**Tuning pore structure of corrosion resistant solid-state-sintered SiC porous ceramics by
particle size distribution and phase transformation**

**Wenming Guo^{a,b,c}, Hanning Xiao^{a*}, Xinghe Yao^d, Jingxiong Liu^a, Jianjun Liang^a,
Pengzhao Gao^a, Guangming Zeng^{b,c}**

a. College of Materials Science and Engineering, Hunan University, Changsha 410082, China;

b. College of Environmental Science and Engineering, Hunan University, Changsha 410082, China;

c. Key Laboratory of Environmental Biology and Pollution Control (Hunan University), Ministry of Education, Changsha 410082, China;

d. China Academy of Launch Vehicle Technology, Beijing 100076, China.

* Corresponding Author: College of Materials Science and Engineering, Hunan University, Changsha 410082, China

Tel: +86 731 88822269 Fax: +86 731 88823554

Email address: wenming_guo@hnu.edu.cn (Wenming Guo)

Abstract

Aggressive process environment like hot corrosive water treatment caused corrosion in ceramic membrane and strength degradation of the membrane support, resulting in the invalidation of membrane filtration. Therefore, new SiC membrane support based on its intrinsic corrosion resistance was prepared by the combination of bi-modal particle size distribution and solid-state-sintering with 1–3 wt.% of B₄C additive at 2150 °C and 2200 °C. By tuning the contents of coarse and fine powder and without introducing any pore-creating agent, the porosity of SiC supports varied in the range of 36–39%, which was suitable for the

membrane filtration. The phase transformation of 6H-SiC to 4H-SiC induced by the addition of B₄C at 2200 °C, accompanied by the formation of large plate-like grains, caused the pore size increased from 11.8 μm to 20.9 μm. The corrosion resistance of solid-state-sintered SiC supports was evaluated by exposing in 90 °C 20 wt.% H₂SO₄ and NaOH solutions for 200 h. The high residual flexural strength ($\geq 92.5\%$) and unchanged microstructure of the solid-state-sintered SiC supports after corrosion demonstrated that the supports exhibited excellent corrosion resistance and would be promising candidates to substitute the traditional ceramic supports.

Keywords: SiC, Microstructure, Corrosion, Membranes, Porous ceramics

1. Introduction

Porous ceramics served as mechanical support for membrane has attracted more and more attentions because of their remarkable properties, such as high separation efficiency, excellent thermal stability, high-pressure resistance, chemical stability, and long service life, compared with polymeric membranes [1–3]. Various materials such as Al₂O₃, mullite, and cordierite, have been introduced as supports [4–6]. However, studies and application data have shown that the porous supports made of these materials present gradual strength degradation when exposed to harsh environments, such as hot corrosive solutions and concentrated basic/acidic solutions, which may result in the damage of membranes under operating pressures [7–10]. The strength-loss ratios of porous high-purity α -Al₂O₃ ceramics corroded in 80 °C 20 wt.% H₂SO₄ and 10 wt.% NaOH solution for 8 h are 9.9% and 8.1% [11, 12], respectively, and these ratios reach 59.7% and 47.5% respectively after deep corrosion for 20 days under similar conditions [13]. Al₂O₃ porous support sintered by using kaolin at 1620 °C by Dong et

al. [9] shows conspicuous loss in strength by 82.65 % when exposed in 10 wt% NaOH solution at 100 °C for 8h, which is owing to the attack of NaOH on the glassy grain boundary and Al₂O₃ grains. By contrast, other low cost ceramic supports such as cordierite and mullite appear more susceptible to dissolution in corrodent [10, 14]. Cordierite is not resistant to thermal acidic solution due to the quick dissolution of Mg and Al [10], while mullite is not stable in hot acidic and basic solutions [15–17].

