Paper

Ultralow-Temperature Field Activated Sintering of Amorphous B₄C under Millimeter Wave Irradiation and Pressure

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ABSTRACT

The millimeter wave pressure sintering featuring power output (W_m) control of amorphous B₄C powders is proposed without the need for any additive, in order to reveal definitely divided densification far below a glass transition. With the constant-rate heating technique, amorphous B₄C containing nanocrystalline rhombohedra can be consolidated without sub- μ m sized pore at 50 MPa. The heating rate used is found to be a dominating variable; with an increase from 10 to 20 K·min⁻¹, it leads to a decrease from 510 to 444 K in full-density temperature. Its second-stage densification is expressed by the constitutive equation of consolidating strain rate $\dot{\epsilon}_n$ via non-Newtonian flow: $\dot{\epsilon}_n = \dot{\epsilon}_0 \sinh(C\sigma_{eff}/RT) \exp(-Q/RT)$ with a decreasing apparent activation energy Q from 42.5 to 35.4 kJ·mol⁻¹ under an increased W_m . The constant-power output technique with a maximum W_m of 2 kW enables one to obtain a considerable initial abrupt shrinkage and subsequently full consolidation at the lowest temperature of 386 K under 98 MPa. The process control of the first-stage densification is then characterized by a neck formation law based on the first order chemical reaction $d(A_f A)/dt = -k_p(A_f A)$ where A is the fraction of a contacted area between particles, having an increasing k_p with W_m .

KEY WORDS

pore-free consolidation, bulk amorphous synthesis, non-Newtonian flow, activation energy, neck formation

1 Introduction

The microwave electric field is one of the most potential process variables of ceramic sintering, by which one can explore every new possibility of ultralow-temperature and/or ultrafast sintering. The enhancement of microwave sintering, accompanied by rapid, internal and/or selective heating, is well-known as a non-thermal effect, which can be explained by surface and bulk transport with a decreased apparent activation energy^{1,2)}. Alternatively, an intelligent millimeter wave pressure system³⁾ was developed as a versatile way of electric field activated thermomechanical testing, especially by which one can manage a powder densification route through judicious selection of multi-process variables. Until now, this processing has been used to activate the densification via visco-plastic flow in amorphous zirconium oxide⁴⁾, and to present an idea of the master field sintering curve for nanocrystalline hydroxyapatite⁵⁾.

In the meantime, the solid state synthesized amorphous B₄C powders⁶⁾ with the excellent surface activity can be fully densified during Newtonian viscous flow prior to a crystallization by pulsed electric current sintering (PECS)⁷⁾. Whereas the millimeter wave pressure-less sintering of commercialized B₄C was used to prepare a highly densified large bulk, but found it difficult to gain a greatly

decreased temperature⁸⁾. So, the millimeter wave pressure-assisted sintering using amorphous B₄C is hoped to provide a process control for artificial neck formation, as suggested in the discharging of rectangular pulses⁷⁾, and full densification via non-Newtonian flow below a glass transition as is the case with high-pressure PECS of amorphous TiAl⁹⁾. The proposed amorphous sintering also lends itself to a further comprehensive understanding of electric field assisted/activated sintering using pulsed current^{10,11)} and direct current¹²⁾.

The current author here is going to set up a process control methodology based a phenomenological law, in terms of manipulative variables of millimeter wave pressure sintering, for the ultralow-temperature full densification of amorphous B₄C in heating.

2 Experimental procedure**2.1 Preparation of amorphous powder**

The amorphous B₄C with 15 mass% nanocrystalline rhombohedra were prepared by mechanically grinding stoichiometric compound particles (the purity of more than 99%, 0.007%Al, 0.002%Ca, 0.03%Fe, 0.02%Na and 0.01%Si, Kojundo Chemical Lab. Co., Ltd.) with a diameter of 0.5 μ m using a temperature-controlled attrition mill for a time period of 504 h¹³⁾. The tank and balls were made of partial stabilized zirconia; the inner wall and the surface were coated with B₄C after several batch milling in an argon atmosphere

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Fig. 1 Schematic diagram of the intelligent millimeter wave pressure sintering system for the bulk amorphous synthesis.

to avoid outer contamination. The powder sample used, having a dispersion of non-flowing particles, is hereafter called amorphous B₄C, since the densification via plastic flow in a matrix of a disordered structure is mainly dealt.

