Microstructure-property relations of eutectic V-Si and V-B alloys

M Regenberg, G Hasemann, C Müller and M Krüger
Otto-von-Guericke University Magdeburg, Faculty of Mechanical Engineering, Institute of Materials and Joining Technology, Universitätsplatz 2, 39106 Magdeburg, Germany
maximilian.regenberg@ovgu.de

Abstract. Subject of the present work are principle studies on V-Si and V-B alloys with hypo- and hypereutectic compositions, regarding the effects of primary phase solidification and the formation of binary microstructures on the resulting mechanical properties. For this purpose, button shaped samples were produced by arc-melting, metallographically prepared and analyzed by scanning electron microscopy. Furthermore, room temperature compression tests have been carried out and the resulting deformations were subsequently analyzed by using electron backscatter diffraction. As a conclusion, the materials’ failure behavior was characterized and discussed with respect to the developed microstructure. The main goal of the experiments conducted is to critically discuss previously published works on the V-Si and V-B alloying systems and to obtain more profound insights on microstructure-property relations with respect to the strengthening effects and plastic deformability.

1 Introduction
Due to the variety of attractive offers in civil aviation, the airplane has become one of the most popular means of transportation in recent decades. According to a statistic of the German VuMA (Arbeitsgemeinschaft Verbrauchs- und Medienanalyse), about 56.5% of residents used the airplane (scheduled flights, charter flights and low-cost airlines combined) to reach their (holiday) destination in the years 2016-2018 [1]. Furthermore, a rising trend in the number of transported passengers (private and business travelers) in air traffic can be stated [2]. These statistic numbers illustrate the growing importance and popularity of air traffic as means of mass transport, while simultaneously the question regarding the environmental compatibility of flying also gains increasing attention. Although the airplane often is the most time-efficient way of traveling, a drawback can be seen if the ecological aspects are considered.
Table 1. Comparison of average emissions from individual means of passenger transport (reference year 2017, collected by the german Federal Environment Agency; emissions are expressed in grams per passenger kilometer (g/Pkm); emissions from the provision and conversion of energy sources into electricity, petrol, diesel and kerosene are included [3].

|                        | Passenger | Coach | Railway, long-dist. | Airplane | Line bus | Railway, local | Tram, subway |
|------------------------|-----------|-------|---------------------|----------|---------|----------------|-------------|
| Greenhouse gases      | (in g/Pkm)| 139   | 32                  | 36²      | 201⁴    | 75             | 60          | 64          |
| Carbon monoxide       | (in g/Pkm)| 0.6   | 0.04                | 0.02     | 0.13    | 0.05           | 0.04        | 0.04        |
| Volatile hydrogen     | (in g/Pkm)| 0.14  | 0.01                | 0        | 0.04    | 0.03           | 0.01        | 0           |
| Nitrogen oxide        | (in g/Pkm)| 0.34  | 0.17                | 0.04     | 0.51    | 0.28           | 0.18        | 0.06        |
| Particular matter     | (in g/Pkm)| 0.004 | 0.003               | 0        | 0.004   | 0.002          | 0.002       | 0           |
| Capacity utilisation  | 1.5 ppl/car| 60%   | 56%                 | 82%      | 21%     | 27%            | 19%         |

¹ Category “Coach” includes buses on irregular schedules (e.g. school trips and coffee cruises) and long-distance bus services
² CO₂, CH₄ and N₂O are stated in CO₂-equivalents
³ Stated emission factors for railways are based on the average energy-mix in Germany
⁴ In consideration of all climate-impacting effects of air traffic (EWF = Emission Weighting Factor = 2)
⁵ Methane not included
⁶ Abrasion not included

Regarding the emissions of greenhouse gases and nitrogen oxides, aircraft turbines show the highest values. Additionally, the numbers of emitted carbon monoxide, volatile hydrogen and particulate matter are comparably high, as can be seen in Table 1. In order to reduce emissions during the operation of flight, the efficiency of the aircraft engine (turbine) has to be increased [4]. To fulfill these requirements the development and application of novel structural materials is inevitable.