SiC is stable in all known aqueous corrosive solutions, except for a slight corrosion in phosphoric acid at 215 °C and in alkaline solution of K₃Fe(CN)₆ at temperatures higher than 100 °C [18]. SiC has high oxidation resistance in certain environments such as passive oxidation by forming a protective SiO₂ layer on top of the SiC surface. High-temperature oxidation enhanced by water above 800 °C is a major disadvantage of SiC [19]. As a result, SiC has been considered as a candidate to substitute traditional ceramic support materials [20, 21]. The corrosion resistance of porous SiC supports prepared through reaction-bonding [22] and liquid-phase-sintering [23] is strongly affected by the residual grain boundary phases, which results in the degradation of mechanical property [24–27]. Recrystallized SiC and solid-state-sintered SiC (with the additives of B/B₄C and C) retain the excellent corrosion resistance of SiC because of their high purity and absence of residual grain boundary phases [24, 28–30]. However, it is difficult to reach high porosity and large pore size for the two SiC materials simply by introducing organic or carbonaceous pore-creating agents so as to meet the high filtration flux of membrane, because the residual carbon originated from the pore-creating agents under inert atmosphere is disadvantageous to the sintering of recrystallized SiC and solid-state-sintered SiC. Instead of pore-creating agents, particle size

distribution of raw materials is generally employed to adjust the porosity and pore size. By this means, porous SiC materials with large pore sizes normally require large particle sizes, and consequently need high sintering temperatures (e.g. 2350 °C for recrystallized SiC with pore size of ~20 μm [31]), which will dramatically enhance the cost of membrane. We have previously prepared a porous SiC support exhibiting high flexural strength by solid-state sintering and phase transformation only with the aid of B₄C at 2200 °C [32]. The phase transformation of 6H-SiC to 4H-SiC altered the morphology of grains, resulting in the change of pore structure. In this study, we will focus on adjusting the pore size of the solid-state-sintered SiC support by particle distribution of raw materials and sintering additive without introducing any pore forming agent.

2. Experimental

2.1 Raw materials

Coarse 6H-SiC powder ($d_{50} = 22.5 \mu\text{m}$) and submicron 6H-SiC powder ($d_{50} = 0.5 \mu\text{m}$) were obtained from Changle Xinyuan Carborundum Micropowder Co., Ltd. (Shandong, China). B₄C powder (~1.0 μm , Jingangzuan Boron Carbide Co., Ltd., China) and polyvinyl alcohol (PVA, average polymerization degree of 1750 ± 50 , Sino Chemical Reagent Co., Ltd., China) were used as sintering agent and temporary binder, respectively.

2.2 Preparation

SiC powders with bi-modal particle size distribution were obtained by mixing the coarse and fine SiC powders in different ratios as shown in Table 1. The starting powder mixtures (SiC:B₄C = 100:1) were blended through ball milling in 1 wt.% PVA solution with SiC grinding balls in a polyurethane jar (400 r/min, 1 h) and then dried by rotary evaporation.

Subsequently, the powder mixtures were sieved with a 60-mesh screen and pressed under 30 MPa. The compacts were loaded into recrystallized SiC crucibles and individually sintered at 2150 °C for 1 h. In addition, C70 series samples doped with 0, 0.5, 1, and 3 wt.% B₄C were prepared similarly, but sintered at 2200 °C for 1 h.

2.3 Characterization

The density of the samples was measured through water displacement technique (Archimedes method). Flexural strength was determined by electronic universal testing machine (AG-50KNIS, Shimadzu, Japan) using three-point bending method in a span of 35 mm at room temperature and loading rate of 0.5 mm/min for samples with a geometrical size of 8 mm × 6 mm × 65 mm. The morphology and microstructures of the samples were observed under a scanning electron microscopy (SEM, JSM6490LV, JEOL, Japan). The mean sizes of grains and necks in SEM photographs of fractured SiC supports were measured by Nano Measurer Software. Pore size distribution was determined by mercury intrusion porosimetry (MIP, PoreMaster-60, Quantachrome, USA). Phase identification was performed through powder X-ray diffraction (XRD, D/max2200 VPC, Rigaku, Japan) by using monochromic Cu K α radiation of 0.15405 nm operated at 40 kV and 40 mA. Corrosion testing was performed by placing the samples in a glass bottle filled with 20 wt.% H₂SO₄ solution or a steel bottle filled with 20 wt.% NaOH solution. The bottles were placed in a water bath maintained at 90 °C. The strength of the samples subjected to H₂SO₄ or NaOH solution was compared with identical unexposed samples. Some samples polished with W5, W3, and W1 diamond before corrosion, were used for microstructure observation after corrosion.

