1 Introduction

Einleitung

The current climate protection plans, such as the Carbon Offset and Reduction Scheme for International Aviation (CORSIA) by the International Civil Aviation Organization (ICAO) [INTE16] or Flightpath 2050 by the European Commission [EURO11], pose challenges for the international aviation industry. Carbon dioxide, nitrogen oxide and noise emissions have to be drastically reduced [EURO11], even though an increased number of passengers is forecasted [BOEI20]. One way of saving fuel and, thus, reducing emissions is to increase the efficiency of aircraft engines. Among others, this is achieved through higher turbine inlet temperatures [AUSM02]. However, the turbine inlet temperature was limited by the limited development potential of metallic superalloys [RAET13]. Thus, also the efficiency of turbines was confined. Therefore, fiber-reinforced non-oxide ceramics (NOx_f) have become the focus of material development in recent years. Due to their high thermal resistance, they enable higher turbine inlet temperatures. Also, cooling air can be reduced or omitted, which leads to an improvement of combustion efficiency and a reduction of emissions. Due to their additional high tolerance to damage and shock loads as well as their low specific weight, they are suitable as construction materials for future engine components [LONG20]. High quality requirements are placed on the manufacture of safety-critical engine components, as component failure is often due to manufacturing defects [NATI98]. This is why the material is mostly machined by grinding [GAVA19]. During grinding, it is important to ensure reduced damage at the surface and the peripheral zone of the workpiece. A time and cost intensive grinding process design results because the material removal mechanisms during grinding, i.e. all processes that lead to the cutting of the material, and the resulting grinding strategies for NOx_f are only insufficiently known.

Therefore, this research work focuses on the analysis of the cause-effect-relationships between the material removal mechanisms in grinding of NOx_f depending on the process input and state variables, the fiber orientation in relation to the grinding direction, the fiber materials and the grinding process parameters. For this analysis, the structure and the material properties at micro and macro scale are analyzed. Subsequently, single grain cutting tests as well as numerical simulations are carried out in order to analyze the fundamental material removal behavior during single grain engagement. The gained knowledge is transferred with additional grinding tests to the multi grain engagement. The obtained knowledge is accumulated within a heuristic descriptive and explanatory model. Thus, this research work contributes to the basic understanding of the material removal behavior of NOx_f during grinding resulting in the derivation of an optimized grinding strategy as a recommendation for action saving time and cost.

Einleitung

Die aktuellen Klimaschutzpläne wie das CARBON OFFSET AND REDUCTION SCHEME FOR INTERNATIONAL AVIATION (CORSIA) der INTERNATIONAL CIVIL AVIATION ORGANIZATION (ICAO) [INTE16] oder FLIGHTPATH 2050 der europäischen Kommission [EURO11] stellen die internationale Luftfahrtindustrie vor Herausforderungen. Kohlenstoffdioxid, Stickoxid- und Lärmemissionen müssen drastisch reduziert werden [EURO11], obwohl ein steigendes Passagieraufkommen prognostiziert wird [BOEI20]. Eine Möglichkeit, Treibstoff einzusparen und damit Emissionen zu reduzieren, ist die Effizienzsteigerung von Flugzeugtriebwerken. Dies wird unter anderem durch höhere Turbineneintrittstemperaturen erreicht [AUSM02]. Allerdings wurden die Turbineneintrittstemperatur durch das begrenzte Entwicklungspotential metallischer Superlegierungen begrenzt [RAET13]. Damit war auch der Wirkungsgrad von Turbinen eingeschränkt. Daher sind faserverstärkte Nichtoxidkeramiken (NOx_f) in den letzten Jahren in den Fokus der Materialentwicklung gerückt. Aufgrund ihres hohen thermischen Widerstands ermöglichen sie höhere Turbineneintrittstemperaturen. Außerdem kann Kühlluft reduziert oder weggelassen werden, was zu einer Verbesserung der Verbrennungseffizienz und einer Reduzierung von Emissionen führt. Aufgrund ihrer zusätzlich hohen Toleranz gegenüber Stoßbelastungen sowie ihres geringen spezifischen Gewichts eignen sie sich als Konstruktionswerkstoff für zukünftige Triebwerkskomponenten [LONG20]. Höchste Qualitätsanforderungen werden an die Herstellung sicherheitskritischer Triebwerkskomponenten gestellt, da das Versagen dieser Komponenten oft auf Fertigungsfehler zurückzuführen ist [NATI98]. Deshalb wird das Material meist durch Schleifen bearbeitet [GAVA19]. Der Schleifprozess muss so ausgelegt werden, dass geringe Werkstückschädigungen an der Oberfläche und der Randzone der Bauteile entstehen. Eine zeit- und kostenintensive Schleifprozessauslegung ist die Folge, da die Zerspanungsmechanismen beim Schleifen, also alle Prozesse, die zur Zerspanung des Materials führen, und die daraus resultierenden Schleifstrategien für NOx_f nur unzureichend bekannt sind.