Reports on low alloyed vanadium-based materials with additions of Cr, Ti or Si for the use in fusion reactor applications (f.e. in self cooled Li blankets) have already been published since the 1990’s [5-7]. Besides that, more recent findings have shown that, due to their low density and high mechanical strength at elevated temperatures up to 1000 °C, V-based materials show a high potential for the use as high temperature structural materials [8, 9]. In particular, the material class of eutectic vanadium-silicon alloys, with operating temperatures up to 700 °C and simultaneously good thermal shock resistance, have proven increased suitability for the potential use as structural components in the field of stationary gas- and aircraft turbines [10, 11]. Extensive research has been carried out by the working group of Krüger et al. in 2016 [8], concerning the mechanical properties of a V-9Si-13B alloy. The samples investigated were produced via a powder metallurgical route and were then compared to the state-of-the-art single-crystal nickel-based superalloy CMSX-4®. The results showed that the V-based alloy can compete with the superalloy, especially in terms of yield stress and the higher specific strength, due to a lower density [8].

In principle, the present study aims to re-evaluate the eutectic compositions of vanadium-silicon and vanadium-boron alloys as they have been stated in the phase diagrams by Smith [12] and more recently by Lima-Kühn et al. [13] (V-Si system), respectively by Spear et al. [14] and Nunes et al. [15] (V-B system) in an experimental way. In addition, the aforementioned authors reported on the obtained microstructure-property relations of the vanadium-based alloys, with regard to present volume fractions of eutectic in the produced samples.
2 Experimental procedures
The alloying systems vanadium-silicon and vanadium-boron were investigated in the presented study. A total of six button-shaped samples with different hypo- and hyper-eutectic alloying compositions (shown in Table 2) were melted in a water-cooled copper crucible, using a non-consumable tungsten electrode in a conventional vacuum-arc-furnace (comparable to Arc Melter MAM-1, Edmund Buehler GmbH, Bodelshausen). The raw materials consisted of vanadium chippings (99.7%), silicon granules (99.99%) and boron granules (99.0%), which were carefully weighed in. During arc-melting, the buttons were flipped and re-melted five times to ensure good homogeneity. The actual chemical alloy composition after melting was verified using inductively coupled plasma optical emission spectroscopy (ICP-OES), see Table 2.

Table 2. Nominal and by Inductively Coupled Plasma-Optical Emission Spectrometry (ICP-OES) measured composition of the V-Si and V-B alloy.

| Nominal composition (at.%) | Si (at.%) | Deviation of Si (%) |
|---------------------------|----------|--------------------|
| V-9.75Si                  | 10.73    | 1.10               |
| V-12.5Si                  | 13.00    | 1.04               |
| V-15.75Si                 | 16.95    | 1.08               |

| Nominal composition (at.%) | B (at.%) | Deviation of B (%) |
|---------------------------|----------|--------------------|
| V-5.5B                    | 5.78     | 1.05               |
| V-11B                     | 11.54    | 1.05               |
| V-14.5B                   | 14.18    | 0.98               |

Figure 1. Hypothetical compositions of the target alloys in the respective parts of the (a) V-Si phase diagram by Smith et al. [12] and (b) V-B phase diagram by Massalski [16].
In order to determine the composition of the samples considering the eutectic and hypo-, respectively hyper-eutectic fractions, the lever rule was used in the vanadium-rich proportions of the V-Si phase diagram according to Smith [12] and the V-B phase diagram compiled by Massalski et al. [16], as can be seen in Figure 1. It must be stated that, according to the displayed vanadium-boron phase diagram, the V-14.5B alloy is expected to solidify at a temperature slightly below the eutectic point, thus developing a hypoeutectic microstructure. In fact this is not the case, since recent investigations carried out by the Nunes work group [13, 14] have shown a more accurate prediction of the invariant reactions, which were taken into account for the present calculations.

The button-shaped samples were cut with a diamond coated saw blade on a cutting machine (Brillant 220, ATM Qness GmbH, Mammelzen) and embedded in epoxy resin. In the course of metallographic preparation, the specimens were subsequently ground on a semi-automatic grinding and polishing machine (Saphir 550, ATM Qness GmbH, Mammelzen) and polished on a single-specimen polishing machine (MiniMet 1000, Buehler, Lake Bluff), using 6 µm, 3 µm and 1 µm polycrystalline diamond suspension consecutively, finishing with a colloidal silica suspension (OP-S). Microstructural investigations were carried out with a scanning electron microscope (SEM; Merlin, Carl Zeiss AG, Oberkochen). In preparation for the SEM analysis, the samples were coated with a thin layer of iridium and afterwards scanned, using the backscattered electron mode (BSE) and wavelength-dispersive X-ray spectroscopy (WDS, Oxford Instruments, Abingdon). The phase fractions of the different microstructures were determined with the optical image analysis software FIJI, using different grey scale values of the images.