3. Results and discussion

3.1 Effects of particle size distribution on the pore structure of porous SiC supports

Fig.1 illustrates the effect of coarse powder content on the density and open porosity of SiC supports after sintering at 2150 °C. The density increases from 2.032 g/cm³ to 2.060 g/cm³ with increasing coarse powder content from 50% (C50) to 55% (C55), however, further increase of coarse powder content to 80% (C80) results in the gradual decrease of the density to 1.959 g/cm³. Particularly, the density varies insignificantly in the range of 50–70% coarse powder content. The porosity variation displays opposite to the density with the increase in coarse powder content. As shown in Fig.1, SiC supports prepared without any pore-creating agents exhibit open porosity in the range of 36–39%, suggest that relative high porosity can be obtained just by adjusting suitable coarse/fine powder ratio, which facilitates the prevention of environmental problems arising from organic pore-creating agents.

The average pore size of SiC supports (Fig. 2) increases from 7.9 μm to 9.1 μm, and the pore size distribution becomes narrower as the coarse powder content increases from 50% to 80%, indicating that an adjustment of coarse/fine particle ratios is effective on adjusting pore size of SiC supports on a small scale.

The fracture surfaces of the SiC supports show similar interconnected porous structure with only coarse connected particles as the framework, and reveal that the original fine particles disappeared after sintering. However, few differences can be detected by careful observation. Large pores can be seen in C50 (Fig. 3a), and the microstructure becomes uniform with the increase of coarse powder content. The mean grain size and neck size obtained from Fig. 3 are illustrated in Fig. 4, both of which decrease with the increase of

coarse powder content, i.e. with the decrease of fine powder content. The microstructure evolution depends on the interstices among coarse particles and the volume content of fine particles. For C50, the interstices among coarse particles are not sufficient to accommodate the fine ones, and the partial fine particles participate in framework formation of green body by making contact with the coarse ones, which will transfer to the necks of coarse particles or be swallowed by the adjacent coarse particles during sintering, leading to the formation of local large pores (Fig. 3a) and local dense large particles. As the coarse powder content increases to 60 wt.%–70 wt.%, interstices among coarse powders can accommodate more fine powders, and the uniformity of pore structure increases. In this range, the fine powders can properly fill the interstices among coarse ones, resulting in relatively high density of green body. For C80, the fine powders are not enough to fill up the interstices among coarse particles, leaving large uniform pores after sintering. Moreover, with the increase of coarse powder content, i.e., the decrease of fine powders content, the sources for grain growth and neck formation decrease, which consequently decreases grain and neck sizes (Fig. 4). As can be seen in Fig. 3, lots of pores larger than 20 μm disagree with the data obtained from mercury intrusion porosimetry (Fig. 2). This discrepancy originates from the “ink bottle” effect of mercury intrusion porosimetry [33], which inflates the volume of the smaller pores due to the presence of larger pores behind the pore throats. Since the porous SiC supports show weak shrinkage during sintering, one can assume the center distance of the adjacent particles is constant. The pore throats increase due to the grain and neck sizes of porous SiC supports decrease with the increase of original coarse powder content (Fig. 4), which is consistent with the decrease average pore size of SiC porous ceramics obtained from mercury

intrusion porosimetry (Fig. 2).

3.2 Effects of phase transformation on the pore structure of SiC porous ceramics

The density and open porosity of C70 doped with different content of B₄C additive after sintering at 2200 °C for 1 h are shown in Fig. 5. The density increases slightly with the B₄C content. The open porosity decreases from 38% to 35% when B₄C content increases from 0% to 3%, which indicate the doping of B₄C has no significant effect on the open porosity of SiC porous ceramics. However, the microstructure of SiC porous ceramics changes notably (Fig. 6). The grains are equiaxed in pure SiC, while larger plate-like grains appear in the SiC porous ceramics doped with B₄C, and the content of plate-like grains increases with the B₄C content. In our previous work [32], we have proved that the phase of the plate-like grains is 4H-SiC, which is transformed from the original 6H-SiC and is beneficial to improve the flexural strength of SiC porous ceramics. Moreover, the increase of plate-like grains in the microstructure with the B₄C content coincides with the variation of characteristic peaks of 4H-SiC in Fig. 7.

The transformation of 6H-SiC to 4H-SiC altered the pore structure of SiC porous ceramics. With the increase of 4H-SiC content, i.e. the increase of plate-like grains, the pore distribution of porous SiC supports shifts to larger interval and the average pore size increases from 11.8 μm to 20.9 μm as shown in Fig. 8, which demonstrates the phase transformation is more effective than the particle size distribution on tuning the pore structure of SiC porous ceramics.

