Lateral Extensions to Nanowires for Controlling Nickel Silicidation Kinetics: Improving Contact Uniformity of Nanoelectronic Devices

Maik Simon*,†,§, Ryo Mizuta‡, Ye Fan‡, Alexander Tahn‖,§, Darius Pohl‖,§, Jens Trommer†,§, Stephan Hofmann‡, Thomas Mikolajick†,§, Walter M. Weber†,§,#

AUTHOR ADDRESS. NaMLab gGmbH, Nöthnitzer Str. 64a, 01187 Dresden, Germany

KEYWORDS: nickel, silicon, nanowires, in situ, silicidation

ABSTRACT. Although widely applied as contacts to nanoelectronic devices, metal silicides in nanostructures suffer from varying composition and growth rate. To study the underlying kinetics and to control reactions, we introduce local volume extensions (“polders”) to silicon nanowires. This method allows to decouple the silicide growth process from variations in the metal supply and to gain a reduced length growth rate as long as silicon reaction volume is available in the polders. In situ analyses are gained by scanning electron microscopy during the anneal to extract the growth rates. A deterministic limitation of silicide growth by nickel flux, NiSi2 reaction rate and the nickel diffusion is observed. The extracted maximal reaction rate at the NiSi2-Si interface allows to determine an activation energy. Subsequent transmission electron microscopy reveals an epitaxial \{111\} NiSi2-Si interface in the \langle011\rangle-oriented nanowire. It is also seen that the polders suppress Ni-rich silicide phases and gives rise to the formation of a single-crystalline Ni-Si phase with a Ni:Si ratio close to 1:1. Retarded growth by the application of polders can almost stop the silicidation in nanowires at
a defined point even for different Ni fluxes. This can help to reduce gate overlap and channel length variation especially in Schottky-junction-based field effect transistors. Geometric optimization of the polder regions with regard to largest impact is discussed.

1. Introduction

Silicides are compounds between silicon and various transition or rare earth metals which can be formed by heat-induced solid-state reaction (“silicidation”). Silicidation is widely applied to create highly conductive and sometimes even magnetic contacts to silicon with a minor level of defects or contaminations at the junction. Thus, it is commonly used to create gate and/or source/drain contacts of for example spin-based field-effect transistors (FETs), FET-based sensors or commercial metal-oxide-semiconductor FETs (MOSFETs). Especially NiSi is a preferred contact material because of its good electrical conductivity, low film stress, low processing temperature and a reduced silicon consumption during silicidation in comparison to other silicides like CoSi2 or TiSi2. Beyond this, all stable Ni-silicide phases have in common that the metal and not the silicon is the dominant diffusing species. This avoids a channel depletion by the Kirkendall effect, as well as the formation of shorts bridging to the gate. Moreover, it allows that these silicides can intrude over large distances into silicon. This ability is of special interest for Schottky-barrier MOSFETs (SB-MOSFETs) and reconfigurable FETs (RFETs) with nano-scaled channels made from Si fins, nanowires or nanosheets. Here, nickel silicidation allows to form Schottky contacts in a self-aligned manner directly below a gate. The intrusive silicide growth can also be used to segregate dopants in front of the silicide-silicon interface in order to form abrupt junctions with reduced contact resistivity, e.g. in tunnel FETs (TFETs) and metallic-source/drain MOSFETs. With decreasing device dimensions, the precise self-aligned placement of Schottky junctions becomes more important but also challenging. Although knowledge of silicidation in bulk and thin film silicon exists, it is insufficient to describe the silicidation kinetics in
modern nanometer-scaled and sometimes even three-dimensionally extended structures. First, because in nanometer dimensions new parameters like the Si-channel surface, volume and faceting and the local metal contact quality and geometry get important. Second, because the stress inside confined nanochannels presumably changes the formation energy of phases, e.g. NiSi$_2$ is not found below silicidation temperatures of 750°C in bulk but identified as the leading phase in nanowires even at 300°C.$^{14}$ Yet, a strong variation of the silicide length even among equally processed nanochannels usually makes it difficult to reveal the effects of these parameters.

