Anisotropic Deformation Behavior of CuNi10Fe1Mn Alloy with Continuous Columnar Grain Structure (Postprint)

Authors: LIU Yongkang, Huang Haiyou, Xie Jianxin

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

Abstract

The mechanical properties and deformation behavior of continuous columnar crystal structure CuNi10Fe1Mn alloy plate specimens during tensile testing along the direction parallel to columnar crystal growth (SD) and the direction perpendicular to columnar crystal growth (PD) were investigated, and the influence of anisotropy in the continuous columnar crystal structure on the mechanical properties and deformation behavior of the alloy was discussed. The results indicate that specimens along both SD and PD exhibit a [100] preferred orientation, but along SD the orientations of all grains are distributed near [100] with an average Taylor factor m value of approximately 2.17, whereas along PD the orientations of some grains are scattered between [001]-[011], exhibiting certain orientation scatter, with an m value of approximately 2.93. Microstructural characteristics such as low orientation scatter and absence of transverse grain boundary constraints result in uniform stress distribution during deformation along SD, with all grains deforming consistently along the tensile direction; the yield strength and tensile strength are 85 MPa and 215 MPa respectively, the elongation after fracture reaches 42%, and the fracture exhibits typical ductile characteristics. Due to the presence of certain orientation scatter and transverse grain boundaries, PD specimens exhibit obvious grain boundary stress concentration and surface undulation during deformation, the yield strength significantly increases to 115 MPa, the elongation after fracture decreases to 36%, and the fracture is of mixed type. The anisotropy in grain orientation and grain boundary distribution in the continuous columnar crystal structure CuNi10Fe1Mn alloy is the cause of the anisotropy in deformation behavior.

Full Text

Anisotropic Deformation Behavior of Continuous Columnar-Grained CuNi10Fe1Mn Alloy

LIU Yongkang[1], **HUANG Haiyou**[1,2], **XIE Jianxin**[1,2]

[1] Key Laboratory for Advanced Materials Processing of Ministry of Education, Institute for Advanced Materials and Technology, University of Science and Technology Beijing, Beijing 100083, China
[2] Beijing Laboratory of Metallic Materials and Processing for Modern Transportation, University of Science and Technology Beijing, Beijing 100083, China

Correspondent: XIE Jianxin, professor, Tel: (010)62332254, E-mail: jxxie@mater.ustb.edu.cn

Supported by: National Key Technology Research and Development Program of China (No.2011BAE23B00), National Natural Science Foundation of China (No.51104015) and Independent Research Program of State Key Laboratory for Advanced Metals and Materials (No.2012Z-12)

Manuscript received: 2014-07-04, in revised form 2014-09-23

Abstract

In continuous unidirectional solidification, a unidirectional heat transfer condition can be established between the melt and solidifying metal to control grain growth along the solidification direction (SD). This method produces continuous columnar-grained (CCG) polycrystalline alloys without transverse grain boundaries, featuring highly oriented texture and straight grain boundary morphology. High texture orientation significantly improves deformation consistency among grains, while straight grain boundaries reduce the number of strain components requiring coordination during deformation. Consequently, CCG alloys exhibit high plasticity and excellent ductility along the SD. For example, CCG CuNi10Fe1Mn alloy achieves tensile elongation exceeding 40%. However, as noted, CCG alloys possess extremely anisotropic microstructures. This work investigates the mechanical properties and deformation behavior of CuNi10Fe1Mn alloy with tensile directions along both the SD and perpendicular to the solidification direction (PD), using electron back-scatter diffraction (EBSD) and digital image correlation (DIC) techniques to study microstructural anisotropy effects.

Results indicate that both SD and PD samples exhibit [100] preferred orientation. All grains in SD samples (Taylor factor m=2.17) cluster near [100], while some grains in PD samples (m=2.93) scatter between [001]-[011], showing greater orientation dispersion. SD samples feature low orientation dispersion and no horizontal grain boundaries, promoting uniform stress distribution and consistent grain deformation along the tensile direction, with yield strength of

85 MPa, tensile strength of 215 MPa, and elongation of 42%—characteristic of ductile fracture. PD samples, due to orientation dispersion and horizontal grain boundaries, exhibit pronounced grain boundary stress concentration and zigzag surface morphologies, with yield strength increasing markedly to 115 MPa, elongation decreasing to 36%, and showing mixed fracture characteristics. The anisotropic deformation behavior of CCG CuNi10Fe1Mn alloy originates from anisotropic grain orientation and grain boundary distribution.