3.3 Corrosion resistance of solid-state-sintered SiC supports in hot acidic/basic solutions

The corrosion resistance of the C70 series SiC supports sintered with various B₄C

contents at 2200 °C is evaluated in this study. Residual strengths of SiC supports after 200 h of corrosion in 20 wt.% H₂SO₄ and NaOH solutions at 90 °C are shown in table 2. The average strengths of supports all remain $\geq 92.5\%$ of their initial strengths, suggesting that SiC supports sintered with or without B₄C additive possess the desired corrosion resistance in hot acidic/basic solutions, which is considerably superior to the supports reported in previous literature [11–17]. Moreover, a representative SiC support sintered with 1 wt.% B₄C at 2200 °C shows un conspicuous variation of flexural strength as the corrosion time increases (Fig. 9). Slight corrosion can be detected at the necks between the large grains in pure SiC support as shown in Fig. 10a and 10b, but corrosion trace is absent in the SiC supports sintered with B₄C (Fig. 10c–f). Porous solid-state-sintered SiC support exhibits corrosion resistance comparable with the dense solid-state-sintered SiC ceramics [24], indicating the intrinsic excellent corrosion resistance of solid-state-sintered SiC, which may profit from the high corrosion resistance of SiC, B₄C, and the solid solute of B₄C in SiC.

4. Conclusions

Porous solid-state-sintered SiC membrane supports with relative high open porosity of 35–39% are successfully prepared just by employing suitable content of coarse and fine SiC powders without introducing any pore creating agents. Adjusting the raw particle distribution is only effective on adjusting pore size of SiC supports on a small scale, however, adjusting the additive B₄C content at 2200 °C is an efficient way of tuning the pore size without affecting porosity. Dramatic change in pore size from 11.8 μm to 20.9 μm is obtained by controlling the B₄C content. The porous solid-state-sintered SiC supports with B₄C as sintering additive exhibit residual strength higher than 92.5% after corrosion in hot

acidic/basic solution for 200 h, which is much superior to traditional ceramic support materials.

Acknowledgements

The authors are grateful for the financial support from the National Science Foundation of China (Grant Nos. 51302076, 51372078) and the China Postdoctoral Science Foundation (Grant No. 2013M531783).

References

- [1] P.B. Belibi, M.M.G. Nguemtchouin, M. Rivallin, J.N. Nsami, J. Sieliechi, S. Cerneaux, M.B. Ngassoum, M. Cretin, Microfiltration ceramic membranes from local Cameroonian clay applicable to water treatment, *Ceram. Int.* 41 (2015) 2752–2759.
- [2] S.R. Hosseinabadi, K. Wyns, V. Meynen, R. Carleer, P. Adriaenssens, A. Buekenhoudt, B.V. Bruggen, Organic solvent nanofiltration with Grignard functionalized ceramic nanofiltration membranes, *J. Membr. Sci.* 454 (2014) 496–504.
- [3] S. Kroll, L. Treccani, K. Rezwan, G. Grathwohl, Development and characterisation of functionalised ceramic microtubes for bacteria filtration. *J. Membr. Sci.* 2010 (365) 447–455.
- [4] Q. Chang, Y. Wang, S. Cerneaux, J. Zhou, X. Zhang, X. Wang, Y. Dong, Preparation of microfiltration membrane supports using coarse alumina grains coated by nano TiO₂ as raw materials, *J. Eur. Ceram. Soc.* 34 (2014) 4355–4361.
- [5] J. Cao, X. Dong, L. Li, Y. Dong, S. Hampshire, Recycling of waste fly ash for production of porous mullite ceramic membrane supports with increased porosity, *J. Eur. Ceram. Soc.* 34 (2014) 3181–3194.