Standard cylindrical compression samples [19] of 3 mm in length and 1.8 mm in diameter were prepared by electro discharge machining (EDM) and surface-cleaned (removal of a burr and the surface tarnish) subsequently, using 800 and 1200 grit SiC paper. In addition, a 3 h lapping process with sandblasting granules (glass-beads) in ethanol was carried out, to avoid any impact of the surface condition on the subsequent mechanical tests. Room temperature compression tests were performed on a universal testing machine (Type 1381, Instron, Norwood) with a maximum load capacity of 30 kN and a load cell with an operating range of ± 10 kN. According to DIN 501006 [19], a constant cross-head speed corresponding to an initial (engineering) strain rate of $\dot{\varepsilon} = 10^{-3}$ s$^{-1}$ was used. As results of the compression tests, the compressive yield stress functions were determined using the 0.2% offset method and displayed in engineering compressive stress-strain curves. After testing, macroscopic images of the samples were obtained on a digital microscope (Keyence VHX-5000, Osaka). Additionally, the specimens were embedded in epoxy resin, grinded down to about half their cylindrical diameter and prepared afterwards, following the same steps as described above (see preparation of the button-shaped samples). Subsequently, the mechanical deformation was determined by means of local misorientation maps, using electron backscatter diffraction (EBSD, Oxford Instruments, Abingdon).

3 Results and discussion

The as-cast microstructure of the V-Si alloys is displayed in Figure 2. Significant differences between the primary phases and the eutectic areas were obtained in the alloying system, which are emphasized by different grey values. The stated values for the phase fractions represent the results of the optical image analysis. In Figure 2 (a) the V-9.75Si microstructure, consisting of primary solidified vanadium solid solution (V$_{SS}$) phase (≈ 60 vol.% phase fraction) and surrounding eutectic matrix is present, whereas Figure 2 (b) displays the eutectic arrangement of the V-12.5Si alloy and Figure 2 (c) depicts the hypereutectic V-15.75Si microstructure, containing primary solidified V$_3$Si dendrites (≈ 40 vol.% phase fraction) and eutectic microstructure. The results of the SEM analysis, regarding the V-B microstructures are displayed in Figure 3. The hypoeutectic V-5.5B alloy (Figure 3 (a)) shows fractions of V$_{SS}$ phase (≈ 60 vol.% phase fraction) and eutectic matrix, whereas the V-11B alloy in Figure 3 (b) is considered as near-eutectic since small proportions of V$_3$B$_2$ phase (< 3 vol.% phase fraction) were obtained in these specimens. The hypereutectic alloy V-14.5B (Figure 3 (c)) depicts a fraction of primary V$_3$B$_2$ (≈ 14 vol.% phase fraction) and a respectively larger fraction of the eutectic matrix, compared to alloy V-5.5B. It is noteworthy, that the vanadium-boron alloys formed a non-facettted-facetted
eutectic in contrast to the vanadium-silicon alloys which form a lamellar eutectic. This can be explained by deviating solid-liquid interfacial energies of the phases, different melting entropies of the eutectic phases and kinetic coefficients for coarse surfaces or different local supercooling effects [20-22]. The present deviations from the calculated compositions regarding both alloying systems can be explained by the rapid cooling conditions of the crucible, which prevents the diffusion of the elements and thus the formation of phases under equilibrium condition (as predicted by the lever rule) and the apparent inaccuracies in the published phase diagrams.

Figure 2. SEM-BSE images of the (a) hypoeutectic alloy V-9.75Si, (b) eutectic alloy V-12.5Si and (c) hypereutectic alloy V-15.75Si microstructures.