Daher werden in dieser Forschungsarbeit die Ursache-Wirkungs-Beziehungen zwischen den Zerspanungsmechanismen beim Schleifen von NOx_f in Abhängigkeit von Prozesseingangs- und Zustandsgrößen, der Faserorientierung in Bezug auf Ritz- und Schleifrichtung, Fasermaterialien und Schleifprozessparameter analysiert. Für diese Analyse werden die Struktur und die Materialeigenschaften der untersuchten NOx_f im Mikro- und Makromaßstab analysiert. Anschließend werden grundlegende Ritzuntersuchungen sowie numerische Simulationen durchgeführt, um das grundlegende Zerspanungsverhalten beim Einzelkorneingriff zu analysieren. Die gewonnenen Erkenntnisse werden mit zusätzlichen Schleifversuchen auf den Mehrkorneingriff übertragen. Das gewonnene Wissen wird in einem heuristischen Beschreibungs- und Erklärungsmodell zusammengefasst. Somit trägt diese Forschungsarbeit zum grundlegenden Verständnis des Zerspanungsverhaltens von NOx_f beim Schleifen bei, woraus die Ableitung einer optimierten Schleifstrategie als zeit- und kostensparende Handlungsempfehlung resultiert.

2 State of the art

Stand der Technik

The state of the art provides the basis and justification for the subject of this research work. Section 2.1 focusses on the basics of ceramics, their classification, structures and resulting properties. The classification and properties of fiber-reinforced ceramics, also called ceramic matrix composites (CMCs), are presented in Section 2.2. The fundamentals of ductile and brittle material removal behavior in grinding are discussed in Section 2.3. Section 2.4 focuses on the state of the art of grinding both monolithic and fiber-reinforced non-oxide ceramics. Finally, the knowledge gaps identified in the state of the art are summarized in a conclusion and the economical and scientific problem statements are derived in Section 2.5.

2.1 Basics of ceramics

Grundlagen der Keramiken

About 30,000 years ago, ceramics as the first artificial material was created by shaping and solidifying soils [SALM07, p. 3]. Due to the long history of ceramics, there are many different definitions for this material. However, there is no universally accepted definition for the term ceramics. The term ceramics can be either assigned to a class of materials, which can be distinguished from metals or plastics, or to an entire branch of industry dealing with ceramics [SALM07, p. 2]. HAASE defined the term ceramics as materials which are created by forming a powder and solidifying the shape through the application of high temperatures [HAAS68]. HENNICKE specified the definition and defined ceramics as materials which are "inorganic, non-metallic, and poorly soluble in water and at least 30 % crystalline. Ceramics are formed from a raw material at room temperature and receive their typical material properties through a heat treatment, usually above $T = 800$ °C" [HENN67]. Meanwhile, according to WERNER ET AL., all non-metallic and inorganic materials belong to the group of ceramics [WERE16, p. 323]. The difference between ceramic and metallic materials is defined with the help of the temperature coefficient of electrical resistance α_R [WERE16, p. 323], or by the type of chemical bond [RÖSL12, p. 16].

