture, both textures can nucleate and grow preferentially in the deformed matrix [?]. Consequently, route I material, with its larger cold rolling deformation and more numerous coarse particles, shows more pronounced PSN effects during solution treatment, resulting in higher intensities of CubeND and P textures. H is a shear texture that primarily appears on sheet surfaces during rolling and can be retained during recrystallization due to difficult nucleation and growth; route II material contains a small amount of H texture.

Overall, due to the appropriate combination of a certain number of coarse particles and relatively few fine particles in the cold-rolled state, route II produces a recrystallized alloy sheet with more randomly oriented and CubeND-oriented grains, while containing fewer detrimental Cube and Goss textures. This results in excellent stamping formability with r reaching 0.6187, which is higher than both the 0.5391 value for route I and the maximum value of 0.578 reported in literature [?] obtained by fully exploiting PSN effects from Fe-rich particles. Additionally, Δr is also lower (Table 1).

2.5 Microstructure Evolution Model

To better illustrate the effects of different thermomechanical processes on microstructure and texture, [Figure 8: see original paper] presents schematic diagrams of the microstructure evolution during processing. Both routes start with the same hot-rolled microstructure containing certain amounts of Fe-rich and Mg2Si phases (Figs. 8a and e), followed by different cold rolling and annealing sequences. In route I, direct intermediate annealing after hot rolling results in low stored energy and incomplete recrystallization, with limited growth of existing and precipitated fine Mg2Si and Si particles (Fig. 8b). Subsequent cold rolling of the soft matrix does not effectively fragment coarse second-phase particles, leaving numerous coarse particles and fine Mg2Si and Si particles in the cold-rolled state (Fig. 8c). The severe deformation around these numerous, relatively large coarse particles leads to pronounced PSN effects during solution treatment. However, the large cold rolling deformation also facilitates formation of other detrimental orientations such as Cube and Goss nuclei (Fig. 8d). Consequently, the recrystallized microstructure contains grains with multiple orientations and large aspect ratios (Fig. 6a), resulting in strong planar anisotropy.

In route II, the hot-rolled sheet first undergoes cold rolling deformation, which not only fragments the coarse second-phase particles to some extent but also, due to the larger stored strain energy, promotes rapid growth of existing and precipitated particles during subsequent intermediate annealing, yielding larger particle sizes than in route I (Fig. 8f). The annealed sheet is relatively soft, so cold rolling does not significantly alter the distribution of differently sized particles (Fig. 8g). With smaller cold rolling deformation before solution treatment, the PSN effect from coarse particles is less pronounced than in route I, and formation of other oriented nuclei is also reduced (Fig. 8h). Although the number of fine particles providing effective pinning is lower than in route I, resulting in slightly larger recrystallized grain size, most recrystallized grains are equiaxed with fewer texture components, predominantly random orientations (Fig. 6b). Therefore, the alloy exhibits higher r-value and reduced anisotropy.

Conclusions

1. Variations in thermomechanical processing have essentially no effect on the strength and strain-hardening exponent n of T4P-pre-aged alloys, but significantly influence the average plastic strain ratio r, planar anisotropy Δr, and elongation in different directions. Alloy sheets obtained by cold rolling before intermediate annealing after hot rolling (processing II) demonstrate better formability than those directly annealed after hot rolling (processing I), with r reaching 0.6187 and markedly reduced anisotropy.

2. Although the particles distributed in Al-Mg-Si-Cu alloy matrix mainly include Fe-rich phase particles and Mg2Si particles, their sizes and distribution states change significantly with thermomechanical processing. The 1 mm cold-rolled sheet from processing I contains numerous coarse and fine particles, resulting in fine but elongated recrystallized grains after solution treatment. In contrast, processing II yields reduced coarse particle size and increased fine particle size, leading to slightly larger but essentially equiaxed recrystallized grains after solution treatment.
3. The texture components and contents of solution-treated alloy sheets vary considerably with thermomechanical processing. Processing I produces multiple recrystallization texture components including CubeND, P, Cube, and Goss textures with relatively high intensities, while processing II results in only weak CubeND, Cube, and H textures.
4. Different thermomechanical processing routes exert important influences on the comprehensive properties, microstructure, and texture evolution of Al-Mg-Si-Cu alloy sheets, based on which a schematic model of microstructure evolution during thermomechanical processing has been proposed.

References

[?] Yu Z Q, Lin Z Q, Zhao Y X. Mater Des, 2007; 28: 203
[?] Gao Q M, Cao L S, Yu X D. Light Alloy Fabr Technol, 2009; 37: 30 (高琪妹, 曹力生, 于晓丹. 轻合金加工技术, 2009; 37: 30)
[?] Engler O, Hirsch J. Mater Sci Forum, 1996; 217: 479
[?] Bennett T A, Petrov R H, Kestens L A I. Scr Mater, 2010; 63: 461
[?] Troeger L P, Starke Jr E A. Mater Sci Eng, 2000; A293: 19
[?] Peng X Y, Guo M X, Wang X F, Cui L, Zhang J S, Zhuang L Z. Acta Metall Sin, 2015; 51: 169 (彭祥阳, 郭明星, 汪小峰, 崔莉, 张济山, 庄林忠. 金属学报, 2015; 51: 169)
[?] Lievers W B, Pilkey A K, Lloyd D J. Mater Sci Eng, 2003; A361:
[?] Lankford W T, Snyder S C, Bauscher J A. Trans Am Soc Met, 1950; 42: 1197
[?] Hirofumi I, Takayuki T. Mater Trans, 2007; 48: 2014
[?] Leu D K. Int J Mach Tools Manuf, 1997; 37: 201
[?] Humphreys F J. Acta Metall, 1977; 25: 1323
[?] Orsund R, Nes E. Scr Mater, 1988; 22: 665
[?] Davies R K, Randle V, Marshall G J. Acta Mater, 1998; 46: 6021
[?] Bennett T A, Petrov R H, Kestensv L A I. Scr Mater, 2010; 62: 78
[?] Higginson R L, Aindow M, Bate P S. Mater Sci Eng, 1997; A225:
[?] Ito K, Musick R, Lucke K. Acta Metall, 1983; 31: 2137
[?] Jensen D J, Hansen N, Humphreys E J. Acta Metall, 1985; 33:
[?] Hansen N, Jensen D J. Metall Trans, 1986; 17A: 253
[?] Hirsch J, Nes E, Lucke K. Acta Metall, 1987; 35: 427
[?] Kashihara K, Inagaki H. Metall Trans, 2009; 50A: 528
[?] Humphreys F J. Acta Metall, 1977; 25: 1323

[?] Engler O. Mater Sci Technol, 1996; 12: 859

Note: Figure translations are in progress. See original paper for figures.