2.2 Millimeter wave consolidation featuring power output control

Fig. 1 illustrates the instrumented millimeter wave pressure sintering system for the bulk amorphous synthesis. This apparatus provides the means of controlling the process variables of temperature T , applied stress σ , heating rate dT/dt where t is the time, millimeter wave power output W_m and the plunger displacement Z . The experiment using the microwave of oscillation tube with the frequency of 28 GHz was well-appointed to meet test conditions for a macroscopic analysis of multi-variable controlled sintering of far from equilibrium ceramics⁵⁾. Two power output controlled methods were adopted, by which one can vary the electric field E applied to an amorphous compact: the constant-rate heating technique was performed with 10 and 20 K·min⁻¹ (0.167 and 0.333 K·s⁻¹) at an applied stress of 50 MPa, and the constant-power output technique was conducted under a W_m of 2 kW at the maximum under 98 MPa.

Fig. 2 is the outlook of the specially designed cylindrical die set⁵⁾. The alumina die which transmits microwave has an inner diameter of 20 mm and an outer diameter of 40 mm. The amorphous B₄C powders with a weight of 3 g were packed into the die and then consolidated in a vacuum; the uniaxial loading using a digital servo-control system (Instron 8800) was applied through the upper and lower plungers for a more uniform stress profile inside a flat compact. The die set was surrounded by a cylindrical alumina board to reduce the dissipation of a generated heat and a temperature gradient inside a powder compact. The die temperature was measured via thermocouple at the center of a flat end of the alumina disk with a thickness 2 mm situated beneath a compact, and was recognized as an average sintering temperature from knowing its inverse parabolic distribution across a radius of the compact and a thickness of the disk.

The true displacement, corrected for thermal expansion of graphite plungers, is used to determine the sample size and then its

Fig. 2 Alumina die set used for millimeter wave sintering with two-way loading.

density ρ . The relative density D is here used as an experimental measure and calculated from the following equation:

$$D = H_f D_f / (Z_f - Z + H_f) \quad (1)$$

where D_f is the final relative density of sintered bulk, H_f is the final height at room temperature and Z_f is the final true displacement. The true strain rate $\dot{\epsilon}$ during consolidation of a powder compact in a die¹⁴⁾ is defined by

$$\dot{\epsilon} = -(dH)/(Hdt) = (1/D)(dD/dt) \quad (2)$$

where H is the instantaneous height of a compact. $\dot{\epsilon}$ {equal to the normalized densification rate $(1/\rho)(d\rho/dt)$ } is applicable to an isoconfigurational glass.

2.3 Characterization of consolidated bulk

The microstructure of a surface of an amorphous B_4C bulk was observed with an optical microscope. The density of an amorphous B_4C bulk is obtained from its weight and volume. The consolidated B_4C was characterized by X-ray diffraction using Cu K α radiation in the theta-theta diffractometer with a high-speed one-dimensional detector (Rigaku Ultima IV). The average crystallite size d was estimated using the Scherrer formula of $d=0.9\lambda/(B\cos\theta)$ where B is the X-ray peak broadening at half-maximum, λ is the X-ray wavelength (0.1542 nm) and θ is the scattering angle.

3 Results

3.1 Amorphous sintering under constant-rate heating

Fig. 3 shows the temperature of the compact and the apparent displacement Z' versus time at a heating rate of $10\text{ K}\cdot\text{min}^{-1}$ under an applied stress of 50 MPa. This measurement was taken in a real time during the millimeter wave sintering of the amorphous B_4C powders. The amorphous B_4C exhibits a two-step increase in the apparent displacement up to 510 K after loading to a maximum load $P_{\max}$ necessary to reach the level of 50 MPa. This process is accompanied by a rapid shrinkage at ambient temperatures and a sigmoidal curve wherein a rate process would underlie below a glass transition. Further heating leads to a linear decrease in Z' resulting from thermal expansion of the sample and plungers as depicted by a dashed line. In every trial of consolidation up to approximately 873 K, the thermal expansion of the amorphous B_4C bulk caused a fracturing of the alumina die without any indication of further shrinkage. Fig. 4 shows the outlook of the consolidated amorphous B_4C with a diameter of 20 mm as described in Fig. 3. The amorphous B_4C bulk is a cylindrical flat disk with a shiny surface and sharp edges in avoidance of any visible crack and void.