There are no discrete molecules in ceramics. However, there are spatial arrangements of one or more types of atoms, either ordered as a crystal lattice or randomly as glass [WERE16, p. 323]. The type, quantity, size, shape, orientation and distribution of these phases including pores are referred to as the microstructure. The microstructure influences physical or chemical properties of ceramics [SALM07, pp. 144–146].

The types of bonds, which determine the cohesion of the microstructure, are divided into primary bonds and secondary bonds. Chemical effects cause primary bonds. They have a higher bond energy in comparison to the secondary bonds, which are based on physical interactions. However, most bonds rarely consist of just one of the bond types. Instead, they form hybrid bonds, because several bond types exist parallel. Table 2-1 summarizes the different bond types and their bond energy as well as their range [SALM07, pp. 9–15].

Table 2-1: Overview of primary and secondary bond types [SALM07, p. 10]
Übersicht über Primär- und Sekundärbindungen [SALM07, p. 10]

Interaction type	Bond energy E_b [kJ/mol]	Range r_b [10^{-12} m]
Primary bond type		
Covalent	60–700	100–200
Ionic	600–1,100	up to 300
Metallic	110–350	100–200
Secondary bond type		
VAN DER WAALS linkage	0.08–40	300–500
LEWIS-acid-base-interactions	<80	<300
Hydrogen bridge bond	< 0	<200

Primary bonds of ceramics are characterized by ionic and covalent bonds, which result in a full outer electron shell [RÖSL12, p. 16]. These bond types cause strong bonding forces, which lead to high stability, a high YOUNG'S modulus E and a low coefficient of thermal expansion α, as well as high brittleness [TIET94, pp. 4–6].

2.1.1 Classification of ceramics

Klassifizierung von Keramiken

Ceramics are classified regarding their structure, their chemical composition and their scope of application. However, the use of this nomenclature can lead to overlaps. One nomenclature distinguishes ceramics according to their grain size d_g. The limit is approximately 0.1 mm < d_g < 0.2 mm. If the grain size of the ceramics is smaller than the limit ceramics are called fine ceramics. Coarse ceramics consist of structural components larger than that limit [INFO03, pp. 18–20]. With regard to their application, ceramics are divided into refractory ceramics and technical ceramics [SALM07]. The softening point defines refractory ceramics and their resulting applications. If the softening point of ceramics is above T_s > 1,500 °C, they are called refractory products, at temperatures above T_s > 1,800 °C they are called highly refractory products [ELST07, p. 719]. Technical ceramics are defined as ceramics which have been developed for technical applications in order to distinguish them from utility ceramics such as crockery or sanitary articles and from building ceramics like bricks or tiles. Technical ceramics are subdivided into traditional technical ceramics and advanced ceramics (Figure 2-1). Following SALMANG ET AL., refractory ceramics are also considered traditional technical ceramics. Spark plugs, grinding wheels or wear parts are considered as traditional technical ceramics. Advanced ceramics are further divided into functional ceramics and structural ceramics, which are attributed to high strength, toughness and thermal shock resistance [SALM07, pp. 783–784]. Regarding their chemical composition, silicate ceramics containing silicon oxide, oxide ceramics composed of metallic elements and oxygen and non-oxide ceramics containing oxygen-free compounds are highly relevant to technical use [RÖSL12, p. 16].

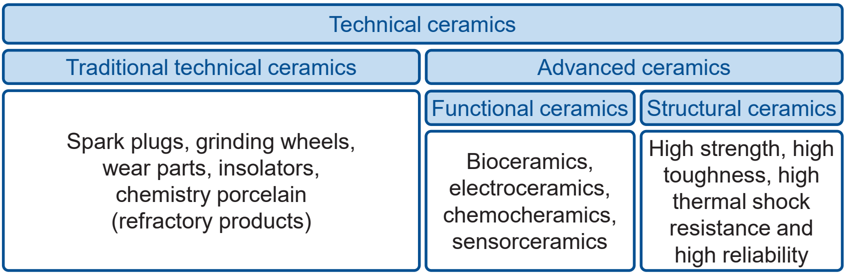

Figure 2-1: Classification of technical ceramics following [SALM07, p. 783]
Einteilung der technischen Keramiken