In the following, a novel nanoscaled structure is presented which allows one to observe the silicidation rate in different silicon geometries while excluding a divergence of Ni supply that could originate for example from a varying quality of individual Ni contacts. An in situ analysis using scanning electron microscopy (SEM) is applied in order to not only observe the final result of the silicidation but to clearly determine the actual diffusion and reaction kinetics in different geometries independently from possible reaction onset-delays, temperature ramp-up and cool-down effects or switchings between reaction- and diffusion-limited growth. A TEM study further reveals, that the polders inhibit the growth of Ni-rich silicide phases and give rise to the formation of a single-crystalline nickel silicide with a Ni:Si ratio close to 1:1. Finally, the gained insights into the silicidation in complex nanostructures are applied to reduce the variation of the silicide length in nanowires by lateral extensions. Multiple extension geometries are compared in a model and their effectiveness is proven experimentally. Alltogether, we show that polders are a useful tool to create nanoelectronic devices with uniform contacts, both in terms of silicide phase and silicide length.

2. Background

Nickel silicidation of a nanowire is driven by the diffusion of Ni into the silicon and then through the silicidized nanowire unto the interface where it forms a new layer of silicide (see
Figure 1a). In a simplified model that assumes the formation of only a single silicide phase, the presence of an inexhaustible Ni source and the absence of Si diffusion, the grown silicide axial length $L$ within the Si nanowire is initially “reaction-limited” over time $t$ according to

$$L = r \cdot t,$$

(1)

with $r$ being the length-related rate of silicide formation.$^{14,15}$ Nevertheless, with increasing distance between the Ni-source and the silicide-silicon interface, the Ni concentration at this interface must drop due to diffusion. Eventually, the Ni concentration at the interface is so strongly reduced that the incoming Ni flux determines the growth rate, yielding a slower “diffusion-limited” silicidation rate according to

$$L^2 = \kappa \cdot t,$$

(2)

where $\kappa$ is the effective rate with which Ni diffuses through all present silicide phases.$^{14,15}$ Several studies have already analyzed the silicidation reaction over time in bottom-up grown nanowires.$^{14-20}$ The transition from reaction- to diffusion-limited growth regimes has been observed in several of them for relatively long anneal times.$^{14,17-20}$ Additionally, it has been claimed that a limited metal supply, e.g. due to an interfacial oxide between the silicon and the Ni source, can lead to an even smaller and ultimately even almost constant growth rate.$^{20,21}$

Figure 1. (a) Schematic of the nickel silicide formation in a nanowire in side view. Ni atoms are dissolved through the interface into the nanowire where they diffuse to the interface with the silicon and then react to form a new layer of silicide. (b) SEM top-view on array of 14 nm
high (011)-oriented Si nanowires after Ni silicidation, revealing a large silicide length variability $\Delta L$. (c) On the same chip as the structures in (b), a Ni-supplied nanowire is connected to a 14 nm thin silicon area in which a rectangular Si-rich silicide area with much shorter $L$ is formed. Ni-rich silicide formation stops at the junction between silicon area and nanowire.

Diffusion in single crystal nanowires is assumed to occur either by volume diffusion through the crystal or surface diffusion at the interface to the surrounding oxide, as grain boundary diffusion can be excluded. Some studies reporting post-reaction (ex situ) analyses on nanowires showed a trend of longer silicide intrusion with decreasing cross-section. Different surface to volume ratios amongst these wires have been used to ascribe this observation to either volume or surface diffusion. However, a large wire-to-wire variation of silicide lengths has hampered a systematic analysis and even led to opposing interpretations.

In fact, it is commonly observed that the silicide length varies largely among bottom-up as well as top-down fabricated nanowires after equal processing which may be attributed to a variation of Ni contact geometry and quality. Figure 1b shows an exemplary array of top-down fabricated nanowires patterned from a silicon-on-insulator wafer with silicide lengths varying by as much as $\Delta L = 0.74 \mu \text{m}$ or 57% compared to the maximal length value. Such variations make it difficult to reveal how parameters such as strain, crystal direction, temperature or nanowire diameter will influence the silicidation kinetics. As a consequence, the precise placement of silicide-silicon junctions for example in transistors with nanostructured channels is challenging.