Fig. 5 shows the consolidating strain rate versus temperature for the amorphous B_4C as in Fig. 4. The two-stage densification process under millimeter wave radiation is confirmed by $\dot{\epsilon}$ in the engineering sense of the term. In the first-stage densification denoted as I, $\dot{\epsilon}$ shows a steep rise up to a maximum level of $1\cdot 10^{-3}\text{ s}^{-1}$ followed by a rapid decrease. This behavior is not concerned with an inhomogeneous flow that is characterized by extremely low levels of strain rate sensitivity exponent⁷⁾ and activation energy¹⁵⁾. So, one must look elsewhere for a mechanism involved; i.e. neck formation via pulsed spark^{7,16)} is mentioned as a promising candidate. In the second-stage densification denoted as II, $\dot{\epsilon}$ gradually increases followed by a decrease from 408 K. This dependence is coupled to non-Newtonian flow, and is unlike Newtonian viscous flow with the stress exponent of unity and a higher activation energy of $420\text{ kJ}\cdot\text{mol}^{-1}$ ⁷⁾.

Fig. 6 shows the millimeter wave sintering processes of the amorphous B_4C compact for heating rates of 10 and $20\text{ K}\cdot\text{min}^{-1}$ at an applied stress of 50 MPa: (a) the relative density against temperature and (b) the power output versus processing time. The

Fig. 3 Apparent displacement and temperature versus time at a heating rate of $10\text{ K}\cdot\text{min}^{-1}$. Measurements are taken during millimeter wave pressure sintering of the amorphous B_4C powders under an applied stress of 50 MPa.

Fig. 4 Cylindrical shaped amorphous B_4C bulk sample synthesized by millimeter wave pressure sintering as shown in Fig. 3.

Fig. 5 Millimeter wave consolidating strain rate against temperature for the amorphous B_4C powder compact under $10\text{ K}\cdot\text{min}^{-1}$.

discrete densification process clearly appears for two different heating rates; it should be noted that the relative density D_0 at the starting of stable sintering is close to 0.64 that is the predicted value of the densest random packing of spherical particles. For the heating rate increased to $20\text{ K}\cdot\text{min}^{-1}$, the relative density at the completion of the first-stage densification attains a higher level of 0.84 even in a shorter period, and the temperature necessary for pore-free consolidation is decreased to 444 K, contrary to conventional sintering. Concomitantly, the power output, as increased with time according to a nearly square root law, has

Fig. 6 Millimeter wave pressure sintering process of the amorphous B₄C under 50 MPa in the cases of the heating rate of 10 and 20 K·min⁻¹: (a) relative density versus temperature and (b) power output versus processing time.

a nearly double level in an overall process of densification. The heating rate used with the same die and sample volume is so found to be a manipulative parameter of two-stage densification under electric field.

Fig. 7 shows the optical micrograph of consolidated amorphous B₄C for 20 K·min⁻¹. There is no visible void, as commonly observed in a sintered bulk, throughout a glassy surface, except for some straight scratches with a width range 0.3 to 9 μm, close to each of abrasive grain diameters of a series of sand papers used. This indicates that the amorphous B₄C is fully densified avoiding sub-μm sized pore within the resolution of optical microscopy. Fig. 8 shows X-ray diffraction patterns of the amorphous B₄C bulk synthesized at 533 K using 20 K·min⁻¹ and attrition-milled amorphous B₄C powders. The consolidated B₄C shows no significant change of the first amorphous X-ray peak in comparison with that of as-milled powders, together with a constant crystallite size of 10 nm for rhombohedral particles. Referring to an increased relative integrated peak intensity of 2.1 at 1573 K⁷⁾, one can say that the major structural relaxation is suppressed during the second-stage densification.

3.2 Amorphous sintering under constant-power output

Fig. 9 shows the apparent displacement and the temperature against time under the maximum constant power output of 2 kW. This measurement was taken during the millimeter wave sintering of the amorphous B₄C powders after loading up to a higher

Fig. 7 Optical micrograph of a surface of the consolidated amorphous B₄C bulk sample using 20 K·min⁻¹.

Fig. 8 X-ray diffraction patterns of consolidated amorphous B₄C bulk at 533 K and as attrition-milled amorphous powders.