- [6] D. Kim, A. Kellogg, E. Livaich, B. A. Wilhite, Towards an integrated ceramic micro-membrane network: Electroless-plated palladium membranes in cordierite supports, *J. Membr. Sci.* 340 (2009) 109–116.
- [7] T.V. Gestel, C. Vandecasteele, A. Buekenhoudt, C. Dotremont, J. Luyten, R. Leysen, B. Van der Bruggena, G. Maes, Alumina and titania multilayer membranes for nanofiltration: preparation, characterization and chemical stability, *J. Membr. Sci.* 207 (2002) 73–89.
- [8] U. Aust, S. Benfer, M. Dietze, A. Rost, G. Tomandl, Development of microporous ceramic membranes in the system $\text{TiO}_2/\text{ZrO}_2$, *J. Membr. Sci.* 281 (2006) 463–471.
- [9] Y. Dong, B. Lin, J. Zhou, X. Zhang, Y. Ling, X. Liu, G. Meng, S. Hampshire, Corrosion resistance characterization of porous alumina membrane supports, *Mater. Charact.* 62 (2011) 409–418.
- [10] Y. Dong, X. Feng, D. Dong, S. Wang, J. Yang, J. Gao, X. Liu, G. Meng, Elaboration and chemical corrosion resistance of tubular macro-porous cordierite ceramic membrane supports, *J. Membr. Sci.* 34 (2007) 65–75.
- [11] W. Qin, C. Peng, M. Lv, J. Wu, Preparation and properties of high-purity porous alumina support at low sintering temperature, *Ceram. Int.* 40 (2014) 13741–13746.
- [12] W. Qin, B. Lei, C. Peng, J. Wu, Corrosion resistance of ultra-high purity porous alumina ceramic support, *Mater. Lett.* 144 (2015) 74–77
- [13] G. Dong, H. Qi, N. Xu, Effect of activated carbon doping on structure and property of porous alumina support, *J. Chin. Ceram. Soc.* 40 (2012) 844–850.
- [14] G. Chen, X. Ge, Y. Wang, W. Xing, Y. Guo, Design and preparation of high permeability

- porous mullite support for membranes by in-situ reaction, *Ceram. Int.* 41 (2015) 8282–8287.
- [15] M. Herrmann, F.L. Toma, L.M. Berger, G. Kaiser, C.C. Stahr, Comparative study of the corrosion resistance of thermally sprayed ceramic coatings and their bulk ceramic counterparts, *J. Eur. Ceram. Soc.* 34 (2014) 493–504.
- [16] D. Wu, F. Mao, S. Wang, Z. Zhou, A low-temperature fabrication route for enhancing mechanical properties and corrosion resistance of porous mullite ceramics through homogeneously mullite sol-coating method, *J. Ceram. Process. Res.* 14 (2013) 677–681.
- [17] S. Kurajica, L. Curkovic, E. Tkalcec, V. Mandic, Acid corrosion behavior of sol-gel-prepared mullite ceramics with and without addition of lanthanum, *J. Am. Ceram. Soc.* 96 (2013) 923–927.
- [18] D. Zhuang, J.H. Edgar, Wet etching of GaN, AlN, and SiC: a review, *Mater. Sci. Eng. R* 48 (2005) 1–46.
- [19] P. Colomban, The corrosion of ceramic-matrix composites, *Mater. Sci. Forum* 251–254 (1997) 833–844.
- [20] M. Facciotti, V. Boffa, G. Magnacca, L. B. Jørgensen, P.K. Kristensen, A. Farsi, K. König, M.L. Christensen, Y. Yue, Deposition of thin ultrafiltration membranes on commercial SiC microfiltration tubes, *Ceram. Int.* 40 (2014) 3277–3285.
- [21] W. Deng, X. Yu, M. Sahimi, T.T. Tsotsis, Highly permeable porous silicon carbide support tubes for the preparation of nanoporous inorganic membranes, *J. Membr. Sci.* 451 (2014) 192–204.
- [22] V.S. Kiselov, P.M. Lytvyn, V.O. Yukhymchuk, A.E. Belyaev, S.A. Vitusevich,