3. Results and discussion

3.1 Observation of silicide growth in nanowires with lateral extensions
The key observation of this paper is that Ni silicide intrusion in terms of length $L$ is significantly reduced when the reaction front area becomes much larger than the cross section of the supplying nanowire. For example, Figure 1b shows silicides which grew 0.56-1.31 $\mu$m into regular nanowires. By contrast, the silicide in another nanostructure on the same chip shown in Figure 1c grows only 0.29 $\mu$m into forward direction because the Ni is injected through a nanowire into a wide thin film area. In this case it is also observed that the silicide in the (100)-oriented monocrystalline Si film does not grow circularly but in a rectangular shape into the $\langle 011 \rangle$ directions around the injection point, at least for small intrusion. Coincidently, the extension inhibits the growth of bright and uneven silicides (see Figure 1c) in the nanowire beyond the junction. These silicides are commonly characterized as Ni-rich silicides.$^{17,28}$

A similar propagation retardation effect as in Figure 1c is utilized in hydraulic engineering: Floods in rivers can be reduced when distinct retention polder areas behind dikes are intentionally flooded. By analogy, we use the insights gained by Figure 1c to create nanowires that are only locally extended by lateral “polder” areas to temporarily reduce the silicide growth rate in terms of length until the polders are “filled”.

An in situ analysis is used to reveal the silicidation kinetics in a nanowire with bar-shaped polders (Figure 2a) when annealed at 450°C. Since large top-down fabricated structures are hard to be entirely in situ examined by transmission electron microscopy (TEM), the analysis is conducted by scanning electron microscopy (SEM) on a heating stage. Figure 2b shows the evolution of silicide lengths $L$ in the polders and the nanowire over time (video available online). The measurement precision is approximately ±30 nm. We did not observe the silicidation before it reaches the junction point where Ni can diffuse into forward and sideward directions. Thus we define this point to be the origin ($L = 0$). A 385 nm long feed nanowire segment connects this point to the Ni source. It is observed that $L$ grows with different rates: with a constant rate until the polders are filled in regime 1, with a higher
constant rate in regime 2 and finally in regime 3 with a decaying rate slower than in regime 2 but higher than in regime 1.

**Figure 2.** Silicide formation in a 17 nm high nanowire with bar polders at 450°C. (a) SEM top-view image after all anneals. (b) Silicide length evolution over time gathered by in situ SEM. The inset shows the measured distances after 10.6 min of in situ silicidation. (c) Accumulated silicidized area in top-view over time based on the local polder/wire width. (d) Arrhenius plot of highest length growth rates extracted from earlier reports on silicidation in [111] bottom-up grown nanowires and in regime 2 in (b). An activation energy $E_A$ of
2.2 eV per \{111\} plane/monolayer is extracted, considering only the data points with high growth rates (filled symbols).

**Figure 3.** Side view TEM and FFT analysis of a length cut through the nanowire. (a) \(\{111\}\)-oriented silicide/silicon interface. (b) A magnified section of (a) shows that the interface grew epitaxially. (c) An FFT analysis of the area marked in (a) reveals that the interface consists of NiSi\(_2\) and that the \(\{111\}\) axis is the normal to the interfacial plane in (b). The zone axis is [011]. (d) Typical section of the silicided nanowire between polder and silicide/silicon. The measured thicknesses are generally 22.7 nm ± 2 nm. (e) FFT of the area marked in (d) reveals a crystal structure with a 90° angle when seen from the [011] zone axis.

Since the silicide grows with almost stable rates but different slopes in the three regimes, additional information is necessary to distinguish the different growth regimes. For this purpose, the accumulated silicide area in top view (silicide length × local structure width) is evaluated in Figure 2c. The areal reaction rate drops from regime 1 to 2 after theolders are completely silicidized while still being linear in both regimes. In other words, less silicide is formed although the Ni supply presumably stays constant. The constant but high length
growth rate $r$ indicates a reaction limitation for regime 2. In Figure 2d, $r$ is compared to the maximal reported growth rates of earlier in situ studies on silicidation in [111]-oriented oxidized nanowires. Those wires also feature an epitaxial \{111\} NiSi$_2$/silicon interface as identified here by Fast Fourier Transform (FFT) in a TEM post-reaction analysis (Figure 3a,b,c and S2, Table S1 and S2).\textsuperscript{17–19} The good agreement between earlier reports and this work also indicates that the electron irradiation of 40 $\mu$A/cm$^2$ at 5 kV had no major impact on the silicidation kinetics. Please note that data from ex situ silicidation studies are not included in Figure 2d even though some of them reported higher effective growth rates.\textsuperscript{23–25} The reason is that only for in situ measurements we can definitely exclude contributions of temperature ramping, temperature overshoots or removed oxide shell close to the Ni pad. The meta-study of growth rates also reveals that some growth rates (open symbols) are much lower than what has been reported in other studies for the same or similar temperatures. The remaining datapoints (filled symbols), however, form a linear trend. The slope of this Arrhenius plot of $-8.0 \times 10^3$ nm$\cdot$K gives an activation energy of