Fig. 9 Apparent displacement and temperature versus time at the constant power output of 2 kW. Measurements are taken during the millimeter wave sintering of the amorphous B₄C powders at 98 MPa.

applied stress of 98 MPa. The amorphous B_4C with a weight of 1.5 g, heated with approximately $100\text{ K}\cdot\text{min}^{-1}$, exhibits a large rapid increment in the displacement, having the highest $\dot{\epsilon}$ of $2.6\cdot 10^{-2}\text{ s}^{-1}$, and approaches monotonously to full density at the lowest temperature of 386 K and in a short period of time for 56 s. Additionally, the field-emission scanning electron microscopy with energy dispersive X-ray spectroscopy confirmed that rapidly sintered amorphous B_4C had no apparent spatial distribution of sub- μm sized pore except for a lack of $0.5\text{ }\mu\text{m}$ sized particles (on the order of 0.13%) along large scratches on its surface where oxidation is occasionally likely to occur. Fig. 10 shows the relative density of the amorphous B_4C bulk against time under a W_m of 2 kW at 98 MPa. This figure includes a good correlation between D and t in the first stage densification under the W_m of 0.51 ± 0.01 and $0.7\pm 0.02\text{ kW}$ for heating rates of 10 and $20\text{ K}\cdot\text{min}^{-1}$ respectively. For 2 kW, the relative density shows an accelerative increase up to a maximum level of 0.95 and in a time period of 8 s. So, the constant-power output technique could provide a base for ultralow-temperature and ultrafast sintering, which is classified as flash sintering^{12,17)}.

Fig. 11 shows the temperature necessary for full densification of amorphous B_4C versus power output and applied stress. This figure compares a full-density temperature range 1073 to 1523 K in PECS⁷⁾ via Newtonian viscous flow in metastable equilibrium amorphous B_4C (with 2% rhombohedra) subjected to full structural relaxation, and 965 K in millimeter wave micro-pressure (0.6 MPa) sintering of amorphous B_4C (3% rhombohedra) without a die. One can see a synergistic decrease in the full-density temperature of far below a glass transition by increasing power output and applied stress. The amorphous B_4C bulk has a density of $2.1\text{ Mg}\cdot\text{m}^{-3}$, which is nearly equal to that in PECS⁷⁾. ρ for an amorphous matrix is also estimated as $2.02\text{ Mg}\cdot\text{m}^{-3}$, using a mixture rule with $2.52\text{ Mg}\cdot\text{m}^{-3}$ for rhombohedra; this value is 5% higher than $1.92\text{ Mg}\cdot\text{m}^{-3}$ for a powder mixture of amorphous carbon ($2.1\text{ Mg}\cdot\text{m}^{-3}$) and boron ($1.73\text{ Mg}\cdot\text{m}^{-3}$).

4 Discussion

4.1 A process control model of amorphous neck formation

A phenomenological law of the neck formation via millimeter wave sparking under applied stress is based on a linear chemical reaction, analogous to the case of rectangular pulses^{18,19)}, and is so described by

$$d(A_f A)/dt = -k_p(A_f A) \quad (3)$$

where A is the fraction of a contacted area between particles, A_f is the final contacted area fraction, and k_p is the rate constant. Eqn. (3) is then rewritten as the regression equation of

$$-\text{Ln}\{(A_f A)/(A_f A_i)\} = k_p t \quad (4)$$

where A_i is the initial contacted area fraction. In order to obtain A against t at a certain k_p , eqn. (4) is transformed into the following equation:

Fig. 10 Relative density versus time for amorphous B_4C under $W_m = 2\text{ kW}$. This figure compares the results of heating rates of 10 and $20\text{ K}\cdot\text{min}^{-1}$.

Fig. 11 Temperature necessary for full densification of amorphous B_4C versus power output and applied stress. This figure includes the cases of PECS and millimeter wave micro-pressure sintering.

$$A = A_f(A_f A_i)/\exp(k_p t) \quad (5)$$

Fig. 12 shows the numerical analysis of determining the optimized k_p in the first-stage densification of amorphous B_4C under 2 kW: (a) logarithmic $(A_f A_i)/(A_f A)$ where $A_f = 1$ and $A_i = 0.0015$ as a function of time, and (b) calculated A using eqn. (5) against D , referring to Fig. 10. This figure compares the results of $W_m = 0.51$ and 0.7 kW .