Silicate ceramics are dominant in fine ceramic products. The essential components of these multi-phase materials are clay and kaolin, feldspar and soapstone as silicate carriers. In the manufacturing process, a high proportion of over twenty weight percent of glass phase is generated, whose main component is silicon oxide (SiO_2). Due to the relatively low sintering temperatures, good process control and high availability of natural raw materials, silicate ceramics are significantly cheaper than oxide or non-oxide ceramics. Silicate ceramics are used in heating technology, measurement and control technology, process and environmental technology, high and low voltage technology with typical applications such as insulators, fuse cartridges, catalysts, housings and in a variety of applications in electrical installation technology [INFO03, pp. 20–21].

Basic components of oxide ceramics are either oxides of a single element or oxygen ions in combination with cations of different elements in their crystal lattice, which are called mixed oxides. Aluminium oxide (Al_2O_3), zirconium oxide (ZrO_2) and magnesium oxide (MgO) are examples for oxide ceramics with single elements. Spinel ($MgAl_2O_4$), barium titanate ($BaTiO_3$) and zinc iron oxide ($ZnFe_2O_4$) contain oxygen ions in combination with corresponding cations [SALM07, p. 784]. The major bond type of oxide ceramics are ion bonds [SCTT91]. The proportion of glass phase of oxide ceramics is close or equal to zero. Raw materials for oxide ceramics are manufactured synthetically and have a high degree of purity. Oxide ceramics are often used in electrical engineering as well as structural material [INFO03, pp. 21–22].

Non-oxide ceramics are based on compounds of boron (B), carbon (C), nitrogen (N) and silicon (Si). Non-oxide ceramics have a high proportion of covalent bonds [INFO03, p. 22]. These covalent bonds ensure an improved YOUNG'S modulus E, high strength σ_f and high hardness H, combined with high operating temperatures T_o, high corrosion resistance and high wear resistance of non-oxide ceramics [SALM07, pp. 888–889]. Silicon carbide (SiC), silicon nitride (Si_3N_4), aluminum nitride (AlN), boron carbide (B_4C) and boron nitride (BN) are relevant examples of non-oxide ceramics for application [INFO03, p. 22] in aerospace and turbine industry [HÜLS14, p. 139].

The investigation of non-oxide CMC materials consisting of carbon fiber and silicon carbide fiber reinforced silicon carbide (C/SiC, SiC/SiC) is the focus of this work. In order to create a deeper understanding of these investigated materials, the structure and the properties of SiC and C are described in the Sections 2.1.2 and 2.1.3.

2.1.2 Structure and properties of silicon carbide

Aufbau und Eigenschaften von Siliciumcarbid

Next to silicon nitride (Si_3N_4), SiC is the most important non-oxid ceramic material [SALM07, p. 897]. SiC is the only solid-state combination of the two elements Si and C. The high degree of covalent bonding results in the high hardness and the outstanding chemical resistance to acid or lye of SiC. SiC bears not only covalent bonds. Due to the different electronegativity of the elements silicon and carbon, SiC also has ionic bond components [KAIS14, pp. 5–7].

Depending on the modification, SiC occurs in a hexagonal modification as α-SiC (Figure 2-2 (a)) or in a cubic modification as β-SiC (b). The cubic phase is often referred to as the low-temperature modification and the hexagonal phase is considered as the high-temperature modification, due to the observation that β-SiC transforms into α-SiC at around T_{Trans} = 2,100 °C [SALM07, p. 897]. The exact transition temperature is influenced by the surrounding atmosphere. Pure α-SiC is colorless whereas pure β-SiC appears in yellow color. Due to the polycrystalline structure of SiC and various impurities in the crystals, SiC material appears greenish, gray-blue or black. Pure SiC does not melt but only dissociates at T_{Disso} = 2,540 °C [HÜLS14, p. 141].