KEY WORDS continuous columnar grain, Cu-Ni alloy, anisotropy, mechanical property, deformation behavior

1. Introduction

In continuous directional solidification, establishing unidirectional heat transfer conditions between melt and solidifying metal controls grain growth along the crystallographic easy-growth direction, producing alloys with grains aligned parallel to the solidification direction, free of grain boundaries perpendicular to that direction (transverse grain boundaries), and possessing highly oriented microstructures. These represent special single-crystal or polycrystalline alloys. For cubic metallic materials, crystals grow fastest along <001> directions, facilitating preparation of CCG alloys with strong <001> texture. Strong texture significantly enhances deformation consistency during tensile loading along the grain direction, while straight grain boundaries reduce the number of coordinated strain components, and elimination of transverse grain boundaries decreases dislocation-grain boundary interactions. Consequently, CCG alloys demonstrate high plasticity and superior ductility along the solidification direction compared with conventional polycrystalline materials. For instance, our previous work produced 17.28 mm diameter pure Cu wire that could be cold-drawn to 19.7 m diameter at room temperature without intermediate annealing, achieving a cumulative deformation of 7.69$×$10ˆ7%. CCG CuNi10Fe1Mn alloy tubes exhibited tensile elongation of 49%, comparable to conventional polycrystalline pure Cu. A CCG CuNi10Fe1Mn alloy tube (10 mm diameter, 1.8 mm wall thickness) was processed into capillary tubes (1.0 mm diameter, 0.2 mm wall thickness) through multi-pass cold rolling and drawing at room temperature, achieving a total cross-section reduction of 98.9%.

These characteristics enable development of short-process, high-efficiency manufacturing technologies combining continuous directional solidification with severe plastic deformation. For example, CCG CuNi10Fe1Mn alloy tubes produced by hot-cold combined mold continuous casting exhibit high surface quality and excellent cold workability, enabling direct large-deformation continuous cold rolling to produce various condenser tubes. Compared with conventional extrusion-rolling processes, this approach significantly shortens the process flow while substantially improving production efficiency and yield.

However, as described above, CCG alloys exhibit extreme microstructural anisotropy. Investigating how this extreme anisotropy affects mechanical properties and deformation behavior is crucial for improving performance, selecting appropriate processing methods, and establishing reasonable processes. Therefore, this study focuses on CuNi10Fe1Mn alloy, employing full-field strain analysis to investigate strain distribution and evolution during tensile deformation of samples oriented parallel and perpendicular to the columnar grain growth direction, combined with microstructural observation and orientation analysis.

2. Experimental Methods

High-purity metals ($ $99.9%) were used as raw materials, with compositions prepared according to ASTM B111 standards. Continuous columnar-grained CuNi10Fe1Mn alloy plates (85 mm × 45 mm cross-section) were prepared by directional solidification. Chemical analysis revealed the actual composition (wt.%): Ni 10.70, Fe 1.35, Mn 0.70, balance Cu.

Two groups of dog-bone tensile specimens were machined by wire electrical discharge machining, with tensile directions parallel and perpendicular to the columnar grain growth (solidification) direction, designated SD (solidification direction) and PD (perpendicular direction) specimens, respectively. All specimens had a gauge length of 20 mm and cross-section of 3 mm × 5 mm. Uniaxial tensile tests were conducted at room temperature on an MTS-810 universal testing machine at a strain rate of 10^-3 s^-1.

Phase analysis was performed using a D/max-RB 12 kW rotating anode X-ray diffractometer (XRD, CuKα radiation). Before tensile testing, specimen surfaces (5 mm × 30 mm) were mechanically ground and polished with sandpaper and cloth, then etched for ~5 s in a solution of CH3COOH:HNO3:H2O (2:2:1 volume ratio) for metallographic observation using a KEYENCE VHX-2000 3D optical microscope (OM). Specimens were subsequently electropolished (electrolyte: H2SO4:H2O:H3PO4 = 1:3:6 volume ratio; 5.8-6.0 V, 0.9-1.0 A, 140-150 s) for EBSD orientation analysis using a TESCAN MIRA3 LMH thermal field emission scanning electron microscope (SEM) equipped with an EBSD detector. Data processing employed HKL Channel 5 orientation analysis software.