$$E_A = -\left( -8.0 \times 10^3 \text{ nm}\cdot\text{K} \cdot 8.617 \times 10^{-5} \frac{\text{eV}}{\text{K}} \cdot (0.313 \text{ nm})^{-1} \right) = 2.2 \text{ eV}$$

per transformed \{111\} Si plane (plane distance 0.313 nm).

Knowing that regime 2 is reaction-limited, the constant but lower silicidation rate $r$ in regime 1 must originate from the limited supply of Ni that is distributed by diffusion from the feed nanowire segment into all three directions. Although such diffusion-limited silicidation usually results in a square-root growth over time, a very gentle decay of $r$ can still appear like a linear growth.\textsuperscript{20,21} Note that regime 1 is not the very first regime in this nanowire. Before reaching the polders, we expect that the silicide has been growing in a reaction-limited regime in the feed nanowire segment. Then, the polders interrupt this initial regime but regime 2 is its continuation. However, we did not observe this in situ.
Figure 2b also shows that the silicide in regime 1 grows faster in the wider polders than in the forward nanowire. This is counterintuitive as the gradient of diffusing Ni should be equal in all three branches. It also contradicts earlier observations of a longer silicide for thinner nanowires,\textsuperscript{17,22-24} ascribed to a stronger surface diffusion of Ni along the silicide/oxide interface due to the larger surface to volume ratio.\textsuperscript{26} However, those experiments have been conducted at lower temperatures and/or for much shorter times. In these cases, a reaction-limited growth \((L \propto t)\) with wire-to-wire differences in reaction on-set times or Ni supply is more likely than a diffusion-limited growth \((L \propto t^{0.5})\) that typically sets on much later.\textsuperscript{18,19} As opposed to this, we do assign an artificially supply-limited diffusion to rule the growth in regime 1. A preferential surface diffusion along the \{011\} planes is unlikely, since this would result in a faster growth in narrow nanowires which we have neither observed in this experiment nor in ex situ inspections of ordinary nanowires not shown here. However, even for a volume diffusion of Ni, the growth rate in the polder should not be higher than in the nanowire. A possible explanation could be an unfavorable diffusion or silicide formation close to the side walls. In fact, several other studies have reported that Ni-silicide grows faster in wires with thin instead of thick oxide shells, attributed to a reduced interfacial energy barrier or to enhanced Ni diffusion due to less compressive stress.\textsuperscript{17,28,29}

For a large distance between Ni source and silicide-silicon interface, the Ni diffusion gradient inside the nanowire flattens. Given the evident reduction in growth rate between regimes 2 and 3 we attribute the Ni diffusion to be the limiting factor for the silicidation rate in regime 3. The growth rate further decayed during a second in situ anneal carried out under the same conditions (Figure S1) which confirms the diffusion-limination of growth. Considering also the feed nanowire segment as part of the diffusion path, a median diffusion coefficient \(\kappa = 8.3 \times 10^{-15} \text{ m}^2\text{s}^{-1}\) is obtained here which agrees well with earlier reports.\textsuperscript{14} In the course of the second anneal, a bulge formed before the polder (Figure 2a) which has been observed earlier in nanowires as a result of repeated anneals.\textsuperscript{21} A TEM and FFT analysis of a
length cut through the complete 4.45 µm long silicidized nanowire reveals that approximately 85% of it consists continuously of presumably single-crystalline nickel silicide (Figure 3d,e, S3, Table S3). Analyzing the reciprocal spacings from TEM taken at different angles, we found the shortest reciprocal spacing among the four resulting FFTs as 1.8 nm\(^{-1}\) (see marked reflections in Fig. S4+f at stage tilt -25° and +16°). This converts to 0.55 nm in real space. Yet, this does not fit to any plane spacing of known nickel silicide crystals except for NiSi\(_2\). However, we can exclude cubic NiSi\(_2\) since (i) the (100) reflections are forbidden and (ii) the symmetry of the FFT in Figure S4b does not fit to a cubic crystal structure. Since we found the 0.55 nm spacing under two sample tilts, we expect it to be stemming from at least two large lattice parameters. From the obliquity of the FFTs we assume a monoclinic space group. An EDX scan along the TEM lamella shows that nickel and silicon are close to ratio 1:1 in this segment (Figure S5). The observed \(\kappa\) can be likely attributed to the Ni diffusion through this predominantly growing phase.