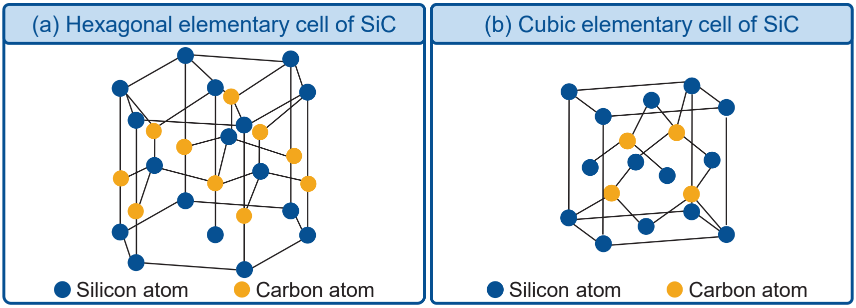

Figure 2-2: Elementary cells of SiC [SALM07, p. 898]
Elementarzellen von SiC

Each Si atom has a four-fold bond with C as its nearest neighbor. As a consequence, C has a four-fold bond with Si. This results in a structure of Si-C bilayers consisting of Si-C_4 and Si_4-C tetrahedrons [KAIS14, pp. 6–7]. The Si-C bilayers in a stacking sequence of ABCABC result in the cubic β-SiC, also called Zincblende structure (3C). The stacking sequence of ABAB is called Wurtzite structure (2H) and present at α–SiC. The resulting structures are classified according to the Ramsdell classification scheme, in which numbers provide information about the quantity of different bilayers and letters give information about the present modification. Nevertheless, the stacking sequence of these tetrahedron layers can vary within wide ranges

[SALM07, pp. 897–898], which results in extremely pronounced polytypes of SiC [KAIS14, pp. 6–7]. Polytypes are defined as a structures with the same chemical composition differing only along one crystallographic axis [KAIS14, pp. 6–7]. More than 250 different polytypes of SiC are known [CHEU06]. Since all polytypes consist of the same structural layers and since these are densely packed, their physical properties vary only slightly. Generally, α-SiC is a mixture of different polytypes [SALM07, pp. 897–899].

According to the ACHESON-process, the technical production of SiC is carried out using quartz sand (SiO_2) and coke or carbon (C) at temperatures of up to $T_{Acheson}$ = 2,400 °C. In this case, a thermal reduction takes place according to equation (2.1). Carbon monoxide (CO) is created as a side product [SALM07, p. 900].

$$SiO_2 + 3\,C \rightarrow SiC + 2\,CO \tag{2.1}$$

The reaction proceeds through various intermediate stages. The resulting SiC is available as α-SiC [SALM07, pp. 900–904]. Cubic β-SiC is mainly produced from an aqueous sugar solution and silica gel using the so-called GENERAL ELECTRIC process [PREN63], which ensures a very even molecular distribution of the C and the Si compounds. After drying the mixture, thermal reduction takes place at T_{GE} = 1,800 °C. Besides this, SiC can be produced in a highly pure form by means of synthetic processes [GENE72; SCHW78]. A synthetic method of SiC production is the pyrolysis of organosilicon precursors, which is mainly used for fiber production. The precursors already contain the substances to be synthesized in the stoichiometric ratio of SiC. Cubic β-SiC is usually obtained by pyrolysis. The ceramic moldings produced in the pyrolysis have a high porosity of $20\,\% < P < 25\,\%$ due to the associated gas evolution. A reduction in porosity is necessary [TIET94, p. 31], but also expensive [SALM07, p. 906]. Due to the high proportion of covalent bonds, high effort is required to process SiC into dense bodies. Various types of materials are derived, which have very specific properties and resulting applications. Abbreviations, material types and specific properties are listed in Table 2-2.