Full-field strain distribution during tension was analyzed using digital image correlation (DIC). Before testing, specimens were spray-painted to create trackable speckle patterns. Tensile loading was performed using the MTS-810 machine while two 4M CCD cameras (2358$×$1728 pixels) captured images. ARAMIS 3D image analysis software calculated full-field strain data and strain contour maps.

2. Results

2.1 Microstructure

[Figure 1: see original paper] shows OM images of continuous columnar-grained

CuNi10Fe1Mn alloy specimens parallel and perpendicular to the columnar grain axis. The average columnar grain diameter is ~1.5 mm, with grains containing numerous dendrites. Long primary dendrites grow along the solidification direction, arranged parallel with an average spacing of ~200 m. Secondary dendrites grow perpendicular to primary dendrites at 90° angles, appearing as "cross" shapes in transverse sections (Fig. 1b).

XRD analysis results are shown in Fig. 1c, with lattice constants a=b=c=0.35923 nm, $\alpha=\beta=\gamma=90°$ used for phase identification. The sharp (200) diffraction peak indicates strong [100] preferred orientation along the solidification direction. Combined dendrite morphology analysis reveals that the α-phase with fcc structure has primary dendrites growing along [100] and secondary dendrites along four <001> directions perpendicular to the primary dendrites, forming the "cross" shape shown in Fig. 1b.

2.2 Mechanical Properties

Typical tensile stress-strain curves for SD and PD specimens are shown in [Figure 2: see original paper]. Compared with PD specimens, SD specimens exhibit higher plasticity but lower yield and tensile strengths. Statistical analysis of five tensile tests for each group yielded average mechanical properties: SD specimens achieved 42% average elongation, with yield and tensile strengths of 85 MPa and 215 MPa, respectively; PD specimens showed yield and tensile strengths of 115 MPa and 229 MPa, respectively, with 36% average elongation. Both elongation and yield strength demonstrate anisotropy. However, compared with sand casting and centrifugal casting, both SD and PD specimens exhibit high plasticity.

To quantify plastic anisotropy, we define the plastic anisotropy parameter h as:

$$h = \frac{\bar{\delta}_{SD} - \bar{\delta}_{PD}}{\bar{\delta}_{SD}} \times 100\%$$

where $\bar{\delta}_{SD}$ and $\bar{\delta}_{PD}$ represent average elongations of SD and PD specimens, respectively. Calculation shows PD specimen elongation is 14.3% lower than SD specimens, corresponding to a plastic anisotropy of 14.3%.

2.3 Sample Surface and Fracture Morphology

Post-deformation surface and fracture morphologies are shown in [Figure 3: see original paper]. SD specimen front surfaces display banded undulations parallel to the loading axis (Fig. 3a), with amplitude increasing near the fracture. Side surfaces remain relatively smooth without obvious surface relief. PD specimens exhibit significant undulations along the tensile direction on both front and side surfaces (bamboo-like deformation, Fig. 3b). Comparison with OM images confirms these undulations occur along grain boundaries, indicating inconsistent intergranular deformation and strain differences between adjacent grains, manifesting as grain-scale surface relief.

SD specimen fractures show pronounced necking with numerous uniformly distributed dimples (Fig. 3c), characteristic of ductile fracture. PD specimens exhibit less necking, larger areas of river patterns, fewer and non-uniformly distributed dimples, indicating mixed fracture characteristics (Fig. 3d). Comparison of necking and fracture locations with columnar grain boundaries reveals that necking and fracture occur within grains, not along grain boundaries.

2.4 Full-Field Strain During Tension

Full-field strain analysis further elucidated deformation behavior differences between SD and PD specimens. Camera recordings captured speckle pattern displacements on specimen surfaces during continuous loading, with software analysis revealing strain evolution. Strain distribution contour maps were selected at five characteristic strain levels: initial plastic deformation point (I), plastic deformation stages (II and III), macroscopic stress maximum (IV), and necking stage (V) [Figure 4: see original paper]. When local strain \$ \$40%, contour maps display red; for detailed local strain distribution, axial centerline data (L1 and L2) are extracted in Figs. 4c and d. The local strain value Y is calculated from surface spot displacements; after necking, unstable deformation in the necked region causes local strain 突变.