Figure 3. SEM-BSE images of the (a) hypoeutectic alloy V-5.5B, (b) near-eutectic alloy V-11B and (c) hypereutectic alloy V-14.5B microstructures.
The results of the room temperature compression tests are displayed in Figure 4. In order to provide a statistical evaluation, 4-6 samples of each alloy were tested, which provided a high reproducibility of the results. Considering the V-Si alloys (Figure 4 (a)), a strong correlation between the Si content and the compressive strength, as well as the maximum plastic deformation until failure was observed. This is in good agreement with the varying degrees of bulging of the samples. The hypoeutectic alloy V-9.75Si shows the highest grade of plastic deformation ($\varepsilon_{p,max} \approx 14\%$), respectively lowest strength ($\sigma_{0.2} \approx 849$ MPa), which is attributed to the high fraction of ductile $V_3Si$ phase in the alloy composition [23]. It is stated that the sample failed in a ductile regime, which is indicated by shear failure along the diagonal axis of the cylindrical specimens during the compression tests. The eutectic alloy V-12.5Si shows an increase in strength ($\sigma_{0.2} \approx 914$ MPa) and reduced plastic deformability ($\varepsilon_{p,max} \approx 7.5\%$), compared to V-9.75Si, which is caused by higher fractions of the intermetallic $V_3Si$ phase. Additionally, less bulging and shear plastic deformability of the eutectic specimen was found. The highest mechanical strength ($\sigma_{0.2} \approx 1202$ MPa) and the lowest plastic deformability ($\varepsilon_{p,max} \approx 3\%$) values were measured for the hypoeutectic alloy V-15.75Si, which possesses the highest percentage of the $V_3Si$ phase compared to all alloys tested. As mentioned before, the increasing phase fraction of intermetallic silicides may be interpreted as the origin for the increase in strength and embrittlement of the samples. Regarding the failure mechanism, it must be noted that there is significantly less ductility, even though the crack propagates in a diagonal direction, no measurable bulging of the cylinder and no shear behavior as such has been observed.

Considering the results of the compression-loaded V-B alloys (Figure 4 (b)), the dependency between alloyed B content and resulting mechanical properties cannot be derived as clear as those from the V-Si samples. Specifically the compressive stress-strain curves of the eutectic alloy V-11B and the hypereutectic alloy V-14.5B exhibit overlays and show minor differences concerning the compressive strength. The lowest strength ($\sigma_{0.2} \approx 272$ MPa) values were obtained from the V-5.5B samples, and an almost ideal plastic behavior was determined. The samples of the near eutectic V-11B alloy exhibited comparatively high strength values ($\sigma_{0.2} \approx 460$ MPa) and good plastic deformability ($\varepsilon_{p,max} \approx 20\%$). Again, it is noteworthy that the results are to be viewed in a more differentiated way, since the present material failure is still close to ideal plastic behavior and thus, crack characterization is not applicable. The bulging of the V-11B specimen is also distinctive and a deformation alongside the diagonal axis indicates failure by shear plastic deformation of the compression-tested samples. In comparison with the hypo- and hypereutectic vanadium-boron alloys, the results of the V-14.5B tests have shown higher compressive yield strength values ($\sigma_{0.2} \approx 548$ MPa) at high plastic deformations ($\varepsilon_{p,max} \approx 22\%$). Taking the corresponding results of the compression tests into account, it can be stated that the graphs show less similarity to an ideal-plastic curve. Additionally, bulging of these samples is less significant and the crack induced by the deformation along the diagonal points out to a minor brittle fracture behavior. Based on the observations, it can be concluded that a tendency towards lower deformability with increasing boron content (hence an increasing content of intermetallic vanadium boride phase) is present.
Figure 4. Engineering compressive stress-strain curves of the examined V-Si (top) and V-B (bottom) alloys and photographs of typical samples after compression testing, see inset.
The EBSD misorientation maps, see Figure 5, are in good agreement with the results obtained from the compression tests. Regarding the low alloyed V-9.75Si sample (Figure 5 (a)), it can be stated that the energy introduced during compression is almost entirely transferred via the deformation of the vanadium-rich fraction of the eutectic (green color-coded areas). In contrast, the V<sub>SS</sub> primary dendrites show almost no deformation (blue areas), merely a slight green veil can be noted around the grains which is an indicator for piled-up dislocations at the grain boundaries. Further attention must be paid to the emerging thin green lines on the primary dendrites, which are presumed to be dislocation slip bands caused by dislocation pile-ups and local stress peaks at the grain boundaries. On the one hand, the origin of the present cracks in the two-phase eutectic region can be described as a result of stress-peaks in the specimens, induced by the strong deformation of the V<sub>SS</sub>-rich matrix (leading to the aforementioned piled-up dislocations at the grain boundaries). On the other hand, the cracks in the V<sub>3</sub>Si grains indicate a stress relief mechanism, determined by a break-up of the hard, intermetallic phase [11]. The misorientation map of the V-12.5Si alloy (Figure 5 (b)) shows an evenly distributed deformation behavior of the V<sub>SS</sub> over the entire sample. The present fraction of intermetallic V<sub>3</sub>Si is larger compared to the hypoeutectic V-Si alloy and the proportion of vanadium solid solution is smaller, respectively, resulting in an increase of the strength and a decrease of the ductility. Moreover, it is mentioned that an irregularity in the distribution of misorientations can be seen in areas where the V<sub>3</sub>Si phase increasingly solidified in larger, rod shaped structures. The deformation obtained in the regions between the dendrites is smaller, compared to the sections with a higher ratio of the globular silicide phase. This observation supports the assumption, that the alloy’s mechanical properties may be adjusted by a combination of the strength enhancing V<sub>3</sub>Si phase and the ductile V<sub>SS</sub> matrix. According to the previous results, the deformation behavior of the V-15.75Si sample is in good agreement with the literature [11,24,25]. Due to the higher silicon content in the hypereutectic specimen, the fraction of intermetallic V<sub>3</sub>Si phase is larger in comparison with the hypo- and fully eutectic alloys. The EBSD image of alloy V-15.75Si (Figure 5 (c)) shows a similar correlation regarding the distribution of the force through deformation in the matrix phase and the breaking up of the silicide phases in case of stress overload, as stated before [11].