The optimized numerical calculation is carried out, taking account of that the neck formation during the discharging of rectangular pulses occurs via bulk transport, having the approach of two spherical amorphous particles under applied stress, which is characterized by a simple geometric relationship to largely different $A^{20)}$. The A (calculated with A_i of 0.07 at 1 s) versus D relation is found to converge on the concave curved line where the boundary condition that A goes to 1 at $D=1$ is satisfied with the most proper value of $2.2\cdot 10^{-1}\text{ s}^{-1}$ for the level of k_p . For this procedure, it is convenient to express a concave relation of A on D explicitly

Fig. 12 Optimized numerical calculation, based on the neck formation law, of the first-stage densification under the constant power output of 2 kW for amorphous B₄C: (a) $-\ln\{(A_r - A)/(A_r - A_i)\}$ as a function of time, and (b) fraction of the contacted area between particles as a function of relative density. This figure compares the results in the cases for $W_m = 0.51$ and 0.7 kW. The dashed line represents the computed concave curve using an approximation function of $A = (1/D^m)(D - D_0)/(1 - D_0)$ for millimeter wave neck formation, and the solid line represents the convex curve based on $A = az/4\pi r^2 = D(D - D_0)/(1 - D_0)$ for powders densification by particle deformation.

as an approximation function of $A = (1/D^m)(D - D_0)/(1 - D_0)$ where m is the exponent. The computed concave curve for chosen m and D_0 is depicted as the dashed line. For comparison, in the case for the densification of powders by particle (with the radius r) deformation, the total contact area normalized by the surface area for a compact with $0.64 < D < 0.9$ is written as an equation of $A = az/4\pi r^2 = D(D - D_0)/(1 - D_0)$ where a is the average contact area and z is the average number of contacts per particle²⁰. This A - D relation is represented as the solid convex curve in Fig. 12 (b). For an increase in W_m from 0.51 to 0.7 and 2 kW, the A - D curve has a larger concavity identified with $m = 0.7, 1.6$ and 2.9 , and is unlike the single master curve; k_p is increased from 1.4 to $9 \cdot 10^{-3}$ and $2.2 \cdot 10^{-1} \text{ s}^{-1}$ according to an equation of $k_p = \exp(-3328 \text{ kW} \cdot \text{s}^{-1}/W_m)$. In addition, for $W_m = 2$ kW, D deviates from the concave curve at 8 s, following a rapid increase up to a level of the densification via plastic flow that is isotropic in nature.

The physical meaning underlying above mentioned equations is discussed. Eqn. (3) says that successive neck formation occurs only at newly diffused spark sites in possible sparking area

($A_r - A$) avoiding a growth of existing necks. Also, the concave function form of $A(D)$, that is an accelerating increase in D which is characterized by m , implies that with time, an overlapping area between particles increases at decreasing sites. Then, an exponential power output dependence of $k_p(W_m)$ may come from increments in site number of neck formation and local temperature, according to an Arrhenius relationship, with W_m . Besides, the transition of A from concave to convex function is related to an abrupt growth of present necks under higher effective stresses at a risen sintering temperature. The foregoing could be self-consistent at least for the fundamental physics of artificial neck formation. A further study on an exact formulation of concave function and $k_p(W_m)$ will be served for a detail discussion on a mechanism involved.

4.2 A field activated amorphous consolidation mechanism

The constitutive equation of the millimeter wave consolidating strain rate $\dot{\epsilon}_n$ that non-Newtonian flow contributes under a high effective stress σ_{eff} here is expressed with

$$\dot{\epsilon}_n = \dot{\epsilon}_0 \sinh(C\sigma_{\text{eff}}/RT) \exp(-Q/RT) \quad (6)$$

where $\dot{\epsilon}_0$ is a pre-exponential term, C is the constant, Q is the apparent activation energy for non-Newtonian flow and R is the gas constant. In order to deduce Q and $\dot{\epsilon}_0$, eqn. (6) then becomes the following relationship of the form:

$$\ln\{\dot{\epsilon}_n / \sinh(C\sigma_{\text{eff}}/RT)\} = \ln\dot{\epsilon}_0 - Q/RT \quad (7)$$

For a lower density compact with $0.64 < D < 0.9$ under an isostatic pressing, the effective stress applied to a contacted area between particles²⁰ is given by an equation of $\sigma_{\text{eff}} = \sigma(1 - D_0)/\{D^2(D - D_0)\}$ where D_0 is the relative density at the onset of consolidation. The use of this effective stress seems permissible for a compact completing major particle rearrangements, with an aspect ratio (diameter divided by height) of more than three, in constrained compression where a pseudo-hydrostatic pressure holds.