SD specimen strain distribution (Figs. 4a and c) shows uniform strain distribution throughout the deformation zone at 12% macroscopic strain (position II). With increasing load, localized stress concentration appears in the specimen center (contour maps III and IV). Beyond 30% macroscopic strain (position IV), necking occurs in the stress concentration region, macroscopic stress decreases, while local stress concentration intensifies (contour map V). Fracture occurs at ~42% macroscopic strain through transgranular fracture in the localized stress concentration region.

PD specimen strain distribution (Figs. 4b and d) differs significantly. Multiple localized stress concentration regions appear early in the plastic stage (6% macroscopic strain, contour map II). With increasing load, strain differences between deformation zones increase (curves II-IV in Fig. 4d), and local stress concentration intensifies. Beyond 30% macroscopic strain, necking occurs in the central stress concentration region, macroscopic stress decreases, and local stress concentration continues increasing until fracture.

[Figure 5: see original paper]a shows maximum local strain Ymax evolution along the tensile direction for both specimen types. Both exhibit similar patterns: during uniform deformation, Ymax increases slowly with macroscopic strain; at ~30% macroscopic strain, necking initiates and Ymax increases rapidly until fracture. At equivalent macroscopic strains, SD specimens show smaller Ymax values. Considering SD specimens also have lower tensile strength, this indicates lower flow stress both globally and locally. Pre-fracture Ymax values are similar for both specimen types, suggesting comparable local stresses at failure. Comparison of PD fracture locations with columnar grain boundaries reveals that grain boundaries are not preferential crack propagation paths; fracture locations show no direct correspondence with grain boundaries. Both specimen types undergo transgranular fracture with identical fracture behavior and fracture resistance.

[Figure 5: see original paper]b shows the difference ΔY between maximum and minimum local strains. SD specimens exhibit smaller strain differences, indicating better strain (stress) distribution uniformity and higher deformation consistency.

3. Discussion

3.1 Microstructural Anisotropy

[Figure 1: see original paper] demonstrates that CCG microstructural anisotropy includes both grain boundary distribution and grain orientation anisotropy. During SD tension, all grain boundaries are parallel to the deformation direction; during PD tension, all grain boundaries are perpendicular. XRD spectra (Fig. 1c) confirm strong [100] preferred orientation along the solidification direction. To further analyze orientation anisotropy, EBSD orientation analysis statistically examined over 20 grains along both directions.

[Figure 6: see original paper] shows inverse pole figures (IPFs) along SD and PD. Both exhibit [100] preferred orientation, but SD grains cluster near [100] while PD grains scatter between [001]-[011], showing orientation dispersion. This orientation characteristic results from unidirectional heat transfer conditions during continuous directional solidification: all columnar grains grow along easy-growth <001> directions under constrained conditions, forming strong [100] orientation along SD. PD crystal growth lacks constraint, allowing random orientation in the plane perpendicular to [100] or rotation about SD (Fig. 6c). However, Fig. 6b shows PD also exhibits some [100] orientation, though with greater dispersion than SD. [Figure 6: see original paper]d shows Schmid factor contour lines in the orientation triangle for fcc metals. Schmid factors for SD grains range 0.41-0.48, while PD grains range 0.41-0.50.

3.2 Orientation Dependence of Deformation Behavior

For fcc CuNi10Fe1Mn alloy, with Cu and Ni forming infinite solid solutions, the material is a single-phase solid solution with medium-low stacking fault energy. At low strain rates and room temperature, the primary deformation mechanism is dislocation slip with 12 possible slip systems. Different grain orientations activate varying numbers and difficulty levels of slip systems, with Schmid factors indicating slip system activation ease. PD grains show larger Schmid factors (Fig. 6d), suggesting lower critical shear stress for slip activation and easier deformation. However, mechanical testing (Fig. 2) shows PD yield strength is significantly higher than SD. This occurs because Schmid factors only reflect single-grain slip activation difficulty, while polycrystalline deformation is strongly influenced by neighboring grain misorientation (deformation consistency). Figs. 6a and b show PD has greater orientation dispersion, resulting in lower deformation consistency and grain boundary stress concentration (Fig. 4b).