In general, a limited supply of Ni suppresses the growth of Ni-rich silicides in nanowires. For example, NiSi is predominantly formed when Ni is supplied by a point contact to a Ni-nanowire instead of a large Ni contact or when a platinum interlayer is inserted between Si nanowire and Ni source.\(^{29,30,31}\) In our work almost the complete nanowire beyond the polder consists of a single silicide phase with Ni:Si ratio close to 1:1 – apart from a short segment of Ni\(_2\)Si close to the polder and NiSi\(_2\) close to the leading interface. We attribute this phase stabilization to the polder. Presumably, the polders temporarily absorb excess interstitial Ni so that the Ni concentration in the nanowire is too low to allow for the formation of Ni-rich phases. The observation that bright and volumetric Ni-rich silicides stop growing at the entry to the wider Si areas is made throughout a large number of structures, including that in Figures 1c. Therefore we propose that area extensions to nanowires are a useful and novel way to retard or even suppress the formation of metal rich silicides, in particular of Ni-rich silicides. This can be beneficial especially for electronic devices where a volumetric
expansion by silicidation as in the Ni-rich bulges is undesired. Since for such devices also a low silicide resistivity is crucial, the phase should be analysed also in this regard in future works.

3.2. Application of lateral extensions for silicide length control in nanowires

The presented results successfully demonstrate a reduction in silicide length growth by appending lateral polders which is utilized in the following to reduce the variation of silicide length between nanowires of similar geometry. This is explained by two model nanowires (NW1 and NW2) having equal width $w$ but different silicide growth rates for example due to differently strong Ni supplies at the wire ends (Figure 4a). Their silicide lengths $L_1$ and $L_2$ are defined as the distance between the Ni source edge and the silicide-silicon interface in forward direction. For simplicity, we assume a constant volumetric silicidation rate $\nu = dV/dt$ within each wire, i.e. no decay of Ni diffusion with increasing distance to the source is considered. NW1 has a factor two higher $\nu$ than NW2 due to a larger Ni flux $J_{\text{diff}}$. The modeled difference in silicide length $\Delta L = L_1 - L_2$ drawn as dashed lines in Figure 4b therefore grows linearly over time. A small $\Delta L$ is intended and may be easiest to achieve for a short anneal time. However, many applications require a considerable silicide intrusion for dopant segregation or self-aligned positioning of a Schottky junction, as described in the introduction.
Figure 4. Control of forward directed silicide length $L$ by polders. (a) Schematic of a nanowire without (left) and with (right) polder, where the silicide (grey) grows to length $L_2$. For the case of higher silicidation rates the silicide length is increased by $\Delta L$ to $L_1$. The polder splits the Ni diffusion flux $J_{\text{diff}}$ into three directions. (b) Expected evolution of $L$ in two nanowires (lower plot), where NW1 has twice the silicidation rate as NW2 due to better Ni supply. Polders give NW2 time to catch up to NW1 so that $\Delta L = L_1 - L_2$ (upper plot) can be reduced for a target length $e$ (vertical lines). (c) In the same model, pan- and mushroom-shaped polders reduce the minimal achievable $\Delta L$ for $L > e$. $e$ is equal in all geometries.