Figure 5. SEM images (top) and color-coded EBSD misorientation maps regarding vanadium (bottom); From left to right: (a) V-9.75Si, (b) V-12.5Si, (c) V-15.75Si.
The misorientation maps of the vanadium-boron alloy are displayed in Figure 6. The hypoeutectic V-5.5B sample (Figure 6 (a)) shows a homogeneous distribution of the introduced deformation via the eutectic phase fraction. At high magnitudes, small amounts of cracked, rod-shaped $V_3B_2$ crystallites were obtained in the microstructure. This supports the assumptions stated above (concerning the failure mechanism of the V-Si alloys), that a stress relief mechanism, determined by a break-up of the hard, intermetallic phase is also present. At this point, the attention is drawn to a special feature, present in Figure 6 (a). A primary grain (which can hardly be identified as such in the band contrast image on the left side) in the upper section of the image seems strongly permeated by the eutectic structure. This can be attributed to a residual amount of melt that was accumulated during cooling, and it solidified partly as primary grain and subsequently as eutectic microstructure. Furthermore, the green veils on the blue colored $V_{SS}$ grains are noticeable: It is challenging to differentiate whether these are dislocation belts (similar to the V-9.75Si sample) or sample preparation defects, induced by grinding and polishing. The eutectic microstructure of the V-B alloys is present in a fine non-faceted-faceted arrangement. Due to the resolution limitation of the EBSD detector, a more precise characterization of this area was not possible. The images of the V-11B compression samples (Figure 6 (b)) display a regular distribution of deformation across the vanadium-rich fractions of the eutectic microstructure, comparable to that of the eutectic V-Si specimens. The present vanadium boride facets and the primary crystal remains almost undeformed. The $V_3B_2$ crystal depicted in Figure 6 (b) shows a green veil at its phase boundaries, indicating a pile-up of dislocations. Based on the fine eutectic microstructure and the strong deformability, conclusions were drawn regarding the high levels of plastic deformation present in some of the tested samples: Similar to the V-15.75Si alloy, the misorientation map of the hypereutectic V-14.5B sample (Figure 6 (c)) indicates a failure behavior, determined by a distribution of the deformation between the V-rich $V_{SS}$ phase and the intermetallic phase. The plastic deformation is illustrated in green (V-rich phase), while the grey ($V_3B_2$ phase) fractions show cracks, induced by stress-induced overload.