For alloys deforming by dislocation slip, tensile stress σ relates to critical resolved shear stress τ through:

$$\sigma = m\tau$$

where m is the Taylor factor. For fcc single crystals, m values for different orientations can be obtained from [Figure 6: see original paper]d. For conventional fcc polycrystalline structures, the average Taylor factor is m=3.06. Conventional cast polycrystalline CuNi10Fe1Mn alloy has yield strength ~120 MPa, yielding critical shear stress τ=39.2 MPa.

Experimental results show SD specimen yield strength of 85 MPa, giving m_{SD}=2.17; PD specimen yield strength of 115 MPa gives m_{PD}=2.93. According to Taylor plasticity theory, m reflects polycrystalline resistance to plastic deformation. The SD m value approaches the theoretical minimum of 2.0, indicating low deformation resistance, while the higher PD m value indicates greater resistance. Both values are lower than conventional polycrystalline structures, giving CCG CuNi10Fe1Mn alloy lower deformation resistance than conventional polycrystalline alloys.

To further investigate microstructural anisotropy effects, deformed microstructures were analyzed ([Figure 7: see original paper]). Figures 7a and b show OM images and EBSD orientation maps after 15% tensile strain. Slip lines and deformation bands are visible, with slip trace analysis revealing activated slip systems. After deformation, different oriented grains activate slip systems sequentially, creating low-angle boundaries within grains and near grain boundaries.

The SD specimen observation region contains three grains (labeled 1, 2, 3) with orientations near <001> along the tensile direction. The two grain boundaries show misorientations of 23.1° and 40.3° (Fig. 7c). Slip trace statistics (Table 1) show: Grain 1 contains two intersecting slip trace sets at ~39° and 43° to the tensile axis; Grain 2 contains three non-intersecting trace sets, mostly at 57° with minor sets at 36° and 63°; Grain 3 shows one trace set at 44°. Trace angles enable estimation of activated slip systems and corresponding Schmid factors (Table 1). Results show activated slip systems in grains 1 and 2 have Schmid factors of 0.41, while grain 3 reaches 0.46. Similar Schmid factors indicate slip activation at comparable stress levels. Uniform slip trace distribution and consistent contrast demonstrate high deformation consistency during SD deformation. Additionally, SD grain boundaries are parallel to the tensile direction, with columnar grains continuous through the specimen, promoting uniform grain deformation, straight post-deformation specimens, and minimal intergranular relief (Fig. 3a).

The PD specimen region contains five grains with orientations [071], [031], [021], [010], and [100] from left to right, showing misorientations of 16.4°, 9.0°, 17.5°, and 21.6° (Fig. 7d). Significant differences in slip line traces and micro-deformation structures are evident between grains: grains 2 and 4 show shallow traces, while grains 1, 3, and 5 show deep traces with abundant substructures near grain boundaries, reflecting strong deformation inhomogeneity. Trace analysis (Table 1) shows PD specimens have two primary trace angles relative to the tensile axis, poor intergranular deformation consistency, and require activation of more slip systems. Different slip systems show large Schmid factor variations; in grain 3, activation of only b and c slip systems (Schmid factor 0.46) cannot satisfy deformation consistency, requiring additional a and d systems (Schmid factor 0.39). Therefore, PD deformation exhibits orientation dispersion, varying deformation difficulty between grains, and under transverse grain boundary constraints, forced deformation continuity at grain boundaries generates local inhomogeneous deformation (Fig. 7d) and stress concentration (Fig. 4b), producing pronounced bamboo-like surface undulations (Fig. 3b).

In summary, deformation behavior anisotropy along SD and PD results from CCG CuNi10Fe1Mn alloy microstructural anisotropy. However, microstructural anisotropy does not significantly affect plasticity, yielding low plastic anisotropy. This may arise from: (1) despite [001]-[011] orientation dispersion along PD, a large proportion of grains remain near [100], giving both SD and PD specimens [100] preferred orientation (Fig. 6); and (2) although grain boundary distribution is anisotropic, grain boundaries are not preferential crack propagation paths (Figs. 3 and 4), and transverse grain boundaries do not significantly impede PD elongation. Both specimen types show identical fracture behavior—transgranular fracture. Therefore, CCG CuNi10Fe1Mn alloy exhibits small elongation differences between SD and PD, with plastic anisotropy below 15%.