As drawn in Figure 4a, the lateral addition of polders of width $p$ at distance $d$ from the Ni source splits the Ni flux $J_{\text{diff}}$ into three directions until the complete length $x$ of the polders is transformed into silicide. As a result, the growth rate $r$ in terms of length will slow down by a factor 3 although the volumetric growth rate $\nu$ remains constant.
The final silicide-silicon interface is intended not to be located in the polder segment \((d < L_{1,2} < e)\) but in the nanowire beyond \((L_{1,2} > e)\). Vertical lines in Figure 4b mark the times at which the silicide of NW2 reaches \(L_2 = e\) with and without polder. In the upper plot it can be seen that \(\Delta L\) is much larger for the case without polders than with polders at the same point in time. The reason for this can be seen in the lower plot. Between the silicide reaching the beginning of the polder \((L_2 = p, \text{ time } = 2 \text{ for NW2})\) and the silicide filling up the polders \((L_2 = p+x, \text{ time } = 8 \text{ for NW2})\), the silicide growth rate \(r\) is reduced. After the polders are filled, the high \(r\) is reestablished, as experimentally demonstrated in Figure 2b before. For sufficiently large \(x\), a time window (here between time steps 2 and 4) exists for which the polder of NW1 is not yet filled so that the silicide is still growing with decreased rate while the silicide in NW2 has reached beyond the critical length \(L = e\). During this time, \(L_1\) and \(L_2\) are closer to each other than without polders, hence the desired substantial reduction in silicide length variability \(\Delta L\) is achieved.

The rate \(r\) can be reduced even more if the silicide-silicon interface area is further extended. In a geometry as in Figure 1c, the interface length of the rectangular silicide area grows ideally to \(w+4L_0\) where \(L_0\) is the forward silicide length inside the rectangle (Figure S6). With a fixed \(J_{\text{diff}}\) and hence a fixed \(v\) this means that the length growth rate \(r\) towards the forward direction decreases with growing \(L_0\) according to Equation (S5).

Based on this model, even smaller \(\Delta L\) is gained when the polder is small at the connection to the nanowire but widens within a distance \(s\). Such pan- and mushroom-shaped structures with “handle” width \(p = w\) and length \(s = 2w\) are displayed in Figure 4c. For all geometries \(e\) is set to \(8w\). As long as \(L < d+s\), the reaction is retarded in the same manner as for a bar-shaped polder. However, as soon as the silicide reaches further than \(d+s\), the interface area increases even more and leads to a further overall decreasing \(r\). This approach almost “stops”
the forward silicidation upon a targeted distance $d+s$ so that $\Delta L$ is reduced compared to the bar-shaped polder at a much lower area consumption.

According to Figure 4b the application of polders should not only reduce $\Delta L$ for a fixed targeted $L$ but also reduce the achieved $L$ for a fixed time. Indeed, pan-shaped polders as in Figure 5a limit $L$ reliably to a value of 47 nm beyond the polder in median, although $e$ varies between 260 and 435 nm for the examined structures. This can be applied to create uniform silicide lengths in a transistor with several parallel nanowires, as Figure 5b shows. Neighboring nanowires can thereby share polder areas. The achieved control over silicide length is crucial to create for example Schottky-junction-based FETs with uniform channel length or minimal gate overlaps. Still, for dense nanowire arrays as in Figure 5c, a polder bar is more feasible. The reduced $L$ compared to an unrestricted silicidation clearly indicates that the polder reduces the length growth rate by splitting the available amount of unreacted Ni to an extended silicide-silicon interface area. This and the possibility for transversal Ni diffusion between the wires result in a reduced $\Delta L$. For achieving even an even better equalization of $L$ longer polders can be employed.

![Figure 5](image.png)

**Figure 5.** SEM top-view images of 14 nm high Si nanostructures after Ni silicidation. (a) Pan-shaped polders limit and homogenize $L$ (cloud of dots) to a value closely beyond it, even for varying $e$. The Ni pad edge is marked by a dashed line. (b) Parallel nanowires with shared
polders. The use of polders uniformly reduces $L$ to shortly beyond the polders. (c) A polder bar connecting a dense array of parallel nanowires results in reduced $L$ and $\Delta L$.

4. Conclusions

Lateral extensions (“polders”) to nanowires simplify the research on and the application of Ni silicidation in Si-nanowire devices. The polders can temporarily switch the silicide growth from a reaction- to a diffusion-limited regime and stop Ni-rich silicides at the junction point. This allows to identify the Ni reaction rates and diffusion coefficients by in situ SEM. The diffusion appears to be considerably faster in wide than in narrow nanowires, as opposed to prior ex situ observations of Ni silicidation in nanowires. Polders suppress the growth of Ni-rich silicides so that a single-crystalline silicide phase with Ni:Si ratio close to 1:1 forms. They can be also applied to almost stop the silicidation at a certain point inside a nanowire so that slower silicidizing nanowires have sufficient time to catch up. This can be used to create nanoelectronic devices with uniform channel lengths or gate overlaps. Based on a model to describe the polder effect, the polder geometries are optimized to gain ideal control over the silicide length as proven on chip.