Figure 6. SEM images (top) and color-coded EBSD misorientation maps regarding vanadium (bottom); From left to right: (a) V-5.5B, (b) V-11B, (c) V-14.5B.
In addition to the alloy characteristics discussed above, a bar chart is presented in Figure 7, showing the experimentally obtained compression yield strengths. When taking the values, displayed in Table 3 into account, a clearly differentiated increase with respect to the Si and B contents in the alloy composition was observed. In comparison, the $\sigma_{0.2}(\text{Si})$ values are generally higher than the $\sigma_{0.2}(\text{B})$ numbers, resulting in a difference in the overall (compressive) strength of $\sim 430$ MPa, regarding the hypoeutectic alloys, $\sim 440$ MPa, referred to the eutectic alloys and $\sim 520$ MPa, concerning the hypereutectic alloys. The displayed results, again, indicate, that the mechanical properties of both alloying systems are determined by the present fraction of eutectic (ductile) and intermetallic (brittle but high strength) microstructure in proportion by volume. Furthermore, the compressive strength of the samples was not taken for evaluation because of the partly present ideal plastic failure behavior, leading to intersections with the fracture stresses and therefore resulting in misleading or imprecise conclusions.

**Table 3.** $\sigma_{0.2}$ values of the V-Si and V-B alloys tested.

| Alloy   | V-9.75Si | V-12.5Si | V-15.75Si | V-5.5B | V-11B | V-14.5B |
|---------|----------|----------|-----------|--------|-------|---------|
| $\sigma_{0.2}$ (MPa) | ~745     | ~840     | ~1015     | ~272   | ~400  | ~490    |
| Standard deviation (MPa) | 65       | 41       | 105       | 43     | 59    | 33      |

**Figure 7.** Comparison of the compressive yield strengths of the V-9.75Si, V-12.5Si, V-15.75Si, V-5.5B, V-11B and V-14.5B samples.

**4 Summary and conclusion**

In the present work, different alloys from the system vanadium-silicon and vanadium-boron were produced by an arc-melting process. The microstructures were designed to contain either $\sim 50\%$ of the primary phases in comparison to the respective eutectic alloy composition, or a 100% eutectic microstructure, evaluated with computer aided image analysis. Room temperature compression tests on cylindrical samples were carried out and the mechanical properties such as compressive yield stress and deformation, were investigated. Based on the compression tests, the room temperature plasticity was determined for all samples, which was influenced by the different alloying elements and the different volume
fractions of the primary phases. Grain misorientation maps, displaying orientation changes by calculating the misorientation between each pixel and the surrounding pixels, thus highlighting strongly deformed regions, were generated with EBSD analysis. The maps indicate that the applied mechanical stress is absorbed by deformation of the V\textsubscript{SS} phase, whilst the high-strength intermetallic silicide and boride phases act as strength enhancing components. Dislocation pile-ups on the phase boundaries and resulting cracks were stated as the criteria of failure in almost all samples. An exception is alloy V-5.5B, which exhibits an almost ideal-plastic deformation behavior and thus, no cracking in the investigated range of deformation was observed. It can be noted, that the present findings are in good agreement with previous investigations on comparable alloy systems.

The results of the present work indicate a high potential of vanadium-based materials with respect to structural applications. To gain a more profound understanding of the deformation mechanisms, and/or on the dislocation movement in those alloys, transmission electron microscopic analysis (TEM) may be a helpful tool, and more V-B samples with higher fractions of boron should be investigated.

Acknowledgements
The present research was funded by the German Research Foundation (DFG) under the grant number 410338871 and was in part conducted within the context of the International Graduate School MEMORIAL at Otto-von-Guericke-University (OVGU) Magdeburg, Germany, kindly supported by the European Structural and Investment Funds (ESF) under the program "Sachsen-Anhalt WISSENSCHAFT Internationalisierung" (project no. ZS/2016/08/80646). Financial support of the Methodisch-Diagnostisches Zentrum Werkstoffprüfung (MDZWP) e. V., Magdeburg, Germany is greatly acknowledged. We kindly thank Dr. E. Wessel (IEK-2, FZ Jülich) for providing the SEM images and the EBSD analysis.