4. Conclusions

1. CCG CuNi10Fe1Mn alloy specimens exhibit [100] preferred orientation along both SD and PD. All SD grain orientations cluster near [100], while PD grain orientations scatter between [001]-[011] with some orientation dispersion.

2. Average Taylor factors are ~2.17 for SD and ~2.93 for PD, both lower than conventional polycrystalline structures (3.07). SD deformation shows high consistency with yield strength 85 MPa, tensile strength 215 MPa, elongation 42%, and ductile fracture. PD deformation yields yield strength 115 MPa, tensile strength 229 MPa, elongation 36%, and mixed fracture.

3. Anisotropic grain orientation and grain boundary distribution in CCG CuNi10Fe1Mn alloy cause deformation behavior anisotropy.

References

[1] Ohno A, Soda H, McLean A, Yamazaki H. Adv Mater, 1990; 28: 161
[2] Onaka S, Kato H, Hashimoto S, Miura S. J Jpn Inst Met, 1995; 59:
[3] Zhang Z M, Lü T, Xu C J, Guo X F. Acta Metall Sin (Engl Lett), 2008; 21: 275
[4] Ji D P, Liu X F, Xie J X, Yu J W, Li W H, Rong M L. Acta Metall Sin, 2006; 42: 1243
[5] Wang Y, Huang H Y, Xie J X. Mater Sci Eng, 2011; A530: 418
[6] Xie J X, Wang Y, Huang H Y. Chin J Nonferrous Met, 2011; 21: 2324
[7] Gao K W, Liu M Y, Zou F L, Pang X L, Xie J X. Mater Sci Eng, 2010; A527: 4750
[8] Gan C L, Liu X F, Huang H Y, Xie J X. Mater Sci Eng, 2013; A579: 202
[9] Zhang H, Xie J X, Wang Z D. J Univ Sci Technol Beijing (Engl Ed), 2004; 11: 240
[10] Gan C L, Liu X F, Huang H Y, Xie J X. Acta Metall Sin, 2010; 46: 1549
[11] Mei J, Liu X H, Jiang Y B, Xie J X. Chin J Nonferrous Met, 2012; 22: 2529
[12] Xie J X, Lou H F, Wang Z D, Hu P X, Zhang H, Dong Y Z, Kang J L, Miao G W, Fu X Z, Yan M, Guan B H, Jiang X L. Chin Pat, 200710065281.9, 2007
[13] Xie J X, Mei J, Liu X H, Liu X F. Chin Pat, 201010501407.4, 2011
[14] Mei J, Liu X H, Xie J X. Int J Min Met Mater, 2012; 19: 339
[15] Fu Y B, Yan Z M, Li T J, Chen P, Cheng Y F, Yin G M. Mater Des, 2009; 30: 4478
[16] Li X T, Guo Z X, Zhao X W, Wei B, Chen F B, Li T J. Foundry, 2007; 56: 691
[17] Daly S, Ravichandran G, Bhattacharya K. Acta Mater, 2007; 55:
[18] Efstathiou C, Sehitoglu H, Carroll J, Lambros J, Maier H J. Acta Mater, 2008; 56: 3791
[19] Morgeneyer T F, Taillandier-Thomas T, Helfen L, Baumbach T, Sinclair I, Roux S, Hild F. Acta Mater, 2008; 69: 78
[20] Zou F L, Gao K W, Zhu Q F, Xie J X. Acta Metall Sin, 2008; 44: 297
[21] Bradley A J, Cox W F, Goldschmidt H J. J Inst Met, 1941; 67: 189
[22] DKI German Copper Institute Booklet. Copper Nickel Alloys: Properties, Processing, Application. http://www.copper.org/applications/marine/cuni/txt_{DKI}, 2010
[23] Courtney T H. Mechanical Behavior of Materials. 2nd Ed, New York: McGraw-Hill Companies Inc., 2000: 159
[24] Li X T, Zhao X W, Wei B, Chen F B, Yan Z M, Li T J. Chin J Nonferrous Met, 2007; 17: 922
[25] Zaefferer S, Kuo J C, Zhao Z, Winning M, Raabe D. Acta Mater, 2003; 51: 4719