5. Experiment Section

The nanostructures are defined by electron beam lithography (EBL) on a 20 nm thin (100) silicon-on-insulator wafer piece and structured by a reactive ion etching utilizing SF$_6$, O$_2$ and CHF$_3$.\textsuperscript{32} The nanowires are defined by EBL in $\langle 011 \rangle$ direction, using the $\langle 011 \rangle$-oriented cleaving edges of the wafer as reference. A short etching in diluted HF is performed to remove native oxide around the structure and subsequently a thermal oxide is grown in the oxygen atmosphere by a rapid thermal annealing (RTA) at 875°C for 5 min (Figure 2 and 3) or 10 min (Figure 1, 5). The Ni pads are structured by a lift-off process.\textsuperscript{33} Preliminary to the sputter deposition of 45 nm Ni, thermal oxide is locally removed by buffered HF. The
structures are then annealed for 60 s (Figure 1) or 40 s (Figure 2, 3 and 5) at 450°C by RTA in forming gas atmosphere.

For the in situ annealing in Figure 2, the sample was loaded into a “ZEISS Gemini300” SEM onto a “Kammrath & Weiss” in situ heater and contacted by a thermocouple. The sample was heated at a pressure of less than 10⁻⁵ mbar from room temperature to 250°C with a ramp rate of 50°Cmin⁻¹ and then from 250°C to 450°C with slower rates of 10°Cmin⁻¹ and with three intermediate 5–10 min stabilization breaks to avoid temperature overshooting. The images were then recorded at 450°C at a frame rate of 30 min⁻¹. The temperature of the devices during in situ annealing was stable to within 0.2°C as measured by the thermocouple. The thermocouple is accurate to 30°C of the actual temperature, as determined by melting a salt with a known melting temperature (Na₂WO₄) on a SiO₂ substrate using the in situ heater. The temperature distribution is visually confirmed to be uniform across the full active area of the heater, using this same melting calibration.
Supporting Information

In situ video of silicidation in nanowire with lateral extensions
Further analysis of regime 3 by a second anneal
Phase determination by FFT
EDX scan along nanowire lamella
Model describing silicidation rate in nanowires and area extensions

AUTHOR INFORMATION

Corresponding Author

*Maik Simon – NaMLab gGmbH, Nöthnitzer Straße 64a, 01187 Dresden, Germany; e-mail: maik.simon@namlab.com

Authors

†NaMLab gGmbH, Nöthnitzer Str. 64a, 01187 Dresden, Germany

§Technische Universität Dresden, Center for Advancing Electronics Dresden (CfAED), 01162 Dresden, Germany

‡Department of Engineering, University of Cambridge, Trumpington Street, Cambridge CB2 1PZ, United Kingdom

‖Dresden Center for Nanoanalysis (DCN), Technische Universität Dresden, Helmholtzstraße 18, 01062 Dresden, Germany

#Current affiliation is: Institut für Festkörperelektronik, Technische Universität Wien, Gußhausstraße 25-25a, 1040 Wien, Austria
Author Contributions

The manuscript was written through contributions of all authors. All authors have given approval to the final version of the manuscript.

Funding Sources

Parts of this work are supported by the “Deutsche Forschungsgemeinschaft (DFG)” in the framework of “ReproNano” (WE 4853/1-3, MI 1247/6-2) and of the cluster of excellence “CfAED” (EXC 1056). Further support originates from the EPSRC Cambridge NanoDTC (EP/L015978/1, EP/P005152/1) and the German Academic Exchange Service within the frame of the “IPID4all” program and the Graduate Academy of TU Dresden.

ACKNOWLEDGMENT

The authors would like to thank Uwe Mühle (Fraunhofer IKTS, now with Robert Bosch Semiconductor Manufacturing Dresden GmbH) for TEM imaging and Jan Gärtner (NaMLab) for image evaluation.

REFERENCES