- Synthesis and properties of porous SiC ceramics, *J. Appl. Phys.* 107 (2010) 093510.
- [23] M. Fukushima, Y. Zhou, H. Miyazaki, Y. Yoshizawa, K. Hirao, Y. Iwamoto, S. Yamazaki, T. Nagano, Microstructural Characterization of Porous Silicon Carbide Membrane Support With and Without Alumina Additive, *J. Am. Ceram. Soc.* 89 (2006) 1523–1529.
- [24] D.A. Ray, S. Kaur, R.A. Cutler, D.K. Shetty, Effects of additives on the pressure-assisted densification and properties of silicon carbide, *J. Am. Ceram. Soc.*, 91 (2008) 2163–2169.
- [25] S. Ding, Y. Zeng, D. Jiang, Preparation and properties of in situ reaction bonded porous SiC ceramics, *J. Inorg. Mater.* 21 (2006) 1397–1403.
- [26] W. Kim, H.S. Hwang, J.Y. Park, Corrosion behavior of reaction-bonded silicon carbide ceramics in high-temperature water, *J. Mater. Sci. Lett.* 21 (2002) 733–735.
- [27] K.R. Mikeska, S.J. Bennison, S.L. Grise, Corrosion of ceramics in aqueous hydrofluoric acid, *J. Am. Ceram. Soc.* 83 (2000) 1160–1164.
- [28] M. Herrmann, J. Schilm, G. Michael, Corrosion behavior of different technical ceramics in acids, basic solutions and under hydrothermal conditions, *Key Eng. Mater.* 264–268 (2004) 877–880.
- [29] M. Herrmann, E. Kluge, C. Rödel, A. McKie, F. van Staden, Corrosion behaviour of silicon carbide–diamond composite materials in aqueous solutions, *J. Eur. Ceram. Soc.* 34 (2014) 2143–2151.
- [30] C. Badini, F. de Haan, E. Celasco, P. Hähner, C.M. Vega Bolivar, S. Biamino, M. Pavese, P. Fino, Corrosion behavior of SiC laminate under decomposed sulfuric acid at 850°C, *J.*

Am. Ceram. Soc., 95 (2012) 2627–2634.

- [31] Z.Z. Yi, Z.P. Xie, Y. Huang, J.T. Ma, Y.B. Cheng, Study on gelcasting and properties of recrystallized silicon carbide, *Ceram. Int.*, 28 (2002) 369–376.
- [32] W. Guo, H. Xiao, J. Liu, J. Liang, P. Gao, G. Zeng, Effects of B₄C on the microstructure and phase transformation of porous SiC ceramics, *Ceram. Int.*, 41 (2015) 11117–11124.
- [33] A.B. Abell, K.L. Willis, D.A. Lange, Mercury intrusion porosimetry and image analysis of cement-based materials, *J. Colloid Interface Sci.*, 211 (1999) 39–44.

Table 1 Compositions of coarse and fine powders for preparing supports (wt.%)

Samples	Coarse powder	Fine powder
C50	50	50
C55	55	45
C60	60	40
C65	65	35
C70	70	30
C75	75	25
C80	80	20

Table 2 Residual strengths of SiC supports after corrosion in 90 °C 20 wt.% H₂SO₄ solution
and NaOH solution for 200 h (%)

B ₄ C content (wt.%)	Sintering temperature (°C)	Residual strength	
		NaOH	H ₂ SO ₄
0	2200	93.1	93.1
0.5	2200	92.5	95.8
1	2150	95.3	98.6
1	2200	93.1	98.9
3	2200	97.9	100.2

Fig. 1 Density and open porosity of SiC supports with different contents of coarse powder

Fig.2 Pore size distribution and average pore size of SiC supports with different coarse powder contents determined by mercury intrusion porosimetry

Fig.3 SEM photographs of SiC supports sintered at 2150 °C for 1 h (a-C50, b-C60, c-C70 and d-C80)

Fig. 4 Mean sizes of grains and necks in SiC supports

Fig. 5 Density and open porosity of C70 SiC supports with different contents of B₄C sintered at 2200 °C

Fig. 6 Microstructure of C70 SiC supports sintered at 2200 °C with 0 (a), 0.5 (b), 1 (c), and 3 (d) wt.% of B₄C

Fig. 7 XRD spectra of SiC supports with different B₄C contents sintered at 2200 °C

Fig. 8 Pore size distribution and average pore size of SiC supports with different B₄C contents determined by mercury intrusion porosimetry

Fig. 9 Flexural strength variation of the SiC supports sintered with 1wt.% B₄C at 2200 °C for 1 h after hot acidic (a) and basic (b) solution corrosion for different time

Fig. 10 SEM photographs of SiC supports sintered with 0 (a, b), 1 wt.% (c, d) and 3 wt.% (e, f) B₄C at 2200 °C for 1 h after corrosion in hot acidic (a, c and e) and basic (b, d and f) solution for 200 h

Tuning pore size and microstructure

High corrosion resistance