References
[1] IFAK; GFK Media and Communication Research; forsa marplan 2019 Meistgenutzte Verkehrsmittel zum Erreichen des Urlaubsziels in Deutschland in den Jahren 2017 bis 2019 1
[2] Statistisches Bundesamt 2019 Anzahl der beförderten Personen im Luftverkehr in den Jahren 2004 bis 2018 in Deutschland 1
[3] Umweltbundesamt 2018 Vergleich der durchschnittlichen Emissionen einzelner Verkehrsmittel im Güterverkehr - Bezugsjahr: 2018 (Umweltbundesamt)
[4] Scheskey E and Kral M 2003 Flugzeugtriebwerke: Kolben- und Gasturbinentriebwerke ; Aufbau, Wirkungsweise und Betriebsverhalten (Rhombos-Verlag)
[5] Smith D L, Chung H M, Loomis B A, Matsui H, Votinov S and Van Witzenburg W 1995 Development of vanadium-base alloys for fusion first-wall-blanket applications Fusion Eng. Des. 29 399–410
[6] Zinkle S J, Matsui H, Smith D L, Rowcliffe A F, Osch E van, Abe K and Kazakov V A 1998 Research and development on vanadium alloys for fusion applications J. Nucl. Mater. 258–263 205–14
[7] Ehrlich K 2003 Die Entwicklung von Strukturmaterialien für die Kernfusion Materwiss. Werkstech. 34 39–48
[8] Krüger M 2016 High temperature compression strength and oxidation of a V-9Si-13B alloy Scr. Mater. 121 75–8
[9] Krüger M and Köppe-Grabow B 2017 Prozessabhängige mikrostrukturausbildung von V\textsubscript{ss}-V3Si-V5SiB2-werkstoffen Prakt. Metallogr. Metallogr. 54 293–307
[10] Krüger M 2016 Innovative metallische Hochtemperaturwerkstoffe (Otto-von-Guericke-Universität Magdeburg)
[11] Bei H, George E P, Kenik E A and Pharr G M 2004 Microstructures and mechanical properties of V-V3Si eutectic composites Zeitschrift für Met. 95 505–12
[12] Smith J F 1989 Si-V (Silicon-Vanadium) Phase Diagrams of Binary Vanadium Alloys (ASM International) pp 261–7
[13] De Lima-Kühn B B, Da Silvaa A A A P, Suzukia P A, Coelho G C and Nunes C A 2016 Microstructural characterization of as-cast V-Si alloys and reevaluation of the invariant reactions involving the liquid phase of the V-Si system Mater. Res. 19 1122–6
[14] Spear K E, Liao P K and Smith J F 1989 B-V (Boron-Vanadium) Phase Diagrams of Binary Vanadium Alloys ed J F Smith (Metals Park, Ohio, United States of America: ASM International) pp 20–6
[15] Nunes C A, De Lima B B, Coelho G C, Rogl P and Suzuki P A 2004 On the stability of the V5B6-phase J. Alloys Compd. 370 162–6
[16] Massalski T B, Okamoto H, Subramanian P R and Kacprzak L 1990 Binary Alloy Phase Diagrams (Materials Park Ohio: ASM International)
[17] de Lima B B, Coelho G C, Suzuki P A, Nunes C A and Rogl P 2004 Evaluation of the invariant reactions of the V-B system J. Phase Equilibria Diffus. 25 134–9
[18] Pinto da Silva A A A, Chaia N, Ferreira F, Carvalho Coelho G, Fiorani J M, David N, Vilasi M and Nunes C A 2017 Thermodynamic modeling of the V-Si–B system Calphad Comput. Coupling Phase Diagrams Thermochim. 59 199–206
[19] DIN 50106 2016 Prüfung metallischer Werkstoffe – Druckversuch bei Raumtemperatur 1–10
[20] Elliott R 1984 Eutectic solidification Metall. Trans. A 65 85–92
[21] Karma A and Plapp M 2004 New insights into the morphological stability of eutectic and peritectic coupled growth JOM 56 28–32
[22] Sahm P R, Egry I and Volkmann T 1999 Schmelze, Erstarrung, Grenzflächen ed P R Sahm, I Egry and T Volkmann (Berlin, Heidelberg: Springer Berlin Heidelberg)
[23] Bauer G, Güther V, Hess H, Otto A, Roidl O, Roller H, Sattelberger S, Köther-Becker S and Beyer T 2017 Vanadium and Vanadium Compounds Ullmann’s Encycl. Ind. Chem. 38 1–22
[24] Hasemann G, Müller C, Grüner D, Wessel E and Krüger M 2019 Room temperature plastic deformability in V-rich V–Si–B alloys Acta Mater. 175 140–7
[25] Müller C, Hasemann G, Regenberg M, Betke U and Krüger M 2019 Microstructure and Compression Properties of V SS -V 3 B 2 Eutectic Alloys in the V-Si-B System Materials (Basel). 13 1–12, 2100.