Table 2-2: SiC material types and specific properties [SALM07, pp. 906–922]
SiC-Werkstofftypen und spezifische Eigenschaften

Abbreviation	Material type Unit	Porosity P [vol. %]	YOUNG'S modulus E [GPa]	Hardness H [GPa]	Fracture toughness K_{IC} [$MPam^{0.5}$]
SSiC	Sintered	< 2	410	23–24	2.5–4.7
LPSSiC	Liquid phase sintered	< 1	420	22–23	4–6
HPSiC	Hot-pressed	< 0.1	440	22–25	3–4
HIPSiC	Hot-isostat.-pressed	0	450	25–27	3–4
CBSiC	Vitrified-bonded	20	100	-	1–2
NBSiC	Nitride-bonded	20	100	-	-
RSiC	Recrystallized	20	250	-	1–2.5
RBSiC	Reaction-bonded	5-20	250	-	1.5–2.5
SiSiC	Silicon infiltrated	0	400	20–22	3–4.7

In general, SiC materials are characterized by high hardness H, good thermal conductivity λ and good temperature strength σ_T. The properties of the individual SiC material types vary due to the different composition and structure influenced by type of binder phase, porosity and grain size [SALM07, pp. 906–922].

Figure 2-3 shows the bending strength σ_B as a function of temperature T. CBSiC has a relatively low bending strength σ_B, which increases with increasing SiC content. Although the remaining porosity of RSiC has a decisive negative influence on the bending strength σ_B, RSiC shows higher bending strength σ_B than CBSiC. In comparison to that, SiSiC has improved bending strength σ_B values. The loss in bending strength σ_B at T = 1,400 °C is caused by the presence of metallic silicon, which has a melting point of T_M = 1,414 °C. LPSiC, HIPSiC and HPSiC have the highest bending strength σ_B of this material class [SALM07, p. 919].

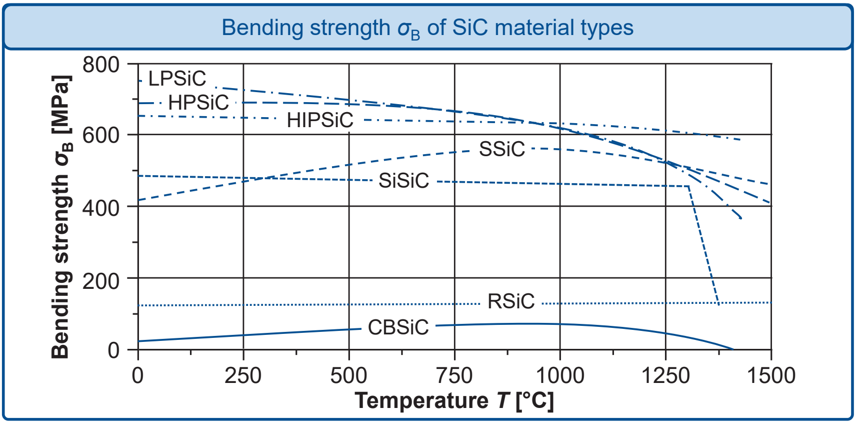

Figure 2-3: Bending strength σ_B of SiC material types [SALM07, p. 919]
Biegefestigkeit von SiC Werkstoffen

The chemical reaction behavior of SiC is useful for various applications. SiC can be used in oxidizing atmospheres due to a protective SiO_2 layer forming on the SiC surface. The diffusion of the oxygen through this layer then becomes limiting for the described reaction. The presence of water increases the reaction rate. For example, SiC is used as a grain material at grinding wheel manufacturing. Likewise, the rollers for high-temperature roller furnaces are made of extruded RSiC or SSiC. Due to their high purity and residual porosity, these materials show high performance regarding heat conduction and are resistant to thermal shock. SiC is also used for heating elements and heat exchangers up to temperatures of T = 1,500 °C [SALM07, p. 920]. Moreover, SiC is used as fiber and reinforcement material in CMC applications. CMC materials with SiC fibers are especially applied as high-temperature components in hot sections of engines for power and propulsion technologies above application temperatures of metallic alloys of T_{App} > 1,100 °C [DICA05, pp. 33–34]. Structure and properties of SiC fibers are presented in Section 2.2.2.