Fig. 13 shows the optimized numerical calculation using eqn. (7) for the second-stage densification in the constant-rate heating of amorphous B₄C. In the plot of $\ln\{\dot{\epsilon}_n / \sinh(C\sigma_{\text{eff}}/RT)\}$ with an appropriately selected C versus reciprocal temperature, a linearity is seen over nearly the entire range of $\dot{\epsilon}$ as in Fig. 5; hence Q is deduced in a temperature range 343 to 417 K. For a series of C , thus-obtained Q shows a variation, consistent with a change of the square of the correlation coefficient r in a least squares method as shown in Fig. 13 (b). Taking the most proper value of $1.6 \cdot 10^{-5} \text{ J} \cdot \text{mol}^{-1} \cdot \text{Pa}^{-1}$ for C at the maximum r^2 of 0.9918 , Q is determined to be $42.5 \text{ kJ} \cdot \text{mol}^{-1}$ for amorphous B₄C under a W_m of $0.55 \pm 0.03 \text{ kW}$ in the case of $10 \text{ K} \cdot \text{min}^{-1}$, and concomitantly $\dot{\epsilon}_0$ is deduced at 294 s^{-1} . This low Q level, which is comparable to $50.1 \text{ kJ} \cdot \text{mol}^{-1}$ for amorphous Ti₅₀Al₅₀⁹⁾, is acceptable for non-Newtonian flow that could proceed via correlated motion of atoms.

At the increasing heating rate to $20 \text{ K} \cdot \text{min}^{-1}$, the pressure sintering strain rate holds a higher level than that of $10 \text{ K} \cdot \text{min}^{-1}$

Fig. 13 Numerical analysis of the second-stage densification under the constant-rate heating of amorphous B_4C : (a) sintering strain rate reduced by the term of effective stress component $\sinh(C\sigma_{eff}/RT)$ versus reciprocal temperature; (b) apparent activation energy and coefficient of determination against C .

over a testing temperature regime. That is, under a W_m increased to 0.8 ± 0.05 kW, Q is decreased to $35.4 \text{ kJ} \cdot \text{mol}^{-1}$ with an acceptable r^2 of 0.9732, and at the same time $\dot{\epsilon}_0$ decreases to 122 s^{-1} . The second-stage field assisted densification is therefore described by an equation of $\dot{\epsilon}_n = \dot{\epsilon}_0(W_m) \sinh(C\sigma_{eff}/RT) \exp\{-Q(W_m)/RT\}$, where $\dot{\epsilon}_n$ is treated as a subordinate variable of $\dot{\epsilon}_0$ and Q , as in the case of PECS²¹⁾. This equation could provide the process control in terms of any of the variables involved. Also, one can simply say that the accelerated densification by an increased W_m comes from energizing a disordered structure and subsequently lowering a thermally activated barrier in non-Newtonian flow mobility.

5 Conclusions

The millimeter wave pressure sintering combined with process control techniques of constant-rate heating and constant-power output is proposed as a method of synthesizing a fully densified amorphous B_4C through the definitely divided densification far below a glass transition. The first-stage rapid densification is phenomenologically described by a law of neck formation based on the first-order chemical reaction, $d(A_r-A)/dt = -k_p(A_r-A)$ with $k_p = \exp(-3328 \text{ kW} \cdot \text{s}^{-1}/W_m)$. The field assisted second-stage densification via non-Newtonian flow is fairly well expressed with $\dot{\epsilon}_n = \dot{\epsilon}_0(W_m) \sinh(C\sigma_{eff}/RT) \exp\{-Q(W_m)/RT\}$ with a decreasing Q from

42.5 to $35.4 \text{ kJ} \cdot \text{mol}^{-1}$ under an increased W_m by the increased heating rate from 10 to $20 \text{ K} \cdot \text{min}^{-1}$. With the proposed technique, ultralow-temperature sintering of amorphous B_4C can be achieved with the minimum full density temperature of 386 K in heating.

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